Investigations of low dimensional magnetic structures for memory and storage device applications

Kumar, Durgesh

2020

Kumar, D. (2020). Investigations of low dimensional magnetic structures for memory and storage device applications. Doctoral thesis, Nanyang Technological University, Singapore.

<https://hdl.handle.net/10356/146139>

<https://doi.org/10.32657/10356/146139>

This work is licensed under a Creative Commons Attribution-NonCommercial 4.0 International License (CC BY-NC 4.0).

Downloaded on 26 Apr 2023 12:35:55 SGT

**NANYANG
TECHNOLOGICAL
UNIVERSITY**

SINGAPORE

**INVESTIGATIONS OF LOW DIMENSIONAL MAGNETIC
STRUCTURES FOR MEMORY AND STORAGE DEVICE
APPLICATIONS**

DURGESH KUMAR

SCHOOL OF PHYSICAL AND MATHEMATICAL SCIENCES

2020

**INVESTIGATIONS OF LOW DIMENSIONAL MAGNETIC
STRUCTURES FOR MEMORY AND STORAGE DEVICE
APPLICATIONS**

DURGESH KUMAR

SCHOOL OF PHYSICAL AND MATHEMATICAL SCIENCES

A thesis submitted to the Nanyang Technological
University in partial fulfilment of the requirement for the
degree of Doctor of Philosophy

2020

Statement of Originality

I hereby certify that the work embodied in this thesis is the result of original research done by me except where otherwise stated in this thesis. The thesis work has not been submitted for a degree or professional qualification to any other university or institution. I declare that this thesis is written by myself and is free of plagiarism and of sufficient grammatical clarity to be examined. I confirm that the investigations were conducted in accord with the ethics policies and integrity standards of Nanyang Technological University and that the research data are presented honestly and without prejudice.

16/07/2020

Date

A handwritten signature in black ink, appearing to read 'Durgesh Kumar', with a horizontal line extending from the end of the signature.

DURGESH KUMAR

Supervisor Declaration Statement

I have reviewed the content and presentation style of this thesis and declare it of sufficient grammatical clarity to be examined. To the best of my knowledge, the thesis is free of plagiarism and the research and writing are those of the candidate's except as acknowledged in the Author Attribution Statement. I confirm that the investigations were conducted in accord with the ethics policies and integrity standards of Nanyang Technological University and that the research data are presented honestly and without prejudice.

17/07/2020

Date

S. N. PIRAMANAYAGAM

Authorship Attribution Statement

The chapter 4 of this thesis contains material from three papers published in the following peer-reviewed journals in which I am listed as an author. However, chapter 5 of this thesis partially contains material from one paper published in the following peer-reviewed journal in which I am listed as an author.

The first half of chapter 4 is published as Durgesh Kumar, Surbhi Gupta, Tianli Jin, Razia Nongjai, Kandasami Asokan and S. N. Piramanayagam, Modification of Structural and Magnetic Properties of Masked Co-Pt Films Induced by High-Energy Ion Implantation, IEEE Magnetics Letters 9, 1-5 (2018).

DOI: 10.1109/LMAG.2017.2768324

and

Durgesh Kumar, Surbhi Gupta, Tianli Jin, Razia Nongjai, Kandasami Asokan and S. N. Piramanayagam, Tailoring the structural and magnetic properties of masked CoPt thin films using ion implantation, AIP Advances 8 (5), 056504 (2018).

DOI: 10.1063/1.5007767

The contributions of the co-authors are as follows:

- A/Prof S. N. Piramanayagam proposed the idea, discussed the results and supervised the project.
- I co-designed the study with A/Prof S. N. Piramanayagam.
- Tianli helped during the AGFM of the samples.
- Dr Razia and Dr Asokan from Inter-University Accelerator Center, New Delhi (India) helped in ion-implanting the samples.
- All the other experimental and simulation work including sample fabrication, experimental characterizations, and simulations were conducted by me.
- I analyzed the data with the help of A/Prof. S. N. Piramanayagam, and Dr Surbhi.
- I prepared the manuscript drafts. A/Prof S. N. Piramanayagam provided his inputs time to time during the preparation of the manuscript draft. The manuscript was revised with the help from all the co-authors.

The second half of chapter 4 is published as Durgesh Kumar, Surbhi Gupta, Kim Kong Tham, Razia Nongjai, Shin Saito, Kandasami Asokan and S. N. Piramanayagam Ion-Implantation-Induced Disorder in FePt-C Thin Films, IEEE Transactions on Magnetics, 55 (3), 1-5 (2019).

DOI: 10.1109/TMAG.2018.2867912

The contributions of the co-authors are as follows:

- A/Prof S. N. Piramanayagam proposed the idea, discussed the results and supervised the project.
- I co-designed the study with A/Prof S. N. Piramanayagam.
- Dr Tham and Prof Saito from Tohoku University (Japan) helped during the torque measurements of the samples. I visited Tohoku University for carrying out the torque measurements.
- Dr Razia and Dr Asokan from Inter-University Accelerator Center, New Delhi (India) helped in ion-implanting the samples.
- All the other experimental work including sample fabrication, and experimental characterizations were conducted by me.
- I analyzed the data with the help of A/Prof. S. N. Piramanayagam, and Dr Surbhi.
- I prepared the manuscript drafts. A/Prof S. N. Piramanayagam provided his inputs time to time during the preparation of the manuscript draft. The manuscript was revised with the help from all the co-authors.

The chapter 5 is partially published as Durgesh Kumar, Pinaki Sengupta, Rachid Sbiaa, and S. N. Piramanayagam, Spin Transfer Torque Induced Domain Wall Oscillations in Ferromagnetic Nanowire with a Nanoscale Dzyaloshinskii-Moriya Interaction Region, Journal of Magnetism and Magnetic Materials, 507, 166807 (2020).

DOI: 10.1016/j.jmmm.2020.166807

The contributions of the co-authors are as follows:

- A/Prof S. N. Piramanayagam proposed the idea, discussed the results and supervised the project.
- I co-designed the study with A/Prof S. N. Piramanayagam.
- All the micromagnetic simulations were conducted by me.
- I analyzed the data with the help of A/Prof. S. N. Piramanayagam, A/Prof Pinaki and A/Prof Rachid.
- I prepared the manuscript drafts. A/Prof S. N. Piramanayagam provided his inputs time to time during the preparation of the manuscript draft.

The manuscript was revised with the help from all the co-authors.

16/07/2020

Date

DURGESH KUMAR

Abstract

The data storing devices are of great importance in the age of big data. Amongst several data storage devices, the hard disk drives (HDDs) share the maximum (~80% of the total data). Therefore, continuous advancements in the areal density of HDDs are required. However, the areal density growth of current HDD technology is not proportional to the generated big-data, and hence, alternative technologies need to be researched. The media industry is eyeing a potential alternative, namely, heated-dot magnetic recording (HDMR). However, one of the major challenges of HDMR is the optimization of a patterning technique suitable for the mass production of media. Between physical and magnetic patterning processes, magnetic patterning suits more for mass production of the recording disks, which requires ion implantation of non-magnetic species through the hard mask. Moreover, the usage of self-assembly based lithography technique for fabrication of the hard mask enhances the suitability of magnetic patterning for the mass production of recording disks.

Therefore, we have studied the magnetic patterning of $\text{Co}_{80}\text{Pt}_{20}$ and FePt-C films. The self-assembly of polystyrene- polydimethylsiloxane di-block copolymer, which resulted in the patterns as small as ~20 nm, was utilized to fabricate the hard mask on the magnetic samples. The non-magnetic $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ions with high and low energies (HE and LE) were utilized. Regarding $\text{Co}_{80}\text{Pt}_{20}$ samples, the structural and magnetic properties are more sensitive to LE $^{14}\text{N}^+$ and HE $^{40}\text{Ar}^+$ implantations. The change in magnetic properties might be arising from an increase in A or a decrease in K_u or both. However, for FePt-C samples, all the implantation cases except LE $^{40}\text{Ar}^+$ ion implantation are significantly affecting the structural and magnetic properties.

Domain wall (DW) memory is another potential replacement of the present storage scheme. Although there has been a significant advancement in the writing, DW motion, and reading operations of the DW memory, there are still many unsolved issues. One of them is lack of controlled and reliable motion of DWs. The concept of artificial pinning sites has been proposed for controlled and reliable DW motion. The artificial pinning sites are formed by locally changing the physical shape or the magnetic properties of the memory unit. There have been a few proposed concepts of fabricating artificial pinning sites in the past. However, the most common problem with these concepts is the non-uniform pinning strength of pinning sites.

We have studied the concepts of (a) square wave shaped nanowire, and (b) local introduction of interfacial Dzyaloshinskii-Moriya interaction (*i*DMI) to fabricate pinning sites using micromagnetic simulations. The optimum pinning strength of the former was found to be limited in a small range of current density values. However, the latter shows the DW pinning in a large range of current density, geometrical, and magnetic parameters. Moreover, the pinning strength was found to be tunable as a function of the above-mentioned parameters. In addition, we have also observed the DW oscillations in the nanoscale *i*DMI region for a certain range of parameters. We have developed a theoretical model that excellently explains the nature of studied DW oscillations.

The operational power consumption is another problem for DW memories. The experimental demonstration of spin-orbit torque (SOT) driven DW motion has accelerated the research towards reducing the power of DW devices. However, there has not been much success in the past. Noticeably, the β -W has been the most common choice for the HM layer due to its high SOT efficiency. However, β -W invites the problem of larger defects in the ferromagnetic (FM) layer, and hence, larger operational power. Thus, we have studied a novel design, where an α or $(\alpha+\beta)$ -W layer has been sandwiched between β -W and FM layer. The proposed insertion reduced the intrinsic pinning of the FM layer and helped to reduce the current density by a factor of 10^4 for driving the DWs. This observation lays the foundation for ultra-low power DW devices. In addition to memory applications, these devices were further studied for emerging neuromorphic computing.

Acknowledgements

First and foremost, I would like to express my sincere gratitude towards my PhD supervisor Associate Professor S. N. Piramanayagam (Prem) for his incessant supervision during the course of this thesis. His continuous assessment and encouragement kept me motivated throughout this journey. He not only taught me critically evaluating and finding solutions of the research problems but also writing, presentation skills and all those things, which are prerequisites for an independent researcher. I also thank him for arranging my research trips to overseas universities to carry out experiments. I wish every PhD student gets a supervisor like him. I also thank Prof Shiva Prasad (retired Professor of IIT Bombay, India) for motivating me to join Prof Prem's group for my PhD degree. I would also like to thank my thesis advisory committee members Prof Christos Panagopoulos and Prof Rajdeep S. Rawat for their valuable inputs aimed towards the betterment of my research skills.

I am thankful to Singapore Government and Nanyang Technological University (NTU) for providing me an opportunity to carry out my PhD at NTU. Besides, I am grateful to them for financial assistance during my PhD.

Next, I would like to thank Facility for Analysis Characterisation Testing & Simulation (FACTS) lab for their help in the usage of X-ray diffraction (XRD) tools. A special mention to Dr Pio and Dr Weiling for several useful discussions during XRD experiments. I would also like to thank Nanyang NanoFabrication Center (N2FC) for helping me to utilize reactive ion etching (RIE), scanning electron microscope (SEM) and thickness profilometer machines during my experiments. I am also thankful to Division's central facilities (Division of Physics and Applied Physics, NTU) and corresponding staffs. A special thanks to Dr Hongjie (NTU staff in the past) for teaching me atomic force microscopy (and magnetic force microscopy) tool.

I would like to thank Dr Asokan and his team from Inter-University Accelerator Centre (IUAC), India for helping me in low and high energy ion implantations. I would also like to thank Prof Xiaoxi Liu and Chaung Ma from Shinshu university, Japan for their help in experiments during my short visit to Shinshu University. I am thankful to Prof Shin Saito and Dr Kim Kong Tham from Tohoku University for teaching and allowing me to use torque magnetometry and several fruitful discussions. Besides, I am thankful to Prof P. S. Anil Kumar and his students Sreekar, Banibrato, Ajesh and Monalisha for my smooth and pleasant research visit to Indian Institute of Science, India.

Next, I thank our collaborators Dr Sze Ter Lim and Dr Chung Hong Jing in Institute of Materials Research and Engineering (A*STAR) for active collaboration and several fruitful discussions.

A special appreciation to all the current and past members of SpinLab for providing a very friendly environment in the lab. I thank to Dr Tianli Jin for helping me in several experiments time to time. I also thank to Mr Sabpreet Bhatti for enormous number of professional and personal discussions. A special thanks to the new addition in the lab, Mr JianPeng, for your hard work over the last one year. I wish you a great PhD journey ahead. I also thank Nicolo, William, Laura and all the short-term students in our lab for their help and support. I thank Dr Mojtaba Ranjbar and Dr Surbhi Gupta for their valuable suggestions during early days of my PhD. I am fortunate that I share a great bond with Dr Varun, who taught me several lessons of research life. Thank you, Dr Varun for all your love and care. Although I never met Dr Shreya (past postdoc of SpinLab), I took inspiration from her work during my initial days of PhD. Thus, I thank Dr Shreya too.

I would also like to thank Dr Gan Weiliang and Mr Gerard Lim from Prof Lew Wen Siang's lab for discussions on MOKE and VSM, respectively. I thank IEEE magnetism society for financial support during my participation in the summer school in Spain.

Next, I would like to thank my friends Pratyush and Saurabh for being alongside whenever needed. I would also like to thank Poulomi, Abhishek, Bhartendu, Udvas, and Piyush for keeping my emotions controlled during the tough times.

Last but not the least, I would like to thank my grandparents, parents, sister and all the extended family members for their unconditional love, care, faith and everything they have given to me. This would not have been possible without their constant support. In the end, I would like to thank Almighty God.

Thank you!

(Durgesh Kumar)

Table of Contents

Abstract	8
Acknowledgements	10
Table of Contents	12
List of Abbreviations	19
List of Symbols	22
List of Publications	24
Chapter 1 Introduction.....	28
1.1 Background	28
1.1.1 Hard Disk Drives (HDDs)	28
1.1.2 Domain Wall Memory	32
1.2 Research Objectives	35
1.2.1 Effects of Ion Implantation on Structural and Magnetic Properties of Masked Magnetic Thin Films	36
1.2.2 Artificial Pinning Sites for Controlled Domain Wall Motion	37
1.2.3 Observation of Ultra-low Pinning Field for Energy Efficient Domain Wall Motion	40
1.3 Organization of Thesis	42
References.....	44
Chapter 2 Literature Review	49
2.1 Ferromagnetism	49
2.1.1 Exchange Energy	50

2.1.2	Anisotropy Energy	51
2.1.3	Zeeman Energy	51
2.1.4	Magnetostatic Energy	52
2.1.5	Stress Energy	52
2.1.6	Magnetostriction Energy	53
2.2	Hard Disk Drives	54
2.2.1	Longitudinal Magnetic Recording (LMR)	54
2.2.2	Perpendicular Magnetic Recording (PMR)	56
2.2.3	Energy-assisted Magnetic Recording (EAMR)	59
2.2.4	Heated-dot Magnetic Recording (HDMR)	61
2.3	Domain Wall Memory	63
2.3.1	Writing or Nucleation of Domains	65
2.3.2	Moving the Domain Walls in Racetracks	68
2.3.2.1	Magnetic Field Driven Domain Wall Motion	68
2.3.2.2	Electric Current Driven Domain Wall Motion	72
2.3.2.2.1	Spin-transfer Torque (STT)	72
2.3.2.2.2	Spin-orbit Torque (SOT)	76
2.3.3	Reading the Information	82
2.3.4	Challenges of Domain Wall memory Device	83
2.4	Conclusions	85
	References	86

Chapter 3 Experimental and Simulation Methods 94

3.1	Thin Film Deposition	95
3.1.1	Magnetron Sputtering	95
3.2	Characterization Tools	97
3.2.1	X-ray Diffraction (XRD)	97

3.2.2	Atomic Force Microscopy (AFM) and Magnetic Force Microscopy (MFM)	98
3.2.3	Vibrating Sample Magnetometer (VSM).....	101
3.2.4	Alternating Gradient Force Magnetometer (AGFM).....	102
3.2.5	Torque Magnetometry	103
3.2.6	Magneto-optic Kerr Microscopy.....	106
3.3	Device Fabrication: Fabrication of Dot Patterns for Ion Implantation Through Hard Mask	109
3.3.1	Self-assembly of Di-block Copolymers.....	109
3.3.2	Process Flow of Fabricating Dot Patterns.....	112
3.3.3	Optimization of Experimental Parameters for Self-assembly of Di-block Copolymers	113
3.3.4	Ion Implantation.....	116
3.3.5	Transport of Ions in Matter (TRIM)	117
3.4	Device Fabrication: Fabrication of Domain Wall Devices	118
3.4.1	Sample Cleaning and Prebaking	119
3.4.2	Spin Coating of Photoresist and Post-baking	119
3.4.3	Optical Lithography for Device Fabrication and Development	119
3.4.4	Ion Milling	121
3.4.5	Resist Removal, Sample Cleaning, and Prebaking.....	121
3.4.6	Spin Coating, Post-baking, Optical Lithography for Electrode Fabrication and Development.....	122
3.4.7	Electrode Deposition and Lift-off.....	122
3.5	Measurement Assembly for Domain Wall Devices	123
3.6	Sheet Resistance Measurements.....	125
3.7	Micromagnetic Simulations	127
3.7.1	Object-oriented Micromagnetic Framework (OOMMF).....	127
3.7.2	Mumax3	128

3.8	Conclusions	129
	References.....	130
Chapter 4	Effects of Ion Implantation on Structural and Magnetic Properties of Masked Magnetic Thin Films	134
4.1	Introduction.....	134
4.2	Tailoring the Structural and Magnetic Properties of Masked Co ₈₀ Pt ₂₀ Thin Films Using High and Low Energy Ion Implantation.....	137
4.2.1	Influence of Ion Implantation on Structural Properties	139
4.2.1.1	X-ray Diffraction (XRD) Results.....	139
4.2.1.2	TRIM Simulation and Cross-sectional TEM Results	141
4.2.2	Influence of Ion Implantation on Magnetic Properties	144
4.2.2.1	Magnetic Hysteresis (M-H) Loop Measurements.....	144
4.2.2.2	Magnetic Force Microscopy (MFM) Results	148
4.2.2.3	Micromagnetic Simulations	152
4.2.2.4	Estimation of Anisotropy Constant Using “Area Method” ...	153
4.2.3	Summary on CoPt samples	155
4.3	Ion Implantation Induced Disorder in FePt-C Thin Films.....	156
4.3.1	Influence of Ion Implantation on Structural Properties	158
4.3.1.1	X-ray Diffraction (XRD) Results.....	158
4.3.1.2	TRIM Simulation Results	160
4.3.2	Influence of Ion Implantation on Magnetic Properties	164
4.3.2.1	Magnetic Hysteresis (M-H) Loop Measurements.....	164
4.3.2.2	Torque Magnetometry Results.....	166
4.3.2.3	Magnetic Force Microscopy (MFM) Results	169
4.4	Conclusions.....	171
	References.....	172

Chapter 5	Artificial Pinning Sites for Controlled Domain Wall Motion.....	180
5.1	Introduction.....	180
5.2	Domain Wall Pinning in Square Wave Shaped Nanowire	185
5.2.1	Simulation Model.....	186
5.2.2	Domain Wall Dynamics in Reference Nanowire.....	187
5.2.3	Domain Wall Dynamics in Square Wave Shaped Nanowire.....	188
5.2.4	Summary	192
5.3	Domain Wall Pinning in a Nanowire with Nanoscale Interfacial Dzyaloshinskii–Moriya Interaction (<i>iDMI</i>) Region	193
5.3.1	Introduction.....	193
5.3.2	Simulation Model.....	195
5.3.3	Domain Wall Dynamics for Reference Nanowire	196
5.3.4	Domain Wall Dynamics in Samples with Nanoscale Interfacial Dzyaloshinskii–Moriya Interaction (<i>iDMI</i>) Region	197
5.3.5	Domain Wall Pinning at Multiple Pinning Sites.....	202
5.4	Domain Wall Oscillations in a Nanowire with Nanoscale Interfacial Dzyaloshinskii–Moriya Interaction (<i>iDMI</i>) Region	204
5.4.1	Dependence of Oscillation Frequency on External Stimulus and Geometrical Parameters	206
5.4.2	Dependence of Oscillation Frequency on Magnetic Parameters	208
5.4.3	Analytical Model for Observed Domain Wall Oscillations.....	210
5.4.4	Proposed Device	215
5.5	Conclusions.....	217
	References.....	219
Chapter 6	Observation of Ultra-low Pinning Field for Energy Efficient Domain Wall motion.....	226
6.1	Introduction.....	226

6.2	Optimization of Thickness for Dual Spin Hall Layers	231
6.2.1	Magnetic Measurements	231
6.2.2	Electrical Measurements	234
6.3	Results on Dual Spin Hall Layers	236
6.3.1	Characterizations of Thin Films	236
6.3.2	Characterizations of the Domain Wall Devices	241
6.3.3	Underlying Mechanisms of Observed Low Current Density	250
6.3.4	Comparison of Energy Consumption.....	259
6.4	Conclusions.....	262
	References	263
 Chapter 7 Ultra-low Power Domain Wall Devices for Neuromorphic Computing		269
7.1	Introduction.....	269
7.2	Synaptic Performance of Meander devices	273
7.2.1	Stochastic Synaptic Behaviour	273
7.2.2	Transition from Stochastic to Non-stochastic Synaptic Behaviour ..	274
7.2.3	Non-stochastic Synaptic Behaviour	276
7.2.4	Other Synaptic Performances	279
6.5	Conclusions.....	283
	References	284
 Chapter 8 Conclusions and Future Work		287
8.1	Conclusions of the Thesis	287
8.2	Future Works	293
8.2.1	Ultra-low Dimensional Domain Wall Devices	293

8.2.2	Experimental Observation of Domain Wall Dynamics in Square Wave Shaped Nanowire and Nanowire with Nanoscale Interfacial Dzyaloshinskii–Moriya interaction (<i>iDMI</i>) Region.....	294
8.2.3	On-chip Neurons and Synapses for Neuromorphic Computing	296
	References.....	299

List of Abbreviations

AFM	Atomic Force Microscopy
AGFM	Alternating Gradient Force Magnetometer
ASD	As deposited
BPMR	Bit-patterned Magnetic Recording
BPSI	Bits Per Square Inch
CMOS	Complementary Metal Oxide Semiconductor
CPU	Central Processing Unit
DBCP	Di-block Copolymer
DC	Direct Current
DI	De-ionized
DSA	Directed Self-assembly
DW	Domain Wall
EAMR	Energy-assisted Magnetic Recording
EB	Exabyte
EBL	Electron Beam Lithography
fcc	Face-centered Cubic
fct	Face-centered Tetragonal
FFT	Fast Fourier Transform
FM	Ferromagnetic
FORC	First-order Reversal Curves
GMR	Giant Magnetoresistance
GPU	Graphical Processing Unit
HAMR	Heat-assisted Magnetic Recording
hcp	Hexagonal-close Packed
HDD	Hard Disk Drive
HDMR	Heated-dot Magnetic Recording
HM	Heavy Metal
HP	High Pressure
<i>i</i> DMI	Interfacial Dzyaloshinskii-Moriya Interaction
IMA	In-plane Magnetic Anisotropy
IOTs	Internet of Things

IP	In-plane
IPA	Isopropyl Alcohol
LLG	Landau-Lifshitz-Gilbert
LMR	Longitudinal Magnetic Recording
LP	Low Pressure
MAMR	Microwave-assisted Magnetic Recording
MB	Megabyte
MFM	Magnetic Force Microscopy
M-H	Magnetic Hysteresis
MOKE	Magneto-optical Kerr Effect
MRAM	Magnetic Random-access Memory
MTJ	Magnetic Tunnel Junction
NIL	Nano-imprint Lithography
NIST	National Institute of Standards and Technology
OOMMF	Object-oriented Micromagnetic Framework
OOP	Out-of-plane
PDMS	Polydimethylsiloxane
PMA	Perpendicular Magnetic Anisotropy
PMR	Perpendicular Magnetic Recording
PS	Polystyrene
RAMAC	Random-access Method of Accounting and Control
REE	Rashba-Edelstein Effect
RF	Radio Frequency
RIE	Reactive Ion Etching
SA	Self-assembly
SAF	Synthetic Antiferromagnets
SHA	Spin Hall Angle
SHE	Spin Hall Effect
SIMS	Secondary Ion Mass Detector
SMR	Shingled Magnetic Recording
SNR	Signal to Noise Ratio
SOC	Spin-orbit Coupling
SOT	Spin-orbit Torque
SRIM	Stopping and Range of Ions in Matter

STO	Spin Torque Oscillator
STT	Spin-transfer Torque
TB	Terabyte
TDMR	Two-dimensional Magnetic Recording
TEM	Tunneling Electron Microscopy
THF	Tetrahydrofuran
TMR	Tunnel Magnetoresistance
TRIM	Transport of Ions in Matter
VRM	Vibrating Reed Magnetometer
VSM	Vibrating Sample Magnetometer
XRD	X-ray Diffraction
ZB	Zettabyte

List of Symbols

a	Lattice Constant
a_0	Bohr Radius
α	Damping Constant
α_R	Rashba Parameter
A	Exchange Constant
β	Non-adiabaticity Factor of STT
d	Interplanar Spacing for Crystalline Materials
δ	Domain Wall Width
Δ	Domain Width
D	DMI Parameter
H_{eff}	Effective Magnetic Field
e	Electronic Charge
ϵ_{ij}	Magnetoelastic Strain Tensor
E	Energy
f	Frequency of Oscillations
g	Landé g-Factor
$\hbar$	Reduced Planck's Constant
H	Applied Magnetic Field
H_c	Coercivity
H_d	Demagnetization Field
H_{DMI}	DMI Field
H_W	Walker Breakdown Field
J	Current Density
J^*	Exchange Coupling Constant
J_C	Charge Current
J_S	Spin Current
k	Wavevector
K_B	Boltzmann Constant
K_u	Anisotropy Constant
l_{ex}	Exchange Length
λ	Constant for Quantifying the Strength of Side-jump Scattering

μ	Mobility of Domain Walls
μ_0	Permeability of Free Space
μ_B	Bohr Magneton
m_e	Electronic Mass
M_r	Remanence
M_s	Saturation Magnetization
p	Angular Momentum of Conduction Electrons
P	Polarization of Spin Current
ρ	Resistivity
R	Resistance
σ	Polarization of Accumulated Spins in Spin-Hall Effect
S_i	Spin of i^{th} Electron
τ	Torque
θ_{SH}	Spin Hall Angle
χ	Flory-Huggins Interaction Parameter
Z	Atomic Number

List of Publications

Journal Publications

1. Ultra-low Power Domain Wall Device for Spin-based Neuromorphic Computing, **Durgesh Kumar**, Chung Hong Jing, Chan JianPeng, Tianli Jin, Lim Sze Ter, Rachid Sbiaa, S.N. Piramanayagam, arXiv:2007.12357, 2020 (In communication).
2. Enhanced Spin–Orbit Torque Efficiency in Pt/Co/Ho Devices Using a Second Interface Ho Layer, Tianli Jin, Wai Cheung Law, **Durgesh Kumar**, Feilong Luo, Qi Ying Wong, Gerard Lim, Xuan Wang, Wen Siang Lew and S. N Piramanayagam, APL Materials, 8(11), 111111, 2020.
3. Nickel ferrite embedded polyvinylidene fluoride composite based flexible magneto-electric systems, Avinash Chaurasiya, Pikeshe Pal, Joseph V Vas, **Durgesh Kumar**, S.N. Piramanayagam, Anil K Singh, Rohit Medwal and Rajdeep Singh Rawat, Ceramics International, 46(16), 25873-25880, 2020.
4. Spin Transfer Torque Induced Domain Wall Oscillations in Ferromagnetic Nanowire with a Nanoscale Dzyaloshinskii-Moriya Interaction Region, **Durgesh Kumar**, Pinaki Sengupta, Rachid Sbiaa, and S. N. Piramanayagam, Journal of Magnetism and Magnetic Materials, 507, 166807, 2020.
5. Domain Wall Motion Control for Racetrack Memory Applications, **Durgesh Kumar**, Tianli Jin, S. Al Risi, Rachid Sbiaa, Wen Siang Lew and S. N. Piramanayagam, IEEE Transactions on Magnetics, 55(3), 1-8, 2019.
6. Ion-Implantation-Induced Disorder in FePt-C Thin Films, **Durgesh Kumar**, Surbhi Gupta, Kim Kong Tham, Razia Nongjai, Shin Saito, Kandasami Asokan and S. N. Piramanayagam, IEEE Transactions on Magnetics, 55 (3), 1-5, 2019.
7. Nanoscale Compositional Modification in Co/Pd Multilayers for Controllable Domain Wall Pinning in Racetrack Memory, Tianli Jin, **Durgesh Kumar**, Weiliang Gan, Mojtaba Ranjbar, Feilong Luo, Rachid Sbiaa, Xiaoxi Liu, Wen Siang Lew and S. N. Piramanayagam, physica status solidi (RRL)-Rapid Research Letters, 12 (10) 1800197, 2018.
8. Broadband strip-line ferromagnetic resonance spectroscopy of soft magnetic CoFeTaZr patterned thin films, Surbhi Gupta, **Durgesh Kumar**, Tianli Jin, Razia Nongjai,

- Kandasami Asokan, Abhijit Ghosh, M. Aparnadevi, P. Suri and S. N. Piramanayagam, AIP Advances 8 (5), 056125, 2018.
9. Tailoring the structural and magnetic properties of masked CoPt thin films using ion implantation, **Durgesh Kumar**, Surbhi Gupta, Tianli Jin, Razia Nongjai, Kandasami Asokan and S. N. Piramanayagam, AIP Advances 8 (5), 056504, 2018.
 10. Modification of Structural and Magnetic Properties of Masked Co–Pt Films Induced by High-Energy Ion Implantation, **Durgesh Kumar**, Surbhi Gupta, Tianli Jin, Razia Nongjai, Kandasami Asokan and S. N. Piramanayagam, IEEE Magnetics Letters 9, 1-5, 2018.
 11. Domain Wall Memory: Physics, Materials and Devices, **Durgesh Kumar**, Tianli Jin, Rachid Sbiaa, Mathias Kläui, Subhankar Bedanta, Shunsuke Fukami, Dafine Ravelosona, See-Hun Yang, Xiaoxi Liu and S. N. Piramanayagam, (in communication).
 12. Domain wall dynamics in (Co/Ni)_n nanowire with anisotropy energy gradient for neuromorphic computing applications, William Wai Lum Mah, **Durgesh Kumar**, Tianli Jin, S.N. Piramanayagam (in communication)
 13. Domain wall device with a graded anisotropy as artificial neuron for neuromorphic computing, William Wai Lum Mah, Tianli Jin, **Durgesh Kumar**, S.N. Piramanayagam (in communication)

Conference Presentations

14. Ultra-low Power Domain Wall Device for Spin-based Neuromorphic Computing, **Durgesh Kumar**, Chung Hong Jing, Chan JianPeng, Tianli Jin, Lim Sze Ter, Rachid Sbiaa, and S.N. Piramanayagam, 9th MRS-S Conference on Advanced Materials, Virtual Meeting, Poster Presentation (Poster- 047). *The paper won the best multimedia poster award.*
15. Recent Developments in Spin-Based Neuromorphic Computing, S.N. Piramanayagam, **Durgesh Kumar**, Hong Jing Chung, Chan JianPeng, Tianli Jin, Rachid Sbiaa, Sze Ter Lim, 5th International Conference on Emerging Electronics-2020, Virtual Meeting (Host: IIT Delhi, India), *Invited Talk by Prof. Piramanayagam.*
16. Developments in Spin-orbit Torque based Magnetic Tunnel Junctions for Neuromorphic Computing, S.N. Piramanayagam, **Durgesh Kumar**, Hong Jing Chung,

- Chan JianPeng, Tianli Jin, Rachid Sbiaa, Sze Ter Lim, The Magnetic Recording Conference (TMRC)- 2020, Virtual Meeting, *Invited Talk by Prof. Piramanayagam.*
17. Spintronic devices for neuromorphic computing, S. N. Piramanayagam, Nicolo R. Sernicola, Tianli Jin, **Durgesh Kumar**, Wen Siang Lew, Rachid Sbiaa, Sze Ter Lim, The Magnetic Recording Conference (TMRC)- 2019, Minneapolis, MN (U.S.A.), *Invited Talk by Prof. Piramanayagam.*
 18. Spin Transfer Torque Induced Domain Wall Oscillations in Ferromagnetic Nanowire with a Nanoscale Dzyaloshinskii-Moriya Interaction Region, **Durgesh Kumar**, Pinaki Sengupta, Rachid Sbiaa and S. N. Piramanayagam, International Conference on Materials for Advanced Technologies (ICMAT)- 2019, Singapore, Oral Presentation (Symposium F).
 19. High-Frequency Domain Wall Oscillations in Ferromagnetic Nanowire with a Nanoscale Dzyaloshinskii-Moriya Interaction (DMI) Region, **Durgesh Kumar**, Pinaki Sengupta, Rachid Sbiaa and S. N. Piramanayagam, Materials Research Society- Singapore Summer School on Magnetic and Spintronic Materials-2019, Singapore, Poster Presentation (S-05).
 20. Domain wall pinning in square wave shaped magnetic nanowires, **Durgesh Kumar** and S. N. Piramanayagam, Joint MMM-Intermag-2019, Washington DC (U.S.A.), Poster Presentation (GN-04).
 21. Large Uniaxial Anisotropy $\text{Co}_{50}(\text{Pt}_{1-x}\text{Rh}_x)_{50}$ Thin Films for Microwave Assisted Magnetic Recording, **Durgesh Kumar**, Varun Chaudhary, Tianli Jin, Kim Kong Tham, Shin Saito and S. N. Piramanayagam, Joint MMM-Intermag-2019, Washington DC (U.S.A.), Poster Presentation (EK-09).
 22. Domain wall pinning at novel geometrical design in magnetic nanowire, **Durgesh Kumar**, Varun Chaudhary, Tianli Jin and S. N. Piramanayagam, Magnetics symposium IEEE Magnetics Society Singapore Chapter- 2018, Oral Presentation (student presentations-08).
 23. Ion implantation induced disorder in FePt-C films, **Durgesh Kumar**, Surbhi Gupta, Kim Kong Tham, Razia Nongjai, Shin Saito, Kandasami Asokan and S. N. Piramanayagam, The Magnetic Recording Conference (TMRC)- 2018, Milpitas, CA (U.S.A.), Poster Presentation (P1-10).
 24. Large Uniaxial Anisotropy $\text{Co}_{50}(\text{Pt}_{1-x}\text{Rh}_x)_{50}$ Thin Films for Microwave Assisted Magnetic Recording, **Durgesh Kumar**, Tianli Jin, Kim Kong Tham, Shin Saito and S.

- N. Piramanayagam, The Magnetic Recording Conference (TMRC)- 2018, Milpitas, CA (U.S.A.), Poster Presentation (P1-14).
25. Tailoring the structural and magnetic properties of masked CoPt thin films using ion implantation, **Durgesh Kumar**, Surbhi Gupta, Tianli Jin, Razia Nongjai, Kandasami Asokan and S. N. Piramanayagam, Conference on Magnetism and Magnetic Materials (MMM)- 2017, Pittsburgh (U.S.A.), Poster Presentation (BQ-04).
 26. Study of ion implantation based magnetic patterning process for heated-dot magnetic recording” **Durgesh Kumar**, Surbhi Gupta, Tianli Jin, Razia Nongjai, Kandasami Asokan, and S. N. Piramanayagam, Magnetism symposium IEEE Magnetism Society Singapore Chapter- 2017, Oral Presentation (student presentations-09). ***The paper won the second prize in best presentation competition.***
 27. Modification of Structural and Magnetic Properties of Masked Co–Pt Films Induced by High-Energy Ion Implantation, **Durgesh Kumar**, Surbhi Gupta, Tianli Jin, Razia Nongjai, Kandasami Asokan and S. N. Piramanayagam, IEEE Magnetism Society Summer School- 2017, Universidad Internacional Menéndez Pelayo, Santander (Spain), Poster Presentation (student presentations).

Others

28. Easter eggs on a sandy beach, **Durgesh Kumar**, Nicolo R. Sernicola, S. N. Piramanayagam, Intermag- 2018, Singapore, Art in Magnetism (participation).
29. Perm’algae, **Durgesh Kumar**, Nicolo R. Sernicola, Varun Chaudhary and S. N. Piramanayagam, Joint MMM-Intermag- 2019, Art in magnetism (participation).

Chapter 1. Introduction

1.1. Background

A massive amount of digital data is produced every minute in today's world. With the advent of technological platforms such as web, media, communication, and internet of things (IOTs), the trend of digital data has increased rapidly in recent years [1]. According to a report, 44 ZB data is expected to be produced globally in 2020 [2, 3]. This number was approximately 2 ZB back in 2010 [2, 3]. Moreover, the global data produced in a calendar year is expected to reach ~175 ZB in 2025 (equivalent to ~463 EB data every day in 2025) [2, 3]. For more than two decades, magnetic hard disk drives (HDDs) have been playing the major role of data storage devices. Currently, magnetic HDDs offer the storage of approximately 75-80% of the total data [2]. Although alternative devices such as solid-state drives are replacing HDDs in laptops and other portable devices, a large portion of the global data is still stored in HDDs. For example, it is expected that in 2025, HDDs will still store ~50-60% of the total digital data [2]. Therefore, continuous growth in areal density, a measure of the storage capacity of HDDs, is essential.

1.1.1. Hard Disk Drives (HDDs)

The underlying principle of HDDs is magnetic recording [4, 5], where a certain magnetization of a group of grains (a bit) is considered as an information “0” or “1”. The magnetization from one direction to the other is changed by applying an external magnetic field. The principles of magnetic recording were described by Oberlin Smith in 1878 [6]. However, the first device, namely “Telegraphone”, based on these principles was patented in 1898 by a Danish researcher Valdemar Poulsen [6]. Although the design of the machine was ready, the device was never commercialized after a hard work of 20 years and abandoned in 1918. After 10 years, the first magnetic tape-based storage was introduced, which was later in 1934 named magneto-phone [7]. The era of modern magnetic recording started when IBM introduced the first-ever HDD named random-access method of accounting and control (RAMAC) hard disk drive in 1956

[8]. The total storage capacity of these devices was 5 MB with an areal density of 2 kbps (kilobits per square inch). The history of magnetic recording is briefly illustrated in Figure 1.1 (a).

Figure 1.1. The (a) history and (b) technological roadmap (adopted from [9, 10]) of magnetic recording technology.

Until the early 1980s, when the thin film deposition technique was initiated [7], the magnetic recording was based on the particulate media. In this media, the magnetic particles were pasted on a substrate using glue-like solvents. Soon after the development of thin film deposition techniques, the media industry shifted to thin films. The magnetic recording was then based on the longitudinal magnetic recording (LMR), where magnetizations point in a direction parallel to the disk surface [11]. A maximum areal density of 130 Gbps was achieved using the LMR technology [12]. Conventionally, a reduction in bit (rather grain) size is considered as the way to increase the areal density. However, reduction in grain size (beyond the grain size for the

areal density of 130 Gbpsi) was hindered by several issues such as superparamagnetic limit, transition noise, demagnetizing field, and limited writing field [7].

To overcome these obstacles the idea of perpendicular magnetic recording (PMR) technology was proposed. Although the idea of PMR was proposed in 1975 by researchers from Tohoku University, it took almost 30 years for the industry to ship the first commercial HDD based on PMR technology [13]. In PMR media, the magnetic layer with magnetization pointing perpendicular to the disk surface is utilized. A maximum areal density of 1 Tbpsi was achieved using this technology. A further increase in areal density was impeded by the famous media trilemma problem [7, 12]. The media trilemma problem refers to the problem of (a) signal to noise ratio (SNR), (b) thermal stability, and (c) writability [14].

The areal density was pushed slightly by using the concept of two-dimensional magnetic recording (TDMR) and/or shingled magnetic recording (SMR) technologies [15, 16]. The TDMR technology was based on making the narrower data tracks on HDD media and, hence increasing the areal density. In this situation, the reading may become difficult, which can be avoided by using an array of reading heads to read the information either from one track or several neighboring tracks. This technology was predicted to increase the areal density by 5-10% of existing areal density [17]. However, SMR is a technology that helps in developing the narrower tracks. Here the data tracks are developed in such a way that the adjacent tracks overlap one another in going towards cross-track direction [18].

To achieve a higher jump in areal density, the concept of energy-assisted magnetic recording (EAMR) was proposed [19]. The EAMR technology can be classified into (a) heat-assisted magnetic recording (HAMR), [20] and (b) microwave-assisted magnetic recording (MAMR) [21-23]. The HAMR technology uses an additional laser source incorporated with the writing head. While writing, the laser source heats the bit to a temperature higher than the Curie temperature. This effectively reduced the anisotropy constant of the bit, which needs to be

altered. Therefore, a lower writing field is needed to write the information. Therefore, it allows the usage of a material with higher magnetic anisotropy than the existing recording media and also allows further reduction in the grain size. Recently, Seagate has introduced the HAMR based HDD in the market with an areal density of 2 Tbps [9]. The upcoming versions of HAMR based HDD may show an areal density of 4 Tbps or more. On the other hand, Western Digital and Toshiba anticipated in MAMR based HDD and predicted an areal density closer to 4 Tbps [24]. This technology is expected to be introduced in the market towards the end of the present year [25, 26]. In MAMR, a spin-torque oscillator (STO), which can generate a microwave field, is attached to the writing head. The generated microwave field can be tuned to the resonant frequency of the recording material and hence makes the switching of magnetization easier.

However, the predicted growth in areal density may not be enough to meet the requirement of exponentially increasing digital data. Therefore, researchers are looking for a successor technology. As indicated in Figure 1.1 (b), the heated-dot magnetic recording (HD MR) is the next recording technology and is predicted to increase the areal density to 10 Tbps and beyond [10]. The HD MR technology combines the bit-patterned magnetic recording (BPMR) and HAMR. All the recording technologies discussed till now use a continuous layer as a recording medium. This is composed of magnetic grains separated by less magnetic (or non-magnetic) grain boundaries. Several magnetic grains constitute one bit of recorded information. However, in BPMR, the continuous media is patterned into the magnetic islands. The magnetic islands are separated by the non-magnetic regions. The thermal stability of BPMR depends on the volume of the one magnetic island (unlike continuous media where thermal stability depends on the volume of a magnetic grain) [27]. In general, the volume of a magnetic island, which constitutes one bit of information, is much larger than that of a magnetic grain and hence effectively increases the thermal stability. Moreover, it provides a scope of further decreasing

the size of magnetic islands (compared to island size in targeted HDMR media) and hence further increase in areal density. It is important to note that thermal stability defines the lifetime of data storage devices [7, 12], which will be discussed in detail in the next chapter.

However, one of the major challenges of the HDMR technology is the optimization of a suitable lithography technique for the mass production of these devices. There have been two main approaches for the fabrication of patterned media for HDMR, namely, (a) physical patterning and (b) magnetic patterning [27]. In the physical patterning process, the dot patterns are prepared on a resist layer using a lithography tool such as electron beam lithography (EBL), nano-imprint lithography (NIL), and directed self-assembly. Subsequently, the patterns are transferred from the resist layer to the recording layer using the dry etching process. However, in magnetic patterning, first, the magnetic patterns are formed on the resist layer. Then, the ion implantation of non-magnetic ions is performed through the mask patterns. Magnetic patterning is advantageous over physical patterning as magnetic patterning (i) locally modifies the magnetic properties, (ii) minimizes the need for disk planarization, (iii) is suitable for mass production of recording devices to meet consumer demand, and (iv) is cheaper and less time-consuming. Owing to these advantages, we have followed the route of magnetic patterning.

1.1.2. Domain Wall Memory

The domain wall (DW) racetrack memory is another potential alternative to the current data storage scheme. It can facilitate a large storage capacity to meet the demands of the rapidly growing digital data [28]. The idea of DW memory was conceived by IBM researcher S. Parkin in 2008 [29]. Since then, the field of DW memory has evolved a lot and a summary of major breakthroughs is presented in Figure 1.2 [30]. A DW memory device is a parallel array of several ferromagnetic nanowires lying in the plane or perpendicular to the plane of the substrate. The data is encoded in the magnetic domains magnetized in a certain direction. These magnetic domains can be moved back and forth to write further information. Like any other

storage device, the working principle of DW memory is also categorized into, writing, storing, moving and reading [31].

Figure 1.2. The evolution of domain-wall racetrack memory devices (partially adopted from [30]).

The writing can be performed using (a) nucleation pads, (b) local Oersted field from the non-magnetic metal crossbar, and (c) spin-transfer torque (STT) from a magnetic tunnel junction (MTJ) structure [32-34]. For the case of nucleation pads, large nucleation pads are fabricated at one or both ends of the nanowire. After saturating the device with a large magnetic field, a reverse field, called the nucleation field, is applied to insert a DW in the nanowire. This method was mostly used for the nanowires with in-plane magnetic anisotropy (IMA). Secondly, for the case of nanowires with perpendicular magnetic anisotropy (PMA), a non-magnetic metal crossbar is fabricated at one end of the nanowire. When a current is sent through this crossbar, an Oersted field surrounding the crossbar is generated. This local field can be used to deterministically nucleate the domains. However, the STT method is the most popular choice for nucleating the domains. At one end of the nanowire, an MTJ is fabricated whose free layer is constituted by the nanowire (memory unit). When a current is sent through the MTJ, the spin-polarized current reaches the local area of the soft layer (area beneath the MTJ). This spin-

polarized current switches the magnetization of the nanowire locally and hence inserts a DW in the nanowire.

Once the DW is inserted in the ferromagnetic nanowire, it can be moved using several means. Conventionally, the magnetic field was used to move the DWs [35]. However, this is a non-local method of moving the DWs and is not suitable for devices at a chip level. Later, it was found that the electrical currents can move the DWs along a nanowire. The electrons from the applied current interact with the electrons of the host atoms and transfer angular momentum according to the laws of adiabatic and non-adiabatic torques of STT [36, 37]. Further research in the field led to the popularization of the phenomenon called spin-orbit torque (SOT) induced DW motion [38]. When a current is sent through the heavy metal (HM) with high spin-orbit coupling, deposited in the vicinity of the ferromagnetic (FM) layer, the FM layer experience a torque called SOT. The bulk spin-Hall effect and interfacial Rashba Edelstein effect is the reason for such a torque. The details of these mechanisms will be discussed in the next chapter. The research studies revealed that the SOT is more power efficient technique than STT and hence is more favorable for DW devices [39].

The presence or absence of particular information is sensed by the reading MTJ, which is fabricated at the other end of the nanowire (one end is occupied by a writing element) [34, 40-42]. Similar to the writing element, the FM nanowire makes the soft layer of the MTJ. Depending on the relative orientations of the magnetizations in the fixed and free layer, a particular bit of information is recognized. In summary, the information is first written and then moved to the parking area and parked. When one needs to read that information, it is moved in the vicinity of the reading element. Therefore, unlike the case of HDD, the writing and reading elements are static in DW memory.

Although the research community has made tremendous advancements after the advent of DW memory, there are still some challenges that need to be addressed. These are (a) controlled and

reliable motion of DW, (b) thermal stability of the DW memory, (c) large intrinsic pinning defects, (d) large operational power, (e) larger size of the DW memory devices, and (f) lack of efficient motion of multiple DWs [28]. We have picked the problems of controlled and reliable DW motion, power consumption, and intrinsic pinning field in this thesis.

1.2. Research Objectives

Keeping the trend of big data in mind, we need to research potential alternatives to the present storage scheme. As mentioned earlier, HDMR is the successor of the current HDD technology. However, considering the urgency of coming up with new big data storing devices, researchers are developing alternatives to HDDs. Between a couple of racetrack ideas, DW motion-based racetrack memory is the preferred contender. Since the present thesis focuses on memory and storage research, therefore, we have studied the self-assembly and ion-implantation based magnetic patterning process in the initial phases. While we have studied DW memory devices in the later phases of the thesis. The experimental and simulation tools have been utilized to study and understand the physical and technological insight into the proposed concepts. We have deposited all the thin films used in this study using the magnetron sputtering process. These thin-film samples have been studied at two stages. Firstly, the thin film samples were characterized using several experimental tools to understand the structural, electrical, and magnetic properties. In the next stage, the devices from thin film samples were fabricated and studied. Besides, the micromagnetic simulations using the object-oriented micromagnetic framework (OOMMF) and Mumax3 were performed to theoretically probe the corresponding characteristics. The details of the experimental and simulation tools and the corresponding results will be discussed in the upcoming chapters of this thesis.

1.2.1. Effects of Ion Implantation on Structural and Magnetic Properties of Masked Magnetic Thin Films

Inspired by the advantages, we have pursued the path of the magnetic patterning process. The patterns on the resist layer, which will subsequently act as a hard mask, can be fabricated using EBL, NIL, and self-assembly of di-block copolymers (DBCP) techniques. Amongst these lithography techniques, the self-assembly process is the most advantageous one [27]. The self-assembly process can result in smaller patterns (as small as 3 nm) with minimum size distribution in comparison to other lithography techniques [43]. On top of that, the self-assembly process is time and cost-efficient, which is one of the major requirements of the media industry [27]. Therefore, we have performed the self-assembly of polystyrene (PS)-polydimethylsiloxane (PDMS) DBCP to fabricate the mask patterns on the magnetic samples. The reason for choosing the PS-PDMS DBCP is its high Flory-Huggins interaction parameter and easier selective etching [27]. These parameters help in obtaining smaller and uniform patterns.

After patterning the hard mask, we have performed the ion-implantation of non-magnetic $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ions. These non-magnetic ions are expected to mostly interact with the recording material in the unmasked region and hence changing the structural and magnetic properties locally. We have chosen two distinct energies, high (40 keV for $^{14}\text{N}^+$ and 100 keV for $^{40}\text{Ar}^+$ ions) and low (7.5 keV for $^{14}\text{N}^+$ and 4.5 keV for $^{40}\text{Ar}^+$ ions) energy for both the ions. The reason for choosing the high and low energy is to study the effect of ion implantation when the ions stop in the magnetic layer and in the layers below the magnetic layer, respectively.

Most importantly, we have performed the magnetic patterning on hexagonal closed packed $\text{Co}_{80}\text{Pt}_{20}$ and $L1_0$ face-centered tetragonal FePt-C magnetic samples. Both these materials are the supreme choices of media researchers due to their high perpendicular magneto-crystalline anisotropy [7]. After fabricating the samples, we performed the structural and magnetic analysis of the reference as well as patterned samples using experimental and micromagnetic

simulation tools. The details of the results and conclusions drawn from this study will be presented in chapter 4.

Novelty, Practical Industrial Significance, and Limitations: As highlighted, magnetic patterning is advantageous over physical patterning in several ways. The magnetic patterning involves ion-implantation of non-magnetic species through a hard mask, which requires a suitable lithography technique. The usage of conventional lithography methods such as electron beam lithography and nano-imprint lithography are appreciated at laboratory level research. However, they cannot meet the industrial demands in terms of cost and time. Therefore, a lithography method suited for mass production needs to be developed. Thus, we are utilizing self-assembly of di-block copolymers, which can result in patterns as small as 3 nm with good uniformity of patterns. With the exiting capabilities, we can fabricate patterns of sizes ~20 nm with good uniformity. The novelty of this study is using self-assembly as a technique for fabricating the hard mask. Additionally, we have researched ions with two distinct energies; (a) low energy and (b) high energy. The reason for doing this is to study the effects of ion implantations when non-magnetic ions stay with the magnetic contents in the magnetic layer and when they pass through the magnetic layer and stay in layers below the magnetic layer. Although this method is of great importance for practical applications and we have witnessed significant results, further optimizations are required.

1.2.2. Artificial Pinning Sites for Controlled Domain Wall Motion

The DW motion is stochastic in nature [44-46]. This means it is very difficult to stop a DW at a certain position in a controlled and reliable manner. However, for DW memory devices, the controlled DW motion and parking is very important. This is because one wants the uniform bit sizes to retain the stored information. Moreover, the stochastic DW motion may result in the collapse of neighboring domains and loss of the stored information. The controlled and reliable DW motion can be achieved by using the concept of artificial pinning sites [28]. In

addition to controlled DW motion, the artificial pinning sites also help in improving the thermal stability of the DW device [47]. In artificial pinning sites, the energy landscape for DW motion is locally altered at specific positions. This can be done either by changing the shape or magnetic properties at the positions of pinning sites [28]. Thus, depending on the nature of pinning sites, they can be categorized into (a) geometrical (refers to the change in physical shape) and (b) non-geometrical (refers to the change in magnetic properties) pinning sites. This part of DW research is quite young and only a few concepts have been studied.

In the past, researchers have proposed the concept of notches (or anti-notches) and staggered nanowire and demonstrated successful pinning of DWs [29, 48]. These concepts fall under the geometrical pinning site category. On the other hand, the non-geometrical pinning using local metal diffusion, ion implantation, and local exchange coupling has been studied [49-52]. However, the most common problem with these concepts is the non-uniform pinning strength of pinning sites. This may result in the non-deterministic pinning of DWs at successive pinning sites. To achieve the deterministic pinning, the increasing strength of pinning sites in going from the present pinning site to the next has been proposed. This, however, needs an increasing magnitude of external stimulus for driving DWs and hence limits the number of pinning sites. This, in turn, will limit the storage capacity of the DW devices. In order to avoid these problems, we have studied two distinct concepts of fabricating pinning sites.

The first concept is based on fabricating the square wave shaped nanowire in place of the straight nanowire. The Laplace pressure at the corners of the square wave shaped nanowire opposes the torque from the driving current and hence helps in DW pinning [53, 54]. We have performed detailed micromagnetic simulations to study this concept and the results will be discussed in chapter 5.

The second design, however, is based on introducing the interfacial Dzyaloshinskii-Moriya interaction (*i*DMI) in a nanoscale region. The effective field from the nanoscale *i*DMI changes

the energy landscape for the DW in the nanoscale region [55, 56]. In other words, the introduction of nanoscale *i*DMI creates either a step or well in the total energy of the system (depending on the sign of *i*DMI) in this nanoscale region and hence stops the DW. The DW can only be depinned by applying a higher magnitude of the driving force. We have studied this concept in detail and the results are discussed in the second half of chapter 5.

Novelty, Practical Industrial Significance, and Limitations: The field of artificial pinning sites for DW memory devices is not well explored. In general, geometrical pinning sites show variation in the size of pinning sites. This results in the stochastic nature of DW depinning from the successive pinning sites. Therefore, non-geometrical pinning sites are preferred. However, for the better practical performance of these pinning sites, new methods need to be researched. This way we think that pinning sites based on nanoscale *i*DMI show significant importance. In particular, this is a method based on magnetic modifications, where issues like lateral straggle and interlayer diffusion (issues for ion-implantation and metal diffusion) can be avoided. And hence, there is more freedom in shrinking the DW devices.

In terms of practical significance, we believe that the pinning sites in the form of nanoscale *i*DMI can directly be implemented in DW memory devices. Moreover, they can be used in realizing stochastic and non-stochastic synapses for neuromorphic computing applications. The virtue of DW oscillations is also significant for synaptic applications. However, a few more research iterations are required to make it fully practical.

The major limitation of fabrication of pinning sites using square wave shaped nanowire design is the limited window of current density where significant pinning was observed. However, the only limitation for DW pinning in nanoscale *i*DMI could be the experimental verification of the proposed idea. Nevertheless, we believe that this idea can easily be realized in experiments. In fact, we are already working on Pt-based materials to realize the experimental counterpart of this study. Since DW oscillations in the nanoscale *i*DMI region were observed for

significantly high values of DMI. At present, it may be slightly difficult to find heavy metal that can provide such a high DMI. However, theoretical studies suggest that such higher DMI values are possible and we are trying to realize materials with such DMI values.

1.2.3. Observation of Ultra-low Pinning Field for Energy Efficient Domain Wall Motion

Another issue with the DW memory is the large operational power. Several methods such as utilizing gate voltage, simultaneous application of electrical current and magnetic field (along easy axis direction), simultaneous application of currents current in crossbar geometry, etc. have been studied for reducing the power consumption [57-63]. However, the reported reduction in power consumption is not significant [57-63]. More recently, it has been demonstrated that SOT on DWs from adjacent HM layer results in more efficient DW motion with a low power consumption [64]. In general, the efficiency of SOT (and hence power consumption) is quantified by the spin Hall angle (θ_{SH}), which is the ratio of the charge current flowing in HM to the spin current diffused in the adjacent FM layer [65]. Several materials such as Au, Nb, Mo, Pd, Pt, Ta, W, IrCu, CuBi, $W_{1-x}Ta_x$, etc. have been studied for increasing θ_{SH} [66-69]. Out of these materials β -W has been the choice of many research studies due to higher θ_{SH} . The reported θ_{SH} ranges from ~ 0.2 to ~ 0.6 for β -W, with the resistivity values ranging from ~ 100 to $\sim 600 \mu\Omega \text{ cm}$ [70-72]. In general, a high Ar gas pressure and low power during the deposition of W layer result in β -phase in W thin films [72]. An increment in resistivity in a range from 100 to 300 $\mu\Omega \text{ cm}$ is reported to be the basis for increasing the θ_{SH} [73]. However, high resistivity will add to higher power consumption, which is not desirable. Moreover, the β -W forms a granular structure, which may be translated to the FM layer. This may result in an FM layer with higher defects and hence a higher pinning field. To avoid this situation, we have studied a novel material engineering where a W layer, deposited at low Ar gas pressure, was sandwiched between β -W layer and FM layer. The results of this study are

presented in detail in chapter 6 of this thesis. The proposed spin Hall layer is also extended to realize low power neuromorphic devices. These extended results are presented in chapter 7 of this thesis.

Novelty, Practical Industrial Significance, and Limitations: As discussed, β -W exhibits the highest spin-Hall angle amongst conventional spin Hall materials. However, β phase in W is achieved when W is deposited at a low deposition rate and high Ar gas pressure. Such deposition conditions result in granular W film. This results in the generation of defects in the thin FM layer, which causes stochastic DW motion and increases the critical current density. This in turn increases the operational power of DW devices. To avoid these, we have proposed to deposit a low power and low Ar gas deposited W film between β -W and ferromagnetic layer. By doing this, we have achieved ultra-low-power DW memory devices. Moreover, our DW devices mimic synaptic functions for emerging neuromorphic computing.

As far as practical significance is concerned, we have shown the functions of the DW memory unit. The low power virtue can be combined with other functions of DW memory for its practical realization. However, some problems like thermal stability (lifetime of stored information) and synchronous motion of multiple DWs hinder the practical realization of DW devices. From applications in neuromorphic computing, we are taking further steps to deposit industrial level magnetic tunnel junction (MTJ) structures where the existing structure will act as a free layer and spin Hall layer. Once this is achieved, we plan to realize three-terminal low power synthetic neurons and synapses.

One limitation of the current study is that the magnetic field (for instance, positive) supports DW motion only in one direction (corresponding to the positive magnetic field, motion related to positive current). To realize the DW motion in the opposite direction the supporting magnetic field needs to be reversed. It is noticeable that the external magnetic field must be avoided in practical applications. Meanwhile, the required magnetic field is very small and can be realized

using the stray field from the reference layer of MTJ, which will be unidirectional. We are making efforts to engineer our DW devices to achieve DW motion in opposite direction at low power in existing DW devices.

1.3. Organization of Thesis

The present thesis is organized in the following manner,

Chapter 1 delivers the background and current state of art in the field of magnetism (or spintronics)-based memory and storage devices. Moreover, it highlights the present obstacles and how can they be conquered. In the later part of the chapter, the goals of this thesis have been stated.

Chapter 2 details the literature survey of the field, which serves in understanding the current state-of-art as well as the results presented in the thesis.

Chapter 3 lists the experimental and simulation techniques used in this thesis. It starts with a brief introduction of the thin film deposition and characterization techniques. Later, it discloses the discussion about the lithography techniques used to fabricate patterns in the present research work. Towards the end, it discusses the measurement assembly for DW devices and micromagnetic simulators.

Chapter 4 presents the results of magnetic patterning on $\text{Co}_{80}\text{Pt}_{20}$ and FePt-C magnetic alloys. In the first half, the results of magnetic patterning on the $\text{Co}_{80}\text{Pt}_{20}$ films have been discussed. However, in the second half, the results of magnetic patterning of the FePt-C thin films have been presented. Both the subchapters start with the discussion on the process followed during the magnetic patterning and then unveil the characteristics of the patterned samples.

Chapter 5 discusses the results of DW pinning (or DW pinning and oscillations) in square wave shaped nanowire and nanowire with nanoscale i DMI region. First, the micromagnetic simulation results of DW pinning in the square wave shaped nanowire have been presented. Subsequently, the results of DW pinning and DW oscillations in the nanowire with the

nanoscale *i*DMI region have been discussed. Besides, the analytical model prepared to explain the simulation results of DW oscillations in the nanoscale *i*DMI region have also been discussed.

Chapter 6 is dedicated to the research discussion about our results on ultra-low-power DW motion induced by SOT from the dual W layer. Initially, the stacks with a single W layer were studied to optimize the preparation parameters for dual-layer samples. Subsequently, the results at the thin film as well as device level for the dual W spin Hall layer samples have been discussed.

Chapter 7 is an extension of chapter 6 that highlights the usefulness of studied spin Hall layer in realizing low power synaptic devices. The results of stochastic, intermediate, and non-stochastic DW motion in the meander nanowire have been presented. Besides, more specific synaptic functions of proposed meander device have been discussed towards the end of this chapter.

Chapter 8 gives a summary of the thesis in the first half. In the latter half, it discusses the future outlooks of the research results presented in the thesis.

References

- [1] <https://ied.eu/project-updates/the-4-industrial-revolutions/>.
- [2] <https://www.seagate.com/files/www-content/our-story/trends/files/idc-seagate-dataage-whitepaper.pdf>.
- [3] <https://www.weforum.org/agenda/2019/04/how-much-data-is-generated-each-day-cf4bddf29f/>.
- [4] J. L. Tsai, G. Varvaro, C. Pi, J. Y. Chen and Y. T. Wu, "CoCrPt@(TiO₂, CoO) granular thin films grown on Ru/Ni_xPd_{100-x}/NiTa (x= 20, 50, 80)," *Thin Solid Films*, vol. 693, p. 137683, 2020.
- [5] H. N. Bertram, *Theory of magnetic recording*. Cambridge University Press, 1994.
- [6] E. D. Daniel, C. D. Mee and M. H. Clark, *Magnetic recording: the first 100 years*. John Wiley & Sons, 1998.
- [7] S. Piramanayagam and T. C. Chong, *Developments in data storage: materials perspective*. John Wiley & Sons, 2011.
- [8] A. Moser, K. Takano, D. T. Margulies, M. Albrecht, Y. Sonobe, Y. Ikeda, S. Sun and E. E. Fullerton, "Magnetic recording: advancing into the future," *Journal of Physics D: Applied Physics*, vol. 35, no. 19, p. R157, 2002.
- [9] <https://blog.seagate.com/craftsman-ship/hamr-next-leap-forward-now/>.
- [10] http://idema.org/?page_id=5868.
- [11] D. Weller and M. F. Doerner, "Extremely high-density longitudinal magnetic recording media," *Annual review of materials science*, vol. 30, no. 1, pp. 611, 2000.
- [12] S. Piramanayagam, "Perpendicular recording media for hard disk drives," *Journal of Applied Physics*, vol. 102, no. 1, p. 2, 2007.
- [13] S.-I. Iwasaki and K. Takemura, "An analysis for the circular mode of magnetization in short wavelength recording," *IEEE Transactions on Magnetics*, vol. 11, no. 5, pp. 1173, 1975.
- [14] H. Richter and S. Harkness, "Media for magnetic recording beyond 100 Gbit/in.²," *MRS bulletin*, vol. 31, no. 5, pp. 384, 2006.
- [15] https://documents.westerndigital.com/content/dam/doc-library/en_us/assets/public/western-digital/collateral/white-paper/white-paper-shingled-magnetic-recording-helioseal-technology.pdf.
- [16] <https://www.forbes.com/sites/tomcoughlin/2018/04/29/two-dimensional-magnetic-recording-and-other-hdd-news/#37e25fc921d0>.
- [17] <https://www.anandtech.com/show/10470/the-evolution-of-hdds-in-the-near-future-speaking-with-seagate-cto-mark-re/4>.
- [18] R. Wood, M. Williams, A. Kavcic and J. Miles, "The feasibility of magnetic recording at 10 terabits per square inch on conventional media," *IEEE Transactions on Magnetics*, vol. 45, no. 2, pp. 917, 2009.
- [19] B. Pathem, X.-C. Guo, F. Rose, N. Wang, K. Komvopoulos, E. Schreck and B. Marchon, "Carbon overcoat oxidation in heat-assisted magnetic recording," *IEEE transactions on magnetics*, vol. 49, no. 7, pp. 3721, 2013.
- [20] H. Katayama, M. Hamamoto, J. Sato, Y. Murakami and K. Kojima, "New developments in laser-assisted magnetic recording," *IEEE Transactions on Magnetics*, vol. 36, no. 1, pp. 195, 2000.
- [21] J. G. Zhu, X. C. Zhu and Y. H. Tang, "Microwave assisted magnetic recording," *IEEE Transactions on Magnetics*, vol. 44, no. 1, pp. 125, 2008.
- [22] Y. H. Tang and J. G. Zhu, "Narrow Track Confinement by AC Field Generation Layer in Microwave Assisted Magnetic Recording," *IEEE Transactions on Magnetics*, vol. 44, no. 11, pp. 3376, 2008.

