

**NANYANG
TECHNOLOGICAL
UNIVERSITY**

SINGAPORE

**INVESTIGATIONS ON CONTROL OF DOMAIN
WALL MOTION USING SYNTHETICALLY
TEXTURED MAGNETIC NANOSTRUCTURES**

JIN TIANLI

**SCHOOL OF PHYSICAL AND MATHEMATICAL
SCIENCES**

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WALL MOTION USING SYNTHETICALLY
TEXTURED MAGNETIC NANOSTRUCTURES**

JIN TIANLI

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

2019

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.

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06/06/2019

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Date



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S. N. Piramanayagam

Authorship Attribution Statement

This thesis contains materials chapter 4 and chapter 5 published in the following peer-reviewed journals in which I am listed as an author. Chapter 6 published in the peer-reviewed journals and the 2018 INTERMAG conference in which I am listed as an author.

The chapter 4 is partially published as Tianli Jin, Weiliang Gan, Chuang Ma, Rudolf Schaefer, Funan Tan, Wen Siang Lew, Ivan Soldatov, Xiaoxi Liu, Wai Cheung Law, S. N. Piramanayagam*, “Nanoscale modification of magnetic properties for effective domain wall pinning”, *J. Magn. Mater.*, **475**, 70 (2019). DOI: <https://doi.org/10.1016/j.jmmm.2018.11.114>.

The rest is published as “Tuning magnetic properties for domain wall pinning via localized metal diffusion”, Tianli Jin, Mojtaba Ranjbar, Shikun He, Wai Cheung Law, Tiejun Zhou, Wen Siang Lew, Xiaoxi Liu and S. N. Piramanayagam*, *Sci. Rep.*, **7**, 16208 (2017). DOI: <https://doi.org/10.1038/s41598-017-16335-z>.

The contributions of the co-authors are as follows:

- A/Prof S. N. Piramanayagam provided the initial project direction.
- I fabricated the thin film and devices, performed the measurements and micromagnetic simulations. S. N. Piramanayagam, Mojtaba Ranjbar and I analyzed the data and prepared the manuscript drafts.
- Shikun He and Tiejun Zhou assisted in FMR and TEM measurements.
- Funan Tan and Wai Cheung Law coded the Labview which the measurement process ran on. They also assisted during the measurement setup.

- Weiliang Gan assisted in the micromagnetic simulation coding.
- Chuang Ma, Xiaoxi Liu, Rudolf Schaefer and Ivan Soldatov assisted in MOKE measurement.
- A/Prof Lew Wen Siang provides the electrical measurement setup.

The manuscript was revised with all of the authors.

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The contributions of the co-authors are as follows:

- A/Prof S. N. Piramanayagam and Mojtaba Ranjbar suggested the materials area.
- I fabricated device, performed the measurements and micromagnetic simulations.
- A/Prof S. N. Piramanayagam and I wrote the drafts of the manuscript. The manuscript was revised together with Rachid Sbiaa, Xiaoxi Liu.
- Mojtaba Ranjbar carried out the thin film deposition process.
- Durgesh Kumar performed the X-ray diffraction (XRD) measurement and transport of ions in matter (TRIM) simulation.
- Weiliang Gan, Feilong Luo and Wen Siang Lew assisted in polar Kerr images measurement. They also assisted in analyzing the data.
- B⁺ implantation was carried out in Nanyang NanoFabrication Centre.

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The contributions of the co-authors are as follows:

- I designed the study, fabricated the device, and performed the measurements and micromagnetic simulations.
- Funan Tan carried out the thin film deposition process using magnetron sputtering.
- A/Prof S. N. Piramanayagam, Funan Tan and I wrote the drafts of the manuscript.
- Calvin Ang Ching Ian and Weiliang Gan assisted in simulation coding and MOKE measurements.
- Prof Jiangwei Cao and A/Prof Wen Siang Lew assisted in analyzing the data.

The manuscript was revised with all of the authors.

26/05/2019

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Abstract

Domain wall motion-based devices such as racetrack memory have been proposed as promising candidates for high capacity, non-volatile information storage. In these devices, multiple domain walls can be propagated through nanowires at a speed of several hundred meters per second using the magnetic field, electrical currents and both, thus allowing data to be written, read and processed efficiently. A major challenge towards the commercial realization for domain wall memories is the stochasticity of domain wall motion. Such stochastic behavior limits the reliability and hence maximum data density. To overcome this issue, researchers have proposed and investigated the concept of forming artificial pinning centers at specific positions along the nanowire, such as creating notches to alter energy barriers at specific positions and forming exchange bias. In this thesis, I investigated the use of synthetic magnetic textures to control and pin domains in nanowires. The synthetic magnetic textures can be achieved by using non-magnetic metal diffusion into magnetic nanowires or ions implantation.

In NiFe nanowire, the orthogonal Cr crossbars as overlayer on NiFe nanowire were deposited. Then the device was annealed at 400 °C for one hour to allow Cr atoms diffusion into NiFe nanowire at the cross positions, which causes the change of the magnetic properties of NiFe in the saturation magnetization (M_s) and the damping constant (α). We performed wide-field magneto-optical Kerr effect (MOKE) and magnetoresistance measurements to evaluate the pinning field about 100 Oe. Subsequently, as a second concept, we studied the boron ions (B^+) implantation effect into Co/Pd multilayer-based nanowires. We have observed that B^+ implantation can reduce the anisotropy

energy of the Co/Pd multilayers and change the easy-axis of magnetization from perpendicular to in-plane direction. In (Co/Pd)₁₀ nanowire, the pinning field, realized by B⁺ implantation, can be as high as 900 Oe.

We also demonstrated the formation of pinning centers by using exchange interaction between films with perpendicular magnetic anisotropy (PMA) and in-plane magnetic anisotropy (IMA) to create locally tilted magnetization of the PMA layer. The thickness of a Pt spacer layer between the PMA layer and the IMA layer was varied to tune the orientation of magnetization. For Pt layer thickness below 2 nm, the magnetization switching behavior of the PMA and IMA layers suggests the presence of exchange interaction between those two layers. I also carried out micromagnetic simulations to further study the tilted angle change by tuning the exchange coupling strength and the depinning current with various tilted angles. The simulation results indicate that the depinning current density decreased with larger tilted magnetization angle and that the tilted angle can be optimized by tuning the exchange coupling strength between PMA and IMA layers.

As a summary, a synthetic magnetic texture can be achieved by modifying the magnetic properties of nanowire using various methods, such as thermal diffusion of non-magnetic species and local ion implantation. These magnetically modified regions are supposed to form the pinning sites. In comparison with lithographically fabricated notches along a nanowire, which is non-uniform and increases the local heating at the position of the notches, the methods proposed in this study, can potentially be used in future domain wall memory applications.

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List of abbreviations

AI	Artificial intelligence
IoT	Internet of Things
HDD	Hard disk drive
HAMR	Heat-assisted magnetic recording
MARM	Microwave-assisted magnetic recording
SNR	Signal-to-noise ratio
SSD	Solid states drive
DWRM	Domain wall racetrack memory
DW	Domain wall
Tbpsi	Terabits per square inch
STO	Spin torque oscillator
GMR	Giant magnetoresistance
TMR	Tunnel magnetoresistance
MTJ	Magnetic tunnel junction
STT	Spin-transfer torque
SOT	Spin-orbit torque
SOC	Spin-orbit coupling
IMA	In-plane magnetic anisotropy
PMA	Perpendicular magnetic anisotropy
HH	Head to head
TT	Tail to tail
LLG	Landau–Lifshitz–Gilbert
LLGS	Landau-Lifshitz-Gilbert–Slonczewski
VSM	Vibrating sample magnetometer
AGFM	Alternating gradient force magnetometer
FMR	Ferromagnetic resonance
VNA	Vector network analyzer
DUT	Device under test

RF	Radio Frequency
AFM	Atomic force microscopy
SPM	Scanning probe microscopy
MFM	Magnetic force microscopy
SEM	Scanning electron micrograph
TEM	Transmission electron microscopy
XPS	X-ray photoelectron spectroscopy
EDX	Energy dispersive X-ray
XRD	X-Ray diffraction
TRIM	Transport of ions in matter
MOKE	Magneto-optical Kerr effect
EBL	Electron beam lithography
AMR	Anisotropy magnetoresistance
MR	Magnetoresistance
DC	Direct current
T_{an}	Annealing temperature
SI	Shallow-implantation
MI	Medium-implantation
DI	Deep-implantation
OOP	Out-of-plane
IP	In-plane
SFD	Switching field distribution
DMI	Dzyaloshinskii-Moriya interaction

List of publications

Journal publications

- 1, “*Nanoscale modification of magnetic properties for effective domain wall pinning*”

Tianli Jin, Funan Tan, Wai Cheung Law, Weiliang Gan, Ivan Soldatov, Rudolf Sch äfer, Chuang Ma, Xiaoxi Liu, Wen Siang Lew, and S. N. Piramanayagam*

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- 2, “*Synaptic element for neuromorphic computing using magnetic domain wall device with synthetic pinning sites*”

Tianli Jin, Weiliang Gan, Funan Tan, N. S. Sernicola, Wen Siang Lew, and S.N. Piramanayagam*

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- 3, “*Tilted magnetisation for domain wall pinning in racetrack memory*”

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- 4, “*Nanoscale compositional modification in Co/Pd multilayers for controllable domain wall pinning in racetrack memory*”

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- 5, “*Tuning magnetic properties for domain wall pinning via localized metal diffusion*”

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Scientific Reports, **7**, 16208 (2017).

6, “Switching domain wall motion on and off using a gate voltage for novel transistor applications”

Chuang Ma,[†] **Tianli Jin**,[†] Xiaoxi Liu*, and S. N. Piramanayagam*

Applied Physics Letters, **113**, 232401 (2018) (**Co-first author**).

7, “Staggered magnetic nanowire devices for effective domain wall pinning in racetrack memory”

M. Al Bahri, B. Borie, **Tianli Jin**, R. Sbiaa*, M. Kläui, and S. N. Piramanayagam

Physical Review Applied, **11**, 024023 (2019).

8, “Perpendicular Magnetic Anisotropy in Co/Pt multilayers induced by hcp-Ho at 400 °C”

Wai Cheung Law, **Tianli Jin**, Xin Tong Zhu, Ramesh Rao Nistala, Naganivetha Thiyagarajah, Chim Seng Seet, Wen Siang Lew*

Journal of Magnetism and Magnetic Materials, **477**, 124 (2019).

9, “Electric field control on single and double Pt/Co heterostructure for enhanced thermal stability”

Funan Tan, G. J. Lim, **Tianli Jin**, H. X. Liu, F. Poh, and W. S. Lew*

Journal of Magnetism and Magnetic Materials, **490**, 165448 (2019).

10, “Domain Wall Memory: Physics, Materials and Devices”

Durgesh Kumar, **Tianli Jin**, Rachid Sbiaa, Mathias Kläui, Subhankar Bedanta, Xiaoxi Liu, and S.N. Piramanayagam*

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11, “Controlled spin-torque driven domain wall motion using staggered magnetic nanowires”

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Physical Review Applied, submitted.

12, “Domain Wall Motion Control for Racetrack Memory Applications”

Durgesh Kumar, **Tianli Jin**, S. Al Risi, R. Sbiaa, Wen Siang Lew, S.N. Piramanayagam*

IEEE. Transactions on Magnetics, **55**, 2300708 (2019).

13, “*High temperature ferromagnetic resonance study on pMTJ stacks with diffusion barrier layers*”

Wai Cheung Law, T. Tahmasebi, Funan Tan, **Tianli Jin**, Weiliang Gan, R. R. Nistala, X. T. Zhu, Z. Q. Mo, H. W. Teo, C. S. Seet, A. See, S. N. Piramanayagam, and W. S. Lew*

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14, “*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*

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15, “*Spin orbit torques induced magnetization reversal through asymmetric domain wall propagation in Ta/CoFeB/MgO structures*”

Jiangwei Cao*, Yifei Chen, **Tianli Jin**, Weiliang Gan, Ying Wang, Yuqiang Zheng, Hua Lv, Susana Cardoso, Dan Wei, Wen Siang Lew

Scientific Reports, **8**, 1355 (2018).

16, “*Broadband strip-line ferromagnetic resonance spectroscopy of soft magnetic CoFeTaZr patterned thin films*”

S. Gupta, D. Kumar, **T. L. Jin**, R. Nongjai, K. Asokan, A. Ghosh, M. Aparnadevi, P. Suri, and S. N. Piramanayagam*

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17, “*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*

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Conference presentations

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21, “*Tilted magnetization by exchange interaction for domain wall pinning in racetrack memory*”

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22, “*Tailoring magnetic texture for domain wall motion control in nanowires of Co/Pd multilayers*”

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23, “*Domain wall pinning studies by using synthetic magnetically textured nanostructure*”

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24, *“Tuning magnetodynamic properties in NiFe film with localized diffusion for domain wall devices”*

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25, *“Novel method to pin domain walls for domain wall memory”*,

Tianli Jin, ..., S. N. Piramanayagam*,

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26, *“Tailoring magnetic properties of NiFe diffused with non-magnetic metals for spin torque oscillators”*,

Tianli Jin, ..., S. N. Piramanayagam*,

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Chapter 1 Introduction

1.1 Background

Electronic devices have been an integral part of our daily lives. They will continue to play a significant role, especially with the advent of artificial intelligence (AI), 5G communication technology and the Internet of Things (IoT) [1-6]. One key aspect of electronic devices is the memory and storage systems, which have been developed using different technologies to satisfy the growing digital data demands. Hard disk drives (HDDs) [7], optical drives (like CD/DVD) [8, 9], NAND flash memories (like USB flash drives, solid-state drives (SSDs)) [10, 11], and different types of random access memories (RAMs) [10, 12-15] are used for storage and memory. However, it is important to note that each memory device has its pros and cons. Thus, different memory devices are used for different purposes, playing specific roles in data processing as well as storage. For example, hard disk drives are extensively used to store the enormous digital content, based on their high capacity and low price per bit compared to others. Whereas, RAMs are widely used as the components of register and cache memory, owing to their high operation speed and energy efficiency.

1.1.1 Hard disk drives

Over the last few decades, HDDs have been widely used in various computing and consumer electronic devices. This technology is based on reversing the magnetization of a certain area of a recording medium using an external magnetic field, which was first introduced by Oberlin Smith in 1878 [16]. The recording medium is a granular magnetic thin film, where a set of grains store a bit. In a simpler sense, the two directions of the magnetizations can be considered as “0”

or “1” of the binary information. In the state-of-the-art technology, the bits are so closely packed that the signals are analog-like and complex processing is carried out to decode 1s and 0s [17]. The reading/writing heads sense the direction of magnetization through measuring the magnetoresistance to read bits and reversing the direction of the grains to write bits. Figure 1.1 shows the schematic diagram for HDDs.

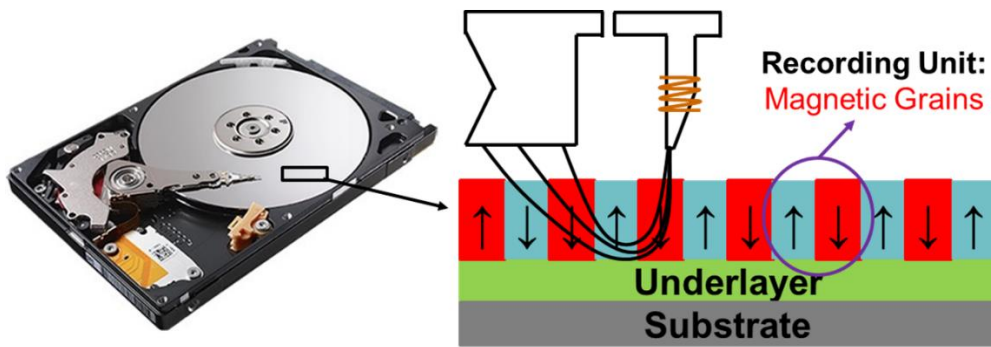


Figure 1.1. (Left) The photograph of HDD. (Right) Schematic diagram for HDD with perpendicular magnetization. The recording unit is magnetic grains with a different orientation, which points up and down recording the bits “1” and “0”. The reading head senses the direction of magnetization through measuring the magnetoresistance. Writing head switches the direction of magnetization in the grain to write bits.

The advent of major breakthroughs of HDDs can be categorized into four steps [18]. Firstly, breakthroughs in reading technology. The read head based on giant magnetoresistance (GMR) and tunnel magnetoresistance (TMR) can read the weaker magnetic features reliably owing to its high magnetoresistance ratio reaching ~600% at room temperature (in laboratory demonstrations) [19], compared to the read head based on anisotropy magnetoresistance (AMR), which magnetoresistance ratio is only 1~2% [20]. The breakthroughs in reading technology led to a period of rapid areal density increases exceeding 100% per year [21-26]. The reading technology moved from inductive heads to AMR and to GMR current-in-plane spin valves, which were subsequently replaced by

current-perpendicular-to-plane (CPP) magnetic tunneling junction (MTJ) readers, which yields even higher signal by improving the signal-to-noise ratio (SNR) [20, 26, 27].