- [23] S. Batra and W. Scholz, "Reduction in Switching Field for a Granular Perpendicular Medium Using Microwave Assisted Magnetic Recording," *IEEE Transactions on Magnetics*, vol. 44, no. 11, pp. 3392, 2008.
- [24] <https://spectrum.ieee.org/computing/hardware/lasers-vs-microwaves-the-billiondollar-bet-on-the-future-of-magnetic-storage>.
- [25] <https://blog.westerndigital.com/the-need-for-energy-assisted-recording-technology/>.
- [26] https://www.theregister.co.uk/2018/12/07/toshiba_goes_to_mamr/.
- [27] R. A. Griffiths, A. Williams, C. Oakland, J. Roberts, A. Vijayaraghavan and T. Thomson, "Directed self-assembly of block copolymers for use in bit patterned media fabrication," *Journal of Physics D: Applied Physics*, vol. 46, no. 50, p. 503001, 2013.
- [28] D. Kumar, T. Jin, S. Al Risi, R. Sbiaa, W. S. Lew and S. Piramanayagam, "Domain wall motion control for racetrack memory applications," *IEEE Transactions on Magnetics*, vol. 55, no. 3, pp. 1, 2018.
- [29] S. S. Parkin, M. Hayashi and L. Thomas, "Magnetic domain-wall racetrack memory," *Science*, vol. 320, no. 5873, pp. 190, 2008.
- [30] S. Parkin and S.-H. Yang, "Memory on the racetrack," *Nature nanotechnology*, vol. 10, no. 3, pp. 195, 2015.
- [31] Y. Xu, D. Awschalom and J. Nitta, *Handbook of spintronics*. Springer Netherlands, 2016.
- [32] C. C. Faulkner, M. D. Cooke, D. A. Allwood, D. Petit, D. Atkinson and R. P. Cowburn, "Artificial domain wall nanotrap in $\text{Ni}_{81}\text{Fe}_{19}$ wires," *Journal of applied physics*, vol. 95, no. 11, pp. 6717, 2004.
- [33] T. A. Moore, I. Miron, G. Gaudin, G. Serret, S. Auffret, B. Rodmacq, A. Schuhl, S. Pizzini, J. Vogel and M. Bonfim, "High domain wall velocities induced by current in ultrathin Pt/Co/AlOx wires with perpendicular magnetic anisotropy," *Applied Physics Letters*, vol. 93, no. 26, p. 262504, 2008.
- [34] N. Ben-Romdhane, W. Zhao, Y. Zhang, J.-O. Klein, Z. Wang and D. Ravelosona, "Design and analysis of racetrack memory based on magnetic domain wall motion in nanowires," in *2014 IEEE/ACM International Symposium on Nanoscale Architectures (NANOARCH)*, 2014: IEEE, pp. 71.
- [35] T. A. Moore, P. Möhrke, L. Heyne, A. Kaldun, M. Kläui, D. Backes, J. Rhensius, L. J. Heyderman, J.-U. Thiele and G. Woltersdorf, "Magnetic-field-induced domain-wall motion in permalloy nanowires with modified Gilbert damping," *Physical Review B*, vol. 82, no. 9, p. 094445, 2010.
- [36] S. DuttaGupta, S. Fukami, C. Zhang, H. Sato, M. Yamanouchi, F. Matsukura and H. Ohno, "Adiabatic spin-transfer-torque-induced domain wall creep in a magnetic metal," *Nature Physics*, vol. 12, no. 4, pp. 333, 2016.
- [37] A. Thiaville, Y. Nakatani, J. Miltat and Y. Suzuki, "Micromagnetic understanding of current-driven domain wall motion in patterned nanowires," *EPL (Europhysics Letters)*, vol. 69, no. 6, p. 990, 2005.
- [38] D. Bang, J. Yu, X. Qiu, Y. Wang, H. Awano, A. Manchon and H. Yang, "Enhancement of spin Hall effect induced torques for current-driven magnetic domain wall motion: Inner interface effect," *Physical Review B*, vol. 93, no. 17, p. 174424, 2016.
- [39] K. Cai, Z. Zhu, J. M. Lee, R. Mishra, L. Ren, S. D. Pollard, P. He, G. Liang, K. L. Teo and H. Yang, "Ultrafast and energy-efficient spin-orbit torque switching in compensated ferrimagnets," *Nature Electronics*, vol. 3, no. 1, pp. 37, 2020.
- [40] S. Parkin, X. Jiang, C. Kaiser, A. Panchula, K. Roche and M. Samant, "Magnetically engineered spintronic sensors and memory," *Proceedings of the IEEE*, vol. 91, no. 5, pp. 661, 2003.

- [41] S. Fukami, T. Suzuki, K. Nagahara, N. Ohshima, Y. Ozaki, S. Saito, R. Nebashi, N. Sakimura, H. Honjo and K. Mori, "Low-current perpendicular domain wall motion cell for scalable high-speed MRAM," in *2009 Symposium on VLSI Technology*, 2009: IEEE, pp. 230.
- [42] A. Annunziata, M. Gaidis, L. Thomas, C. Chien, C. Hung, P. Chevalier, E. O'sullivan, J. Hummel, E. Joseph and Y. Zhu, "Racetrack memory cell array with integrated magnetic tunnel junction readout," in *2011 International Electron Devices Meeting*, 2011: IEEE, pp. 24.3. 1.
- [43] R. Ruiz, H. Kang, F. A. Detcheverry, E. Dobisz, D. S. Kercher, T. R. Albrecht, J. J. de Pablo and P. F. Nealey, "Density multiplication and improved lithography by directed block copolymer assembly," *Science*, vol. 321, no. 5891, pp. 936, 2008.
- [44] G. Meier, M. Bolte, R. Eiselt, B. Krüger, D.-H. Kim and P. Fischer, "Direct imaging of stochastic domain-wall motion driven by nanosecond current pulses," *Physical review letters*, vol. 98, no. 18, p. 187202, 2007.
- [45] M.-Y. Im, L. Bocklage, P. Fischer and G. Meier, "Direct observation of stochastic domain-wall depinning in magnetic nanowires," *Physical review letters*, vol. 102, no. 14, p. 147204, 2009.
- [46] T. Hayward, "Intrinsic nature of stochastic domain wall pinning phenomena in magnetic nanowire devices," *Scientific reports*, vol. 5, p. 13279, 2015.
- [47] T. Suzuki, S. Fukami, K. Nagahara, N. Ohshima and N. Ishiwata, "Current-Driven Domain Wall Motion, Nucleation, and Propagation in a Co/Pt Multi-Layer Strip with a Stepped Structure," *IEEE Transactions on Magnetism*, vol. 44, no. 11, pp. 2535, 2008.
- [48] M. Al Bahri, B. Borie, T. L. Jin, R. Sbiaa, M. Kläui and S. Piramanayagam, "Staggered magnetic nanowire devices for effective domain-wall pinning in racetrack memory," *Physical Review Applied*, vol. 11, no. 2, p. 024023, 2019.
- [49] T. Jin, M. Ranjbar, S. He, W. C. Law, T. Zhou, W. Lew, X. Liu and S. Piramanayagam, "Tuning magnetic properties for domain wall pinning via localized metal diffusion," *Scientific reports*, vol. 7, no. 1, pp. 1, 2017.
- [50] T. Jin, D. Kumar, W. Gan, M. Ranjbar, F. Luo, R. Sbiaa, X. Liu, W. S. Lew and S. Piramanayagam, "Nanoscale compositional modification in Co/Pd multilayers for controllable domain wall pinning in racetrack memory," *physica status solidi (RRL)–Rapid Research Letters*, vol. 12, no. 10, p. 1800197, 2018.
- [51] T. Jin, F. Tan, C. C. I. Ang, W. Gan, J. Cao, W. S. Lew and S. Piramanayagam, "Tilted magnetisation for domain wall pinning in racetrack memory," *Journal of Magnetism and Magnetic Materials*, vol. 489, p. 165410, 2019.
- [52] I. Polenciuc, A. Vick, D. Allwood, T. Hayward, G. Vallejo-Fernandez, K. O'Grady and A. Hirohata, "Domain wall pinning for racetrack memory using exchange bias," *Applied Physics Letters*, vol. 105, no. 16, p. 162406, 2014.
- [53] X. Zhang, N. Vernier, W. Zhao, H. Yu, L. Vila, Y. Zhang and D. Ravelosona, "Direct observation of domain-wall surface tension by deflating or inflating a magnetic bubble," *Physical Review Applied*, vol. 9, no. 2, p. 024032, 2018.
- [54] Y. Zhang, X. Zhang, N. Vernier, Z. Zhang, G. Agnus, J.-R. Coudeville, X. Lin, Y. Zhang, Y.-G. Zhang and W. Zhao, "Domain-Wall Motion Driven by Laplace Pressure in Co–Fe–B/MgO Nanodots with Perpendicular Anisotropy," *Physical Review Applied*, vol. 9, no. 6, p. 064027, 2018.
- [55] D. Kumar, P. Sengupta, R. Sbiaa and S. Piramanayagam, "Spin transfer torque induced domain wall oscillations in ferromagnetic nanowire with a nanoscale Dzyaloshinskii–Moriya interaction region," *Journal of Magnetism and Magnetic Materials*, vol. 507, p. 166807, 2020.

- [56] A. Thiaville, S. Rohart, É. Jué, V. Cros and A. Fert, "Dynamics of Dzyaloshinskii domain walls in ultrathin magnetic films," *EPL (Europhysics Letters)*, vol. 100, no. 5, p. 57002, 2012.
- [57] C. Song, B. Cui, F. Li, X. Zhou and F. Pan, "Recent progress in voltage control of magnetism: Materials, mechanisms, and performance," *Progress in Materials Science*, vol. 87, pp. 33, 2017.
- [58] U. Bauer, M. Przybylski and G. S. Beach, "Voltage control of magnetic anisotropy in Fe films with quantum well states," *Physical Review B*, vol. 89, no. 17, p. 174402, 2014.
- [59] A. Schellekens, A. Van den Brink, J. Franken, H. Swagten and B. Koopmans, "Electric-field control of domain wall motion in perpendicularly magnetized materials," *Nature communications*, vol. 3, no. 1, pp. 1, 2012.
- [60] W. Lin, N. Vernier, G. Agnus, K. Garcia, B. Ocker, W. Zhao, E. E. Fullerton and D. Ravelosona, "Universal domain wall dynamics under electric field in Ta/CoFeB/MgO devices with perpendicular anisotropy," *Nature communications*, vol. 7, no. 1, pp. 1, 2016.
- [61] C. Ma, T. Jin, X. Liu and S. N. Piramanayagam, "Switching domain wall motion on and off using a gate voltage for domain wall transistor applications," *Applied Physics Letters*, vol. 113, no. 23, p. 232401, 2018.
- [62] T. Koyama, K. Ueda, K.-J. Kim, Y. Yoshimura, D. Chiba, K. Yamada, J.-P. Jamet, A. Mougin, A. Thiaville and S. Mizukami, "Current-induced magnetic domain wall motion below intrinsic threshold triggered by Walker breakdown," *Nature nanotechnology*, vol. 7, no. 10, p. 635, 2012.
- [63] K.-J. Kim, Y. Yoshimura, W. S. Ham, R. Ernst, Y. Hirata, T. Li, S. Kim, T. Moriyama, Y. Nakatani and T. Ono, "Energy-efficient writing scheme for magnetic domain-wall motion memory," *Applied Physics Express*, vol. 10, no. 4, p. 043002, 2017.
- [64] C. O. Avci, E. Rosenberg, M. Baumgartner, L. Beran, A. Quindeau, P. Gambardella, C. A. Ross and G. S. Beach, "Fast switching and signature of efficient domain wall motion driven by spin-orbit torques in a perpendicular anisotropy magnetic insulator/Pt bilayer," *Applied Physics Letters*, vol. 111, no. 7, p. 072406, 2017.
- [65] A. Hirohata, K. Yamada, Y. Nakatani, L. Prejbeanu, B. Diény, P. Pirro and B. Hillebrands, "Review on Spintronics: Principles and Device Applications," *Journal of Magnetism and Magnetic Materials*, vol. 509, p. 166711, 2020.
- [66] X. Sui, C. Wang, J. Kim, J. Wang, S. Rhim, W. Duan and N. Kioussis, "Giant enhancement of the intrinsic spin Hall conductivity in β -tungsten via substitutional doping," *Physical Review B*, vol. 96, no. 24, p. 241105, 2017.
- [67] Q. Hao and G. Xiao, "Giant spin Hall effect and switching induced by spin-transfer torque in a W/Co 40 Fe 40 B 20/MgO structure with perpendicular magnetic anisotropy," *Physical Review Applied*, vol. 3, no. 3, p. 034009, 2015.
- [68] P. Deorani, J. Son, K. Banerjee, N. Koirala, M. Brahlek, S. Oh and H. Yang, "Observation of inverse spin Hall effect in bismuth selenide," *Physical Review B*, vol. 90, no. 9, p. 094403, 2014.
- [69] Y. Fan, P. Upadhyaya, X. Kou, M. Lang, S. Takei, Z. Wang, J. Tang, L. He, L.-T. Chang and M. Montazeri, "Magnetization switching through giant spin-orbit torque in a magnetically doped topological insulator heterostructure," *Nature materials*, vol. 13, no. 7, p. 699, 2014.
- [70] L. Neumann, D. Meier, J. Schmalhorst, K. Rott, G. Reiss and M. Meinert, "Temperature dependence of the spin Hall angle and switching current in the nc-W (O)/CoFeB/MgO system with perpendicular magnetic anisotropy," *Applied Physics Letters*, vol. 109, no. 14, p. 142405, 2016.

- [71] Y. Takeuchi, C. Zhang, A. Okada, H. Sato, S. Fukami and H. Ohno, "Spin-orbit torques in high-resistivity-W/CoFeB/MgO," *Applied Physics Letters*, vol. 112, no. 19, p. 192408, 2018.
- [72] C. Zhang, S. Fukami, K. Watanabe, A. Ohkawara, S. DuttaGupta, H. Sato, F. Matsukura and H. Ohno, "Critical role of W deposition condition on spin-orbit torque induced magnetization switching in nanoscale W/CoFeB/MgO," *Applied Physics Letters*, vol. 109, no. 19, p. 192405, 2016.
- [73] C.-F. Pai, L. Liu, Y. Li, H.-W. Tseng, D. C. Ralph and R. A. Buhrman, "Spin transfer torque devices utilizing the giant spin Hall effect of tungsten," *Applied Physics Letters*, vol. 101, no. 12, p. 122404, 2012.

Chapter 2. Literature Review

As this thesis focuses on the study of magnetic recording and domain wall (DW) memory devices, therefore, it is imperative to understand the basic concepts and major research in the field. The present chapter summarizes the relevant literature survey related to the magnetic recording and domain wall memory devices. Firstly, the underlying physics of ferromagnetic materials and related concepts will be introduced. Subsequently, the focused literature regarding magnetic recording and DW devices will be discussed.

2.1. Ferromagnetism

To appreciate the physical and technological investments in the field of magnetic recording and DW devices, it is important to first understand the basics of ferromagnetism. The first theoretical model, which was based on classical physics, to understand ferromagnetism was developed by P. Weiss in 1906 [1]. According to this model, there exists an interaction field, called the molecular field, between the atomic moments, which aligns them in a particular direction. The theory successfully explained the phenomenon of spontaneous magnetization and the existence of ferromagnetic order below the Curie temperature. This means a ferromagnetic material converts to paramagnetic material above the Curie temperature. However, it remained unanswered how an external magnetic field can rotate the magnetization of a magnetic substance in the presence of a much stronger molecular field ($\sim 10^8 \times$ higher). To answer this, Weiss postulated the idea of magnetic domains, where the molecular field is only limited to individual domains. Thus, aligning the atomic moments in a domain parallel to each other. However, the atomic moments in different domains point in different directions.

Although molecular field theory explained some of the concepts of ferromagnetism, the origin of the molecular field was unclear. The origin of the molecular field was later explained quantum mechanically by Heisenberg [2]. The molecular field was found to originate from short-ranged exchange energy between electronic spins. Thus, ferromagnetism was made well

understood using Heisenberg's theory of exchange interaction. However, it is very difficult to theoretically obtain properties of a certain magnetic system using Heisenberg's theory. Therefore, later, the micromagnetic theory was developed [3, 4]. The micromagnetic theory considers a magnetic system as a continuous medium with certain boundaries. A magnetic system is the result of local or global minimization of the total energy corresponding to that system. In micro-magnetism, the total energy of a magnetic system can be written as [5],

$$E_{total} = E_{exchange} + E_{anisotropy} + E_{Zeeman} + E_{magnetostatic} + E_{stress} + E_{magnetostriction} \quad (2.1)$$

Here, the energy terms in the right-hand side of Eq. 2.1 stand for exchange, anisotropy, Zeeman, magnetostatic, stress, and magnetostriction energies.

2.1.1. Exchange Energy

Exchange interaction is the quantum mechanical interaction that comes from the symmetry of wave functions of the system under the interchange of spins. The exchange interaction favors the parallel alignment of neighboring spins. For a system of two electrons with spins S_1 and S_2 , the exchange energy is given by [6],

$$e_{exchange} = -2J^* (\vec{S}_1 \cdot \vec{S}_2) \quad (2.2)$$

Where, J^* is the exchange coupling constant and usually is constant throughout the material system. The positive sign of J^* refers to a ferromagnetic system with the parallel alignment of the spins. For a magnetic system with an enormous number of spins (say n), Eq. 2.2 can be written as [6],

$$E_{exchange} = -2J^* \sum_{(i=n)>j} \vec{S}_i \cdot \vec{S}_j \quad (2.3)$$

The exchange interaction is short-ranged and dominantly favors parallel alignment of spins over a length scale called the exchange length, given by [7],

$$l_{exchange} = \sqrt{\frac{2A}{\mu_0 M_s^2}} \quad (2.4)$$

Where A is the exchange constant and is given by,

$$A = \frac{J^*}{a^2} \quad (2.5)$$

Here, μ_0 , M_s and a are permeability of free space, saturation magnetization, and lattice constant, respectively.

2.1.2. Anisotropy Energy

If the physical properties of a system are direction-dependent, the system is called anisotropic. Similarly, if one measures the magnetic properties of a magnetic material in different directions and observes them be different, the system is called magnetically anisotropic. The magnetic materials get magnetized easily along a certain direction called the easy axis. Similarly, the direction along which magnetizing the material is the hardest is called the hard axis. The remanent magnetization is also the lowest in the hard direction. There can be various sources of anisotropy in magnetic materials, like,

1. Magnetocrystalline anisotropy
2. Shape anisotropy
3. Interface anisotropy
4. Growth-induced anisotropy

The major consequence of the anisotropy is that the spins always try to align along the anisotropy direction. Usually, the strength of anisotropy is measured in terms of anisotropy constants ($K_1, K_2, \dots$). The corresponding energy is referred to as anisotropy energy.

2.1.3. Zeeman Energy

When an external magnetic field H is applied to a magnetic moment m , it tries to rotate the magnetic moment in a direction parallel to the applied magnetic field. The corresponding energy is given by [8],

$$e_{Zeeman} = -\vec{m} \cdot \vec{H} \quad (2.6)$$

For a system with n number of magnetic moments, Eq. 2.6 can be written as a sum of individual interaction energies,

$$E_{Zeeman} = -\sum_{i=1}^n \vec{m}_i \cdot \vec{H} \quad (2.7)$$

2.1.4. Magnetostatic Energy

The magnetostatic energy arises from the dipolar interactions of the magnetic moments, which can be written as [5],

$$E_{magnetostatic} = \frac{\vec{m}_i \cdot \vec{m}_j}{r_{ij}^3} - \frac{3(\vec{m}_i \cdot \vec{r}_{ij})(\vec{m}_j \cdot \vec{r}_{ij})}{r_{ij}^5} \quad (2.8)$$

Where m_i and m_j are the interacting atomic magnetic moments and r_{ij} is the spacing between them. Magnetostatic interaction is a long-ranged energy and favors the antiparallel configuration of the magnetic moments (or spins). For a magnetic material, the total magnetic energy needs to be summed over all individual dipolar interactions.

In other words, consider a rectangular stripe with magnetization pointing along the length of the stripe. Magnetic free poles will be generated at the two ends of the stripe. Therefore, there will be a non-zero stray field in the neighborhood of the stripe. However, Maxwell's 2nd law says, $\vec{\nabla} \cdot \vec{B} = 0$ or $\vec{\nabla} \cdot \vec{H}_d = -\vec{\nabla} \cdot \vec{M}$ and hence a demagnetization field will be generated inside the magnetic sample, which will oppose the magnetization direction. Thus, the magnetic moments (or spins) are under this demagnetization field and the corresponding interaction is a Zeeman-like interaction, which is given by [5],

$$E_{magnetostatic} = -\frac{1}{2}\mu_0 \sum_{i=1}^n \vec{M} \cdot \vec{H}_d \quad (2.9)$$

2.1.5. Stress Energy

When external stress is applied to the magnetic system, it induces strain energy in the system.

The corresponding energy term is given by [5, 9],

$$E_{stress} = -\sum_{i,j} \sigma_{ij} \epsilon_{ij} \quad (2.10)$$

Where σ_{ij} and ϵ_{ij} are the applied stress and magnetoelastic strain tensor, respectively.

2.1.6. Magnetostriction Energy

The magnetostriction of the ferromagnetic material may result in local stress in the ferromagnetic material and the corresponding energy is given by [5, 9],

$$E_{magnetostriction} = \frac{1}{2} \int (\vec{p}_e - \vec{\epsilon}) \cdot \vec{c} \cdot (\vec{p}_e - \vec{\epsilon}) d^3r \quad (2.11)$$

Where p_e , ϵ , and c stand for the deviation from the magnetostrictive state, strain introduced in a freely deformed state, and elastic constant tensor, respectively.

Usually, the energies due to exchange interaction, anisotropy, and magnetostatic interaction are present in a ferromagnetic system. The other energy terms arise when external perturbations such as the magnetic field and stress are applied. Therefore, the magnetic state of a certain magnetic material is the result of local or absolute minimization of total energy (exchange+ anisotropy+ magnetostatic energy) of the system.

Magnetic materials may form domains due to the competition between the magnetostatic energy and the exchange energy of the system. The exchange energy, which is the short-range energy, tries to keep the neighboring atomic moments in a mutually parallel configuration. This leads to an increase in the magnetostatic energy, which is relatively long-range energy. As a result, the system tries to orient two or more moments in different directions. Thus, the balance between these two energies results in the formation of domains. The different orientations of the neighboring domains increase the exchange energy in the transition region if the neighboring moments change abruptly. Therefore, to minimize the exchange energy, the atomic moments gradually change their direction in going from one domain to the other. However, the increase in the relative orientation of the atomic moments from the anisotropy direction increases the anisotropy energy of the system. Thus, the minimization of the anisotropy energy requires the minimization of spins pointing away from the anisotropy

direction. Therefore, the balance between exchange energy and anisotropy energy results in the formation of a transition region between the two domains. This region is known as the DW. Moreover, the mutual competition of exchange energy and magnetocrystalline energy determines the width, type, and surface energy of the DWs [10].

In materials with out-of-plane anisotropy, the DWs can be of two types (a) Bloch, and (b) Néel walls. In Bloch DWs, the spins inside the DW rotate in the yz -plane (refer to Figure 2.1) [9, 11]. Moreover, the Bloch type DWs are degenerate as clockwise as well as counterclockwise rotations are equally probable. However, the Néel type of DW structure is favored when the thickness of the sample becomes comparable to the thickness of the DW. In this case, the spins inside the DW rotate in the xy -plane. Figure 2.1 is the schematic presentation of Bloch and Néel walls.

Figure 2.1. The schematic representation of (a) Bloch and (b) Néel type DW.

2.2. Hard Disk Drives

2.2.1. Longitudinal Magnetic Recording (LMR)

The era of modern hard disk drives (HDDs) started when IBM introduced the very first HDD called random access method of accounting and control (RAMAC) with an areal density (number of bits packed in an area of one square inch) of 2 kbps (kilobits per square inch) in

1956 [12-14]. Until the development of thin-film deposition techniques, the magnetic recording continued to use magnetic particulate media. In this method, the magnetic particles were coated on the substrate using some glue-like solvent (binder). The development of thin-film deposition techniques revolutionized magnetic recording technology as this can be utilized to deposit more uniform and thin films. Moreover, complex stacks of thin films can be deposited, and magnetic properties can be altered according to the needs in a relatively easier manner.

Initially, the particulate media as well as the recording medium prepared using thin film deposition techniques were based on longitudinal magnetic recording (LMR) technology. Figure 2.2 shows the schematic of the LMR media. In this technology, the magnetic moments are lying in the plane parallel to the substrate. An inductive head was used to switch the magnetization of a bit (a group of magnetic grains) from one direction to the other. Moreover, the stray field from the transition region between two bits was used to identify the written information. With the introduction of thin film deposition techniques, initially, Co-based alloys were used as a recording medium in LMR technology. Epitaxial growth of Co-alloys in the film plane direction was achieved by using Cr-alloy based underlayer. This technology achieved an areal density of ~100 Gbps. To increase the areal density further, the reduction of bit size was proposed. However, this was limited due to the superparamagnetic limit, which results in the flipping of magnetic moments without the application of the magnetic field.

An increase in areal density to 130 Gbps was achieved by using antiferromagnetically coupled magnetic media [15, 16]. The technology, however, suffered from several additional issues while increasing the areal density further. These are, high transition noise due to the zig-zag nature of the transition region (between two bits), high demagnetizing field due to in-plane orientation of magnetic moments, and limited writing fields at higher areal densities. These problems resulted in the emergence of perpendicular magnetic recording (PMR) technology.

Figure 2.2. The schematic of longitudinal magnetic recording technology.

2.2.2. Perpendicular Magnetic Recording (PMR)

The idea of PMR technology was proposed in 1975 by researchers from Tohoku University [17]. In PMR media, the magnetizations point in a direction perpendicular to the disk surface. This, in turn, required the invention of materials with high perpendicular magnetic anisotropy. Materials such as CoCr, CoPtNb, CoCrPt-oxide were developed over several decades. PMR technology also solves the problem of superparamagnetism to some extent. Moreover, since the magnetic moments are directed in a perpendicular direction, the problem of the demagnetizing field at high bit-densities was negligible. Besides, the PMR technology uses a soft-underlayer below the recording layer, which forms a magnetic image of the writing head and hence, helps in increasing the writing field. All of these factors allow the formation of smaller bits. Despite these advantages, the community took almost 30 years in coming up with a suitable design, and the first HDD based on PMR technology was introduced in 2006 [15, 16, 18]. Figure 2.3 (a) shows the schematic of HDD based on PMR technology. The PMR technology also uses Co-based alloys with easy axis in (002) directions. Moreover, Ru based dual underlayers are used to promote the (002) axis texturing as well as optimized segregation between the magnetic grains. Several attempts have been made to increase the area density of

PMR media. However, a maximum areal density of 1 Tbps/cm² could be achieved using this technology.

Figure 2.3. Schematic of (a) perpendicular magnetic recording (PMR) technology and (b) media trilemma problem.

Further increment in areal density is limited by the media trilemma problem [15, 16]. The media trilemma problem refers to the challenge of optimizing Signal to Noise ratio (SNR), thermal stability, and writability (as shown in Figure 2.3 (b) simultaneously. This can be understood in the following manner. The increment in areal density is related to the reduction in the size of the bits. This can be achieved in two ways, (a) reducing the number of grains constituting one bit, and (b) reducing the size of the magnetic grains keeping the number of grains per bit the same. This is schematically illustrated in Figure 2.4 below. However, the SNR is related to the number of grains per bit through the following relation [16],

$$SNR = 10 \log_{10}(n) \quad (2.12)$$

Where n is the number of grains constituting one bit. Therefore, reducing the number of magnetic grains per bit will reduce the SNR, and hence is not favorable. Thus, reducing the size of the grains is the only way to reduce the size of the bit.

Figure 2.4. Proposed ways to reduce the size of a bit, (i) reduce the number of grains per bit and (ii) reduce the size of the magnetic grains, and hence to increase the areal density.

However, the thermal stability, which decides the lifetime of a data storage device, is given by [15],

$$E_B = \frac{K_u V}{K_B T} \quad (2.13)$$

Where, K_u , V , K_B , and T stand for the anisotropy constant, the volume of grains, Boltzmann constant, and temperature, respectively. Reducing the size of the grains will reduce the volume of grains (V) and hence the thermal stability factor. The media industry considers $E_B > 60$, for an HDD media with a lifetime of 10 years. A way to retain the thermal stability, while reducing the size of the grains is increasing the anisotropy constant value.

However, the writing field needed to switch the magnetization of the bits is [14-16],

$$H_W \propto \frac{2K_u}{M_s} \quad (2.14)$$

Therefore, increasing K_u will need a higher writing field from the write head. However, with the current technology, a higher writing field is a problem. Therefore, increasing areal density suffers from these three inter-related problems. As described in chapter 1, the extended PMR (PMR with two-dimensional magnetic recording and/or shingled magnetic recording) technology was proposed to increase the areal density. However, the increment in areal density is not enough to meet the large data demands. Therefore, the concept of energy-assisted magnetic recording was proposed.

2.2.3. Energy-assisted Magnetic Recording (EAMR)

To overcome the problem of the media trilemma and further increase the areal density, the concept of heat-assisted magnetic recording (HAMR) was proposed. The original idea of HAMR was proposed separately by Katayama and Saga in 1999 [19, 20]. Figure 2.5 (a) shows the schematic of the HAMR media. It uses a laser source attached to the writing head. The laser source is accommodated with a dedicated assembly, which helps to focus the laser beam to the required bit. While writing a particular bit, the bit is heated to a temperature close to its Curie temperature. An increase in the temperature effectively reduces the coercivity of the material (as indicated in Figure 2.5 (b) and (c)). The magnetization of a bit with lower coercivity can be flipped with lower writing fields. Once the information is written, the corresponding bit can be cooled to room temperature and hence helping it gain its original coercivity. This can be done by depositing a heat sink layer below the recording layer (magnetic layer). The heat sink layer extracts the heat from the recording layer and cools it back to room temperature. This allows the usage of materials with a higher anisotropy constant (higher than used in PMR media) and hence provides a scope to reduce the bit sizes.

Figure 2.5. (a) Schematic of heat-assisted magnetic recording (HAMR) technology, (b) the pictorial representation of HAMR working principle, and (c) magnetic depiction of the concept proposed in (b).

To realize a HAMR medium, materials with high magnetocrystalline anisotropy constant such as SmCo, FePt, and NdFeB are required. Moreover, the material should have a sensitive relationship with the temperature. Among these materials, the FePt binary alloy is the most common choice and most of the research regarding HAMR was performed on this material [16]. As discussed in the previous chapter, Seagate has recently launched an HDD based on HAMR technology with an areal density of 2 Tbpsi [21]. The areal density is expected to go 4 Tbpsi or more in the future [22, 23]. Despite these physical and technological advantages, HAMR has several issues towards increasing the areal density further. Some of the major challenges are as follows. (a) The bit size is limited by the reduction in the laser spot size below 40 nm [24]. (b) The repeated heating and cooling deteriorates the physical properties of the

media and hence limits the lifetime [25]. (c) The lifetime of near-field transducers, which helps in achieving the small laser spots, is another issue [26, 27].

Microwave-assisted magnetic recording (MAMR) has been proposed as another replacement of PMR media. The original concept was proposed by Zhu et al. [28]. In this technology, an alternating magnetic field of microwave frequency range is applied in addition to the writing field. Here, the writing field from the electromagnet is applied along the easy axis of the recording medium. However, the microwave field is applied in a direction perpendicular to the writing field. Such a microwave field is generated from a spin-torque oscillator attached to the writing head [29]. The microwave field excites a precessional motion of the magnetization and helps in switching the magnetization at a reduced value of the writing field. This allows the usage of a material with higher anisotropy and hence provides a scope of bit size reduction. Recently, Western Digital and Toshiba have announced the shipping of MAMR based HDD towards the end of this year with an areal density closer to 4 Tbpsi [30, 31]. However, the proposed areal density may not be able to meet the growing big data demands. Therefore, alternative technologies need to be researched.

2.2.4. Heated-dot Magnetic Recording (HDMR)

Heated-dot magnetic recording (HDMR) has been proposed as a technology to offer areal densities of 10 Tbpsi and above [21]. According to the prediction of the ASTC roadmap, this technology may take over the current PMR and/or EMAR based HDD in 2025 [21]. The basic principle of HDMR technology is a combination of bit-patterned media recording (BPMR) and HAMR. As discussed, the conventional magnetic recording media uses a granular media where several magnetic grains form one bit. The BPMR media, however, uses a patterned media, where a continuous film is transformed into a patterned film consisting of dot-like patterns. Every dot pattern acts as one bit for information storage. In this technology, every magnetic dot or magnetic island is a single domain and needs to be switched in one go. Therefore, the

thermal stability factor depends on the volume of magnetic islands (unlike conventional media, where grain volume is considered for estimation of thermal stability). In addition, in current technology, approximately 20-30 grains constitute one bit. Therefore, fabricating the bit patterned media with an island size comparable to the current bit will increase the thermal stability by $\sim 20\text{-}30\times$. This provides the makers with a huge scope for reducing the size of the magnetic bits and hence areal density.

Figure 2.6. Schematic of heated-dot magnetic recording (HDMR) technology.

Moreover, the writing process of HDMR is the same as that of HAMR technology. Therefore, a material with a higher magnetic anisotropy can be used as a recording medium. Figure 2.6 shows the schematic diagram of an HDD based on HDMR technology. The technology, however, has several issues to be addressed. One of the leading challenges for this concept is to optimize a suitable lithography technique for fabricating smaller magnetic islands (10 nm

and less) with cost and time efficiency [32]. The other issues are the reduction of the switching field distribution and synchronization of the writing head and bit location [33-43]. In the present thesis, we have raised the fabrication issue and attempted to address it using the magnetic patterning process.

2.3. Domain Wall Memory

Although the spintronics community is researching DW devices for more than two decades [44-51], the idea of DW based memory device was proposed in 2008 by a team in IBM lead by S. Parkin [52]. The DW memory has shown the potential to offer a large data storage capacity and could be a potential replacement for the current PMR and/or HAMR based HDD [53-55]. Like any other memory or storage device, the DW memory should also be able to perform (a) writing, (b) moving and storing, and (b) reading the information [52, 56, 57]. This has been schematically shown in Figure 2.7 (a). In DW memory, the information is encoded in the magnetic domains separated by thin DWs. The magnetizations of the domains in two different directions (for instance, along $+z$ or $-z$; where z is the direction perpendicular to sample surface) can be considered as the information “0” or “1”. The information can be written by locally reversing the magnetization of the corresponding domains. The stored information can be read using a reading element. Unlike the HDD, the reading and writing elements in the DW are stationary and the information is moved back and forth analogous to racing cars on the racetrack [52, 56, 57]. This can be achieved using an external magnetic field or external electrical current or both. The ability of magnetic domains to retain its magnetization until an external perturbation is applied can be used to store the information. The schematic of the DW racetrack memory is presented in Figure 2.7 (b).

Figure 2.7. (a) The requirements of the DW memory device, (b) schematic of DW memory devices, and (c) proposed a 3-D array of racetracks for future data storage (adopted from [52, 56, 57]).

The DW memory can be fabricated (i) horizontally or (ii) vertically. In the horizontal configuration, the DW racetracks are fabricated parallel to the substrate. However, in the vertical configuration, the racetracks are fabricated perpendicular to the substrate. Parkin, in his first demonstration, proposed that an array of DW racetracks (or nanowires) can be fabricated in the vertical configuration and hence offer a huge storage capacity [52, 56, 57]. This is schematically illustrated in Figure 2.7 (c) [52, 56, 57]. However, researchers are struggling to fabricate vertical nanowires. Although the vertical configuration is difficult to

achieve, there is no limit on stacking several layers of an array of horizontally fabricated nanowires. Since the reading and writing elements are static, the consecutive layers can be placed close to each other (say in a separation of a few 10s of micrometers or lesser). This is one of the biggest advantages of DW memory devices over HDDs, where the number of platters (storing disks) are limited [53].

Besides, the DW memory devices are power efficient compared to the HDDs [58]. In hard disk drives, the mechanical rotations of the disks using a motor consume a lot of power. Secondly, the reading and writing elements are also moved using motors and actuators to access the information, which is also power consuming. Thirdly, the magnetic field is used to write the information, which is considered as more power-consuming than newly researched spin-transfer and spin-orbit torque mechanisms. However, in the case of DW memory devices, there is no mechanical motion involved as reading and writing elements are static. Moreover, power-efficient STT and SOT have been proposed to move the domains or DWs in the DW memory. Besides, the DW memory devices are time efficient compared to the HDDs. The functionalities and corresponding research progress involved in the successful operation of DW memory devices will be discussed in the coming sections.

2.3.1. Writing or Nucleation of Domains

The first requirement of any memory or storage device is writing the information that needs to be stored. In DW memory devices, the writing can be performed in several ways, which will be described in this section. One of the oldest methods for writing information is to nucleate reverse domains using nucleation pads [59-61]. Large pads of the magnetic material (same as of memory element) are fabricated at one end or both the ends of the nanowire. After saturating the DW device with a large magnetic field along the easy axis direction, a nucleation field (field close to the coercivity of the device) in the opposite direction (with respect to saturation field) is applied. This results in the emergence of a reversed magnetic domain. The reason for

using a large nucleation field is to ensure the nucleation from the nucleation pads as the larger size will have a larger probability of having a defect (or nucleation point). Once the reversed domain is nucleated, a small magnetic field is applied and the DW is inserted in the nano-track. The direction of the nucleation field can be altered depending on the need of writing a bit “0” or “1”. Although this is one of the easiest ways of nucleation, this method is not suitable for the chip level devices as this is a non-local effect. In other words, the nucleation process may alter the bits stored in neighboring nanowires. In addition, having large nucleation pads reduce the density of devices in a chip. The pictorial representation of this method of writing is presented in Figure 2.8 (a).

Later, researchers have studied an alternative method based on the Oersted field for nucleating the domains. In this method, a nonmagnetic metal nanowire is fabricated in a transverse configuration with respect to the memory unit. When a current is applied, an Oersted field is generated surrounding the magnetic racetrack. This magnetic field can be utilized to reverse the domains locally. The direction of the Oersted field can be reversed by reversing the direction of the current in the nonmagnetic crossbar. Several research studies have demonstrated the nucleation of the reversed domain using this method [62-64]. Figure 2.8 (b) shows the schematic of the Oersted field-based nucleation technique. Although this method generates a local magnetic field and is useful for basic studies, it is not suitable for commercial applications due to high power consumption.

Figure 2.8. The schematics of the ways of nucleating the domains using (a) nucleation pads, (b) Oersted field, and (c) STT through MTJ.

Meanwhile, the most appropriate method for nucleating the domains was proposed and demonstrated [65-67]. This method is power efficient as well as suitable for on-chip devices. This method involves a magnetic tunnel junction (MTJ), where the memory unit acts as a free layer. The magnetic tunnel junction is made up of a fixed layer, a free layer, and an insulating layer. The magnetization of the fixed layer is quenched, and a high magnetic field or current is needed to switch the magnetization. However, the free layer is made up of a magnetically softer layer (like CoFeB) and magnetization can be switched at lower values of external stimulus. The insulating layer (such as MgO) acts as a quantum tunneling barrier. When a current is sent through this MTJ the current with the majority of electrons with one type of spin reaches the free layer. These majority electrons exert a torque on the magnetization of the free layer and

switch the magnetization locally. By changing the direction of the current, the nucleation of both types of domains (say up and down domain for PMA nanowire) can be achieved. Thus, a DW can be inserted in the nano-track. A schematic of this method is shown in Figure 2.8 (c). After writing, moving the DWs for achieving the reading and further writing is important. The research advancements made in this area of DW memory devices will be described in the upcoming sub-section.

2.3.2. Moving the Domain Walls in Racetracks

The static configuration of reading and writing elements requires that the DWs (or domains) are moved back and forth along the nano-track for reading/writing operations. Therefore, DW motion is an important aspect of DW devices. The DW in racetracks can be moved using an external magnetic field as well as an external electrical current. The details of these methods will be discussed in the coming sub-sections.

2.3.2.1. Magnetic Field Driven Domain Wall Motion

In the initial days of DW memory research, the magnetic field was considered as a driving force for the DWs [68-74]. To understand the magnetic field-driven DW dynamics, one needs to understand the micromagnetic formalism in dynamic situations (a continuation of discussions in section 1). The dynamics of DWs upon the application of an external magnetic field can be understood from the Landau-Lifshitz-Gilbert (LLG) equation [75, 76]. The equation was originally formulated by Landau and Lifshitz and was later modified by Gilbert. The dynamics of a local magnetization in the presence of an externally applied field is given by,

$$\frac{\partial \vec{m}}{\partial t} = -\gamma_0 \vec{m} \times \vec{H}_{eff} + \alpha \vec{m} \times \frac{\partial \vec{m}}{\partial t} \quad (2.15)$$

Here, m is local magnetization, H_{eff} is the total effective magnetic field acting on m . This includes the external magnetic field as well as internal fields such as anisotropy, exchange, and

magnetostatic fields. α is a damping parameter and $\gamma_0 = \frac{\mu_0 g e}{2m_e}$. Where μ_0 , g , e , m_e are permeability of free space, Landé g-factor, electronic charge, and electronic mass respectively. The first term on the right-hand side of the LLG equation stands for the precessional motion of the magnetization around the effective magnetic field. On the other hand, the second term tries to damp the precessional motion and aligns the magnetization along the direction of the effective magnetic field. Hence, the first and second terms are named precessional and damping terms, respectively. This phenomenon is schematically presented in Figure 2.9 (a) and (b).

Figure 2.9. The schematic diagrams of the (a) precessional and (b) damping motion of magnetization in the presence of an effective magnetic field.

The magnetic field-driven DW motion exhibits three distinct phases in the DW velocity vs applied field graph. For extremely low applied field, the DW velocity increases linearly with the magnetic field. Let us understand this formalism for Bloch type DW, where the spin at the center of the DW points along the $+y$ axis (Figure 2.10). Moreover, the external magnetic field is applied along the $+z$ -axis. The torque $-\gamma_0 \vec{m} \times \vec{H}_{eff}$ results in a small canting of the DW towards the $-x$ -axis. This generates a demagnetization field (H_d), which opposes the canting of the DW. The torque from this demagnetization field $-\gamma_0 \vec{m} \times \vec{H}_d$ results in a steady motion of

the DW until a critical field occurs. This critical field is called the Walker field [77]. In this region, the DW velocity is given by [68],

$$v = \mu H_{eff} \quad (2.16)$$

Where μ is the mobility of the DW. When the applied magnetic field increases slightly more than the Walker field. The DW cannot maintain its canting angle and starts precessing about the $+z$ axis. This results in a periodic change in DW structure between Bloch and Néel wall and DW velocity drops abruptly. This is called the Walker breakdown [77]. In this region, the DW motion is governed by $\alpha \vec{m} \times \frac{\partial \vec{m}}{\partial t}$ torque and the DW velocity is given by,

$$v = \alpha^2 \mu H_{eff} \quad (2.17)$$

Upon further increment in the applied field, much higher DW velocities are achieved. The occurrence of these three regimes is schematically shown in Figure 2.10.

Several micromagnetic and experimental studies have been performed to study the presence of the three regimes in the magnetic field-driven DW dynamics [68, 78-82]. Lee et al. studied the field-induced DW dynamics in NiFe wires using micromagnetic simulations and observed the three distinct regimes in the DW velocity *vs* applied field plot [83]. Kondou et al. studied the existence of the Walker breakdown for Permalloy wires experimentally using magnetoresistance measurements [78]. Figure 2.11 illustrates the experimental observation of DW velocity *vs* applied field curve [68]. The study confirms the presence of The Walker breakdown at ~ 5 Oe. However, the magnetic field is not suitable for on-chip DW devices due to the non-local effect and power deficiency. Therefore, alternative ways to move the DWs must be highlighted.

Figure 2.10. The schematic of DW motion below and above the Walker field, driven by an external magnetic field (partially inspired from [84]).

Figure 2.11. The experimental evidence of the presence of three regions in field-driven DW motion (adopted from [68]).

2.3.2.2. Electric Current Driven Domain Wall Motion

The electrical current can also be used for moving the DWs. This is a power-efficient method compared to the usage of a magnetic field. Moreover, the incorporation of electrical current with spintronic devices is suitable at the on-chip device level. Two types of torques, namely, (a) spin-transfer torque (STT) and spin-orbit torque (SOT) can be generated from electrical current. However, they are generated in different configurations. The phenomenon of current-driven DW motion will be detailed in the upcoming sub-section.

2.3.2.2.1. Spin-transfer Torque (STT)

When an electrical current is applied across a ferromagnetic micro(nano)wire, the electrons from the applied current interacts with the conduction electrons of the host material. This results in the magnetization switching and hence the DW motion [85-87]. To understand the underlying physics of this interaction, the LLG equation has been modified with two additional terms, as [85],

$$\frac{\partial \vec{m}}{\partial t} = -\gamma_0 \vec{m} \times \vec{H}_{eff} + \alpha \vec{m} \times \frac{\partial \vec{m}}{\partial t} - [\vec{u} \cdot \vec{\nabla}] \vec{m} + \beta \vec{m} \times [\vec{u} \cdot \vec{\nabla}] \vec{m} \quad (2.18)$$

Where $u = \frac{JPg\mu_B}{2eM_s}$, with the following meaning of the parameters. P , μ_B , e , M_s , β , g , and J stand for the polarization of spin current, Bohr magneton, electronic charge, saturation magnetization, non-adiabaticity factor, Landé g-factor, and current density, respectively. The first and second terms in newly added terms stand for adiabatic and non-adiabatic spin-transfer torque (STT).

Let us first understand the underlying physics of adiabatic torque. Like the field-driven DW motion case, consider the case of Bloch wall, with the magnetization of the central spin along the y-axis (Figure 2.12). Moreover, the current is applied along -x-direction. We will first consider the case when the magnitude of the current is small. Since the magnetization in the left side of the DW is along the +z-direction, the spins from the applied current will be polarized

in the $+z$ direction (the majority of spins). These spins will exert a torque, which is analogous to damping like torque, on the DW and will try to align it in $+z$ direction. This damping like torque ($\alpha \vec{m} \times \frac{\partial \vec{m}}{\partial t}$) will induce a precession of the DW in the clockwise direction. Consider the case, when the spin is slightly tilted in the $+x$ direction. The stray field coming from this tilting will induce a demagnetizing field in the $-x$ -direction, which will try to restore the initial configuration of DW. Moreover, the torque from this demagnetization field ($-\gamma_0 \vec{m} \times \vec{H}_{eff}$) will be along the $-z$ -axis and induce the precession of DW in the anti-clockwise direction. The co-existence of these opposing torques will result in no motion of the DW. This is known as intrinsic pinning. When the applied current is increased beyond a threshold current, the damping like torque starts dominating and a non-zero DW velocity is achieved. The schematic illustration of this discussion is shown in Figure 2.12.

Figure 2.12. The schematic of DW motion induced by adiabatic spin-transfer torque [84].

While comparing the theoretical and experimental results of STT induced DW motion, the researchers observed discrepancies [85, 88]. To avoid these, the idea of adding the non-adiabatic torque was proposed. As can be seen in Eq. 2.18, the non-adiabatic torque is orthogonal to the adiabatic torque. Moreover, the adiabatic torque possesses the symmetry of the magnetic field. Therefore, non-adiabatic torque is also called field-like torque. In the

presence of adiabatic as well as non-adiabatic torque, the STT-driven DW dynamics depend on the inter-balance between damping like (corresponding to adiabatic) and field-like (corresponding to non-adiabatic torque) torques.

Figure 2.13. The micromagnetic results of STT driven DW motion in perfect nanowire (no defects) for different values of the non-adiabaticity factor at a damping constant of 0.02 [85].

Thiaville et al. performed micromagnetic simulations to understand the STT induced DW dynamics in various cases as shown in Figure 2.13 [85]. The presented simulations were performed for perfect nanowire with in-plane anisotropy. Moreover, the value of the damping constant parameter (α) used was 0.02. For $\beta=0$ (pure adiabatic case), the DW only moves after a certain threshold current (as discussed in the previous section) [85, 89]. However, for non-zero values of β , the DW takes a linear relation with the current density until a critical current density. In this regime, the DW velocity is given by [85],

$$v = \frac{\beta \left(\frac{JPg\mu_B}{2eM_s} \right)}{\alpha} \quad (2.19)$$

At current densities slightly larger than this critical current density, the DW velocity drops abruptly. This behavior is similar to field-driven DW motion and is hence known as Walker breakdown. Further, an increase in the current results in higher DW velocities.

After realizing the potential advantages of STT for driving the DWs, several material systems were studied experimentally for efficient DW motion. Here, the term efficient represents a faster, low power, deterministic, and reproducible DW motion. Initially, soft magnetic materials such as Permalloy were extensively used for experiments of STT driven DW motion. In times when STT for driving the DWs was not so popular, a group of L. Berger experimentally demonstrated the DW motion when current pulses were passed through Permalloy films [86, 87]. Later, Yamaguchi et al. demonstrated the DW motion in Permalloy wire using electrical current [90]. They observed the DW moving along the electron flow and explained the DW motion due to spin-transfer torque. However, the DW velocity observed in this study was relatively low (~ 6 m/s at 1.3×10^{12} A/m²). Subsequently, an increased DW velocity of 110 m/s at a current density of 1.5×10^{12} A/m² was reported for Permalloy wires by Hayashi et al. [91]. A DW velocity of ~ 200 m/s at a J value of 5.8×10^{11} A/m² was reported when an electrical current was combined with a small magnetic field applied along an easy axis [92]. More recently, a much higher DW velocity of 800 m/s and the presence of Walker breakdown was demonstrated by Boone et al. [93, 94].

Higher storage capacity can be obtained by replacing the in-plane magnetized materials with perpendicular magnetized (PMA) materials. Therefore, the research community started focusing on STT-driven DW motion for PMA materials. The first experimental demonstration of DW motion driven by STT for this class of materials was presented by Ravelosona et al. [95]. They performed DW experiments on the Co/Pt based system. Fukami et al. researched the STT driven DW motion in Ta/CoFeB/MgO system and observed a relatively low critical current density of 6.2×10^{11} A/m² for DW motion [96]. A DW velocity of the order of a few

tens of m/s was observed in this study. In the heavy metal (HM)/CoFeB/MgO system, the flow of electrons in the ferromagnetic, as well as the HM layer, reduces the efficiency of STT. Usage of thin HM layer (say 1 nm) has been proposed as a solution to observe the STT induced DW motion [96, 97]. The adiabatic torque from STT is the main driving force for DW motion in this type of system.

Koyama et al. researched a DW velocity of 55 m/s for a current density of 1.3×10^{12} A/m² in (Co/Ni)_n samples when subjected to STT [98]. Several other experimental studies reported a similar range of DW velocities for STT driven DW dynamics [99-101]. However, faster and more power-efficient DW motion can be observed by replacing STT with SOT. Therefore, the field of SOT has accelerated in the recent past.

2.3.2.2.2. *Spin-orbit Torque (SOT)*

When an electrical current is passed through the HM layer, it exerts a torque on the FM layer in HM/FM structure. This torque is known as spin-orbit torque (SOT) [102-104]. The origin of this SOT is the spin Hall effect (SHE) in bulk HM layer and Rashba-Edelstein effect (REE) at HM/FM interface [105]. Moreover, SHE and REE are attributed to materials with high spin-orbit coupling (SOC) [105]. In an atom configuration, the coupling of orbital and spin angular momentum of an electron orbiting around the nucleus gives rise to an interaction, called SOC interaction. The corresponding interaction energy is given by [5],

$$E_{SO} \approx \frac{\mu_0 \mu_B^2 Z^4}{4\pi a_0^3} \quad (2.20)$$

Where Z and a_0 are atomic number and Bohr radius of the given atomic configuration. Eq. 2.20 indicates that the SOC increases with the atomic number. Therefore, HMs with a higher atomic number is one of the choices for studies involving SOT.

When a charge current is passed through a non-magnetic metal with high SOC (HMs like Pt, Ta, W), the electrons undergo spin-dependent scattering. This results in the accumulation of electrons with opposite spins at the extreme edges of the metal in a direction transverse to the

direction of charge flow. This, in turn, generates a pure spin current between the edges of accumulated spins. This process is called the SHE as shown in Figure 2.14 (a) [106]. The spin-dependent scattering arises from two distinct mechanisms. First, the conduction electrons interact with the effective field generated due to the high SOC of the host material. The Hamiltonian of this interaction has the form [107],

$$\vec{H}_1 = -\frac{1}{2}\mu_B\vec{B}_{eff}(\vec{k})\cdot\vec{S} \quad (2.21)$$

Here, k is the wave vector corresponding to the charge current and S is the spin operator. However, B_{eff} is an effective field that originated due to SOC. From Eq. 2.21, one can see that the interaction energy is spin-dependent and opposite spins will scatter in opposite directions. This interaction is intrinsic, which depends on material properties, and is known as skew scattering. Second, an impurity or defect may also lead to spin-dependent scattering, known as side-jump scattering, and give rise to SHE. The corresponding Hamiltonian is written as [107],

$$\vec{H}_2 = \lambda\vec{S}\cdot(\vec{k}\times\vec{\nabla}V) \quad (2.22)$$

Here, V is the potential generated due to a defect in the material. Moreover, the quantity λ is related to the strength of the scattering. The SHE is quantified by [105],

$$\vec{J}_S = \frac{\hbar}{2e}\theta_{SH}(\vec{J}_C\times\vec{\sigma}) \quad (2.23)$$

Where J_S , $\hbar$, θ_{SH} , J_C and σ are spin current, reduced Planck constant, spin-Hall angle, charge current, and polarization of accumulated spins respectively. The spin-Hall angle is an intrinsic property of the material, which quantifies the ability of charge to spin current generation of a given material. Moreover, the sign of the spin-Hall angle determines the direction of the generated spin current.

In addition to SHE, the SOC gives rise to the REE. In multilayer structures with inversion asymmetry (like HM/FM/oxide), the interface SOC results in an electric field “ $\vec{E}$ ” pointing in the direction of inversion asymmetry. The conduction electrons, which are traveling near the

interface experience this electric field. If “ $\vec{p}$ ” is the momentum of the conduction electrons, they experience a magnetic field too. The direction of the magnetic field is given by the “ $\vec{E} \times \vec{p}$ ” direction. Moreover, the interaction between the conduction electrons and the magnetic field results in the rotation of the spins of the conduction electrons along the direction of the magnetic field. This is called the interfacial REE (refer to Figure 2.14 (b)) and the corresponding interaction Hamiltonian is given by [105],

$$H_R = \frac{\alpha_R}{\hbar} (\vec{E} \times \vec{p}) \cdot \vec{\sigma} \quad (2.24)$$

Here, α_R is the Rashba parameter.

Figure 2.14. Schematic illustration of (a) SHE and (b) REE.

Corresponding to SHE and REE, the torque acting on the FM layer can be separated into two parts. The spin current generated due to bulk SHE diffuses into the FM layer and exerts a Slonczewski like STT on the spins in the FM layer. The corresponding torque tries to align the spins in the FM layer along the direction of the polarization of accumulated spins and shows the symmetry of damping-like torque. Hence, this torque is termed as damping-like torque and is given by [108, 109],

$$\vec{\tau}_{DL} \sim \vec{m} \times (\vec{\sigma} \times \vec{m}) \quad (2.25)$$

The equivalent field acting on the spins in the FM layer can be written as,

$$\vec{H}_{DL} \sim (\vec{\sigma} \times \vec{m}) \quad (2.26)$$

On the contrary, the asymmetric accumulation of spins due to REE interacts with the spins in the FM layer through exchange coupling. This results in the precession of spins with respect to the polarization of asymmetrically accumulated spins. This says that the torque exhibits the field-like symmetry and is hence termed as field-like torque, which is given by [108, 109],

$$\vec{\tau}_{FL} \sim (\vec{\sigma} \times \vec{m}) \quad (2.27)$$

Similar to the previous case, the corresponding equivalent field can be written as,

$$\vec{H}_{FL} \sim \vec{\sigma} \quad (2.28)$$

Moore et al. studied the current driven DW dynamics in 500 nm wide wires of Pt/Co/AlO_x PMA material and observed a high DW velocity of ~130 m/s [110]. Moreover, the DW was found to be moving along the direction of the current density. At that time, the observation of such a high velocity was unclear. Later, Miron et al. experimentally demonstrated that the presence of inversion asymmetry in the Pt/Co/AlO_x structure increases the efficiency of non-adiabatic STT efficiency (close to 1), which may arise due to the Rashba effect, and hence increases the DW velocities [111]. However, a detailed understanding of the underlying origin of the Rashba effect was still lacking. In his subsequent study, Miron et al. demonstrated the field-like torque originating from the electrical current [112]. The Rashba effect due to structural inversion asymmetry and high SOC induces the field-like SOT in the FM layer. A high Rashba field of the order of 10⁻⁸ Oe/(A/m²) was estimated in this study. Later, the same group demonstrated an even higher DW velocity of 400 m/s at a J of $\sim 3 \times 10^{12}$ A/m² in Pt/Co/AlO_x wires [113]. Extremely high non-adiabatic and field-like SOTs were explained to be the reasons for such high DW velocities. Several other studies reported similar results in a similar family of material structures [114, 115].

Around the same time, Liu et al. reported that SOT from SHE in the Pt layer can switch the magnetization in the Co layer in Pt/Co PMA structures [116]. Slightly later, the same group studied the Pt/Co/ AlO_x structure and demonstrated magnetization switching because of SHE in the Pt underlayer [117]. Later, SHE-induced SOT-driven DW dynamics was demonstrated by Haazen et al. [104]. However, an external in-plane (IP) magnetic field (along the length of the wire) was applied to activate the torque because of SHE. In structures like HM/FM with PMA, the direction of polarization of spins accumulated in HM is along the y -axis (refer to Figure 2.14 (a)). Therefore, the damping like torque will be zero for Bloch type DWs (with magnetization in DW along the y axis). Consequently, an external magnetic field is required to rotate the magnetization of the DW along the x -axis.

The requirement for an external magnetic field can be avoided by introducing the interfacial Dzyaloshinskii-Moriya interaction (i DMI). The i DMI introduces the Néel DW with the magnetization pointing along the x -axis. Emori et al. studied that Slonczewski-like or damping-like torque from SHE can solely drive the DWs in wires of Pt/CoFe/MgO and Ta/CoFe/MgO materials [118]. The i DMI from the interface stabilizes the Néel walls and activates the damping-like torque. The fixed chirality induces more uniform and highly efficient DW motion [118]. Several other material systems such as synthetic antiferromagnets (SAF) and ferrimagnets have been studied and SOT was found to drive the DWs at much higher velocities [119-121]. The major achievements in DW dynamics research can be summarized in the following table.

Table 2.1. Summary of major research regarding DW dynamics.

Study	Material	Driving Force	DW velocity	Remarks
Beach et al. [68]	Ta/ Ni ₈₀ Fe ₂₀ / Ta [IMA]	Magnetic Field	~150 m/s @ ~70 Oe	One of the first experimental observations of Walker Breakdown
Hayashi et al. [91]	Fe/ AlO _x / Ni ₈₁ Fe ₁₉ / TaN/ Ru [IMA]	Electrical Current	~110 m/s @ ~1.5×10 ¹² A/m ²	High DW velocity was reported with only STT (zero fields)
Beach et al. [122]	Ta/ Ni ₈₀ Fe ₂₀ / Ta [IMA]	Electrical Current & Magnetic Field	~200 m/s @ ~63 Oe and 5.8×10 ¹¹ A/m ²	Enhancement in DW velocity by combining the field along the easy axis and STT.
Koyama et al. [98]	Ta/ Pt/ (Co/Ni) _n / Co/ Pt [PMA]	Electrical Current	~60 m/s @ ~1.3×10 ¹² A/m ²	The DW dynamics was found to be independent of the magnetic field.
Moore et al. [110]	Pt/ Co/ AlO _x [PMA]	Electrical Current	~130 m/s @ ~1.8×10 ¹² A/m ²	Rashba Effect (one component of SOT) started coming into the picture
Liu et al. [116]	Ta/ CoFeB based IMA & PMA	Electrical Current with small IP field	Magnetization Switching	The second component of SOT (SHE) was researched
Emori et al. [118]	Pt & Ta /CoFe/ MgO [PMA]	Electrical Current	~1 m/s @ ~1×10 ¹¹ A/m ²	SOT on Chiral DWs started gaining attention
Yang et al. [121]	(Co/Ni) _n / Ru/ (Co/Ni) _n [PMA SAF]	Electrical Current	~750 m/s @ ~3×10 ¹² A/m ²	THE highest DW velocity (at that time) was reported
Cai et al. [120]	Pt/ CoGd/ TaO _x / SiO ₂ [PMA]	Electrical Current	~6 km/s @ 4.2×10 ¹¹ A/m ²	SOT driven ultrafast DWs in Ferrimagnets

2.3.3. Reading the Information

Reading the stored information is another requirement of a memory device. In the case of DW memory devices, the reading operation can be performed using a magnetic tunnel junction (MTJ). The spin-dependent tunneling through the insulating layer gives rise to high and low resistance states. When the magnetizations of both the layers are parallel to each other, more majority electrons pass through and hence a low resistance state is achieved. On the other hand, when the magnetizations of fixed and free layers are antiparallel to each other, fewer majority electrons tunnel, which results in a high resistance state.

Figure 2. 15. (a) The schematic depiction of the reading mechanism in DW memory device and (b) illustration of low and high resistance states for MTJ. The two resistance states represent two distinct bits for the DW memory device.

The reading element, whose free layer is the memory element of the DW memory device, is usually fabricated at a certain place along the nanowire. The bit which needs to be read is

pushed above or below the insulating and fixed layer. Depending on their relative magnetizations, a high or low resistance state (corresponding to “0” or “1”) is achieved. This phenomenon is schematically explained in Figure 2.15 (a) and (b). Annunziata et al. demonstrated reading operation in NiFe nanowire (memory element) using an IP reading element [123]. Later, Romdhane et al. reported the reading in the PMA racetrack using CoFeB/MgO/Ta-based MTJ [124]. The MTJ serves as a memory unit in magnetic random-access memories (MRAM). Dedicated research at university as well as at industry is happening to improve the capabilities of MTJ. Nowadays, a TMR ratio of ~200% can be easily achieved in the industry, which is good enough for the reading operation of DW memory devices [125].

2.3.4. Challenges of Domain Wall Memory Device

Despite phenomenal advancements in writing, moving, and reading techniques for DW memory devices, there are several issues that need to be addressed. They are,

1. The controlled and reliable motion of DW
2. Thermal stability of the DW memory
3. Large intrinsic pinning defects
4. Large operational power
5. The larger size of the DW memory devices
6. Lack of efficient motion of multiple DWs

In the present thesis, we have made an attempt at the issues of controlled and reliable DW motion, operational power, and intrinsic pinning defects. The details of the literature survey and the results of our study will be presented in the corresponding chapters of the thesis. However, researchers are making efforts to solve other issues. For example, the concept of artificial pinning sites has been proposed for increasing the thermal stability of DW devices [126-131]. The higher thermal stability is required to store the information for a long period (similar to the case of HDDs, discussed earlier in this chapter). Moreover, the concept of chiral

DWs has been demonstrated for the sequential motion of multiple DWs [132]. Meanwhile, there is not enough research for scaling the DW devices down to nanometer sizes. In summary, further efficient research needs to be performed on the issues of thermal stability, the motion of multiple DWs, and DW sizes. The research performed to date may not serve the purpose for the on-chip fabrication of DW devices.

2.4. Conclusions

Considering the scope of the thesis, this chapter provides an introduction to the underlying physics of magnetic HDDs and DW memory devices. Initially, we have introduced the micromagnetic theory and concepts of domain formation in FM materials. Besides, we have introduced the type of DWs that exist in the FM systems. Subsequently, we discussed the history and state-of-the-art of magnetic HDDs. We have also talked about the future HDD technology and its challenges that led us to study chapter 4 of this thesis. Towards the end of this chapter, we have introduced DW memory devices as a potential replacement for current HDD technology. Besides, we have discussed the writing, storing, moving, and reading operations of DW memory devices and corresponding research advancements. Finally, we have listed the challenges of DW memory devices that need to be solved for making this a practical device. This leads us to study chapters 5, 6, and 7 of this thesis.

References

- [1] P. Weiss, "La variation du ferromagnétisme avec la température," *Comptes Rendus*, vol. 143, pp. 1136, 1906.
- [2] W. Heisenberg, "Zur theorie des ferromagnetismus," in *Original Scientific Papers Wissenschaftliche Originalarbeiten*: Springer, 1985, pp. 580.
- [3] J. Fidler and T. Schrefl, "Micromagnetic modelling - the current state of the art," *Journal of Physics D: Applied Physics*, vol. 33, no. 15, pp. R135, 2000.
- [4] D. Suess, J. Fidler and T. Schrefl, "Micromagnetic simulation of magnetic materials," *Handbook of Magnetic Materials*, vol. 16, pp. 41, 2006.
- [5] J. M. Coey, *Magnetism and magnetic materials*. Cambridge university press, 2010.
- [6] S. Blundell, "Magnetism in condensed matter," ed: American Association of Physics Teachers, 2003.
- [7] <https://math.nist.gov/~MDonahue/talks/mcsd20030916-exchange-a4.pdf>.
- [8] <http://hyperphysics.phy-astr.gsu.edu/hbase/quantum/zeeman.html>.
- [9] B. D. Cullity and C. D. Graham, *Introduction to magnetic materials*. John Wiley & Sons, 2011.
- [10] A. Goldman, *Handbook of modern ferromagnetic materials*. Springer Science & Business Media, 2012.
- [11] F. Bloch, "Zur Theorie des Austauschproblems und der Remanenzerscheinung der Ferromagnetika," *Zeitschrift für Physik*, vol. 74, no. 5-6, pp. 295, 1932.
- [12] E. D. Daniel, C. D. Mee and M. H. Clark, *Magnetic recording: the first 100 years*. John Wiley & Sons, 1998.
- [13] A. Moser, K. Takano, D. T. Margulies, M. Albrecht, Y. Sonobe, Y. Ikeda, S. Sun and E. E. Fullerton, "Magnetic recording: advancing into the future," *Journal of Physics D: Applied Physics*, vol. 35, no. 19, p. R157, 2002.
- [14] S. N. Piramanayagam, Lecture Notes of Prof S. N. Piramanayagam (SPMS-PAP-747, NTU Singapore).
- [15] S. Piramanayagam, "Perpendicular recording media for hard disk drives," *Journal of Applied Physics*, vol. 102, no. 1, p. 2, 2007.
- [16] S. Piramanayagam and T. C. Chong, *Developments in data storage: materials perspective*. John Wiley & Sons, 2011.
- [17] S.-I. Iwasaki and K. Takemura, "An analysis for the circular mode of magnetization in short wavelength recording," *IEEE Transactions on Magnetics*, vol. 11, no. 5, pp. 1173, 1975.
- [18] S.-I. Iwasaki, "Perpendicular magnetic recording—Its development and realization," *Journal of magnetism and magnetic materials*, vol. 324, no. 3, pp. 244, 2012.
- [19] H. Katayama, M. Hamamoto, J.-I. Sato, Y. Murakami and K. Kojima, "New developments in laser-assisted magnetic recording," *IEEE transactions on magnetics*, vol. 36, no. 1, pp. 195, 2000.
- [20] H. Saga, H. Nemoto and M. Takahashi, "New recording method combining thermo-magnetic writing and flux detection," *Japanese journal of applied physics*, vol. 38, no. 3S, p. 1839, 1999.
- [21] <https://blog.seagate.com/craftsman-ship/hamr-next-leap-forward-now/>.
- [22] D. Weller, G. Parker, O. Mosendz, E. Champion, B. Stipe, X. Wang, T. Klemmer, G. Ju and A. Ajan, "A HAMR Media Technology Roadmap to an Areal Density of 4 Tb/in²," *IEEE transactions on magnetics*, vol. 50, no. 1, pp. 1, 2013.
- [23] Y. Chen, S. Leong, B. Xu and H. Yang, "Challenges and implementation approaches of shingled heat assisted magnetic recording (HAMR) for 10 Tb/In² data storage," in

Magnetics Symposium 2014-Celebrating 50th Anniversary of IEEE Magnetics Society (MSSC50), 2014: IEEE, pp. 1.

- [24] S. Greaves, Y. Kanai and H. Muraoka, "Magnetization switching in energy assisted recording," *IEEE Transactions on Magnetics*, vol. 48, no. 5, pp. 1794, 2012.
- [25] C. Sun, D. Xu, D. Brewes, J. Chen, S. Heald and G. Chow, "Investigation of heat-assisted magnetic recording media films in four dimensions," *IEEE Transactions on Magnetics*, vol. 49, no. 6, pp. 2510, 2013.
- [26] K. Kuriyama, M. J. Chabalko, Y. Kong, Y. Luo, T. E. Schlesinger and J. A. Bain, "Modeling of Polarization Effects in Au Nanodots Excited With InAs Quantum Dot Emitters for Use as a HAMR Heat Source," *IEEE Transactions on Magnetics*, vol. 49, no. 7, pp. 3560, 2013.
- [27] E. B. Quirk, A. Gamble, R. Hussin, G. Slovin, Y. Kong, T. E. Schlesinger, J. A. Bain, K. Kuriyama and Y. Luo, "A Process for Transferring and Patterning InAs Quantum Dot Optical Gain Media for HAMR Near Field Optical Sources," *IEEE Transactions on Magnetics*, vol. 49, no. 7, pp. 3564, 2013.
- [28] J.-G. Zhu, X. Zhu and Y. Tang, "Microwave assisted magnetic recording," *IEEE Transactions on Magnetics*, vol. 44, no. 1, pp. 125, 2008.
- [29] S. Okamoto, N. Kikuchi, M. Furuta, O. Kitakami and T. Shimatsu, "Microwave assisted magnetic recording technologies and related physics," *Journal of Physics D: Applied Physics*, vol. 48, no. 35, p. 353001, 2015.
- [30] <https://blog.westerndigital.com/the-need-for-energy-assisted-recording-technology/>.
- [31] https://www.theregister.co.uk/2018/12/07/toshiba_goes_to_mamr/.
- [32] M. Ranjbar, "Challenges in Bit Patterned Media," *Journal of Electronics & Embedded Engineering*, vol. 1, no. 1, p. 555552, 2016.
- [33] W. F. Brown Jr, "Thermal fluctuations of a single-domain particle," *Physical Review*, vol. 130, no. 5, p. 1677, 1963.
- [34] C. A. Ross, "Patterned magnetic recording media," *Annual Review of Materials Research*, vol. 31, no. 1, pp. 203, 2001.
- [35] B. Terris and T. Thomson, "Nanofabricated and self-assembled magnetic structures as data storage media," *Journal of physics D: Applied physics*, vol. 38, no. 12, p. R199, 2005.
- [36] T. Thomson, G. Hu and B. Terris, "Intrinsic distribution of magnetic anisotropy in thin films probed by patterned nanostructures," *Physical Review Letters*, vol. 96, no. 25, p. 257204, 2006.
- [37] O. Hellwig, A. Berger, T. Thomson, E. Dobisz, Z. Bandic, H. Yang, D. Kercher and E. Fullerton, "Separating dipolar broadening from the intrinsic switching field distribution in perpendicular patterned media," *Applied Physics Letters*, vol. 90, no. 16, p. 162516, 2007.
- [38] M. H. Kryder, E. C. Gage, T. W. McDaniel, W. A. Challener, R. E. Rottmayer, G. Ju, Y.-T. Hsia and M. F. Erden, "Heat assisted magnetic recording," *Proceedings of the IEEE*, vol. 96, no. 11, pp. 1810, 2008.
- [39] P. Dürrenfeld, F. Gerhard, M. Ranjbar, C. Gould, L. Molenkamp and J. Åkerman, "Spin Hall effect-controlled magnetization dynamics in NiMnSb," *Journal of Applied Physics*, vol. 117, no. 17, p. 17E103, 2015.
- [40] R. Sbiaa, I. Al-Omari, P. R. Kharel, M. Ranjbar, D. J. Sellmyer, J. Åkerman and S. N. Piramanayagam, "Temperature effect on exchange coupling and magnetization reversal in antiferromagnetically coupled (Co/Pd) multilayers," *Journal of Applied Physics*, vol. 118, no. 6, p. 063902, 2015.

- [41] R. Sbiaa, M. Ranjbar and J. Åkerman, "Domain structures and magnetization reversal in Co/Pd and CoFeB/Pd multilayers," *Journal of Applied Physics*, vol. 117, no. 17, p. 17C102, 2015.
- [42] Y. Wei, P. Svedlindh, M. Kostylev, M. Ranjbar, R. K. Dumas and J. Åkerman, "Measuring acoustic mode resonance alone as a sensitive technique to extract antiferromagnetic coupling strength," *Physical Review B*, vol. 92, no. 6, p. 064418, 2015.
- [43] Y. Yin, F. Pan, M. Ahlberg, M. Ranjbar, P. Dürrenfeld, A. Houshang, M. Haidar, L. Bergqvist, Y. Zhai and R. K. Dumas, "Tunable permalloy-based films for magnonic devices," *Physical Review B*, vol. 92, no. 2, p. 024427, 2015.
- [44] D. A. Allwood, G. Xiong, C. Faulkner, D. Atkinson, D. Petit and R. Cowburn, "Magnetic domain-wall logic," *Science*, vol. 309, no. 5741, pp. 1688, 2005.
- [45] H. Williams, W. Shockley and C. Kittel, "Studies of the propagation velocity of a ferromagnetic domain boundary," *Physical Review*, vol. 80, no. 6, p. 1090, 1950.
- [46] B. Alessandro, C. Beatrice, G. Bertotti and A. Montorsi, "Domain-wall dynamics and Barkhausen effect in metallic ferromagnetic materials. I. Theory," *Journal of applied physics*, vol. 68, no. 6, pp. 2901, 1990.
- [47] V. G. Bar'yakhtar, B. Ivanov and M. V. Chetkin, "Dynamics of domain walls in weak ferromagnets," *Soviet Physics Uspekhi*, vol. 28, no. 7, p. 563, 1985.
- [48] S. Zhang and Z. Li, "Roles of nonequilibrium conduction electrons on the magnetization dynamics of ferromagnets," *Physical review letters*, vol. 93, no. 12, p. 127204, 2004.
- [49] J. Shibata, G. Tatara and H. Kohno, "A brief review of field-and current-driven domain-wall motion," *Journal of Physics D: Applied Physics*, vol. 44, no. 38, p. 384004, 2011.
- [50] D. Bhowmik, M. E. Nowakowski, L. You, O. Lee, D. Keating, M. Wong, J. Bokor and S. Salahuddin, "Deterministic domain wall motion orthogonal to current flow due to spin orbit torque," *Scientific reports*, vol. 5, p. 11823, 2015.
- [51] D. Mancilla-Almonacid, R. Jaeschke-Ubiergo, A. S. Núñez and S. Allende, "Ultrafast domain wall propagation due to the interfacial Dzyaloshinskii–Moriya interaction," *Nanotechnology*, vol. 31, no. 12, p. 125707, 2020.
- [52] S. S. Parkin, M. Hayashi and L. Thomas, "Magnetic domain-wall racetrack memory," *Science*, vol. 320, no. 5873, pp. 190, 2008.
- [53] D. Kumar, T. Jin, S. Al Risi, R. Sbiaa, W. S. Lew and S. Piramanayagam, "Domain wall motion control for racetrack memory applications," *IEEE Transactions on Magnetics*, vol. 55, no. 3, pp. 1, 2018.
- [54] M. Hayashi, L. Thomas, R. Moriya, C. Rettner and S. S. Parkin, "Current-controlled magnetic domain-wall nanowire shift register," *Science*, vol. 320, no. 5873, pp. 209, 2008.
- [55] M. H. Kryder and C. S. Kim, "After hard drives—What comes next?," *IEEE Transactions on Magnetics*, vol. 45, no. 10, pp. 3406, 2009.
- [56] <https://www.quora.com/When-will-racetrack-memory-be-available-for-sale>.
- [57] S. Parkin and S.-H. Yang, "Memory on the racetrack," *Nature nanotechnology*, vol. 10, no. 3, pp. 195, 2015.
- [58] G. Malinowski, O. Boulle and M. Kläui, "Current-induced domain wall motion in nanoscale ferromagnetic elements," *Journal of Physics D: Applied Physics*, vol. 44, no. 38, p. 384005, 2011.
- [59] D. Kumar, C. H. Jing, C. JianPeng, T. Jin, L. S. Ter, R. Sbiaa and S. Piramanayagam, "Ultra-low Power Domain Wall Device for Spin-based Neuromorphic Computing," *arXiv preprint arXiv:2007.12357*, 2020.

- [60] D. Atkinson, D. A. Allwood, G. Xiong, M. D. Cooke, C. C. Faulkner and R. P. Cowburn, "Magnetic domain-wall dynamics in a submicrometre ferromagnetic structure," *Nature materials*, vol. 2, no. 2, pp. 85, 2003.
- [61] H. Mohammed, S. A. Risi, T. L. Jin, J. Kosel, S. Piramanayagam and R. Sbiaa, "Controlled spin-torque driven domain wall motion using staggered magnetic wires," *Applied Physics Letters*, vol. 116, no. 3, p. 032402, 2020.
- [62] D. Chiba, G. Yamada, T. Koyama, K. Ueda, H. Tanigawa, S. Fukami, T. Suzuki, N. Ohshima, N. Ishiwata and Y. Nakatani, "Control of multiple magnetic domain walls by current in a Co/Ni nano-wire," *Applied Physics Express*, vol. 3, no. 7, p. 073004, 2010.
- [63] N. Ohshima, T. Koyama, H. Tanigawa, M. Kotsugi, T. Ohkouchi, D. Chiba, T. Kinoshita and T. Ono, "Real space observation of current-induced magnetic domain wall displacement in Co/Ni nano-wire by photoemission electron microscopy," *Journal of Physics: Condensed Matter*, vol. 23, no. 38, p. 382202, 2011.
- [64] K. Ueda, D. Chiba, T. Koyama, G. Yamada, H. Tanigawa, S. Fukami, T. Suzuki, N. Ohshima, N. Ishiwata and Y. Nakatani, "Current-induced domain wall motion in Co/Ni nano-wires with different Co and Ni thicknesses," in *Journal of Physics: Conference Series*, 2011, vol. 266, no. 1: IOP Publishing, p. 012110.
- [65] M. Foerster, O. Boulle, S. Esefelder, R. Mattheis and M. Kläui, "Domain Wall Memory Device," *Handbook of Spintronics*, pp. 1, 2014.
- [66] Y. Zhang, W. Zhao, J.-O. Klein, D. Ravelsona and C. Chappert, "Ultra-high density content addressable memory based on current induced domain wall motion in magnetic track," *IEEE transactions on magnetics*, vol. 48, no. 11, pp. 3219, 2012.
- [67] W. Zhao, D. Ravelosona, J.-O. Klein and C. Chappert, "Domain wall shift register-based reconfigurable logic," *IEEE Transactions on Magnetism*, vol. 47, no. 10, pp. 2966, 2011.
- [68] G. S. Beach, C. Nistor, C. Knutson, M. Tsoi and J. L. Erskine, "Dynamics of field-driven domain-wall propagation in ferromagnetic nanowires," *Nature materials*, vol. 4, no. 10, pp. 741, 2005.
- [69] K. Sixtus and L. Tonks, "Propagation of large Barkhausen discontinuities," *Physical Review*, vol. 37, no. 8, p. 930, 1931.
- [70] R. De Blois, "Magnetization processes: reversals and losses," *J. Appl. Phys.*, vol. 29, no. 3, pp. 459, 1958.
- [71] D. Smith, "A magnetic shift register employing controlled domain wall motion," *IEEE Transactions on Magnetism*, vol. 1, no. 4, pp. 281, 1965.
- [72] D. Herman Jr, B. Argyle, P. Trouilloud, B. Petek, L. Romankiw, P. Andricacos, S. Krongelb, D. Rath, D. Canaperi and M. Komsa, "Study of field-driven wall-configuration conversions for laminated Permalloy in the easy-axis state," *Journal of Applied Physics*, vol. 63, no. 8, pp. 4036, 1988.
- [73] K. Klaassen and J. Van Peppen, "Field-driven domain-wall jumps in thin-film heads," *Journal of applied physics*, vol. 69, no. 8, pp. 5417, 1991.
- [74] W. Li, Z. Chen, D. Wen, D. Chen, Z. Fan, M. Zeng, X. Lu, X. Gao and M. Qin, "Rotating magnetic field driven antiferromagnetic domain wall motion: Role of Dzyaloshinskii-Moriya interaction," *Journal of Magnetism and Magnetic Materials*, vol. 497, p. 166051, 2020.
- [75] T. L. Gilbert, "A phenomenological theory of damping in ferromagnetic materials," *IEEE transactions on magnetics*, vol. 40, no. 6, pp. 3443, 2004.
- [76] L. Landau and E. Lifshitz, "On the theory of the dispersion of magnetic permeability in ferromagnetic bodies," in *Perspectives in Theoretical Physics*: Elsevier, 1992, pp. 51.
- [77] N. L. Schryer and L. R. Walker, "The motion of 180 domain walls in uniform dc magnetic fields," *Journal of Applied Physics*, vol. 45, no. 12, pp. 5406, 1974.

- [78] K. Kondou, N. Ohshima, S. Kasai, Y. Nakatani and T. Ono, "Single shot detection of the magnetic domain wall motion by using tunnel magnetoresistance effect," *Applied physics express*, vol. 1, no. 6, p. 061302, 2008.
- [79] J. Yang, C. Nistor, G. Beach and J. Erskine, "Magnetic domain-wall velocity oscillations in permalloy nanowires," *Physical Review B*, vol. 77, no. 1, p. 014413, 2008.
- [80] M. Hayashi, L. Thomas, C. Rettner, R. Moriya and S. S. Parkin, "Direct observation of the coherent precession of magnetic domain walls propagating along permalloy nanowires," *Nature Physics*, vol. 3, no. 1, p. 21, 2007.
- [81] M. Hayashi, L. Thomas, C. Rettner, R. Moriya and S. S. Parkin, "Real time observation of the field driven periodic transformation of domain walls in Permalloy nanowires at the Larmor frequency and its first harmonic," *Applied Physics Letters*, vol. 92, no. 11, p. 112510, 2008.
- [82] S. Glathe, R. Mattheis and D. Berkov, "Direct observation and control of the Walker breakdown process during a field driven domain wall motion," *Applied Physics Letters*, vol. 93, no. 7, p. 072508, 2008.
- [83] J.-Y. Lee, K.-S. Lee, S. Choi, K. Y. Guslienko and S.-K. Kim, "Dynamic transformations of the internal structure of a moving domain wall in magnetic nanostripes," *Physical Review B*, vol. 76, no. 18, p. 184408, 2007.
- [84] S.-H. Yang and S. Parkin, "Novel domain wall dynamics in synthetic antiferromagnets," *Journal of Physics: Condensed Matter*, vol. 29, no. 30, p. 303001, 2017.
- [85] A. Thiaville, Y. Nakatani, J. Miltat and Y. Suzuki, "Micromagnetic understanding of current-driven domain wall motion in patterned nanowires," *EPL (Europhysics Letters)*, vol. 69, no. 6, p. 990, 2005.
- [86] P. Freitas and L. Berger, "Observation of s-d exchange force between domain walls and electric current in very thin Permalloy films," *Journal of Applied Physics*, vol. 57, no. 4, pp. 1266, 1985.
- [87] C. Y. Hung and L. Berger, "Exchange forces between domain wall and electric current in permalloy films of variable thickness," *Journal of applied physics*, vol. 63, no. 8, pp. 4276, 1988.
- [88] G. Tatara and H. Kohno, "Theory of current-driven domain wall motion: Spin transfer versus momentum transfer," *Physical review letters*, vol. 92, no. 8, p. 086601, 2004.
- [89] Z. Li and S. Zhang, "Domain-wall dynamics and spin-wave excitations with spin-transfer torques," *Physical review letters*, vol. 92, no. 20, p. 207203, 2004.
- [90] A. Yamaguchi, T. Ono, S. Nasu, K. Miyake, K. Mibu and T. Shinjo, "Real-space observation of current-driven domain wall motion in submicron magnetic wires," *Physical review letters*, vol. 92, no. 7, p. 077205, 2004.
- [91] M. Hayashi, L. Thomas, C. Rettner, R. Moriya, Y. B. Bazaliy and S. S. Parkin, "Current driven domain wall velocities exceeding the spin angular momentum transfer rate in permalloy nanowires," *Physical review letters*, vol. 98, no. 3, p. 037204, 2007.
- [92] G. Beach, C. Knutson, C. Nistor, M. Tsoi and J. Erskine, "Nonlinear domain-wall velocity enhancement by spin-polarized electric current," *Physical Review Letters*, vol. 97, no. 5, p. 057203, 2006.
- [93] C. Boone, J. Katine, M. Carey, J. Childress, X. Cheng and I. Krivorotov, "Rapid domain wall motion in permalloy nanowires excited by a spin-polarized current applied perpendicular to the nanowire," *Physical review letters*, vol. 104, no. 9, p. 097203, 2010.
- [94] L. Heyne, M. Kläui, D. Backes, T. Moore, S. Krzyk, U. Rüdiger, L. J. Heyderman, A. F. Rodríguez, F. Nolting and T. O. Mendes, "Relationship between nonadiabaticity and

- damping in permalloy studied by current induced spin structure transformations," *Physical review letters*, vol. 100, no. 6, p. 066603, 2008.
- [95] D. Ravelosona, D. Lacour, J. Katine, B. Terris and C. Chappert, "Nanometer scale observation of high efficiency thermally assisted current-driven domain wall depinning," *Physical review letters*, vol. 95, no. 11, p. 117203, 2005.
 - [96] S. Fukami, T. Suzuki, Y. Nakatani, N. Ishiwata, M. Yamanouchi, S. Ikeda, N. Kasai and H. Ohno, "Current-induced domain wall motion in perpendicularly magnetized CoFeB nanowire," *Applied Physics Letters*, vol. 98, no. 8, p. 082504, 2011.
 - [97] S. DuttaGupta, S. Fukami, C. Zhang, H. Sato, M. Yamanouchi, F. Matsukura and H. Ohno, "Adiabatic spin-transfer-torque-induced domain wall creep in a magnetic metal," *Nature Physics*, vol. 12, no. 4, pp. 333, 2016.
 - [98] T. Koyama, D. Chiba, K. Ueda, H. Tanigawa, S. Fukami, T. Suzuki, N. Ohshima, N. Ishiwata, Y. Nakatani and T. Ono, "Magnetic field insensitivity of magnetic domain wall velocity induced by electrical current in Co/Ni nanowire," *Applied Physics Letters*, vol. 98, no. 19, p. 192509, 2011.
 - [99] H. Tanigawa, T. Suzuki, S. Fukami, K. Suemitsu, N. Ohshima and E. Kariyada, "Thickness dependence of current-induced domain wall motion in a Co/Ni multi-layer with out-of-plane anisotropy," *Applied Physics Letters*, vol. 102, no. 15, p. 152410, 2013.
 - [100] H. Tanigawa, T. Koyama, G. Yamada, D. Chiba, S. Kasai, S. Fukami, T. Suzuki, N. Ohshima, N. Ishiwata and Y. Nakatani, "Domain wall motion induced by electric current in a perpendicularly magnetized Co/Ni nano-wire," *Applied Physics Express*, vol. 2, no. 5, p. 053002, 2009.
 - [101] H. Tanigawa, K. Suemitsu, S. Fukami, N. Ohshima, T. Suzuki, E. Kariyada and N. Ishiwata, "Effect of device temperature on domain wall motion in a perpendicularly magnetized Co/Ni wire," *Applied physics express*, vol. 4, no. 1, p. 013007, 2011.
 - [102] I. M. Miron, T. Moore, H. Szabolcs, L. D. Buda-Prejbeanu, S. Auffret, B. Rodmacq, S. Pizzini, J. Vogel, M. Bonfim and A. Schuhl, "Fast current-induced domain-wall motion controlled by the Rashba effect," *Nature materials*, vol. 10, no. 6, pp. 419, 2011.
 - [103] X. Zhao, Y. Liu, D. Zhu, M. Sall, X. Zhang, H. Ma, J. Langer, B. Ocker, S. Jaiswal and G. Jakob, "Spin-orbit torque driven multi-level switching in He⁺ irradiated W-CoFeB-MgO Hall bars with perpendicular anisotropy," *Applied Physics Letters*, vol. 116, no. 24, p. 242401, 2020.
 - [104] P. Haazen, E. Murè, J. Franken, R. Lavrijsen, H. Swagten and B. Koopmans, "Domain wall depinning governed by the spin Hall effect," *Nature materials*, vol. 12, no. 4, pp. 299, 2013.
 - [105] R. Ramaswamy, J. M. Lee, K. Cai and H. Yang, "Recent advances in spin-orbit torques: Moving towards device applications," *Applied Physics Reviews*, vol. 5, no. 3, p. 031107, 2018.
 - [106] J. Hirsch, "Spin hall effect," *Physical Review Letters*, vol. 83, no. 9, p. 1834, 1999.
 - [107] T. Schäpers, *Semiconductor spintronics*. Walter de Gruyter GmbH & Co KG, 2016.
 - [108] S. Fukami, T. Anekawa, C. Zhang and H. Ohno, "A spin-orbit torque switching scheme with collinear magnetic easy axis and current configuration," *Nature Nanotechnology*, vol. 11, no. 7, p. 621, 2016.
 - [109] A. Hirohata, K. Yamada, Y. Nakatani, L. Prejbeanu, B. Diény, P. Pirro and B. Hillebrands, "Review on Spintronics: Principles and Device Applications," *Journal of Magnetism and Magnetic Materials*, vol. 509, p. 166711, 2020.
 - [110] T. A. Moore, I. Miron, G. Gaudin, G. Serret, S. Auffret, B. Rodmacq, A. Schuhl, S. Pizzini, J. Vogel and M. Bonfim, "High domain wall velocities induced by current in

- ultrathin Pt/Co/AlOx wires with perpendicular magnetic anisotropy," *Applied Physics Letters*, vol. 93, no. 26, p. 262504, 2008.
- [111] I. Miron, P.-J. Zermatten, G. Gaudin, S. Auffret, B. Rodmacq and A. Schuhl, "Domain wall spin torquemeter," *Physical review letters*, vol. 102, no. 13, p. 137202, 2009.
 - [112] I. M. Miron, G. Gaudin, S. Auffret, B. Rodmacq, A. Schuhl, S. Pizzini, J. Vogel and P. Gambardella, "Current-driven spin torque induced by the Rashba effect in a ferromagnetic metal layer," *Nature materials*, vol. 9, no. 3, pp. 230, 2010.
 - [113] I. M. Miron, T. Moore, H. Szabolcs, L. D. Buda-Prejbeanu, S. Auffret, B. Rodmacq, S. Pizzini, J. Vogel, M. Bonfim and A. Schuhl, "Fast current-induced domain-wall motion controlled by the Rashba effect," *Nature materials*, vol. 10, no. 6, p. 419, 2011.
 - [114] K.-S. Ryu, L. Thomas, S.-H. Yang and S. S. Parkin, "Current induced tilting of domain walls in high velocity motion along perpendicularly magnetized micron-sized Co/Ni/Co racetracks," *Applied Physics Express*, vol. 5, no. 9, p. 093006, 2012.
 - [115] T. Koyama, H. Hata, K.-J. Kim, T. Moriyama, H. Tanigawa, T. Suzuki, Y. Nakatani, D. Chiba and T. Ono, "Current-induced magnetic domain wall motion in a Co/Ni nanowire with structural inversion asymmetry," *Applied Physics Express*, vol. 6, no. 3, p. 033001, 2013.
 - [116] L. Liu, C.-F. Pai, Y. Li, H. Tseng, D. Ralph and R. Buhrman, "Spin-torque switching with the giant spin Hall effect of tantalum," *Science*, vol. 336, no. 6081, pp. 555, 2012.
 - [117] L. Liu, O. Lee, T. Gudmundsen, D. Ralph and R. Buhrman, "Current-induced switching of perpendicularly magnetized magnetic layers using spin torque from the spin Hall effect," *Physical review letters*, vol. 109, no. 9, p. 096602, 2012.
 - [118] S. Emori, U. Bauer, S.-M. Ahn, E. Martinez and G. S. Beach, "Current-driven dynamics of chiral ferromagnetic domain walls," *Nature materials*, vol. 12, no. 7, p. 611, 2013.
 - [119] Z. Yu, Y. Zhang, Z. Zhang, M. Cheng, Z. Lu, X. Yang, J. Shi and R. Xiong, "Domain-wall motion at an ultrahigh speed driven by spin-orbit torque in synthetic antiferromagnets," *Nanotechnology*, vol. 29, no. 17, p. 175404, 2018.
 - [120] K. Cai, Z. Zhu, J. M. Lee, R. Mishra, L. Ren, S. D. Pollard, P. He, G. Liang, K. L. Teo and H. Yang, "Ultrafast and energy-efficient spin-orbit torque switching in compensated ferrimagnets," *Nature Electronics*, pp. 1, 2020.
 - [121] S.-H. Yang, K.-S. Ryu and S. Parkin, "Domain-wall velocities of up to 750 m s⁻¹ driven by exchange-coupling torque in synthetic antiferromagnets," *Nature nanotechnology*, vol. 10, no. 3, p. 221, 2015.
 - [122] G. Beach, C. Knutson, C. Nistor, M. Tsoi and J. L. Erskine, "Nonlinear domain-wall velocity enhancement by spin-polarized electric current," *Physical Review Letters*, vol. 97, no. 5, p. 057203, 2006.
 - [123] A. Annunziata, M. Gaidis, L. Thomas, C. Chien, C. Hung, P. Chevalier, E. O'sullivan, J. Hummel, E. Joseph and Y. Zhu, "Racetrack memory cell array with integrated magnetic tunnel junction readout," in *2011 International Electron Devices Meeting*, 2011: IEEE, pp. 24.3. 1.
 - [124] N. Ben-Romdhane, W. Zhao, Y. Zhang, J.-O. Klein, Z. Wang and D. Ravelosona, "Design and analysis of racetrack memory based on magnetic domain wall motion in nanowires," in *Proceedings of the 2014 IEEE/ACM International Symposium on Nanoscale Architectures*, 2014: ACM, pp. 71.
 - [125] S. Bhatti, R. Sbiaa, A. Hirohata, H. Ohno, S. Fukami and S. Piramanayagam, "Spintronics based random access memory: a review," *Materials Today*, vol. 20, no. 9, pp. 530, 2017.
 - [126] T. Suzuki, S. Fukami, K. Nagahara, N. Ohshima and N. Ishiwata, "Evaluation of scalability for current-driven domain wall motion in a Co/Ni multilayer strip for

- memory applications," *IEEE Transactions on Magnetics*, vol. 45, no. 10, pp. 3776, 2009.
- [127] S. Fukami, T. Suzuki, K. Nagahara, N. Ohshima and N. Ishiwata, "Large thermal stability independent of critical current of domain wall motion in Co/Ni nanowires with step pinning sites," *Journal of Applied Physics*, vol. 108, no. 11, p. 113914, 2010.
 - [128] K.-J. Kim, R. Hiramatsu, T. Koyama, K. Ueda, Y. Yoshimura, D. Chiba, K. Kobayashi, Y. Nakatani, S. Fukami and M. Yamanouchi, "Two-barrier stability that allows low-power operation in current-induced domain-wall motion," *Nature communications*, vol. 4, p. 2011, 2013.
 - [129] K.-J. Kim, R. Hiramatsu, T. Moriyama, H. Tanigawa, T. Suzuki, E. Kariyada and T. Ono, "Tradeoff between low-power operation and thermal stability in magnetic domain-wall-motion devices driven by spin Hall torque," *Applied Physics Express*, vol. 7, no. 5, p. 053003, 2014.
 - [130] S. Fukami, J. Ieda and H. Ohno, "Thermal stability of a magnetic domain wall in nanowires," *Physical Review B*, vol. 91, no. 23, p. 235401, 2015.
 - [131] R. Sbiaa and M. Al Bahri, "Constricted nanowire with stabilized magnetic domain wall," *Journal of Magnetism and Magnetic Materials*, vol. 411, pp. 113, 2016.
 - [132] J.-S. Kim, M.-A. Mawass, A. Bisig, B. Krüger, R. M. Reeve, T. Schulz, F. Büttner, J. Yoon, C.-Y. You and M. Weigand, "Synchronous precessional motion of multiple domain walls in a ferromagnetic nanowire by perpendicular field pulses," *Nature communications*, vol. 5, p. 3429, 2014.

Chapter 3. Experimental and Simulation Methods

The experimental and simulation techniques employed to perform research studies of this thesis will be introduced in this chapter. In addition, the details of the initiation and optimization of the lithography technique based on self-assembly of di-block copolymers (DBCP) will be discussed. In the initial days of my Ph.D. candidature, we (I with the help of Prof Prem) initiated and optimized this lithography technique in our lab. The thin films used during the various research studies carried out during the course of this Ph.D. were prepared using the magnetron sputtering process. The characterizations of the thin films were performed using X-ray diffraction (XRD), atomic force microscopy (AFM), magnetic force microscopy (MFM), vibrating sample magnetometer (VSM), alternating gradient force magnetometer (AGFM), cross-sectional tunneling electron microscope (TEM), torque magnetometry and magneto-optic Kerr microscopy techniques. The devices were fabricated using self-assembly, optical lithography, reactive ion etching, and ion milling techniques. The self-assembly of DBCP was employed to fabricate the hard mask on the top of CoPt and FePt media samples. This is to facilitate the masked ion implantation on these samples to locally modify the structural and magnetic properties. For the theoretical understanding of the ion implantation process, the Transport of Ions in Matter (TRIM) program from Stopping and Range of Ions in Matter (SRIM) simulations were performed. The electrical measurements on the thin films and domain wall (DW) devices were performed using Keithly 6221 current source and Keithly 2182A nano-voltmeter in association with Kerr microscopy. To understand the magnetization statics and dynamics of the studied samples, micromagnetic simulations using the object-oriented micromagnetic framework (OOMMF) and Mumax3 simulators were performed.

3.1. Thin Film Deposition

3.1.1. Magnetron Sputtering

Sputtering, a physical vapor deposition process, is a well-used technique for depositing thin films in the field of Spintronics [1-4]. As depicted in Figure 3.1, an application of high voltage results in a discharge plasma of neutral gas atoms, ions, and electrons in the sputtering chamber. Typically, Ar gas is used to host the plasma during the sputtering process. The positively charged Ar^+ ions hit the target surface and undergo statistical scattering. When the energy transferred from the Ar^+ ions is more than the binding energy of surface atoms of the target material, the target atoms get ejected. Subsequently, these target atoms travel and stick on the substrate forming a thin film of the target material.

In magnetron sputtering, magnets are arranged in the back surface of the target holder in such a way that the flux lines are parallel to the target surface. The secondary electrons generated during the bombardment of Ar^+ ions on the surface of the target experience simultaneous electric as well as magnetic fields. These mutually perpendicular electric and magnetic fields result in the spiral motion of secondary electrons. This enhances the probability of collision of these electrons with other Ar gas atoms near the target surface. A larger collision probability increases the probability of ionization of Ar gas. This way, the higher sputtering rates at lower Ar gas pressure is achieved. Figure 3.1 shows the schematic of the magnetron sputtering system.

The sputtering process can be categorized as DC and RF sputtering process [1]. In DC sputtering process, a DC voltage is applied to generate the plasma. This is used for depositing the metallic thin films. On the other hand, the RF sputtering process is used to deposit the insulator films, where an RF voltage is applied across the electrodes (as shown in Figure 3.1). For instance, if DC power is applied for the case of insulating targets, the Ar^+ ions start accumulating at the surface due to the low conductivity of the target and repel the further

coming Ar^+ ions. In order to avoid this situation, RF power is applied. The continuous bombardment of Ar^+ ions on the surface of the target results in an increase in the temperature. This, in turn, may result in the melting of the target. Therefore, cooling water is continuously supplied to maintain the temperature.

The studied magnetic thin film stacks Ta (5 nm)/ Ru (15 nm)/ $\text{Co}_{80}\text{Pt}_{20}$ (14 nm), NiTa (35 nm)/MgO (8 nm)/ FePt-C (8 nm) and W (2- 8 nm)/CoFeB (1 nm)/ MgO (~1 nm)/ Ru (2 nm) were prepared using DC/ RF sputtering process. Moreover, all other supplementary thin films (like thin films for thickness optimization, deposition condition optimization, etching rate optimization, etc.) and electrode pads were also deposited using DC/RF magnetron sputtering process.

Figure 3.1. The schematic illustration of the magnetron sputtering process.

3.2. Characterization Tools

3.2.1. X-ray Diffraction (XRD)

X-ray diffraction (XRD) is a widely used technique for the structural analysis of crystalline samples [5]. Figure 3.2 (a) shows the schematic diagram of the XRD technique. The X-ray beam incidents on the surface of the samples and gets diffracted. We have used θ - 2θ geometry, where both the X-ray source as well as detector move in such a way that the diffracted beam makes an angle 2θ with respect to the incident beam. Where θ is the incident angle of X-ray with respect to the sample plane. The X-rays incident on the sample and gets diffracted from the crystalline planes present in the sample. The diffracted beam is detected by the X-ray detector. When the diffracted beam fulfills the condition,

$$2d \sin\theta = n\lambda \quad (3.1)$$

a corresponding diffraction peak appears in the intensity vs 2θ plot [5]. Here, d is interplanar spacing, λ is the wavelength of X-rays (1.54442 Å), and n is the order of diffraction peak. This condition is known as Bragg's Law [5]. The obtained intensity vs 2θ is then compared with the standard database from Joint Committee on Powdered Diffraction Standard (JCPDS) [6]. This is to find out the possible crystal structure of the sample.

In this thesis, we have used the Pananalytical XPert Pro XRD machine, which uses a $Cu-K_\alpha$ X-ray source, to characterize the media samples ($Co_{80}Pt_{20}$ and FePt-C). Figure 3.2 (b) is a typical XRD plot. This corresponds to Xe ion (Energy: 897.6 MeV and Ion Fluence: 1×10^{11} ions/cm²) irradiated Ta/Ru/ $Co_{80}Pt_{20}$ media sample [7]. The XRD graph shows the occurrence of Ru hexagonal close-packed (HCP) (002) and CoPt HCP (002) peaks and confirms that both Ru and CoPt are in HCP crystal structure, which is highly textured along a direction perpendicular to the film plane.

Figure 3.2. (a) The schematic of the X-ray diffraction technique and (b) example of the typical XRD graph. This graph corresponds to Xe ion (897.6 MeV energy and 1×10^{11} ions/cm² ion fluence) irradiated Ta/Ru/Co₈₀Pt₂₀ sample [7].

3.2.2. Atomic Force Microscopy (AFM) and Magnetic Force Microscopy (MFM)

The atomic force microscopy (AFM), invented by Swiss scientist Gerd Binnig and colleagues in 1986, is a well-used technique for characterizing the surfaces of different kinds of samples [8, 9]. The working principle of AFM is based on the interatomic forces between the tip, attached at the end of the cantilever (as shown in Figure 3.3 (a)), and the sample surface to be measured [9-11]. The tip scans the sample surface and follows the surface topography due to interatomic forces between the tip and the sample surface. As a result, the cantilever bends accordingly. A laser beam originated from a laser source, incidents on the cantilever, and gets reflected. The reflected laser beam is detected by a position-sensitive photodiode. As the cantilever bends according to the topography of the sample, different position spots are detected by the photodetector. The detected signal from the photodiode is then processed by the computer processing system and a topographical image is obtained.

The AFM operates in three different modes namely, (a) contact mode, (b) non-contact mode, and (c) tapping mode. The difference between these operating modes comes from the height difference between the tip and the sample surface. In the contact mode, the tip flies at a height

of the order of a few angstroms or lesser from the sample surface. Therefore, the force between the tip and the sample surface is repulsive. On the other hand, in the non-contact mode, the tip flies at a height of tens to hundreds of angstroms from the sample surface. In this case, the inter-atomic force between the tip and the sample surface is attractive and mainly comes from the long-range van der Waals force. The height in the tapping mode is intermediate to contact and non-contact modes. Figure 3.3 (a) shows the schematic display of the AFM measurement system. Figure 3.3 (b) shows the typical AFM image obtained after scanning the surface of a sample [7]. The image is related to the dot patterns obtained after performing nano-imprint lithography (NIL) on Si substrate. The image was taken during the installation and optimization of the NIL technique in our lab [7].

For the visual understanding of the magnetic properties of a magnetic material, magnetic force microscopy (MFM) is employed [10-13]. Unlike AFM, a magnetic tip is used to observe the magnetic images in MFM. In other words, the AFM tip undergoes a thin magnetic coating on the top for being utilized as an MFM tip. When the tip travels in the vicinity of the magnetic sample (~ 50 nm), the stray field coming from the magnetic sample interacts with the magnetic tip through dipolar interaction. To avoid the inter-atomic force between the tip and the sample, the height of the tip is adjusted accordingly. Depending on whether the magnetic interaction force is attractive or repulsive, the tip and hence cantilever moves downward or upward. The reflected laser beam experiences a corresponding deflection in the photodetector (like AFM) and the output of the photodetector generates the resulting magnetic domain map after passing through the computer processing scheme. For the samples with perpendicular magnetic anisotropy (PMA), the up (down) magnetic domains generate outgoing (incoming) magnetic flux lines. These flux lines directly interact with the magnetic tip and result in a domain image with up-down domains. However, in the samples with in-plane (IP) anisotropy, the domains are oriented in the sample plane and the stray field only comes from the positions of domain

walls (DWs) and hence only DWs are imaged clearly. A Cypher AFM (and MFM) from Asylum Research was used to observe the topography and domain mapping in this thesis. The AFM was mostly utilized for determining the thickness (a few nanometer-thin films) and surface morphology of magnetic and non-magnetic thin films. Moreover, the domain imaging of PMA films and devices was obtained using the MFM mode of the machine. Figure 3.3 (c) shows a typical MFM domain map of PMA thin film [7, 14]. This image corresponds to a perpendicularly magnetized $\text{Co}_{50}(\text{Pt}_x\text{Rh}_{1-x})_{50}$ thin film. Here, red and green colors represent the magnetic domains pointing up and down directions, respectively. The regions with intermediate color (mostly yellowish) represent the DW separating two neighboring domains.

Figure 3.3. (a) The schematic illustration of atomic force microscopy, (b) typical AFM image obtained after scanning the surface of a sample [7]. This image corresponds to the dot patterns prepared using nano-imprint lithography on Si substrate. (c) a typical MFM image of PMA samples [7, 14]. This corresponds to a perpendicularly magnetized $\text{Co}_{50}(\text{Pt}_x\text{Rh}_{1-x})_{50}$ thin films.

3.2.3. Vibrating Sample Magnetometer (VSM)

The vibrating sample magnetometer (VSM) was pioneered by S. Foner in 1955 [15]. It is used for the magnetic characterizations of the magnetic samples. The working principle of VSM is based on Faraday's law of electromagnetic induction [15, 16], described as [17],

$$E \propto -\frac{d\phi}{dt} \quad (3.2)$$

Where E stands for induced electromotive force (emf) and $\frac{d\phi}{dt}$ is the rate of change of magnetic flux as a function of time. The sample is mounted on the sample rod and placed between the poles of an electromagnet. The applied magnetic field from the electromagnets aligns the magnetic moments of the magnetic sample along the applied magnetic field. This results in the generation of the stray field at the surface of the sample. When the sample is vibrated in a direction perpendicular to the magnetic field i.e. along the y-axis using the vibrator assembly (see Figure 3.4 (a)), the flux passing through the pick-up coil, placed in the vicinity of the sample, alters. The alteration in the magnetic flux induces an emf in the pick-up coil. The induced emf, in general, depends on (i) frequency of vibration, (ii) amplitude of vibration, (iii) applied magnetic field, and (iv) magnetic moment of the sample. However, with the proper calibration of the system, the magnetic moment associated with the system can be obtained. Figure 3.4 (a) shows the schematic illustration of the VSM instrument.

The Lakeshore 8600 VSM system was utilized for obtaining the magnetic hysteresis (M-H) loops along in-plane (IP) and out-of-plane (OOP) directions (with respect to the applied magnetic field) of the studied magnetic thin films in the present thesis. From IP and OOP M-H loops, the magnetic properties such as coercivity, saturation magnetization, remanence, anisotropy field can be evaluated. Figure 3. 4 (b) shows the M-H loop of the perpendicularly magnetized (Co/Ni)₃ multilayer films in the easy axis (OOP) direction [18].

Figure 3.4. (a) The schematic illustration of the vibrating sample magnetometer (VSM) instrument and (b) typical magnetic hysteresis (M-H) loop measured using the VSM system. The M-H loop corresponds to the perpendicularly magnetized (Co/Ni)₃ multilayer system along the easy axis direction [18].

3.2.4. Alternating Gradient Force Magnetometer (AGFM)

Alternating gradient force magnetometer (AGFM) or alternating gradient magnetometer (AGM) or vibrating reed magnetometer (VRM) is another member of the family of magnetometry [19, 20]. The AGFM system is more sensitive than the vibrating sample magnetometer system. In AGFM, the sample is attached to one end of the sample holder (rather a rod) and placed between the electromagnetic coils. Simultaneous direct current (DC) and alternating gradient magnetic fields are applied to the sample. The field gradient applies an alternating force on the sample, which is given by,

$$F_z = m_z B_z \frac{dH_{gradient}}{dz} \quad (3.3)$$

Where m_z , B_z , and H_{gradient} stand for the magnetic moment of the sample, DC field and gradient field applied to the sample, respectively. This alternating force makes the sample and hence the sample rod oscillating at a certain frequency. When the frequency of oscillations is made equal to the resonant frequency of the system, the amplitude of vibrations enhances by several orders. These oscillations are detected by the piezoelectric system and converted to voltage signals, which are proportional to the magnetic moment of the sample [21]. There are, however, some limitations to this measurement system, particularly the smaller sample size, which reduces the achievable signal. The MicroMag AGM 2900 system (Lakeshore) was used to obtain the IP and OOP hysteresis loops of the $\text{Co}_{80}\text{Pt}_{20}$ thin film samples. A schematic of the AGFM system is shown in Figure 3.5.

Figure 3.5. The schematic depiction of alternating gradient force magnetometer (AGFM).

3.2.5. Torque Magnetometry

The torque magnetometry is one of the most accurate methods to find out the anisotropy direction and the value of the anisotropy constant for a magnetic sample [19, 22]. The basic

principle of torque magnetometry relies on the fact that, when a magnetic sample is placed in a magnetic field, the field exerts a torque (or force) to align the magnetization along the field direction [19, 23, 24]. An application of a magnetic field in a direction away from the easy axis results in aligning the magnetization along the direction of the magnetic field. However, the anisotropy tries to pull it back along the easy axis direction. Therefore, the torque needed to align the magnetization along a certain direction (parallel to applied magnetic field), and hence the corresponding anisotropy energy (rather a torque) can be estimated. In the following section, the working design of the currently available torque magnetometers will be explored. A pair of electromagnets are used to apply a large magnetic field to the sample. When the magnetic sample, attached to the sample holding rod, is placed between the poles of an electromagnet, it exerts a torque on the sample. An electromagnetic coil, known as the compensation coil, is attached to the other end of the sample holding rod. This compensation coil is placed between a permanent magnet of known strength. By sending a current into the compensation magnet, the torque on the sample can be compensated. This is done accurately by using an assembly of a laser source, a mirror, and two photodetectors placed according to Figure 3.6 (a). Therefore, the amount of current sent to the compensating magnet is proportional to torque on the magnetic sample. The electromagnet assembly is rotated from -180° to $+180^\circ$ for estimating the torque acting on the sample in all directions. Moreover, the whole experiment is usually done for different field values for accurate estimation of the anisotropy direction as well as anisotropy constant. Figure 3.6 (a) shows a schematic representation of a torque magnetometer system. From the torque measurements, we obtain a graph between the torque and the rotation angle as a function of a magnetic field. If E_{anis} is the anisotropy energy of the system and L is the torque on the sample, they take the following relationship [25, 26],

$$L = -\frac{dE_{anis}}{d\theta} \quad (3.4)$$

Where θ is the rotation angle. Since, E_{anis} depends on the anisotropy constant (or constants depending on the magnetic system under consideration), the anisotropy constant (or constants) can be deduced. Note, the most accurate estimation occurs from the torque curves measured at the fields larger than the saturation field of the sample.

Figure 3.6. (a) The schematic diagram of torque magnetometer and (b) exemplary magnetic torque curve for a PMA sample [7, 14].

In this thesis, we have used a TAMAKAWA torque magnetometer of model “TM-TR2050-HGC” for estimating the anisotropy constant of FePt-C samples. This torque magnetometer can go to a maximum field of 25 kOe. Moreover, this is a highly sensitive torque magnetometer and can measure minute torques. Figure 3.6 (b) is a typical torque curve, measured using the mentioned system, for a sample with perpendicular magnetic anisotropy (PMA) [7, 14]. The torque curve qualitatively confirms the presence of PMA as the first peak in the torque curve appears in the first quadrant [19]. Moreover, the presence of only one peak between an angle

of 0° to 90° confirms the uniaxial anisotropy [19]. Upon further processing this torque curve, the quantitative value of the anisotropy constant can be deduced.

3.2.6. Magneto-optic Kerr Microscopy

Magneto-optic Kerr effect (MOKE) microscopy is an alternative system to observe magnetic domain patterns in real-time and to measure the magnetic hysteresis loops of magnetic thin films. When the thickness of the sample is too less (say 1 nm or less), the VSM system may not be able to measure the M-H loop accurately. In such situations, the MOKE system is very useful, though it does not measure the absolute value of the magnetic moment of the magnetic system [27-29]. Moreover, while measuring the micro(or nano)meter-sized devices, the MOKE system becomes crucial (within the resolution of the MOKE system). Besides, the MOKE system is very useful for measuring the real-time changes (due to external perturbations such as a magnetic field or electrical current or both) in the magnetic properties of the system.

The working principle of the MOKE system is based on the Kerr effect, named after discoverer J. Kerr [30-34]. A light beam with linear polarization incidents on the magnetic sample and interacts with the magnetization of the sample. As a result, the polarization of the reflected light is rotated with respect to the polarization of the incident beam. The extent of rotation of the polarization of reflected light depends on the magnetization of the sample. The reflected light passes through the analyzer, whose transmission axis is, in general, rotated by 90° with respect to the polarizer. After this, the light falls on the photodetector and it reads the intensity of the falling light and the lock-in amplifier converts this intensity to the corresponding voltage signal. From this voltage signal, the magnetization of the sample is deduced. Figure 3.7 (a) shows the schematic of the MOKE system.

The incident light can be p-polarized or s-polarized depending on whether the electric field vector is in or perpendicular to the plane of incidence [35]. When a p-polarized light gets reflected from the magnetic sample, the reflected light consists of p as well as s-components,

where the s-component is usually out of phase with respect to the p-component. This results in the elliptical polarization of the reflected light. Moreover, the transverse axis of the reflected light is also rotated with respect to the transverse axis of the incident light. A similar phenomenon happens with the incident light with s-polarization. This phenomenon is called Kerr rotation or Kerr ellipticity. Depending on the relative configuration of the magnetization and incident light, the Kerr effect can be categorized into (a) Longitudinal, (b) Transverse, and (c) Polar Kerr effect [35]. In the transverse Kerr effect, the magnetization of the sample is along the film plane and in the plane of incidence of the laser beam. In the longitudinal Kerr effect, the magnetization in the film plane while perpendicular to the plane of the incident for the laser beam. However, in polar MOKE, the magnetization points in a direction perpendicular to the film plane and in the plane of incidence.

In this thesis, we have utilized the Magvision Kerr system to perform the experiments on thin films as well as devices. An ultrabright 2200 lumens LED was used as a light source in the system. Moreover, a filter with narrow bandwidth was used to achieve highly polarized light, important for high contrast. The OOP magnet attached to the system can reach a maximum magnetic field of 1 T. On the other hand, the IP magnet can reach a maximum field of 0.65 T. The IP magnet has a small OOP magnet assembly, which can generate an OOP field of ~ 200 Oe. This assembly is useful for performing DMI and SOT measurements, where simultaneous IP and a small OOP field are required. Figure 3.7 (b) shows an exemplary MOKE measurement. This corresponds to the M-H loop and corresponding Kerr microscopy images (major ones) for a PMA sample [7].

Figure 3.7. (a) The schematic illustration of the magneto-optic Kerr effect (MOKE) system and (b) exemplary magnetic hysteresis loop and Kerr microscopy images (at various stated fields) for a PMA sample [7].

3.3. Device Fabrication: Fabrication of Dot Patterns for Ion Implantation Through Hard Mask

To fabricate the hard mask for ion implantation in CoPt and FePt samples through the mask, we utilized the self-assembly of the DBCP lithography technique. For this, we initiated and optimized the self-assembly process in our lab. As mentioned previously, this mask pattern aids the advantage of creating local modifications in the structural and magnetic properties in the way of preparing patterned films for applications in heated-dot magnetic recording.

3.3.1. Self-assembly of Di-block Copolymers

The block copolymers have the capability to self-assemble into regular and periodic nanopatterns [36, 37]. The block copolymers are of various types such as di-block, tri-block, and star-block, the di-block copolymers (DBCP), however, are the simplest ones [38]. The DBCPs are made up of two polymer chains (for instance, A and B, as shown in Figure 3.8 (a)), whose physical properties are dissimilar to each other. Therefore, they are purely phase-separated from each other. However, when these dissimilar polymers constitute a DBCP, they get attached through strong covalent bonds. Moreover, the microphase separation between the two components of DBCP is energetically unfavorable in normal conditions. In other words, one needs to supply the energy from outside to achieve the microphase separation. Researchers have studied the annealing process results in the microphase separation between the two components of DBCP [38-40]. Two types of annealing processes, namely, thermal annealing and solvent annealing have been studied for this purpose. However, previous studies indicate that the solvent annealing is preferred as it results in complete microphase separation [38, 41, 42]. The solvent annealing reduces the glass transition temperature of the polymer chains. This results in the enhancement of the mobility of constituent polymer chains and hence induces phase separation.

Figure 3.8. (a) The schematic representation of two components of A and B of DBCP and (b) the shape of patterns as a function of volume fraction.

Besides, the Flory-Huggins interaction parameter (χ), degree of polymerization (N), and volume fraction of the constituent polymers are important parameters that govern the morphology, orientation, and the ordering of the resulting structural arrays [43-48]. The size of the patterns is decided by the parameters χ and N through the following relation,

$$d = aN^{\frac{2}{3}}\chi^{\frac{1}{6}} \quad (3.5)$$

Here, a is the segment size of the monomer unit which forms the pattern. In addition, the χ parameter plays an important role in microphase separation. Moreover, the volume fraction of the individual polymer is a responsible parameter for the shape of the patterns. Consider V_A and V_B are the volume fractions of the polymer A and B respectively. When V_B is less than 21%, dot patterns of polymer B are obtained in the matrix of polymer A. Similarly, when V_B is in-between 21% and 33%, the cylindrical patterns of component B are achieved in the matrix of A. However, when V_B exceeds 33%, but less than 63%, the alternative lamellar patterns of polymers A and B are achieved. They are schematically shown in Figure 3.8 (b).

In the present case, the Polystyrene (PS)-Polydimethylsiloxane (PDMS) DBCP, as shown in Figure 3.9, were utilized. The reasons for choosing PS-PDMS DBCP are, (a) organic-inorganic

blocks, (b) easier selective reactive ion etching (RIE), and (3) high χ parameter of 0.26 [41, 42]. These advantages give rise to the better microphase separation of the two components upon solvent annealing and hence more uniform structures. Subsequently, the selection of suitable solvents for the solvent annealing process was performed. This was done by comparing Hansen's solubility parameter for PS, PDMS, and solvents to be used for microphase separation. Hansen's solubility parameter is an estimation of the degree of interaction between the materials [49, 50]. In other words, when Hansen's solubility parameter of two different species is close to each other, the interaction probability is higher for the corresponding species.

Figure 3.9. The schematic representation of respective components of PS-PDMS di-block copolymer.

Table 3.1 is a list of Hansen's solubility parameters for Tetrahydrofuran (THF), toluene, heptane, PS, and PDMS. The Hansen's solubility parameter of THF and PS are very close to each other and therefore, PS is soluble in THF. Therefore, THF will provide more chain mobility to PS than PDMS. However, Hansen's solubility parameters of toluene and heptane are very close to that of the PS and PDMS, respectively. Therefore, PS is soluble in toluene, and PDMS is soluble in heptane. A mixture of toluene and heptane will provide kinetic energy to both PS and PDMS and induce them to move away from each other. Hence, the mixture of

toluene and heptane was considered for microphase separation of PS-PDMS DBCP. After deciding the experimental ingredients, the optimization of the parameters for obtaining the uniform dot patterns was initialized.

Table 3.1. Hansen’s solubility parameter for indicated solvents, PS, and PDMS.

Solvent/Polymer	Hansen’s Solubility Parameter [(MPa) ^{1/2}]
THF	18.6 [50, 51]
Toluene	18.2 [49, 50]
Heptane	15.3 [50, 52]
PS	18.5 [50]
PDMS	15.5 [50]

3.3.2. Process Flow of Fabricating Dot Patterns

The patterning process includes the following steps,

- I. Preparation of PS-PDMS DBCP and toluene solution at room temperature.
- II. Spin coating of the prepared solution on the sample under study. Before spin-coating, the samples are cleaned thoroughly using DI water, acetone, and isopropyl alcohol (IPA). The details of the cleaning process will be discussed in sub-section 3.4.
- III. Solvent annealing of the spin-coated sample in the presence of toluene and heptane mixture in a closed beaker. The solvent annealing process results in the microphase separation of PS and PDMS. The dots of PDMS (volume fraction of PDMS is selected accordingly, details are provided later) dots were formed in the matrix of PS. Moreover, a thin layer of PDMS was also formed.
- IV. Reactive ion etching (RIE) using CF₄ was performed to etch the thin PDMS layer on the top.

V. Lastly, RIE using O_2 ions was performed to etch the PS matrix. This process resulted in the PDMS dots on the columns of PS.

This process is schematically presented in Figure 3.10. The experimental parameters such as the concentration of PS-PDMS DBCP in toluene, spin coating conditions, solvent annealing conditions, etc. are optimized after several iterations of experiments. The AFM was performed to observe and test the final product of the whole process.

Figure 3.10. Process flow of fabricating dot-patterns using self-assembly of PS-PDMS DBCP.

3.3.3. Optimization of Experimental Parameters for Self-assembly of Di-block Copolymers

We have optimized all the experimental conditions involved in the process of achieving the regular and uniform dot patterns. After a wide literature survey, we have chosen the following conditions as a starting point, molecular weight: 51.5 kg/mol, the concentration of PS-PDMS in toluene: 1.5 wt%, spin coating speed: 4500 rpm, spin coating time: 30 sec, ratio of toluene and heptane in solvent annealing mixture: 6:1, solvent annealing time: 16 h and RIE recipe:

CF₄ plasma (3sccm) at 50W for 5 s, followed by O₂ plasma (20sccm) at 90W for 30 s. However, we could not get a perfect microphase separation. Subsequently, we varied the parameters to several values for achieving uniform dot patterns.

Figure 3.11. Evolution of dot-patterns during process optimization.

Figure 3.11 (a)- (d) shows the evolution of dot patterns during the parameter optimization. Figure 3.11 (a) corresponds to the conditions; molecular weight: 51.5 kg/mol, the concentration of PS-PDMS in toluene: 1.5 wt%, spin coating speed: 4500 rpm, spin coating time: 30 sec, ratio of toluene and heptane in solvent annealing mixture: 6:1, solvent annealing time: 16 h and RIE recipe: CF₄ plasma (3sccm) at 50W for 5 s, followed by O₂ plasma (20sccm) at 90W for 20 s. The resulted patterns lack microphase separation and partially separated patterns were obtained. When spin coating speed was changed to 6000 rpm, keeping all other parameters unchanged, better patterns were observed as shown in Figure 3.11 (b). However, the microphase separation is still partial. Moreover, the long-range patterning is missing. This inspired us to change the concentration of PS-PDMS in toluene. Better patterns were observed when the concentration was increased to 2.5 wt% as indicated by Figure 3.11 (c). When the spin coating time was decreased to 20 sec, the best patterns of all were observed as shown in Figure 3.11 (d). It should be noticed that several iterations of experiments were performed to identify the right experimental parameter at each step.

Table 3.2. *Tabulation of optimization and best-optimized parameters used for obtaining uniform and regular nanodots from self-assembly of PS-PDMS di-block copolymers.*

Parameter	Studied Conditions	Best Optimized Conditions
Molecular Weight	51.5 kg/mol (16% volume fraction of PDMS) 54 kg/mol (22% volume fraction of PDMS)	51.5 kg/mol (16% volume fraction of PDMS) 54 kg/mol (22% volume fraction of PDMS)
Concentration of PS-PDMS in PS-PDMS and Toluene Mixture	1 to 3 wt% in a difference of 0.5 wt%	2.5 wt%
Spin Coating Speed	4500, 5400, 6000, 6600 7500 rpm (acceleration was adjusted accordingly to get uniform films)	6000 rpm
Spin Coating Time	20, 30 s	20 s
Ratio of Toluene and Heptane in Toluene+ Heptane Mixture for Solvent Annealing	4:1, 5:1, 6:1, 7:1	6:1
Solvent Annealing Time	16 h	16 h
CF ₄ and O ₂ RIE	1. CF ₄ plasma (3sccm) at 50W for 5 s, followed by O ₂ plasma (20sccm) at 90W for 30 s 2. CF ₄ plasma (3sccm) at 100W for 5 s, followed by O ₂ plasma (20sccm) at 90W for 20 s 3. CF ₄ plasma (3sccm) at 50W for 5 s, followed by O ₂ plasma (20sccm) at 90W for 20 s	CF ₄ plasma (3sccm) at 50W for 5 s, followed by O ₂ plasma (20sccm) at 90W for 20 s

During the optimization process, we have also utilized PS-PDMS DBCP with a molecular weight of 54 kg/mol. Notably, we have observed that the best-optimized parameters are the same for the case of PS-PDMS with a molecular weight of 51.5 as well as 54 kg/mol. This is because they are almost similar in properties. We have used the best-optimized conditions for preparing the mask patterns for CoPt and FePt media samples. The above table shows the list of parameters studied during the parameter optimization process. Additionally, the observed best-optimized conditions are also indicated.

3.3.4. Ion Implantation

After fabricating the mask patterns on the media samples, we performed the ion implantation of non-magnetic species. The ion implantation is a way to alter the structural and magnetic properties of a magnetic system in a controlled way [53]. The implantation parameters such as type and energy of the ions, fluence of implantation are chosen according to the requirements. The interaction between the foreign ions and host atoms depends on the type of ions and their energy. For the case of highly energetic lighter ions (ions with lower atomic weights), the ions impart their energy to the electrons of the host atoms and excite them. The probability of changing the structural and magnetic properties is less in such implantations. However, for the case of low energetic heavier ions (ions with larger atomic weights), the ions undergo elastic collisions with the nuclei of host atoms. This process results in the energy transfer from implanted ions to host atoms and the displacement of host atoms happens. These displaced atoms undergo several collisions with other atoms before losing most of their energy. Therefore, a cascaded scattering process happens in the host lattice and concludes in a significant lattice disorder and defect generation. During the cascaded scattering process, the temperature of the host lattice rises to very high values (of the order of 10^4 to 10^6 °C). Although the system comes in thermal equilibrium in a time span of picoseconds, it significantly enhances the diffusion of defects, introduced during cascaded scattering [54].