Secondly, owing to some problems in longitudinal technology, such as the large demagnetizing fields and the superparamagnetic limit, a revolution took place in magnetic recording media, where the magnetization orientation was changed to perpendicular with respect to the disk plane. As a result, the areal density has been increased to 1 Tbps (Terabits per square inch) [28-30]. Thirdly, the development of highly anisotropic materials, *i.e.* high K_u , which provides a possibility of achieving small grains size with a high thermal stability. For example, iron-platinum alloy (FePt) with grains size about 8 nm have been developed, so that the smaller bit cell enables more bits to be crammed into a disk surface, which increases the areal density of data and the capacity [21-23]. Although the high anisotropy media can solve the thermal stability problem, the issue of writability comes into picture. As described by the media trilemma, sketched in Figure 1.2, a high writing field is required to switch the data bit of the high K_u material [7, 12, 13, 31].

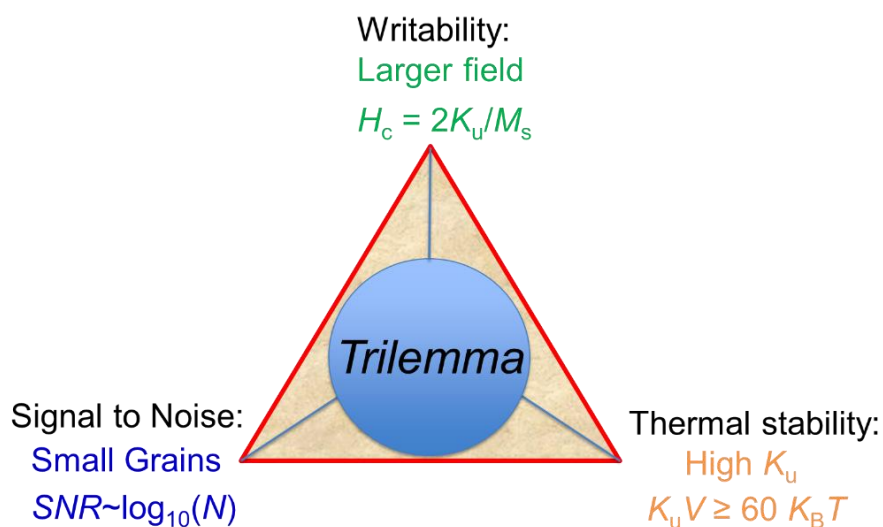


Figure 1.2. An illustration of the magnetic recording trilemma [31].

Lastly, there has been a significant improvement in writing technologies. As we discussed, FePt is the best candidate material as the recording layer, but the required writing field is enormous for FePt. Recently, Seagate reported the achievement of the high capacity of 16 Tbps in the 3.5-inch HDD platform, using heat-assisted magnetic recording (HAMR) technology to write information [32]. In HAMR technology, a small laser is attached to the writing head, to heat a tiny spot on the disk and reduce the coercivity of magnetic grains, where the data can be written by a small magnetic field [33, 34]. Moreover, another technology to achieving high areal density was proposed by Jimmy Zhu using microwave-assisted magnetic recording (MAMR) [35]. In this technology, the microwave fields located near to the write pole are emitted by a spin torque oscillator (STO). This microwave field causes the magnetization precession at a high speed, thus the higher K_u perpendicular anisotropy recording medium is able to be written at lower magnetic fields. Figure 1.3 shows the areal density growth through technological and materials innovations, where CAGR means a compounded annual growth rate [36].

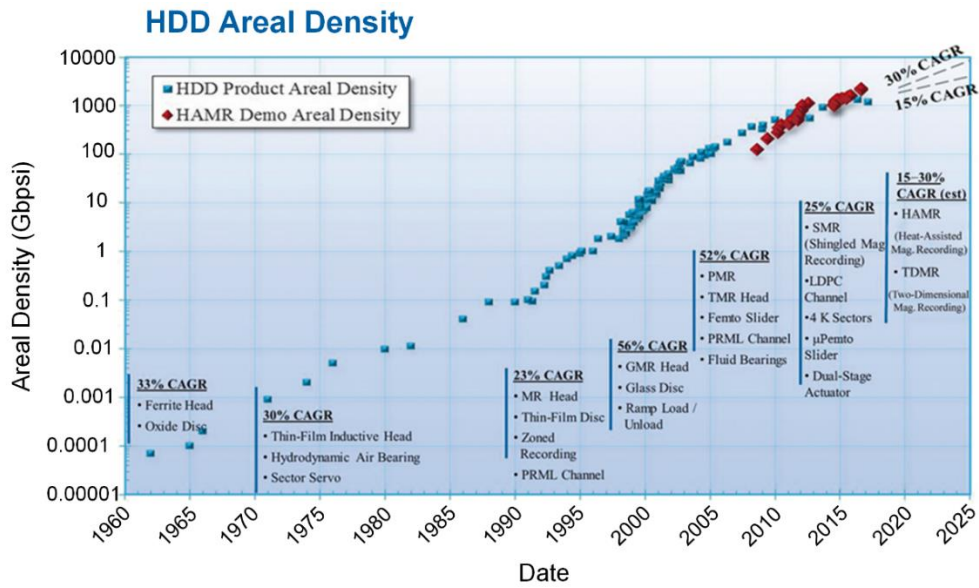


Figure 1.3. Areal density growth has occurred through a steady progression of technological and materials innovations. CAGR means a compounded annual growth rate [36].

Despite the emergence of technologies such as HAMR and MAMR, increasing areal density above 25 Tbps is still a big challenge for HDD technology. Moreover, the areal density growth has been slow and with such a growth rate, magnetic recording technology cannot compete with solid state drives (SSD) [37]. Also, the writing and reading heads rely on the relative mechanical motion between the heads and disks, the slow access time and the undesired high energy consumption are some drawbacks of HDDs. It is difficult to enable HDDs to fulfill the demand of the booming consumer information industry to pursue the small, portable and energy efficient products. At the same time, solid state memory devices are investigated for high-speed and energy saving applications.

1.1.2 Domain wall racetrack memory

Considering these deficiencies of HDDs, Stuart Parkin from IBM proposed the domain wall racetrack memory (DWRM) as an alternative for storing vast information [38]. It opened a new prospect of potential non-volatile storage-class memory, which has all the advantages such as higher speed, almost unlimited endurance, and random access of information [39]. When proposed, it attracted widespread interest both in physics and application.

In DWRM, magnetic domains store the information in a three-dimensional magnetic strip. Hundreds of millions of nanowires arranged vertically, like trees in a forest, have the potential for storing vast amounts of data as shown in Figure 1.4(a). The coherent motion of those domains-or equivalently, the domain walls (DWs) is essential for reading and writing. For example, domains are moved by a pulse current and the information can be written by passing them over a write head, where a fringe magnetic field (H) generated by write current can change the magnetization direction. High sensitive tunnel magnetoresistance (TMR) senses devices, called read element can be integrated to read information as shown in Figure 1.4(b).

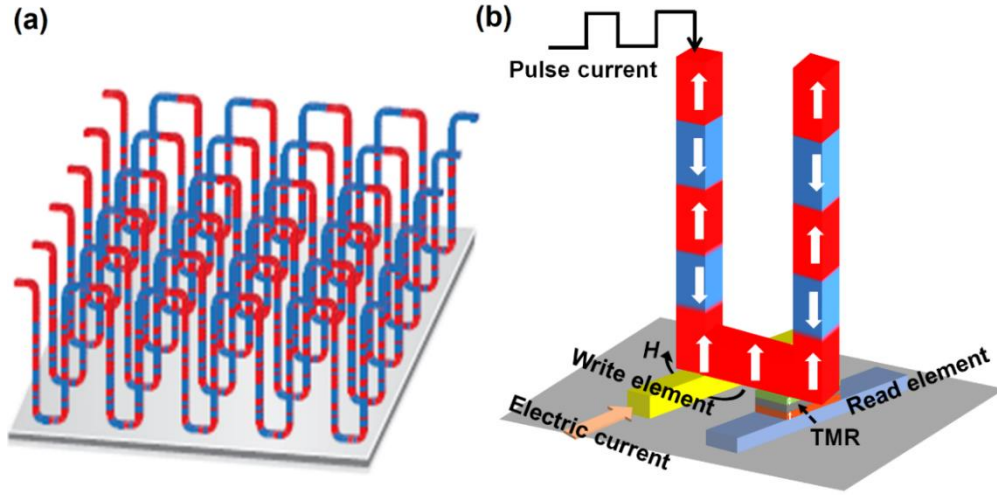


Figure 1.4. (a) Three-dimensional domain wall racetrack memory (DWRM) consisting of arrays of vertical racetracks on silicon. (b) A diagram of the nanowire consisting of the write element and read element [38].

1.2 Research objectives

Domain wall racetrack memory has been proposed and investigated as the non-volatile storage class memory devices. In this sense, it should be able to follow the key requirements of a memory device, such as writing, storing, addressing and reading. When we mention addressing, it means that the domains or equivalently DWs must be moved along the racetrack for reading/writing bits. Many groups have studied the behavior of DWs motion and reported the stochastic and random propagation, which spread over a large distance in the range of micrometers. The lack of precise control over DW displacement has become one of the main limitations of DWRM application [39-42]. It is necessary to form regions where domain walls can be pinned effectively and uniformly. Therefore, researchers have worked on this topic for a decade. There have been various methods proposed for creating pinning sites in the nanowires to control the DW motion. Those approaches are based on altering the local physical or magnetic properties of the nanowire.

Modification of the nanowire shape to pin domain wall, such as the creation of notches in ferromagnetic nanowires, have been investigated in the past [43-46]. Recently, staggered nanowire has been proposed as another method to pin the domain wall, which is based on fabricating the nanowire with a small off-set in either one direction or both [47]. However, these methods used for domain walls pinning are suffering from a complex fabrication process and non-uniform pinning strength. Polenciuc *et al.* have pinned domains by forming crossbars of antiferromagnetic material [48]. However, this is a growing field, and various requirements have to be fulfilled by the proposed pinning approaches.

In this work, I have investigated the creation of synthetic magnetic textures using various approaches to pin domain walls. The synthetic magnetic texture is achieved by localized diffusion of non-magnetic metal into ferromagnetic nanowires by annealing-induced mixing or by ion implantation to modify the properties of the ferromagnetic nanowire. These magnetically modified regions are supposed to form the pinning site to control DW motion effectively. I have also proposed forming a tilted magnetization by the exchange coupling between two layers with in-plane and out-of-plane magnetization to control DW propagation in the DW devices with perpendicular magnetic anisotropy.

In this thesis, the investigation is carried out by a three-fold approach: 1. Use of micromagnetic simulations to predict and understand the pinning. 2. Tuning and understanding the magnetic, magneto-dynamic and structural properties in thin film level. 3. Fabrication of nanowires with pinning sites and studying the DW propagation and pinning effect by magnetoresistance measurements and Kerr microscopic observations.

Firstly, Permalloy ($\text{Ni}_{80}\text{Fe}_{20}$), which shows excellent soft magnetic properties, was chosen to investigate the effect of the first approach, viz., local metal diffusion. Diffusion of various metals including Ta, Cr, Cu, and Ru, was studied by annealing the thin films at different temperatures for one hour. The static and dynamic properties of NiFe were characterized at thin film level to understand the effect of metal diffusion. The $\text{Ni}_{80}\text{Fe}_{20}$ devices were fabricated without and with cross pinning sites to demonstrate that this method could be useful in domain wall memory devices. In addition, micromagnetic simulations were performed to investigate the pinning effect in the NiFe nanowires with different properties modification regions.

Secondly, pinning investigations were carried out in Co/Pd multilayers system with a high-anisotropy, which is suitable for high-density device applications. The prepared Co/Pd multilayers displayed a high perpendicular magnetic anisotropy (PMA) and the size of the domain wall could be tuned by changing the repeat number of bilayers. Attempts to pin domain walls using localized elemental diffusion of Cr was not found suitable for this system. However, local modification of magnetic properties using the implantation of Boron ions (B^+) to change the magnetic anisotropy was proved to be effective. Domain wall pinning in devices has been demonstrated at the B^+ implanted positions. To understand further, micromagnetic simulations were carried out to study the pinning effect in Co/Pd multilayers nanowire with a pinning site formed by reducing the magnetic anisotropy.

As another method, I investigated localized tilting of magnetization, to pin the domain walls. In order to form pinning sites with tilted magnetization, I deposited Co/Ni multilayers as a perpendicular magnetic anisotropy (PMA) layer,

whose anisotropy is not too strong. A thick Co layer with in-plane magnetic anisotropy (IMA) was grown on the top of Co/Ni layer to tilt the magnetization direction of Co/Ni multilayer diverged from z-direction. A Pt layer with varying thickness as the spacer layer was used to modify the exchange strength between the PMA and IMA layers. Micromagnetic simulations were carried out to understand the pinning effect with different tilting angles.

1.3 Organization of thesis

The thesis is organized as follows;

Chapter 1 gives a brief introduction of the challenges of existing magnetic storage technologies, such as HDDs, which provide the necessity to come up with the new potential alternative technology. This chapter also provides the motivation and objective of this thesis.

Chapter 2 describes the details of domain wall properties in ferromagnetic materials. It covers the domain wall dynamics, followed by a detailed overview of the current state-of-the-art in domain wall memory research. The motion of the domain wall and achieving high-speed motion, by the magnetic field and electric current are discussed.

Chapter 3 introduces the experimental techniques and details of the simulation methods used in this thesis. This chapter includes a brief introduction of thin film deposition techniques, device fabrication procedures and the different characterization techniques used for thin film and device investigations.

Chapter 4 presents and discusses the results of tuning the magnetic properties of NiFe by using different elements diffusion. Firstly, the investigations on thin film level are described. Later, the simulation work related

to DW pinning with magnetic properties modification is described. At last, the experimental results to show the DW pinning effect by Cr diffusion into NiFe devices are described.

Chapter 5 discusses the magnetic texture change by ion implantation using B^+ . The magnetic properties of $(Co/Pd)_N$ samples with different repeat numbers N of two bilayers were investigated before and after B^+ implantation. The simulation results are discussed later. At the end of this chapter, the DW dynamics, as driven by the magnetic field in $(Co/Pd)_{10}$ wires, are presented.

Chapter 6 describes the investigations of the pinning effect in Co/Ni multilayers by tilting its magnetization direction. Firstly, the micromagnetic simulation which predicts the possibility of the pinning effect in the tilted magnetization position is described. Then the dependence of depinning magnetic field and current for different tilt angles are discussed. Next, the experimental study where the DW pinning in Co/Ni PMA wires with Co IMA crossbars is discussed. The tilt angle in devices was estimated using Mumax3 simulation by varying exchange coupling strength.

Chapter 7 gives a brief conclusion about domain wall pinning effect investigated in this thesis by using synthetic magnetic texture structures. Suggestions for future work are provided at the end.

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Chapter 2 Literature Review

2.1 Introduction

As this thesis focuses on domain wall memory, it is essential to review and understand domain wall properties and dynamics for developing domain wall devices with excellent performance. In this chapter, we discuss the domain wall formation and properties in two types of magnetization structures, which exhibit in-plane and out-of-plane magnetic anisotropy respectively. After the fundamental knowledge is presented, we discuss domain wall dynamics and motion driven by the magnetic field and electrical current, as well as the domain wall pinning.

2.2 Domain formation and domain walls

In a magnetic material, magnetic domains are the regions in which the magnetization exhibits a uniform ordering and direction. However, a large region with a uniform magnetization throughout the ferromagnetic materials corresponds to a higher magnetostatic energy. To reduce the energy, the magnetization of the whole material is arranged randomly with opposite directions in neighboring regions, forming multi-domains [1]. Four types of energies contribute to the properties of a ferromagnetic material: Zeeman energy, exchange energy, anisotropy energy, and demagnetizing energy. The domain formation occurs mainly due to the competition between the demagnetizing energy (E_d), and the exchange energy (E_{ex}) of the system. The E_{ex} tries to keep the neighboring magnetic moments lying parallel to each other. At the same time, demagnetizing energy tries to orient large clusters of magnetic moments in random directions. Thus, the balance between these two energies results in the formation of the domains. It is crucial to note that the different orientations of the

neighboring domains increase the exchange energy. In order to minimize E_{ex} , the magnetic moments gradually change their directions in aligning from one domain to the other. However, at the same time, the increase in the relative orientation of the magnetic moments from the anisotropy direction increases the anisotropy energy of the system. Thus, the minimization of the anisotropy energy requires the minimization of spins away from the anisotropy direction. Thus, the competition between exchange energy and anisotropy energy results in the formation of a transition region between the two domains. This region is known as a domain wall (DW).