The ion implantation process can be theoretically estimated using the Transport of Ions in Matter (TRIM) program from Stopping and Range of Ions in Matter (SRIM) simulations [55-57]. Following experimental needs, the type and energy of ions, fluence of implantations can be estimated from these simulations. The $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ions were utilized for ion implantation into CoPt and FePt samples in the present thesis. All the implantations were performed by our collaborators at in Inter-University Accelerator Center (New Delhi, India). While discussing the experimental findings of masked ion implantations, the details of implantation parameters will be presented.

3.3.5. Transport of Ions in Matter (TRIM)

The Transport of Ions in Matter (TRIM) program from Stopping and Range of Ions in Matter simulations is an open-source software, developed by J. Zeigler [55-57]. The program uses statistical algorithms based on Monte Carlo simulations to estimate the implanted ions-host atoms interactions mediated by electronic excitations or nuclear collisions or both [55, 56]. This program can calculate the distribution of implanted ions, distributions of host atoms after cascaded scattering, target damage, sputtering of the target, ionization, production of phonons, etc. However, we have utilized TRIM software to calculate the distribution of implanted ions and distributions of host atoms after cascaded scattering in the present study. To perform simulations, one needs to input the following entries,

- a) The ion to be implanted
- b) The energy of the implanted ion
- c) The fluence of the implanted ion
- d) The implantation angle
- e) The layer structure of the host material with detailed composition of every layer

Figure 3.12 shows an exemplary image of the distribution of the implanted ions and the distributions of the host atoms. The simulations have been performed for Si/SiO₂/ Ni₅₀Fe₅₀ (10

nm)/Pt (5 nm) structure, where $^{102}\text{Ru}^+$ ions (energy: 20 keV and fluence: 5000) were implanted. The higher accuracy in the simulations is obtained by choosing a higher number of implanted ions (fluence). Figure 3.12 (a) shows the distribution of implanted ions and Figure 3.12 (b) shows the distributions of host atoms after ions-atoms interactions. The implanted ions are mostly concentrated in Pt and $\text{Ni}_{50}\text{Fe}_{50}$ layers. Moreover, the host atoms are also distributed in the respective layers. However, there is a hint of intermixing of Pt, Ni, and Fe atoms across the Pt/ $\text{Ni}_{50}\text{Fe}_{50}$ interface.

Figure 3.12. TRIM simulations to calculate the (a) distribution of implanted ions and (b) distributions of host atoms after cascaded scattering for $^{102}\text{Ru}^+$ ion-implantation on $\text{Si}/\text{SiO}_2/\text{Ni}_{50}\text{Fe}_{50}$ (10 nm)/Pt (5 nm) layer structure [7].

3.4. Device Fabrication: Fabrication of Domain Wall Devices

In order to perform the DW motion experiments on DW devices, DW device fabrication was performed. Prior to device fabrication, magnetic multilayers (of interest) were deposited using DC/RF magnetron sputtering process (as discussed before) in all the cases of device fabrication. The device fabrication includes the following steps.

3.4.1. Sample Cleaning and Prebaking

The first step in the device fabrication process is the proper cleaning of the sputter-deposited samples. The cleaning process was initialized by the ultrasonication of the samples in de-ionized (DI) water for 3 mins. This is to remove all the loose materials intact to the surface of the samples. Subsequently, the samples were ultrasonicated for 3 mins in acetone to remove all the organic contaminants from the sample surface. Lastly, the samples were ultrasonicated in IPA for 3 mins to remove the residues of acetone. After all the ultrasonication baths, the samples were dried using the pure N gas. To further avoid any moisture on the sample surface, the samples were prebaked at 110 °C for 1 min.

3.4.2. Spin Coating of Photoresist and Post-baking

Subsequently, the photoresist was spin-coated on the top of the samples. The spin coating conditions used were, speed: 3500 rpm, acceleration: 2000 rpm/sec, time: 30 sec. The positive resists AZ MIR 701 and AZ 1505 were used for the fabrication of DW devices. These spin coating conditions resulted in the resist thickness of ~700- ~800 nm. After the spin coating, samples were post-baked at 110 °C for 1 min for ensuring the complete evaporation of all the liquid residues.

3.4.3. Optical Lithography for Device Fabrication and Development

The word lithography has been taken from the Greek words lithos and graphia [58]. Where lithos means the stone and graphia means to write. Therefore, lithography literally means writing on stones. However, in the context of micro and/or nanofabrication, lithography refers to the development of devices of various shapes and sizes (according to the requirement) on the sample under investigation.

Usually, the sample with a photoresist on the top is exposed to the light through a hard mask. The mask contains the geometry of the device. For the case of positive resist, the exposed area of the resist becomes more soluble. Once the exposed sample is dipped in the developer, it

removes the exposed resist. However, in the present case, a MicroWriter ML[®]3 of Durham Magneto Optics optical lithography tool, which is a maskless lithography tool, was utilized. This machine uses a software-based mask, which recognizes certain pixels. The area which needs to be exposed is kept “white” and the rest of the area is kept “black”. However, in the case of the negative resist, the exposed area becomes harder and unexposed resist goes away upon development. The schematic of the optical lithography process is shown in Figure 3.13.

Figure 3.13. The schematic depiction of optical lithography with the virtual mask. By virtual mask, I mean the absence of a physical mask. Besides, the exposure is controlled using pixel recognition by the tool's software.

During the fabrication of the devices, the input image file was designed in such a way that the devices were made in black color and the rest of the area in white color. During exposure, the light interacted in the regions of white in the image file and made the resist in this area more soluble. The exposure time and exposure dose were set on the basis of the dose optimization process, carried out before the fabrication of each sample. Subsequently, the samples were developed in AZ 726 MIF developer. This process resulted in the resist only at the place of devices (in the shape of devices).

3.4.4. Ion Milling

The previous step resulted in the resist mask at the place of devices. The samples were then ion milled to transfer the patterns from the resist layer to the sputtered films. The ion milling time was optimized by etching the test samples and measuring the etched thickness using AFM. In order to understand the ion milling process, the underlying working principle is briefly explained below.

Ion milling, also known as reverse sputtering, is a dry etching process used to etch the thin film samples. A highly energetic beam of inert gas atoms (for instance Ar gas atoms) is bombarded on the surface of the sample, which needs to be etched. The sample is rotated at constant RPM to ensure a uniform etching of the sample. Moreover, the sample is placed at a certain angle with respect to the ion beam to minimize the re-deposition on the sample surface. Usually, the ion-milling systems are designed with a secondary ion mass detector (SIMS) to identify the elements at the surface of the sample.

3.4.5. Resist Removal, Sample Cleaning, and Prebaking

The ion milling process resulted in the devices (to be fabricated) of the material under study. However, this process usually leaves some residual resist on the top of the devices. Therefore, the residual resist was cleaned by the ultrasonication of the sample in the Technistrip P1331 stripper. To ensure the cleanliness of the sample, we again cleaned and pre-baked the samples

following the steps described in point 1. For electrical measurements of the devices, one needs electrode pads of some highly conductive metal such as Au or Cu. Therefore, one more round of the fabrication process was performed.

3.4.6. Spin Coating, Post-baking, Optical Lithography for Electrode Fabrication and Development

To fabricate the electrodes, the steps 3.4.1 to 3.4.3 were repeated. However, for the case of electrodes, electrodes and the background were made in white and black color, respectively. After the exposure and development, the resist from the place of electrodes was removed and the rest of the portion was covered with the resist layer. The experimental conditions and materials used were the same as above.

3.4.7. Electrode Deposition and Lift-off

Subsequently, the electrode material Ta (5 nm)/Cu (90 nm)/ Ta (5nm) was deposited. The bottom and top Ta layers are for adhesion and prevention from the oxidization of the Cu layer, respectively. The Ta/Cu/Ta is in direct contact with the Si substrate and nucleation pads of the device in the exposed area. However, in the rest of the area the electrode material deposited on the top of the resist. After depositing the electrode material, samples were kept in the Technistrip P1331 stripper. As a result, the resist and the electrode material on resist flew off and electrode material only remains in the exposed area. This process is called the lift-off process. This way, the DW devices with electrode pads were obtained as an end product. Figure 3.14 illustrates the complete process flow for the fabrication of DW devices. Once the devices were fabricated, the device testing and actual experiments were conducted.

Figure 3.14. Process flow of fabrication of DW devices with electrode pads.

3.5. Measurement Assembly for Domain Wall Devices

This section gives a brief overview of the measurement assembly for observing magnetic field or electrical current or both induced DW motion in DW devices. After fabricating the devices, we mount the devices on the chip carrier (shown in Figure 3.15) using double-sided tape. The chip carrier is designed in such a way that one end of the internally connected wire can be attached to the device and the other end to the break-out box. There are five pairs of such internally connected wires. The connection between the device and the chip carrier is established using the wire-bonding machine, which comes with an optical microscope (refer to Figure 3.15). This machine is very useful for connecting wires in devices with very small electrodes (typically, $300\text{ }\mu\text{m} \times 300\text{ }\mu\text{m}$ in our lab). Once the device is physically and electrically mounted on the chip carrier, the (chip carrier + device) assembly is mounted on the stage of Kerr microscopy. Depending on the experimental requirements, the connections are made between the MOKE system, current source, and nano-voltmeter.

To perform, the DW motion experiments usually the first step is to measure the M-H loop of the device to extract the coercivity of the device. This is because the nucleation of the reversed domain happens very close or at the coercivity value. In the present thesis, we have performed all the DW measurements for samples with PMA, therefore the discussion is centered only on PMA samples (rather devices). For the DW insertion, first, the device is saturated using a high OOP magnetic field (say +500 Oe). Subsequently, a reversal field (also called nucleation field) is applied to nucleate a reverse domain and hence to insert DW in DW devices. To facilitate the optimized DW insertion, large nucleation pads (usually much larger than the width of micro or nanowire) are fabricated at one or both ends of the wire. The dimensions of the nucleation pads may vary for different systems. The nucleation of the reversed domain happens in the nucleation pad and expands to insert a DW in the nanowire. After the nucleation of a reversed domain, the external stimuli are applied and real-time changes in magnetic configurations can be captured using Kerr microscopy. Moreover, the nano-voltmeter can also be used to record the change in resistance of the DW device due to DW motion, Hall voltage, etc. (according to needs). We have utilized Keithly 6221 current source for supplying the current to the DW devices and Keithly 2182A nano-voltmeter. For the schematic representation of the measurement set-up, please refer to Figure 3.15.

Figure 3.15. The schematic illustration of measurement assembly for DW devices [7].

3.6. Sheet Resistance Measurements

The sheet resistance measurements were performed on the thin film samples related to the study in chapter 6 (and chapter 7). Although the measurements were performed on thin films, the preparation of samples and measurements require optical lithography, current source, and nano-voltmeter. Hence, the details of measurements are being discussed after introducing the above-mentioned techniques. The sheet resistance was measured using the standard four-probe method [59]. For this, four electrode pads of dimensions $300\text{ }\mu\text{m} \times 300\text{ }\mu\text{m}$ were fabricated using optical lithography. Subsequently, the Ta (5 nm)/Cu (90 nm)/Ta (5 nm) electrode material was deposited and then lift-off was performed. The center to center spacing between the adjacent pads was fixed at 0.25 cm. Further, the current (I) was sent through the outer

electrodes, and voltage (V) was measured between the inner electrodes. From V vs I graph, the sheet resistance (R) was estimated. Figure 3.16 shows the schematic diagram of the standard four-probe arrangement.

Figure 3.16. The schematic illustration of the standard four-probe method for measuring the sheet resistance of the thin film samples.

To estimate the resistivity from measured sheet resistance, the following relationship was utilized [60],

$$\rho = \frac{RA}{L} \quad (3.6)$$

Where A and L stand for cross-sectional area and spacing between two neighboring electrodes. Since, A can be written as,

$$A = t \times a \quad (3.7)$$

Where t is the total thickness of the film stack and a is the length of the cross-section (as indicated in Figure 3.16). Inserting these values in the above Eq. results in,

$$\rho = \frac{Rta}{L} \quad (3.8)$$

Upon inserting the corresponding values, the resistivity (ρ) can be obtained.

3.7. Micromagnetic Simulations

As described in the previous chapter, the statics and dynamics of a magnetic system are governed by the Landau–Lifshitz–Gilbert (LLG) equation. Micromagnetic simulators play an important role in understanding the magnetic system by computationally solving the LLG equation for certain geometrical and magnetic parameters. In the present research studies, we have used two kinds of micromagnetic simulators, namely, (a) Object Oriented Micromagnetic Framework (OOMMF) and (b) Mumax3. The OOMMF has been used to simulate the $\text{Co}_{80}\text{Pt}_{20}$ PMA system and the results are discussed in chapter 4. On the other hand, Mumax3 has been used to simulate the rest of the systems during the course of the thesis.

3.7.1. Object-oriented Micromagnetic Framework (OOMMF)

The OOMMF micromagnetic simulator has been developed by the National Institute of Standards and Technology (NIST) [61, 62]. The major contribution during the development of this software was provided by M. Donahue and D. Porter [61, 62]. The simulator uses the finite difference method for solving the LLG equation as a function of time as well as space parameters. The input code for solving a certain problem in OOMMF is written in the C++ language. Moreover, it provides the user, the independence of modifying the code according to the use. In general, to perform the micromagnetic simulations using OOMMF, one has to define the geometry of the system, the magnetic parameters, external stimuli like a magnetic field or electrical current, or both (if needed), etc. The system to be simulated is usually divided into several identical subsystems, called cells, and one cell is considered as a single magnetic moment or spin. The cell size must be chosen smaller than the exchange length of the magnetic system under consideration for computational accuracy. The program uses evolvers for performing step by step computation for a given system. These evolvers are of two kinds, (a) time evolvers, which use LLG dynamics, and (b) minimization evolvers, which locate local

minima in the energy surface through direct minimization techniques. The function of the evolvers is controlled by various drivers and can be defined according to the needs. Like other computation software, the simulations get finished once the total energy of the system reaches the minimum energy state unless stopped halfway. The OOMMF generates, (a) scalar quantities such as energies, magnetization, etc., and (b) vector field display, which gives spatial distribution of the magnetization as an output.

3.7.2. Mumax3

The Mumax3 software is developed by V. Waeyenberge and his team at Ghent University [63-65]. Like OOMMF, Mumax3 also uses the finite difference method for computing the LLG equation. This is an open-source software (like OOMMF). However, Mumax3 has a more user-friendly interface than OOMMF. The program provides many in-built definitions, which can directly be used in the user-written codes to perform several kinds of simulations. This simulation software is a graphical processing unit (GPU) accelerated program and hence ~100 times faster than the central processing unit (CPU) based simulation programs. Besides, this simulator consumes less memory and several simulations can be run simultaneously. Similar to OOMMF, any given geometry is divided into several small cells with a cell size less than the exchange length of the system. According to the requirements, one needs to define the geometrical, magnetic, and perturbative parameters (and others if required) for a certain simulation case. The magnetization statics or dynamics for each cell is recalculated by solving the LLG equation after every time step using the finite difference method until the total energy of the system reaches its minima. Like OOMMF, the scalar and vector output files are generated once the simulation is finished.

3.8. Conclusions

The experimental and computational tools to achieve the goals of this thesis have been introduced in this chapter. All the thin film stacks were prepared using DC/RF magnetron sputtering process. The characterization tools such as XRD, AFM, MFM, VSM, AGFM, Torque magnetometer, MOKE (in association with Kerr microscopy), and four-probe measurements employed to investigate thin film properties, were briefly discussed. Subsequently, the details of device fabrication methodologies and related experimental and computational techniques were discussed. These include self-assembly of PS-PDMS DBCP, ion-implantation, TRIM simulations, maskless optical lithography, and ion milling. Later, the measurement assembly for domain wall devices was described. In the end, the micromagnetic simulators such as OOMMF and Mumax3 were briefly introduced, used to simulate the statics and dynamics of magnetic systems studied in this thesis.

References

- [1] K. Wasa, I. Kanno and H. Kotera, *Handbook of sputter deposition technology: fundamentals and applications for functional thin films, nano-materials and MEMS*. William Andrew, 2012.
- [2] A. Herdt, *Exploring the electronic properties of novel spintronic materials by photoelectron spectroscopy*. Forschungszentrum Jülich, 2012.
- [3] S. Piramanayagam and T. C. Chong, *Developments in data storage: materials perspective*. John Wiley & Sons, 2011.
- [4] E. Y. Tsymbal and I. Žutić, *Spintronics Handbook: Spin Transport and Magnetism: Volume Two: Semiconductor Spintronics*. CRC press, 2019.
- [5] B. D. Cullity, *Elements of X-ray Diffraction*. Addison-Wesley Publishing, 1956.
- [6] P. Rajagiri, T. Parida, B. Sahu, V. Narayanan, S. Prasad and K. Ramanathan, "Enhancement of magnetization in substituted Zn ferrite thin films," *Journal of Magnetism and Magnetic Materials*, vol. 499, p. 166200, 2020.
- [7] D. Kumar and S. N. Piramanayagam, *Property of SpinLab, NTU (copyright: S. N. Piramanayagam)*, 2020.
- [8] G. Binnig and H. Rohrer, "Scanning tunneling microscopy," *Surface science*, vol. 126, no. 1-3, pp. 236, 1983.
- [9] G. Binnig, C. F. Quate and C. Gerber, "Atomic force microscope," *Physical review letters*, vol. 56, no. 9, p. 930, 1986.
- [10] U. Hartmann, "Magnetic force microscopy," *Annual review of materials science*, vol. 29, no. 1, pp. 53, 1999.
- [11] U. Kaiser, A. Schwarz and R. Wiesendanger, "Magnetic exchange force microscopy with atomic resolution," *Nature*, vol. 446, no. 7135, pp. 522, 2007.
- [12] R. B. Proksch, T. Schäffer, B. Moskowitz, E. Dahlberg, D. A. Bazylinski and R. B. Frankel, "Magnetic force microscopy of the submicron magnetic assembly in a magnetotactic bacterium," *Applied Physics Letters*, vol. 66, no. 19, pp. 2582, 1995.
- [13] R. Sbiaa, C. Z. Hua, S. Piramanayagam, R. Law, K. O. Aung and N. Thiyagarajah, "Effect of film texture on magnetization reversal and switching field in continuous and patterned (Co/Pd) multilayers," *Journal of Applied Physics*, vol. 106, no. 2, p. 023906, 2009.
- [14] Durgesh Kumar, Tianli Jin, Kim Kong Tham, Shin Saito and S. N. Piramanayagam, "Large Uniaxial Anisotropy $\text{Co}_{50}(\text{Pt}_{1-x}\text{Rh}_x)_{50}$ Thin Films For Microwave Assisted Magnetic Recording," *Presentation at TMRC*, 2018.
- [15] S. Foner, "Versatile and sensitive vibrating-sample magnetometer," *Review of Scientific Instruments*, vol. 30, no. 7, pp. 548, 1959.
- [16] D. Smith, "Development of a Vibrating-Coil Magnetometer," *Review of Scientific Instruments*, vol. 27, no. 5, pp. 261, 1956.
- [17] N. R. Sree Harsha, Anupama Prakash and D. P. Kothari, *The Foundations of Electric Circuit Theory*: IOP Publishing, 2016.
- [18] D. Kumar and S. N. Piramanayagam, "Domain Wall Pinning in Square Wave Shaped Magnetic Nanowires," *Presentation at Joint MMM-Intermag 2019 Conference*, 2019.
- [19] B. D. Cullity and C. D. Graham, *Introduction to magnetic materials*. John Wiley & Sons, 2011.
- [20] K. O'Grady, V. Lewis and D. Dickson, "Alternating gradient force magnetometry: applications and extension to low temperatures," *Journal of applied physics*, vol. 73, no. 10, pp. 5608, 1993.
- [21] P. Flanders, "An alternating-gradient magnetometer," *Journal of Applied Physics*, vol. 63, no. 8, pp. 3940, 1988.

- [22] F. Ono, "Torque curve measurements in HCP Co in very low fields," *IEEE Transactions on Magnetics*, vol. 23, no. 5, pp. 2428, 1987.
- [23] J. Rigue, D. Chrischon, A. De Andrade and M. Carara, "A torque magnetometer for thin films applications," *Journal of magnetism and magnetic materials*, vol. 324, no. 8, pp. 1561, 2012.
- [24] S. Brialmont, J.-F. Fagnard and P. Vanderbemden, "A simple torque magnetometer for magnetic moment measurement of large samples: Application to permanent magnets and bulk superconductors," *Review of Scientific Instruments*, vol. 90, no. 8, p. 085101, 2019.
- [25] C. Williams, J. Krebs, F. Rachford, G. Prinz and A. Chaiken, "Torque measurements of the magnetic anisotropy energy of antiferromagnetic-coupled Fe/Cr/Fe layers," *Journal of magnetism and magnetic materials*, vol. 110, no. 1-2, pp. 61, 1992.
- [26] X. Wang, R. Wu, D.-s. Wang and A. Freeman, "Torque method for the theoretical determination of magnetocrystalline anisotropy," *Physical Review B*, vol. 54, no. 1, p. 61, 1996.
- [27] Z. Qiu, J. Pearson and S. Bader, "Additivity of the magneto-optic Kerr signal in ultrathin Fe (110)/Ag (111) superlattices," *Physical Review B*, vol. 45, no. 13, p. 7211, 1992.
- [28] Z. Qiu and S. D. Bader, "Surface magneto-optic Kerr effect," *Review of Scientific Instruments*, vol. 71, no. 3, pp. 1243, 2000.
- [29] J. R. Hampton, J.-L. Martinez-Albertos and H. D. Abruña, "Development of a versatile SMOKE system with electrochemical applications," *Review of scientific instruments*, vol. 73, no. 8, pp. 3018, 2002.
- [30] R. Bjørn, K. B. Pedersen and L. Diekhöner, "Development of a Surface Magneto-Optic Kerr Effect Setup," *Master's Thesis*, 2011.
- [31] J. Kerr, "On Rotation of the Plane of Polarization by Reflection from the Pole of a Magnet," *Philosophical Magazine and Journal of Science*, vol. 3, no. 19, pp. 321, 1877.
- [32] J. Kerr, "A new relation between electricity and light: dielectric media birefringent (second paper)," *Philosophical Magazine and Journal of Science*, vol. 50, no. 333, pp. 446, 1875.
- [33] P. Weinberger, "John Kerr and his effects found in 1877 and 1878," *Philosophical Magazine Letters*, vol. 88, no. 12, pp. 897, 2008.
- [34] J. Kerr, "A new relation between electricity and light: Dielectrified media birefringent," *Philosophical Magazine and Journal of Science*, vol. 50, no. 332, pp. 337, 1875.
- [35] <https://people.bath.ac.uk/vkv23/English/RechercheT14a.htm>.
- [36] A.-C. Shi and B. Li, "Self-assembly of diblock copolymers under confinement," *Soft Matter*, vol. 9, no. 5, pp. 1398, 2013.
- [37] R. Ruiz, E. Dobisz and T. R. Albrecht, "Rectangular patterns using block copolymer directed assembly for high bit aspect ratio patterned media," *ACS NANO*, vol. 5, no. 1, pp. 79, 2011.
- [38] R. A. Griffiths, A. Williams, C. Oakland, J. Roberts, A. Vijayaraghavan and T. Thomson, "Directed self-assembly of block copolymers for use in bit patterned media fabrication," *Journal of Physics D: Applied Physics*, vol. 46, no. 50, p. 503001, 2013.
- [39] O. Hellwig, J. Bosworth, E. Dobisz, D. Kercher, T. Hauet, G. Zeltzer, J. Risner-Jamtgaard, D. Yaney and R. Ruiz, "Bit patterned media based on block copolymer directed assembly with narrow magnetic switching field distribution," *Applied Physics Letters*, vol. 96, no. 5, p. 052511, 2010.
- [40] T. R. Albrecht, D. Bedau, E. Dobisz, H. Gao, M. Grobis, O. Hellwig, D. Kercher, J. Lille, E. Marinero and K. Patel, "Bit patterned media at 1 Tdot/in² and beyond," *IEEE transactions on magnetics*, vol. 49, no. 2, pp. 773, 2013.

- [41] N. Kihara, R. Yamamoto, N. Sasao, T. Shimada, A. Yuzawa, T. Okino, Y. Ootera, Y. Kamata and A. Kikitsu, "Fabrication of 5 Tdot/in.² bit patterned media with servo pattern using directed self-assembly," *Journal of Vacuum Science & Technology B, Nanotechnology and Microelectronics: Materials, Processing, Measurement, and Phenomena*, vol. 30, no. 6, p. 06FH02, 2012.
- [42] A. Kikitsu, T. Maeda, H. Hieda, R. Yamamoto, N. Kihara and Y. Kamata, "5 Tdots/in² bit patterned media fabricated by a directed self-assembly mask," *IEEE transactions on magnetics*, vol. 49, no. 2, pp. 693, 2013.
- [43] T. H. Epps III and R. K. O'Reilly, "Block copolymers: controlling nanostructure to generate functional materials—synthesis, characterization, and engineering," *Chemical Science*, vol. 7, no. 3, pp. 1674, 2016.
- [44] M. J. Fasolka and A. M. Mayes, "Block copolymer thin films: physics and applications," *Annual Review of Materials Research*, vol. 31, no. 1, pp. 323, 2001.
- [45] R. A. Segalman, "Patterning with block copolymer thin films," *Materials Science and Engineering: R: Reports*, vol. 48, no. 6, pp. 191, 2005.
- [46] J. N. Albert and T. H. Epps, "Self-assembly of block copolymer thin films," *Materials Today*, vol. 13, no. 6, pp. 24, 2010.
- [47] S. Kim, C. M. Bates, A. Thio, J. D. Cushen, C. J. Ellison, C. G. Willson and F. S. Bates, "Consequences of surface neutralization in diblock copolymer thin films," *ACS nano*, vol. 7, no. 11, pp. 9905, 2013.
- [48] C. M. Bates, T. Seshimo, M. J. Maher, W. J. Durand, J. D. Cushen, L. M. Dean, G. Blachut, C. J. Ellison and C. G. Willson, "Polarity-switching top coats enable orientation of sub–10-nm block copolymer domains," *Science*, vol. 338, no. 6108, pp. 775, 2012.
- [49] V. Y. Senichev and V. V. Tereshatov, "General principles governing dissolution of materials in solvents," *Handbook of solvents*, pp. 101, 2001.
- [50] J. Brandrup, E. H. Immergut, E. A. Grulke, A. Abe and D. R. Bloch, *Polymer handbook*. Wiley New York etc, 1989.
- [51] K. Kilk, N. Savest, J. Hruľjova, E. Tearo, S. Kamenev and V. Oja, "SOLVENT SWELLING OF DICTYONEMA OIL SHALE," *Oil Shale*, vol. 27, no. 1, 2010.
- [52] S. Rasappa, L. Schulte, D. Borah, H. Hulkkonen, S. Ndoni, T. Salminen, R. Senthamaraikanan, M. A. Morris and T. Niemi, "Morphology evolution of PS-b-PDMS block copolymer and its hierarchical directed self-assembly on block copolymer templates," *Microelectronic Engineering*, vol. 192, pp. 1, 2018.
- [53] J. Fassbender and J. McCord, "Magnetic patterning by means of ion irradiation and implantation," *Journal of Magnetism and Magnetic Materials*, vol. 320, no. 3-4, pp. 579, 2008.
- [54] G. S. Was, "Ion beam modification of metals: compositional and microstructural changes," *Progress in surface science*, vol. 32, no. 3-4, pp. 211, 1989.
- [55] J. Ziegler, J. Biersack and U. Littmark, "The Stopping and Range of Ions in Solids (Pergamon, New York, 1985)," See [www. SRIM. org](http://www.srim.org) for SRIM version, 2008.
- [56] <http://www.srim.org/index.htm#HOMETOP>.
- [57] J. Ziegler, "Stopping of energetic light ions in elemental matter," *Journal of applied physics*, vol. 85, no. 3, pp. 1249, 1999.
- [58] T. Adams and R. Layton, "Introductory MEMS: Fabrication and Applications. Concordia University," ed: University of California Press, Oakland, CA, 2009.
- [59] C. Zhang, S. Fukami, K. Watanabe, A. Ohkawara, S. DuttaGupta, H. Sato, F. Matsukura and H. Ohno, "Critical role of W deposition condition on spin-orbit torque induced magnetization switching in nanoscale W/CoFeB/MgO," *Applied Physics Letters*, vol. 109, no. 19, p. 192405, 2016.

- [60] http://spiff.rit.edu/classes/phys213/lectures/resist/resist_long.html.
- [61] <http://math.nist.gov/oommf/>.
- [62] M. Donahue and D. Porter, "OOMMF user's guide, version 1.0, National Institute of Standards and Technology," *Interagency Report NISTIR*, 1999.
- [63] A. Vansteenkiste, J. Leliaert, M. Dvornik, F. Garcia-Sanchez and B. Van Waeyenberge, "The design and verification of mumax3, aip advances 4, 107133 (2014)," *CAS Article*.
- [64] A. Vansteenkiste and B. Van de Wiele, "MuMax: A new high-performance micromagnetic simulation tool," *Journal of Magnetism and Magnetic Materials*, vol. 323, no. 21, pp. 2585, 2011.
- [65] J. Leliaert, M. Dvornik, J. Mulkers, J. De Clercq, M. Milošević and B. Van Waeyenberge, "Fast micromagnetic simulations on GPU—recent advances made with," *Journal of Physics D: Applied Physics*, vol. 51, no. 12, p. 123002, 2018.

Chapter 4. Effects of Ion Implantation on Structural and Magnetic Properties of Masked Magnetic Thin Films

The present chapter is attributed to the study of the “self-assembly + ion-implantation” based magnetic patterning process for applications in magnetic recording. This research study was performed on two highly anisotropic alloys and therefore, has been divided into two sub-studies. The title of the first sub-study is “Tailoring the Structural and Magnetic Properties of Masked Co₈₀Pt₂₀ Thin Films Using High and Low Energy Ion Implantation” and has been partially published as “Kumar *et al.*, *IEEE Magnetics Letters*, 9, 1-5 (2018)” and “Kumar *et al.*, *AIP Advances*, 8(5), 056504 (2018)”. On the other hand, the title of the second sub-study is “Ion Implantation Induced Disorder in FePt-C Thin Films” and has been published as “Kumar *et al.*, *IEEE Transactions on Magnetics*, 55(3), 1-5 (2019)”. [Reproduced and/or inspired from mentioned papers, with the permission of respective publishers]

4.1. Introduction

As mentioned in earlier chapters, optimizing the suitable fabrication technique is one of the major issues with the heated-dot magnetic recording (HDMR) technology [1-4]. Conventionally, the physical patterning process has been proposed for the fabrication of patterned media as shown in Figure 4.1 (a) [2, 5-9]. In the physical patterning process, the patterns are made in the resist layer and subsequently, transferred to the magnetic layer using the dry etching process. The pattern fabrication on the resist layer has been studied using various lithography techniques such as electron beam lithography (EBL), nano-imprint lithography (NIL), and directed self-assembly (DSA) [8-11]. However, fabricating smaller patterns (say 10 nm and smaller) with minimum possible size distribution is challenging using EBL [12-14]. EBL has been studied for patterning the recording layer with a feature size of 10 nm using hydrogen silsesquioxane resist [15]. However, this technique is very expensive and time-consuming [16]. For instance, currently, the industry makes ~600 PMR disks/hour. On

the other hand, fabrication of one EBL based HDMR disk may take several days. Moreover, the addition of several intermediate steps in the preparation of one HDMR disk increases the cost significantly. Alternatively, NIL can be used for the physical patterning of the magnetic layer. It has been reported that the magnetic layer with a feature size of ~25 nm can be fabricated using this technique [10, 17-20]. However, achieving defect-free pattern transfer in this process is very crucial. Therefore, DSA, which is time and cost-efficient, is a potential alternative technique [21-25].

The physical patterning process requires the deposition of a non-magnetic material (between two consecutive magnetic islands) for the planarization of the disk, which is very crucial for the safe flight of the reading/writing head [26]. Therefore, using DSA as a replacement of other lithography techniques in the physical patterning process minimizes the usefulness of this technique. Thus, it is very essential to adopt an alternative technique for the mass fabrication of the magnetic layers.

Alternatively, the magnetic patterning process involves patterning the resist layer using DSA (considering its advantages over EBL and NIL) and then fabrication of magnetic islands using ion implantation of nonmagnetic species (refer to Figure 4.1 (b)) [27-30]. Here, the patterned structures (say pillars of resist) act as a mask, and the implanted nonmagnetic ions mostly interact with the magnetic material in the unmasked region and hence, alter the properties of the magnetic layer in the unmasked region. This process does not require the additional planarization process [31]. Ion implantation has been shown to be very useful for local modification of magnetic properties. The process of ion implantation was first studied by Chappert et al. in 1998 [32]. The structural and magnetic properties of several systems, in particular alloy systems, depend on the relative concentration of atoms [27, 33]. The ion implantation can be used to modify the relative composition and hence the structural and magnetic properties. Moreover, ion implantation can alter the atomic arrangement, interface

structure, and crystalline phase of the system [27, 34-36]. This will also lead to a change in the magnetic properties of the material. Several researchers have studied the ion implantation process for fabricating the nanopatterns [30, 37-43]. However, most of the studies used conventional lithography technique for fabricating the mask pattern [29-31, 44-48]. As stated in “Chapter 3: Experimental Techniques”, the self-assembly technique is capable of fabricating patterns as small as 3 nm, which is good enough for HDMR at 10 Tbps [2, 49, 50]. Therefore, we studied this distinctive combination of SA and ion implantation for local modification of magnetic properties. The detailed results of this study will be presented in this chapter.

Figure 4.1. The schematics of (a) physical and (b) magnetic patterning processes.

4.2. Tailoring the Structural and Magnetic Properties of Masked $\text{Co}_{80}\text{Pt}_{20}$ Thin Films Using High and Low Energy Ion Implantation

The other major aspect of HDMR technology is investigating materials with high perpendicular magnetic anisotropy [51, 52]. One such material system is a $\text{Co}_{100-x}\text{Pt}_x$ alloy, which can exhibit high magneto-crystalline anisotropy [53, 54]. However, it has been researched that the $\text{Co}_{100-x}\text{Pt}_x$ alloy is very sensitive to Pt concentration [55-57]. For $x < 25$, a highly anisotropic hexagonal closed packed (hcp) structure is favored [54-57]. On the other hand, for $x > 25$, the stacking faults and face-centered cubic (fcc) structure leads to alloy with much lower anisotropy [58, 59]. Since the structural and magnetic properties largely depend on x [59-61], we have chosen $\text{Co}_{80}\text{Pt}_{20}$ to achieve a highly anisotropic hcp structure [54, 62, 63]. Moreover, the sensitive dependence on x is also useful in our ion implantation-based patterning process. The ion implantation may alter the composition of Co and Pt by inducing a cascaded interaction of atoms and ions (interaction of atoms with foreign ions and other host atoms) and hence may result in a change of the local properties.

Figure 4.2. The process flow of sample preparation. The schematic of deposited film stack $\rightarrow$ fabrication of mask pattern on the top of film stack using self-assembly technique $\rightarrow$ ion implantation of non-magnetic species ($^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ions) through the mask (from left to right).

Figure 4.2 presents the utilized “glass substrate/ Ta (5 nm)/ Ru (15 nm)/ Co₈₀Pt₂₀ (14 nm)” stack in the present study. The Ta and Ru layers act as the underlayers to promote a perpendicular magnetic anisotropy of the Co₈₀Pt₂₀ film [64]. In addition, Ta also serves as an adhesion layer between the glass substrate and upper layers [64]. It is noticeable that the Ru layer also grows in hcp structure in certain deposition conditions and the heteroepitaxial growth of Co₈₀Pt₂₀ grains on Ru grains enhances the perpendicular magnetic anisotropy of the Co₈₀Pt₂₀ film [64]. After depositing the film stack, we fabricated a mask pattern using self-assembly of polystyrene (PS)- polydimethylsiloxane (PDMS) di-block copolymers (DBCP) for facilitating local magnetic modification through ion implantation. The details of the self-assembly process have already been discussed in chapter 3.

Figure 4.3. (a) AFM image of the dot pattern (mask) obtained using self-assembly of PS-PDMS di-block copolymers, and (b) Grain (dot) size distribution of the mask pattern.

To check the nature of the mask pattern, we scanned the sample surface using AFM. Figure 4.3 (a) shows the AFM image of the mask pattern, which confirms a good uniformity and periodicity of the dot patterns. The scanned images were then processed using “Gwyddion” and “Image J” computer software for obtaining the dot diameter and center to center spacing between the dots [65, 66]. Figure 4.3 (b) shows the histogram for the distribution of dot sizes corresponding to Figure 4.3 (a). The histogram also confirms a good uniformity in the dot sizes.

The average dot diameter was calculated to be (27 ± 4) nm (corresponding to an average dot pattern area of about 576 nm^2) with the center to center spacing of 45 nm. These numbers are quite close to what can be achieved using these DBCPs and are better than what we can achieve within the capabilities of our lab. Subsequently, the ion implantation of $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ions was carried out. The detailed ion-implantation conditions are tabulated below.

Table 4.1. Parameters used for ion-implantation of FePt-C samples.

Implanted Ion	Energy (keV)	Ion Fluence (ions/cm ²)
$^{14}\text{N}^+$	40 (high energy)	1×10^{15} , 5×10^{15} , 1×10^{16} , 2×10^{16}
	7.5 (low energy)	1×10^{15} , 5×10^{15} , 1×10^{16} , 5×10^{16}
$^{40}\text{Ar}^+$	100 (high energy)	1×10^{15} , 5×10^{15} , 1×10^{16} , 2×10^{16}
	4.5 (low energy)	1×10^{15} , 5×10^{15} , 1×10^{16} , 5×10^{16}

Both types of ions were implanted with two different energies as indicated in the table. The aim of choosing low and high energy in both cases is to study the effect of implanted ions when they stop in the magnetic layer (low energy) and when they stop in the layers below the magnetic layer (high energy). Although the high energy ions are expected to stop in the layers below the magnetic layer, they pass through the magnetic layer and may cause a significant change in structural and magnetic properties.

After magnetic patterning of the thin film samples, we characterized all the samples to probe the possible structural and magnetic changes in the samples. The corresponding results will be detailed in the upcoming sub-sections.

4.2.1. Influence of Ion Implantation on Structural Properties

4.2.1.1. X-ray Diffraction (XRD) Results

In order to probe the structural status of the implanted samples, we first performed X-ray diffraction (XRD) measurements on pristine and implanted samples. The results are presented

in Figure 4.4. The XRD pattern of the pristine sample (red curves) shows the presence of Ru (002) and Co (002) peaks at 2θ values of $\sim 42.20^\circ$ and $\sim 42.84^\circ$, respectively. This confirms the texturing of the magnetic layer as well as the underlayer in the (002) direction [67]. For high energy $^{14}\text{N}^+$ ion implanted sample, the Ru (002) peak shifts towards higher 2θ values with a maximum shift for the highest fluence of 2×10^{16} ions/cm². However, the Co (002) peak shifts towards lower 2θ values for higher fluences. As a result, both peaks merge for the fluences of 1×10^{16} and 2×10^{16} ions/cm². On the other hand, for low energy $^{14}\text{N}^+$ ion-implanted samples, both peaks shift towards lower 2θ values. Moreover, the rate of shift of Co (002) is higher, which results in the merging of both peaks for higher ion fluences of 1×10^{16} and 5×10^{16} ions/cm².

For the high energy $^{40}\text{Ar}^+$ ion implantation case, the Ru (002) peak is almost unaltered for all fluences. The Co (002) peak, however, shifts towards lower 2θ values, and the shift of this peak is much larger compared to previous cases. As a result, Co (002) appears at 2θ values smaller than that of Ru (002) peak for the fluence of 2×10^{16} ions/cm². For the low energy $^{40}\text{Ar}^+$ ion implantation case, the structural changes are minimal. Only the Co (002) peak slightly shifts towards lower 2θ values and the shift increases with the increase in ion fluence.

The results of high energy $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ion implantations may arise because of (a) intermixing of Co and Pt atoms in Ru layer and vice versa at Ru/CoPt interface [68, 69] or (b) occupation of interstitial positions by implanted ions [70] or (c) both. In the case of low energy $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ion implantations, the intermixing is not significant. This is because intermixing at Ru/CoPt interface should cause the shift of the Ru (002) peak towards higher angles, which is not observed in these cases. Therefore, the occupation of implanted ions in the interstitial positions could be a reason for these observations. It is noticeable that for low energy $^{40}\text{Ar}^+$ ion implantation, this effect is only limited to the magnetic layer while for low energy $^{14}\text{N}^+$ ion implantation the implanted ions are affecting CoPt as well as the Ru layer. To further

understand these attributions, we performed the “Transport of Ions in Matter (TRIM)” simulations and cross-sectional tunneling electron microscopy (TEM) [71].

Figure 4.4. XRD patterns of Ta/ Ru/ Co₈₀Pt₂₀ stacks upon implantation of (a) high, (b) low energy $^{14}\text{N}^+$ ions; (c) high, and (d) low energy $^{40}\text{Ar}^+$ ions.

4.2.1.2. TRIM Simulation and Cross-sectional TEM Results

In the first step, we performed TRIM simulations to estimate where ions stop in various cases of implantations. For high energy $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ion implantations, the ions stop in the substrate as shown in Figure 4.5 (top panel). Moreover, the number of ions staying in the magnetic layer is more for the case of high energy $^{40}\text{Ar}^+$ ions implantation. However, for low energy $^{14}\text{N}^+$ ion implantation, the ions are mainly staying in the magnetic layer, while some ions are also stopping ion the Ru underlayer. Different from all these cases, the ions are concentrated only in the magnetic layer for low energy $^{40}\text{Ar}^+$ ion implantation case.

Figure 4.5. TRIM Simulations for estimating the ion ranges for different cases of study.

Figure 4.6. The number density of displaced Co, Pt, and Ru host atoms due to implantation of (a) high energy $^{14}\text{N}^+$, (b) low energy $^{14}\text{N}^+$, (c) high energy $^{40}\text{Ar}^+$, and (d) low energy $^{40}\text{Ar}^+$ ions

In the next step, we performed TRIM simulations to find out the distribution of host atoms after cascaded scattering. These results are displayed in Figure 4.6 (a)- (d). There is a finite number

of Co, Pt, and Ru atoms crossing the Ru/CoPt interface in all the cases except the low energy $^{40}\text{Ar}^+$ ion implantation case. The intermixing is quite significant for high energy $^{40}\text{Ar}^+$ ion implantation compared to low and high energy $^{14}\text{N}^+$ ion implantation. Moreover, the type of species ($^{14}\text{N}^+$ or $^{40}\text{Ar}^+$) also significantly affects the observed results. In all the cases (except low energy $^{40}\text{Ar}^+$ ion implantation), the intermixing of Co atoms is more than that of Pt, which can significantly affect the composition of CoPt alloy. These results are in good agreement with our XRD results.

Figure 4.7. The cross-sectional TEM for (a) pristine, (b) high energy $^{14}\text{N}^+$ implanted sample (fluence: 2×10^{16} ions/cm²) and high energy $^{40}\text{Ar}^+$ ion implanted sample (fluence: 2×10^{16} ions/cm²).

In order to experimentally confirm the intermixing process, we performed cross-sectional tunneling electron microscopy (TEM) [72]. For these experiments, we have chosen three samples: (i) pristine sample, (ii) high energy $^{14}\text{N}^+$ implanted sample (highest fluence), and (iii) high energy $^{40}\text{Ar}^+$ implanted sample (highest fluence). The results are presented in Figure 4.7. For a pristine sample, the atoms are mostly contained in the respective layers with minimum diffusion across the interfaces. For the implanted samples, however, the atoms intermix across the interfaces with more intermixing for the $^{40}\text{Ar}^+$ ion implanted sample. Moreover, the

implanted ions are also found in all the layers of the Ta/Ru/CoPt stack. These results are in exceptionally good agreement with our predictions from simulations and hence support our explanation of the XRD results. After understanding the structural changes, we performed magnetic measurements on these samples and results will be discussed in coming sub-sections.

4.2.2. Influence of Ion Implantation on Magnetic Properties

4.2.2.1. *Magnetic Hysteresis (M-H) Loop Measurements*

To understand the influence of ion implantation on the magnetic properties of the films, the magnetic hysteresis (M-H) loop measurements were carried out first. Figure 4.8 shows the M-H loops of the pristine and $^{14}\text{N}^+$ ion-implanted (high and low energy) samples. The M-H loops of pristine and high and low energetic $^{14}\text{N}^+$ ion-implanted samples state the presence of perpendicular magnetic anisotropy (PMA) in all the samples. However, the coercivity of the pristine sample in the OOP direction is quite low (106 Oe). This is because the grains of CoPt film are highly exchange-coupled [64, 73]. A high-pressure deposition of Ru underlayer, magnetic film, and segregants is required to decouple the grains in the magnetic layer leading to high coercivity in the film [64]. However, when the sample is to be fabricated for HDNR, highly coupled media is required as the decoupling between the islands (not grains) is achieved by patterning the films. Therefore, we intentionally fabricated all the films at low pressure to achieve higher exchange coupling between the magnetic grains. In this case, nucleation and domain wall (DW) motion are the dominant modes of magnetization reversal and hence the observed coercivity is quite low [74].

Figure 4.8. Magnetic hysteresis loops of (a) pristine, (b)-(e) high energy $^{14}\text{N}^+$ implanted sample with increasing fluence (as mentioned in the figure) and (f) plot of coercivity vs ion fluence for high energy $^{14}\text{N}^+$ implantation. (g)-(j) M-H loops of low energy $^{14}\text{N}^+$ ion-implanted samples and (k) the coercivity vs ion fluence plot for low energy $^{14}\text{N}^+$ ion implantation.

For the high energy $^{14}\text{N}^+$ ion implantation case, the shape of hysteresis loops slightly changes with the increase in the ion fluence. However, the changes are not much visible in the M-H

loops as the pristine films are highly exchange-coupled. Therefore, we extracted the coercivity from the hysteresis loops and plotted them as a function of ion fluence. For the ion fluence of 1×10^{15} ions/cm², the coercivity is quite large as compared to that of the pristine sample. However, as ion fluence increases, the coercivity approaches the coercivity of the pristine sample. On the other hand, for the low energy $^{14}\text{N}^+$ ion implantation case, the implantation does not influence the magnetic properties much until a fluence of 1×10^{16} ions/cm². However, for ion fluence of 5×10^{16} ions/cm², the shape of the M-H loop changes significantly. Moreover, the coercivity is increased to 322 Oe, which is significant in comparison to other previously studied cases.

These observed results can be understood in the following manner. For low energy implantations, a greater number of foreign ions are interacting with the atoms in the magnetic layer (Figure 4.5). Moreover, the unmasked region is directly exposed to the ions and the unmasked region is almost protected from the foreign ions. The Co atoms have more chances to be displaced as they are lighter compared to Pt atoms. The incoming ions either displace the Co atoms in the lateral direction from the unmasked region to the masked region or in the Ru layer. Since the intermixing is (lesser compared to high energy $^{40}\text{Ar}^+$ ion implantation) not significant enough to result in visible changes, lateral displacement is expected to be dominant. This lateral displacement is expected more for low energy ion implantation as a larger number of $^{14}\text{N}^+$ ions are concentrated in the magnetic layer. Therefore, low energetic $^{14}\text{N}^+$ sit in the unmasked region and increase the exchange decoupling between magnetic islands in the masked region and hence result in increased coercivity.

Figure 4.9. Magnetic hysteresis loops of (a) pristine, (b)-(e) high energy $^{40}\text{Ar}^+$ implanted sample with increasing fluence (as mentioned in the figure) and (f) plot of coercivity vs ion fluence for high energy $^{40}\text{Ar}^+$ implantation. (g)-(j) M-H loops of low energy $^{40}\text{Ar}^+$ ion-implanted samples and (k) the coercivity vs ion fluence plot for low energy $^{40}\text{Ar}^+$ ion implantation.

In the next step, we performed M-H loop measurements for the $^{40}\text{Ar}^+$ ion-implanted samples as shown in Figure 4.9 (a)-(k). For high energy $^{40}\text{Ar}^+$ ion implantation, the shape of the

hysteresis loop shows significant change only for the highest fluence. The coercivity value for this case is also higher (204 Oe) compared to other $^{40}\text{Ar}^+$ ion implantation cases (refer to Figure 4.9 (f)). However, for low energy case, the changes are not quite visible. Moreover, the coercivity is almost constant as a function of ion fluence. This is because the foreign ions are concentrated only in the magnetic layer and they sit in the interstitial positions (mainly in the unmasked region). As a result, we observed a slight increase in the coercivity for all the fluences, but the increase in the fluence does not affect the process. On the other hand, for high energy implanted case, the intermixing is large. Moreover, this intermixing effect is noticeable only for the highest fluence of 2×10^{16} ions/cm². The magnetic Co and Pt atoms are pushed into the Ru layer in the unmasked region. Hence, there is exchange decoupling amongst the magnetic grains below the masked region through the nonmagnetic unmasked region. In order to further understand these attributions, we measured the domain maps for all the samples using MFM.

4.2.2.2. Magnetic Force Microscopy (MFM) Results

Figure 4.10 and 4.11 show the results from MFM scans for the pristine and $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ion-implanted CoPt samples. All the MFM images were taken in the as-prepared state (not subjected to any magnetic field), over the $5\text{ }\mu\text{m} \times 5\text{ }\mu\text{m}$ area of the samples. The MFM domain map of the pristine sample shows very clear stipe domains, which is a signature of the sample with large perpendicular magnetic anisotropy [75]. The stripe domains were observed in all the $^{14}\text{N}^+$ implanted samples suggesting that the PMA was present in all the samples. For the high energy $^{14}\text{N}^+$ ion implantation case, the domain size is increasing from a fluence of 5×10^{15} ions/cm². Inspired by these results, we measured the domain width and domain wall width for these samples. The domain, as well as domain wall width, was found to be increasing with the increase in the ions fluence. We repeated the domain and DW width measurements for low

energy $^{14}\text{N}^+$ ion implantation case as well. A more noticeable increase was observed in this case.

Like the $^{14}\text{N}^+$ implantation study, we evaluated the domain width and domain wall width for $^{40}\text{Ar}^+$ ion-implanted samples. For high and low energy implantations, the domain and DW width increase, however, the increase is not very significant in the low energy implantation case. These observations could be interpreted as follows. Firstly, ion implantation would cause a lateral straggle of the Co atoms, which will effectively increase the exchange coupling in certain regions (mostly the regions under the masks of PDMS/SiO₂) [76]. At the same time, ion implantation of high energetic particles ($^{40}\text{Ar}^+$) will decrease the perpendicular anisotropy of the magnetic film due to the changes in the relative concentration of Co and Pt. The relative changes in these properties will lead to the observation of larger magnetic domains.

For the ultrathin films, for which the domain width is much larger than the film thickness, the domain width (Δ) is proportional to [77],

$$\Delta \propto \left(d \sqrt{\frac{A}{K_u}} \right)^{\frac{1}{2}} \quad (4.1)$$

Where Δ is the domain width, A is the exchange constant K_u is the anisotropy constant and d is the thickness of the film. We have derived this formula for our films taking the inspiration from the reference [77]. This model is applicable to our films, where the thickness is about 15 nm and the domain width is 200 nm. Moreover, the domain wall width (δ) is proportional to [78],

$$\delta \propto \sqrt{\frac{A}{K_u}} \quad (4.2)$$

From these equations, the increase of domain, as well as domain wall width, can be explained on the basis of an increase in A or decrease in K_u or both. To understand the role of A and K_u on these changes, OOMMF simulations and further experimental investigations were carried out.

Figure 4.10. MFM domain maps of (a) pristine, (b)- (e) high energetic $^{14}\text{N}^+$ and (f)- (i) low energetic $^{14}\text{N}^+$ ion-implanted samples. (j) The plots of domain width and domain wall width as a function of ion fluence for all studied cases.

Figure 4.11. MFM domain maps of (a) pristine, (b)- (e) high energetic $^{40}\text{Ar}^+$ and (f)- (i) low energetic $^{40}\text{Ar}^+$ ion-implanted samples. (j) The plots of domain width and domain wall width as a function of ion fluence for all studied cases.

4.2.2.3. Micromagnetic Simulations

In order to understand the findings from the MFM domain maps, we performed micromagnetic simulations using “two dimensional NIST Object Oriented Micromagnetic Framework (OOMMF) simulation program [79]. The geometrical and magnetic parameters used during the simulations are tabulated in table 4.2. We performed simulations for various values of M_s , A , and K_u and displayed some of the key results.

Table 4.2. The geometrical and magnetic parameters used during the micromagnetic simulations.

Parameter	Value
Sample Dimensions	1 μm $\times$ 1 μm $\times$ 15 nm
Cell Size	5 nm $\times$ 5 nm $\times$ 5 nm
Saturation Magnetization (M_s)	5×10^5 – 2.5×10^6 A/m
Exchange Constant (A)	5×10^{-12} – 25×10^{-12} J/m
Anisotropy Constant (K_u)	1×10^5 – 5×10^6 J/m ³ (along (0, 0, 1), i.e. along perpendicular to film surface
Damping Constant (α)	0.15

In the first step, we performed the simulations for various values of exchange constant (A) ranging from 10 pJ/m to 50 pJ/m as shown in Figure 4.12 (a). The K_u , M_s , and α values were fixed at 5×10^5 J/m³, 5×10^5 A/m, and 0.15 respectively. As the value of A increases, the domain size as well as the DW width increases. In the next step, we fixed A , M_s , and α values at fixed at 13 pJ/m, 5×10^5 A/m, and 0.15, respectively, and varied K_u . As can be seen in Figure 4.12 (b), a decrement in the anisotropy constant results in an increase in domain width and DW width. Moreover, the smallest simulated A and largest simulated K_u (in separate simulations) result in more circular domains. Therefore, the results of MFM can be believed to be arising

from (a) an increase in A or (b) a decrease in K_u or (c) both. In order to further understand this, we estimated the anisotropy constant using the area method [80].

Figure 4.12. The depiction of domain maps obtained from micromagnetic simulations when (a) A is varied and (b) K_u is varied.

4.2.2.4. Estimation of Anisotropy Constant Using “Area Method”

In the “area method”, anisotropy constant is experimentally obtained from hysteresis loops by comparing the difference of energy “ E ” to saturate the films along IP and OOP directions [80, 81], where

$$E = \int_0^M H(M)dM \quad (4.3)$$

Figure 4.13 summarizes the anisotropy constant values for different cases. The anisotropy constant for as-deposited samples is $4.32 \times 10^5 \text{ J/m}^3$. All of the implanted samples except the high-energy $^{40}\text{Ar}^+$ ion implanted sample, show the anisotropy constant value very close to that of the pristine sample. However, the estimated value of K_u for high energy $^{40}\text{Ar}^+$ ion implanted sample is $6.5 \times 10^4 \text{ J/m}^3$. Therefore, the decrease in K_u is the dominant cause for the increasing

trend of the domain and DW width for high energy $^{40}\text{Ar}^+$ ion implantations. In other cases, the increase in A is believed to be the dominant cause of the observed results. Although the changes in A can be found out using first-order reversal curves (FORC), these films are highly exchange coupled and the changes observed in exchange coupling may not be that clear in FORC as in previous studies [76].

Figure 4.13. Estimated anisotropy constant values for different samples as mentioned.

4.2.3. Summary of CoPt samples

In this set of investigations, we have studied the effect of high and low energy $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ion implantation on masked $\text{Co}_{80}\text{Pt}_{20}$ films. In order to achieve masked implantation, we fabricated the mask pattern using self-assembly of PS-PDMS DBCP. XRD results reveal a significant change in the structural properties for the highest ion fluence in all the implantation cases except low energy $^{40}\text{Ar}^+$ ion implantations. This is explained to be arising due to the mixing of Co and Pt atoms in the Ru layer and vice versa and occupancy of foreign atoms in the interstitial positions. These results were further supported by TRIM simulations and cross-sectional TEM. The lower fluences do not show noticeable changes. The structural changes were also translated to magnetic properties and significant changes were observed. However, the changes are more for the case of high energy $^{40}\text{Ar}^+$ and low energy $^{14}\text{N}^+$ ion implantations than remaining. MFM and hysteresis loop measurement results indicate that the observed results might be arising from an increase in A or a decrease in K_u or both. These results were further confirmed by micromagnetic simulations.

4.3. Ion Implantation Induced Disorder in FePt-C Thin Films

The $L1_0$ face-centered tetragonal (fct) FePt binary alloy is the most suitable candidate for achieving recording media with a high perpendicular magnetic anisotropy (PMA) [82-85]. A high magnetic anisotropy value of the order of $\sim 10^7$ erg/cc can be obtained for $L1_0$ fct-FePt alloys [86-88]. Such a high anisotropy can result in stable grains of size as small as 3 nm in ~ 10 nm thick magnetic films [89-91]. In A-B binary alloys, the equilibrium between A-A, B-B, and A-B interactions (of the order of 1 eV) results in a certain phase [33, 92]. In particular, for FePt binary system, the competition among, (a) formation enthalpy of the ordered system, (b) mixing enthalpy of random alloy, and (c) strain energy from atomic size mismatch stabilizes a certain phase. Therefore, a relative atomic composition and heat treatment during or after the deposition decide a particular phase in the FePt system. The $L1_0$ fct-phase occurs when the atomic concentrations of Fe and Pt atoms are 50% each [86]. Hence, we have chosen equal concentrations of Fe and Pt during the deposition of our films. Moreover, when $\text{Fe}_{50}\text{Pt}_{50}$ films are deposited at room temperature, a disordered face-centered cubic (fcc) phase is achieved [86]. Therefore, heating at a high temperature is required to get the $L1_0$ fct-phase in $\text{Fe}_{50}\text{Pt}_{50}$ films.

Note, equiatomic FePt is the premier choice of recording material for HAMR technology [93]. In the past, several seed layers and segregants (in the recording layer) have been investigated for achieving high magnetocrystalline anisotropy and smaller grain sizes in $L1_0$ fct FePt system [88-91, 94-104]. These studies indicate that the typical FePt media design consists of “substrate/ NiTa (20-100 nm)/ heat sink layer (20-200 nm)/ MgO (5-20 nm)/ FePtX-Y (6-12 nm). As mentioned earlier, one of the major requirements for HDMR media is the recording material with high magneto-crystalline anisotropy. Therefore, we have utilized a similar material stack for studying the effects of magnetic patterning for HDMR applications.

The thin-film stack “glass substrate/ NiTa (35 nm)/MgO/ (8 nm)/ FePt-C (8 nm)” was deposited using DC/RF magnetron sputtering process. The samples were annealed at 600 °C before depositing the FePt-C layer. Moreover, the purpose of using MgO was to promote perpendicular magnetic anisotropy in the magnetic layer [105, 106]. Subsequently, we fabricated mask patterns on the top of the film stack for facilitating local magnetic modification through ion implantation. The mask pattern was deposited using the self-assembly process (the process and the parameters are the same as used for fabricating mask patterns for Co₈₀Pt₂₀ films, discussed in the previous sub-chapter). Figure 4.14 shows the schematic of the sample preparation scheme.

Figure 4.14. The process flow of creating local changes in the structural and magnetic properties.

Figure 4.15. (a) AFM image of the dotted mask on the top of FePt-C samples and (b) dot size distribution of the mask pattern (inset: 3D version of the AFM image).

To examine the quality of the mask pattern, we performed the AFM of the samples. Figure 4.15 (a) shows the corresponding image of the self-assembled dot pattern created over the films. From the AFM observations, an average dot diameter of (20.5 ± 2.7) nm was estimated. Moreover, the center-to-center distance of 28.3 nm was obtained using the FFT of image 4.15 (a). After obtaining the samples with mask patterns on the top, the ion implantation of $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ions were carried out. Considering the results in the study with the $\text{Co}_{80}\text{Pt}_{20}$ magnetic layer, we have chosen only the highest fluence available in all the studied cases. The details of ion implantation conditions are provided in the following table.

Table 4.3. Parameters used for ion-implantation of FePt-C samples.

Implanted Ion	Energy (keV)	Ion Fluence (ions/cm ²)
$^{14}\text{N}^+$	40 (high energy)	2×10^{16}
	7.5 (low energy)	5×10^{16}
$^{40}\text{Ar}^+$	100 (high energy)	2×10^{16}
	4.5 (low energy)	5×10^{16}

4.3.1. Influence of Ion Implantation on Structural Properties

4.3.1.1. X-ray Diffraction (XRD) Results

In the first step, we performed XRD experiments on pristine (i.e. sample before implantation), high and low energy $^{14}\text{N}^+$, and $^{40}\text{Ar}^+$ implanted samples. The results are presented in Figure 4.16. The presence of MgO (002) peak in the XRD scans of pristine sample confirms the strong texture along (002) direction in the MgO layer. The presence of the MgO (002) peak is crucial as this peak promotes the texture of the magnetic layer in the perpendicular direction (with respect to film plane), important for large PMA in the magnetic layer [107, 108]. Moreover, the presence of superlattice FePt (001) peak in the XRD patterns confirms the epitaxial growth of FePt-C films on the MgO films with $L1_0$ fct ordered phase [107, 108]. However, the presence

of FePt (111) peak in the pristine sample suggests the simultaneous presence of a chemically disordered fcc phase [107, 108]. This indicates the growth of our samples is not comparable to high-quality HAMR FePt-C media fabricated in the industry. Although the pristine sample is partially ordered, they serve the purpose of this academic study.

For, the high energy $^{14}\text{N}^+$ implanted sample, the FePt (111) peak shifts towards lower 2θ values, suggesting the increase in the interplanar spacing d corresponding to FePt (111) peak [109]. This could be because of the occupancy of $^{14}\text{N}^+$ ions in the interstitial positions. Moreover, the MgO (002) peak has disappeared, indicating the lowering of (002) texture in the MgO layer. This could be attributed to damage of the MgO layer or interlayer diffusion of the materials at the MgO/FePt-C interface. The presence of a FePt (001) peak confirms that the partially ordered fct phase is still present in this sample. However, for the low energy $^{14}\text{N}^+$ implantation case, the shift in the FePt (111) peak towards the lower angle is larger compared to the high energy case. The other peaks are present with minimal effect of ion implantation for this case. A larger number of ions are possibly occupying the interstitial positions in this case. This also suggests that low energy $^{14}\text{N}^+$ ions are largely stopping in the FePt-C layer. While high energy $^{14}\text{N}^+$ ions are affecting the MgO layer in addition to the FePt-C layer.

On the other hand, for high and low energy $^{40}\text{Ar}^+$ ion-implantation case, the absence of FePt (001) peak suggests the degradation of partially ordered phase of FePt-C layer [110]. Moreover, the MgO (002) and FePt (111) peaks shift towards lower angles with more shift in the high energy $^{40}\text{Ar}^+$ implantation case. These results also hint that the low energy $^{40}\text{Ar}^+$ ions mostly stay in the magnetic layer while high energy $^{40}\text{Ar}^+$ ions pass through the magnetic layer (like the $^{14}\text{N}^+$ ion-implantation case). These results can be further understood from the TRIM simulation results [71].

Figure 4.16. XRD patterns of FePt-C pristine and ion-implanted samples.

4.3.1.2. TRIM Simulation Results

Figure 4.17 shows the distribution of implanted ions for all cases of study. For high energy $^{14}\text{N}^+$ as well as $^{40}\text{Ar}^+$ ion implantation, the ions stop in the glass substrate. Moreover, the concentration of ions stopping in the magnetic layer is less than that of stopping in the layer below the magnetic layer. However, for low energy $^{14}\text{N}^+$ implantation, the ions partially stop in the magnetic layer. Moreover, the concentration of ions stopping in the magnetic layer is much larger than high energy implantation cases. For low energy $^{40}\text{Ar}^+$ ion implantation, the ions mostly stop in the magnetic layer, though the concentration is smaller than the low energy $^{14}\text{N}^+$ implantation case. These results are in good agreement with our speculations from the XRD results.

Figure 4.17. The distribution of implanted ions (Δn) for different studied cases as shown in the figure.