In a one-dimensional model, DW energy can be given by as an integral as follows:

$$E_{dw} = \int_{-w/2}^{+w/2} [E_{anis}(x) + E_{ex}(x)] dx \quad (2.1)$$

where E_{anis} and E_{ex} represent the anisotropy and exchange energies respectively. The boundary conditions represent the start and end positions of the domain wall with a width of w . Using a variational approach to find the minimum energy, this gives the value of w as:

$$w = \pi \sqrt{\frac{2A_{ex}}{K_u}} \quad (2.2)$$

Solving the integral of Equation (2.1) would result in $E_{dw} = E_{anis}w/2 + E_{ex}\pi^2/w$. By substituting Equation (2.2) to the solution of Equation (2.1), the DW energy is given to be:

$$E_{dw} = \pi \sqrt{2E_{ex}E_{anis}} \quad (2.3)$$

This simple model from Equation (2.1) shows that E_{ex} would increase the DW width w , whereas E_{anis} increases the energy required for the DW, which

decreases the DW width w . The balance of the energy terms introduced thus far in a ferromagnetic system can be expressed simply in an illustration as shown in Figure 2.1, where E_{demag} and E_{dw} represent the energy terms for demagnetization and domain wall respectively. Although DWs reduce the E_{demag} , the DWs themselves require energy, which is related to the intrinsic E_{ex} and E_{anis} of the material as shown in Equation (2.3).

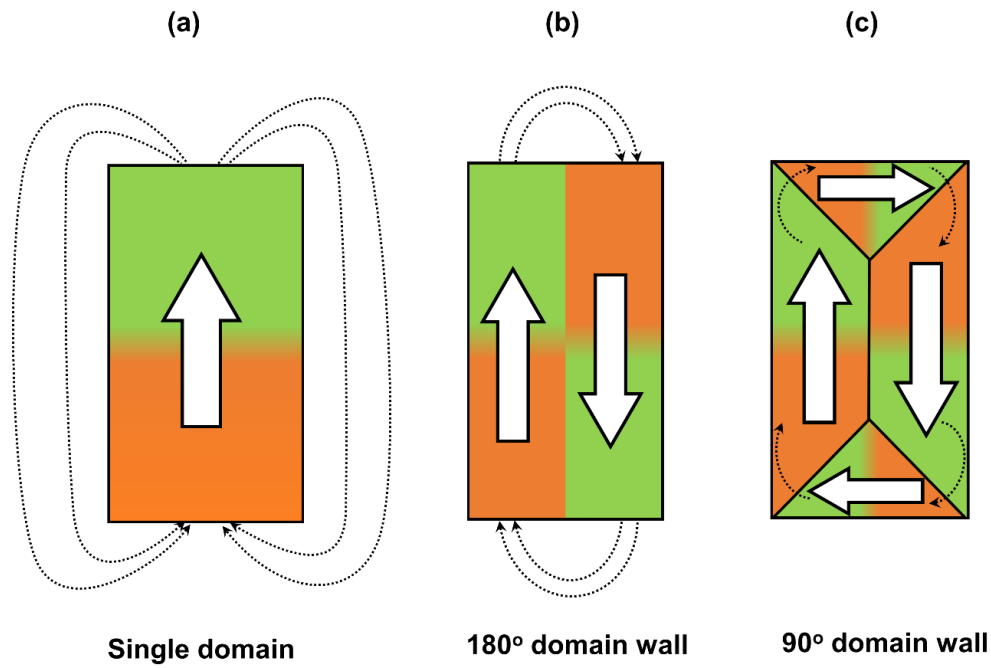


Figure 2.1. Schematic diagram illustrating the single domain and multidomain structures. (a) Single domain results in enormous demagnetization energy. (b) A two-domain structure and forming a domain wall lower the demagnetization field. (c) The multidomain structure reduces the demagnetization energy as most of the flux lines are closed, while resulting in a heightened domain wall energy.

Domain wall dimensions and its energy exhibit different properties in different materials with various anisotropy energy constant K_u . In in-plane (IP) magnetic anisotropy materials, such as Permalloy ($\text{Ni}_{80}\text{Fe}_{20}$), Co, Fe, and CoFe alloy, have low K_u , resulting in a larger DW width of ~ 100 nm. They usually form head-to-head (H-H) and tail-to-tail (T-T) DWs, as illustrated in Figure 2.2,

which mainly depends on the thickness of the thin film. Thiaville and Nakatani theoretically predicted two types of spin structures: transverse wall and vortex wall in IP magnetized materials [2]. Figure 2.3(a) shows a schematic of the transverse DW. In this case, the magnetic spin rotates its direction in the plane of the film. In order to further reduce the energy, the structure of the domain wall becomes symmetric respect to the vertical axis, which yields a V-shaped domain wall. Whereas the magnetic spins in the vortex wall core yield a clockwise or anticlockwise shape around the vortex core.

Figure 2.3(b) shows a schematic of the anti-clockwise vortex DW, in which case, the magnetization in the vortex core points in either out-of-plane direction. McMichael and Donahue performed some simulations to verify that the geometry and the magnetic properties of material take the dominant roles in domain wall formation [3]. Figure 2.4 shows the DW configuration as a function of geometry in soft magnetic materials. The formation of spin texture is more sensitive to the width of the device and the thickness of the film. In Permalloy devices, with a narrower width and larger thickness, transverse spin textures are favored. On the contrary, vortex spin textures are favored in the devices with a larger width, while the thickness does not play an important role in forming this spin texture. In Co devices, thin films with the thickness below 20 nm, are favored to form transverse DWs. In addition, the films with larger thickness t ($t > 20$ nm) and width w ($w > 250$ nm) are favored to form vortex DWs. In soft magnetic materials, DWs configuration is much complex, which is determined by the minimization of the total energy [3, 4]. In a particular geometry, DW configuration can be defined by width (w) $\times$ thickness (t) $\approx$ constant, which depends on the material properties [5].

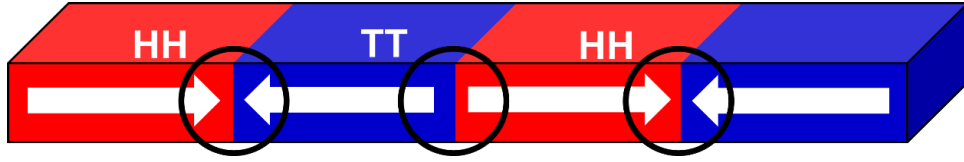


Figure 2.2. Schematic diagram illustrating the head to head (H-H) and tail to tail (T-T) DWs.

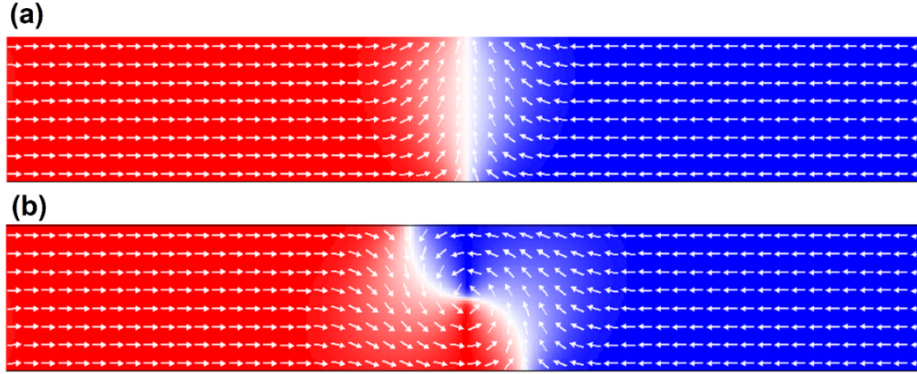


Figure 2.3. The schematic of the (a) transverse DW and (b) anti-clockwise vortex DW [5].

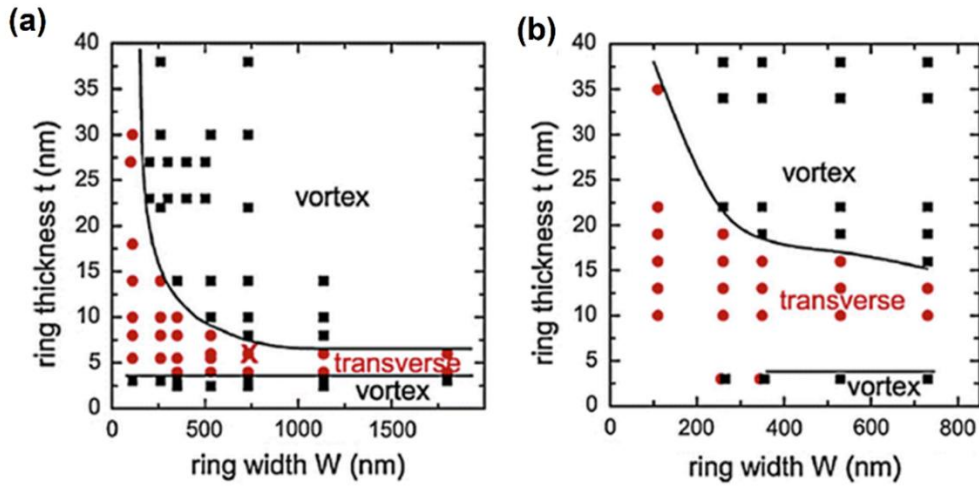


Figure 2.4. Experimental depiction of the vortex and transverse domain wall in (a) $\text{Ni}_{80}\text{Fe}_{20}$ and (b) Co rings [5].

Out-of-plane (OOP) magnetized materials with strong PMA, have much larger anisotropy energy, K_u . In addition, the domain wall width is much smaller, down to ~ 10 nm. The spin structure of domain walls is much robust. A higher spin-orbit coupling and higher domain wall velocities are observed in these

materials compared to IP soft magnetic materials. As discussed, the competition between E_{ex} and E_{anis} results in the domain walls of finite thickness and structure. Depending on the configuration of the domain walls, the domain wall can be generally classified as of two types: (1) Bloch wall and (2) Néel wall.

In the Bloch wall, the magnetization in the domain wall gradually rotates about an axis which is normal to the plane of the domain wall. First explained by Bloch in 1932, this is a typical domain wall structure in thick films [6, 7]. However, the Néel wall is favored when the thickness of the sample becomes comparable to the dimensions of the domain wall. In this case, the free poles near the domain wall become significant which can lead to the change in the domain wall structure and magnetization rotates within the plane of the sample. Figure 2.5 is the schematic illustration of the Bloch wall and Néel wall.

For perpendicularly magnetized materials, two types of candidates have been mainly researched: (1) ultrathin magnetic multilayers, where PMA originates from the interface. Examples of such materials are Co/Pt, Co/Pd and Co/Ni multilayers and (2) thicker (usually 5~60 nm) magnetic films where PMA comes from magnetocrystalline anisotropy, such as ordered FePt, and CoPt alloys. Due to a large anisotropy energy, the domain wall width is much smaller (1~10 nm) in these materials.

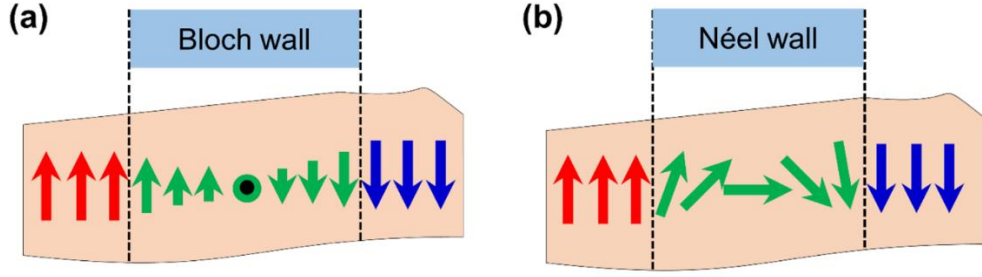


Figure 2.5. Demonstration of (a) Bloch wall and (b) Néel wall; intermediate spin structure between two domains in ferromagnetic materials. The arrows represent the magnetization.

In the Bloch wall, the magnetic moments rotate out of the plane whereas in the Néel wall the magnetization changes the direction by an in-plane rotation. The Bloch wall width can be expressed as [8]:

$$w = \sqrt{A / K_u} \quad (2.4)$$

where A is the exchange stiffness and K_u is the uniaxial anisotropy constant.

The Néel wall width is expressed as:

$$w = \sqrt{A / K_d} \quad (2.5)$$

where A is the exchange stiffness and $K_d = M_s^2 / 2\mu_0$ for thin film. M_s is the saturation magnetization and μ_0 is the vacuum permeability [9].

2.3 Domain wall dynamics

Domain wall dynamics is one of the fascinating phenomena in the fundamental aspects of magnetism as well as applications, which has drawn significant attention of researchers. In order to develop spintronics devices based on domain wall motion, an understanding of the DW dynamics is critical. The efficient forces to drive DW motion is by utilizing the magnetic field (H) and electric current (j_e), as shown in Figure 2.6. The magnetic field helps to switch magnetization and push DW to move in two opposite directions. While an

electric current drives the DW motion along with the electron flow, by spin momentum transfer of electrons to magnetic moments.

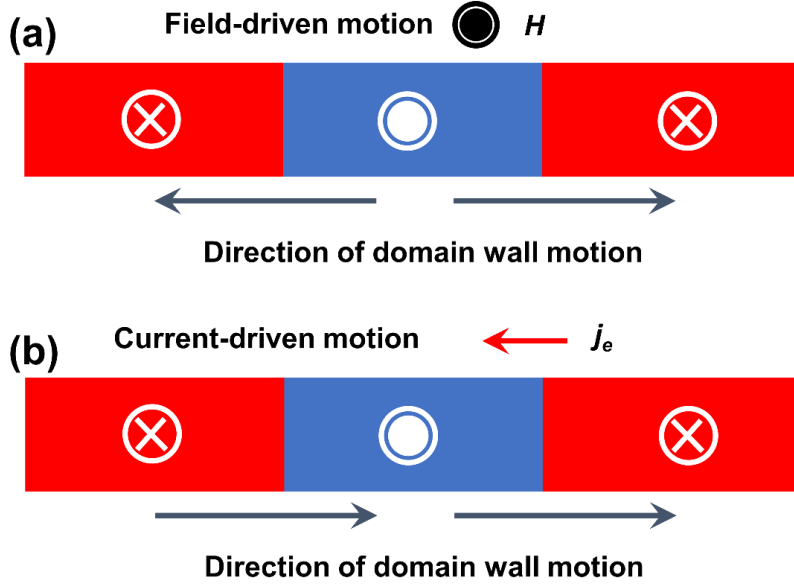


Figure 2.6. DW motion is driven by (a) a magnetic field and (b) an electric current.

When an external magnetic field exists, the magnetization $\mathbf{M}$ will experience a torque and precess around the magnetic field, which phenomenon can be expressed as [10]:

$$\frac{\partial \mathbf{M}}{\partial t} = -\gamma \mathbf{M} \times \mathbf{H}_{\text{eff}} \quad (2.6)$$

The magnetic field is usually written as an effective magnetic field $\mathbf{H}_{\text{eff}}$. The magnetic moment $\mathbf{M}$ precesses around $\mathbf{H}_{\text{eff}}$, and this phenomenon is known as Larmor precession. γ represents the absolute value of the gyromagnetic ratio. The effective field $\mathbf{H}_{\text{eff}}$ combines an external magnetic field, the demagnetizing field, and additional anisotropy terms. $\mathbf{H}_{\text{eff}}$ is not constant but depends on magnetization at the moment.

This basic model was firstly proposed by Landau and Lifshitz [10]. Since the magnetization cannot precess indefinitely, an additional phenomenological dissipation term was added which is dependent on the time-derivative of the

magnetization, called damping term. Gilbert showed that the magnitude of the magnetization is conserved through the following expression, named as Landau-Lifshitz-Gilbert (LLG) equation [10, 11]:

$$\frac{\partial \mathbf{M}}{\partial t} = -\gamma \mathbf{M} \times \mathbf{H}_{\text{eff}} + \frac{\alpha}{M_s} (\mathbf{M} \times \frac{\partial \mathbf{M}}{\partial t}) \quad (2.7)$$

where α is the phenomenological Gilbert damping parameter. The precession frequency around $\mathbf{H}_{\text{eff}}$ of typically several GHz and the damping with a typical time constant of several nanoseconds. Those two terms describe the magnetization from relaxation toward equilibrium.