In addition to the concentration of interacting foreign ions, the corresponding momentum associated with traveling ions also affects the host atoms. When the ions are implanted in the film (or film stack), the maximum kinetic energy E_p transferred from foreign ion to the atom in the film is given by [111],

$$E_p = \frac{4MmE}{(M+m)^2} \quad (4.4)$$

Here M , m , and E are the atomic weight of the displaced atom in the film, atomic weight of implanted ion, and the energy of implanted ion, respectively.

Table 4.4. *Estimated interaction parameters for different cases of the present study.*

Case	Ions stopping in FePt-C Layer (%)	Energy transferred to Fe atoms (keV)	Energy transferred to Pt atoms (keV)
High energy $^{14}\text{N}^+$	3	25.6	10
Low energy $^{14}\text{N}^+$	20	4.8	1.9
High energy $^{40}\text{Ar}^+$	3	97.3	56.5
Low energy $^{40}\text{Ar}^+$	85	4.4	2.5

The changes in structural properties can be qualitatively explained in the following manner. Using TRIM simulations, we estimated the fraction of ions stopping in the FePt-C layer. Moreover, using Eq. 4.4, we calculated the amount of energy transferred to the Fe and Pt atoms in various studied cases. These estimated numbers are summarized in table 4.4. For low energy $^{40}\text{Ar}^+$ ion-implantation, 85% of foreign ions are stopping in the magnetic layer. However, the total energy transferred to Fe and Pt atoms is quite low. Although they are transferring ~98% and ~56% of the total energy to Fe and Pt atoms respectively, the energy of incident ions itself is quite low (4.5 keV). Therefore, they may sit in interstitial positions and enlarge the interplanar spacing for fct (001) peak and result in the conversion from fct to fcc structure. Similarly, for the high energy $^{40}\text{Ar}^+$ ion-implantation case, the energy transferred to the host Fe and Pt atoms are highest, however, only 3% of the total incident ions are interacting with Fe and Pt atoms. This could lead to similar observations. Although the energy transferred to Fe and Pt atoms for low energy $^{14}\text{N}^+$ ions is comparable to low energy $^{40}\text{Ar}^+$ ions, only 20% of implanted ions are interacting with host Fe and Pt atoms. Moreover, the size (related to atomic weight) of $^{14}\text{N}^+$ ions is smaller, so the effect of an increase in interplanar spacing is more relevant for fcc (111) peak. On the other hand, for high-energy $^{14}\text{N}^+$ ions, the fraction of implanted ions interacting with Fe and Pt atoms are the lowest (like high energy $^{40}\text{Ar}^+$ ions). In addition, the energy transferred is much smaller compared to high-energy $^{40}\text{Ar}^+$ ion implantations. Thus, these are showing the least effects on the FePt-C magnetic layer.

Figure 4.18. The distribution of host atoms after cascaded scattering (Δm) for different studied cases as indicated.

We have also performed TRIM simulations to observe the distribution of host atoms after cascaded scattering in all the cases. These results are summarized in Figure 4.18. From these results, one can see that the maximum intermixing of atoms across the interfaces of FePt-C/MgO and MgO/NiTa happened for the high energy $^{14}\text{N}^+$ implantation case. Moreover, except for low energy $^{40}\text{Ar}^+$ ion implantation, a sufficient number of implanted ions are staying in the MgO layer. Therefore, the MgO (002) peak is least unmoved for low energy $^{40}\text{Ar}^+$ ion implantation, and maximum effects were recorded for high energy $^{14}\text{N}^+$ implantations. The results of TRIM simulations and theoretical calculations are in line with our observations in XRD results.

4.3.2. Influence of Ion Implantation on Magnetic Properties

4.3.2.1. *Magnetic Hysteresis (M-H) Loop Measurements*

To further understand the results, we performed magnetic hysteresis loop measurements for pristine and $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ion-implanted samples (Figure 4.19). The M-H loops of the pristine samples indicate that the coercivity in the OOP direction is quite large (5.2 kOe), which confirms the granular nature of the samples [112]. Moreover, it also suggests a large magnetic anisotropy of the sample. However, the remanence is almost the same in both OOP as well as IP directions as shown in Figure 4.19 (a). Moreover, the coercivity in the IP direction (4.8 kOe) is also comparable to that in the OOP direction. This suggests the simultaneous presence of the soft fcc phase in the sample. In other words, there are some magnetic moments pointing slightly away from the OOP direction present in the system. Further, the observed kink at zero magnetic fields also confirms this attribution [113-115].

Next, we measured the M-H loops for the implanted samples as shown in Figure 4.19 (b) to (e). The remanence in the IP direction was larger than that in the OOP direction in all the cases except the low energy $^{40}\text{Ar}^+$ ion implanted sample. This suggests the change of easy axis from OOP direction to IP direction [116]. Moreover, except for the low energy $^{40}\text{Ar}^+$ ion implanted sample, the coercivity in both directions has decreased a lot compared to the pristine sample, indicating that there might be a reduction in the anisotropy of the samples. Moreover, the saturation magnetization has also decreased in all the cases, which could be because of the intermixing of atoms across the FePt-C/MgO interface. However, for the $^{40}\text{Ar}^+$ ion implantation case, the effects are lesser. Thus, the lateral (dominantly) and vertical displacement of Fe atoms can be a reason for the decreased coercivity of the implanted samples. Figure 4.20 shows the coercivity, remanence, and saturation magnetization for pristine and implanted samples.

Figure 4.19. The magnetic hysteresis loops of FePt-C pristine and ion-implanted samples.

Figure 4.20. The variation of saturation magnetization (M_s), IP and OOP coercivity (H_c), and IP and OOP remanence (M_r) for FePt-C pristine and ion-implanted samples.

4.3.2.2. Torque Magnetometry Results

Subsequently, we measured the torque curves using torque magnetometry for estimating the anisotropy constant of the pristine and implanted samples [117-119]. Figure 4.21 (a)–(e) shows the torque curves of the pristine and ion-implanted samples. The pristine sample exhibits the periodicity of 180° of peaks in the torque curve, which signifies the presence of uniaxial anisotropy in the system [120]. Moreover, the appearance of the first peak in the first quadrant suggests that the OOP direction is the easy axis for this sample [120]. The occurrence of other less intense peaks in the torque curve of the pristine sample is the signature of the fact that there are some atomic moments, which are away from the easy axis (OOP) direction. These results are in good agreement with our results from M-H loops of the pristine sample.

Figure 4.21. The torque curves of (a) pristine, (b) high energy $^{14}\text{N}^+$, (c) low energy $^{14}\text{N}^+$, (d) high energy $^{40}\text{Ar}^+$ and (e) low energy $^{40}\text{Ar}^+$ ion implanted samples.

Except for the low energy $^{40}\text{Ar}^+$ ion-implanted sample, all the other implanted samples show the appearance of the first peak in the fourth quadrant. However, the periodicity of 180° of the peaks is maintained. This means that the uniaxial anisotropy is still present, but the direction of the easy axis has changed from OOP to IP direction. Also, the torque curves do not have any other less intense peaks except for the low energy $^{14}\text{N}^+$ ion-implantation case, which means

that most of the magnetic moments are in the IP direction. It is important to note that the amplitude of the peaks, which is related to the anisotropy constant value of these samples, is different for all of the cases. For the low energy $^{40}\text{Ar}^+$ ion-implantation case, the torque curves have a periodicity of 180° , the first peak is in the first quadrant and the intensity of peaks corresponding to atomic moments away from the easy axis direction has diminished. This implies that the relative number of atomic moments oriented in the easy axis (OOP direction) has increased.

Figure 4.22 shows the graph of anisotropy constant for the pristine and implanted samples. For FePt pristine and low energy $^{40}\text{Ar}^+$ ion-implanted samples, a hysteretic nature was seen in the torque curves, which means that 25 kOe (maximum available field for experimental setup) is not sufficient to saturate the samples. Thus, the value of the anisotropy constant for these two samples may be a little different from the estimated value. The samples having IP easy axis direction has a negative value of the anisotropy constant, which is a well-known fact in the magnetism community [121].

Figure 4.22. The bar chart showing the change in anisotropy constant (K_u), estimated from torque measurements, for FePt-C pristine and implanted samples.

Figure 4.23. The MFM scans of (a) pristine, (b) high energy $^{14}\text{N}^+$, (c) low energy $^{14}\text{N}^+$, (d) high energy $^{40}\text{Ar}^+$ and (e) low energy $^{40}\text{Ar}^+$ ion implanted samples.

4.3.2.3. Magnetic Force Microscopy (MFM) Results

In order to visually observe the effects of ion implantation on the domain patterns of FePt-C samples, we scanned an area of $5\ \mu\text{m} \times 5\ \mu\text{m}$ for all the samples using MFM. These measurements were performed at zero external magnetic fields. The results are shown in Figure 4.23 (a)- (e). The MFM image of the pristine samples (Figure 4.23 (a)) shows domains with a clear contrast, which is a characteristic feature of magnetic films with PMA [36, 122, 123]. However, for the ion-implanted case, the contrast of the domain pattern has diminished in all

the cases, except for low energy $^{40}\text{Ar}^+$ ion implantation samples. This indicates that the PMA is still present in this sample [124]. However, all the other samples have turned to IP anisotropic. These results are in agreement with our previous results.

4.4. Conclusions

In this chapter, the effect of ion implantation on masked CoPt and FePt samples was studied. The results of the CoPt samples were summarized in section 4.2.3. However, the results of ion implantation on the masked FePt-C magnetic thin films are summarized here. The reference sample consists of fct hard and fcc soft phases. For the $^{14}\text{N}^+$ ion-implantation case, the FePt (111) peak shifts toward low 2θ values, which leads to an increase in the inter-planar spacing for the planes corresponding to FePt (111) peak. We also observed a decrease in coercivity (both in IP and OOP directions), remanence (both in IP and OOP directions), and saturation magnetization, which is later explained due to a decrease in the anisotropy constant of the samples. On the other hand, after $^{40}\text{Ar}^+$ ion implantation, the FePt (001) phase is absent, which indicates the transformation into fcc soft phase. Also, coercivity and saturation magnetization has decreased. This can be due to occupancy of implanted ions in the interstitial positions and also the displacement of Fe atoms in the lateral direction (dominantly) and across the FePt-C and MgO interface and leading to the decrease in the anisotropy constant. This attribution is well supported by the torque curves and MFM domain maps. Although the effects of ion implantation of two different species are quite reasonable, the mask pattern seems to be less effective while stopping the incoming ions. Nevertheless, this study unveils many new results and presents a method to alter the structural and magnetic properties of the equiatomic FePt system.

References

- [1] A. Kikitsu, Y. Kamata, M. Sakurai and K. Naito, "Recent progress of patterned media," *IEEE Transactions on Magnetics*, vol. 43, no. 9, pp. 3685, 2007.
- [2] A. Kikitsu, "Prospects for bit patterned media for high-density magnetic recording," *Journal of Magnetism and Magnetic Materials*, vol. 321, no. 6, pp. 526, 2009.
- [3] D. S. Kuo, K. Y. Lee, X. Yang, S. Xiao, Y. Hsu, Z. Yu, M. Feldbaum, T. Klemmer, Y. Kubota, J. Thiele, P. Steiner, K. Wago, S. Dallorto and D. Olynick, "Heated Dot Magnetic Recording Media- Path to 10 Tdots/in²," in *2016 International Conference of Asian Union of Magnetics Societies (ICAUMS)*, pp. 1, 2016.
- [4] M. T. Kief and R. H. Victora, "Materials for heat-assisted magnetic recording," *MRS Bulletin*, vol. 43, no. 2, pp. 87, 2018.
- [5] D. Wachenschwanz, W. Jiang, E. Roddick, A. Homola, P. Dorsey, B. Harper, D. Treves and C. Bajorek, "Design of a manufacturable discrete track recording medium," *IEEE Transactions on Magnetics*, vol. 41, no. 2, pp. 670, 2005.
- [6] K. Hattori, K. Ito, Y. Soeno, M. Takai and M. Matsuzaki, "Fabrication of discrete track perpendicular media for high recording density," *IEEE transactions on magnetics*, vol. 40, no. 4, pp. 2510, 2004.
- [7] B. D. Terris, T. Thomson and G. Hu, "Patterned media for future magnetic data storage," *Microsystem Technologies*, vol. 13, no. 2, pp. 189, 2007.
- [8] T. R. Albrecht, H. Arora, V. Ayanoor-Vitikkate, J. Beaujour, D. Bedau, D. Berman, A. L. Bogdanov, Y. Chapuis, J. Cushen, E. E. Dobisz, G. Doerk, H. Gao, M. Grobis, B. Gurney, W. Hanson, O. Hellwig, T. Hirano, P. Jubert, D. Kercher, J. Lille, Z. Liu, C. M. Mate, Y. Obukhov, K. C. Patel, K. Rubin, R. Ruiz, M. Schabes, L. Wan, D. Weller, T. Wu and E. Yang, "Bit-Patterned Magnetic Recording: Theory, Media Fabrication, and Recording Performance," *IEEE Transactions on Magnetics*, vol. 51, no. 5, pp. 1, 2015.
- [9] B. D. Terris, "Fabrication challenges for patterned recording media," *Journal of Magnetism and Magnetic Materials*, vol. 321, no. 6, pp. 512, 2009.
- [10] R. Sbiaa and S. N. Piramanayagam, "Patterned media towards nano-bit magnetic recording: Fabrication and challenges," *Recent patents on nanotechnology*, vol. 1, no. 1, pp. 29, 2007.
- [11] R. L. White, R. M. H. Newt and R. F. W. Pease, "Patterned media: a viable route to 50 Gbit/in/sup 2/ and up for magnetic recording?," *IEEE Transactions on Magnetics*, vol. 33, no. 1, pp. 990, 1997.
- [12] R. Ruiz, H. Kang, F. A. Detcheverry, E. Dobisz, D. S. Kercher, T. R. Albrecht, J. J. de Pablo and P. F. Nealey, "Density multiplication and improved lithography by directed block copolymer assembly," *Science*, vol. 321, no. 5891, pp. 936, 2008.
- [13] X. Yang, S. Xiao, W. Wu, Y. Xu, K. Mountfield, R. Rottmayer, K. Lee, D. Kuo and D. Weller, "Challenges in 1 Teradot/in.² dot patterning using electron beam lithography for bit-patterned media," *Journal of Vacuum Science & Technology B: Microelectronics and Nanometer Structures Processing, Measurement, and Phenomena*, vol. 25, no. 6, pp. 2202, 2007.
- [14] J. K. W. Yang, Y. Chen, T. Huang, H. Duan, N. Thiyagarajah, H. K. Hui, S. H. Leong and V. Ng, "Fabrication and characterization of bit-patterned media beyond 1.5 Tbit/in²," *Nanotechnology*, vol. 22, no. 38, p. 385301, 2011.
- [15] I.-B. Baek, J.-H. Yang, W.-J. Cho, C.-G. Ahn, K. Im and S. Lee, "Electron beam lithography patterning of sub-10 nm line using hydrogen silsesquioxane for nanoscale device applications," *Journal of Vacuum Science & Technology B: Microelectronics*

- and Nanometer Structures Processing, Measurement, and Phenomena, vol. 23, no. 6, pp. 3120, 2005.
- [16] N. GAUR, "Investigations on advanced magnetic recording media," Ph.D. Thesis, 2012.
 - [17] P. Candeloro, A. Gerardino, E. Di Fabrizio, S. Cabrini, G. Giannini, L. Mastrogiacomo, M. Ciria, R. C. O'Handley, G. Gubbiotti and G. Carlotti, "Patterned magnetic permalloy and nickel films: fabrication by electron beam and x-ray lithographic techniques," *Japanese journal of applied physics*, vol. 41, no. 8R, p. 5149, 2002.
 - [18] X. Yang, S. Xiao, W. Hu, J. Hwu, R. van de Veerdonk, K. Wago, K. Lee and D. Kuo, "Integration of nanoimprint lithography with block copolymer directed self-assembly for fabrication of a sub-20 nm template for bit-patterned media," *Nanotechnology*, vol. 25, no. 39, p. 395301, 2014.
 - [19] M. Malloy and L. Litt, "Technology review and assessment of nanoimprint lithography for semiconductor and patterned media manufacturing," *Journal of Micro/Nanolithography, MEMS, and MOEMS*, vol. 10, no. 3, p. 032001, 2011.
 - [20] X. Yang, Y. Xu, C. Seiler, L. Wan and S. Xiao, "Toward 1Tdot/in.² nanoimprint lithography for magnetic bit-patterned media: Opportunities and challenges," *Journal of Vacuum Science & Technology B: Microelectronics and Nanometer Structures Processing, Measurement, and Phenomena*, vol. 26, no. 6, pp. 2604, 2008.
 - [21] J. K. W. Yang, B. Cord, H. Duan, K. K. Berggren, J. Klingfus, S.-W. Nam, K.-B. Kim and M. J. Rooks, "Understanding of hydrogen silsesquioxane electron resist for sub-5-nm-half-pitch lithography," *Journal of Vacuum Science & Technology B: Microelectronics and Nanometer Structures Processing, Measurement, and Phenomena*, vol. 27, no. 6, pp. 2622, 2009.
 - [22] S. Park, D. H. Lee, J. Xu, B. Kim, S. W. Hong, U. Jeong, T. Xu and T. P. Russell, "Macroscopic 10-Terabit-per-Square-Inch Arrays from Block Copolymers with Lateral Order," *Science*, vol. 323, no. 5917, p. 1030, 2009.
 - [23] T. P. Bigioni, X.-M. Lin, T. T. Nguyen, E. I. Corwin, T. A. Witten and H. M. Jaeger, "Kinetically driven self assembly of highly ordered nanoparticle monolayers," *Nature Materials*, vol. 5, no. 4, pp. 265, 2006.
 - [24] W. F. van Dorp, B. van Someren, C. W. Hagen, P. Kruit and P. A. Crozier, "Approaching the Resolution Limit of Nanometer-Scale Electron Beam-Induced Deposition," *Nano Letters*, vol. 5, no. 7, pp. 1303, 2005.
 - [25] S. Xiao, X. Yang, S. Park, D. Weller and T. P. Russell, "A Novel Approach to Addressable 4 Teradot/in.² Patterned Media," *Advanced Materials*, vol. 21, no. 24, pp. 2516, 2009.
 - [26] C. Choi, Y. Yoon, D. Hong, Y. Oh, F. E. Talke and S. Jin, "Planarization of patterned magnetic recording media to enable head flyability," *Microsystem technologies*, vol. 17, no. 3, pp. 395, 2011.
 - [27] J. Fassbender and J. McCord, "Magnetic patterning by means of ion irradiation and implantation," *Journal of Magnetism and Magnetic Materials*, vol. 320, no. 3-4, pp. 579, 2008.
 - [28] D. L. Kuo, Xinwei; Gui, Jing, "Patterning longitudinal magnetic recording media with ion implantation," *United States Patent 6864042*, 2005.
 - [29] M. S. Martín-González, F. Briones, J. M. García-Martín, J. Montserrat, L. Vila, G. Faini, A. M. Testa, D. Fiorani and H. Rohrmann, "Nano-patterning of perpendicular magnetic recording media by low-energy implantation of chemically reactive ions," *Journal of Magnetism and Magnetic Materials*, vol. 322, no. 18, pp. 2762, 2010.
 - [30] K. Sato, A. Ajan, N. Aoyama, T. Tanaka, Y. Miyaguchi, K. Tsumagari, T. Morita, T. Nishihashi, A. Tanaka and T. Uzumaki, "Magnetization suppression in Co/Pd and

- CoCrPt by nitrogen ion implantation for bit patterned media fabrication," *Journal of Applied Physics*, vol. 107, no. 12, p. 123910, 2010.
- [31] T. Kato, D. Oshima and S. Iwata, "Ion Irradiation for Planar Patterning of Magnetic Materials," *Crystals*, vol. 9, no. 1, Jan 2019, Art no. 27, doi: 10.3390/cryst9010027.
 - [32] C. Chappert, H. Bernas, J. Ferré, V. Kottler, J.-P. Jamet, Y. Chen, E. Cambril, T. Devolder, F. Rousseaux and V. Mathet, "Planar patterned magnetic media obtained by ion irradiation," *Science*, vol. 280, no. 5371, pp. 1919, 1998.
 - [33] O. Gutfleisch, J. Lyubina, K. H. Müller and L. Schultz, "FePt hard magnets," *Advanced Engineering Materials*, vol. 7, no. 4, pp. 208, 2005.
 - [34] N. Gaur, K. Pandey, S. Maurer, S. Piramanayagam, R. Nunes, H. Yang and C. Bhatia, "Magnetic and structural properties of CoCrPt–SiO₂-based graded media prepared by ion implantation," *Journal of Applied Physics*, vol. 110, no. 8, p. 083917, 2011.
 - [35] N. Gaur, S. Piramanayagam, S. Maurer, R. Nunes, S. Steen, H. Yang and C. Bhatia, "Ion implantation induced modification of structural and magnetic properties of perpendicular media," *Journal of Physics D: Applied Physics*, vol. 44, no. 36, p. 365001, 2011.
 - [36] N. Gaur, S. Kundu, S. Piramanayagam, S. Maurer, H. Tan, S. Wong, S. Steen, H. Yang and C. Bhatia, "Lateral displacement induced disorder in L1 0-FePt nanostructures by ion-implantation," *Scientific reports*, vol. 3, p. 1907, 2013.
 - [37] B. Terris, M. Albrecht, G. Hu, T. Thomson and C. Rettner, "Recording and reversal properties of nanofabricated magnetic islands," *IEEE transactions on magnetics*, vol. 41, no. 10, pp. 2822, 2005.
 - [38] Y. Chen, T. Huang, S. Leong, S. Hu, K. Ng, Z. Yuan, B. Zong, B. Liu and V. Ng, "A study of multirow-per-track bit patterned media by spinstand testing and magnetic force microscopy," *Applied Physics Letters*, vol. 93, no. 10, p. 102501, 2008.
 - [39] L. Ding, T. Chen, Y. Liu, M. Yang, J. Wong, K. Liu, F. Zhu and S. Fung, "The influence of the implantation dose and energy on the electroluminescence of Si⁺-implanted amorphous SiO₂ thin films," *Nanotechnology*, vol. 18, no. 45, p. 455306, 2007.
 - [40] C. Flynn, M. Ondrejcek and W. Swiech, "Synthesis of pans and mesas using a beam of self-ions," *Journal of Physics: Condensed Matter*, vol. 20, no. 39, p. 395001, 2008.
 - [41] C. White, S. Withrow, K. Sorge, A. Meldrum, J. Budai, J. Thompson and L. Boatner, "Oriented ferromagnetic Fe-Pt alloy nanoparticles produced in Al₂O₃ by ion-beam synthesis," *Journal of applied physics*, vol. 93, no. 9, pp. 5656, 2003.
 - [42] C. White, S. Withrow, J. Budai, L. Boatner, K. Sorge, J. Thompson, K. Beaty and A. Meldrum, "Ferromagnetic FePt nanoparticles formed in Al₂O₃ by ion implantation," *Nuclear Instruments and Methods in Physics Research Section B: Beam Interactions with Materials and Atoms*, vol. 191, no. 1-4, pp. 437, 2002.
 - [43] S. Y. Chou, M. S. Wei, P. R. Krauss and P. B. Fischer, "Single-domain magnetic pillar array of 35 nm diameter and 65 Gbits/in.² density for ultrahigh density quantum magnetic storage," *Journal of Applied Physics*, vol. 76, no. 10, pp. 6673, 1994.
 - [44] C. Chappert, H. Bernas, J. Ferré, V. Kottler, J. P. Jamet, Y. Chen, E. Cambril, T. Devolder, F. Rousseaux, V. Mathet and H. Launois, "Planar Patterned Magnetic Media Obtained by Ion Irradiation," *Science*, vol. 280, no. 5371, p. 1919, 1998.
 - [45] H. Bernas, "Planar Patterned Magnetic Media Obtained by Ion Irradiation," APS, p. V22-001, 2001.
 - [46] J. Ferré, C. Chappert, H. Bernas, J. P. Jamet, P. Meyer, O. Kaitasov, S. Lemerle, V. Mathet, F. Rousseaux and H. Launois, "Irradiation induced effects on magnetic properties of Pt/Co/Pt ultrathin films," *Journal of Magnetism and Magnetic Materials*, vol. 198-199, pp. 191, 1999.

- [47] E. Suharyadi, T. Kato, S. Tsunashima and S. Iwata, "Magnetic Properties of Patterned Co/Pd Nanostructures by E-Beam Lithography and Ga Ion Irradiation," *IEEE Transactions on Magnetics*, vol. 42, no. 10, pp. 2972, 2006.
- [48] D. Oshima, T. Kato and S. Iwata, "Switching Field Distribution of MnGa Bit Patterned Film Fabricated by Ion Beam Irradiation," *IEEE Transactions on Magnetics*, vol. 54, no. 2, pp. 1, 2018.
- [49] C. Vogler, C. Abert, F. Bruckner, D. Suess and D. Praetorius, "Heat-assisted magnetic recording of bit-patterned media beyond 10 Tb/in²," *Applied Physics Letters*, vol. 108, no. 10, p. 102406, 2016.
- [50] A. Venugopal, A. Ghoreyeshi and R. H. Victora, "High-density shingled heat-assisted recording using bit-patterned media subject to track misregistration," *IEEE Transactions on Magnetics*, vol. 53, no. 11, pp. 1, 2017.
- [51] K. Hono, Y. Takahashi, G. Ju, J.-U. Thiele, A. Ajan, X. Yang, R. Ruiz and L. Wan, "Heat-assisted magnetic recording media materials," *MRS Bulletin*, vol. 43, no. 2, p. 93, 2018.
- [52] M. H. Kryder, E. C. Gage, T. W. McDaniel, W. A. Challener, R. E. Rottmayer, G. Ju, Y.-T. Hsia and M. F. Erden, "Heat assisted magnetic recording," *Proceedings of the IEEE*, vol. 96, no. 11, pp. 1810, 2008.
- [53] D. J. Sellmyer and R. Skomski, *Advanced magnetic nanostructures*. Springer Science & Business Media, 2006.
- [54] Y. Yamada, T. Suzuki, H. Kanazawa and J. C. Österman, "The origin of the large perpendicular magnetic anisotropy in Co₃Pt alloy thin films," *Journal of Applied Physics*, vol. 85, no. 8, pp. 5094, 1999.
- [55] G. Harp, D. Weller, T. Rabedeau, R. Farrow and M. Toney, "Magneto-optical Kerr spectroscopy of a new chemically ordered alloy: Co₃Pt," *Physical review letters*, vol. 71, no. 15, p. 2493, 1993.
- [56] Y. Yamada, W. Van Drent, E. Abarra and T. Suzuki, "High perpendicular anisotropy and magneto-optical activities in ordered Co₃Pt alloy films," *Journal of applied physics*, vol. 83, no. 11, pp. 6527, 1998.
- [57] R. Bandhu, R. Sooryakumar, R. C. Farrow, D. Weller, M. Toney and T. Rabedeau, "Elastic properties of chemically ordered Co₃Pt thin films," *Journal of applied physics*, vol. 91, no. 5, pp. 2737, 2002.
- [58] A. Kashyap, K. Garg, A. Solanki, T. Nautiyal and S. Auluck, "Electronic structure, Fermi surface, and Curie temperature calculations for the Co-Pt system," *Physical Review B*, vol. 60, no. 4, p. 2262, 1999.
- [59] S. N. Piramanayagam, B. Varghese, Y. Yang, W. K. Lee and H. K. Tan, "Investigations of stacking fault density in perpendicular recording media," *Journal of Applied Physics*, vol. 115, no. 24, p. 243901, 2014.
- [60] B. Varghese, S. N. Piramanayagam, Y. Yang, S. K. Wong, H. K. Tan, W. K. Lee and I. Okamoto, "Equiatomic CoPt thin films with extremely high coercivity," *Journal of Applied Physics*, vol. 115, no. 17, p. 17b707, 2014.
- [61] Y. Yang, B. Varghese, H. K. Tan, S. K. Wong and S. N. Piramanayagam, "Investigations of Stacking Faults in Stacked Granular Perpendicular Recording Media With a High-Anisotropy CoPt Layer," *IEEE Transactions on Magnetics*, vol. 50, no. 11, p. 3101404, 2014.
- [62] S. C. Chen, C. F. Lu, R. Z. Chen, C. M. Wang and P. C. Kuo, "Epitaxial Co₈₀Pt₂₀Films with Columnar Grains and Perpendicular Magnetic Anisotropy at Ambient Temperature," in *2016 International Conference of Asian Union of Magnetics Societies (ICAUMS)*, 2016, pp. 1.

- [63] S. C. Chen, S. U. Jen, R. Z. Chen, C. F. Lu, C. M. Wang and P. C. Kuo, "Effect of Cu, Cu/Ru, or Ru/Cu seed-layer on perpendicular magnetic anisotropy of Co₈₀Pt₂₀ films," *Journal of Magnetism and Magnetic Materials*, vol. 459, pp. 106, 2018.
- [64] S. N. Piramanayagam, "Perpendicular recording media for hard disk drives," *Journal of Applied Physics*, vol. 102, no. 1, p. 011301, 2007.
- [65] <http://gwyddion.net/download.php>.
- [66] <https://imagej.nih.gov/ij/download.html>.
- [67] R. Hara, K. Hayakawa, K. Ebata and R. Sugita, "Effect of aging and annealing on perpendicular magnetic anisotropy of ultra-thin CoPt films," *AIP Advances*, vol. 6, no. 5, p. 056117, 2016.
- [68] D. Makarov, F. Klimenta, S. Fischer, F. Liscio, S. Schulze, M. Hietschold, M. Maret and M. Albrecht, "Nonepitaxially grown nanopatterned Co–Pt alloys with out-of-plane magnetic anisotropy," *Journal of Applied Physics*, vol. 106, no. 11, p. 114322, 2009.
- [69] O. Dragos-Pinzaru, A. Ghemes, H. Chiriac, N. Lupu, M. Grigoras, S. Riemer and I. Tabakovic, "Magnetic properties of CoPt thin films obtained by electrodeposition from hexachloroplatinate solution. Composition, thickness and substrate dependence," *Journal of Alloys and Compounds*, vol. 718, pp. 319, 2017.
- [70] N. Gaur, S. L. Maurer, R. W. Nunes, S. N. Piramanayagam and C. S. Bhatia, "Ion-Implantation Studies on Perpendicular Media," *Journal of Nanoscience and Nanotechnology*, vol. 11, no. 3, pp. 2619, 2011.
- [71] J. Ziegler, "Stopping of energetic light ions in elemental matter," *Journal of applied physics*, vol. 85, no. 3, pp. 1249, 1999.
- [72] T. L. Jin, M. Ranjbar, S. K. He, W. C. Law, T. J. Zhou, W. S. Lew, X. X. Liu and S. N. Piramanayagam, "Tuning magnetic properties for domain wall pinning via localized metal diffusion," *Scientific Reports*, vol. 7, no. 1, p. 16208, 2017.
- [73] J. E. E. Baglin, A. J. Kellock, B. D. Terris, D. K. Weller and L. Folks, "Patterned magnetic recording media containing chemically-ordered FePt or CoPt," ed: Google Patents, 2001.
- [74] J. M. Coey, *Magnetism and magnetic materials*. Cambridge university press, 2010.
- [75] S. Piramanayagam, M. Ranjbar, R. Sbiaa, A. Tavakkoli KG and T. Chong, "Characterization of high-density bit-patterned media using ultra-high resolution magnetic force microscopy," *physica status solidi (RRL)–Rapid Research Letters*, vol. 6, no. 3, pp. 141, 2012.
- [76] N. Gaur, S. N. Piramanayagam, S. L. Maurer, S. E. Steen, H. Yang and C. S. Bhatia, "First-Order Reversal Curve Investigations on the Effects of Ion Implantation in Magnetic Media," *IEEE Transactions on Magnetics*, vol. 48, no. 11, pp. 2753, 2012.
- [77] Z.-P. Shi, "Magnetic domain structure in thin films with large perpendicular anisotropy," *Journal of Physics: Condensed Matter*, vol. 4, no. 11, p. L191, 1992.
- [78] R. C. O'Handley, *Modern Magnetic Materials: Principles and Applications*. 1999, p. 768.
- [79] M. Donahue and D. Porter, "OOMMF user's guide, version 1.0, National Institute of Standards and Technology," *Interagency Report NISTIR*, 1999.
- [80] B. D. Cullity and C. D. Graham, *Introduction to magnetic materials*. John Wiley & Sons, 2011.
- [81] H. Meng, W. Lum, R. Sbiaa, S. Lua and H. Tan, "Annealing effects on CoFeB-MgO magnetic tunnel junctions with perpendicular anisotropy," *Journal of applied physics*, vol. 110, no. 3, p. 033904, 2011.
- [82] B. Dieny, M. Chshiev, B. Charles, N. Strelkov, A. Truong, O. Fruchart, A. Hallal, J. Wang, Y. K. Takahashi and T. Mizuno, "Impact of intergrain spin-transfer torques due

- to huge thermal gradients in heat-assisted magnetic recording," *IEEE Transactions on Magnetics*, vol. 54, no. 12, pp. 1, 2018.
- [83] G. C. Varvaro, F.; , "Ultra-High-Density Magnetic Recording," *New York: Jenny Stanford Publishing*, 2016.
 - [84] S. Pisana, O. Mosendz, G. J. Parker, J. W. Reiner, T. S. Santos, A. T. McCallum, H. J. Richter and D. Weller, "Effects of grain microstructure on magnetic properties in FePtAg-C media for heat assisted magnetic recording," *Journal of Applied Physics*, vol. 113, no. 4, p. 043910, 2013.
 - [85] S. D. Granz and M. H. Kryder, "Granular L10 FePt (001) thin films for Heat Assisted Magnetic Recording," *Journal of Magnetism and Magnetic Materials*, vol. 324, no. 3, pp. 287, 2012.
 - [86] S. Piramanayagam and T. C. Chong, *Developments in data storage: materials perspective*. John Wiley & Sons, 2011.
 - [87] D. Weller and A. Moser, "Thermal effect limits in ultrahigh-density magnetic recording," *IEEE Transactions on Magnetics*, vol. 35, no. 6, pp. 4423, 1999.
 - [88] A. Q. Wu, Y. Kubota, T. Klemmer, T. Rausch, C. Peng, Y. Peng, D. Karns, X. Zhu, Y. Ding, E. K. C. Chang, Y. Zhao, H. Zhou, K. Gao, J. Thiele, M. Seigler, G. Ju and E. Gage, "HAMR Areal Density Demonstration of 1+ Tbpsi on Spinstand," *IEEE Transactions on Magnetics*, vol. 49, no. 2, pp. 779, 2013.
 - [89] D. Weller, O. Mosendz, G. Parker, S. Pisana and T. S. Santos, "L10 FePtX–Y media for heat-assisted magnetic recording," *physica status solidi (a)*, vol. 210, no. 7, pp. 1245, 2013.
 - [90] X. Wang, K. Gao, H. Zhou, A. Itagi, M. Seigler and E. Gage, "HAMR Recording Limitations and Extendibility," *IEEE Transactions on Magnetics*, vol. 49, no. 2, pp. 686, 2013.
 - [91] G. Ju, Y. Peng, E. K. C. Chang, Y. Ding, A. Q. Wu, X. Zhu, Y. Kubota, T. J. Klemmer, H. Amini, L. Gao, Z. Fan, T. Rausch, P. Subedi, M. Ma, S. Kalarickal, C. J. Rea, D. V. Dimitrov, P. Huang, K. Wang, X. Chen, C. Peng, W. Chen, J. W. Dykes, M. A. Seigler, E. C. Gage, R. Chantrell and J. Thiele, "High Density Heat-Assisted Magnetic Recording Media and Advanced Characterization—Progress and Challenges," *IEEE Transactions on Magnetics*, vol. 51, no. 11, pp. 1, 2015.
 - [92] T. Massalski, P. Subramanian, H. Okamoto and L. Kacprzak, "Binary Alloy Phase Diagrams, Vols. 1–3, ASM," *International Materials Park, OH*, p. 2018, 1990.
 - [93] D. Weller, G. Parker, O. Mosendz, A. Lyberatos, D. Mitin, N. Y. Safonova and M. Albrecht, "Review Article: FePt heat assisted magnetic recording media," *Journal of Vacuum Science & Technology B*, vol. 34, no. 6, p. 060801, 2016.
 - [94] J. S. Chen, B. C. Lim, J. F. Hu, B. Liu, G. M. Chow and G. Ju, "Low temperature deposited L10 FePt–C (001) films with high coercivity and small grain size," *Applied Physics Letters*, vol. 91, no. 13, p. 132506, 2007.
 - [95] S. D. Granz, K. Barmak and M. H. Kryder, "Granular L10 FePt:X (X = Ag, B, C, SiO_x, TaO_x) thin films for heat assisted magnetic recording," *The European Physical Journal B*, vol. 86, no. 3, p. 81, 2013.
 - [96] S. Chen, X. Shu, Q. Xie, C. Zhou, J. Zhou, J. Deng, R. Guo, Y. G. Peng, G. Ju and J. Chen, "Structure, magnetic and thermal properties of FePt–C–BN granular films for heat assisted magnetic recording," *Journal of Physics D: Applied Physics*, vol. 53, no. 13, p. 135002, 2020.
 - [97] L. Shi, C. Dames, J. R. Lukes, P. Reddy, J. Duda, D. G. Cahill, J. Lee, A. Marconnet, K. E. Goodson, J.-H. Bahk, A. Shakouri, R. S. Prasher, J. Felts, W. P. King, B. Han and J. C. Bischof, "Evaluating Broader Impacts of Nanoscale Thermal Transport Research," *Nanoscale and Microscale Thermophysical Engineering*, vol. 19, no. 2, pp. 127, 2015.

- [98] O. Mosendz, S. Pisana, J. W. Reiner, B. Stipe and D. Weller, "Ultra-high coercivity small-grain FePt media for thermally assisted recording (invited)," *Journal of Applied Physics*, vol. 111, no. 7, p. 07B729, 2012.
- [99] B. S. D. C. S. Varaprasad, Y. K. Takahashi, J. Wang, T. Ina, T. Nakamura, W. Ueno, K. Nitta, T. Uruga and K. Hono, "Mechanism of coercivity enhancement by Ag addition in FePt-C granular films for heat assisted magnetic recording media," *Applied Physics Letters*, vol. 104, no. 22, p. 222403, 2014.
- [100] L. Zhang, Y. K. Takahashi, A. Perumal and K. Hono, "L10-ordered high coercivity (FePt)Ag-C granular thin films for perpendicular recording," *Journal of Magnetism and Magnetic Materials*, vol. 322, no. 18, pp. 2658, 2010.
- [101] T. Shiroyama, B. S. D. C. S. Varaprasad, Y. K. Takahashi and K. Hono, "Microstructure and Magnetic Properties of FePt-Cr₂O₃ Films," *IEEE Transactions on Magnetics*, vol. 50, no. 11, pp. 1, 2014.
- [102] J. A. Christodoulides, P. Farber, M. Dannl, H. Okumura, G. C. Hadjipanaysi, V. Skumryev, A. Simopoulos and D. Weller, "Magnetic, structural and microstructural properties of FePt/M (M=C,BN) granular films," *IEEE Transactions on Magnetics*, vol. 37, no. 4, pp. 1292, 2001.
- [103] B. S. D. C. S. Varaprasad, Y. K. Takahashi, A. Ajan and K. Hono, "Electrically conductive (Mg_{0.2}Ti_{0.8})O underlayer to grow FePt-based perpendicular recording media on glass substrates," *Journal of Applied Physics*, vol. 113, no. 20, p. 203907, 2013.
- [104] L. Zhang, Y. K. Takahashi, K. Hono, B. C. Stipe, J. Juang and M. Grobis, "FePtAg-C Nanogranular Film as Thermally Assisted Magnetic Recording (TAR) Media," *IEEE Transactions on Magnetics*, vol. 47, no. 10, pp. 4062, 2011.
- [105] B. Lim, J. Chen and G. Chow, "Interfacial Effects of MgO Buffer Layer on Perpendicular Anisotropy of L-10 FePt Films," *IEEE transactions on magnetics*, vol. 42, no. 10, pp. 3017, 2006.
- [106] J. Wang, S. Hata, Y. Takahashi, H. Sepehri-Amin, B. C. S. Varaprasad, T. Shiroyama, T. Schrefl and K. Hono, "Effect of MgO underlayer misorientation on the texture and magnetic property of FePt-C granular film," *Acta Materialia*, vol. 91, pp. 41, 2015.
- [107] A. Perumal, L. Zhang, Y. Takahashi and K. Hono, "FePtAg-C nanogranular films fabricated on a heat resistant glass substrate for perpendicular magnetic recording," *Journal of Applied Physics*, vol. 108, no. 8, p. 083907, 2010.
- [108] B. C. S. Varaprasad, M. Chen, Y. Takahashi and K. Hono, "L-10-Ordered FePt-Based Perpendicular Magnetic Recording Media for Heat-Assisted Magnetic Recording," *IEEE transactions on magnetics*, vol. 49, no. 2, pp. 718, 2013.
- [109] B. Cullity and S. Stock, "Elements of X-ray Diffraction (Prentice Hall, New Jersey, 2001), p. 170."
- [110] T. Aoyama, I. Sato, H. Ito and S. Ishio, "B ion implantation effects on L10 chemical ordering of FePt thin films," *Journal of magnetism and magnetic materials*, vol. 287, pp. 209, 2005.
- [111] T. Hasegawa, G. Li, W. Pei, H. Saito, S. Ishio, K. Taguchi, K. Yamakawa, N. Honda, K. Ouchi and T. Aoyama, "Structural transition from L-10 phase to A 1 phase in FePt films caused by ion irradiation," *Journal of applied physics*, vol. 99, no. 5, p. 053505, 2006.
- [112] J. Tsai, Y. Chen, C. Pi, Y. Wu, G. Varvaro and C. Chang, "Magnetic and Microstructural Properties of [FePt-Mg (Ti, Ta, Zr, Nb, B) O] Granular Films," *IEEE Transactions on Magnetics*, vol. 54, no. 11, pp. 1, 2018.
- [113] J. P. Liu, Y. Liu, C. Luo, Z. Shan and D. J. Sellmyer, "Magnetic hardening in FePt nanostructured films," *Journal of applied physics*, vol. 81, no. 8, pp. 5644, 1997.

- [114] C. Chen, O. Kitakami, S. Okamoto and Y. Shimada, "Ordering and orientation of CoPt/SiO₂ granular films with additive Ag," *Applied Physics Letters*, vol. 76, no. 22, pp. 3218, 2000.
- [115] Z. Pan, J. Lin, T. Zhang, S. Karamat, T. Tan, P. Lee, S. Springham, R. Ramanujan and R. Rawat, "Lowering of L10 phase transition temperature of FePt thin films by single shot H⁺ ion exposure using plasma focus device," *Thin Solid Films*, vol. 517, no. 8, pp. 2753, 2009.
- [116] T. Hasegawa, T. Yamazaki and S. Ishio, "Correlation between crystal planes and disordering of ordered L10 FePt structure caused by ion irradiation," *Materials Letters*, vol. 178, pp. 243, 2016.
- [117] S. Saito, D. Hasegawa, D. Djayaprawira and M. Takahashi, "Evaluation for separating magnetic properties of columnar structure and initial layer in CoCr-based perpendicular thin-film media," *Journal-Magnetics Society of Japan*, vol. 25, no. 4/2, pp. 583, 2001.
- [118] M. Matsumoto, S. Sharmin, J. Inoue, E. Kita and H. Yanagihara, "Large Negative Uniaxial Magnetic Anisotropy in Epitaxially Strained Nickel Ferrite Films," *IEEE Transactions on Magnetics*, vol. 53, no. 11, pp. 1, 2017.
- [119] Y.-S. Zhang, W. He, J. Tang, S. S. Ahmad, W. Zhang, Y. Li, X.-Q. Zhang and Z.-H. Cheng, "The effect of annealing temperature on the magnetic anisotropy in Co ultrathin film on MgO(001) substrate," *AIP Advances*, vol. 7, no. 5, p. 056326, 2017.
- [120] F. Ono, "Torque curve measurements in HCP Co in very low fields," *IEEE Transactions on Magnetics*, vol. 23, no. 5, pp. 2428, 1987.
- [121] D.-T. Quach, Q.-H. Tran, K. Møhave and D.-H. Kim, "Perpendicular magnetic anisotropy and the magnetization process in CoFeB/Pd multilayer films," *Journal of Physics D: Applied Physics*, vol. 47, no. 44, p. 445001, 2014.
- [122] L.-C. Garnier, M. Marangolo, M. Eddrief, D. Bisero, S. Fin, F. Casoli, M. G. Pini, A. Rettori and S. Tacchi, "Stripe domains reorientation in ferromagnetic films with perpendicular magnetic anisotropy," *Journal of Physics: Materials*, vol. 3, no. 2, p. 024001, 2020.
- [123] A. Stellhorn, A. Sarkar, E. Kentzinger, M. Waschk, P. Schöffmann, S. Schröder, G. Abuladze, Z. Fu, V. Pipich and T. Brückel, "Control of the stripe domain pattern in L10-ordered FePd thin films," *Journal of Magnetism and Magnetic Materials*, vol. 476, pp. 483, 2019.
- [124] M. Jaafar, R. Sanz, J. McCord, J. Jensen, R. Schäfer, M. Vázquez and A. Asenjo, "Pattern-induced magnetic anisotropy in FePt thin films by ion irradiation," *Physical Review B*, vol. 83, no. 9, p. 094422, 2011.

Chapter 5. Artificial Pinning Sites for Controlled Domain Wall Motion

Controlled domain wall (DW) motion is extremely important for DW devices. This can be achieved using artificial pinning sites. Despite the serious urgency, the current state-of-the-art of this aspect is still in its younger stage. In this thesis, we have explored a couple of ideas using micromagnetic simulations for this purpose. This chapter is dedicated to an in-depth discussion of these. Some of the results presented in this chapter have been published as “Kumar *et al.*, *Journal of Magnetism and Magnetic Materials*, 507, 166807 (2020)”.
[Reproduced and/or inspired from mentioned paper, with the permission of the corresponding publisher]

5.1. Introduction

When an external stimulus (or two simultaneous stimuli) is applied to move the DWs back and forth, it is very difficult to stop the DWs at specific positions [1-4]. In a deposited magnetic thin film, there might exist some energy barriers or energy wells due to the variation in the properties of the deposited film. These regions act as pinning centers and they may be far away from each other. As a result, in a tiny device, there may not be any pinning center and the DW motion may be abrupt and not be controllable. The only way to stop the DWs controllably and repeatably at certain positions in small devices is by fabricating artificial pinning sites along the nano-track [1-6]. These artificial pinning sites also help in achieving uniform bit lengths in DW devices. Moreover, by packing the pinning sites closely, the storage density of DW devices can be enhanced. Artificial pinning sites have also been proposed for higher thermal stability of DW devices [7-9]. As mentioned in chapter 2, the thermal stability determines the lifetime of the data and hence of the DW device.

The artificial pinning sites can be categorized into (i) geometrical pinning sites, which are based on altering the physical shape of the nanowire at the position of pinning sites [5, 10, 11], and (ii) non-geometrical pinning sites, which are based on changing the magnetic properties locally

[5, 12, 13]. In the early days of DW memory research, researchers have proposed the fabrication of notches for parking the DWs at the pinning sites. Ono et al. fabricated semicircular double-sided notches in the nanowires of NiFe/Cu/NiFe. They demonstrated the pinning of DWs at the pinning position by measuring the giant magnetoresistance (GMR) of the devices [14, 15]. Later, Kläui et al. demonstrated DW pinning at triangular notches in the ring-shaped NiFe and Co nanowires through magnetoresistance measurements [16, 17]. Hayashi et al. reported DW pinning in triangular notches in straight NiFe nanowires [18]. Several other researchers studied DW pinning by forming triangular notches along the nanowires [19-21]. Parkin et al. fabricated a series of notches at different positions of a NiFe wire and showed subsequent pinning at different pinning centers upon the application of electrical current [6]. In other studies, Endo et al. reported DW pinning in double-sided notches and Himeno et al. studied the same in antisymmetric notches [22, 23].

The formation of notches at specific positions changes the shape anisotropy in materials such as NiFe and Co and higher energy is needed to move the DWs from these constrictions. There are, however, some serious issues with the notches. Firstly, they may have variations in shapes and may result in different pinning strengths of pinning sites. Secondly, if notches have the smallest feature size of X , achieved with a certain lithography technique, the width of the nanowire must be larger than X (say, $3X$). Therefore, notches pose a problem in the miniaturization of the DW devices and hence a problem towards increasing the storage capacity. Moreover, the constriction in the form of notches increases the problem of Joule heating of the DW devices at the notch positions due to the size-dependent electrical resistance. In addition to all the above disadvantages, notches have been proposed and studied for materials with in-plane (IP) anisotropy, which are not suitable for DW devices in many ways. These materials offer lower storage capacity (due to wider DWs), lower DW velocity, lower thermal stability (due to lower anisotropy), and higher power consumption compared to the

materials with perpendicular magnetic anisotropy (PMA). Therefore, notches are not the preferred contenders in the research community in current times.

Figure 5.1. The schematics of fabrication of pinning sites in the form of (a) triangular notches and (b) stepped or staggered nanowire. These methods fall under geometrical pinning sites. Schematic presentations of non-geometrical pinning sites by utilizing (c) local metal diffusion, (d) ion-implantation of non-magnetic ions, and (e) exchange coupling between in-plane magnetized crossbar and out-of-plane magnetized domain wall track.

Recently, Bahri and Sbiaa conceived the idea of the stepped (staggered) nanowire for creating the pinning sites [24]. Two neighboring segments of the nanowire connect with the small offset at the junctions. They successfully demonstrated the pinning of a DW at the steps. Later, they explored this theoretical concept experimentally [25]. Both of these studies were performed for IP (NiFe) material. Therefore, it became essential to test the concept in out-of-plane (OOP)

materials. Kumar et al. performed micromagnetic simulations and demonstrated DW pinning in $(\text{Co/Ni})_n$ multilayer system (OOP) using the idea of a stepped nanowire [5]. Subsequently, Mohammed et al. reported the experimental evidence of DW pinning in a stepped nanowire for a $(\text{Co/Ni})_n$ multilayer system [26]. In a couple of more studies, DWs were found to pin at the steps in a stepped nanowire geometry [27, 28]. For the case of NiFe, the change in shape anisotropy (similar to the case of geometrical notches), at the adjoint of two neighboring segments of a stepped nanowire is one reason for the DW pinning. Other than this, the DW surface tension and hence the DW surface energy is altered at the geometrical pinning centers and hence higher energy is needed to push the DWs from the geometrical pinning centers [29]. For the $(\text{Co/Ni})_n$ multilayer system, the latter is the main cause for DW pinning as there is no (or insignificant) shape anisotropy for such materials. Although this method showed good results from a DW pinning point of view, an increasing pinning strength is required at successive pinning centers, which will limit the storage capacity of the devices. In general, the geometrical methods may be less favorable because of fabrication difficulties. Therefore, researchers have studied a few non-geometrical methods.

In 2014, Polenciuc et al. researched that the exchange bias from an antiferromagnetic crossbar can be utilized to form the pinning sites along the CoFe ferromagnetic microwire [30]. Later, Jin et al. used the concept of local metal diffusion at the parking sites to modify the magnetic properties locally [31]. For this, they used the crossbars of non-magnetic metals on the top of ferromagnetic wires (NiFe) and annealed them at a higher temperature. This resulted in the diffusion of non-magnetic metal atoms in the ferromagnetic wire locally. These magnetically modified regions were found to show the pinning of the DWs. However, this method is not suitable for multilayer systems where interface anisotropy is utilized to promote the OOP anisotropy in the system. This is because the annealing at higher temperatures gives rise to the

interface mixing (say interlayer mixing of Co and Ni in $(\text{Co/Ni})_n$ multilayer system) and hence reduces or even results in the loss of OOP anisotropy.

Therefore, Jin et al. utilized ion-implantation of non-magnetic B^+ ions locally on a track of $(\text{Co/Pd})_n$ multilayer system with OOP anisotropy [32, 33]. Successful pinning of DWs was observed in this method. However, this method has the problem of the lateral straggle of non-magnetic ions or atoms (spatial movement of ions or atoms), which poses a limitation on packing pinning sites very close to each other. The lateral straggle of atoms also happens when a non-magnetic metal is diffused into an FM layer and hence adds another problem to the metal diffusion method. More recently, Jin et al. proposed that the exchange coupling between an in-plane magnetized crossbar and an out-of-plane magnetized track can be used to form pinning sites [13]. They reported DW pinning in this method while using magnetic field as well as electrical current (separate experiments) as a driving force for DWs. In both experiments, increasing strength of the driving force was used to push the DW from subsequent pinning centers, which may limit the number of parking sites along the nanowire.

Inspired by the advantages and disadvantages of these methods, we have studied two distinct methods of creating pinning sites. The first of these falls under the geometrical pinning site, while the second comes under the non-geometrical pinning site. In this chapter, we will present an in-depth discussion of these studies.

5.2. Domain Wall Pinning in Square Wave Shaped Nanowire

First, we have proposed and studied the idea of square wave shaped nanowires. The neighboring segments of the proposed design meet at an angle of 90° to each other and eventually form the shape of a square wave (therefore, the name square wave shaped nanowire). The schematic diagram of the design is presented in Figure 5.2 (c). The intersection of two adjacent segments of square wave shaped nanowires was expected to present different energy for the DW and should pin the DW. As stated earlier for PMA materials, the DW surface energy is the reason for DW pinning in geometrical pinning sites [29]. In a geometrical expansion (say anti-notch type pinning site), the DW, which can be considered as an elastic membrane, experiences a Laplace pressure ($P_{Laplace}$). This Laplace pressure results in contraction of the magnetic domain. This means that at the position of the pinning site, the $P_{Laplace}$ hinders the further motion of DW. Therefore, increased energy is needed to further move the DW. In the present case, the corner of two neighboring segments of a square wave shaped nanowire presents a geometrical enlargement. Therefore, there will be a $P_{Laplace}$, which will exert pressure along $-x\cos\theta$ (Figure 5.2 (a)). The spin-transfer torque (STT), which is the driving force in the present study, can be considered as the other source of the pressure. Only when the pressure due to STT (P_{STT}) along $+x\cos\theta$ will exceed the $P_{Laplace}$, the DW will depin from the pinning center (corresponding corner). This phenomenon is described in Figure 5.2 (a). Detailed micromagnetic simulations were performed to study this idea and results will be presented and discussed in the upcoming sub-sections of this chapter.

Figure 5.2. (a) The illustration describing the concept behind the proposed design. The schematics of the (b) reference nanowire and (c) proposed square wave shaped nanowire with respective dimensions.

5.2.1. Simulation Model

To study the proposed idea, we performed micromagnetic simulations using the Mumax3 software [34-36]. For comparison, we have simulated a straight (or reference) nanowire with the dimensions $500 \text{ nm} \times 40 \text{ nm} \times 5 \text{ nm}$. For the square wave shaped nanowire, the width and thickness of the nanowire were kept the same as that of the reference nanowire. The length is divided into three segments. The dimensions of all the segments of the square wave shaped nanowire were kept the same. This is to facilitate equal bit lengths. The schematic of the reference nanowire and proposed square wave shaped nanowire is shown in Figure 5.2 (b) and (c). The cell size was set at $5 \text{ nm} \times 5 \text{ nm} \times 5 \text{ nm}$, which is less than the exchange length (7 nm) for the magnetic parameters used in the simulations [34, 37-39]. The magnetic parameters such as saturation magnetization (M_s), exchange constant (A), anisotropy constant (K_u), damping parameter (α), and non-adiabaticity of STT (β) were chosen as $8 \times 10^5 \text{ A/m}$, $2 \times 10^{-11} \text{ J/m}$, $5.3 \times 10^5 \text{ J/m}^3$ (0, 0, 1), 0.02 and 0.02 respectively [13, 40-42]. For the reference nanowire, the current was applied in the $-x$ direction and hence the electrons were set to flow in the $+x$ direction. On

the other hand, for the square wave shaped nanowire, the current was applied in the $-x$ direction in the first and third segment and in the $-y$ -direction in the second segment. At the first and second corners, the current was set in the $-x\cos 45^\circ$ direction. In all the cases the DW was found to move in a direction opposite to the current flow, which is the signature of STT driven DW dynamics [43].

5.2.2. Domain Wall Dynamics in Reference Nanowire

Firstly, we have simulated the DW motion for the reference (straight) nanowire. We performed simulations for different current densities starting from 1×10^{11} A/m² to 1×10^{12} A/m². The DW starts from the left region and moves to the other end of the nanowire in subsequent simulation times. Figure 5.3 (a) shows the snapshots of the DW position at different simulation times for the current density (J) of 8×10^{11} A/m². Here, blue and red colors represent the magnetization in the $+z$ and $-z$ directions, respectively. The DW was nucleated at the left end of the nanowire (initial condition, $t = 0$ ns) and moved with a velocity of 29.7 m/s at a J of 8×10^{11} A/m². Similar results (motion of DW from the left end to right end) were obtained for all the other current densities. From these results, we calculated the DW velocities for different simulated current densities. Figure 5.3 (b) shows the plot of DW velocity as a function of J . The DW velocity linearly increases with J , which can be explained in the following manner. For STT-induced DW motion, the DW velocity takes the following relation with the current density [43].

$$v = \left(\frac{\beta}{\alpha}\right) \frac{JPg\mu_B}{2eM_s} \quad (5.1)$$

Where P , g , μ_B , and e are the polarization of spin current, Landé g factor, Bohr magneton, and electronic charge, respectively. All the other parameters have the same meaning as stated previously. In the present case, $\beta = \alpha$ and therefore, the above expression reduces to,

$$v = \frac{JPg\mu_B}{2eM_s} \quad (5.2)$$

In the above expression, one can see that the DW velocity depends linearly on the current density J .

Figure 5.3. (a) The snapshots of DW motion in reference nanowire at different simulation times for a current density of $8 \times 10^{11} \text{ A/m}^2$ and (b) the plot of DW velocity as a function of current density (J).

5.2.3. Domain Wall Dynamics in Square Wave Shaped Nanowire

In the next step, we performed simulations for the square wave shaped nanowire. Figure 5.4 (a), (b), and (c) show the graphs of DW position vs simulation time (t) for J values of $3 \times 10^{11} \text{ A/m}^2$, 8×10^{11} , and $1 \times 10^{12} \text{ A/m}^2$, respectively. The DW gets pinned at the first pinning site with no further motion for the current densities from $1 \times 10^{11} \text{ A/m}^2$ to $3 \times 10^{11} \text{ A/m}^2$. As it can be seen in Figure 5.4 (a), the DW starts moving from the left end of the nanowire for $3 \times 10^{11} \text{ A/m}^2$ and gets pinned at the first pinning site. Even after applying the current for a long time, the DW does not move from this position. The corresponding snapshots are depicted in Figure 5.5 (a). However, for current densities larger than $3 \times 10^{11} \text{ A/m}^2$, the DW stops only for some time at pinning centers and then depins. As an example, for $8 \times 10^{11} \text{ A/m}^2$ the DW starts moving from the left and gets pinned at the first pinning site for some time. Further application of the current results in the depinning of the DW from the first pinning site. After sweeping the second

segment of the nanowire, the DW pins at the second pinning site for some time and then depins. Subsequently, the DW moves towards the right end of the nanowire. This behavior can be seen in Figure 5.4 (b) and 5.5 (b). In Figure 5.4 (b), x and y represent the DW motion in segments along x and y directions respectively. Therefore, when the DW moves in a segment along the x -direction, the DW position along the y -direction is constant and vice versa.

Figure 5.4. The plot of DW position vs simulation time for (a) $3 \times 10^{11} \text{ A/m}^2$, $8 \times 10^{11} \text{ A/m}^2$, and (c) $1 \times 10^{12} \text{ A/m}^2$. (d) The graph of pinning time as a function of current density at two studied pinning sites.

Figure 5.5. The snapshots of DW motion for square-wave shaped nanowire at different simulation times for current densities of (a) $3 \times 10^{11} \text{ A/m}^2$ and (b) $8 \times 10^{11} \text{ A/m}^2$.

Similar results were observed for J of 1×10^{12} A/m², except the pinning time at pinning centers decreases further. For more clarity, we plotted the pinning time at the two pinning sites (abbreviated as corner 1 and corner 2) as a function of J (Figure 5.4 (d)). The pinning time at both the pinning centers decreases with the increase in current density, which is an expected result. Moreover, the stay time at corner 1 is slightly more than that of corner 2. A wise tuning of current density and pulse width is needed to observe multi-resistance states in this design. The current must be used to push the DW until the pinning site and then the current must be set to zero. For further motion, the current must again be switched on and so on.

The observed results can be categorized into two separate categories. The first category ranges from a current density of 1×10^{11} A/m² to 3×10^{11} A/m². In this range, the $P_{Laplace}$ is much larger than P_{STT} at the first pinning site, therefore, P_{STT} cannot depin the DW from the first pinning site. The second category starts from 4×10^{11} A/m² and ends at 1×10^{12} A/m². In this range, P_{STT} becomes comparable (or even larger) to $P_{Laplace}$. As the current density increases P_{STT} increases and becomes increasingly dominant (as J increases) compared to $P_{Laplace}$. Therefore, the pinning time decreases as J increases. Moreover, the parking time at pinning site 2 is slightly smaller than that at pinning site 1 for all J values (from $J = 4 \times 10^{11}$ A/m² to 1×10^{12} A/m²). This could be attributed to a slightly smaller $P_{Laplace}$ at pinning site 2. This may be due to variation in the environment for pinning sites 1 and 2. This means pinning site 1 is followed by another pinning site whereas this is not the case for pinning center 2. It is important to note that the nature of DW and DW dynamics is identical at both the pinning centers.

5.2.4. Summary

We have observed the linear scaling of DW velocity with the current density for the reference nanowire, which is in good agreement with the existing micromagnetic theory. For the square wave shaped nanowire, we observed that the proposed design offers pinning at the corners. However, the range of current densities for which $P_{Laplace} \gg P_{STT}$ is quite small. The other J values may not be suitable as they can only pin the DW for smaller times. In order to observe better pinning, we shifted to another concept, which comes under the non-geometrical pinning sites.

5.3. Domain Wall Pinning in a Nanowire with Nanoscale Interfacial

Dzyaloshinskii–Moriya Interaction (*i*DMI) Region

5.3.1. Introduction

The non-collinear interaction between spins [44-49], described as Dzyaloshinskii-Moriya interaction (DMI) has picked up the interest of researchers in the recent past [50-61]. Chiral DWs that exhibit ultra-high-speed and skyrmions are some highlights of DMI-induced phenomena [62-67]. In particular, for DW devices in spintronic applications, magnetic systems with interfacial DMI (*i*DMI) result in more rigid, stable, and Chiral DWs, which move faster and more efficiently. Emori et al. studied current-induced DW dynamics in Pt/CoFe/MgO and Ta/CoFe/MgO films and observed DW motion due to Slonczewski-like torque from the spin Hall Effect [68]. The *i*DMI results in the formation of Néel walls with a fixed chirality (instead of Bloch walls) and hence showed larger DW velocity. Ryu et al. deduced a similar conclusion for Pt/Co/Ni/Co/Pt films [67]. Yang et al. studied current-driven DW dynamics in Co/Ni/Co/Ru/Co/Ni/Co/Pt synthetic antiferromagnetic coupled structure, where chiral DWs were stabilized due to *i*DMI from the Pt layer [69]. A higher DW velocity of 750 m/s was observed in this study. Since the DWs with the same chirality repel each other, they can be pushed synchronously along a nanowire. Kim et al. demonstrated the movement of several chiral DWs simultaneously upon the application of an out-of-plane pulsed magnetic field [70]. In all the studies involving *i*DMI so far, the investigations were focused on a ferromagnetic (FM) layer, which is entirely under *i*DMI. We decided to investigate a distinctive scenario, where *i*DMI acts only in a narrow region of a ferromagnetic (FM) nanowire. This nanoscale *i*DMI is expected to result in some new physics useful for fundamental understanding as well as practical applications.

In micromagnetic theory, the total energy of a magnetic material is the sum of the anisotropy energy, exchange energy, and magnetostatic energy [36, 71, 72],

$$E_{total} = E_{anisotropy} + E_{exchange} + E_{magnetostatic} = E_1 \quad (5.3)$$

Here, we have neglected the Zeeman energy and energy due to thermal effects. When a heavy metal (HM) layer is introduced in the vicinity of an FM layer, the high spin-orbit coupling (SOC) in the HM layer and structural inversion asymmetry at the HM/FM interface induces an *i*DMI in the FM layer. In the HM/FM structure, the *i*DMI is the antisymmetric exchange interaction, which favors the perpendicular orientation of neighboring spins in the FM layer [47, 73]. Moreover, the *i*DMI between two neighboring spins is mediated by an atom in the HM layer with high SOC and given by $E_{DMI} = \vec{D}_{ij} \cdot (\vec{S}_i \times \vec{S}_j)$. In the presence of *i*DMI, the total energy of the system modifies to [74],

$$E_{total} = E_{anisotropy} + E_{exchange} + E_{magnetostatic} + E_{DMI} = E_2 \quad (5.4)$$

Where E_{DMI} is the energy due to *i*DMI in the FM layer. The *i*DMI significantly affects the DW statics as well as dynamics. In general, the materials with perpendicular magnetic anisotropy (PMA) host the Bloch type DWs. The presence of *i*DMI changes the DW structure in these materials. As the *i*DMI increases, the DW structure changes from Bloch wall to mixed Bloch and Néel DW, and beyond a critical value, the DW structure changes to pure Néel type DW [75]. In other words, the presence of *i*DMI lifts the degeneracy in DW energy and changes the structure to pure Néel DW after a certain critical value. The *i*DMI behaves like an effective magnetic field in a direction perpendicular to the DW surface and rotates the magnetization in the DW. Secondly, the *i*DMI tilts the DW surface and the extent of this tilt was found to be proportional to the magnitude of *i*DMI [75].

In the present study, the *i*DMI was introduced only in a nanoscale region. Therefore, the energy vs *x*- co-ordinate curve resembles Figure 5.6 (a). Introduction of nanoscale *i*DMI forms the “step” in energy. In this study, we have considered a -ve *i*DMI value. Therefore, the total energy E_2 becomes less than E_1 . However, the total energy of the system is always negative, and hence energy-*x* curve takes the forms as in Figure 5.6 (a). When the DW reaches the nanoscale *i*DMI

region, it experiences higher energy and gets pinned in the nanoscale *i*DMI region. we have performed detailed micromagnetic simulations to study this physical concept and the results will be presented in the upcoming sections.

Figure 5.6. (a) The energy profile as a function of x -coordinate (along the length of nanowire) for the proposed concept and (b) The schematic of the proposed concept with respective dimensions. Here, nanoscale *i*DMI is introduced in the middle of the nanowire.

5.3.2. Simulation Model

To systematically study this idea, we performed micromagnetic simulations. We have taken the dimensions of the nanowire as $300 \text{ nm} \times 30 \text{ nm} \times 7 \text{ nm}$, as shown in Figure 5.6 (b). The cell size was fixed at $4.7 \text{ nm} \times 3.75 \text{ nm} \times 3.5 \text{ nm}$. The cell size is smaller than the exchange length for the given magnetic parameters. The magnetic parameters such as saturation magnetization (M_s), exchange constant (A), anisotropy constant (K_u), damping constant (α), non-adiabatic parameter (β) were chosen as $6.55 \times 10^5 \text{ A/m}$, $1 \times 10^{-11} \text{ J/m}$, $4 \times 10^5 \text{ J/m}^3$, 0.02, 0.02 respectively. These values are experimentally viable for the $(\text{Co/Ni})_n$ multilayer system [41, 76-81]. The *i*DMI was introduced in the middle of the FM nanowire with various widths (abbreviated as “ d ” in Figure 5.6 (b)) such as 5 nm, 10 nm, 20 nm, 30 nm, 33 nm, 35 nm, 37 nm, and 40 nm. The *i*DMI values used in the simulations were varied from -0.5 mJ/m^2 to -3 mJ/m^2 with a step of -0.5 mJ/m^2 . It is important to see that two atomic layers can be considered along the z -axis.

Therefore, in practical situations, the *i*DMI originates at the interface (as the name suggests) of HM and FM and it dilutes upon going away from the interface. However, in a situation where the FM layer is ultrathin (only two atomic layers), *i*DMI should not decrease much. Therefore, we have considered a uniform *i*DMI along the z -axis (in the nanoscale region).

5.3.3. Domain Wall Dynamics in Reference Nanowire

Like the previous case, we have simulated the DW motion for the reference nanowire at first, where no nanoscale DMI region is present. Figure 5.7 (a) displays the graph of the z -component of the magnetization (m_z), which corresponds to the DW position, with simulation time. The DW moves from the left end of the nanowire to the right end of the nanowire without any stopping. Figure 5.7 (b) is the plot of DW velocity as a function of J . The DW velocity linearly increases with J , which is consistent with our previous explanation (Eq. 5.2). Figure 5.5 (c) shows the screenshots of the DW position at different simulation times for the J of 1×10^{11} A/m². Here, blue and red colors represent the magnetization in $+z$ and $-z$ directions, respectively. As mentioned, the DW is nucleated at the left end of the nanowire (initial condition) and moved with a constant current of 1×10^{11} A/m².

Figure 5.7. (a) The plot of m_z (DW position) vs simulation time at different current densities for reference nanowire, (b) the DW velocity profile as a function of J for reference nanowire, and (c) snapshots of DW positions at different simulation times for $J = 1 \times 10^{11} \text{ A/m}^2$.

5.3.4. Domain Wall Dynamics in Samples with Nanoscale Interfacial Dzyaloshinskii–Moriya Interaction (i DMI) Region

In the next step, we introduced a nanoscale i DMI region in the middle of the nanowire. We first performed the simulations for a d value of 5 nm and i DMI constant of -0.5 mJ/m^2 . We varied the current density from $6 \times 10^{10} \text{ A/m}^2$ to $5 \times 10^{12} \text{ A/m}^2$. For smaller current densities (from $6 \times 10^{10} \text{ A/m}^2$ to $1 \times 10^{11} \text{ A/m}^2$), the DW starts from the left end and gets pinned at the nanoscale i DMI region. Only after applying a higher current density ($2 \times 10^{11} \text{ A/m}^2$ and above), the DW gets depinned from the pinning site and travels towards the right end of the nanowire.

Figure 5.8 (a) shows the snapshots at different simulation times for $d=5$ nm, $i\text{DMI} = -0.5$ mJ/m², and $J = 6 \times 10^{10}$ A/m², where the DW moves until the pinning position and gets pinned at the pinning site without further movement. Figure 5.8 (b) summarizes the pinning and depinning for this set of simulations. Next, we increased the $i\text{DMI}$ value to -1 mJ/m² keeping all the other parameters identical to the discussed case. Similar findings were obtained for this case as well (Figure 5.8 (c) and (d)).

Figure 5.8. Illustration of DW pinning for a nanowire with nanoscale $i\text{DMI}$ for (a) and (b) $d=5$ nm, $i\text{DMI}$ constant $= -0.5$ mJ/m² and (c) and (d) $d=5$ nm, $i\text{DMI}$ constant $= -1$ mJ/m². The yellow lines in (a) and (c) represent the middle of the nanowire and hence the middle of the pinning site.

We then extended our simulations for different d and i DMI values ($d = 5, 10, 20, 30$ nm and i DMI = -0.5, -1.0, -1.5, -2.0, -2.5 and -3.0 mJ/m²). Figure 5.9 (a) illustrates the DW pinning in all the simulated cases for the J value of 1×10^{11} A/m². The DW gets pinned at the smaller current densities and gets depinned after applying a higher current. The DW pinning for a large range of d and i DMI values is one of the central advantages of the proposed concept. From these simulations, we estimated the depinning current for all the simulated cases (Figure 5.9 (b)). For low d and i DMI values, the DW depins at a J value of 2×10^{11} A/m². However, the depinning current is larger for larger d and i DMI values. An increase in d and i DMI values will increase the width and height of the energy barrier formed by nanoscale i DMI. Therefore, a higher current will be required to depin the DW from the pinning site.

Figure 5.9. (a) Illustration of DW pinning for a nanowire with nanoscale i DMI for all simulated d and i DMI values at J of 1×10^{11} A/m² and (b) the plot of depinning current as a function of i DMI for different d values.

For sufficiently larger d and i DMI values, the DW was found to oscillate for a certain range of J . The DW dynamics can be divided into three regions as a function of d and i DMI values. These are (a) DW pinning region, (b) Intermediate region, and (c) DW pinning and DW

oscillation region. Figure 5.10 is the phase diagram showing these three regions. For the $iDMI$ value smaller than -2 mJ/m^2 ($-0.5, -1.0, -1.5, -2.0 \text{ mJ/m}^2$), the DW gets pinned for smaller currents and get depinned for currents above a certain threshold, for all values of d . This region is referred to as the DW pinning region. For clarity, we have not shown the graph for $iDMI$ values less than -2 mJ/m^2 as it comes under the pinning region. For the $iDMI$ value of -2.5 mJ/m^2 , DW pinning was observed for d values of 5 nm and 10 nm . However, for $iDMI$ value of -2.5 mJ/m^2 and d value of 20 nm , the DW oscillations were observed only for one J value. For J smaller than this, the DW gets pinned and for J larger than this, the DW gets depinned and moves to the extreme right end. The d and $iDMI$ values for which, the DW oscillations were observed only for one current density value were included in the intermediate region. The DW pinning as well as DW oscillation region was observed for the d values between 30 nm to 40 nm and $iDMI$ values between -2.5 to -3 mJ/m^2 . In this region, the DW was found oscillating for a large range of current values. We will discuss these results later in sub-section 5.4.

Figure 5.10. The phase diagram of DW dynamics as a function of d and DMI values. The DW oscillations were observed from a current density of $5 \times 10^{11} \text{ A/m}^2$ to $1 \times 10^{12} \text{ A/m}^2$ for all the cases except for $d = 40 \text{ nm}$ and $iDMI$ constant of -3 mJ/m^2 . For $d = 40 \text{ nm}$ and $iDMI$ constant of -3 mJ/m^2 , the oscillations were observed from a J value of $2 \times 10^{11} \text{ A/m}^2$ to $1 \times 10^{12} \text{ A/m}^2$.

As shown in Figure 5.11 (a) to (d), when the DW enters the i DMI active region, the structure of the DW changes but it may not reach a stable (or minimum energy) state (due to finite d). In addition, the DW tilts in this region, and the lower edge of the DW gets pinned to the beginning of the (i DMI introduced) pinning region. The upper end moves following the tilting of the DW. As stated earlier, the DW tilting depends on the magnitude of the i DMI constant. Sufficient tilting, which supports the DW oscillation only occurs with higher values of i DMI. Moreover, the nature of tilting of the DW i.e. whether the DW tilts towards the $-yx$ plane or the $+yx$ plane, depends on the sign of i DMI. Since the $-ve$ i DMI favors the tilting in $-yx$ plane, therefore, the lower end of the DW gets pinned and other end moves. As the upper end slightly crosses the pinning region (the other end of i DMI active region), the DW experiences different energy (or torque acting on the DW) as there is no i DMI (Figure 5.6 (a and b)). If the torque from STT is not enough to depin the lower edge from the lower end of pinning site (starting point of i DMI active region), the DW comes back, and this process continues and results in oscillations. It is important to note that the DW moves in the $+x$ direction because of the presence of STT and not because of the existence of i DMI. However, the i DMI acts as a restoring force for the DW within a certain range of J , d , and i DMI values. The whole phenomenon is schematically illustrated in Figure 5.11 (a) to (d).

Figure 5.11. (a) Initial state: When the DW (white intermediate region) is far from the nanoscale DMI region, the Bloch DW is formed. The effect of H_{DMI} on the DW in subsequent times; (b) As soon as the DW enters the DMI introduced region, the center of the DW (black arrow) points in the $+x$ direction, and the Bloch wall changes to Néel wall. In this case, the angle between the DMI field and the central spin of DW is zero. In the subsequent time intervals depicted in (c) and (d), the DW bends in the plane spanned by the x -axis and y -axis. In all the cases, one end of the DW is pinned while the other end of the DW is moving.

5.3.5. Domain Wall Pinning at Multiple Pinning Sites

Taking the findings of the single pinning site as a reference, we introduced five pinning sites along the nanowire. The spacing between the consecutive pinning sites was set at 40 nm and the width of one pinning site is fixed at 10 nm. The i DMI strength was chosen to be -1.5 mJ/m^2 for every pinning site. We applied a pulsed current of amplitude $1 \times 10^{12} \text{ A/m}^2$ and pulse width 1.9 ns. The pulse width was optimized based on DW's travel time between two pinning centers. The schematic of the design is presented in Figure 5.12 (a). Figure 5.12 (b) and (c) represent the DW position (m_z) vs simulation time graph and snapshots showing the DW pinning at different pinning sites, respectively. The red and blue colors in Figure 5.12 (c) represent the

magnetization up and down, respectively. Moreover, P_i represents the i^{th} pinning site in the same figure.

Figure 5.12. (a) The schematic of the nanowire with five pinning sites with corresponding dimensions and spacing between consecutive pinning sites, (b) The plot of DW position vs simulation time showing the subsequent pinning at different pinning sites, and (c) corresponding snapshots at corresponding simulation times. Here, P_i is the i^{th} pinning site and red and blue colors represent magnetization up and down respectively.

After the application of the first pulse, the DW reaches from its initial position to the first pinning site and stays there until the application of the second pulse. After the second pulse, the DW depins from the first pinning site and pins at the second pinning site and stays there until the third current pulse. This process continues and, in the end, DW depins from the last pinning site and travels towards the right end of the nanowire.

5.4. Domain Wall Oscillations in a Nanowire with Nanoscale Interfacial Dzyaloshinskii–Moriya Interaction (*i*DMI) Region

In the earlier discussions, we noticed that there is a certain range of parameters, where the DW oscillates. We extended our simulations to explore these DW oscillations in detail. First, we will discuss the nature of observed DW oscillations. Subsequently, we will discuss the dependence of oscillation frequency on various parameters such as J , d , magnetic parameters, and *i*DMI constant. Moreover, we have developed an analytical model to describe the simulated DW oscillations. In the end, we will also discuss the concept of a device proposed on the basis of observed results, which could potentially be useful for many applications, in particular, for emerging Neuromorphic computing.

The variation of z -component of magnetization (m_z) as a function of simulation time (t) was used to describe the DW oscillation. Figure 5.13 (a) shows the DW oscillation for the nanowire with d of 30 nm, *i*DMI strength of -3 mJ/m^2 , and a J of $5 \times 10^{11} \text{ A/m}^2$. Once the DW reaches the DMI region, the lower end of the DW gets pinned at the lower edge of the DMI region, and the other end of the DW oscillates back and forth with a certain frequency. The nature of observed simulations is consistent with our explanation presented in section 5.3.4. In order to estimate the frequency of oscillations, we performed the fast Fourier transform (FFT) of the m_z vs t data. The inset of Figure 5.13 (a) shows the FFT of the oscillation data, where the first-order frequency peak appears at 4.19 GHz. Figure 5.13 (b) is the -zoomed in- version of Figure 5.13 (a), where some of the major points have been highlighted. Figure 5.13 (c) depicts the

corresponding snapshots of DW positions at different simulation times. As can be seen, the DW enters the DMI modified region at t of 2.75 ns and reaches the other end at $t \sim 2.96$ ns. At $t \sim 3.03$ ns, the DW moves back and m_z attains the same value as it was at $t = 2.75$ ns. This cycle repeats for subsequent simulation times.

Figure 5.13. (a) The depiction of DW oscillations in local magnetic modification from m_z vs simulation time (t) graph for $d = 30$ nm, $\text{DMI} = -3$ mJ/m² and $J = 5 \times 10^{11}$ A/m². Although we have performed simulations for 100 ns, the oscillations - zoomed in for 5 ns - are shown for clarity. Inset: The fast Fourier transform (FFT) of the waveforms, which shows the occurrence of first-order frequency peak at 4.19 GHz. (b) Further -zoomed in- version of Figure 5.13 (a), highlighting the major points in m_z vs t graph, and (c) corresponding snapshots of the DW position at different simulation times. Here, blue and red colors represent magnetization in $+z$ and $-z$ directions respectively.

5.4.1. Dependence of Oscillation Frequency on External Stimulus and Geometrical Parameters

The DW dynamics strongly depends on the driving force, geometrical, and magnetic parameters. To investigate the DW oscillations in detail, we studied the influence of these parameters on the oscillation frequency of DW. Firstly, we simulated the DW oscillations for various values of J at limiting values of d and i DMI constant (boundary values of the range where DW oscillation occurs). The results are summarized in Figure 5.14. In the DW pinning+ DW oscillation regime, the DW oscillations were observed from a current density of 5×10^{11} A/m² to 1×10^{12} A/m² for all the cases except for $d = 40$ nm and i DMI constant of -3 mJ/m². For $d = 40$ nm and i DMI constant of -3 mJ/m², the oscillations were observed from a J value of 2×10^{11} A/m² to 1×10^{12} A/m². However, for comparison, we have presented the results for a J from 5×10^{11} A/m² to 1×10^{12} A/m² in all the cases. The oscillation frequency increases linearly with the current density J for all the cases except for $d = 30$ nm and i DMI constant of -3 mJ/m². This result can be understood from Eq. 5.2, which states the linear increase in DW velocity with J . A faster-moving DW in the nanoscale i DMI region will oscillate faster and hence result in a higher oscillation frequency. However, for $d = 30$ nm and i DMI constant of -3 mJ/m², the d value may not be able to support the posed tilting of DW and oscillations are not happening in the equilibrium state. In this set of simulations, the frequency can be tuned from a minimum of 1.82 GHz to 4.19 GHz by tuning J . Based on these simulation results, we fixed the J value at 9×10^{11} A/m² for all subsequent simulations.

Figure 5.14. The frequency vs current density (J) graph for (c) $d = 30$ nm, (d) $d = 40$ nm, for two different i DMI values (limiting i DMI values for DW oscillations) of -2.5 and -3 mJ/m².

Figure 5.15. The plot of the frequency of DW oscillations as a function of d for i DMI values of -2.5 and -3 mJ/m².

In the next step, we simulated the dependence of the frequency of oscillation on the width of the nanoscale i DMI region (d). Figure 5.15 shows the dependence of oscillation frequency on different d values ranging from 30 to 40 nm. Here, we fixed the i DMI constant at -2.5 , -3 mJ/m², and J at 9×10^{11} A/m². As we can see, except for $d = 30$ nm, the frequency is independent of the d value. As $d = 30$ nm is the onset of DW oscillations and may not be able to induce the

energetically favored DW oscillations (i.e. DW may not reach the minimum energy state). These results suggest that the DW oscillations may not have a geometrical origin.

5.4.2. Dependence of Oscillation Frequency on Magnetic Parameters

The DW dynamics is strongly affected by the magnetic parameters. Therefore, it is important to study the dependence of the frequency of oscillations on magnetic parameters. All the simulations were performed for $d=40$ nm, $i\text{DMI}=-3$ mJ/m² and $J=9\times 10^{11}$ A/m². In the first case, we varied K_u for different values of M_s and keeping A fixed (as shown in Figure 5.16 (a)). The oscillation frequency increases with the increase in K_u . Moreover, the frequency decreases as M_s increases for every value of K_u . On contrary, when K_u was fixed at 4×10^5 J/m³ and A was increased, a decreasing trend of the frequency of oscillation was obtained for every simulated value of M_s . Like the previous case, the frequency decreases as saturation magnetization increases for every value of A (refer to Figure 5.16 (b)). Next, we have chosen the A and K_u values, which results in the higher frequencies, and performed the simulations for various values of M_s . The results are presented in Figure 15.6 (c). For $K_u=4\times 10^5$ J/m³ and $A=0.9\times 10^{11}$ J/m, we observed an oscillation for some simulation time and then depinning of the DW from $i\text{DMI}$ introduced region. By tuning the magnetic parameters, the frequency can be tuned from 2.8 GHz to 4.97 GHz. Ion implantation of non-magnetic species can be used to alter the magnetic parameters in experimental situations [5, 32, 82-88].

Figure 5.16. The dependence of the frequency of oscillations on (a) anisotropy constant (K_u) and (b) exchange constant (A) for various values of saturation magnetization (M_s) and (c) the plot for the frequency of oscillations as a function of saturation magnetization (M_s).

Recent studies pointed out that when an HM layer is deposited adjacent to the FM layer, K_u also gets modified [89-91]. The increase or decrease in K_u depends on the interface materials (HM/FM) used. Since we have introduced $iDMI$ in the nanoscale region, the local anisotropy (K_u^{DMI}) may also change. In order to incorporate all possible situations, we varied K_u^{DMI} for a large range of values for three different $iDMI$ constant values. The other parameters such as M_s , A , α , β were taken as 6.55×10^5 A/m, 1×10^{-11} J/m, 0.02, 0.02, respectively. The anisotropy constant outside the nanoscale $iDMI$ region was chosen as 4×10^5 J/m³. The results of this set of simulations are presented in Figure 5.17. In all simulations, the oscillations were observed only in a certain band of anisotropy values. The range of K_u , where oscillation occurs, was

found to be proportional to the value of $iDMI$. As the K_u value in the nanoscale $iDMI$ region increases, the frequency of oscillations decreases for all $iDMI$ values. The frequency could be tuned from 1.47 GHz to 3.96 GHz, by choosing a suitable K_u^{DMI} and $iDMI$ value. The increase in frequency with $iDMI$ value is consistent with the micromagnetic theory and will be discussed later. However, the change in anisotropy constant in the nanoscale region (K_u^{DMI}) forms another simultaneous energy well (or energy barrier), and with the increase in K_u^{DMI} , the height of the energy barrier increases, which may be a reason for the decreasing trend of oscillation frequency in this set of simulations.

Figure 5.17. The plot of the frequency of oscillation vs anisotropy constant in nanoscale $iDMI$ region (K_u^{DMI}) for different values of $iDMI$ strengths.

5.4.3. Analytical Model for Observed Domain Wall Oscillations

In order to understand the results in detail, we have modelled the dynamics of the DW with that of a single spin (spin at the center of DW) in a Zeeman field (analogous to H_{DMI}). Fig. 5.18 shows the schematic of the analytical model. The energy of the spin in this Zeeman-like field is given by [92, 93],

$$U = -\mu_0 (\vec{m} \cdot \vec{H}_{DMI}) \quad (5.5)$$

$$U = -\mu_0 m H_{DMI} \cos \theta \quad (5.6)$$

Figure 5.18. The conceptual schematic of our analytical model. The DW is considered as a single spin (black arrow in upper figure), which experiences a field due to iDMI in the nanoscale region.

Taylor's expansion of $\cos \theta$ is (for small θ) [94],

$$\cos \theta = 1 - \frac{\theta^2}{2} + \frac{\theta^4}{24} - \dots \quad (5.7)$$

Neglecting the higher-order terms and combining Eq. 5.6 and 5.7 will result in,

$$U = -\mu_0 m H_{DMI} \left(1 - \frac{\theta^2}{2}\right)$$

$$U = -\mu_0 m H_{DMI} + \mu_0 m H_{DMI} \frac{\theta^2}{2} \quad (5.8)$$

The first term is a constant, therefore U can be written as,

$$U \approx \frac{1}{2} (\mu_0 m H_{DMI}) \theta^2 \quad (5.9)$$

Which is of the form of $U = \frac{1}{2}kx^2$ and hence represents a harmonic oscillator equation. Note, here we are interested in angular displacement θ . Therefore, the i DMI does not change the magnitude of the spin, but only changes the angular orientation of the spin.

Following Eq. (5.9), the magnitude of the force acting on this spin becomes,

$$F = -(\mu_0 m H_{DMI})\theta \quad (5.10)$$

Which is of the form of $F=-kx$ with $k = \mu_0 m H_{DMI}$, All the other terms except H_{DMI} are constant, which means the restoring force for the oscillating DWs mainly depends on H_{DMI} . Here, we are considering the angular displacement of the center spin of the DW, where one end is pinned and only the other end is moving. The expression of F in Eq. 5.10 gives the force acting on the moving end of the center spin of DW. This simply will be a torque if we consider the overall effect on the center spin of DW. Following Eq. 5.10, the frequency of oscillation can be approximated as,

$$\omega \propto \sqrt{H_{DMI}} \text{ (with } \omega = 2\pi f \text{)} \quad (5.11)$$

Since STT drags the DW in the $+x$ direction (in our case), the force due to STT is in the $+x$ direction. At point “a”, the force due to DMI, $F_{DMI}=0$ (since $\theta=0$, from Eq. 5.10). Therefore, F_{STT} will drag the DW further, but with further movement, θ will become non-zero. Consequently, there will be a force F_{DMI} in the negative x -direction, which increases with θ . Hence, the DW will oscillate between the boundaries of these two forces.

In order to further verify our model, we simulated the event of DW oscillations for various values of i DMI constant, and the results are presented in Figure 5.19. The simulated frequencies were then fitted with the equation obtained from the analytical model. Except for the two limiting values of i DMI constants, the analytical model perfectly fits our simulated data. This

ensures the validity of our analytical model and speculation of considering the H_{DMI} field as restoring force.

Figure 5.19. The plot of frequency vs $(DMI)^{1/2}$ for two different d values. The discrete data points were fitted with the analytical model and exceptionally good fitting was found.

From micromagnetic theory, the $iDMI$ acts like an effective field, given by [75, 95],

$$\vec{H}_{DMI} = \frac{2D}{\mu_0 M_s} \left(\frac{\partial m_z}{\partial x}, \frac{\partial m_z}{\partial y}, -\frac{\partial m_x}{\partial x} - \frac{\partial m_y}{\partial y} \right) \quad (5.12)$$

Here, D is the DMI parameter. This field is an additive term in the expression of the total effective field in the nanoscale $iDMI$ region. The above equation can alternatively be written as [50, 96, 97],

$$H_{DMI} = \frac{2D}{\mu_0 M_s} \left(\frac{\sqrt{K_u - \left(\frac{\mu_0 M_s^2}{2} \right)}}{\sqrt{A}} \right) \quad (5.13)$$

Combining Eqs. 5.11 and 5.13 will result in,

$$f \propto \sqrt{\frac{2D}{\mu_0 M_s} \left(\frac{\sqrt{K_u - \left(\frac{\mu_0 M_s^2}{2}\right)}}{\sqrt{A}} \right)} \quad (5.14)$$

An independent increase in K_u , A , and M_s will result in an increase, decrease, and decrease in H_{DMI} and hence in the same in oscillation frequency. Our simulation results of the dependence of frequency on simulation parameters are in good agreement with Eq. 5.14.

It is a matter of utmost importance to comment on whether the micromagnetic parameters used in this study will lead to the formation of skyrmions in the nanoscale $iDMI$ region. For this purpose, we have calculated the critical DMI energy constant for a few different cases, as shown in Table 5.1 [98, 99]. The critical DMI energy constant is given by,

$$D_c = \frac{4\sqrt{AK}}{\pi} \quad (5.15)$$

1. For all the cases except one result presented in Figure 5.19 ($iDMI = -2.5 \text{ mJ/m}^2$, $A = 1 \times 10^{-11} \text{ J/m}$, $K_u = 4 \times 10^5 \text{ J/m}^3$), the calculated D_c is smaller than the $iDMI$ value used, which suggests that the simulated conditions do not support the formation of skyrmions.
2. For the case of $iDMI = -2.5 \text{ mJ/m}^2$, $A = 1 \times 10^{-11} \text{ J/m}$, $K_u = 4 \times 10^5 \text{ J/m}^3$, the calculated D_c is greater than $|D|$. However, we calculated the core radius of a single skyrmion, which is given by,

$$R = \frac{\Delta}{\sqrt{2\left(1 - \frac{D}{D_c}\right)}} \quad (5.16)$$

Here Δ is DW width. Moreover, the calculated R for this case was found to be 43.5 nm, which is larger than the width of $iDMI$ introduced region (“30 nm × 35 nm” and “30 nm × 40 nm”) [98, 99]. Hence, it can be stated that the simulation parameters used in this study do not support the formation of skyrmions.

Table 5.1. List of calculated critical DMI energy constant (D_c) for different A and K_u values used in the study.

No.	A (10^{-11} J/m)	K_u (10^5 J/m ³)	D_c (mJ/m ²)	Remarks
1	1	4	2.55	Typical condition (parameters are relevant to (Co/Ni) _n multilayer system) [76-81]
2	0.9	3.6	2.29	Minimum A and K_u (parameters are relevant to (Co/Ni) _n multilayers appropriate material engineering using mechanisms such as ion implantation or thermal annealing) [76-81, 100-102]
3	1.2	4	2.79	Maximum A and K_u (parameters are relevant to (Co/Ni) _n multilayers appropriate material engineering using mechanisms such as ion implantation or thermal annealing) [76-81, 100-102]

5.4.4. Proposed Device

Based on the findings of this study, we propose a device as shown in Figure 5.20. The device is composed of an FM nanowire with an HM layer crossbar in the vicinity of FM. A tunnel barrier and a reference FM layer with the perfect matching of spatial co-ordinates can be deposited at the other side of HM. The magnetization of the reference layer will be fixed and following the DW oscillation, a change in tunnel magnetoresistance (TMR) will happen. This change in TMR can be converted to a microwave signal using a dedicated electronic circuit. Thus, this device converts a DC voltage into microwave frequencies. Since the frequency of oscillation can be tuned from a minimum of 1.82 GHz to a maximum of 4.99 GHz by changing

the width of i DMI introduced region (d), the value of i DMI, current density (J), and magnetic parameters such as M_s , A , K_u , etc., a number of such oscillators can be fabricated in a single chip. Thus, a large range of frequencies can be obtained from a single chip.

Figure 5.20. The schematic presentation of the proposed oscillator with the HM crossbar at one side and MTJ at the other side of the FM nanowire.

5.5. Conclusions

In this chapter, we carried out detailed micromagnetic simulations to study the controlled DW motion through pinning and depinning in two separate designs namely, (a) square wave shaped nanowire and (b) nanowire with nanoscale *iDMI*. The former comes under the category of geometrical pinning sites. On contrary, the latter comes under the category of non-geometrical pinning sites. For the case of square wave shaped nanowires, we observed that the corner of the design pin the DW successfully. However, the window of current density values for which pinning strength is significant is quite small. This could limit the virtue of the proposed design. However, a fine-tuning of amplitude and pulse width of the applied current may result in the pinning of DW at the successive pinning sites.

Next, we studied a case, where *iDMI* has been used to form local magnetic modification. This local magnetic modification was expected to facilitate pinning for the DWs. For this purpose, we have introduced the *iDMI* in the nanoscale region. This local magnetic modification hosts DW pinning and DW oscillations for various values of *iDMI* strengths. We observed that DW dynamics can be divided into three distinct phases. For smaller values of *iDMI* and *d*, the DW was found to move from the left end of the nanowire and pin at the nanoscale *iDMI* region. An application of higher currents results in depinning of DWs from the pinning sites. On the other hand, for “*d*” values from 30 nm to 40 nm and *iDMI* strengths of -2.5 to -3 mJ/m², the DW oscillates in the nanoscale *iDMI* region. This regime is called, DW pinning+ DW oscillation regime. In this regime, the DWs were found to oscillate for a large range of current density values. The effective field from the *iDMI* was found to act as a restoring force for the DW. Moreover, the oscillation frequency can be scaled from 1.47 GHz to 4.97 GHz by adjusting the external stimulus, geometrical and magnetic parameters. The regime between these two was defined as an intermediate regime, where the DW only oscillates for a single *J* value. Thus, the proposed concept can host several devices on a single chip performing a range of functionalities

such as pinning of DW, oscillations generating a larger band of frequencies. The experimental realizations of the studied concepts will add more value to this study. We will briefly describe the design of experiments in chapter 8.

References

- [1] U. Bauer, S. Emori and G. S. D. Beach, "Voltage-controlled domain wall traps in ferromagnetic nanowires," *Nature Nanotechnology*, vol. 8, no. 6, pp. 411, 2013.
- [2] N. Lei, T. Devolder, G. Agnus, P. Aubert, L. Daniel, J.-V. Kim, W. Zhao, T. Trypiniotis, R. P. Cowburn, C. Chappert, D. Ravelosona and P. Lecoeur, "Strain-controlled magnetic domain wall propagation in hybrid piezoelectric/ferromagnetic structures," *Nature Communications*, vol. 4, no. 1, p. 1378, 2013.
- [3] F. Nasirpouri, S.-M. Peighambari-Sattari, C. Bran, E. M. Palmero, E. Berganza Eguiarte, M. Vazquez, A. Patsopoulos and D. Kechrakos, "Geometrically designed domain wall trap in tri-segmented nickel magnetic nanowires for spintronics devices," *Scientific reports*, vol. 9, no. 1, p. 9010, 2019.
- [4] H. W. Schumacher, D. Ravelosona, F. Cayssol, J. Wunderlich, C. Chappert, V. Mathet, A. Thiaville, J. Jamet, J. Ferre and R. J. Haug, "Propagation of a magnetic domain wall in the presence of AFM fabricated defects," *IEEE Transactions on Magnetics*, vol. 37, no. 4, pp. 2331, 2001.
- [5] D. Kumar, T. Jin, S. Al Risi, R. Sbiaa, W. Lew and S. Piramanayagam, "Domain wall motion control for racetrack memory applications," *IEEE Transactions on Magnetics*, vol. 55, no. 3, pp. 1, 2018.
- [6] S. S. Parkin, M. Hayashi and L. Thomas, "Magnetic domain-wall racetrack memory," *Science*, vol. 320, no. 5873, pp. 190, 2008.
- [7] K.-J. Kim, R. Hiramatsu, T. Koyama, K. Ueda, Y. Yoshimura, D. Chiba, K. Kobayashi, Y. Nakatani, S. Fukami and M. Yamanouchi, "Two-barrier stability that allows low-power operation in current-induced domain-wall motion," *Nature Communications*, vol. 4, no. 1, pp. 1, 2013.
- [8] K.-J. Kim, R. Hiramatsu, T. Moriyama, H. Tanigawa, T. Suzuki, E. Kariyada and T. Ono, "Tradeoff between low-power operation and thermal stability in magnetic domain-wall-motion devices driven by spin Hall torque," *Applied Physics Express*, vol. 7, no. 5, p. 053003, 2014.
- [9] E. Martinez, L. Lopez-Diaz, O. Alejos and L. Torres, "Thermally activated domain wall depinning in thin strips with high perpendicular magnetocrystalline anisotropy," *Journal of Applied Physics*, vol. 106, no. 4, p. 043914, 2009.
- [10] M. Kläui, H. Ehrke, U. Rüdiger, T. Kasama, R. E. Dunin-Borkowski, D. Backes, L. J. Heyderman, C. A. F. Vaz, J. A. C. Bland, G. Faini, E. Cambril and W. Wernsdorfer, "Direct observation of domain-wall pinning at nanoscale constrictions," *Applied Physics Letters*, vol. 87, no. 10, p. 102509, 2005.
- [11] D. Kumar, C. H. Jing, C. JianPeng, T. Jin, L. Sze Ter, R. Sbiaa and S. N. Piramanayagam, "Ultra-low Power Domain Wall Device for Spin-based Neuromorphic Computing," arXiv:2007.12357, 2020.
- [12] S. Konings, J. Miguel, J. Luigjes, H. Schlatter, H. Luigjes, J. Goedkoop and V. Gadgil, "Lock in of magnetic stripe domains to pinning lattices produced by focused ion-beam patterning," *Journal of Applied Physics*, vol. 98, no. 5, p. 054306, 2005.
- [13] T. Jin, F. Tan, C. C. I. Ang, W. Gan, J. Cao, W. S. Lew and S. Piramanayagam, "Tilted magnetisation for domain wall pinning in racetrack memory," *Journal of Magnetism and Magnetic Materials*, vol. 489, p. 165410, 2019.
- [14] T. Ono, H. Miyajima, K. Shigeto and T. Shinjo, "Magnetization reversal in submicron magnetic wire studied by using giant magnetoresistance effect," *Applied physics letters*, vol. 72, no. 9, pp. 1116, 1998.

- [15] T. Ono, H. Miyajima, K. Shigeto, K. Mibu, N. Hosoi and T. Shinjo, "Propagation of a magnetic domain wall in a submicrometer magnetic wire," *Science*, vol. 284, no. 5413, pp. 468, 1999.
- [16] M. Kläui, C. Vaz, J. Bland, W. Wernsdorfer, G. Faini and E. Cambril, "Domain wall pinning and controlled magnetic switching in narrow ferromagnetic ring structures with notches," *Journal of applied physics*, vol. 93, no. 10, pp. 7885, 2003.
- [17] M. Kläui, C. Vaz, W. Wernsdorfer, E. Bauer, S. Cherifi, S. Heun, A. Locatelli, G. Faini, E. Cambril and L. Heyderman, "Domain wall behaviour at constrictions in ferromagnetic ring structures," *Physica B: Condensed Matter*, vol. 343, no. 1-4, pp. 343, 2004.
- [18] M. Hayashi, L. Thomas, C. Rettner, R. Moriya, X. Jiang and S. S. Parkin, "Dependence of current and field driven depinning of domain walls on their structure and chirality in permalloy nanowires," *Physical review letters*, vol. 97, no. 20, p. 207205, 2006.
- [19] L. Bogart, D. Eastwood and D. Atkinson, "The effect of geometrical confinement and chirality on domain wall pinning behavior in planar nanowires," *Journal of Applied Physics*, vol. 104, no. 3, p. 033904, 2008.
- [20] D. Atkinson, D. Eastwood and L. Bogart, "Controlling domain wall pinning in planar nanowires by selecting domain wall type and its application in a memory concept," *Applied Physics Letters*, vol. 92, no. 2, p. 022510, 2008.
- [21] D. Bedau, M. Kläui, U. Rüdiger, C. A. Vaz, J. A. C. Bland, G. Faini, L. Vila and W. Wernsdorfer, "Angular dependence of the depinning field for head-to-head domain walls at constrictions," *Journal of applied physics*, vol. 101, no. 9, p. 09F509, 2007.
- [22] Y. Endo, Y. Mitsuzuka and M. Yamaguchi, "Study on the trapping of domain wall in an Ni-Fe nanowire with a constricted area," *IEEE transactions on magnetics*, vol. 46, no. 6, pp. 2413, 2010.
- [23] A. Himeno, K. Kondo, H. Tanigawa, S. Kasai and T. Ono, "Domain wall ratchet effect in a magnetic wire with asymmetric notches," *Journal of Applied Physics*, vol. 103, no. 7, p. 07E703, 2008.
- [24] M. Al Bahri and R. Sbiaa, "Geometrically pinned magnetic domain wall for multi-bit per cell storage memory," *Scientific reports*, vol. 6, p. 28590, 2016.
- [25] M. Al Bahri, B. Borie, T. L. Jin, R. Sbiaa, M. Kläui and S. Piramanayagam, "Staggered magnetic nanowire devices for effective domain-wall pinning in racetrack memory," *Physical Review Applied*, vol. 11, no. 2, p. 024023, 2019.
- [26] H. Mohammed, S. A. Risi, T. Jin, J. Kosel, S. Piramanayagam and R. Sbiaa, "Controlled spin-torque driven domain wall motion using staggered magnetic wires," *Applied Physics Letters*, vol. 116, no. 3, p. 032402, 2020.
- [27] S. Fukami, T. Suzuki, K. Nagahara, N. Ohshima and N. Ishiwata, "Large thermal stability independent of critical current of domain wall motion in Co/Ni nanowires with step pinning sites," *Journal of Applied Physics*, vol. 108, no. 11, p. 113914, 2010.
- [28] M. Yamanouchi, J. Ieda, F. Matsukura, S. Barnes, S. Maekawa and H. Ohno, "Universality classes for domain wall motion in the ferromagnetic semiconductor (Ga, Mn) As," *Science*, vol. 317, no. 5845, pp. 1726, 2007.
- [29] X. Zhang, N. Vernier, W. Zhao, H. Yu, L. Vila, Y. Zhang and D. Ravelosona, "Direct observation of domain-wall surface tension by deflating or inflating a magnetic bubble," *Physical Review Applied*, vol. 9, no. 2, p. 024032, 2018.
- [30] I. Polenciuc, A. Vick, D. Allwood, T. Hayward, G. Vallejo-Fernandez, K. O'Grady and A. Hirohata, "Domain wall pinning for racetrack memory using exchange bias," *Applied Physics Letters*, vol. 105, no. 16, p. 162406, 2014.

- [31] T. Jin, M. Ranjbar, S. He, W. C. Law, T. Zhou, W. Lew, X. Liu and S. Piramanayagam, "Tuning magnetic properties for domain wall pinning via localized metal diffusion," *Scientific reports*, vol. 7, no. 1, pp. 1, 2017.
- [32] T. Jin, D. Kumar, W. Gan, M. Ranjbar, F. Luo, R. Sbiaa, X. Liu, W. S. Lew and S. Piramanayagam, "Nanoscale compositional modification in Co/Pd multilayers for controllable domain wall pinning in racetrack memory," *physica status solidi (RRL)–Rapid Research Letters*, vol. 12, no. 10, p. 1800197, 2018.
- [33] T. Jin, W. Gan, F. Tan, N. Sernicola, W. S. Lew and S. Piramanayagam, "Synaptic element for neuromorphic computing using a magnetic domain wall device with synthetic pinning sites," *Journal of Physics D: Applied Physics*, vol. 52, no. 44, p. 445001, 2019.
- [34] A. Vansteenkiste, J. Leliaert, M. Dvornik, F. Garcia-Sanchez and B. Van Waeyenberge, "The design and verification of Mumax3," *AIP Advances*, vol. 4, no. 10, p. 107133, 2014.
- [35] A. Vansteenkiste and B. Van de Wiele, "MuMax: A new high-performance micromagnetic simulation tool," *Journal of Magnetism and Magnetic Materials*, vol. 323, no. 21, pp. 2585, 2011.
- [36] J. Leliaert, M. Dvornik, J. Mulkers, J. De Clercq, M. V. Milošević and B. Van Waeyenberge, "Fast micromagnetic simulations on GPU—recent advances made with Mumax³," *Journal of Physics D: Applied Physics*, vol. 51, no. 12, p. 123002, 2018.
- [37] G. S. Abo, Y. Hong, J. Park, J. Lee, W. Lee and B. Choi, "Definition of Magnetic Exchange Length," *IEEE Transactions on Magnetics*, vol. 49, no. 8, pp. 4937, 2013.
- [38] H. Neal Bertram and C. Seberino, "Numerical simulations of hysteresis in longitudinal magnetic tape," *Journal of Magnetism and Magnetic Materials*, vol. 193, no. 1, pp. 388, 1999.
- [39] I. S. Dubitskiy, A. V. Syromyatnikov, N. A. Grigoryeva, A. A. Mistonov, N. A. Sapozhnikova and S. V. Grigoriev, "Spin-ice behavior of three-dimensional inverse opal-like magnetic structures: Micromagnetic simulations," *Journal of Magnetism and Magnetic Materials*, vol. 441, pp. 609, 2017.
- [40] D. B. Gopman, C. L. Dennis, P. J. Chen, Y. L. Iunin, P. Finkel, M. Staruch and R. D. Shull, "Strain-assisted magnetization reversal in Co/Ni multilayers with perpendicular magnetic anisotropy," *Scientific Reports*, vol. 6, no. 1, p. 27774, 2016.
- [41] D. Kumar, P. Sengupta, R. Sbiaa and S. Piramanayagam, "Spin Transfer Torque Induced Domain Wall Oscillations in Ferromagnetic Nanowire with a Nanoscale Dzyaloshinskii–Moriya Interaction Region," *Journal of Magnetism and Magnetic Materials*, vol. 507, p. 166807, 2020.
- [42] F. Macià, P. Warnicke, D. Bedau, M. Y. Im, P. Fischer, D. A. Arena and A. D. Kent, "Perpendicular magnetic anisotropy in ultrathin Co/Ni multilayer films studied with ferromagnetic resonance and magnetic x-ray microspectroscopy," *Journal of Magnetism and Magnetic Materials*, vol. 324, no. 22, pp. 3629, 2012.
- [43] A. Thiaville, Y. Nakatani, J. Miltat and Y. Suzuki, "Micromagnetic understanding of current-driven domain wall motion in patterned nanowires," *EPL (Europhysics Letters)*, vol. 69, no. 6, p. 990, 2005.
- [44] J. Vandermeulen, S. A. Nasser, B. Van de Wiele, G. Durin, B. Van Waeyenberge and L. Dupré, "The effect of Dzyaloshinskii–Moriya interaction on field-driven domain wall dynamics analysed by a semi-analytical approach," *Journal of Physics D: Applied Physics*, vol. 49, no. 46, p. 465003, 2016.
- [45] S.-G. Je, D.-H. Kim, S.-C. Yoo, B.-C. Min, K.-J. Lee and S.-B. Choe, "Asymmetric magnetic domain-wall motion by the Dzyaloshinskii–Moriya interaction," *Physical Review B*, vol. 88, no. 21, p. 214401, 2013.

- [46] I. Dzyaloshinskii, "Thermodynamic theory of weak ferromagnetism in antiferromagnetic substances," *Soviet Physics JETP-USSR*, vol. 5, no. 6, pp. 1259, 1957.
- [47] T. Moriya, "Anisotropic superexchange interaction and weak ferromagnetism," *Physical review*, vol. 120, no. 1, p. 91, 1960.
- [48] F. Freimuth, S. Blügel and Y. Mokrousov, "Berry phase theory of Dzyaloshinskii–Moriya interaction and spin–orbit torques," *Journal of physics: Condensed matter*, vol. 26, no. 10, p. 104202, 2014.
- [49] E. Martinez, S. Emori and G. S. Beach, "Current-driven domain wall motion along high perpendicular anisotropy multilayers: The role of the Rashba field, the spin Hall effect, and the Dzyaloshinskii-Moriya interaction," *Applied Physics Letters*, vol. 103, no. 7, p. 072406, 2013.
- [50] A. Thiaville, S. Rohart, É. Jué, V. Cros and A. Fert, "Dynamics of Dzyaloshinskii domain walls in ultrathin magnetic films," *EPL (Europhysics Letters)*, vol. 100, no. 5, p. 57002, 2012.
- [51] M. Bode, M. Heide, K. Von Bergmann, P. Ferriani, S. Heinze, G. Bihlmayer, A. Kubetzka, O. Pietzsch, S. Blügel and R. Wiesendanger, "Chiral magnetic order at surfaces driven by inversion asymmetry," *Nature*, vol. 447, no. 7141, p. 190, 2007.
- [52] P. Ferriani, K. Von Bergmann, E. Vedmedenko, S. Heinze, M. Bode, M. Heide, G. Bihlmayer, S. Blügel and R. Wiesendanger, "Atomic-scale spin spiral with a unique rotational sense: Mn monolayer on W (001)," *Physical review letters*, vol. 101, no. 2, p. 027201, 2008.
- [53] S. Meckler, N. Mikuszeit, A. Preßler, E. Vedmedenko, O. Pietzsch and R. Wiesendanger, "Real-space observation of a right-rotating inhomogeneous cycloidal spin spiral by spin-polarized scanning tunneling microscopy in a triple axes vector magnet," *Physical review letters*, vol. 103, no. 15, p. 157201, 2009.
- [54] S. Heinze, K. Von Bergmann, M. Menzel, J. Brede, A. Kubetzka, R. Wiesendanger, G. Bihlmayer and S. Blügel, "Spontaneous atomic-scale magnetic skyrmion lattice in two dimensions," *Nature Physics*, vol. 7, no. 9, p. 713, 2011.
- [55] M. Belmeguenai, J.-P. Adam, Y. Roussigné, S. Eimer, T. Devolder, J.-V. Kim, S. M. Cherif, A. Stashkevich and A. Thiaville, "Interfacial Dzyaloshinskii-Moriya interaction in perpendicularly magnetized Pt/Co/AlO_x ultrathin films measured by Brillouin light spectroscopy," *Physical Review B*, vol. 91, no. 18, p. 180405, 2015.
- [56] R. Soucaille, M. Belmeguenai, J. Torrejon, J.-V. Kim, T. Devolder, Y. Roussigné, S.-M. Chérif, A. Stashkevich, M. Hayashi and J.-P. Adam, "Probing the Dzyaloshinskii-Moriya interaction in CoFeB ultrathin films using domain wall creep and Brillouin light spectroscopy," *Physical Review B*, vol. 94, no. 10, p. 104431, 2016.
- [57] D.-H. Kim, D.-Y. Kim, S.-C. Yoo, B.-C. Min and S.-B. Choe, "Universality of Dzyaloshinskii-Moriya interaction effect over domain-wall creep and flow regimes," *Physical Review B*, vol. 99, no. 13, p. 134401, 2019.
- [58] M. Perini, S. Meyer, B. Dupé, S. von Malottki, A. Kubetzka, K. von Bergmann, R. Wiesendanger and S. Heinze, "Domain walls and Dzyaloshinskii-Moriya interaction in epitaxial Co/Ir (111) and Pt/Co/Ir (111)," *Physical Review B*, vol. 97, no. 18, p. 184425, 2018.
- [59] A. Qaiumzadeh, I. A. Ado, R. A. Duine, M. Titov and A. Brataas, "Theory of the Interfacial Dzyaloshinskii-Moriya Interaction in Rashba Antiferromagnets," *Physical Review Letters*, vol. 120, no. 19, p. 197202, 2018.
- [60] T. Srivastava, M. Schott, R. Juge, V. Krizakova, M. Belmeguenai, Y. Roussigné, A. Bernand-Mantel, L. Ranno, S. Pizzini and S.-M. Chérif, "Large-voltage tuning of Dzyaloshinskii–Moriya interactions: A route toward dynamic control of skyrmion chirality," *Nano Letters*, vol. 18, no. 8, pp. 4871, 2018.

- [61] R. A. Gallardo, D. Cortés-Ortuño, R. E. Troncoso and P. Landeros, "Spin waves in thin films and magnonic crystals with Dzyaloshinskii-Moriya interactions," *arXiv:1903.04288*, 2019.
- [62] A. Fert, V. Cros and J. Sampaio, "Skyrmions on the track," *Nature nanotechnology*, vol. 8, no. 3, p. 152, 2013.
- [63] N. Romming, C. Hanneken, M. Menzel, J. E. Bickel, B. Wolter, K. von Bergmann, A. Kubetzka and R. Wiesendanger, "Writing and deleting single magnetic skyrmions," *Science*, vol. 341, no. 6146, pp. 636, 2013.
- [64] B. Dupé, M. Hoffmann, C. Paillard and S. Heinze, "Tailoring magnetic skyrmions in ultra-thin transition metal films," *Nature communications*, vol. 5, p. 4030, 2014.
- [65] S. D. Pollard, J. A. Garlow, J. Yu, Z. Wang, Y. Zhu and H. Yang, "Observation of stable Néel skyrmions in cobalt/palladium multilayers with Lorentz transmission electron microscopy," *Nature communications*, vol. 8, p. 14761, 2017.
- [66] I. M. Miron, K. Garello, G. Gaudin, P.-J. Zermatten, M. V. Costache, S. Auffret, S. Bandiera, B. Rodmacq, A. Schuhl and P. Gambardella, "Perpendicular switching of a single ferromagnetic layer induced by in-plane current injection," *Nature*, vol. 476, no. 7359, p. 189, 2011.
- [67] K.-S. Ryu, L. Thomas, S.-H. Yang and S. Parkin, "Chiral spin torque at magnetic domain walls," *Nature nanotechnology*, vol. 8, no. 7, p. 527, 2013.
- [68] S. Emori, U. Bauer, S.-M. Ahn, E. Martinez and G. S. Beach, "Current-driven dynamics of chiral ferromagnetic domain walls," *Nature materials*, vol. 12, no. 7, p. 611, 2013.
- [69] S.-H. Yang, K.-S. Ryu and S. Parkin, "Domain-wall velocities of up to 750 m s^{-1} driven by exchange-coupling torque in synthetic antiferromagnets," *Nature nanotechnology*, vol. 10, no. 3, p. 221, 2015.
- [70] J.-S. Kim, M.-A. Mawass, A. Bisig, B. Krüger, R. M. Reeve, T. Schulz, F. Büttner, J. Yoon, C.-Y. You and M. Weigand, "Synchronous precessional motion of multiple domain walls in a ferromagnetic nanowire by perpendicular field pulses," *Nature communications*, vol. 5, no. 1, p. 3429, 2014.
- [71] H. Kronmüller, "General Micromagnetic Theory and Applications," in *Materials Science and Technology*, 2020, pp. 1.
- [72] S. Sundara Mahalingam, B. V. Manikandan and S. Arockiaraj, "Review – Micromagnetic Simulation Using OOMMF and Experimental Investigations on Nano Composite Magnets," *Journal of Physics: Conference Series*, vol. 1172, no. 1, p. 012070, 2019.
- [73] I. Dzyaloshinsky, "A thermodynamic theory of "weak" ferromagnetism of antiferromagnetics," *Journal of Physics and Chemistry of Solids*, vol. 4, no. 4, pp. 241, 1958.
- [74] J. Leliaert, M. Dvornik, J. Mulkers, J. De Clercq, M. Milošević and B. Van Waeyenberge, "Fast micromagnetic simulations on GPU—recent advances made with," *Journal of Physics D: Applied Physics*, vol. 51, no. 12, p. 123002, 2018.
- [75] L. Buda-Prejbeanu, "Modelling of spintronic devices: from basic operation mechanisms toward optimization," Thesis Dissertation, 2017.
- [76] S. Gupta, R. Sbiaa, M. Al Bahri, A. Ghosh, S. Piramanayagam, M. Ranjbar and J. Akerman, "[Co/Ni] multilayers with robust post-annealing performance for spintronics device applications," *Journal of Physics D: Applied Physics*, vol. 51, no. 46, p. 465002, 2018.
- [77] S. Fukami, Y. Nakatani, T. Suzuki, K. Nagahara, N. Ohshima and N. Ishiwata, "Relation between critical current of domain wall motion and wire dimension in perpendicularly magnetized Co/Ni nanowires," *Applied Physics Letters*, vol. 95, no. 23, p. 232504, 2009.

- [78] C. Burrowes, A. P. Mihai, D. Ravelosona, J. V. Kim, C. Chappert, L. Vila, A. Marty, Y. Samson, F. Garcia-Sanchez, L. D. Buda-Prejbeanu, I. Tudosa, E. E. Fullerton and J. P. Attane, "Non-adiabatic spin-torques in narrow magnetic domain walls," *Nature Physics*, vol. 6, no. 1, pp. 17, 2010.
- [79] T. Koyama, D. Chiba, K. Ueda, K. Kondou, H. Tanigawa, S. Fukami, T. Suzuki, N. Ohshima, N. Ishiwata, Y. Nakatani, K. Kobayashi and T. Ono, "Observation of the intrinsic pinning of a magnetic domain wall in a ferromagnetic nanowire," *Nature Materials*, vol. 10, no. 3, pp. 194, 2011.
- [80] K. Yamada and Y. Nakatani, "Excitation of magnetic domain wall velocity in (Co/Ni) nanowires induced by blocking the motion of vertical Bloch lines," *Applied Physics Express*, vol. 8, no. 9, p. 093004, 2015.
- [81] T. Taniguchi, K.-J. Kim, T. Tono, T. Moriyama, Y. Nakatani and T. Ono, "Precise control of magnetic domain wall displacement by a nanosecond current pulse in Co/Ni nanowires," *Applied Physics Express*, vol. 8, no. 7, p. 073008, 2015.
- [82] R. L. Conte, E. Martinez, A. Hrabec, A. Lamperti, T. Schulz, L. Nasi, L. Lazzarini, R. Mantovan, F. Maccherozzi and S. Dhesi, "Role of B diffusion in the interfacial Dzyaloshinskii-Moriya interaction in Ta/Co 20 Fe 60 B 20/MgO nanowires," *Physical Review B*, vol. 91, no. 1, p. 014433, 2015.
- [83] L. H. Diez, M. Voto, A. Casiraghi, M. Belmeguenai, Y. Roussigné, G. Durin, A. Lamperti, R. Mantovan, V. Sluka and V. Jeudy, "Enhancement of the Dzyaloshinskii-Moriya interaction and domain wall velocity through interface intermixing in Ta/CoFeB/MgO," *Physical Review B*, vol. 99, no. 5, p. 054431, 2019.
- [84] J. Fassbender, J. Von Borany, A. Mücklich, K. Potzger, W. Möller, J. McCord, L. Schultz and R. Mattheis, "Structural and magnetic modifications of Cr-implanted Permalloy," *Physical Review B*, vol. 73, no. 18, p. 184410, 2006.
- [85] S. Gupta, D. Kumar, T. L. Jin, R. Nongjai, K. Asokan, A. Ghosh, M. Aparnadevi, P. Suri and S. Piramanayagam, "Broadband strip-line ferromagnetic resonance spectroscopy of soft magnetic CoFeTaZr patterned thin films," *AIP Advances*, vol. 8, no. 5, p. 056125, 2018.
- [86] D. Kumar, S. Gupta, T. Jin, R. Nongjai, K. Asokan and S. Piramanayagam, "Modification of structural and magnetic properties of masked Co–Pt films induced by high-energy ion implantation," *IEEE Magnetism Letters*, vol. 9, pp. 1, 2017.
- [87] D. Kumar, S. Gupta, T. Jin, R. Nongjai, K. Asokan and S. Piramanayagam, "Tailoring the structural and magnetic properties of masked CoPt thin films using ion implantation," *AIP Advances*, vol. 8, no. 5, p. 056504, 2018.
- [88] D. Kumar, S. Gupta, K. K. Tham, R. Nongjai, S. Saito, K. Asokan and S. Piramanayagam, "Ion-implantation-induced disorder in FePt-C thin films," *IEEE Transactions on Magnetism*, vol. 55, no. 3, pp. 1, 2018.
- [89] J. Cho, N.-H. Kim, J. Jung, D.-S. Han, H. J. Swagten, J.-S. Kim and C.-Y. You, "Thermal Annealing Effects on the Interfacial Dzyaloshinskii–Moriya Interaction Energy Density and Perpendicular Magnetic Anisotropy," *IEEE Transactions on Magnetism*, vol. 54, no. 6, pp. 1, 2018.
- [90] N.-H. Kim, D.-S. Han, J. Jung, K. Park, H. J. Swagten, J.-S. Kim and C.-Y. You, "Dependence of interfacial Dzyaloshinskii–Moriya interaction and perpendicular magnetic anisotropy on the thickness of the heavy-metal layer," *Applied Physics Express*, vol. 10, no. 10, p. 103003, 2017.
- [91] W.-Y. Kim, H. K. Gweon, K.-J. Lee and C.-Y. You, "Correlation between interfacial Dzyaloshinskii–Moriya interaction and interfacial magnetic anisotropy of Pt/Co/MgO structures," *Applied Physics Express*, vol. 12, no. 5, p. 053007, 2019.

- [92] B. D. Cullity and C. D. Graham, *Introduction to magnetic materials*. John Wiley & Sons, 2011.
- [93] J. M. Coey, *Magnetism and magnetic materials*. Cambridge university press, 2010.
- [94] <https://mathworld.wolfram.com/TaylorSeries.html>.
- [95] Z. Wang, B. Zhang, Y. Cao and P. Yan, "Probing the Dzyaloshinskii-Moriya interaction via the propagation of spin waves in ferromagnetic thin films," *Physical Review Applied*, vol. 10, no. 5, p. 054018, 2018.
- [96] T. H. Pham, J. Vogel, J. Sampaio, M. Vaňatka, J.-C. Rojas-Sánchez, M. Bonfim, D. Chaves, F. Choueikani, P. Ohresser and E. Otero, "Very large domain wall velocities in Pt/Co/GdOx and Pt/Co/Gd trilayers with Dzyaloshinskii-Moriya interaction," *EPL (Europhysics Letters)*, vol. 113, no. 6, p. 67001, 2016.
- [97] T. Koyama, Y. Nakatani, J. i. Ieda and D. Chiba, "Electric field control of magnetic domain wall motion via modulation of the Dzyaloshinskii-Moriya interaction," *Science advances*, vol. 4, no. 12, p. eaav0265, 2018.
- [98] R. Novak, F. Garcia, E. Novais, J. Sinnecker and A. Guimaraes, "Micromagnetic study of skyrmion stability in confined magnetic structures with perpendicular anisotropy," *Journal of Magnetism and Magnetic Materials*, vol. 451, pp. 749, 2018.
- [99] S. Rohart and A. Thiaville, "Skyrmion confinement in ultrathin film nanostructures in the presence of Dzyaloshinskii-Moriya interaction," *Physical Review B*, vol. 88, no. 18, p. 184422, 2013.
- [100] L. You, R. Sousa, S. Bandiera, B. Rodmacq and B. Dieny, "Co/Ni multilayers with perpendicular anisotropy for spintronic device applications," *Applied Physics Letters*, vol. 100, no. 17, p. 172411, 2012.
- [101] D. Stanescu, D. Ravelosona, V. Mathet, C. Chappert, Y. Samson, C. Beigné, N. Vernier, J. Ferré, J. Gierak and E. Bouhris, "Tailoring magnetism in CoNi films with perpendicular anisotropy by ion irradiation," *Journal of Applied Physics*, vol. 103, no. 7, p. 07B529, 2008.
- [102] S. Mendisch, V. Ahrens, M. Kiechle, A. Papp and M. Becherer, "Perpendicular Nanomagnetic Logic Based on Low Anisotropy Co\Ni Multilayer," *Journal of Magnetism and Magnetic Materials*, vol. 510, p. 166626, 2020.

Chapter 6. Observation of Ultra-low Pinning Field for Energy Efficient Domain Wall motion

Power consumption in moving the domain walls (DWs) is another obstacle in the progress of DW devices. To make the DW concept work, the power consumption while driving the DWs needs to be at least in a range of pJ- fJ per bit [1-3]. Therefore, we have attempted to address the problem of power consumption towards the end of this thesis. For this purpose, we have proposed a novel engineering of the W/CoFeB/MgO multilayer system and demonstrated DW motion at extremely low electrical currents. Several experiments were performed to study the physics behind the proposed concept. The in-depth details of these results will be presented in this chapter.

6.1. Introduction

The low operational power (or energy) is always a prerequisite of any device. In this way, DW devices are also under intense research intending to reduce the operational power consumption [3-6]. A significant portion of total operation power is used while moving the DWs in a nano-track. Here, our main focus is on reducing the power consumption in driving the DWs. Conventionally, an external magnetic field was proposed as a stimulus for driving the DWs along a nanowire [7-13]. However, this method is relatively higher power consuming [14]. Later, researchers proposed a few ways to reduce the power consumption for external magnetic field-driven DWs. One such method is based on applying a gate voltage in addition to the external magnetic field. When a gate voltage is applied, it reduces the perpendicular magnetic anisotropy (PMA) of the material [15, 16]. This reduction in PMA results in a decrease in the critical field for pushing the DWs and/or increases the DW velocity. Schellekens et al. reported an increase in the DW velocity driven by an external magnetic field in Co and CoB PMA nanowires when a gate voltage was applied [17]. After this, Lin et al. also demonstrated the increment in DW velocity from 0.4 to 4 mm/s upon the application of a gate voltage for

magnetic field-driven DW motion in Ta/CoFeB/MgO nanowires [18]. However, the reduction in total power consumption in moving the DWs is not significant. Besides, the external magnetic field produces non-local effects on magnetic elements. Therefore, the external magnetic field is inapt for large scale fabrication of DW devices on a single chip and hence ruled out [19].

Meanwhile, it was observed theoretically as well as experimentally that when an electrical current is passed through a ferromagnetic wire, the conduction electrons from the applied current quantum mechanically transfer the spin angular momentum to the spins of the ferromagnet. This phenomenon results in a torque called spin-transfer torque (STT), which can move the DWs in a ferromagnetic wire [20-22]. We have discussed the phenomenon of STT in the context of DW devices in detail in chapter 2. The STT can be decomposed into adiabatic and non-adiabatic torques. In the sole presence of adiabatic torque, the DW moves only above a threshold current density. On the other hand, non-adiabatic torque is a field-like torque and may result in the DW motion at relatively smaller current densities. The above discussion is attributed to the case of perfect wires i.e. wires with no defects (please refer to Figure 2.13 in chapter 2). The typical current density required to move the DWs in STT is of the order to $\sim 10^{11}$ - 10^{12} A/m² [23, 24], which is higher than the required values (usually 10^{10} A/m² and lower).

The idea of gate voltage was also explored for STT driven DWs. Ma et al. observed a decrease in the critical current density and an increase in the DW velocity when a gate voltage was applied simultaneously with the electrical current [25]. In an alternative study, the application of an external magnetic field of ~ 300 Oe along the easy axis direction was reported to reduce the critical current density from $\sim 4.5 \times 10^{11}$ to $\sim 3 \times 10^{11}$ A/m² [26]. In a separate approach, Kim et al. demonstrated that in a cross-bar geometry, a current in the cross-bar reduces the critical current in the memory element by one order [27]. The reduced energy consumption is still

higher than the current requirements (in a range of pJ-fJ). Therefore, deeper research is needed to address the issue of power consumption.