When an electric current is sent through the magnetic material, an additional torque term will be added. This theory was proposed by Slonczewski and Berger in 1996 [12, 13]. It describes the transfer of the angular momentum of an electron to magnetization M , known as spin-transfer torque (STT). Considering the additional torque, the magnetization dynamics can be expressed by modified the LLG equation to Landau-Lifshitz-Gilbert–Slonczewski (LLGS) equation as:

$$\frac{\partial \mathbf{M}}{\partial t} = -\gamma \mathbf{M} \times \mathbf{H}_{\text{eff}} + \frac{\alpha}{M_s} \mathbf{M} \times \frac{\partial \mathbf{M}}{\partial t} - \frac{b_j}{M_s^2} \mathbf{M} \times (\mathbf{M} \times \frac{\partial \mathbf{M}}{\partial x}) - \frac{c_j}{M_s} \mathbf{M} \times \frac{\partial \mathbf{M}}{\partial x} \quad (2.8)$$

When DW motion is driven by current, the angular momentum of electrons exerts a torque to the magnetization, leading to DW motion. The direction of DW motion is the same as the direction of electron flow. There are two terms behind the angular momentum transfer effect. The one term is called adiabatic STT term, written as $\tau_{\text{STT1}} = \frac{b_j}{M_s^2} \mathbf{M} \times (\mathbf{M} \times \frac{\partial \mathbf{M}}{\partial x})$. This torque plays a significant role in the initial motion of DW. However, it does not determine the terminal velocity of the domain wall motion. This effect was firstly proposed by Bazaliy *et al.* in half-

metal material when they considered the ballistic motion of conducting electrons in 1998 [14]. Later, Tatara and Kohno also explained the similar expression of this torque [15]. The other term is called non-adiabatic STT term defined as $\tau_{STT2} = \frac{c_J}{M_s} \mathbf{M} \times \frac{\partial \mathbf{M}}{\partial x}$, which is entirely different. This torque originated from the mistracking of electron spins and the local magnetic moments [16]. It behaves as a nonuniform magnetic field and can sustain a steady state wall motion. The theory about non-adiabatic torque was introduced by Zhang and Li in 2004 [16]. Here, c_J and b_J have the unit of velocity and c_J is much smaller than b_J ($c_J/b_J = \xi \approx 10^{-2}$). Considering a 1D system with a spin current flowing along the x -direction ($\mathbf{J}_s = J_s \mathbf{e}_x$), $b_J = PJ_s \mu_B / eM_s (1 + \xi^2)$ and $c_J = PJ_s \mu_B \xi / eM_s (1 + \xi^2)$. Although the magnitude of the non-adiabatic torque c_J is about 2 orders of magnitude smaller than adiabatic torque b_J in most ferromagnetic materials, the terminal velocity is inversely proportional to the damping parameter which makes the velocity comparable to c_J [17-19].

2.4 Domain wall velocity

The domain wall velocity was utilized to describe the domain wall motion, which is related to the operational speed of spintronics devices based on domain wall motion. The domain wall velocity is one of the critical parameters to characterize, thus to further improve the performance of the device application. Domain wall velocity shows different properties in IP magnetized and OOP magnetized systems, which will be particularly discussed below.

2.4.1 Domain wall motion driven by magnetic field

Domain wall motion driven by a magnetic field has been observed in many experimental works. In an earlier work, T. Ono *et al.* demonstrated the domain wall propagation in NiFe strip upon the application of a magnetic field, and the domain wall position was sensed using the giant magnetoresistance (GMR) measurement in a NiFe/Cu/NiFe trilayer structure [20]. They fabricated the necks with a width of ~ 350 nm, which is smaller than nanowire width (~ 500 nm), to pin the domain wall at some specific positions. Upon the application of an increasing external magnetic field, the domain walls moved and displayed different GMR values, indicating the DW motion in a different position, as shown in Figure 2.7 [21].

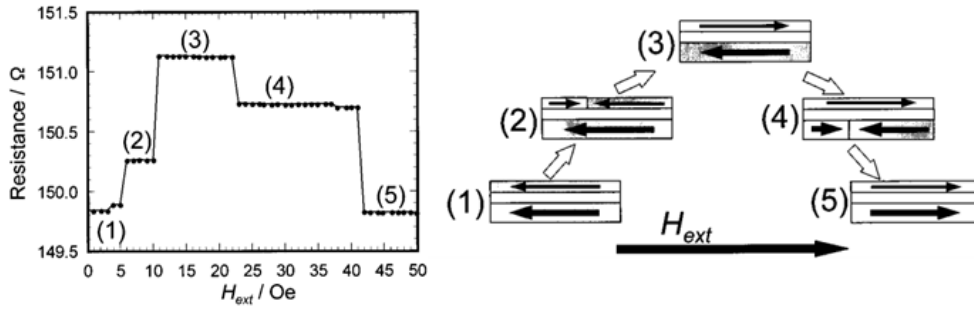


Figure 2.7. (Left) Resistance changes as a function of the externally applied magnetic field. (Right) The schematic diagram of magnetic domain structures deduced from the resistance measurement and the arrow shows the direction of the external field [21].

Atkinson *et al.* measured DW velocity in NiFe nanowires of 200 nm width and showed a non-linear behavior of the DW velocity upon the external magnetic field [22]. At the very high fields, applied in a direction parallel to the magnetization, the DW velocity was found to reach a maximum [22]. In the intermediate values of the applied field, the domain wall velocity showed regions of different slopes. When the magnetic field was applied in the orthogonal direction, a plateau in the domain wall velocity–field graph was observed.

Followed by more theoretical and experimental works, researchers noticed that the DW motion driven by the magnetic field is not increasing consistently with a magnetic field [23]. There exists a Walker breakdown field $H_w = 2\pi\alpha M_s$ (where α is the damping constant, M_s is the saturation magnetization) [24-27]. Below the Walker field H_w , where a steady-state motion is known to exist, it is shown that the domain wall moves smoothly with a velocity, which is determined by $\gamma_0 H_0 w / \alpha$, where γ_0 is gyromagnetic ratio, and w is the domain wall width. Above H_w , the magnetization starts to precess around the field and a periodic component appears in the forward motion of the wall, leading a significant reduction in the DW velocity, as illustrated in Figure 2.8. For much higher fields, $H > H_w$, the DW velocity is proportional to the external magnetic field as $\alpha\gamma_0 H_0 w / (\alpha^2 + 1)$, but the slope is still much smaller compared to the slope for the field below the Walker field H_w because of $\alpha \ll 1$.

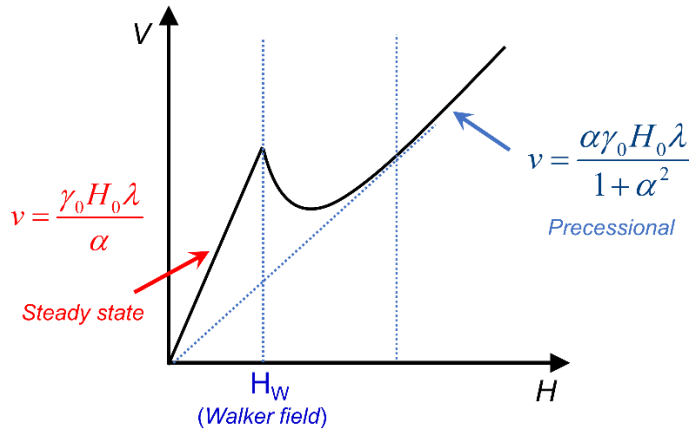


Figure 2.8. Illustration of the magnetic field induced DW motion. The velocity shows Walker breakdown. At a smaller field, the DW velocity increases linearly up to the Walker field. Above this field, transformations about DW configuration set in, which reduce the velocity. For much higher fields, the DW velocity increases again [23].

When we discuss the DW motion, we also have to consider the pinning effect, as they play a key role in the domain wall motion. The existence of defects

in a material can cause DW pinning. Defects can be of any form, like a shape defect, inhomogeneous surface, impurities, and surface strain. The domain is pinned by those defects and cannot be moved unless a larger field is applied to depin the domain wall. The pinning and depinning of the domain wall is a crucial process, corresponding to the reversible and irreversible wall propagation. Considering the pinning effect, domain wall velocity as a dependence of the magnetic field shows three regimes, including creep regime, depinning regime, and flow regime, which is shown in Figure 2.9.

For a low field regime, the domain wall velocity is very slow, near to zero, called creep regime. At a critical value of the applied field, called the depinning field, the domain wall starts precessing around the easy axis and the domain wall velocity is still very low. This regime was named a depinning regime. Domain wall velocity driven by the magnetic field can be expressed as [28, 29]:

$$v = v_0 \exp\left[-\frac{U_c}{k_B T} \left(\frac{H_{dep}}{H}\right)^{1/4}\right] \quad (H < H_{dep}) \quad (2.9)$$

where v_0 is the velocity scaling prefactor related to the lateral length of the small section of the domain wall, U_c is the characteristic energy associated with the disorder-induced pinning potential, k_B is the Boltzmann constant, T is the temperature and H_{dep} is the depinning field below which the creep regime occurs. The exponent $1/4$ fits the data well, which has been determined theoretically and experimentally [28, 29]. When plotting velocity v^4 with an applied field $1/H$, the slope of the linear curve indicates the depinning field of the domain wall motion in the specific system.

By applying a strong field above the depinning field, the Zeeman energy overcomes the pinning potential energy barrier. All the moments within the

domain rotate from their original crystallographic easy axis to the field direction, causing the domain boundaries to expand or domain wall motion. This regime called flow regime. We note that the velocities in the depinning region and flow region are different, due to the presence of the Walker field.

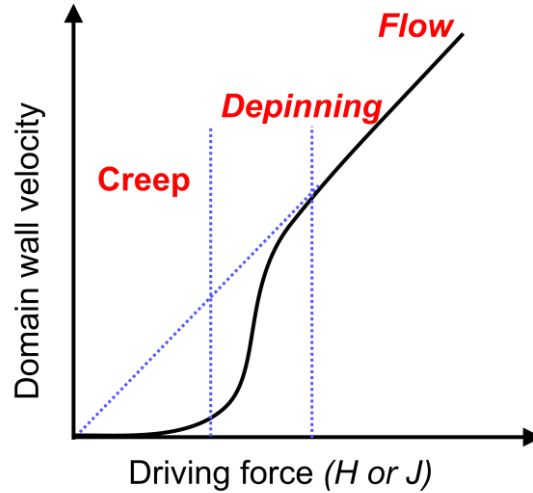


Figure 2.9. Domain wall velocity at three different regimes driven by a field [30].

Berkov *et al.* simulated the field-driven domain wall motion in a 200 nm-wide and 5 nm-thick sample. When the applied field was small, $H = 10$ Oe, a maximum velocity of 650 m/s was reached after a short acceleration, as shown in Figures 2.10 (a-d). Above the Walker breakdown, $H = 20$ Oe the out-of-plane magnetization increased above the critical angle, and the DW velocity decreased drastically. The DW moved backward, causing a reduction in velocity, as shown in Figures 2.10 (j-k). Even for a higher field, the domain wall motion became positive (along positive x-axis direction) and with a linear dependence on the applied magnetic field [31].

The actual behavior of DW dynamics above the Walker breakdown field depends on the cross-section of the nanowire [26, 32]. Beach *et al.* performed the experimental observations on the Walker breakdown process, which demonstrated an obvious reduction in DW velocity in a single NiFe strip at a high

field using the time-resolved MOKE technique [33]. Moreover, Hayashi *et al.* [34, 35] measured the oscillations of the anisotropy magnetoresistance (AMR) signal during the DW motion in NiFe layers, where the amplitude of oscillation indicates the change of DW configuration, such as domain wall type and chirality, for the magnetic field above H_W . Later, a direct observation of the Walker process was investigated by Glathe *et al.* using GMR measurement [36]. These results provide insight understandings about the DW dynamics and the Walker breakdown field of DW motion, which are essential for practical applications.

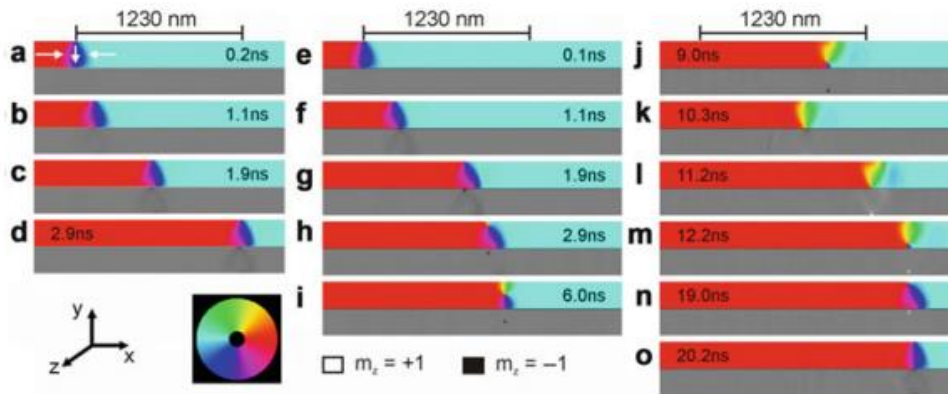


Figure 2.10. Simulation of the domain wall motion with simulating time in NiFe nanowire driven by a magnetic field. (a-d) A domain wall continuously moves upon the small field below the Walker field. (i) Above the Walker breakdown, the out-of-plane magnetization increases. (j-o) A domain wall process and moves backward, then moves forward [36].

2.4.2 Domain wall motion driven by current

Another way to displace the domain walls is through spin torque transfer from an applied current. As shown in Equation (2.8), the last two terms in the LLG equation accounts for the effects of an injected spin-polarized currents. From the LLG equation, the effective spin drift velocity (u) is given by $u = J (pg\mu_B/2eM_s)$, where p , e , M_s , J , g and μ_B stand for spin polarization, electronic charge, saturation magnetization, current density, gyromagnetic factor

and the Bohr magneton of electron spin respectively [37]. The fourth term in the LLG equation gives the information about non-adiabatic transport and spin relaxation where the non-adiabaticity parameter (abbreviated as c_J) is defined as the ratio of the exchange length to the spin flip length [38, 39].

It is important to note that the domain wall dynamics upon the application of the current strongly depends on c_J . For $c_J = 0$ i.e. adiabatic case, there is no steady motion when the current is below this critical current value (J_c). In addition, for an applied current density above the critical current density, domain wall moves fast, with a velocity $v \propto \sqrt{J^2 - J_c^2}$ [40], as shown in Figure 2.11. While for $c_J \neq 0$, the critical current density for domain wall motion reduces to zero and domain wall velocity increases linearly for smaller applied currents. For $c_J = \alpha$, the velocity becomes linear with applied current density. For the cases of $c_J \neq 0$ and $c_J \neq \alpha$, the velocity scales as $c_J u / \alpha$ and increases linearly up to a peak with current density increasing, which was determined as the Walker breakdown current density. Above this value, DW transformations were set in analogous to the field-induced wall motion. Again, for high current densities, the velocity is going to approach u .

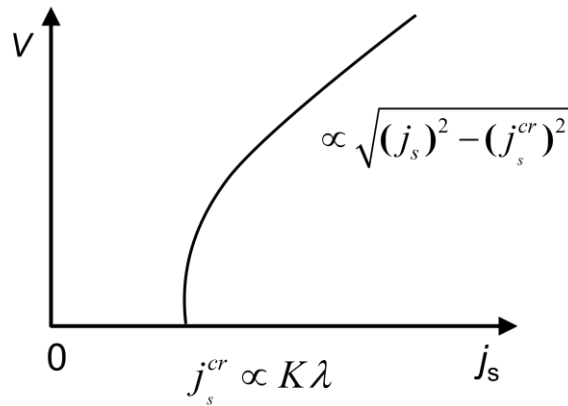


Figure 2.11. Average domain wall velocity as a function of current density, j_s , for the weak pinning case ($V_0 \leq K\lambda$) [40].

Many research groups have experimentally reported current-induced domain wall dynamics [41-50]. In 2006, Beach *et al.* reported the DW velocity enhancements to 125 m/s from 90 m/s driven by the magnetic field under a low current density of 6×10^{11} A/m² in a Permalloy nanowire [49]. Later, Hayashi *et al.* reported the domain wall velocities more than 100 m/s in Permalloy nanowires for a higher current density of 1.5×10^{12} A/m² without the magnetic field [50]. Moreover, some other researches also demonstrated the fast movement of domain walls by a pure current [45, 46, 51].

For higher current densities, a periodic transformation between a vortex and a transverse DW happens. In the case of PMA materials, the transformation happens between a Bloch and a Néel DW. Kläui *et al.* reported the transformation between head-to-head vortex and a transverse DW, when a current density of 2.2×10^{12} A/m² (larger than critical current density) was sent through Permalloy wire [52]. Hayashi *et al.* also reported the real-time observation of the domain wall transformation in Permalloy nanowires with a current density in the range from 0.8×10^{12} to 1.6×10^{12} A/m² [53]. Apart from the IP magnetized materials [33, 54, 55], the perpendicularly magnetized materials, such as Pt/Co/Oxide, Co/Ni multilayer, Co/Pt multilayer, have drawn more and more attention since they were discovered. Similarly, a periodic transformation from Bloch to Néel of domain walls has also been reported for higher current densities in a Pt/Co/AlO_x system [56, 57]. Some behaviors for the DW dynamics of the OOP system are different from that of the IP system. For example, in a narrow (width < 100 nm) and thin (thickness < 10 nm) nanowire, the critical current density j_c becomes independent of pinning strength as well as non-adiabaticity parameter c_J , thus the critical current could reach to a small value [58].

Analogous to the field-induced domain wall motion, current-induced domain wall motion exhibits three regimes: (1) the domain wall motion is slow, which is governed by wall pinning and thermal activation; (2) the domain wall velocity is slightly larger, and the domain wall motion is mainly governed by thermally activated depinning; (3) the domain wall velocity becomes linear with current density, *i.e.* the domain wall velocity becomes much higher. In the experimental studies, Moore and Miron *et al.* reported the linear regime for the current-induced domain wall motion in a Pt/Co/AlO_x nanowire and the maximum DW velocity was ~ 400 m/s for a current density of 3.5×10^{12} A/m² [59-61]. It was also concluded that the non-adiabatic torque is the driving force for the DW motion in this regime [62].

In most of the cases, a domain wall moves in the direction of the electron flow, but a few researchers claimed that DW also moves in the direction of the current flow, which is opposite to the direction of the electron flow. These observations can be explained by the negative current spin polarization, negative non-adiabaticity factor, additional spin-orbit torques [59, 63, 64]. Also, it is important to note that the current acts as a Slonczewski torque and a field-like torque depending on the DW structure. Moreover, one of these results in a high non-adiabaticity factor (which is much higher compared to in-plane magnetized materials), and thus result in a faster domain wall motion. In heavy metal/ferromagnetic/insulator structures, the direction of DW motion was found along the current flowing direction. The force was named as spin-orbit torque (SOT) due to the stronger spin-orbit coupling (SOC) at the interface of heavy metal /ferromagnetic.

Spin-orbit torque (SOT) has been considered as an effective force to drive the DW motion in the ferromagnetic nanowire, as well as in the synthetic antiferromagnetic system. Ultra-fast DW velocity has been reported in synthetic antiferromagnetic structure by Yang in 2015 [65]. This structure comprises two ferromagnetic layers Co/Ni multilayers, which is separated by a thin Ru layer. The net magnetization is zero due to the antiferromagnetic coupling between these two ferromagnetic layers. It was reported that the DW velocity reaches 750 m/s in the synthetic antiferromagnetic system. When current goes through this stack, the exchange coupling torque is produced due to the antiparallel magnetization, which plays an important role to drive the DW motion. Later, Meng *et al.* also reported the threshold current density for the DW displacement decreases with the exchange coupling increasing in a synthetic antiferromagnetic CoFe/Pd multilayers structure, as well as the DW velocity increases accordingly to larger exchange coupling [66]. So far, the domain wall velocity was reported to reach up to 1.5 km/s in the presence of a magnetic field of 1 kOe by Kim *et al.* This ultrafast motion of domain wall was found in ferrimagnetic GdFeCo compounds when the ferromagnet FeCo and antiferromagnet Gd angular momenta balance each other [67]. Later Bläsing *et al.* also reported the domain wall velocity of 800 m/s in Co/Gd ferrimagnetic bi-layers driven by electric current [68]. The larger velocity is due to the maximized exchange coupling torque when the angular momenta of Co and Gd are compensated [68, 69].

Table 2.1 summaries the domain wall velocity driven by an electric current, in different magnetic configuration systems, which have different magnetic anisotropies. Generally, the domain wall width is smaller in material with higher anisotropy, and the domain wall motion is faster in synthetic antiferromagnetic

materials and ferromagnetic systems driven by the spin-orbit torque. In a Co/Ni/Co/Pt system, the velocity is fast due to the presence of the heavy metal Pt, which is helpful to form a chiral domain wall structure [70], as highlighted in the table. In the synthetic antiferromagnetic structure, the exchange coupling torque is more effective to drive domain wall motion besides the spin-orbit torque.

Samples stack	Ni ₈₀ Fe ₂₀	(Co/Ni) _n	Co/Ni/Co/Pt	(CoFe/Pd) _n /CoFe/Ru/Pd/(CoFe/Pd) _n /Ta	Pt/Co/Ni/Co/Ru/Co/Ni/Co	Co/Gd bilayer
Magnetic configuration	In-plane magnetized	Out-of-plane magnetized	Out-of-plane magnetized	Synthetic antiferromagnetism	Synthetic antiferromagnetism	Ferrimagnetic
Domain wall width	~200 nm	~20 nm	~20 nm	~5 nm	~5 nm	~5 nm
Current density	10 ¹² A/m ²	10 ¹¹ A/m ²	10 ¹⁰ A/m ²	10 ¹⁰ A/m ²	10 ¹² A/m ²	10 ¹² A/m ²
Domain wall velocity	100 m/s	100 m/s	400 m/s	150 m/s	750 m/s	800 m/s
Magnetic anisotropy energy density	10 ² J/m ³	10 ⁴ J/m ³	10 ⁴ J/m ³	10 ⁶ J/m ³	10 ⁶ J/m ³	10 ⁶ J/m ³
Method of domain wall driving	Spin-transfer torque	Spin-transfer torque	Spin-orbit torque	Spin-orbit torque	Spin-orbit torque	Spin-orbit torque

Table 2.1. Domain wall properties, domain wall velocity driven by electric current in different magnetic texture systems.

2.5 Domain wall pinning

Based on the previous discussion, domain walls can be propagated through nanowires at speeds of several hundred meters per second using the magnetic field, electrical currents and both. However, the domain wall motion shows a stochastic spread over a large distance of several micrometers, which preclude its applications [71-76]. Therefore, it is essential to control the domain wall motion at some specific positions.

To control the position of the domain wall, the introduction of the pinning sites is an effective method. The pinning sites also help to improve the stability of the stored information. The pinning methods can be classified into two main classes; (a) geometrical pinning sites and (b) non-geometrical pinning sites. The name geometrical pinning sites refer to the modification of nanowire shape using

lithography methods. On the other hand, non-geometrical pinning is based on modifying magnetic properties at certain positions with maintaining the geometry of the nanowire. In this section, we will give a brief review of the study about geometrical pinning sites, as well as the non-geometrical pinning sites reported by other researchers. In chapters 4, 5 and 6, we will systematically discuss the various methods of pinning the domain walls by using synthetic magnetic texture.

2.5.1 Geometrical pinning sites by creating notches

One of the earliest geometrical methods proposed for the domain wall pinning was creating artificial notches at some specific positions along the ferromagnetic nanowire. These notches alter the energy potential and hence domain walls need higher energy to escape from the notches. Kläui *et al.* in 2003 demonstrated the domain wall pinning at the triangular notches in a ring-shaped Permalloy nanowire [77]. They observed that notches created an attractive potential well for head to head domain walls at the site of notches and hence pin the domain wall which was confirmed using magnetoresistance measurements. In their later study, they also reported that notches are also able to obtain efficiently pinning of vortex head-to-head domain walls [78]. The pinning strength of the notch was an order of magnitude larger than the pinning that occurred naturally.

Hayashi *et al.* demonstrated controlled DW motion through a triangular notch in straight Permalloy nanowire [79]. They created a triangular notch at one edge of the nanowire and the depth of the notch was $\sim 30\%$ of the width of the nanowire. Since the magnetization direction in the soft magnetic material depends on the shape of the device, notches artificially change the magnetization

structure locally and hence changes the energy required to move the DWs. They reported the pinning and depinning of DWs at the notches upon the application of the magnetic field as well as an electrical current. A few more studies have been devoted to triangular notches in a Permalloy nanowire [80, 81]. Bogart *et al.* studied the effect of the geometrical confinement of the triangular notches and rectangular notches on the pinning strength [82]. After a single notch study, Endo *et al.* studied the pinning of the DWs in Permalloy nanowires with double notches (at both sides) [83].

In 2008, Parkin *et al.* demonstrated the motion of DWs along a nanowire with several triangular notches experimentally [84]. When a series of the current pulse was applied to the nanowire, DW was pushed by spin-torque transfer mechanism and stopped at the notches. The DWs were found to move in a direction opposite to that of the applied current. Figures 2.12 (a) and (b) show the T-T domain wall and H-H domain wall pinned at the notch. When a series of the current pulse was applied to the nanowire, a domain wall was pushed and stopped at the notch. Figure 2.12(c) clearly shows the controlled motion of the NiFe domain wall in the successive notches. Depending on the current direction, the domain walls are moved in either the left or right direction along the nanowire. In the device illustrated in Figure 2.12(c), ten notches were patterned. However, only eight pinned positions of them were observed. The motion of the domain walls is not completely reliable under the experimental conditions. In some cases, the domain walls do not move, or move only across a notch, or skip a notch while sending a single pulse.

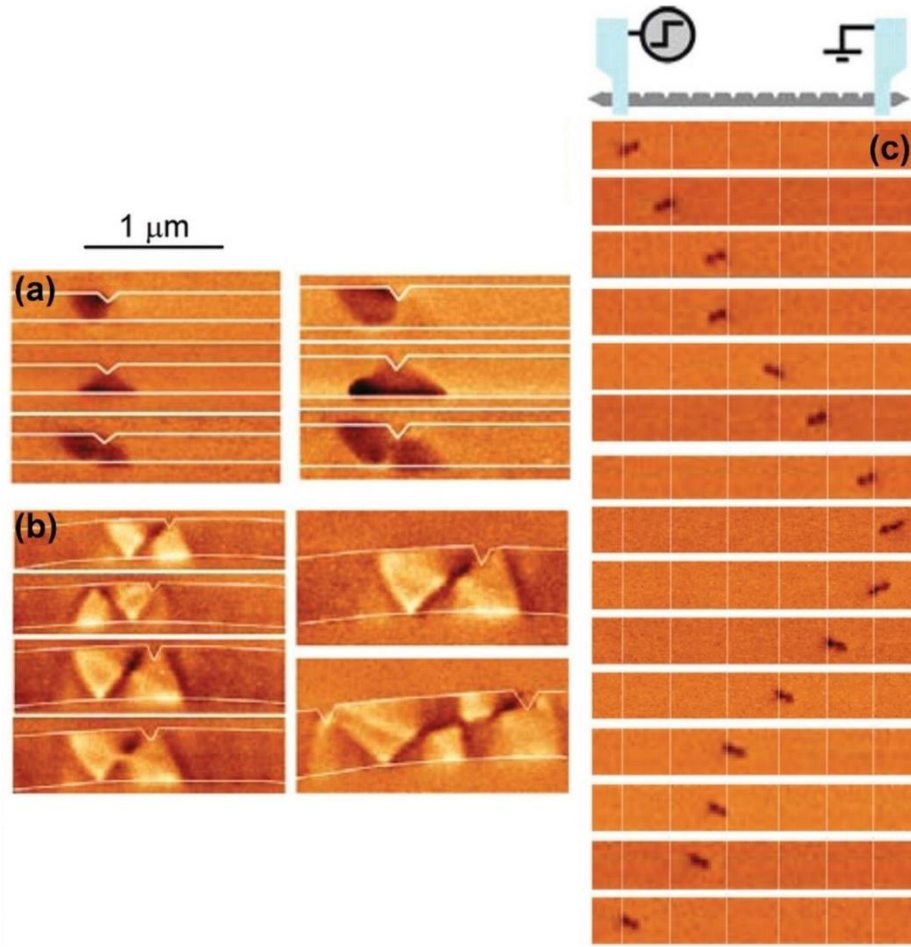


Figure 2.12. (a) and (b) Structure of tail-to-tail DWs and head-to-head DWs observed in a NiFe nanowire. (c) Magnetic force microscopy images of a single tail to-tail vortex DW. For the eight images at the top, pulses with negative polarities were applied. For the seven images at the bottom, pulses with positive polarities were applied [35].

The strength of the pinning sites strongly depends on the shape and dimensions of the notches. However, the notches suffer from variations in shape leading to non-uniform DW pinning strengths. The narrow region also causes Joule heating at the notches due to the increased resistance and current density in the notch position. In addition, if notches are made with the smallest feature size F achievable with a particular lithography technique, the nanowire has to be bigger than F ($3F$, for example). This poses a limitation in the areal density of nanowires.

2.5.2 Geometrical pinning sites based on staggered nanowire

The idea of using a staggered nanowire for controllable domain wall motion was conceived by Bahri *et al.* in 2016 [85]. In this design, two portions of the simple nanowire, which are of the same dimensions, are connected with a small offset in one direction or both. Using micromagnetic simulations, they demonstrated that the domain wall could be pinned in the stepped region and the driven-current density depends on the dimensions of the step. In this study, they simulated the domain wall pinning and depinning based on in-plane magnetic anisotropy (IMA) materials for various values of step dimensions, anisotropy constant and saturation magnetizations and hence presented a phase diagram for a stable domain wall as a function of the dimension of the step.

Recently, Bahri *et al.* demonstrated the effective pinning in a NiFe staggered nanowire experimentally [86]. Figure 2.13(a) shows the schematic of the NiFe nanowire device with multi-staggered nanowire, which has the depth (d) of step and length (l) of step [87]. Figure 2.13(b) displays the DW pinned and depinned at the constriction region driven by the magnetic field. The depinning field is increasing with the larger d shifting. They also pointed out that for $l = 0$, the pinning is stronger compared to that for $l = 50$ nm. By changing the parameters of d and l , the pinning strength can be easily adjusted. As well as the domain stability is enhanced with increasing d value, which can be stabilized about one year for $d = 150$ nm.

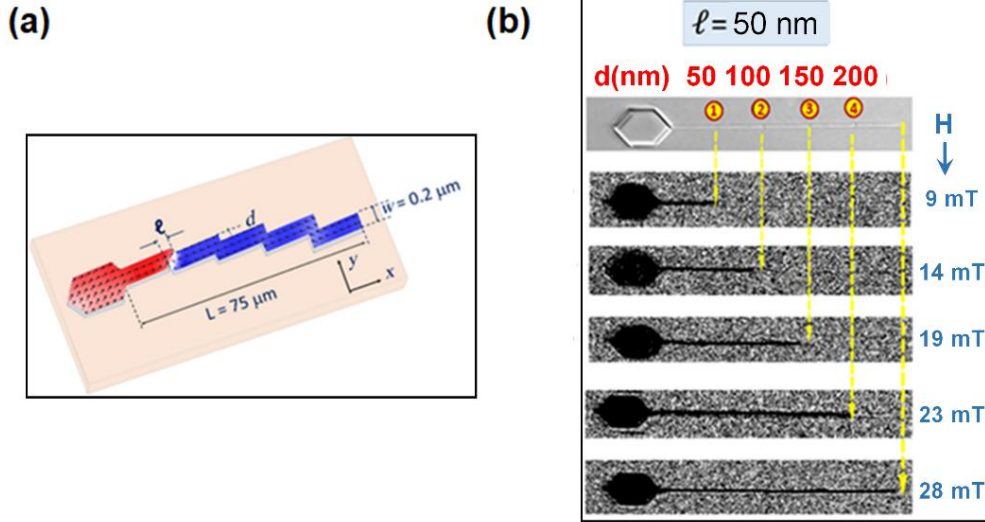


Figure 2.13. (a) Schematic representation of the device with multi-staggered nanowire and (b) the depinning field is increasing with various d from 50 to 200 nm [117].

As a further study, Kumar *et al.* performed micromagnetic simulations for perpendicular magnetic anisotropy (PMA) materials in a similar nanowire geometry [88]. The micromagnetic parameters were chosen for Co/Ni multilayer samples. The study illustrates that it is possible to pin the DW by adjusting the nanowire step length and its width. From these studies, we noted that the staggered design provides control parameters such as the length and width of step to control the domain wall position. Also, the depinning current can be tailored by changing these two parameters.

The geometrically pinned magnetic domain wall is a simple and useful method to control domain wall propagation. However, there are still some drawbacks and limitations. To pattern the notches or the staggered structures, a complicated fabrication process is required. High-resolution patterning and etching processes are necessary to suppress the effect of the edge roughness, which may unintentionally introduce extra pinning. Especially, it becomes quite challenging to fabricate smaller pinning sites in nanowires with a width of few

tens of nm. Therefore, some researchers performed the non-geometrical methods of domain wall pinning.