More recently, researchers have found out that the torque, called spin-orbit torque (SOT), from the adjacent heavy metal (HM) layer in HM/ ferromagnetic (FM) structures, can drive the DWs at even lesser powers [28-30]. As mentioned in chapter 2, the SOT originates from the bulk spin-Hall effect (SHE) in HM and interfacial Rashba-Edelstein effect (REE) at FM/HM interface [31]. The bulk SHE results in the generation of spin current from the charge current. The spin current diffuses into the FM layer and exerts a Slonczewski-like torque on the host spins. However, in material stacks with structural inversion asymmetry, the spin-orbit coupling at the FM/HM interface results in the non-equilibrium accumulation of spins at the interface. These accumulated spins exert a torque on the host spins through the Heisenberg exchange interaction and this effect is known as interfacial REE.

The efficiency of Slonczewski-like torque from SHE depends on the conversion efficiency of the charge current to spin current. This is defined as

$$\theta_{SH} = \frac{|J_s|}{|J_c|} \quad (6.1)$$

and referred to as the spin-Hall angle (SHA or θ_{SH}) [32]. Therefore, a larger spin-Hall angle is a choice for increasing SOT efficiency and hence lowering the current density (and hence operational power) for moving the DWs. Several HM element based spin Hall materials such as Au, Nb, Mo, Hf, Pd, Pt, Ta, W, $W_{1-x}Ta_x$, PtMn have been researched for enhancing θ_{SH} [33-36]. Amongst these materials, a thin W layer with a metastable β -phase (A15 crystal structure) has been the choice of many research studies due to a relatively higher θ_{SH} [33-35, 37-40]. An increment in the resistivity of β -W in a range of 100-300 $\mu\Omega$ cm has been considered as the basis for increasing the θ_{SH} . Researchers have studied several methods such as tuning the deposition parameters (deposition at high pressure and low power), oxygen doping in W layer and alloying W with other heavy metals for increasing the resistivity and hence θ_{SH} [34, 37,

39]. Most of the studies reported the θ_{SH} value in a range of ~ -0.2 to ~ -0.6 [33-35, 37-40]. Figure 6.1 (a) shows a graph of θ_{SH} as a function of the resistivity for the available literature [33-54]. One can see in Figure 6.1 (a) that a higher θ_{SH} is reported for the resistivity values of $\sim 200 \mu\Omega \text{ cm}$ and higher.

Figure 6.1. (a) The graph of spin-Hall angle vs resistivity (reported in relevant studies). A higher θ_{SH} value at minimum possible resistivity is desired for low power DW motion in DW devices and (b) the schematics of conventional and proposed approaches.

However, high resistivity is not favorable, as it will lead to higher power consumption. Moreover, it may inherently add several other issues such as high Joule heating, higher intrinsic defects in the FM layer, and lower tunneling magnetoresistance (TMR). To avoid this situation, we proposed a novel material engineering involving two W spin Hall layers. A low-pressure deposited W thin film, which is expected to reduce the total resistivity [55], was sandwiched between β -W layer and FM layer. The schematic of the design is presented in Figure 6.1 (b). This material engineering was expected to solve the above-mentioned problems while keeping the SHA reasonably high. Before discussing our study, here, we are listing some benchmarking studies that were carried out to reduce the critical current for driving domain walls/ magnetization switching.

Table 6.1. List of recent studies towards reducing the current density for driving the DWs or magnetization switching.

Study	Magnetic Material	Spin Hall Layer	Current Density	Assisting Magnetic Field (H)	Remarks
T. Koyama & T. Ono et al. [26]	(Co/Ni) _n		$\sim 10^{11}$ A/m ²	~ 30 mT	J was reduced from $\sim 4.5 \times 10^{11}$ to $\sim 3 \times 10^{11}$ A/m ² after applying a magnetic field of ~ 30 mT
S. DuttaGupta & H. Ohno et al. [23]	Ta/CoFeB/MgO		$\sim 10^{11}$ A/m ²		Adiabatic STT drives the DWs
Y. Fan & K. Wang et al. [56]	Topologic al Insulator (TI)	Al ₂ O ₃ /Cr-TI/GaAs interface and Cr-TI layer	$\sim 10^8$ A/m ²	zero and non-zero in-plane (IP) H	Gate voltage assisted SOT induced magnetization switching
P. Filippou & S. Parkin et al. [57]	Heusler Films	CoZ' (Z': Al, Ga, Ge, Sn)	$\sim 10^{10}$ - 10^{11} A/m ²		Chiral DWs, driven by electrical current (combined STT and SOT effects)
S. Shi & H. Yang et al. [58]	Ni ₈₁ Fe ₁₉	WTe ₂	$\sim 10^9$ A/m ²		SOT driven magnetization switching
S. Yang, C Garg, & S. Parkin [59]	Synthetic antiferromagnets	Pt and Ru	$\sim 10^{11}$ A/m ²		Overall DW dynamics induced by chiral exchange drag between two current-driven DWs
T. Okuno & T. Ono et al. [60]	GdFeCo Ferrimagnets		$\sim 10^{10}$ A/m ²	~ 85 mT	Non-adiabatic spin transfer torque driven DW motion. The DW motion was assisted by a magnetic field along the easy axis.
S. Vélez & P. Gambardella et al. [61]	Magnetic Insulators	Pt	$\sim 10^{10}$ A/m ²	Switching is assisted by IP H	SOT-driven DW motion with high mobility in Pt/Tm ₃ Fe ₅ O ₁₂ structures
Z. Luo & L. Heyderman et al. [62]	Pt/Co/AlO _x System	Pt	$\sim 10^{11}$ A/m ²		SOT-driven chiral DW motion to perform relevant logic operations

6.2. Optimization of Thickness for Dual Spin Hall Layers

As stated briefly, the low deposition power (P_W) and high Ar gas pressure (p_{Ar}) during deposition of W thin films are required to host the β -phase in W, a prerequisite for high SOT efficiency [37, 38, 40, 41, 53]. However, low P_W and low p_{Ar} during the deposition of W films result in the W films in α or $(\alpha+\beta)$ -phase, which exhibits lower θ_{SH} [55]. In the front of resistivity values, the resistivity of β -W is much higher than that of α -W. To optimize the deposition conditions of W thin film in dual W layer structure, we first prepared and characterized samples with single W layer deposited at low (LP) as well as high (HP) Ar gas pressures. We deposited Si/SiO₂/ W [LP and HP] (t nm)/ Co₄₀Fe₄₀B₂₀ (1 nm)/ MgO (~1 nm)/ Ru (2 nm) stack. Here “ t ” was varied from 2 nm to 8 nm. An Ar gas pressure of 0.78 mTorr and 3.63 mTorr was used during the deposition of LP and HP W thin films, respectively. A low deposition power of 100 W was employed for the deposition of both LP and HP W films. This low deposition power resulted in a very low sputtering rate of 0.03 nm/s. The layer structure (deposition conditions and thickness of individual layers) deposited on the top of LP as well as HP W layer was kept the same in all the samples. After depositing the samples, we characterized them to check the required properties.

6.2.1. Magnetic Measurements

In the first step, we performed the magnetic hysteresis (M-H) loop measurements for all the samples using a vibrating sample magnetometer (VSM). This is to check the presence of perpendicular magnetic anisotropy (PMA) in our samples because the presence of PMA is the first hint (but not the only requirement) of possible β -phase in the W films [63]. Figure 6.2 (a)-(d) shows the in-plane (IP) and out-of-plane (OOP) M-H loops for the LP W (t nm)/X samples, where X stands for Co₄₀Fe₄₀B₂₀/MgO/Ru layers. All the samples exhibit the presence of PMA in as-deposited (ASD) films as the remanence in the OOP direction is larger than that in the IP direction. For a clearer view, we have plotted the -zoomed in- images of OOP M-H loops in

the inset for all the cases. There is clear evidence of variation in coercivity as a function of LP W layer thickness. To see the trend closer, we plotted the variation in coercivity as a function of LP W layer thickness and the results are presented in Figure 6.4. The coercivity first decreases as the LP W layer thickness increases from 2 nm to 4 nm and then increases with the further increase in the thickness.

Figure 6.2. The IP and OOP M-H loops of samples with (a) LP W (2 nm), (b) LP W (4 nm), (c) LP W (6 nm), and LP W (8 nm) spin Hall layers.

In the next step, we measured M-H loops for the HP W (t nm)/X samples, and the results are shown in Figure 6.3 (a)- (d). More interestingly, samples with HP W layer thickness exhibit PMA until 6 nm. However, in the sample HP W (8 nm)/X, the PMA is absent. In other words, there is a transition of anisotropy from OOP direction to IP direction, when the thickness of the

HP W layer is increased beyond 6 nm. The coercivity *vs* thickness of the HP W layer plot (Figure 6.4) shows the same trend as observed for LP W (*t* nm)/ X samples. However, a point by point comparison of coercivity states that the coercivity for HP W/X samples is larger than that in LP W/X samples. This is because the HP deposition conditions result in a more granular structure than that at LP [64, 65]. This granularity is transferred to the thin magnetic layer. The increased granularity decreases the exchange coupling between the magnetic grains and hence results in larger coercivity. Moreover, as the thickness of the W layer increases, the exchange decoupling between the grains in the magnetic layer increases. This could be a reason for the increasing trend of coercivity as a function of W layer thickness.

Figure 6.3. The IP and OOP magnetic hysteresis loops of samples with (a) HP W (2 nm), (b) HP W (4 nm), (c) HP W (6 nm), and HP W (8 nm) layers.

Figure 6.4. The plot of coercivity as a function of LP/HP W layer thickness.

6.2.2. Electrical Measurements

After confirming the presence of PMA in our films (except HP W (8 nm)/X sample), we measured the stack resistivity of all the samples. The stack resistivity (ρ) value has been reported to identify the crystalline phase of the W films in the W/X structure [63]. To estimate the resistivity, we measured sheet resistance using the standard four-probe method [48]. The methodology of the measurement process has been discussed in section 3.6 of chapter 3. Figure 6.5 (a) shows the ρ vs LP W layer thickness plot for single W layer samples. As the LP W layer thickness increases, the resistivity first increases and then decreases. A maximum resistivity was estimated for the sample with LP W (4 nm) spin Hall layer. A stack resistivity of $120\ \mu\Omega\ \text{cm}$ is considered as the separation line from β (A15 crystal structure) and α -phase (body-centered cubic) [63]. Therefore, only LP W (4 nm)/X sample exhibits the β -phase. However, samples LP W (2 nm)/X and LP W (6 nm)/X are in the mixed ($\alpha+\beta$)-phase. The resistivity of the LP W (8 nm)/X sample is extremely small and is expected to be in α -phase.

Figure 6.5. The graph of resistivity vs W layer thickness for (a) LP W and (b) HP W spin Hall layers.

Following the previous measurements, we measured the resistivity of HP W (t nm)/X samples in the next step. Figure 6.5 (b) shows the plot of resistivity vs HP W layer thickness for HP W (t nm)/X. One can see that for “ t ” values of 2, 4, 6 nm, the stack exhibits β -phase. However, the W layer in HP W (8 nm)/X samples is in α -phase. Thus, the transition from complete β -phase to α -phase happens when “ t ” becomes larger than 6 nm. After analyzing the results of magnetic and electrical measurements, one can see that when “ t ” becomes larger than 6 nm, the sample properties change drastically. Therefore, we fixed the total thickness of the dual W layer structure at 6 nm. Moreover, we varied the individual thicknesses of HP and LP W layers (keeping the total thickness at 6 nm). The deposition parameters for HP and LP W were kept the same as used during the deposition of HP and LP W layer depositions in single layer samples. We first performed the characteristic measurements at the thin-film level to test the efficiency of samples. Based on our results at the thin-film level, we fabricated devices for DW motion experiments. Further, we also fabricated the meander microwires on the best dual W layer samples and examined the controlled DW motion in these microwires.

6.3. Results on Dual Spin Hall Layers

We deposited Si/SiO₂/ HP W (t nm)/ LP W (6- t nm)/CoFeB (1 nm)/MgO (~1 nm)/ Ru (2 nm) samples where $t = 2, 3, 4$ nm. We deposited a few samples with intermediate t values (wherever required). We will discuss our results on the thin films as well as devices in the upcoming sub-sections.

6.3.1. Characterizations of Thin Films

In the first step, we performed atomic force microscopy (AFM) experiments on the dual W layer samples. Moreover, we scanned the surfaces of samples with LP W (6 nm) and HP W (6 nm) layers for comparison purposes. We scanned the sample surface of dimension $1\text{ }\mu\text{m} \times 1\text{ }\mu\text{m}$ at different places on every sample. Moreover, we performed AFM measurements on various samples of the same conditions and observed almost similar topography. Here, we present one AFM scan from each sample in Figure 6.6 (a)- (e). In order to better compare the results, we have set the color bar in the same range for all the samples. In going from the sample with LP W (6 nm) layer, the red area and/or the intensity of red color in red regions increases except for the case of the sample with HP W (6 nm) layer. This suggests the increase in the surface roughness as the thickness of the HP W layer increases.

To better understand the results, we extracted the surface roughness from corresponding AFM images using Gwyddion computer software [66]. These results of surface roughness *vs* HP W layer thickness are shown in Figure 6.7. The estimated surface roughness of LP W (6 nm)/X and HP W (6 nm)/X samples were (0.11 ± 0.02) nm and (0.15 ± 0.04) nm, respectively. In the dual-layer samples, the surface roughness linearly increases with the increase in the thickness of the HP W layer. Moreover, this linearity holds when compared with LP W (6 nm)/X and HP W (6 nm)/X samples. With the increase in the HP W layer thickness, the granularity increases and hence will result in an increasing trend in surface roughness [65]. Even though the surface roughness increases with the increase in HP W layer thickness, the surface roughness values

(0.11 nm to 0.17 nm) are suitable enough for usage in commercial devices with magnetic tunnel junction (MTJ) in the vicinity for writing and reading purposes [67-69].

Figure 6.6. The AFM scans of samples with dual W layer of different thickness combinations. For comparison, we have scanned the samples with single W (6 nm) layers deposited at low and high pressure respectively.

Figure 6.7. The surface roughness extracted from AFM images as a function of the thickness of the HP W layer.

Figure 6.8. The magnetic hysteresis loops of samples with (a) HP W (2 nm)/LP W (4 nm), (b) HP W (3 nm)/LP W (3 nm) and (c) HP W (4 nm)/LP W (2 nm) spin-Hall layers. (d) The coercivity vs HP W layer thickness in dual-layer structure. For comparison, we have kept the data points for samples with single LP and HP W (6 nm) samples, respectively.

In the next step, we performed the VSM measurements of samples with dual W spin Hall layers in IP as well as OOP directions, and the results are shown in Figure 6.8 (a)-(c). Similar to the single-layer samples, the PMA is present in all the dual-layer samples in the ASD state. Moreover, the coercivity takes a similar relation as surface roughness with the HP W layer thickness. This confirms our explanation of granularity in HP W films. This granularity is transferred to the magnetic layer on the top and gives rise to an increase in the exchange decoupling between the magnetic grains. Moreover, the presence of PMA also hints towards the presence of β -phase in our dual-layer samples (as discussed before). To get concrete

confirmation of the presence of β -phase in samples with dual W layers, we performed the sheet resistance (and hence resistivity) measurements on these samples.

Figure 6.9 shows the results of resistivity measurements for our samples with dual W layer. For comparison, we have also provided the resistivity values of samples with LP W (6 nm) and HP W (6 nm) spin Hall layers. In this case, we have measured the resistivity of samples with $t = 2, 2.5, 3, 3.5, 3.8,$ and 4 nm (additional intermediate t values) in HP W(t nm)/LP W ($6-t$ nm)/X stack, with a purpose to know the trend of resistivity graph as a function of HP W layer thickness in dual-layer structure. A maximum resistivity of $166 \mu\Omega \text{ cm}$ was observed for the sample HP W (2.5 nm)/LP W (3.5 nm)/X. The resistivity first increases with the increase in the HP layer thickness ($t_{HP \text{ W}}$ in HP W (t nm)/LP W ($6-t$ nm)/X stack) and reaches a maximum when $t_{HP \text{ W}}$ becomes almost equal to $t_{LP \text{ W}}$. Beyond this point the resistivity decreases with the increase in $t_{HP \text{ W}}$. Moreover, the dual-layer sample with $t_{HP \text{ W}} = 2.5$ nm and 3 nm exhibits resistivity much larger than $120 \mu\Omega \text{ cm}$ and hence are ascertained in β -phase. On the other hand, HP W (2 nm)/LP W (4 nm)/X and HP W (3.5 nm)/LP W (2.5 nm)/X are expected to be in mixed ($\alpha+\beta$)-phase as their resistivity values are in the vicinity of $120 \mu\Omega \text{ cm}$ separation line. Since we want to carry on with the dual-layer sample having the optimized resistivity as well as β -phase in spin Hall layer. Therefore, we have chosen HP W (3 nm)/LP W (3 nm)/X sample for further experiments at the thin film and device level. We have also carried out experiments on HP W (2.5 nm)/LP W (3.5 nm)/X samples and will discuss them wherever needed.

Figure 6.9. The graph of resistivity vs thickness of HP W layer in samples dual W spin Hall layers.

Since the modulation in the nature of the W underlayer influences the magnetic properties, we intended to have a closer look at the magnetization reversal process in our samples. For this, we used the Kerr microscopy and recorded the domain patterns while sweeping the OOP magnetic field. The results are presented in Figure 6.10. For all the samples, the magnetization reversal is mediated by DW motion. However, the nature of DW motion is slightly different. This could be attributed to an increase in granularity with an increase in HP W layer thickness. For LP W (6 nm) and HP W (3 nm)/ LP W (3 nm) samples, the complete reversal (after nucleating a reversed region) happens in a narrow field range of 1.8-1.9 Oe. Whereas, the reversal of HP W (6 nm) sample takes over a field range of 8 Oe. In addition, the LP W (6 nm) and HP W (3 nm)/ LP W (3 nm) samples show fewer nucleation sites and the HP W (6 nm) sample shows the most nucleation sites. These results highlight that LP W (6 nm) and HP W (3 nm)/ LP W (3 nm) samples have lesser defects in comparison to HP W (6 nm) sample.

Figure 6.10. Domain evolution while sweeping the OOP magnetic field for thin films (a) LP W (6 nm), (b) HP W (3 nm)/ LP W (3 nm), and (c) HP W (6 nm) with spin Hall layer.

6.3.2. Characterizations of the Domain Wall Devices

To study the DW dynamics upon the application of an electrical current, we fabricated the DW devices of dimensions $50\ \mu\text{m} \times 10\ \mu\text{m}$ as shown in Figure 6.11 (a). First, we fabricated the microwires using optical lithography and ion milling process. Subsequently, we fabricated the electrode pads using the second step of optical lithography, electrode material (Ta (5 nm)/Cu (90 nm)/Ta (5 nm)) deposition, and lift-off process. Once the devices were ready, the electrical connections were made using the wire-bonding process. We have used Keithly 6221 current source for applying the current pulses of $5\ \mu\text{m}$ width and various amplitudes. A Kerr microscopy with polar and longitudinal imaging capabilities was employed to image the domains after the application of current pulses. Based on our results at the thin-film level, we

have chosen HP W (3 nm)/ LP W (3 nm)/X and HP W (2.5 nm)/LP W (3.5 nm)/X samples for device fabrication. Moreover, we fabricated devices from HP W (6 nm)/X and LP W (6 nm)/X samples for comparison purposes. We will discuss the observed results of current-driven DW motion experiments in upcoming sections.

We will first start with the discussion of results of devices made up of LP W (6 nm)/X samples. In order to perform the DW motion experiments, we, first, saturated the device with a large OOP magnetic field of +500 Oe. Subsequently, we applied a nucleation field of -21 Oe for nucleating a negative domain in the microwire. In all the measurements, we were not able to nucleate a reversed domain in the microwire. The domain switches very fast covering the whole microwire upon application of nucleation field. Taking this as the initial state, we started applying an increasing OOP magnetic field. At every magnetic field value, we sent several pulses of current. As one can see in Figure 6.11 (b), there is an emergence of the opposite domain after applying 20 pulses of the current density of magnitude $1 \times 10^9 \text{ A/m}^2$. Here, white and gray colors correspond to magnetization in the -ve and +ve z -axes, respectively. Upon application of subsequent pulses, the movement of the opposite domain is almost negligible. In the next step, we increased the amplitude of J to $1 \times 10^{11} \text{ A/m}^2$. Moreover, we supported the DW motion with a +ve OOP magnetic field. In this case, also, the DW motion and/or complete magnetization switching is not significant.

Figure 6.11. (a) The schematic of the DW device for DW motion measurements. The Kerr microscopy images illustrating the DW motion in the LP (6 nm)/X sample at a current density of (b) 1×10^9 and (c) $1 \times 10^{11} \text{ A/m}^2$. Similarly, the magnetic images of DW motion in HP W (6 nm)/X sample at a J of (d) 1×10^9 and (e) $1 \times 10^{11} \text{ A/m}^2$. Note an OOP magnetic field (as illustrated along with figures) was applied in addition to current pulses.

Subsequently, we performed measurements on the devices made from the HP W (6 nm)/X samples. Similar to the previous case, we first saturated the devices with a large OOP magnetic

field of +500 Oe. For nucleating the negatively magnetized domain, we applied a reversal field of -93 Oe. However, we observed a fast switching of the magnetic domain upon application of the nucleation field in this case as well. Again taking this as the initial state, we applied an increasing +ve magnetic field to supplement the current induced DW motion. Unlike the previous case, magnetization switching does happen in this case. However, the switching is ultrafast and observable DW motion is lacking. Moreover, the OOP magnetic field needed to switch the magnetization is lower for higher values of current densities. These results are presented in Figure 6.11 (d)- (e). Thus, we can say that in the devices made up of a single spin-Hall layer (LP as well as HP), the observable DW motion is lacking.

Continuing the observations in devices with a single spin-Hall layer, we then moved to the devices with dual spin-Hall layers. We first performed measurements on the devices from HP W (3 nm)/LP W (3 nm)/X samples. After saturating the sample with a large +ve OOP magnetic field of 500 Oe, we applied a reversed nucleation field of -17 Oe. Unlike the previous cases, we observed the nucleation of the negatively magnetized domain (as shown in Figure 6.12 (a), very first image). When we applied the current pulses with a magnitude of 1×10^7 A/m², the DW moves in a direction opposite to the current flow. Note, a constant OOP magnetic field of -10 Oe (much smaller than the coercivity of the device (38 Oe)) was applied during the measurements. A total of 37 pulses were applied to push the DW from the upper end (initial position) to the lower end of the device. When we increased the amplitude of the current density to 1×10^9 A/m² (keeping all other conditions identical to the previous case), the DW moves faster and reaches the other end in only 14 pulses. A higher current density of 1×10^{11} A/m² reduces the number of current pulses to only 5 for achieving the same. Thus, for devices with dual spin-Hall layer, a multidomain state was observed. Moreover, a systematic DW motion at very low current densities was observed. These results are novel and inspired us to systematically study the electrical current induced DW motion in dual-W layer samples.

Figure 6.12. The illustration of DW motion in HP W (3 nm)/LP W (3 nm)/X sample for a current density of (a) -1×10^7 , (b) -1×10^9 , and (c) $-1 \times 10^{11} \text{ A/m}^2$. An OOP magnetic field of -10 Oe was applied in all the cases.

In order to systematically study the electrical current induced DW motion in HP W (3 nm)/LP W (3nm)/X devices, we varied the OOP magnetic field and the current densities values to a larger range. We then processed the raw Kerr microscopy images to estimate the DW displacements at various times. From this DW displacement vs time (rather real-time) data, we calculated the DW velocity for all the studied cases. In the first case, we only applied current in both negative (down to up, Figure 6.13 (a) and (b)) as well as positive (up to down, Figure 6.13 (c) and (d)) directions while keeping the external magnetic field zero. The DW does not move after the application of several current pulses. Then we applied a non-zero OOP magnetic field of -5 Oe and repeated the experiments. For +ve J values, the DW does not move at all, while for -ve J values of -1×10^{11} and $-5 \times 10^{11} \text{ A/m}^2$, the DW moves and non-zero DW velocity

was observed. Below this value (in -ve J axis), the DW does not move. When the external magnetic field was increased to -8 Oe, a non-zero DW velocity was observed for current densities larger than -1×10^{10} A/m². Therefore, one can see that with the increase in the magnitude of the magnetic field by 3 Oe, the critical current density to drive the DWs can be reduced by an order of 10^1 . When we increased the external magnetic field to -10 Oe, the DW moves for all the negative current densities. Moreover, the DW moves for the positive J values smaller than 1×10^{11} A/m². These results indicate that the constant negative magnetic field supports the negative current densities and opposes the positive current densities.

To further understand the results, we reversed the direction of the applied magnetic field. For a magnetic field of +10 Oe, the DW was found to move only for the current density of 1×10^{11} A/m² and higher. Upon increasing the magnetic field to 17 Oe, the critical current density can be reduced to 1×10^8 A/m². A further increase in the magnetic field to 18 Oe results in the DW motion for the current densities as low as 1×10^6 A/m². We can see that the +ve magnetic field supports the +ve J and opposes to -ve J . These results are summarized in Figure 6.13 (e)- (h). One possible reason for asymmetric OOP magnetic field values (in -ve and +ve directions) needed for DW motion is a small offset (2-3 Oe) in current (applied to electromagnet) to magnetic field conversion. It should also be noted that the nucleation field needed in positive direction was 27 Oe. From these results, one can say that the electrical current in the dual W layer generates a SOT, which acts as an effective OOP magnetic field, and is either being supported or opposed by the external magnetic field. Moreover, the direction of DW motion as shown in Figure 6.13 (a)- (d) agrees with SOT-driven DW motion in previous reports [70-72].

Figure 6.13. The Kerr microscopy images of HP W (3 nm)/ LP W (3 nm)/X devices showing no DW motion for (a) $H_z = 0$ Oe, $J = -1 \times 10^6$ A/m² and (c) $H_z = 0$ Oe, $J = 1 \times 10^6$ A/m² and DW motion in a direction opposite to the direction of the current density J for (b) $H_z = -10$ Oe, $J = -1 \times 10^6$ A/m² and (d) $H_z = 18$ Oe, $J = 1 \times 10^6$ A/m². (e-f) The DW velocity vs J curve for all J and H_z values studied for HP W (3 nm)/ LP W (3 nm)/X devices. The switching field vs current density plot for (i-j) LP W (6 nm)/X and (k-l) HP W (6 nm)/X devices.

When a sufficient amount of magnetization in the FM layer is orthogonal to the spin current direction, the SOT acts as an effective field and reads for the following form [70],

$$H_{SOT} = -\frac{\hbar\theta_{SH}J}{2|e|M_s t}(\hat{m} \times \hat{\sigma}) \quad (6.2)$$

Here, $\hbar$, θ_{SH} , J , e , M_s and t are reduced Planck constant, spin-Hall angle, current density in heavy metal (HM) layer, electronic charge, saturation magnetization, and the thickness of FM layer, respectively. Besides, $\hat{m}$ and $\hat{\sigma}$ stand for the direction of magnetization inside the DW and spin current. If we consider the direction of the magnetization at the center of our down-up DW is along the +x direction and the direction of the spin current will be in -y-direction (for -ve J). Then considering the -ve sign of spin-Hall angle for W, the magnetic field will be in -z-direction (refer to Figure 6.15). Upon reversing the direction of J , the direction of the SOT effective field will be reversed. Our results are in agreement with this hypothesis. One reason for the magnetization at the center of the down-up DW in +x direction is the presence of DMI which is given by [73],

$$E_{DMI} = -\vec{D}_{ij} \cdot (\vec{S}_i \times \vec{S}_j) \quad (6.3)$$

in our samples. Here D_{ij} is the DMI constant, S_i and S_j are the neighboring spin vectors. This interfacial DMI value stabilizes a left-handed chiral DW ($\downarrow \rightarrow \uparrow$) at the upper end of the microwire. Moreover, the sign of DMI in our samples is +ve, which is in agreement with several reports of the past [41, 74, 75].

We have also performed detailed experiments for the devices with single spin-Hall layers. Since we were not able to observe the nucleation of the reversed domain as well as systematic DW motion in these cases, we listed down the magnetic field values at which the magnetization switches from one direction to the other direction for all the J values. Figure 6.13 (i-j) and (k-l) summarize the results for the LP W (6 nm)/X and HP W (6 nm)/X devices, respectively. The

dotted line in these figures represents the coercive field i.e. the field at which the magnetization switches without any external current. We nucleated the domains in both +ve and -ve directions. As one can see that for smaller current densities (less than $1 \times 10^9 \text{ A/m}^2$), the switching field is exactly equal to the coercive field. For higher currents, the switching happens at lower fields than the coercive field. However, these values are quite close to the coercive field. Although the SOT effective field explanation holds for the HP W (6 nm)/X devices, switching field values suggests that the switching is dominated by the magnetic field. However, for LP W (6 nm)/X devices, the magnetic field seems the main cause for magnetization switching or/and partial switching.

Figure 6.14. The schematic illustration of the action of the effective SOT field in the present study.

6.3.3. Underlying Mechanisms of Observed Low Current Density

To further understand our observations in devices with HP W (3 nm)/LP W (3 nm) dual-spin Hall layer, we further characterized our thin film samples and devices. The observation of such a low current density can be due to (a) an enhanced spin Hall angle or (b) a reduced pinning field in the HP W (3 nm)/ LP W (3 nm) sample or (c) both. To clarify, we measured the spin Hall angle of our devices using the harmonic Hall technique in the first step. To do so we fabricated Hall bar devices using optical lithography and ion milling processes. The spin Hall angle for HP W (3 nm)/ LP W (3 nm) device was found to be -0.14, which is comparable to the existing literature (for comparison, please refer to Figure 6.1 (a)). This value is approximately 1.5- 3 $\times$ larger than our samples with single spin Hall layers (LP W (6 nm) or HP W (6 nm)). These experiments suggest that the observed reduction in current density in our dual-layer device cannot be only due to moderate increment in spin Hall angle value (compared to single spin Hall layer devices). Thus, we measured the pinning field in LP W (6 nm), HP W (3 nm)/ LP W (3 nm), and HP W (6 nm) samples in the next step.

To measure the pinning field, we studied DW dynamics as a function of the OOP magnetic field. For pinning field measurements, we prepared special devices using optical lithography and ion milling. From a 1 cm $\times$ 1 cm sample, we fabricated microwires with various widths. The lengths of all the microwires were varied from 500 μ m to 2000 μ m in the steps of 500 μ m. This is to facilitate the observable DW motion in samples where sharp magnetization switching/ DW motion occurs. A large space between a consecutive group of devices was left for easy nucleation of reversed domains. The optical image of the fabricated devices (devices with length 2000 μ m) is shown in Figure 6.15 (a).

At first, the samples were saturated with a +ve OOP magnetic field of 500 Oe. Then a reversed magnetic field was applied until the nucleation at one edge of the device happens. Subsequently, magnetic field pulses of a certain amplitude were applied, and the DW

displacement was measured. We carried out the experiments for the microwires with widths 10, 20, and 25 μm . Figure 6.15 (b)- (d) show the DW motion at a certain field for the 10 μm wires. From the DW displacement results (for various magnetic fields), we plotted the DW velocity vs OOP magnetic field curves. The data were fitted to obtain the pinning strength. The results of these experiments will be discussed in the upcoming sub-section.

Figure 6.15. (a) The optical image of the devices utilized for pinning field measurements. Illustration of OOP magnetic field-driven DW motion in (b) LP W (6 nm), (c) HP W (3 nm)/LP W (3 nm), and (d) HP W (6 nm) devices. Here, black and grey colors represent the magnetizations down and up, respectively.

As shown in Figure 6.15 (b), LP W (6 nm) sample exhibits the largest DW velocity. However, HP W (6 nm) sample exhibits the smallest DW velocity. Moreover, the OOP magnetic field needed to move the DW in HP W (6 nm) samples is almost 4× larger than LP W (6 nm) and LP W (3 nm)/ HP W (3 nm) samples (Figure 6.15 (d)). Although, the DW velocity in LP W (3 nm)/ HP W (3 nm) dual-layer sample is comparable to that in HP W (6 nm) samples, the OOP magnetic field to achieve this is much smaller (Figure 6.15 (c)).

From DW velocity vs OOP magnetic field plots (Figure 6.16 (a)), we can see that the pinning field for LP W (3 nm)/ HP W (3 nm) sample is 20 Oe, which is comparable to that of LP W (6 nm) sample (15 Oe). However, this pinning field value is much smaller than that of HP W (6 nm) (46 Oe). Note, the pinning field value obtained for the HP W (6 nm) sample is comparable to the values reported in the literature [26, 76, 77]. It is important to note that the values obtained for HP W (6 nm) samples are underestimated. This is because we could not observe a clear DW motion for higher OOP magnetic fields (H_z). In the next step, we calculated the pinning potential in our samples using [78],

$$\ln(v) = \ln(v_0) + \left[\left(-\frac{U_c}{k_B T} \right) (H_{dep})^{\frac{1}{4}} \right] H_z^{-\frac{1}{4}} \quad (6.4)$$

As shown in Figure 6.16 (b), the linear fits of $\ln(v)$ vs $H_z^{-1/4}$ graphs help us in calculating the pinning potential (U_c). HP W (6 nm) and LP W (6 nm) samples had the highest and lowest pinning potentials of 1.6×10^{-19} J and 4.8×10^{-20} J, respectively. Whereas, HP W (3 nm)/ LP W (3 nm) had a pinning potential of about 6.1×10^{-20} J, much lower than that of HP W (6 nm) (and other material systems [79, 80]). To establish our findings with the devices of width 10 μm , we repeated the analysis for the devices with widths of 20 and 25 μm . The results are presented in Figure 6.16 (c)- (f). The findings are similar to that obtained for 10 μm devices. The calculated pinning field and pinning potential values are summarized in table 6.2.

Figure 6.16. The graph of DW velocity vs OOP magnetic field for devices with the width of (a) 10, (c) 20, and (e) 25 μm . The curves are fitted to find out the pinning field for the samples under study. The graph of $\ln(v)$ vs $H_z^{-1/4}$ for devices with the width of (b) 10, (d) 20, and (f) 25 μm . The curves are fitted to find out the pinning potential for the samples under study.

Table 6.2. The list of H_{dep} , and U_c values for different studied devices.

	Device width= 10 μm		Device width= 20 μm		Device width= 25 μm	
Sample	H_{dep} (Oe)	U_c (J) $\times 10^{-20}$	H_{dep} (Oe)	U_c (J) $\times 10^{-20}$	H_{dep} (Oe)	U_c (J) $\times 10^{-20}$
LP W (6 nm)	15	4.8	16	4.8	15	4.3
HP W (3 nm)/ LP W (3 nm)	20	6.1	20	5.6	19	5.2
HP W (6 nm)	46	16.2	44	16	42	17

To understand the reason for the low pinning field in LP W (6 nm) and HP W (3 nm)/ LP W (3 nm) samples, we performed pinning site density measurements in our samples. For these measurements, we measured the hysteresis loops of our samples for different sweeping rates using the polar magneto-optic Kerr effect. We fixed the step size between the consecutive fields and varied the time (delay time) between the consecutive fields. The coercivity decreases as the delay time increases, which is an expected result [81]. From these results, we plotted coercivity as a function of $\ln(dH/dt)$, as shown in Figure 6.17 (a), and fitted with a linear relationship, given by [82, 83],

$$H_c \left(\frac{dH}{dt} \right) = \frac{k_B T}{v_a M_s} \ln \left(\frac{dH}{dt} \right) + constant \quad (6.5)$$

Here, v_a is the activation volume, T is temperature, and k_B is the Boltzmann constant. Other parameters have their usual meanings (mentioned before in this text). From activation volume, the pinning site density can be calculated from [81-83],

$$\rho_{pin} = t_{tot}/v_a \quad (6.6)$$

Here, t_{tot} is the total thickness of the film. Using these relationships, we calculated the pinning site densities for LP W (6 nm), HP W (3 nm)/ LP W (3 nm), and HP W (6 nm) samples, and the results are presented in Figure 6.17 (b). These results indicate that the pinning site density is maximum for HP W (6 nm) sample and minimum for LP W (6 nm) sample. Additionally, pinning site density exhibits intermediate value for HP W (3 nm)/ LP W (3 nm) sample.

Moreover, the estimated pinning site density values are much lower compared to the similar material system [81]. A closer look at the depinning field and pinning site density data suggests that HP W (3 nm)/ LP W (3 nm) sample has lower pinning sites compared to the conventional HP W (6 nm) sample. Moreover, the strength of pinning sites is much lower for HP W (3 nm)/ LP W (3 nm) samples compared to that of HP W (6 nm) sample.

Figure 6.17. (a) The graph of coercivity vs $\ln(dH/dt)$ for LP W (6 nm), HP W (3 nm)/ LP W (3 nm), and HP W (6 nm) samples. (b) The illustration of pinning field density for samples with single and dual-spin Hall layers.

Next, we imaged domain maps of our thin-film samples at a demagnetized state using polar Kerr microscopy. We first saturated the samples using a large (500 Oe) OOP magnetic field and captured the reference image. Subsequently, we demagnetized the sample (starting field= ~ 320 Oe) and captured the domain images (subtracted from reference image). We captured the images at four different places (as shown in Figure 6.18, four different sampling areas) and observed similar domain patterns. From these images, we calculated the domain, and the results are presented in Figure 6.19. A decreasing trend of domain size (and increasing trend of the number of domains) suggests that the DW mediated magnetization reversal is more dominant

for LP W (6 nm) and HP W (3 nm)/ LP W (3 nm) samples. Moreover, there are lesser pinning defects for the case of LP W (6 nm) and HP W (3 nm)/ LP W (3 nm) samples. These results are in good agreement with our previous thin film and device-level results. Therefore, the low current for driving the DWs in the dual-layer sample is attributed to the low pinning field strength, low pinning site density, and moderate value of the spin Hall angle.

Figure 6.18. The domain maps of (a) LP W (6 nm), (b) HP W (3 nm)/ LP W (3 nm), and (c) HP W (6 nm) samples at demagnetized state.

Figure 6.19. The illustration of domain width, deducted from domain maps at demagnetized state, for samples with single and dual-spin Hall layers.

Figure 6.20. The hysteresis loop of the HP W (3 nm)/ LP W (3 nm) device. Note the dimensions of the device are $10 \mu\text{m} \times 50 \mu\text{m}$.

Since the current driven DW motion experiments are supported by the OOP magnetic field, therefore, it is important to understand the role of the OOP magnetic field. For this, we measured the hysteresis loops of the devices. The result is shown in Figure 6.20. As we can see that the OOP magnetic field needed to support the current-driven DW motion in HP W (3 nm)/ LP W (3 nm) sample is much smaller than the device coercivity. Therefore, the OOP magnetic

field of -10 Oe (and 18 Oe) alone should not drive the DW in current-driven DW motion experiments. Moreover, the DW velocity increases as the current increases, which implies that the DW motion is the combined result of current as well as applied OOP magnetic field. At the fields higher than -10 Oe (which is -12 Oe), we observed the current driven DW motion and the results are quite similar except the occurrence of a higher DW velocity at every current value (with respect to the -10 Oe case). Besides, the current driven DW motion experiments are reproducible, which we have tested repeatably on different sets of devices.

Figure 6.21. The plot of coercivity vs current density when an electrical current was applied together with the magnetic field during $M(I)$ - H measurements for the time between consecutive pulses of (a) 0.5 s and (b) 50 ms.

Another concern that may arise is that the thermal energy generated due to the application of current may affect the DW motion results. To address this, we measured $M(I)$ - H loops. In these measurements, while sweeping the OOP magnetic field we applied current pulses of a certain amplitude at every field value. The time between consecutive current pulses was kept at 0.5 and 0.05 s. Moreover, the number of current pulses at every field value was varied to several values as shown in Figure 6.21. For the experiments where more than one pulses are applied at

every field value, the field was applied during all the pulses and then the corresponding data point was collected. The time interval between consecutive fields was also kept the same as the one for consecutive current pulses.

As we can see, the coercivity remains the same at all current densities except $5 \times 10^{11} \text{ A/m}^2$ for a time difference of 0.5 s. This implies the thermal effect at current densities from 10^6 A/m^2 to 10^{11} A/m^2 is negligible. However, for the time difference of 0.05 s, the coercivity decreases (for all J) as the number of current pulses increases. This indicates that thermal energy plays a role and changes the magnetic properties slightly. However, our current driven DW motion experiments resembles to the case “time between consecutive pulses= 0.5 s and one current pulse at every field value”. Therefore, we can say that the thermal effects in our current-driven DW motion experiments are negligible.

6.3.4. Comparison of Energy Consumption

To compare the energy consumption in the proposed scheme as compared to the recent literature, we calculated the energy required in moving one bit by a distance of 20 nm (futuristic length of one bit). The energy consumption can be written as [84],

$$E = I^2 R t \quad (6.7)$$

Where $I = J \times A$, with A as the area of cross-section of the device. R is the resistance of the device and t is the time required to move the DW by 20 nm and can be estimated from, $t = s/v$. Here, $s = 20 \text{ nm}$, and v is the DW velocity. Upon inserting,

$$E = \frac{(JA)^2 R s}{v} \quad (6.8)$$

The values of J , A , and v were picked from the highlighted recent studies, and energy E was calculated. These calculated values are tabulated in table 6.3. For the adiabatic torque from STT driven DW motion in Ta/CoFeB/MgO system, the energy consumed to displace the DW by 20 nm was calculated to be $\sim 0.2 \mu\text{J}$ [23]. However, the same was reduced to $\sim 1 \text{ nJ}$ for SOT-driven DW motion in Ta/CoFeB/MgO system [85]. An IP magnetic field of magnitude 300 Oe

was applied to achieve SOT driven DW motion in this study. The moderate values of damping and field-like torque efficiencies (both -0.056 ± 0.004) resulted in reduced energy consumption in this study. Moreover, the driving mechanisms are entirely different in the above-mentioned systems (references 23 and 85). Note, in systems for SOT-driven DW motion the thickness of the SOT layer is much larger than the one used in reference 23 (thickness of Ta layer: (a) in reference 23, it is 0.5 nm, and (b) in reference 85, it is 8 nm). This allows more current to pass through the SOT layer.

In a case, when the Ta/CoFeB/MgO system is replaced by W-Hf/CoFeB/MgO system, the E reduces to ~ 0.4 nJ [85]. A reduced pinning field and an increased damping-like torque efficiency (-0.143 ± 0.015) in this sample (compared to Ta/CoFeB/MgO in reference 85) is the reason for such a decrement in the energy consumption. The ferrimagnetic systems are the topic of hot research these days, therefore, we also calculated the energy consumption in displacing the DW by 20 nm in such systems, which was found to be ~ 2 pJ [3]. For this particular calculation, we have chosen the CoGe ferrimagnetic system, in which a high DW velocity of ~ 6 km/s was observed [3].

Table 6.3. A comparison of energy required to displace a bit by 20 nm in listed systems with the given driving force.

Driving Force	Order of Energy/Bit
Adiabatic torque	100 nJ [23]
SOT in Ta/CoFeB/MgO system	nJ [85]
SOT in W-Hf/CoFeB/MgO system	100 pJ [85]
SOT in Ferrimagnets	pJ [3]
Present study	aJ*

During all the above calculations, we have taken the maximum DW velocity and corresponding current density value. The distance “ s ” was considered 20 nm in all the calculations. Therefore,

from the DW velocity values, we calculated “ t ” for respective cases using $t = s/v$. The geometrical parameters for DW devices were taken from respective references. Besides, we have assumed the resistance of DW devices as 1 k Ω in all the cases. However, for energy consumption calculation in the present study, we have taken the experimentally measured value of DW device resistance. We calculated the energy consumption in moving the DWs by 18.6 μm (considering the least count of our measurement system) in the present study. The calculated energy was found to be 0.4 fJ, which is $\sim 10^4$ times less than the current state-of-the-art. When we extrapolated this calculation for a displacement of 20 nm, the energy consumption reduced to 0.6 aJ.

6.4. Conclusions

In this chapter, we have studied a novel material engineering in W/CoFeB/MgO based system and reported a reduction in critical current density by $\times 10^4$ compared to the state-of-the-art. These results are extremely impactful for applications of DW devices in DW-based racetrack memory as well as in emerging neuromorphic technology (this aspect will be discussed in detail in the next chapter). The optimization of dual-layer stacks was performed using structural, magnetic, and electrical characterizations at the thin-film level. The insertion of the LP W (3 nm) layer between HP W (3 nm) spin-Hall layer and ferromagnetic CoFeB layer significantly reduces the critical current density for driving the DWs. Such a reduction in current density could be attributed to an enhancement in spin Hall angle of dual-spin Hall layer or reduction in pinning field of FM layer or both. Further experimental investigations revealed that it is a reduction in pinning field in addition to moderate enhancement in spin Hall angle compared to conventional HP W (6 nm)/ X samples which cause the observation of DW motion at such low currents. It was also revealed that the thermal energy due to applied current does not affect the DW motion, especially at low currents. In conclusion, this study is important for low power DW devices and opens the doors for its possible applications in racetrack memory and/or neuromorphic devices.

References

- [1] S. Fukami, M. Yamanouchi, K.-J. Kim, T. Suzuki, N. Sakimura, D. Chiba, S. Ikeda, T. Sugibayashi, N. Kasai and T. Ono, "20-nm magnetic domain wall motion memory with ultralow-power operation," in *2013 IEEE International Electron Devices Meeting*, 2013, pp. 3.
- [2] K.-J. Kim, S. K. Kim, Y. Hirata, S.-H. Oh, T. Tono, D.-H. Kim, T. Okuno, W. S. Ham, S. Kim and G. Go, "Fast domain wall motion in the vicinity of the angular momentum compensation temperature of ferrimagnets," *Nature materials*, vol. 16, no. 12, pp. 1187, 2017.
- [3] K. Cai, Z. Zhu, J. M. Lee, R. Mishra, L. Ren, S. D. Pollard, P. He, G. Liang, K. L. Teo and H. Yang, "Ultrafast and energy-efficient spin-orbit torque switching in compensated ferrimagnets," *Nature Electronics*, vol. 3, no. 1, pp. 37, 2020.
- [4] Y. Wang and H. Yu, "An ultralow-power memory-based big-data computing platform by nonvolatile domain-wall nanowire devices," in *International Symposium on Low Power Electronics and Design (ISLPED)*, 2013, pp. 329.
- [5] L. Sánchez-Tejerina, V. Puliafito, P. K. Amiri, M. Carpentieri and G. Finocchio, "Dynamics of domain-wall motion driven by spin-orbit torque in antiferromagnets," *Physical Review B*, vol. 101, no. 1, p. 014433, 2020.
- [6] C. He, S. A. Razavi, G. Yu, X. Ma, H. Wu, Q. Shao, K. L. Wong, S. Shen, Y. Zhao and Y. Pei, "Study of the perpendicular magnetic anisotropy, spin-orbit torque, and Dzyaloshinskii-Moriya interaction in the heavy metal/CoFeB bilayers with Ir₂₂Mn₇₈ insertion," *Applied Physics Letters*, vol. 116, no. 24, p. 242407, 2020.
- [7] G. S. Beach, C. Nistor, C. Knutson, M. Tsoi and J. L. Erskine, "Dynamics of field-driven domain-wall propagation in ferromagnetic nanowires," *Nature materials*, vol. 4, no. 10, pp. 741, 2005.
- [8] K. Sixtus and L. Tonks, "Propagation of large Barkhausen discontinuities," *Physical Review*, vol. 37, no. 8, p. 930, 1931.
- [9] R. De Blois, "Magnetization processes: reversals and losses," *Journal of Applied Physics*, vol. 29, no. 3, pp. 459, 1958.
- [10] D. Smith, "A magnetic shift register employing controlled domain wall motion," *IEEE Transactions on Magnetics*, vol. 1, no. 4, pp. 281, 1965.
- [11] D. Herman Jr, B. Argyle, P. Trouilloud, B. Petek, L. Romankiw, P. Andricacos, S. Krongelb, D. Rath, D. Canaperi and M. Komsa, "Study of field-driven wall-configuration conversions for laminated Permalloy in the easy-axis state," *Journal of Applied Physics*, vol. 63, no. 8, pp. 4036, 1988.
- [12] K. Klaassen and J. Van Peppen, "Field-driven domain-wall jumps in thin-film heads," *Journal of Applied Physics*, vol. 69, no. 8, pp. 5417, 1991.
- [13] W. Li, Z. Chen, D. Wen, D. Chen, Z. Fan, M. Zeng, X. Lu, X. Gao and M. Qin, "Rotating magnetic field driven antiferromagnetic domain wall motion: Role of Dzyaloshinskii-Moriya interaction," *Journal of Magnetism and Magnetic Materials*, vol. 497, p. 166051, 2020.
- [14] N. Ben-Romdhane, W. Zhao, Y. Zhang, J.-O. Klein, Z. Wang and D. Ravelosona, "Design and analysis of racetrack memory based on magnetic domain wall motion in nanowires," in *2014 IEEE/ACM International Symposium on Nanoscale Architectures (NANOARCH)*, 2014: IEEE, pp. 71.
- [15] C. Song, B. Cui, F. Li, X. Zhou and F. Pan, "Recent progress in voltage control of magnetism: Materials, mechanisms, and performance," *Progress in Materials Science*, vol. 87, pp. 33, 2017.

- [16] U. Bauer, M. Przybylski and G. S. Beach, "Voltage control of magnetic anisotropy in Fe films with quantum well states," *Physical Review B*, vol. 89, no. 17, p. 174402, 2014.
- [17] A. Schellekens, A. Van den Brink, J. Franken, H. Swagten and B. Koopmans, "Electric-field control of domain wall motion in perpendicularly magnetized materials," *Nature communications*, vol. 3, no. 1, pp. 1, 2012.
- [18] W. Lin, N. Vernier, G. Agnus, K. Garcia, B. Ocker, W. Zhao, E. E. Fullerton and D. Ravelosona, "Universal domain wall dynamics under electric field in Ta/CoFeB/MgO devices with perpendicular anisotropy," *Nature communications*, vol. 7, no. 1, pp. 1, 2016.
- [19] S. S. Parkin, M. Hayashi and L. Thomas, "Magnetic domain-wall racetrack memory," *Science*, vol. 320, no. 5873, pp. 190, 2008.
- [20] A. Thiaville, Y. Nakatani, J. Miltat and Y. Suzuki, "Micromagnetic understanding of current-driven domain wall motion in patterned nanowires," *EPL (Europhysics Letters)*, vol. 69, no. 6, p. 990, 2005.
- [21] P. Freitas and L. Berger, "Observation of s-d exchange force between domain walls and electric current in very thin Permalloy films," *Journal of Applied Physics*, vol. 57, no. 4, pp. 1266, 1985.
- [22] C. Y. Hung and L. Berger, "Exchange forces between domain wall and electric current in permalloy films of variable thickness," *Journal of Applied Physics*, vol. 63, no. 8, pp. 4276, 1988.
- [23] S. DuttaGupta, S. Fukami, C. Zhang, H. Sato, M. Yamanouchi, F. Matsukura and H. Ohno, "Adiabatic spin-transfer-torque-induced domain wall creep in a magnetic metal," *Nature Physics*, vol. 12, no. 4, pp. 333, 2016.
- [24] S. Fukami, T. Suzuki, Y. Nakatani, N. Ishiwata, M. Yamanouchi, S. Ikeda, N. Kasai and H. Ohno, "Current-induced domain wall motion in perpendicularly magnetized CoFeB nanowire," *Applied Physics Letters*, vol. 98, no. 8, p. 082504, 2011.
- [25] C. Ma, T. Jin, X. Liu and S. N. Piramanayagam, "Switching domain wall motion on and off using a gate voltage for domain wall transistor applications," *Applied Physics Letters*, vol. 113, no. 23, p. 232401, 2018.
- [26] T. Koyama, K. Ueda, K.-J. Kim, Y. Yoshimura, D. Chiba, K. Yamada, J.-P. Jamet, A. Mougin, A. Thiaville and S. Mizukami, "Current-induced magnetic domain wall motion below intrinsic threshold triggered by Walker breakdown," *Nature nanotechnology*, vol. 7, no. 10, p. 635, 2012.
- [27] K.-J. Kim, Y. Yoshimura, W. S. Ham, R. Ernst, Y. Hirata, T. Li, S. Kim, T. Moriyama, Y. Nakatani and T. Ono, "Energy-efficient writing scheme for magnetic domain-wall motion memory," *Applied Physics Express*, vol. 10, no. 4, p. 043002, 2017.
- [28] I. M. Miron, T. Moore, H. Szambolics, L. D. Buda-Prejbeanu, S. Auffret, B. Rodmacq, S. Pizzini, J. Vogel, M. Bonfim and A. Schuhl, "Fast current-induced domain-wall motion controlled by the Rashba effect," *Nature materials*, vol. 10, no. 6, pp. 419, 2011.
- [29] X. Zhao, Y. Liu, D. Zhu, M. Sall, X. Zhang, H. Ma, J. Langer, B. Ocker, S. Jaiswal and G. Jakob, "Spin-orbit torque driven multi-level switching in He⁺ irradiated W-CoFeB-MgO Hall bars with perpendicular anisotropy," *Applied Physics Letters*, vol. 116, no. 24, p. 242401, 2020.
- [30] P. Haazen, E. Murè, J. Franken, R. Lavrijsen, H. Swagten and B. Koopmans, "Domain wall depinning governed by the spin Hall effect," *Nature materials*, vol. 12, no. 4, pp. 299, 2013.

- [31] R. Ramaswamy, J. M. Lee, K. Cai and H. Yang, "Recent advances in spin-orbit torques: Moving towards device applications," *Applied Physics Reviews*, vol. 5, no. 3, p. 031107, 2018.
- [32] Y. Wang, P. Deorani, X. Qiu, J. H. Kwon and H. Yang, "Determination of intrinsic spin Hall angle in Pt," *Applied Physics Letters*, vol. 105, no. 15, p. 152412, 2014.
- [33] Q. Hao, W. Chen and G. Xiao, "Beta (β) tungsten thin films: Structure, electron transport, and giant spin Hall effect," *Applied Physics Letters*, vol. 106, no. 18, p. 182403, 2015.
- [34] X. Sui, C. Wang, J. Kim, J. Wang, S. Rhim, W. Duan and N. Kioussis, "Giant enhancement of the intrinsic spin Hall conductivity in β -tungsten via substitutional doping," *Physical Review B*, vol. 96, no. 24, p. 241105, 2017.
- [35] Y. Takeuchi, C. Zhang, A. Okada, H. Sato, S. Fukami and H. Ohno, "Spin-orbit torques in high-resistivity-W/CoFeB/MgO," *Applied Physics Letters*, vol. 112, no. 19, p. 192408, 2018.
- [36] R. Bansal, G. Nirala, A. Kumar, S. Chaudhary and P. Muduli, "Large spin hall angle in β -W thin films grown on CoFeB without oxygen plasma," *Spin*, vol. 8, no. 04, p. 1850018, 2018.
- [37] C.-F. Pai, L. Liu, Y. Li, H. Tseng, D. Ralph and R. Buhrman, "Spin transfer torque devices utilizing the giant spin Hall effect of tungsten," *Applied Physics Letters*, vol. 101, no. 12, p. 122404, 2012.
- [38] C.-F. Pai, M.-H. Nguyen, C. Belvin, L. H. Vilela-Leão, D. Ralph and R. Buhrman, "Enhancement of perpendicular magnetic anisotropy and transmission of spin-Hall-effect-induced spin currents by a Hf spacer layer in W/Hf/CoFeB/MgO layer structures," *Applied Physics Letters*, vol. 104, no. 8, p. 082407, 2014.
- [39] K.-U. Demasius, T. Phung, W. Zhang, B. P. Hughes, S.-H. Yang, A. Kellock, W. Han, A. Pushp and S. S. Parkin, "Enhanced spin-orbit torques by oxygen incorporation in tungsten films," *Nature communications*, vol. 7, no. 1, pp. 1, 2016.
- [40] L. Wang, K. Shi, S. Peng, K. Cao, H. Yang, J. Gao, W. Zhao and C. Zhao, "Large spin Hall effect of perpendicularly magnetized β -W/CoFeB/MgO layers with high thermal stability," *Japanese Journal of Applied Physics*, vol. 58, no. 5, p. 050903, 2019.
- [41] J. Torrejon, J. Kim, J. Sinha, S. Mitani, M. Hayashi, M. Yamanouchi and H. Ohno, "Interface control of the magnetic chirality in CoFeB/MgO heterostructures with heavy-metal underlayers," *Nature communications*, vol. 5, no. 1, pp. 1, 2014.
- [42] Q. Hao and G. Xiao, "Giant spin Hall effect and switching induced by spin-transfer torque in a W/Co 40 Fe 40 B 20/MgO structure with perpendicular magnetic anisotropy," *Physical Review Applied*, vol. 3, no. 3, p. 034009, 2015.
- [43] S. Cho, S.-h. C. Baek, K.-D. Lee, Y. Jo and B.-G. Park, "Large spin Hall magnetoresistance and its correlation to the spin-orbit torque in W/CoFeB/MgO structures," *Scientific reports*, vol. 5, p. 14668, 2015.
- [44] S. Cho and B.-G. Park, "Large planar Hall effect in perpendicularly magnetized W/CoFeB/MgO structures," *Current Applied Physics*, vol. 15, no. 8, pp. 902, 2015.
- [45] L. Neumann, D. Meier, J. Schmalhorst, K. Rott, G. Reiss and M. Meinert, "Temperature dependence of the spin Hall angle and switching current in the nc-W (O)/CoFeB/MgO system with perpendicular magnetic anisotropy," *Applied Physics Letters*, vol. 109, no. 14, p. 142405, 2016.
- [46] H. Mazraati, S. Chung, A. Houshang, M. Dvornik, L. Piazza, F. Qejvanaj, S. Jiang, T. Q. Le, J. Weissenrieder and J. Åkerman, "Low operational current spin Hall nano-oscillators based on NiFe/W bilayers," *Applied Physics Letters*, vol. 109, no. 24, p. 242402, 2016.

- [47] T. Okuno, T. Taniguchi, S. Kim, S.-h. C. Baek, B.-G. Park, T. Moriyama, K.-J. Kim and T. Ono, "Temperature dependence of spin Hall magnetoresistance in W/CoFeB bilayer," *Japanese Journal of Applied Physics*, vol. 55, no. 8, p. 080308, 2016.
- [48] C. Zhang, S. Fukami, K. Watanabe, A. Ohkawara, S. DuttaGupta, H. Sato, F. Matsukura and H. Ohno, "Critical role of W deposition condition on spin-orbit torque induced magnetization switching in nanoscale W/CoFeB/MgO," *Applied Physics Letters*, vol. 109, no. 19, p. 192405, 2016.
- [49] W. Skowroński, M. Cecot, J. Kanak, S. Ziętek, T. Stobiecki, L. Yao, S. Van Dijken, T. Nozaki, K. Yakushiji and S. Yuasa, "Temperature dependence of spin-orbit torques in W/CoFeB bilayers," *Applied Physics Letters*, vol. 109, no. 6, p. 062407, 2016.
- [50] J. Kim, P. Sheng, S. Takahashi, S. Mitani and M. Hayashi, "Spin Hall magnetoresistance in metallic bilayers," *Physical review letters*, vol. 116, no. 9, p. 097201, 2016.
- [51] S. Mondal, S. Choudhury, N. Jha, A. Ganguly, J. Sinha and A. Barman, "All-optical detection of the spin Hall angle in W/CoFeB/SiO₂ heterostructures with varying thickness of the tungsten layer," *Physical review B*, vol. 96, no. 5, p. 054414, 2017.
- [52] J. Zhang, C. Garg, T. Phung, C. Rettner, B. P. Hughes, S.-H. Yang, Y. Jiang and S. S. Parkin, "Role of Micromagnetic States on Spin–Orbit Torque-Switching Schemes," *Nano letters*, vol. 18, no. 7, pp. 4074, 2018.
- [53] M. Zahedinejad, H. Mazraati, H. Fulara, J. Yue, S. Jiang, A. Awad and J. Åkerman, "CMOS compatible W/CoFeB/MgO spin Hall nano-oscillators with wide frequency tunability," *Applied Physics Letters*, vol. 112, no. 13, p. 132404, 2018.
- [54] W. Chen, G. Xiao, Q. Zhang and X. Zhang, "Temperature study of the giant spin Hall effect in the bulk limit of β -W," *Physical Review B*, vol. 98, no. 13, p. 134411, 2018.
- [55] W.-B. Liao, T.-Y. Chen, Y. Ferrante, S. S. Parkin and C.-F. Pai, "Current-Induced Magnetization Switching by the High Spin Hall Conductivity α -W," *physica status solidi (RRL)–Rapid Research Letters*, vol. 13, no. 11, p. 1900408, 2019.
- [56] Y. Fan, X. Kou, P. Upadhyaya, Q. Shao, L. Pan, M. Lang, X. Che, J. Tang, M. Montazeri and K. Murata, "Electric-field control of spin–orbit torque in a magnetically doped topological insulator," *Nature nanotechnology*, vol. 11, no. 4, pp. 352, 2016.
- [57] P. C. Filippou, J. Jeong, Y. Ferrante, S.-H. Yang, T. Topuria, M. G. Samant and S. S. Parkin, "Chiral domain wall motion in unit-cell thick perpendicularly magnetized Heusler films prepared by chemical templating," *Nature communications*, vol. 9, no. 1, pp. 1, 2018.
- [58] S. Shi, S. Liang, Z. Zhu, K. Cai, S. D. Pollard, Y. Wang, J. Wang, Q. Wang, P. He and J. Yu, "All-electric magnetization switching and Dzyaloshinskii–Moriya interaction in WTe₂/ferromagnet heterostructures," *Nature nanotechnology*, vol. 14, no. 10, pp. 945, 2019.
- [59] S.-H. Yang, C. Garg and S. S. Parkin, "Chiral exchange drag and chirality oscillations in synthetic antiferromagnets," *Nature Physics*, vol. 15, no. 6, pp. 543, 2019.
- [60] T. Okuno, D.-H. Kim, S.-H. Oh, S. K. Kim, Y. Hirata, T. Nishimura, W. S. Ham, Y. Futakawa, H. Yoshikawa and A. Tsukamoto, "Spin-transfer torques for domain wall motion in antiferromagnetically coupled ferrimagnets," *Nature Electronics*, vol. 2, no. 9, pp. 389, 2019.
- [61] S. Vélez, J. Schaab, M. S. Wörnle, M. Müller, E. Gradauskaite, P. Welter, C. Gutgsell, C. Nistor, C. L. Degen and M. Trassin, "High-speed domain wall racetracks in a magnetic insulator," *Nature communications*, vol. 10, no. 1, pp. 1, 2019.