2.5.3 Non-geometrical pinning sites by using exchange bias

One of the earlier approaches of non-geometrical methods of domain wall pinning was proposed by Polenciuc *et al.* in 2014 [89]. They proposed that an exchange bias field at regular intervals of a nanowire helps to pin the domain walls in a controllable manner. The sample with the structure NiCr (6 nm)/IrMn (5 nm)/CoFe (20 nm) was used for achieving the exchange bias. They achieved exchange bias by fabricating antiferromagnetic wires over and below the ferromagnetic wire and subsequently by field cooling, as shown in Figure 2.14(a). The width of the cross wires was varied in the range of 1 to 2.5 μm , the switching field was tuned from left to right, as shown in Figure 2.14(b). Using this method, they observed the successful domain wall pinning at the pinning sites. The simplicity of design and fabrication makes this method considerable.

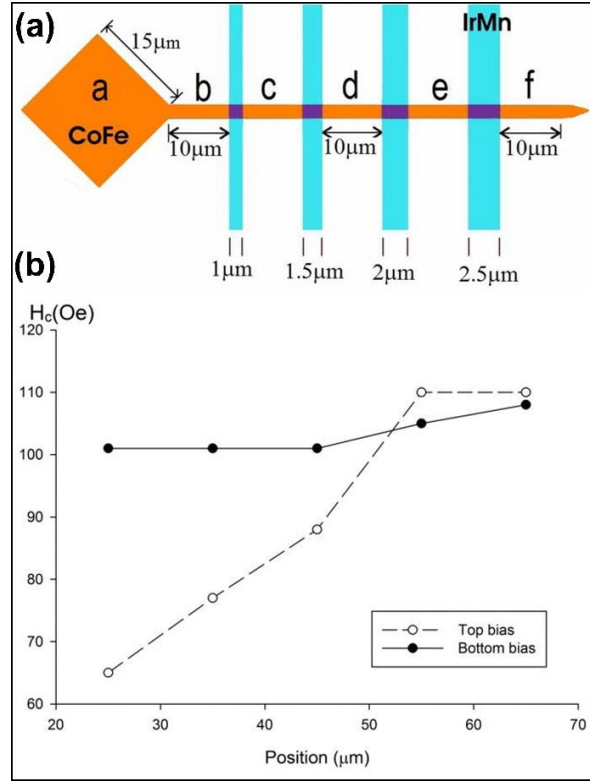


Figure 2.14. (a) Schematic diagram of the ferromagnetic nanowire with antiferromagnetic crossbars pinning sites. (b) The coercivity at the points labeled (b to f) in the racetrack memory with bottom bias (full circles) and top bias (empty circles) configurations [44].

2.5.4 Non-geometrical pinning sites by tuning magnetic properties

The other non-geometrical method of domain wall pinning was to modify the properties of a magnetic nanowire locality. Benitez *et al.* reported Ga^+ ion irradiation on Permalloy nanowires as an effective technique for creating the pinning sites. The technique was shown in Figure 2.15(a) [90]. A similar study was also carried out by D. Burn [91]. In the Au/NiFe nanowire, Ga^+ ions irradiation was used to induce the intermixing between Au and NiFe bilayer. Thus a decrease in magnetization upon the Ga^+ ion irradiation was observed. The depinning field increases with increasing irradiation dose, as shown in Figure 2.15(b).

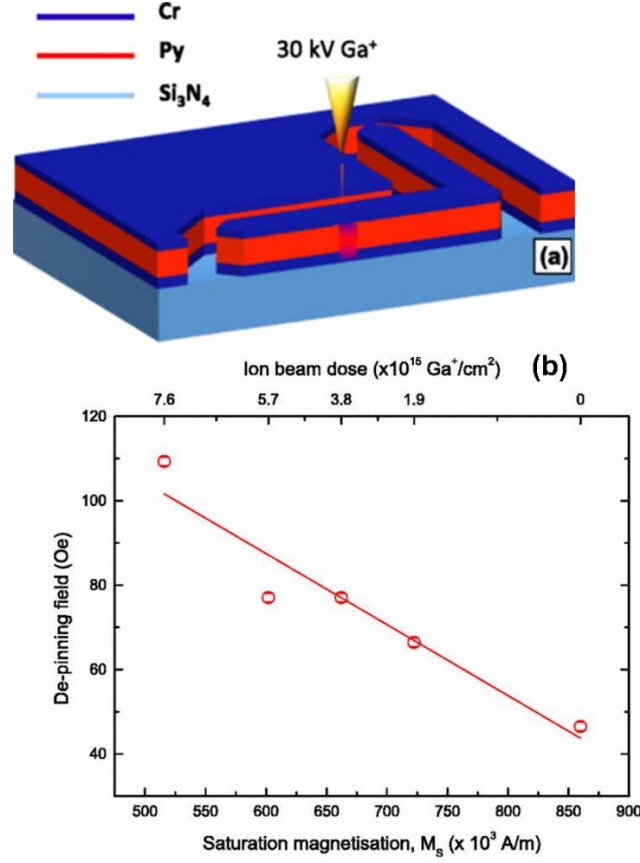


Figure 2.15. (a) Schematic of the ion irradiation fabrication process [90]. (b) Domain wall de-pinning field and the estimated saturation magnetization for various ion doses in an Au/NiFe nanowire [91].

2.6 Conclusion

In this chapter, we discussed domains and domain walls formation which is originated from the minimization of the total energy, resulting in a better stability. Domain wall with a different configuration, such as head-to-head or tail-to-tail, Bloch or Néel walls, has been discussed in different anisotropy energy systems. Domain wall dynamics and velocity have been introduced in in-plane and out-of-plane magnetized systems driven by the magnetic field and electrical current. In the last part, we gave a brief review of domain wall pinning by using geometrical and non-geometrical methods.

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Chapter 3 Experimental details

3.1 Introduction

This chapter mainly focuses on the description of experimental techniques such as the thin film deposition, device fabrication and characterizations of thin films and devices, as well as some other analytical techniques. Several approaches, which were used to modify the magnetic properties of samples to form the synthetic magnetic textures, are introduced. The micromagnetic simulation and nanowire fabrication are followed. Firstly, we introduce the thin film deposition process, which was performed by using magnetron sputtering, followed by the techniques for sample characterization and analyzation. The magnetic characterizations on the thin film level were carried out by alternating gradient force magnetometer (AGFM), vibrating sample magnetometer (VSM), ferromagnetic resonance (FMR), and magnetic force microscopy (MFM). Nanowires were patterned from thin films by using electron-beam lithography (EBL) followed by argon ion milling. Electric and Kerr microscopy measurements on the device level were utilized to probe the DW properties. Simulations were carried out using Mumax3, which is an open-source software and accelerated by the use of graphic cards. At last, we are going to discuss the approaches to modify the magnetic properties of the prepared sample, such as by using thermal annealing, ion-implantation and introducing exchange coupling form another layer.

3.2 Thin film deposition

All thin films, investigated in this thesis, were deposited using high vacuum magnetron sputtering system. Sputtering is a physical vapor deposition process

which employs high voltage to ionize inert gases, such as argon (Ar), to generate plasma. The positively charged ions are accelerated towards a target (cathode) which is biased negatively. The ions can knock the atoms out of the target material. The ejected atoms travel in a line of sight cosine distribution and get deposited on the substrate that is placed adjacent to the sputtering cathode. Conductors and metallic source materials can be deposited using a DC power source whereas insulators are deposited using an RF power source. In the sputtering process, arrays of magnets are placed below the targets, in a variation called magnetron sputtering. It utilizes a closed-loop of magnetic fields around the target to confine and trap electrons. This trapping increases the path length of electrons, causing them to move in a spiral motion. As a result, the efficiency of the ionization process is enhanced and the plasma is generated at a lower pressure. The system is also water-cooled, to reduce the heat generated during the Ar bombardment of the targets. In this thesis, all the metallic thin films were prepared by using DC power and the working principle is shown in Figure 3.1 [1].

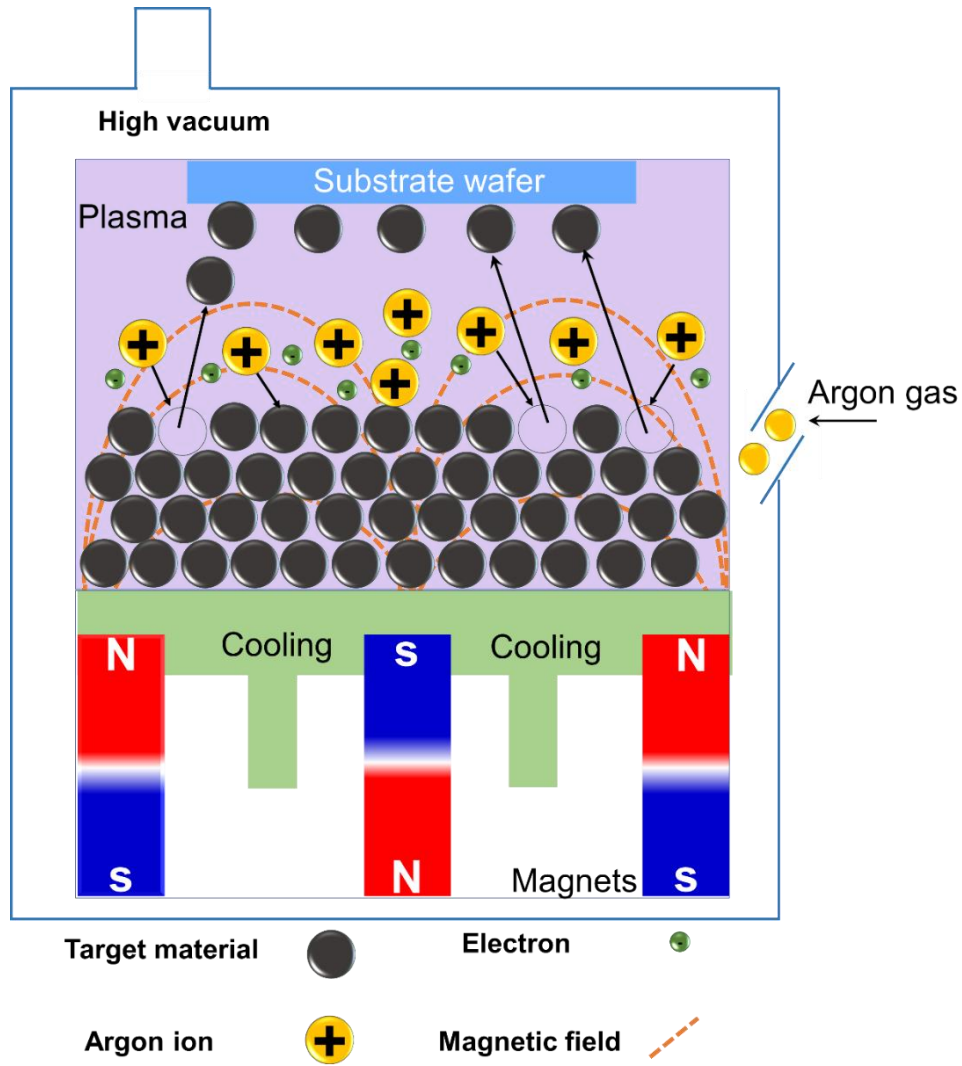


Figure 3.1. Schematic of DC magnetron sputtering [1].

The $\text{Ni}_{80}\text{Fe}_{20}$ samples were deposited on silicon substrates covered by 500 nm thermally oxide SiO_2 layer via DC magnetron sputtering in an AJA system at room temperature. The base pressure of the main chamber was 1.1×10^{-7} Torr. The argon pressure during the deposition of the $\text{Ni}_{80}\text{Fe}_{20}$ layer was fixed at 2 mTorr and the sputtering DC power was 40 W. The deposition rate was 0.12 nm s^{-1} for $\text{Ni}_{80}\text{Fe}_{20}$, which was calibrated by atomic force microscopy (AFM). Different capping layers, like Ta, Cr, Cu, and Ru, were deposited at an argon pressure of 3 mTorr and the DC power of 35 W, separately. Finally, we made the sample structure ($\text{Ni}_{80}\text{Fe}_{20}/\text{X}$ and $\text{X} = \text{Ta, Cr, Cu, and Ru}$), *i.e.*, 10 nm thick

Ni₈₀Fe₂₀ thin film with different non-magnetic metal capping layers, which thickness is 5 nm, as well as Ni₈₀Fe₂₀ thin film without a capping layer as a reference.

Perpendicularly magnetized Co/Pd multilayer films with a stack of substrate/Ta (5 nm)/Cu (5 nm)/Pd (3 nm)/[Co (0.3 nm)/Pd (0.8 nm)]_N/Pd (0.8 nm) and Co/Ni multilayers with the structure of substrate/Ta (5 nm)/Pt (5 nm)/[Co (0.25 nm)/Ni (0.5 nm)]_N/Co (0.25 nm)/Pt were grown on thermally-oxidized Si/SiO₂ substrates using DC magnetron sputtering at room temperature, where *N* indicates the repeating cycles in each film. The bilayers Co and Pd in Co/Pd multilayer samples were grown at the DC power of 40 W, and the deposition rate of Co and Pd is 0.4 nm s⁻¹ and 0.357 nm s⁻¹, respectively. The deposition rate of Co and Ni for Co/Ni multilayer sample is 0.028 nm s⁻¹ and 0.03 nm s⁻¹, respectively. During the deposition procedure, the Ar pressure was kept at 3 mTorr.

3.3 Characterization methods for thin film

3.3.1 Alternating Gradient Force Magnetometer

Alternating gradient force magnetometer (AGFM) is used to measure the force exerted on the magnetized samples in the magnetic field gradient. The alternating gradient field can produce a periodic force which acts on the sample. This alternating force on the sample is proportional to the magnitude of the gradient and the magnetic moment of the sample. The sample placed in the magnetic field is vibrated at the fixed frequency via an electromechanical transducer and mounted on an extension rod attached to a piezoelectric element. The resulting deflection of the extension is transmitted to the piezoelectric sensing element.

The signal detected by the piezoelectric element is much more enhanced by the operating frequency which is at or near to mechanical frequency of the assembly. Software provided by the equipment maker is used to automatically determine mechanical resonance and set the appropriate operating frequency for the study sample [2-4]. The model of AGFM used in the thesis is MicroMag AGM 2900. The operating frequency is around 550 Hz for the $\text{Ni}_{80}\text{Fe}_{20}$ thin film samples. A schematic of AGFM system is shown in Figure 3.2.

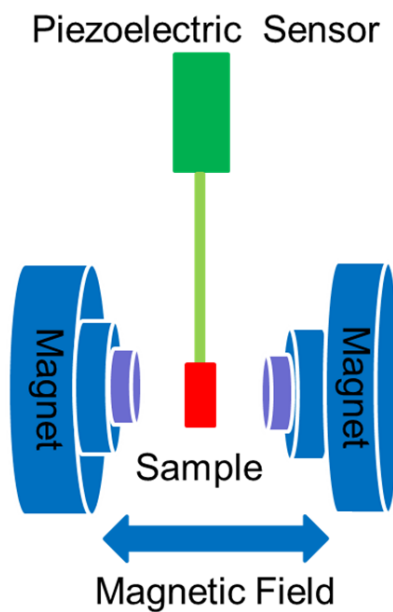


Figure 3.2. Schematic of AGFM system to characterize magnetic properties of the magnetic thin film under investigation.

3.3.2 Vibrating Sample Magnetometer

Vibrating sample magnetometer (VSM) is a measurement equipment to determine the magnetic characteristics of a sample with very high precision (< sub-micron emu). The principle of VSM is based on Faraday's law of induction which accounts for the generation of an electromagnetic force due to a change in magnetic flux. In VSM systems, a set of pickup coils is used to generate a voltage

arising from the time-varying flux. The pickup coils detect the magnetic flux change from the magnetic sample, which can be simply described as follows:

$$B = \mu_0(H + M)A \quad (3.1)$$

where B represents the flux in the coil, μ_0 represents the magnetic permeability, H is the applied external magnetic field, and M is the magnetization of the measured sample. The change in flux can be calculated by measuring the voltage (V) induced, which can be described simply with:

$$V = -N \frac{dB}{dt} \quad (3.2)$$

where V is the induced voltage, and N is the number of turns in the pickup coils. Hence, by measuring the induced voltage from the pickup coil, and with the accurate calibration of the applied magnetic field, H . The flux from the magnetization of the sample, M , can be easily found. Further calibration with a pure nickel sample of known magnetization is required to extract the magnetization of the sample from the induced voltage measured [5-7].

The change in flux is achieved by vibrating the magnetic sample in close proximity to the two-pickup coils as schematically shown in Figure 3.3(a). The oscillating driver provides a sinusoidal vibration at a fixed frequency to the sample, allowing the signal to be picked up by a lock-in amplifier. This enables magnetic signals in the range of 10^{-9} emu to be detected by the Lakeshore 8600 VSM system used in this thesis work. Figure 3.3(b) displays the VSM system used. As with most VSM systems, the pickup coils are placed at the pole caps of the electromagnets. The hysteresis loop can be measured in the film-plane or out-of-plane by orienting the sample in different directions with respect to the applied magnetic field.

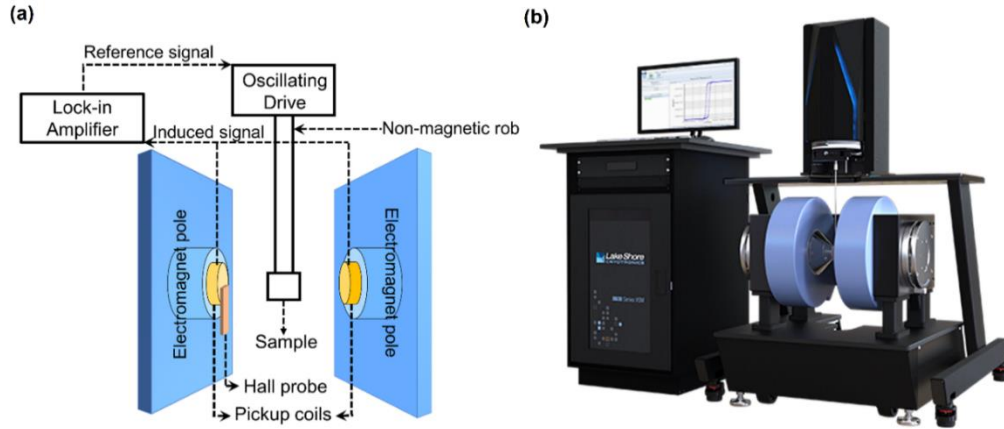


Figure 3.3. (a) Schematic diagram illustrating the working principles of the VSM system and (b) Lakeshore 8600 VSM system for magnetization characterization.

3.3.3 Ferromagnetic Resonance

The ferromagnetic resonance (FMR) technique is an established experimental technique to study and characterize the dynamics of magnetization reversal or oscillations. For this investigation, the sample is placed in a resonance microwave cavity. A Gunn diode provides microwave radiation with a fixed frequency, typically range from 1 to 80 GHz. An electromagnet provides an external magnetic field. The resonance frequency or field at which the absorption takes place is monitored for determining many parameters related to magnetization dynamics.

In modern measurement systems, a Vector Network Analyser (VNA) is usually used for carrying out FMR investigations. A VNA is capable of supplying Radio Frequency (RF) and measuring the RF signal absorbed, as a function of frequency. Alternatively, one can sweep the power level by fixing the frequency at a constant value. In this case, the obtained parameters of the device or network performance will be displayed as a function of power at a fixed frequency. In most experiments, a two-port VNA with the RF signal in the microwave range is

deployed to act as a source and detector for the device under test (DUT). It is used to measure the incident signal, reflected signal and transmitted signal to and from the DUT. These measured signals, also known as scattering parameters (S-parameters), are complex values and contain the magnitude (dB) and phase (degrees). The VNA setup is shown in Figure 3.4 [8, 9].

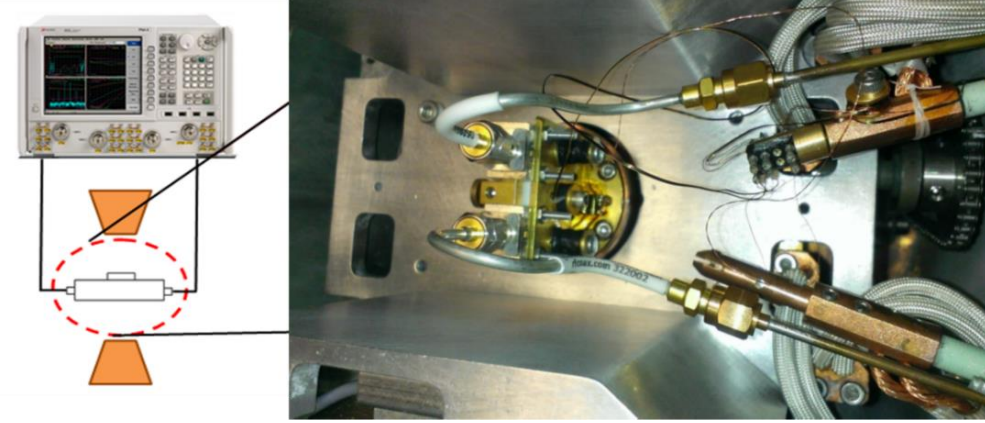


Figure 3.4. The set-up of the FMR system used in the thesis [9].

For a magnetic material, the magnetization will tend to align themselves in the direction of the field in order to minimize their magnetostatics energy. However, as the magnetic moment has its angular momentum, it experiences a torque and precesses around the direction of the field. This precession is known as the Larmor precession. In addition, a damping force eventually settles the magnetization along the field direction. This relation is expressed by the Landau-Lifshitz-Gilbert (LLG) equation [10]:

$$\frac{d\mathbf{M}}{dt} = -\gamma\mu_0\mathbf{M} \times \mathbf{H}_{eff} + \frac{\alpha}{M_s} \left(\mathbf{M} \times \frac{d\mathbf{M}}{dt} \right) \quad (3.3)$$

where $\mathbf{M}$ is the local magnetization, γ is the gyromagnetic ratio, μ_0 is the vacuum magnetic permeability, $\mathbf{H}_{eff}$ is the effective magnetic field in a sample and α is the dimensionless Gilbert damping constant. $\mathbf{H}_{eff}$ comprises of multiple magnetic contributions, such as the Zeeman energy, exchange energy, anisotropy energy

and demagnetizing energy, which collectively determines the static equilibrium orientation of the magnetic dipole. The first term in the LLG equation describes the circular path around the magnetic field, while the second term describes the converging motion towards the effective magnetic field.

By applying an external oscillating force perpendicular to $\mathbf{H}_{\text{eff}}$ over the magnetic moments, such as a radio frequency (RF) electromagnetic signal, we will be able to excite a magnetic moment causing it to precess at a larger angle away from $\mathbf{H}_{\text{eff}}$. If the frequency of the precession coincides with the frequency of the external oscillating force, ferromagnetic resonance occurs, and all magnetic moments will be oscillating in phase. At resonance, maximum energy will be absorbed by the magnetic sample, and this is reflected as a dip in the transmitted energy spectrum of the RF signal. The ferromagnetic resonant frequency, ω_0 , in the applied field is determined by the following equation [11]:

$$f = \frac{\gamma}{2\pi} \left[(H_{\text{appl}} + H_k) \times (H_{\text{appl}} + H_k + M_s) \right]^{1/2} \quad (3.4)$$

For in-plane magnetic system, the LLG equation can be writhed as:

$$\begin{cases} 0 = m_y \frac{M_{\text{eff}}}{M_s} m_z - h_{\text{rf}} m_z \\ \frac{dm_y}{dt} = m_z (H_0 + H_{\text{uni}}) + M_{\text{eff}} m_z - \alpha \frac{dm_z}{dt} \\ \frac{dm_z}{dt} = h_{\text{rf}} M_s - m_y (H_0 + H_{\text{uni}}) + \alpha \frac{dm_y}{dt} \end{cases} \quad (3.5)$$

in which $m_y = m_y e^{-i\omega t}$ and $m_z = m_z e^{-i\omega t}$,

$$\begin{pmatrix} H_0 + H_{\text{uni}} + i\omega\alpha & -i\omega \\ i\omega & H_0 + H_{\text{uni}} + M_{\text{eff}} + i\omega\alpha \end{pmatrix} \begin{pmatrix} m_y \\ m_z \end{pmatrix} = M_s \begin{pmatrix} h_{\text{rf}} \\ 0 \end{pmatrix} \quad (3.6)$$

By substituting Equation (3.5) in $\begin{pmatrix} m_y \\ m_z \end{pmatrix} = A^{-1} \begin{pmatrix} h_{\text{rf}} \\ 0 \end{pmatrix} M_s$, we obtained

$$\begin{pmatrix} m_y \\ m_z \end{pmatrix} = \frac{M_s}{(H_0 + H_{uni})(H_0 + H_{uni} + M_{eff}) - \omega^2 + i\omega\alpha(2(H_0 + H_{uni}) + M_{eff})} \begin{pmatrix} H_0 + H_{uni} + M_{eff} + i\omega\alpha & i\omega \\ -i\omega & H_0 + H_{uni} + i\omega\alpha \end{pmatrix} \begin{pmatrix} h_{rf} \\ 0 \end{pmatrix}$$

The slope of a plot of ΔH vs. f can give rise to the damping parameter based on the following equation [12]:

$$\Delta H = \frac{4\pi}{\gamma} \alpha f + \Delta H_0 \quad (3.7)$$

3.3.4 Magnetic Force Microscopy

Magnetic force microscopy (MFM) is a member of the group of scanning probe microscopy (SPM) techniques. MFM utilizes a magnetized tip to sense the weak fringing magnetic field and map this field over an area of the magnetic sample surface. The tip is usually made by depositing some magnetic material, like Co or CoCrPt alloy, on atomic force microscopy (AFM) tip. Prior to understanding the principle of MFM, the principle of AFM is described.

Usually, AFM operates by measuring the van der Waal's force between a tip and the sample [13, 14]. The AFM tip is attached to a sensitive cantilever, whose tip is brought closer to the surface of the sample. The cantilever is deflected depending on the force of attraction or repulsion, which in turn depends on the distance between the tip and the measured sample. A laser spot reflected from the back of the cantilever is captured by the array of a photodetector to measure the deflection. The deflection of the cantilever provides topography-related information.

In this thesis, the thickness of the thin film was calibrated by AFM measurements. The stepped film was deposited on the Si substrate whose surface

was drawn some straight lines by using a marker pen. Then Isopropyl alcohol (IPA) was used to strip the maker lines. The prepared stepped structure was scanned by an AFM tip. The topography image was obtained, which is able to figure out the step depth, corresponding to the thickness of the thin film.

During MFM measurements, the tip experiences a magnetic interaction in addition to the van der Waals force. To ignore the atomic force, MFM scanning is performed in the lift mode to reduce the effect from van der Waal's interaction, where the tip is lifted 50 nm above the sample to remove the topography information. The defection of the cantilever provides magnetic properties [15-17]. In this thesis, a Bruker D3100V system was used to calibrate the thickness of the thin film and image the domain structure.

Figure 3.5 shows a schematic of MFM. For magnetization along the z -direction, the tip can distinguish the repulsive and attractive force between the tip magnetization and the local magnetization of a sample, which gives the opposite contrast indicating the up magnetization or down magnetization. For a magnetization along in-plane (x - y plane), the stray field from the DW provided the opposite contrasts either from the head to head (H-H) DW and tail to tail (T-T) DW.

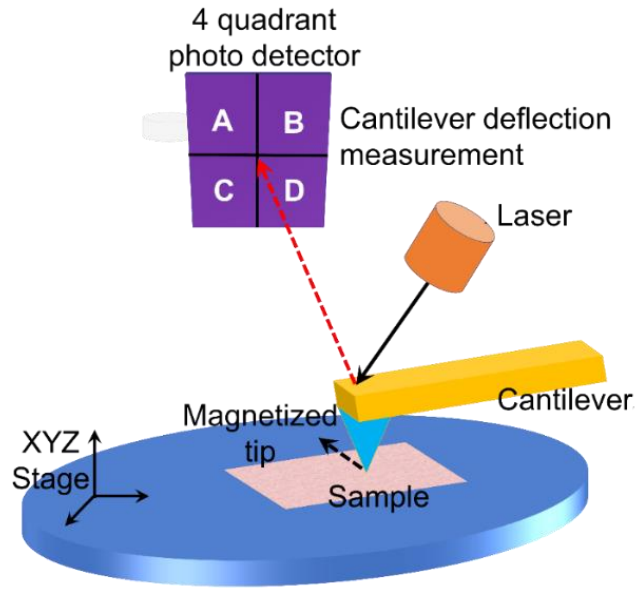


Figure 3.5. Schematic shows the working mechanism of the magnetic force microscopy (MFM).

3.4 Device fabrication

Electron beam lithography (EBL) is a powerful tool to make high-resolution nanostructures. A narrow beam of electrons (e-beam) is used for exposure, as e-beam has a very low wavelength of about 1 nm, which helps to achieve a feature size of 5 nm.

There are some basic parameters chosen for any EBL, like the acceleration voltage of the electrons, the electron beam current, and aperture size. In our group, we are using the commercial Raith system, as shown in Figure 3.6(a). In this system, any value up to 20 kV can be chosen as acceleration voltage, and the aperture size for the electron beam, like 7 μm , 30 μm , 60 μm , 120 μm , *etc.*, can be selected. The larger size of the aperture allows the higher beam current. However, the increased aperture size results in poorer resolution. The pattern layout was designed using GDSII-editor.

In e-beam lithography, the electron beam resist is an important part of the process for exposure, which is the medium for transferring information. Usually, there are two kinds of electron beam resist: negative e-beam resist and positive e-beam resist. In the case of the negative resist, the resist will become harder when exposed by the electron beam and remains at the exposed regions after development. Whereas, in the case of the positive resist, the resist becomes softer after exposure and is washed away during development [18-21]. The choice of resist is based on the process requirement. The schematic of EBL for positive and negative resist is illustrated in Figure 3.6(b).

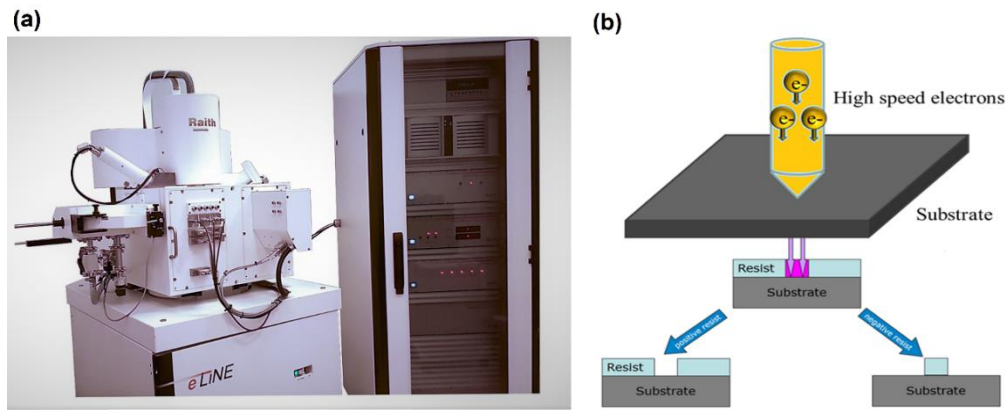


Figure 3.6. (a) Raith EBL system and (b) the schematic of electron beam lithography for positive and negative resist. For positive resist, the region exposed by the electron beam becomes softer and can be removed by the developer process. For negative resist, the region exposed by the electron beam becomes hard and cannot be removed by the developer process.

In order to fabricate nanowire devices with different pinning bars in this study, the process can mainly be divided into three sections. I explain the whole fabrication procedures for manufacturing the $\text{Ni}_{80}\text{Fe}_{20}$ (NiFe) nanowire structure with Cr pinning crossbars, as an example.

3.4.1 Fabricating NiFe nanowire

In order to make Ni₈₀Fe₂₀ device with pinning crossbars, I started with the NiFe film sample of 10 nm thickness grown on SiO₂ (500 nm)/Si wafer of 1 cm by 1 cm. Negative photoresist MaN-2403 was spin-coated upon the sample film to obtain a thin 200 nm resist layer on the sample. The spin coat was followed by 100 °C baking for two minutes to remove the solvent and to achieve the uniform distribution of the e-beam resist on the sample surface. After this, the whole wafer with the resist and the sample films was patterned based on the designed layout. The exposure voltage was 20 keV and the aperture size was 7 µm, and the area dose was 130 µC/cm². Due to the properties of negative e-beam resist, the area exposed by the electron beam became hardened. After the treated wafer was rinsed in separate MaD 525 solutions for 45 seconds twice, and then in DI-water for 2 minutes, only the resist exposed by e-beam was left behind. Then Ar ion milling was carried out to etch the rest of the film unprotected by the e-beam resist, in order to produce the desired nanowire structure. As a final step, the resist was stripped using MR-remover in ultrasonic irrigation for about 10 hours. This process is shown in Figure 3.7(a).