- [62] Z. Luo, A. Hrabec, T. P. Dao, G. Sala, S. Finizio, J. Feng, S. Mayr, J. Raabe, P. Gambardella and L. J. Heyderman, "Current-driven magnetic domain-wall logic," *Nature*, vol. 579, no. 7798, pp. 214, 2020.
- [63] K. Garello, F. Yasin, S. Couet, L. Souriau, J. Swerts, S. Rao, S. Van Beek, W. Kim, E. Liu and S. Kundu, "SOT-MRAM 300nm integration for low power and ultrafast embedded memories," in *2018 IEEE Symposium on VLSI Circuits*, 2018, pp. 81.
- [64] S. Piramanayagam and T. C. Chong, *Developments in data storage: materials perspective*. John Wiley & Sons, 2011.
- [65] S. Piramanayagam, "Perpendicular recording media for hard disk drives," *Journal of Applied Physics*, vol. 102, no. 1, p. 2, 2007.
- [66] <http://gwyddion.net/>.
- [67] L. Martins, J. Ventura, R. Ferreira and P. Freitas, "Optimization of the buffer surface of CoFeB/MgO/CoFeB-based magnetic tunnel junctions by ion beam milling," *Applied Surface Science*, vol. 424, pp. 58, 2017.
- [68] W. Shen, D. Mazumdar, X. Zou, X. Liu, B. Schrag and G. Xiao, "Effect of film roughness in MgO-based magnetic tunnel junctions," *Applied physics letters*, vol. 88, no. 18, p. 182508, 2006.
- [69] J.-Y. Choi, H. Jun, K. Ashiba, J.-U. Baek, T.-H. Shim and J.-G. Park, "Double Pinned Perpendicular-Magnetic-Tunnel-Junction Spin-Valve Providing Multi-level Resistance States," *Scientific reports*, vol. 9, no. 1, pp. 1, 2019.
- [70] J. M. Lee, K. Cai, G. Yang, Y. Liu, R. Ramaswamy, P. He and H. Yang, "Field-Free Spin-Orbit Torque Switching from Geometrical Domain-Wall Pinning," *Nano letters*, vol. 18, no. 8, pp. 4669, 2018.
- [71] S. Emori, U. Bauer, S.-M. Ahn, E. Martinez and G. S. Beach, "Current-driven dynamics of chiral ferromagnetic domain walls," *Nature materials*, vol. 12, no. 7, pp. 611, 2013.
- [72] A. Hrabec, K. Shahbazi, T. A. Moore, E. Martinez and C. H. Marrows, "Tuning spin-orbit torques at magnetic domain walls in epitaxial Pt/Co/Pt_{1-x}Au_x trilayers," *Nanotechnology*, vol. 30, no. 23, p. 234003, 2019.
- [73] A. Thiaville, S. Rohart, É. Jué, V. Cros and A. Fert, "Dynamics of Dzyaloshinskii domain walls in ultrathin magnetic films," *EPL (Europhysics Letters)*, vol. 100, no. 5, p. 57002, 2012.
- [74] L. Buda-Prejbeanu, "Modelling of spintronic devices: from basic operation mechanisms toward optimization," Doctoral dissertation, 2017.
- [75] S. Jaiswal, K. Litzius, I. Lemesch, F. Büttner, S. Finizio, J. Raabe, M. Weigand, K. Lee, J. Langer and B. Ocker, "Investigation of the Dzyaloshinskii-Moriya interaction and room temperature skyrmions in W/CoFeB/MgO thin films and microwires," *Applied Physics Letters*, vol. 111, no. 2, p. 022409, 2017.
- [76] X. Zhao, X. Zhang, H. Yang, W. Cai, Y. Zhao, Z. Wang and W. Zhao, "Ultra-efficient spin-orbit torque induced magnetic switching in W/CoFeB/MgO structures," *Nanotechnology*, vol. 30, no. 33, p. 335707, 2019.
- [77] T. Koyama, D. Chiba, K. Ueda, K. Kondou, H. Tanigawa, S. Fukami, T. Suzuki, N. Ohshima, N. Ishiwata and Y. Nakatani, "Observation of the intrinsic pinning of a magnetic domain wall in a ferromagnetic nanowire," *Nature materials*, vol. 10, no. 3, pp. 194, 2011.
- [78] P. Metaxas, J. Jamet, A. Mougin, M. Cormier, J. Ferré, V. Baltz, B. Rodmacq, B. Dieny and R. Stamps, "Creep and flow regimes of magnetic domain-wall motion in ultrathin Pt/Co/Pt films with perpendicular anisotropy," *Physical review letters*, vol. 99, no. 21, p. 217208, 2007.

- [79] S. A. Siddiqui, S. Dutta, A. Tang, L. Liu, C. A. Ross and M. A. Baldo, "Magnetic domain wall based synaptic and activation function generator for neuromorphic accelerators," *Nano Letters*, vol. 20, no. 2, pp. 1033, 2019.
- [80] K. Shahbazi, J.-V. Kim, H. T. Nembach, J. M. Shaw, A. Bischof, M. D. Rossell, V. Jeudy, T. A. Moore and C. H. Marrows, "Domain-wall motion and interfacial Dzyaloshinskii-Moriya interactions in Pt/Co/Ir (t_{Ir})/Ta multilayers," *Physical Review B*, vol. 99, no. 9, p. 094409, 2019.
- [81] T. Kim, Y. Choi, K. Lee, J. Yoon, J. Cho, C.-Y. You and M. Jung, "Perpendicular magnetic anisotropy of amorphous [CoSiB/Pt] N thin films," *Journal of Applied Physics*, vol. 117, no. 17, p. 17B502, 2015.
- [82] P. Bruno, G. Bayreuther, P. Beauvillain, C. Chappert, G. Lugert, D. Renard, J. Renard and J. Seiden, "Hysteresis properties of ultrathin ferromagnetic films," *Journal of applied physics*, vol. 68, no. 11, pp. 5759, 1990.
- [83] S. Collocott, "Sweep rate dependence of the coercivity and fluctuation field in some RE-Fe (Co)-Al, RE= Nd or Pr, bulk amorphous ferromagnets," *Journal of Applied Physics*, vol. 107, no. 9, p. 09A720, 2010.
- [84] <https://www.ufba.org.nz/images/documents/calculationofelectricalenergy.pdf>.
- [85] K. Fritz, L. Neumann and M. Meinert, "Ultra-low switching current density in all-amorphous W-Hf/CoFeB/TaOx films," *arXiv preprint arXiv:1910.13837*, 2019.

Chapter 7. Ultra-low Power Domain Wall Devices for Neuromorphic Computing

The observation of low current density while driving the DWs in dual W layer devices inspired us to study the low power neuromorphic computing operations. The neuromorphic operations include the realization of functioning synthetic neurons and synapses. In this study, we have proposed the idea of meander wire, where neighboring segments of wire meet at a certain offset. These meeting points help in controlling the DW pinning and depinning from these sites and hence, help to achieve the synaptic functions [1, 2]. Note, they are called artificial pinning sites in conventional spintronics. Therefore, in addition to synaptic performances, they can also be used for controlled and reliable DW motion for DW memory applications (as discussed in Chapter 5). We will present our results of synaptic performances in meander wire design in the coming sections.

7.1. Introduction

Artificial intelligence (AI) is finding widespread applications in several consumer devices such as smartphones, the internet of things, self-driven cars, and many more [3]. However, these intelligent tasks have been accomplished by utilizing intelligent algorithms in present “von Neumann architecture” which is power inefficient [4]. According to a report, which was published five years ago, information, communication, and related technologies consume ~5% of the global energy [5, 6]. Therefore, there is a need to develop an energy-efficient architecture and neuromorphic computing is one such architecture. In recent years, an ample amount of attention has been paid to Neuromorphic or brain-inspired computing [7-13]. The functionalities of the human brain, which needs $\sim 10^4\times$ lesser power than the fastest supercomputer for computing approximately the same amount of data, set an inspiration for neuromorphic computing [14]. The human brain is a complex network of neurons connected

through synapses [15]. The neurons and synapses are the processing and memory units of the human brain. Several material systems, such as phase-change memory, resistive RAM, and spin-based materials have been studied to mimic the functionalities of the human brain in recent years [1, 16-20]. Amongst several charge and spin-based devices for neuromorphic computing, DW devices are one of the least power consuming devices [21]. Therefore, in addition to DW memory, the DW research community is exploring the virtues of DW devices for Neuromorphic computing technology [1, 2, 6].

Figure 7.1 (a) and (b) show the schematics of DW based neurons and synapses [22]. In the DW based neuron, when an input current flows through the spin-Hall layer (dual W layer in the present case), it displaces the DW. As soon as the DW reaches the right end (below or beyond the MTJ structure), the resistance (rather TMR) decreases and the information (encoded in the electronic spins) travels towards the output terminal. This process is called the spiking of the neuron. The synapse, which modulates the input of the post neuron, can be achieved by stopping the DW wall at several places. In the case of the synapse, the MTJ structure is made on the whole of the DW device. By stopping the DW at multiple places, the resistance can be altered, and hence the weight of the information can be altered.

Figure 7.1. The schematic representation of DW based neuromorphic (a) neuron and (b) synapse.

In 2013, Sharad et al. developed a neural network made up of DW devices, which was used for image processing purposes [23]. Later, Sengupta et al. simulated DW devices based all-spin neural network, which turned out 250× energy-efficient than CMOS architecture [24]. Several other domain wall device based neural networks have been studied using numerical simulations and presented [22, 24-37]. However, the experimental realization of the proposed device architectures is lacking. Only a few studies at the fundamental level have been performed. Jin et al. fabricated artificial pinning sites using the ion-implantation method for achieving multi-

resistance states to use it as neuromorphic synapse [2]. In another study multi-resistance states were obtained by increasing the width of the nanowire at certain spacing [38]. Borders et al. fabricated a Hall device and showed multiple Hall resistance states [39]. Recently, Sato et al. fabricated a z-shaped nanowire to achieve the DW oscillations at GHz frequencies for neuromorphic application [40].

Figure 7.2. The schematics of (a) straight or reference and (b) meander microwires. Here, red and blue represent the magnetizations in $+z$ and $-z$ directions, respectively. (c) The top view with the exact number of pinning sites (as in our study) of Figure 7.2 (b). The offset increases in going from the present pinning site to the next pinning site.

Considering an extremely low energy consumption for moving the DW in our engineered material (results presented in previous sub-sections), we proposed the concept of meander nanowire for fabricating ultra-low-power Neuromorphic synapse. In a meander nanowire, the neighboring segments of a wire are connected with a small offset (d). The schematic of the meander nanowire is shown in Figure 7.2. We fixed the width of the wire at $10\ \mu\text{m}$ and length at $240\ \mu\text{m}$. Moreover, we fabricated devices with several d values for optimizing the pinning strength of the constricted pinning centers. However, the length of each segment (to keep the bit lengths identical) is constant in all the devices. The offset increases by 5% (of the width of

wire) in going from the present pinning center to the next pinning center in every device (as shown in Figure 7.2 (c)). If we say that d is 15%, this means that the first pinning center has an offset of 15%, the second pinning center has an offset of 20%, and so on. We have studied three different devices with d values 15%, 40%, and 50%, respectively. We performed the SOT induced DW motion experiments for probing the pinning and depinning process in these devices. We demonstrated the stochastic, intermediate (between stochastic and non-stochastic), and non-stochastic functionalities of neuromorphic synapses through devices with the above-mentioned offsets (d). Besides, we have demonstrated a few specific synaptic functions in our meander devices. We will discuss the results in detail in upcoming sub-sections.

7.2. Synaptic Performance of Meander devices

7.2.1. Stochastic Synaptic Behaviour

First, we experimented with the device with $d = 15\%$ and the results are presented in Figure 7.3 (a) and (b). We fabricated four pinning sites in all the devices. In the first step, the device was saturated in the $+z$ axis with an OOP magnetic field of +500 Oe. Subsequently, a reversed domain was nucleated at one end of the wire using a small reversal field of -15 Oe. Taking this as an initial state, we applied an electric current to observe the DW motion. A negative OOP magnetic field of magnitude 12 Oe was applied during the measurements for this device. The down-up DW starts moving from the upper end of the wire in a direction opposite to the electric current at a J of 1×10^6 A/m² (refer to Figure 7.3 (a)). The nature of DW motion is consistent with our previous results of SOT driven DW motion in the straight nanowire. The DW continues to move from one end to the other end without effective pinning at the pinning centers (P1, P2, P3, and P4). This feature of DW motion could be useful for stochastic neuromorphic synapses. However, the DW stops at some places (flat regions in x vs t graph) along the meander microwire and a higher J of 1×10^7 A/m² to 1×10^8 A/m² was applied to depin and

further moving the DW. This could be because of the presence of extrinsic pinning regions induced by the device fabrication process [41]. Table 7.1 summarizes the domain wall position as a function of time and corresponding current density values (table corresponds to x and t values plotted in Figure 7.3 (b)).

Table 7.1. The table of time “ t ”, domain wall position “ x ”, and corresponding current density “ J ” values for the device with $d= 15\%$.

t (ms)	x (mm)	J (A/m ²)
0 to 1.35	0 to 0.046	1×10^6
1.40 to 2.30	0.046 to 0.108	1×10^7
2.35 to 3.25	0.108 to 0.118	5×10^7
3.30 to 4.35 (end)	0.118 to 0.154	1×10^8

Figure 7.3. (a) The domain maps of DW motion at the positions of pinning centers, captured using Kerr microscopy and (b) the plot of domain wall position “ x ” vs time “ t ” for $d= 15\%$.

7.2.2. The Transition from Stochastic to Non-stochastic Synaptic Behaviour

In the next step, we performed measurements for the device with $d= 40\%$ and results are presented in Figure 7.4 (a) and (b). After saturating the device with an OOP magnetic field of +500 Oe, a nucleation field of -16 Oe was applied to nucleate the DW. A constant magnetic

field of -13 Oe was used during the measurements for this device. Like the case of $d=15\%$, the DW starts moving from the upper end and does not stop at pinning sites P1 and P2. The current density required to carry the DW until the third pinning site is 1×10^6 and 1×10^7 A/m². However, the DW gets pinned at the third pinning site and a depinning current of 5×10^7 A/m² was applied to depin the DW. Further, the DW can be moved in the straight segment between pinning sites 3 and 4 at a small J of 1×10^7 A/m². Subsequently, the DW gets pinned at the fourth pinning site and a higher depinning current density of 1×10^9 A/m² is needed to depin the DW. In summary, the DW moves at very small J values in the straight segment (the region between pinning sites) while a higher current is needed for depinning the DW from pinning sites. Moreover, the depinning current increases as the pinning strength of the pinning sites increases. The details of the current density used at various times during the measurement are listed in table 7.2 (table corresponds to x and t values plotted in Figure 7. 4 (b)).

Table 7.2. The table of time “ t ”, domain wall position “ x ”, and corresponding current density “ J ” values for the device with $d=40\%$.

t (ms)	x (mm)	J (A/m ²)
0 to 0.75	0 to 0.048	1×10^6
0.80 to 1.75	0.048 to 0.107	1×10^7
1.75 to 2.10	0.107 to 0.108 (<i>pinned at 3rd pinning site until 2.10 ms and depins at $t=2.10$ ms, $J=5 \times 10^7$ A/m²</i>)	5×10^7
2.15 to 2.95	0.108 to 0.150	1×10^7
3.10 to 4.35	0.150 (<i>DW pinned at 4th pinning site</i>)	1×10^7 , 5×10^7 , 1×10^8 , 5×10^8 , 1×10^9
4.40 to 4.45 (end)	0.150 to 0.164 (<i>DW depins and moves towards the end</i>)	1×10^9

Figure 7.4. (a) The domain maps of DW motion at the positions of pinning centers, captured using Kerr microscopy and (b) the plot of domain wall position “ x ” vs time “ t ” for $d = 40\%$.

7.2.3. Non-stochastic Synaptic Behaviour

In the next set of meander devices, we increased the offset value “ d ” to 50% and performed similar experiments on those devices. The corresponding experimental results are shown in Figure 7.5. Similar to previous cases, we first saturated the device with an OOP magnetic field of +500 Oe and then nucleated the reversed domain by applying a magnetic field of -15 Oe. After obtaining the initial state, we set a constant magnetic field of -12 Oe and started applying the current pulses. In this case, the DW gets pinned at all the pinning sites. The current density needed to move the DW in the straight portions was either 1×10^6 or 1×10^7 A/m². However, a higher J was needed to depin the DW from the pinning sites. Additionally, the depinning current density increases as the pinning increases. Therefore, an optimum value of pinning strength ($\sim 50\%$) should be set for the non-stochastic synaptic devices. The following table describes the J values used at various phases of DW motion (table corresponds to x and t values plotted in Figure 7.5 (b)).

Table 7.3. The table of time “ t ”, domain wall position “ x ”, and corresponding current density “ J ” values for the device with $d= 50\%$.

t (ms)	x (mm)	J (A/m ²)
0 to 0.35	0 to 0.007	1×10^6
0.40 to 1.00	0.008 to 0.033	1×10^7
1.05 to 3.15	0.033 to 0.034 (<i>pinned at 1st pinning site until 3.15 ms and depins at $t= 3.15$ ms, $J= 1 \times 10^{10}$ A/m²</i>)	$1 \times 10^7, 5 \times 10^7, 1 \times 10^8, 5 \times 10^8, 1 \times 10^9, 5 \times 10^9, 1 \times 10^{10}$
3.20 to 3.75	0.034 to 0.041	1×10^{10}
3.80 to 4.60	0.041 to 0.074	1×10^7
4.65 to 7.35	0.074 to 0.082 (<i>pinned at 2nd pinning site until 7.35 ms and depins at $t= 7.35$ ms, $J= 1 \times 10^{11}$ A/m²</i>)	$1 \times 10^7, 5 \times 10^7, 1 \times 10^8, 5 \times 10^8, 1 \times 10^9, 5 \times 10^9, 1 \times 10^{10}, 5 \times 10^{10}, 1 \times 10^{11}$
7.40 to 8.30	0.082 to 0.119	1×10^7
8.35 to 10.20	0.119 to 0.121 (<i>pinned at 3rd pinning site until 10.20 ms and depins at $t= 10.20$ ms, $J= 1 \times 10^{10}$ A/m²</i>)	$1 \times 10^7, 5 \times 10^7, 1 \times 10^8, 5 \times 10^8, 1 \times 10^9, 5 \times 10^9, 1 \times 10^{10}$
10.25 to 10.70	0.121 to 0.124	1×10^{10}
10.75 to 14.50	0.124 to 0.162	$1 \times 10^7, 5 \times 10^7, 1 \times 10^8, 5 \times 10^8, 1 \times 10^9, 5 \times 10^9, 1 \times 10^{10}, 5 \times 10^{10}$
14.55 to 14.90	0.162 (<i>pinned at 4th pinning site</i>)	5×10^{10}
14.95 to 15.20	0.162 to 0.173 (<i>depinning at $t= 14.95$ ms</i>)	5×10^{10}
15.20 to 15.60 (end)	0.173 to 0.191	1×10^7

Figure 7.5. (a) The domain maps of DW motion at the positions of pinning centers, captured using Kerr microscopy and (b) the plot of domain wall position “ x ” vs time “ t ” for $d = 50\%$.

In addition, we have also performed experiments by reversing the direction of the current (results are not shown here). We observed identical results of pinning and depinning. However, the J required to move the DW in the straight segments as well as the depinning current was found to be relatively higher. This is because when the current direction is reversed, the applied magnetic field opposes the effective field generated from SOT and a higher current is required. However, for the present case (for results presented in the thesis), the applied magnetic field supports the SOT effective field. These results are in very good agreement with our results for the straight nanowire.

As we have discussed in the previous chapter, the pinning sites are extremely important for the controlled motion of DWs in DW racetrack memory. The results obtained for meander nanowire are also significant for controlled DW motion for applications in DW racetrack memory.

7.2.4. Other Synaptic Performances

After observing the stochastic, intermediate, and non-stochastic functions of meander devices, we focused on a few more specific synaptic functions. We have performed experiments to achieve (i) multi-resistance states, (ii) synaptic potentiation, (iii) long term plasticity, and (iv) paired-pulse facilitation (PPF). Multi-resistance states are important in transmitting the weighted signals from pre-synaptic to post-synaptic neuron. Figure 7.6 shows the existence of multi-resistance states through DW displacement vs time graph in our meander DW devices. Although these results are identical to the one presented in Figure 7.5, it is important to highlight that the non-stochastic synapse and multi-resistance functions can be realized from the same set of results.

Figure 7.6. The illustration of multilevel resistance states for the meander wire device with offset $d= 50\%$.

Figure 7.7 shows the synaptic potentiation function for the meander DW device. A continuous application of the current pulses results in continuous DW motion. However, later DW gets pinned at the first pinning site. We performed measurements until 300 pulses of electrical current. The DW position, which is proportional to the conductivity of the soft layer (in full

magnetic tunnel junction) of DW synaptic devices alters as a function of time when an external current is applied.

Figure 7.7. The illustration of (a) potentiation of the meander DW device.

In the next step, we performed measurements for realizing the long term plasticity of the meander devices. The results of long term plasticity are shown in Figure 7.8. We applied current pulses and pushed the DW to the pinning site. Then we switched off the current and observed the DW position in subsequent times. As we propose to observe a small OOP magnetic field required for current-driven DW motion from the reference layer of MTJ. Therefore, we continue to apply a small OOP magnetic field even when the current is switched off. The DW stabilizes at this pinning site for a duration of more than 17100 s. This signifies that our devices are suitable for mimicking the long term synaptic plasticity. This function is also significant for DW memory devices, where it is termed as thermal stability [42, 43]. As we mentioned previously that the thermal stability factor indicates the lifetime of a memory device [42].

Figure 7.8. The illustration of long term plasticity of meander DW device.

Figure 7.9. The depiction of paired-pulse facilitation function for meander synaptic device.

Lastly, we applied pairs of pulses with amplitude 1×10^{11} A/m². The time difference between the consecutive pulses was varied. Moreover, a constant magnetic field of 9 Oe was applied to mimic the stray field arising from the reference layer of the full synaptic device (MTJ with present magnetic layer as its soft layer). We varied the time difference from 5 μ s to 300 s. After every pulse, we captured the Kerr microscopy image. If x_1 and x_2 are the displacements after applying first and second pulse respectively, the PPF (in %) is given by [44],

$$PPF (\%) = \frac{x_2}{x_1} \times 100 \quad (7.1)$$

The results of this set of experiments are shown in Figure 7.9. The PPF remains almost the same as a function of time difference. This could be because the thermal energy from the first pulse does not affect the DW dynamics during the application of the second pulse. Thus, to observe these effects clearly, we should reduce the time difference in the range of nanoseconds, which is restricted due to the unavailability of experimental techniques in our lab.

7.3. Conclusions

The widespread usage of AI demands power-efficient architecture, where intelligent algorithms can be used to achieve intelligent tasks at low power. Neuromorphic computing is attracting attention as a replacement architecture. DW devices have been proposed to construct synthetic neurons and synapses for realizing neuromorphic computing. However, further power reduction is required in DW devices. As described in the previous chapter that our dual-layer device can reduce the current density to drive the DW by an order to 10^4 . To continue, we have proposed a meander wire design where dual W was used as a spin Hall layer and the results of this design were presented in this chapter. We varied the offset to 15%, 40%, and 50% in the meander wires and studied the stochastic, intermediate, and non-stochastic synaptic functions. Moreover, we have presented more specific synaptic functions such as multi-resistance states, synaptic potentiation, long-term plasticity, and PPF. In conclusion, the observation of such energy-efficient synaptic operations is a significant contribution to the field of neuromorphic computing.

References

- [1] J. Grollier, D. Querlioz, K. Camsari, K. Everschor-Sitte, S. Fukami and M. Stiles, "Neuromorphic spintronics," *Nature Electronics*, vol. 3, no. 7, pp. 360, 2020.
- [2] T. Jin, W. L. Gan, F. N. Tan, N. R. Sernicola, W. S. Lew and S. N. Piramanayagam, "Synaptic element for neuromorphic computing using a magnetic domain wall device with synthetic pinning sites," *Journal of Physics D-Applied Physics*, vol. 52, no. 44, p. 5, 2019.
- [3] <https://ebsedu.org/blog/career-opportunities-artificial-intelligence/>.
- [4] E. García-Martín, C. F. Rodrigues, G. Riley and H. Grahn, "Estimation of energy consumption in machine learning," *Journal of Parallel and Distributed Computing*, vol. 134, pp. 75, 2019.
- [5] E. Gelenbe and Y. Caseau, "The impact of information technology on energy consumption and carbon emissions," *Ubiquity (an ACM publication)*, pp. 1, 2015.
- [6] J. Grollier, D. Querlioz and M. D. Stiles, "Spintronic nanodevices for bioinspired computing," *Proceedings of the IEEE*, vol. 104, no. 10, pp. 2024, 2016.
- [7] J. Torrejon, M. Riou, F. A. Araujo, S. Tsunegi, G. Khalsa, D. Querlioz, P. Bortolotti, V. Cros, K. Yakushiji and A. Fukushima, "Neuromorphic computing with nanoscale spintronic oscillators," *Nature*, vol. 547, no. 7664, pp. 428, 2017.
- [8] S. K. Esser, R. Appuswamy, P. Merolla, J. V. Arthur and D. S. Modha, "Backpropagation for energy-efficient neuromorphic computing," in *Advances in neural information processing systems*, 2015, pp. 1117.
- [9] Y. van De Burgt, A. Melianas, S. T. Keene, G. Malliaras and A. Salleo, "Organic electronics for neuromorphic computing," *Nature Electronics*, vol. 1, no. 7, pp. 386, 2018.
- [10] G. Indiveri, B. Linares-Barranco, R. Legenstein, G. Deligeorgis and T. Prodromakis, "Integration of nanoscale memristor synapses in neuromorphic computing architectures," *Nanotechnology*, vol. 24, no. 38, p. 384010, 2013.
- [11] N. K. Upadhyay, H. Jiang, Z. Wang, S. Asapu, Q. Xia and J. Joshua Yang, "Emerging memory devices for neuromorphic computing," *Advanced Materials Technologies*, vol. 4, no. 4, p. 1800589, 2019.
- [12] K. M. Song, J.-S. Jeong, B. Pan, X. Zhang, J. Xia, S. Cha, T.-E. Park, K. Kim, S. Finizio and J. Raabe, "Skyrmion-based artificial synapses for neuromorphic computing," *Nature Electronics*, vol. 3, no. 3, pp. 148, 2020.
- [13] A. Kurenkov, S. Fukami and H. Ohno, "Neuromorphic computing with antiferromagnetic spintronics," *Journal of Applied Physics*, vol. 128, no. 1, p. 010902, 2020.
- [14] <https://www.youtube.com/watch?v=qpxupqADYP0&t=290s>.
- [15] <https://singularity2030.ch/the-human-brain-and-neuromorphic-computing/>.
- [16] T. H. Lee, D. Loke, K. J. Huang, W. J. Wang and S. R. Elliott, "Tailoring transient-amorphous states: towards fast and power-efficient phase-change memory and neuromorphic computing," *Advanced Materials*, vol. 26, no. 44, pp. 7493, 2014.
- [17] S. Nandakumar, M. Le Gallo, I. Boybat, B. Rajendran, A. Sebastian and E. Eleftheriou, "A phase-change memory model for neuromorphic computing," *Journal of Applied Physics*, vol. 124, no. 15, p. 152135, 2018.
- [18] F. Zhou, Z. Zhou, J. Chen, T. H. Choy, J. Wang, N. Zhang, Z. Lin, S. Yu, J. Kang and H.-S. P. Wong, "Optoelectronic resistive random access memory for neuromorphic vision sensors," *Nature nanotechnology*, vol. 14, no. 8, pp. 776, 2019.

- [19] B. Gao, J. Kang, Z. Zhou, Z. Chen, P. Huang, L. Liu and X. Liu, "Metal oxide resistive random access memory based synaptic devices for brain-inspired computing," *Japanese Journal of Applied Physics*, vol. 55, no. 4S, p. 04EA06, 2016.
- [20] C. Münch, R. Bishnoi and M. B. Tahoori, "Reliable in-memory neuromorphic computing using spintronics," in *Proceedings of the 24th Asia and South Pacific Design Automation Conference*, 2019, pp. 230.
- [21] P. Y. Chenyun and A. Naeemi, "An Expanded Benchmarking of Beyond-CMOS Devices Based on Boolean and Neuromorphic Representative Circuits," *IEEE Journal on Exploratory Solid-State Computational Devices and Circuits*, vol. 3, pp. 101, 2017.
- [22] A. Sengupta, Y. Shim and K. Roy, "Proposal for an All-Spin Artificial Neural Network: Emulating Neural and Synaptic Functionalities Through Domain Wall Motion in Ferromagnets," *IEEE Transactions on Biomedical Circuits and Systems*, vol. 10, no. 6, pp. 1152, 2016.
- [23] M. Sharad, D. Fan, K. Yogendra and K. Roy, *Ultra-Low Power Neuromorphic Computing With Spin-Torque Devices* (2013 Third Berkeley Symposium on Energy Efficient Electronic Systems). New York: IEEE, 2013.
- [24] A. Sengupta and K. Roy, "A Vision for All-Spin Neural Networks: A Device to System Perspective," *IEEE Transactions on Circuits and Systems I-Regular Papers*, vol. 63, no. 12, pp. 2267, 2016.
- [25] M. Sharad, C. Augustine and K. Roy, *Boolean and Non-Boolean Computation With Spin Devices* (2012 IEEE International Electron Devices Meeting). New York: IEEE, 2012.
- [26] S. Deb, T. Vatswani, A. Chattopadhyay, A. Basu and X. Y. Fong, "Domain Wall Motion-Based Dual-Threshold Activation Unit for Low-Power Classification of Non-Linearly Separable Functions," *IEEE Transactions on Biomedical Circuits and Systems*, vol. 12, no. 6, pp. 1410, 2018.
- [27] N. Hassan, X. Hu, L. Jiang-Wei, W. H. Brigner, O. G. Akinola, F. Garcia-Sanchez, M. Pasquale, C. H. Bennett, J. A. C. Incorvia and J. S. Friedman, "Magnetic domain wall neuron with lateral inhibition," *Journal of Applied Physics*, vol. 124, no. 15, p. 10, 2018.
- [28] U. Saxena, D. Kaushik, M. Bansal, U. Sahu and D. Bhowmik, "Low-Energy Implementation of Feed-Forward Neural Network With Back-Propagation Algorithm Using a Spin-Orbit Torque Driven Skyrmionic Device," *IEEE Transactions on Magnetics*, vol. 54, no. 11, p. 5, 2018.
- [29] A. Sengupta and K. Roy, "Neuromorphic computing enabled by physics of electron spins: Prospects and perspectives," *Applied Physics Express*, vol. 11, no. 3, p. 7, 2018.
- [30] D. M. Zhang, Y. C. Hou, C. Z. Wang, J. Chen, L. Zeng, W. S. Zhao and Acm, *Hardware Acceleration Implementation of Sparse Coding Algorithm with Spintronic Devices* (Nanoarch'18: Proceedings of the 14th Ieee/Acm International Symposium on Nanoscale Architectures). New York: Assoc Computing Machinery, 2018, pp. 79.
- [31] A. Agrawal and K. Roy, "Mimicking Leaky-Integrate-Fire Spiking Neuron Using Automation of Domain Walls for Energy-Efficient Brain-Inspired Computing," *Ieee Transactions on Magnetics*, vol. 55, no. 1, p. 7, 2019.
- [32] D. Bhowmik, U. Saxena, A. Dankar, A. Verma, D. Kaushik, S. Chatterjee and U. Singh, "On-chip learning for domain wall synapse based Fully Connected Neural Network," *Journal of Magnetism and Magnetic Materials*, vol. 489, p. 11, 2019.
- [33] W. H. Brigner, J. S. Friedman, N. Hassan, L. Jiang-Wei, X. Hu, D. Saha, C. H. Bennett, M. J. Marinella, J. A. C. Incorvia and F. Garcia-Sanchez, "Shape-Based Magnetic Domain Wall Drift for an Artificial Spintronic Leaky Integrate-and-Fire Neuron," *IEEE Transactions on Electron Devices*, vol. 66, no. 11, pp. 4970, 2019.

- [34] W. H. Brigner, X. Hu, N. Hassan, C. H. Bennett, J. A. C. Incorvia, F. Garcia-Sanchez and J. S. Friedman, "Graded-Anisotropy-Induced Magnetic Domain Wall Drift for an Artificial Spintronic Leaky Integrate-and-Fire Neuron," *IEEE Journal on Exploratory Solid-State Computational Devices and Circuits*, vol. 5, no. 1, pp. 19, 2019.
- [35] K. Roy, D. L. Fan, X. Y. Fong, Y. S. Kim, M. Sharad, S. Paul, S. Chatterjee, S. Bhunia and S. Mukhopadhyay, "Exploring Spin Transfer Torque Devices for Unconventional Computing," *IEEE Journal on Emerging and Selected Topics in Circuits and Systems*, vol. 5, no. 1, pp. 5, 2015.
- [36] X. Y. Fong, Y. S. Kim, K. Yogendra, D. L. Fan, A. Sengupta, A. Raghunathan and K. Roy, "Spin-Transfer Torque Devices for Logic and Memory: Prospects and Perspectives," *IEEE Transactions on Computer-Aided Design of Integrated Circuits and Systems*, vol. 35, no. 1, pp. 1, 2016.
- [37] K. Yue, Y. Z. Liu, R. K. Lake and A. C. Parker, "A brain-plausible neuromorphic on-the-fly learning system implemented with magnetic domain wall analog memristors," *Science Advances*, vol. 5, no. 4, p. 9, Apr 2019.
- [38] J. Cai, B. Fang, C. Wang and Z. Zeng, "Multilevel storage device based on domain-wall motion in a magnetic tunnel junction," *Applied Physics Letters*, vol. 111, no. 18, p. 182410, 2017.
- [39] W. A. Borders, H. Akima, S. Fukami, S. Moriya, S. Kurihara, Y. Horio, S. Sato and H. Ohno, "Analogue spin-orbit torque device for artificial-neural-network-based associative memory operation," *Applied Physics Express*, vol. 10, no. 1, p. 013007, 2016.
- [40] N. Sato, K. Schultheiss, L. Körber, N. Puwenberg, T. Mühl, A. Awad, S. Arekapudi, O. Hellwig, J. Fassbender and H. Schultheiss, "Domain wall based spin-Hall nano-oscillators," *Physical review letters*, vol. 123, no. 5, p. 057204, 2019.
- [41] O. Boulle, G. Malinowski and M. Kläui, "Current-induced domain wall motion in nanoscale ferromagnetic elements," *Materials Science and Engineering: R: Reports*, vol. 72, no. 9, pp. 159, 2011.
- [42] S. Fukami, T. Suzuki, K. Nagahara, N. Ohshima and N. Ishiwata, "Large thermal stability independent of critical current of domain wall motion in Co/Ni nanowires with step pinning sites," *Journal of Applied Physics*, vol. 108, no. 11, p. 113914, 2010.
- [43] D. Kumar, T. Jin, S. Al Risi, R. Sbiaa, W. S. Lew and S. Piramanayagam, "Domain wall motion control for racetrack memory applications," *IEEE Transactions on Magnetics*, vol. 55, no. 3, pp. 1, 2018.
- [44] Y.-F. Wang, Y.-C. Lin, I.-T. Wang, T.-P. Lin and T.-H. Hou, "Characterization and modeling of nonfilamentary Ta/TaO_x/TiO₂/Ti analog synaptic device," *Scientific reports*, vol. 5, p. 10150, 2015.

Chapter 8. Conclusions and Future Work

This chapter describes the conclusions of the thesis in its first part and a short discussion of the suggested future works in the second part. The suggested future works are hugely inspired by the research performed during the course of the thesis.

8.1. Conclusions of the Thesis

A huge amount of data is generated every minute in today's world. We -as a researcher, a banker, a doctor, an engineer, or even as an ordinary social media user- are contributing significantly to this increasing data [1]. Moreover, the trend of data increase is expected to enhance further in the future. Currently, the hard disk drives (HDDs) offer to store a larger portion (~75-80% of total data) of the total generated data [2]. Although alternative technologies are replacing HDDs in certain applications, more than half (~50-60% of total data) of the big-data is predicted to be stored in HDDs [2]. Therefore, improvements in the areal density of HDDs are very important. The current HDD industry is using the perpendicular magnetic recording (PMR) technology, whose areal density is limited to 1 Tbpsi [3]. Further increment in the areal density of the PMR technology is hindered by the so-called media trilemma problem [4]. Recently, Seagate has launched heat-assisted magnetic recording (HAMR) in the market [5]. This technology uses a laser supplement during the writing process. The achievable areal density with this product is over 4 Tbpsi [5]. On the other hand, other big media manufacturers such as Western Digital and Toshiba are planning to launch microwave-assisted magnetic recording (MAMR) based HDDs. They are predicted to reach up to an areal density closer to 4 Tbpsi [6, 7]. However, considering the demand for areal density growth due to rapidly increasing big data, successor HDD technology called heated-dot magnetic recording (HDMR) needs to be researched. According to the predicted roadmap (ASTC Technology Roadmap) of the HDDs, HDMR may take over the existing technologies by 2025 [8].

Currently, HDMR technology is under research in the industry as well as academia [9-13]. One of the major challenges of HDMR technology is coming up with a suitable patterning technique. Magnetic patterning is advantageous over physical patterning in many ways [14-16]. The magnetic patterning includes the ion-implantation of non-magnetic species through the hard mask to facilitate the local changes in structural and magnetic properties. Several methods such as conventional electron beam lithography, nano-imprint lithography have been studied for the fabrication of the hard mask. We have, however, researched the self-assembly of di-block copolymers (DBCP) for the fabrication of the hard mask, which resulted in smaller patterns with less size distribution [17].

We first studied $\text{Co}_{80}\text{Pt}_{20}$ alloy, which exhibits the hexagonal close-packed crystal structure and possesses a high perpendicular magnetic anisotropy [18-20]. On top of $\text{Co}_{80}\text{Pt}_{20}$ alloy, we performed self-assembly of polystyrene (PS)- polydimethylsiloxane (PDMS) DBCP to fabricate the mask patterns on the film stack. The self-assembly process resulted in a regular and uniform dot pattern with an average dot size of (27 ± 4) nm. Subsequently, we performed the ion implantation of high (40 keV) and low (7.5 keV) energetic $^{14}\text{N}^+$ and high (100 keV) and low (4.5 keV) energetic $^{40}\text{Ar}^+$ ions. Moreover, various ion fluences of 1×10^{15} , 5×10^{15} , 1×10^{16} , and 2×10^{16} (5×10^{16} ions/cm² for low energy ion implantations) ions/cm² were used in all the implantation cases.

After the fabrication of the samples, we characterized them for probing the structural and magnetic changes. From the XRD measurements, a significant change in structural properties was observed for the highest fluence in all the samples except for the low energy $^{40}\text{Ar}^+$ ion-implanted stack. However, the structural changes were observed to be lower for lower ion fluences in all the cases. We explained them based on the occupancy of implanted ions in interstitial positions or intermixing of Co, Pt, and Ru atoms at the CoPt/Ru interface or both. These results were well supported by the TRIM simulations and the cross-sectional TEM

measurements. The structural changes were also reflected when the magnetic measurements were performed. A significant change in coercivity was observed for the high energy $^{14}\text{N}^+$ and low energy $^{14}\text{N}^+$ ion implantations. The observed results were explained based on the decrease in anisotropy constant K_u or increase in exchange coupling A . These attributions were reflected from the MFM magnetic domain maps. We further confirmed the results from micromagnetic simulations and anisotropy constant calculations.

The equiatomic FePt alloy is the best choice for achieving high perpendicular magnetic anisotropy [21]. Therefore, we implanted NiTa/MgO/FePt-C stacks with high and low energetic $^{14}\text{N}^+$ and $^{40}\text{Ar}^+$ ions through a hard mask, prepared using the self-assembly process. Smaller mask patterns of average dot size (20.5 ± 2.7) nm were fabricated in this case. We chose the highest fluence for all the implantation cases. The XRD results of the pristine sample unveiled the co-existence of face-centered tetragonal (fct) hard and face-centered cubic (fcc) soft-phase in this sample. Although the sample was not as perfect as prepared by industrial researchers, it fulfilled the purpose of the study. For $^{14}\text{N}^+$ ion-implanted samples, the FePt (111) peak shifted towards lower 2θ values, which indicated an increase in interplanar spacing corresponding to this peak. However, for $^{40}\text{Ar}^+$ ion-implanted samples, FePt (001) peak was absent indicating a complete conversion of the sample into the soft phase. Moreover, the magnetic measurements showed a sharp decrease in coercivity and saturation magnetization in all the cases except low energy $^{40}\text{Ar}^+$ ion implantation. Besides, the remanence in OOP direction became less than that in IP direction in all the cases, except low energy $^{40}\text{Ar}^+$ ion implanted sample. This was explained based on a decrease in the anisotropy constant of the sample. These attributions were further confirmed from the torque measurements and MFM domain maps. The results of these studies revealed many novel observations to find a suitable fabrication technique for mass production of the HD MR media. However, further research is needed to improve the magnetic patterning process.

The domain wall (DW) racetrack memory, which has been under intense research for more than two decades, is another potential contender for the replacement of HDDs based on PMR technology [22-26]. Although there has been a lot of scientific advancement in DW motion, writing, and reading of the DWs, there are still many issues, which need to be addressed. These issues are, (i) controlled and reliable motion of DWs, (ii) thermal stability of the DW memory devices, (iii) large intrinsic pinning defects, (iv) large operational power, (v) larger size of the DW memory devices, and (vi) lack of efficient motion of multiple DWs.

Artificial pinning sites have been proposed to solve the problem of controlled and reliable DW motion [27]. Moreover, they help in increasing the thermal stability and hence the lifetime of the DW memory devices [27]. One major problem with most of the existing approaches is the need for increasing the value of the magnetic field or electrical current to push the DWs from the successive pinning sites. This limits the number of pinning sites and hence the storage capacity of the DW memory device. We have proposed and studied two distinct methods for fabricating the artificial pinning sites in this regard.

First, we studied the concept of square wave shaped nanowire. The Laplace pressure on the DW at the corner of two adjoining segments of DW dominates the applied STT on the DW and DW pins at the corner. We observed the DW pinning for a larger range of current densities. However, the pinning strength was found to be weaker for a current density larger than 4×10^{11} A/m². This was because the DW pins for some time at the pinning sites and then depins. Moreover, the pinning time decreased in going from the first pinning site to the second. Therefore, the range of current densities for which pinning strength is significant was observed to be quite small. However, with a wise choice of the pulse width and current density, pinning at multiple pinning sites can be realized. In order to compensate for the drawback of this concept, we moved to another concept.

We found that, when the interfacial Dzyaloshinskii–Moriya Interaction (*iDMI*) is introduced in a nanoscale region of a nanowire, the DW gets pinned in this *iDMI* region. We found that the DW pins for a large range of current densities, the width of the nanoscale *iDMI* region, and the magnitude of *iDMI*. Moreover, the pinning strength of these pinning sites was observed to be much higher. We also performed simulations to achieve the pinning at the multiple successive pinning sites. We found that without increasing the magnitude and pulse width of the current density, DW pins and depins at five successive pinning sites. Besides, we have also observed that the DW oscillates for a certain range of the width of the *iDMI* region, *iDMI* values, and current densities. We systematically studied the origin of these DW oscillations and their dependence on external stimulus, geometrical, and magnetic parameters. Our results were well explained by the micromagnetic theory. Interestingly, the frequency of the DW oscillations can be tuned from 1.47 GHz to 4.97 GHz. We also developed an analytical model, which revealed that the nature of DW oscillations is similar to the classical half-harmonic oscillator, and the frequency of oscillations depends on the square root of the *iDMI* field. These results add great value to the basic understanding as well as applications in the field of DW devices.

In the third project of this thesis, we have attempted to reduce the operational power and large intrinsic pinning defects and to control the DW motion reliably. Researchers have proposed a few methods like gate voltage, simultaneous application of magnetic field, and current to reduce the overall power consumption while moving the DWs [28-30]. However, the observed reduction in power is not significant. In recent years, it has been studied that the spin-orbit torque (SOT) from the bulk spin Hall effect and interfacial Rashba effect in the adjacent heavy metal (HM) can be used to reduce the power consumption while moving the DWs in ferromagnetic (FM) layer (in FM/HM structure) [31]. The power consumption in SOT driven DW motion depends on the efficiency of the SOT, which is related to the efficiency of charge

current to spin current conversion. The factor which exhibits the information about the charge to spin current conversion is called spin Hall angle (θ_{SH}). Researchers have studied several materials to enhance the θ_{SH} and found that β -W gives rise to the highest spin Hall angle. However, the β phase in W films is attributed to high Ar gas pressure (HP) and low deposition power during the making of the thin films. This gives rise to high resistivity in the W layer, which is not good for the low power operation of the DW devices. Moreover, films grown at high deposition pressure are expected to be granular. This may increase the intrinsic defects in the adjacent ferromagnetic layer. To avoid these issues, we proposed a novel material engineering where a W film deposited at low Ar gas pressure (LP) is sandwiched between the β -W and ferromagnetic layer.

To optimize the deposition parameters for the dual W layer structure, we first deposited HP and LP-W (2-8 nm)/CoFeB (1 nm)/MgO (~1 nm)/Ru (2 nm) structure and characterized them. We found out that the stacks with HP W layer thickness less than or equal to 6 nm, are in β phase. Therefore, for dual W layer structures, we decided the total thickness of the W layer to be 6 nm. We found that the samples with the approximately same thickness of LP and HP W in the dual-layer stack exhibit the β phase. Moreover, all the film stacks were found to exhibit perpendicular magnetic anisotropy in the as-deposited state. Based on our results on the thin film level, we fabricated the devices from the dual W layer sample, which shows the β phase. We observed that the DW can be moved at a current density as low as 10^6 A/m². Thus, we reported a reduction of current density by an order of 10^4 compared to existing literature. This is an extremely significant result for the low power DW device.

We also observed the DW pinning and depinning of DWs at multiple pinning sites in meander wire. Like the straight (reference) nanowire, the DW was found to move at very small current densities for the case of meander nanowire as well. These results add significance to low power

DW devices for DW memory applications. Besides, the results are also useful for emerging neuromorphic computing applications.

8.2. Future Works

8.2.1. Ultra-low Dimensional Domain Wall Devices

The dimensions of DW devices involved in most of the current research are 100 nm or more [32, 33]. The DW devices with this size may not achieve the high storage capacity, a touted advantage of DW memory. Therefore, the size of DW devices needs to be reduced. As mentioned earlier that energy-assisted magnetic recording (EAMR) is predicted to achieve an areal density of 4 Tbps. The storage capacity of 40 TB can be attained if one can stack 12 platters in an HDD device. On contrary, if one fabricates a DW device with the following dimension parameter: width of nanowire= width of one domain= width of DW wall= 20 nm, an areal density of 3.4 Tbps can be accomplished. However, there is no restriction on the number of layers of nanowires (no mechanical reading and writing head) for DW devices. Thus, if one fabricates 64 such layers, a storage capacity of ~140 TB can be achieved. Therefore, higher storage capacity is the main advantage of DW racetrack memory [27].

Self-assembly of PS-PDMS DBCP is an excellent method to achieve DW devices with a width of sub-10 nm or lesser [34]. The self-assembly has been reported to fabricate structures as small as 3 nm [34]. In addition to the advantages explained in chapter 2 and 4, the self-assembly process results in better control of the size of the patterns. Moreover, it minimizes the size distribution of the patterns [14]. In the current thesis, we have optimized the self-assembly process for fabricating the dot patterns of average sizes of (20.5 ± 2.7) nm. In relation to DW devices, the lamellar and cylindrical patterns of self-assembly are needed to fabricate writing, memory, and reading elements of DW racetrack memory. Therefore, a separate PS-PDMS DBCP with the required arrangement in the volume fraction of PS and PDMS units is needed.

Moreover, a reduction in the molecular weight of DBCP results in the reduction in the size of the patterns [35]. Thus, one needs to utilize the PS-PDMS DBCP with smaller molecular weight (say 10 or 15 kg/mol).

In addition, the fabrication of DW memory devices will need a multistep lithography process. This is difficult to achieve using the self-assembly process in a controlled manner. However, this is achievable by using the directed self-assembly (DSA) process, where the position of patterns is guided by a supplementing lithography technique [34, 36]. Figure 8.1 shows an exemplary depiction showing the capability of DSA of PS-PDMS DBCP [37, 38]. Several patterns such as nanoholes, nanodots, horizontal nanorods can be fabricated using DSA. Moreover, complex patterns like zig-zag nanowire, cuboids, square wave shaped nanowire, meander nanowire, etc. can also be fabricated using the DSA process [39-43].

Figure 8.1. Exemplary images showing various patterns made using DSA of PS-PDMS DBCP (adopted from [37, 38]).

8.2.2. Experimental Observation of Domain Wall Dynamics in Square Wave Shaped Nanowire and Nanowire with Nanoscale Interfacial Dzyaloshinskii–Moriya Interaction (*i*DMI) Region

We have researched the DW dynamics in square wave shaped nanowire and nanowires with nanoscale *i*DMI region using the micromagnetic simulations. Moreover, we have studied the

dependence of DW dynamics on several related parameters and observed that the DW results in interesting features, in particular for the nanowire with nanoscale *i*DMI region. The next step should be the experimental realization of the discovered concepts. During the simulations, we optimized the material, device parameters, and magnitude of an external stimulus (spin-transfer torque (STT) in our case). Following our findings in simulations, one needs to first deposit the thin films, which are $(\text{Co/Ni})_n$ multilayers as predicted from the simulations. After characterizing the thin films, one should start fabricating the devices. A brief suggestive discussion of the device configuration (to be fabricated) is shown in Figure 8.2. For the case of square wave shaped nanowire, one can directly fabricate devices and electrode pads in two separate lithography steps and put the devices forward for testing.

Figure 8.2. A brief experimental process flow for achieving DW pinning in (i) square wave shaped nanowire and (ii) DW pinning and oscillations in nanowire with nanoscale DMI region.

However, the device fabrication of devices with nanoscale *i*DMI is relatively complex. The first step in device fabrication requires the fabrication of the nanowire constituting the memory unit of the DW device. The next step includes the fabrication of the crossbars of a heavy metal layer for introducing *i*DMI in the nanoscale region. The magnitude of *i*DMI for observing DW pinning is less and can be obtained by fabricating the crossbars of Pt [44-47]. However, for DW oscillations the *i*DMI values are relatively larger. Therefore, one needs to prepare the HM system which can provide the *i*DMI values in the required range. Several studies have proposed a quite high *i*DMI value in the past [45, 46, 48, 49]. Taking the existing research as a starting point, one needs to optimize a crossbar material stack for the proposed study. Once, the crossbar fabrication is done, one can fabricate electrodes and test the devices. Other than the DW memory devices the observations of DW pinning and DW oscillations are useful in making the artificial neuromorphic synapses. This aspect of DW devices will be discussed in the upcoming sub-section.

8.2.3. On-chip Neurons and Synapses for Neuromorphic Computing

The observation of ultra-low-power DW motion (chapters 6 and 7) is a big boost for Spintronic devices, in particular for Neuromorphic devices based on DW motion. We have studied the concept of meander nanowire and observed stochastic and non-stochastic DW pinning in separated devices at low current densities. Moreover, we have observed DW motion in straight nanowire at extremely low current densities of 1×10^6 A/m². These observations are useful in making ultra-low-power stochastic and non-stochastic synapses and neurons, respectively. In summary, we have made individual building blocks of neuromorphic devices operating at ultra-low power. In the next step, one should plan to integrate these on a single chip and test the neuromorphic operations.

Figure 8.3. (a) The schematic illustration of the biological neural network and (b) an example of a neural network made up of the DW devices [50].

A biological neural network consists of neurons and synapses (as discussed before). Several neurons transfer a weighted output to the post neuron, which acts as an input for the post neuron. If the signal needs to be processed further, the neuron fires and transfers the signal through the axon. This process is schematically shown in Figure 8.3 (a) [50]. Figure 8.3 (b) shows an example of a proposed neural network based on DW devices [50]. The synapses can be made up of the meander nanowire and neurons can be made up of the straight nanowire. Moreover, the axon consists of a normal MTJ within the electronic circuit as shown in Figure 8.3 (b) [50]. However, before using the proposed devices for the neuromorphic units, an insulating layer and a fixed layer need to be deposited on these devices. For the meander nanowire, the insulating and fixed layers will cover the whole device. On the contrary, for straight nanowires, the insulating and fixed layers will be deposited at the right corner (please refer to Figure 7.1 in chapter 7).

However, in our experiments, the width of the neuromorphic devices is 10 μm , which needs to be scaled to sub-100 nm or lesser in the next step. With DSA of PS-PDMS DBCP (future work 1), one may also make devices of sub-10 nm sizes [34].

References

- [1] <https://www.socialmediatoday.com/news/what-happens-on-the-internet-every-minute-2020-version-infographic/583340/>.
- [2] <https://www.seagate.com/files/www-content/our-story/trends/files/idc-seagate-dataage-whitepaper.pdf>.
- [3] G. Varvaro, S. Laureti and D. Fiorani, "L10 FePt-based thin films for future perpendicular magnetic recording media," *Journal of magnetism and magnetic materials*, vol. 368, pp. 415, 2014.
- [4] G. Varvaro and F. Casoli, *Ultra-high-density magnetic recording: storage materials and media designs*. CRC Press, 2016.
- [5] <https://blog.seagate.com/craftsman-ship/hamr-next-leap-forward-now/>.
- [6] <https://blog.westerndigital.com/the-need-for-energy-assisted-recording-technology/>.
- [7] https://www.theregister.co.uk/2018/12/07/toshiba_goes_to_mamr/.
- [8] <https://hexus.net/tech/news/storage/123953-seagates-hdd-roadmap-teases-100tb-drives-2025/>.
- [9] W. Tipcharoen, C. Warisarn, A. Kaewrawang and P. Kovintavewat, "Effect of hotspot position fluctuation to writing capability in heated-dot magnetic recording," *Japanese Journal of Applied Physics*, vol. 55, no. 7S3, p. 07MB01, 2016.
- [10] R. Medwal, S. Gautam, S. Gupta, K. H. Chae, K. Asokan, G. R. Deen, R. S. Rawat, R. S. Katiyar and S. Annapoorni, "Self-Stabilized Carbon-L10 FePt Nanoparticles for Heated Dot Recording Media," *IEEE Magnetics Letters*, vol. 9, pp. 1, 2018.
- [11] K. Hono, Y. Takahashi, G. Ju, J.-U. Thiele, A. Ajan, X. Yang, R. Ruiz and L. Wan, "Heat-assisted magnetic recording media materials," *MRS Bulletin*, vol. 43, no. 2, p. 93, 2018.
- [12] W. Tipcharoen, C. Warisarn, D. Tongsoomporn, D. Karns and P. Kovintavewat, "Investigation of writing error in staggered heated-dot magnetic recording systems," *AIP Advances*, vol. 7, no. 5, p. 056511, 2017.
- [13] C. Vogler, C. Abert, F. Bruckner, D. Suess and D. Praetorius, "Heat-assisted magnetic recording of bit-patterned media beyond 10 Tb/in²," *Applied Physics Letters*, vol. 108, no. 10, p. 102406, 2016.
- [14] R. Ruiz, H. Kang, F. A. Detcheverry, E. Dobisz, D. S. Kercher, T. R. Albrecht, J. J. de Pablo and P. F. Nealey, "Density multiplication and improved lithography by directed block copolymer assembly," *Science*, vol. 321, no. 5891, pp. 936, 2008.
- [15] N. GAUR, "Investigations on advanced magnetic recording media," 2012.
- [16] P. Candeloro, A. Gerardino, E. Di Fabrizio, S. Cabrini, G. Giannini, L. Mastrogiacomo, M. Ciria, R. C. O'Handley, G. Gubbiotti and G. Carlotti, "Patterned magnetic permalloy and nickel films: fabrication by electron beam and x-ray lithographic techniques," *Japanese journal of applied physics*, vol. 41, no. 8R, p. 5149, 2002.
- [17] A. Kikitsu, T. Maeda, H. Hieda, R. Yamamoto, N. Kihara and Y. Kamata, "5 Tdots/in² bit patterned media fabricated by a directed self-assembly mask," *IEEE transactions on magnetics*, vol. 49, no. 2, pp. 693, 2013.
- [18] G. Harp, D. Weller, T. Rabedeau, R. Farrow and M. Toney, "Magneto-optical Kerr spectroscopy of a new chemically ordered alloy: Co₃Pt," *Physical review letters*, vol. 71, no. 15, p. 2493, 1993.
- [19] Y. Yamada, W. Van Drent, E. Abarra and T. Suzuki, "High perpendicular anisotropy and magneto-optical activities in ordered Co₃Pt alloy films," *Journal of applied physics*, vol. 83, no. 11, pp. 6527, 1998.

- [20] R. Bandhu, R. Sooryakumar, R. C. Farrow, D. Weller, M. Toney and T. Rabedeau, "Elastic properties of chemically ordered Co₃Pt thin films," *Journal of applied physics*, vol. 91, no. 5, pp. 2737, 2002.
- [21] S. Piramanayagam and T. C. Chong, *Developments in data storage: materials perspective*. John Wiley & Sons, 2011.
- [22] S. S. Parkin, M. Hayashi and L. Thomas, "Magnetic domain-wall racetrack memory," *Science*, vol. 320, no. 5873, pp. 190, 2008.
- [23] L. Gan, S. Chung, K. Aschenbach, M. Dreyer and R. Gomez, "Pulsed-current-induced domain wall propagation in permalloy patterns observed using magnetic force microscope," *IEEE Transactions on Magnetics*, vol. 36, no. 5, pp. 3047, 2000.
- [24] A. Yamaguchi, T. Ono, S. Nasu, K. Miyake, K. Mibu and T. Shinjo, "Real-space observation of current-driven domain wall motion in submicron magnetic wires," *Physical review letters*, vol. 92, no. 7, p. 077205, 2004.
- [25] I. M. Miron, T. Moore, H. Szabolcs, L. D. Buda-Prejbeanu, S. Auffret, B. Rodmacq, S. Pizzini, J. Vogel, M. Bonfim and A. Schuhl, "Fast current-induced domain-wall motion controlled by the Rashba effect," *Nature materials*, vol. 10, no. 6, pp. 419, 2011.
- [26] W. Li, Z. Jin, D. Wen, X. Zhang, M. Qin and J.-M. Liu, "Ultrafast domain wall motion in ferrimagnets induced by magnetic anisotropy gradient," *Physical Review B*, vol. 101, no. 2, p. 024414, 2020.
- [27] D. Kumar, T. Jin, S. Al Risi, R. Sbiaa, W. Lew and S. Piramanayagam, "Domain wall motion control for racetrack memory applications," *IEEE Transactions on Magnetics*, vol. 55, no. 3, pp. 1, 2018.
- [28] A. Schellekens, A. Van den Brink, J. Franken, H. Swagten and B. Koopmans, "Electric-field control of domain wall motion in perpendicularly magnetized materials," *Nature communications*, vol. 3, no. 1, pp. 1, 2012.
- [29] T. Koyama, K. Ueda, K.-J. Kim, Y. Yoshimura, D. Chiba, K. Yamada, J.-P. Jamet, A. Mougin, A. Thiaville and S. Mizukami, "Current-induced magnetic domain wall motion below intrinsic threshold triggered by Walker breakdown," *Nature nanotechnology*, vol. 7, no. 10, p. 635, 2012.
- [30] K.-J. Kim, Y. Yoshimura, W. S. Ham, R. Ernst, Y. Hirata, T. Li, S. Kim, T. Moriyama, Y. Nakatani and T. Ono, "Energy-efficient writing scheme for magnetic domain-wall motion memory," *Applied Physics Express*, vol. 10, no. 4, p. 043002, 2017.
- [31] R. Ramaswamy, J. M. Lee, K. Cai and H. Yang, "Recent advances in spin-orbit torques: Moving towards device applications," *Applied Physics Reviews*, vol. 5, no. 3, p. 031107, 2018.
- [32] Z. Luo, A. Hrabec, T. P. Dao, G. Sala, S. Finizio, J. Feng, S. Mayr, J. Raabe, P. Gambardella and L. J. Heyderman, "Current-driven magnetic domain-wall logic," *Nature*, vol. 579, no. 7798, pp. 214, 2020.
- [33] S. Vélez, J. Schaab, M. S. Wörnle, M. Müller, E. Gradauskaite, P. Welter, C. Gutgsell, C. Nistor, C. L. Degen and M. Trassin, "High-speed domain wall racetracks in a magnetic insulator," *Nature communications*, vol. 10, no. 1, pp. 1, 2019.
- [34] R. A. Griffiths, A. Williams, C. Oakland, J. Roberts, A. Vijayaraghavan and T. Thomson, "Directed self-assembly of block copolymers for use in bit patterned media fabrication," *Journal of Physics D: Applied Physics*, vol. 46, no. 50, p. 503001, 2013.
- [35] T. Xu, H.-C. Kim, J. DeRouchey, C. Seney, C. Levesque, P. Martin, C. Stafford and T. Russell, "The influence of molecular weight on nanoporous polymer films," *Polymer*, vol. 42, no. 21, pp. 9091, 2001.
- [36] X. Gu, I. Gunkel and T. P. Russell, "Pattern transfer using block copolymers," *Philosophical Transactions of the Royal Society A: Mathematical, Physical and Engineering Sciences*, vol. 371, no. 2000, p. 20120306, 2013.

- [37] A. T. KG, K. Gotrik, A. Hannon, A. Alexander-Katz, C. Ross and K. Berggren, "Templating three-dimensional self-assembled structures in bilayer block copolymer films," *Science*, vol. 336, no. 6086, pp. 1294, 2012.
- [38] A. T. KG, S. M. Nicaise, K. R. Gadelrab, A. Alexander-Katz, C. A. Ross and K. K. Berggren, "Multilayer block copolymer meshes by orthogonal self-assembly," *Nature communications*, vol. 7, no. 1, pp. 1, 2016.
- [39] M. Takenaka, S. Aburaya, S. Akasaka, H. Hasegawa, N. Hadjichristidis, G. Sakellariou, Y. Tada and H. Yoshida, "Formation of long-range stripe patterns with sub-10-nm half-pitch from directed self-assembly of block copolymer," *Journal of Polymer Science Part B: Polymer Physics*, vol. 48, no. 22, pp. 2297, 2010.
- [40] J. W. Jeong, W. I. Park, L. M. Do, J. H. Park, T. H. Kim, G. Chae and Y. S. Jung, "Nanotransfer Printing with sub-10 nm Resolution Realized using Directed Self-Assembly," *Advanced Materials*, vol. 24, no. 26, pp. 3526, 2012.
- [41] S.-M. Park, X. Liang, B. D. Harteneck, T. E. Pick, N. Hiroshiba, Y. Wu, B. A. Helms and D. L. Olynick, "Sub-10 nm nanofabrication via nanoimprint directed self-assembly of block copolymers," *ACS nano*, vol. 5, no. 11, pp. 8523, 2011.
- [42] R. G. Hobbs, R. A. Farrell, C. T. Bolger, R. A. Kelly, M. A. Morris, N. Petkov and J. D. Holmes, "Selective sidewall wetting of polymer blocks in hydrogen silsesquioxane directed self-assembly of PS-b-PDMS," *ACS applied materials & interfaces*, vol. 4, no. 9, pp. 4637, 2012.
- [43] J. K. Yang, Y. S. Jung, J.-B. Chang, R. Mickiewicz, A. Alexander-Katz, C. Ross and K. K. Berggren, "Complex self-assembled patterns using sparse commensurate templates with locally varying motifs," *Nature nanotechnology*, vol. 5, no. 4, p. 256, 2010.
- [44] E. C. Corredor, S. Kuhrau, F. Kloodt-Twesten, R. Frömter and H. P. Oepen, "SEMPA investigation of the Dzyaloshinskii-Moriya interaction in the single, ideally grown Co/Pt (111) interface," *Physical Review B*, vol. 96, no. 6, p. 060410, 2017.
- [45] Y. Ishikuro, M. Kawaguchi, N. Kato, Y.-C. Lau and M. Hayashi, "Dzyaloshinskii-Moriya interaction and spin-orbit torque at the Ir/Co interface," *Physical Review B*, vol. 99, no. 13, p. 134421, 2019.
- [46] A. Cao, X. Zhang, B. Koopmans, S. Peng, Y. Zhang, Z. Wang, S. Yan, H. Yang and W. Zhao, "Tuning the Dzyaloshinskii-Moriya interaction in Pt/Co/MgO heterostructures through the MgO thickness," *Nanoscale*, vol. 10, no. 25, pp. 12062, 2018.
- [47] S. Tacchi, R. Troncoso, M. Ahlberg, G. Gubbiotti, M. Madami, J. Åkerman and P. Landeros, "Interfacial Dzyaloshinskii-Moriya interaction in Pt/CoFeB films: effect of the heavy-metal thickness," *Physical review letters*, vol. 118, no. 14, p. 147201, 2017.
- [48] H. Yang, A. Thiaville, S. Rohart, A. Fert and M. Chshiev, "Anatomy of dzyaloshinskii-moriya interaction at Co/Pt interfaces," *Physical review letters*, vol. 115, no. 26, p. 267210, 2015.
- [49] K. Shahbazi, J.-V. Kim, H. T. Nembach, J. M. Shaw, A. Bischof, M. D. Rossell, V. Jeudy, T. A. Moore and C. H. Marrows, "Domain-wall motion and interfacial Dzyaloshinskii-Moriya interactions in Pt/Co/Ir (t Ir)/Ta multilayers," *Physical Review B*, vol. 99, no. 9, p. 094409, 2019.
- [50] A. Sengupta, Y. Shim and K. Roy, "Proposal for an all-spin artificial neural network: Emulating neural and synaptic functionalities through domain wall motion in ferromagnets," *IEEE transactions on biomedical circuits and systems*, vol. 10, no. 6, pp. 1152, 2016.