Ion milling is a dry etching technique where ionized gas bombards the surface of the sample to remove material from it, thereby etching the materials. Figure 3.8 shows the ion milling system AJA ATC-2020-IM, which is used in this work. Ar gas in the chamber was accelerated by a high voltage grid to form a collimated beam of ionized Ar ions. This ionized Ar moved towards the sample and etched the sample. The Ar flow rate into the milling chamber was maintained at 12 sccm and the power of etching was kept at 53 W. The incident angle from the direction of Ar bombardment was fixed at 5°, which helped to avoid any re-

deposition of the material to the side of the etched places. The stage was water-cooled to reduce the heating generated during the Ar bombardment of the sample and the stage.

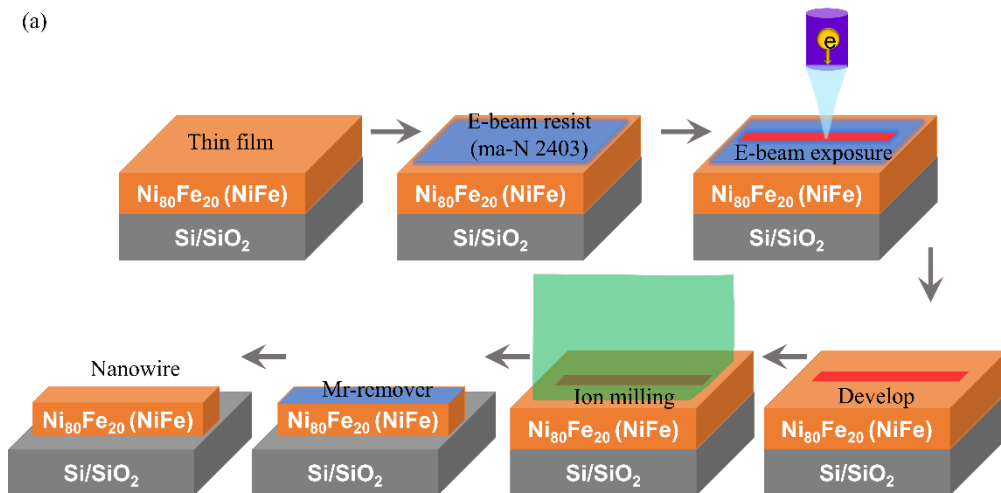
3.4.2 Fabricating Cr crossbar nanowire

The second section of the process is the fabrication of the crossbar nanowire. This step is in continuation of the step as shown in Figure 3.7(b). I started with the wafer already patterned in the first section, where the NiFe nanowire structure of the sample was fabricated. Then, the positive e-beam resist Polymethylmethacrylate (PMMA) layer was spin-coated on the sample. A similar e-beam exposure operation was made with the same electron beam lithography system. The exposure of this section was done according to the coordinates established in the first exposure operation accurately. During this procedure, the acceleration voltage was 10 keV and the aperture size was 30 μm , and the area dose was 130 $\mu\text{C}/\text{cm}^2$. After the exposure, the development was carried out by rinsing the sample in MIBK for 30 seconds and in Isopropyl Alcohol (IPA) for 30 seconds to clean the exposed parts. Then the non-magnetic Cr element was deposited using a magnetron sputter system. Note that, for the study involving B^+ implantation in a Co/Pd multilayer sample, boron ions (B^+) was implanted instead of depositing Cr. The last step is the lift-off of the unwanted Cr or B^+ using acetone and IPA. In the case of Cr diffusion, devices were annealed in a vacuum atmosphere in order to make non-magnetic metal diffusion into the ferromagnetic nanowire at the cross-section to form pinning sites. The details about the annealing temperature and time will be discussed in

Chapter 4. No annealing was carried out in the B⁺ implanted samples. The details about implantation conditions will be present in Chapter 5.

3.4.3 Fabrication of electrode

The final step of the device fabrication is to make contact pads for measuring the transport properties by injecting a pulsed current. The positive e-beam resist was used and developed in MIBK and IPA to wash away the exposed regions after the e-beam exposure. This process is similar to the fabrication of Cr crossbar nanowire, but the larger aperture of 120 μm and the larger area dose of 300 $\mu\text{C}/\text{cm}^2$ were used due to the larger dimension of the electrode. After this step, the electrode film, which consists of tantalum (Ta) with a thickness of 10 nm, copper (Cu) with a thickness of 100 nm and gold (Au) with a thickness of 20 nm from bottom to top were deposited. At last, the sample was cleaned by acetone for 30 seconds and IPA for 30 seconds in ultrasonic irrigation to lift-off the unexposed area, as shown in Figure 3.7(c).



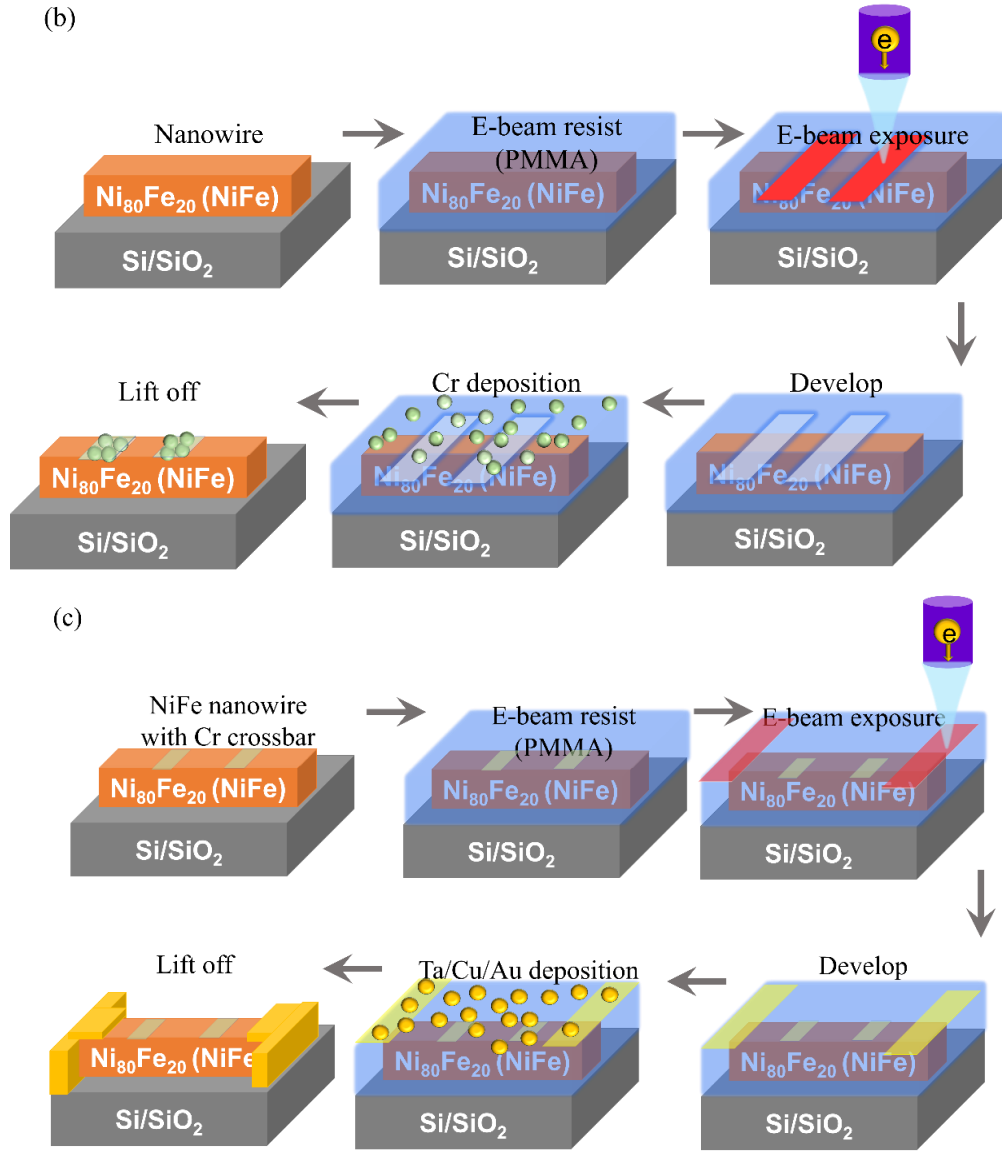


Figure 3.7. Schematic illustration of the nanowire device fabrication process. (a) The process of the nanowire structure fabrication. The sample film is deposited on a SiO₂/Si wafer firstly by magnetron sputtering, and the negative resist is spin-coated next. The wafer is then patterned using electron beam lithography and Ion milling to obtain the nanowire structure. (b) The process of the non-magnetic crossbar and contact pad fabrication. The wafer with the nanowire pattern is capped with positive resist and patterned to attain the crossbar road by electron beam lithography at first. Then the non-magnetic metal is deposited and followed by lift-off. The last step is to make the electrode contact bar. (c) The positive resist was used and patterned by electron beam lithography. A film consisting of Ta/Cu/Au is deposited by magnetron sputtering. Lift-off is done eventually to obtain the final device structure.



Figure 3.8. AJA ion milling system.

3.5 Kerr microscopy

Vibrating Sample Magnetometer (VSM) is a useful tool to quantitatively measure the magnetization of thin films, typically about 2 nm in thickness or above and an area of 4 mm $\times$ 4 mm. However, it is not efficient to measure the magnetization of an ultrathin magnetic film, particularly in nanowire devices, with a thickness below 1 nm. Magneto-Optical Kerr Effect (MOKE) is a more effective method to qualitatively measure the properties for ultrathin film and nanowire devices [22-24]. This technique relies on the Kerr effect. When a polarized light is focused on the magnetized sample surface, it induces a transverse displacement on the electrons and causes a certain rotation of the polarization due to the electric field from the electromagnetic wave. Therefore, the reflected light has a different polarization angle, called Kerr rotation. This rotation between the incident light and the reflected light reflected is related to the magnetization. In this case, the Kerr rotation can be used to detect the magnetization orientation of the magnetic sample.

Figure 3.9(a) shows a schematic of the polar MOKE setup. The main components of the setup are presented in the figure. A linear polarized light is obtained with the light passing through a polarizer. The incident light focused on the surface of the magnetic sample with a specific angle of the polarization. An analyzer is used to detect the Kerr rotation of the reflected lights. A camera is placed instead of a photodetector to get the image of Kerr rotation. The change of the magnetization can be identified from the images. Thus, the domain motions and domain patterns can be characterized by using Kerr microscopy imaging.

Figure 3.9(b) shows the Kerr microscopy used in this thesis. This Kerr microscopy system “MagVision” was built by Vertisis. It uses ultrabright 2200 lumens LED as the white light source, and a narrow bandwidth filter to ensure a high linear polarization for achieving the strong magnetic contrast and the high resolution [25, 26]. The magnetic field can reach about ± 1 T using a closed-flux design without a chiller cooling system. For the stage motion, piezo actuation under remote control is used for fully automatic sample stabilization in the nanometer scale. This Kerr microscopy can measure the magnetic hysteresis loops with the simultaneous capturing of domain images.

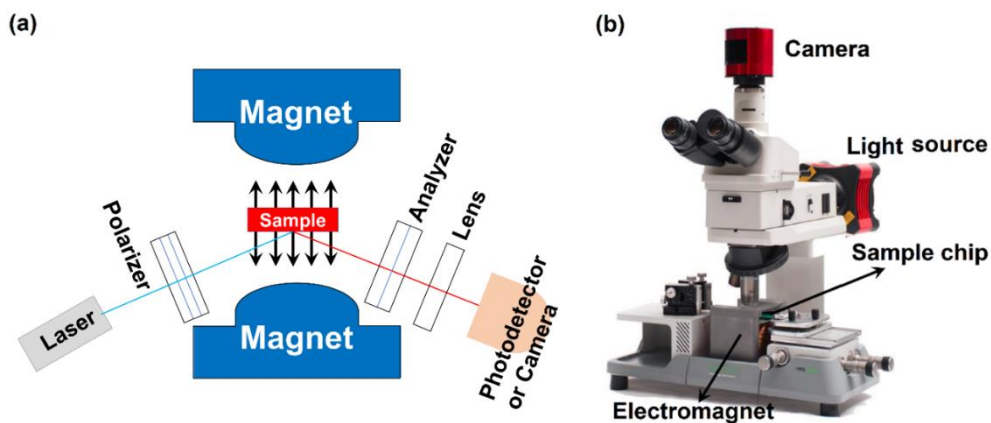


Figure 3.9. (a) Schematic diagram of a polar MOKE setup for magnetization characterization and (b) the Kerr microscopy setup.

3.6 Micromagnetic simulations

To describe the behavior of the magnetization, the LLG equation is widely used by considering the contribution of each term. However, it is typically difficult to analytically solve the LLG equation for a magnetic system. Therefore, micromagnetic simulation tries to calculate it numerically using a finite-difference method. In this study, we performed micromagnetic simulations to understand DW dynamics by using Mumax3, which is an open-source software [27].

Mumax3 is a GPU-accelerated simulation program with a 100 times faster speed compared to CPU based simulations. Moreover, it is optimized for low memory use and even can solve the LLG equation for 16 million finite difference (FD) cells with 2 GB of GPU RAM [28]. Hence, it is an ideal tool with inexpensive GPUs. DW dynamic has been discussed in Chapter 2, based on the LLG equation. In Mumax3, the geometry of the magnetic system is considered into the magnetic simulation by the finite element method. This method discretizes the geometry of the system into small cells. Each cell represents a magnetization which allows the LLG equation to be solved whereby the magnetic field and current for each cell are recalculated after each time step using the finite difference method.

Mumax3 micromagnetic simulation package offers great flexibility in the investigation of magnetization dynamics, both finite difference and finite element methods are approximation methods. Hence, the parameters used for the micromagnetic simulations have to be carefully chosen for the results to have any significance for experimental implementation. For example, the cell size needs to be smaller than the exchange stiffness length of the magnetic material.

Otherwise, any study involving the exchange energy parameter is inaccurate. Furthermore, the discrete cells do not accurately represent a smooth curve, resulting in a staircase-like artifact at the edges. Hence, geometrical shapes with straight lines are generally preferred for micromagnetic simulations, otherwise smaller cell size for curved geometry structures are required which results in longer computation time.

3.7 Approaches to achieve synthetic magnetic texture for controllable pinning

Under the ideal conditions, domain wall propagates without any pinning in the nanowires. While in reality, the pinning is inevitable due to the presence of the defects of the materials and the edge damages of the devices. Those pinning determine the critical magnetic field or current to drive domain wall motion. Besides, it is necessary to introduce the extrinsic pinning to control domain wall propagation and to stabilize the stored information [29-33].

In this thesis, I proposed an idea that the energy barriers can be created by introducing a synthetic magnetically textured structure using a local modification of the magnetic properties. To modify the composition, three approaches have been investigated, which will be discussed one by one.

3.7.1 Thermal annealing

Thermal annealing is useful to relieve the intrinsic stress, which helps to improve the crystal structures and surface roughness in materials. And it also can cause the diffusion between the adjacent layers. In this thesis, thermal annealing was introduced to control the diffusion of elements from the top layer to the bottom

layer. At a high-temperature, the atoms gain thermal energy and move randomly with a longer diffusion length. In a system, comprised of two layers in the material, diffusion of atoms happens from one layer into the other. The diffusion strength is dependent on the differential concentrations of the two layers. In this study, non-magnetic metals were used to diffuse into the $\text{Ni}_{80}\text{Fe}_{20}$ layer after high-temperature annealing to modify the magnetic properties of $\text{Ni}_{80}\text{Fe}_{20}$. In order to understand the annealing effect, the samples were heated up to temperatures, T_{an} , from 100 to 400 °C in the vacuum chamber for avoiding contamination from other gases.

3.7.2 Ion implantation

Ion implantation refers to an experimental procedure in which the target ions were implanted into a lattice using ion beams of specific energy [34-36]. Ion implantation provides a method to dope the desired ion species into a host material. As a result, a change of a crystal structure or composition of the material can be achieved. In the implantation tool, the ion source passes through a mass-analyzer, made of a series of electrostatic and magnetic lenses. The target concentration and implantation depth can be controlled independently. The depth of implantation is mainly determined by the energy of target ions or acceleration voltages. The target concentration is contorted by the fluence (ions/cm²) and the implantation time.

When the ions are implanted into a material, the ions come to a stop due to two mechanisms, classified as nuclear stopping and electronic stopping. In the case of nuclear stopping, the ions undergo elastic collisions with the atomic nuclei of the host lattice atoms and lose energy. There exists a transfer of energy

from the ions to the atoms in the host lattice, resulting in further dissipation towards consecutive recoil atoms. In this process, the target ions are scattered significantly, which causes damage to the crystal structure of the host. In the case of electronic stopping, the motion of ions is slowed down by the charge of the electron cloud. The energy transfer is minimal, and the structural damage is negligible for the host material. Ion implantation is treated as a low-temperature process to tune the properties of the host material and does not involve physical milling.

The calculation for the mixing when some ions implanted into the magnetic nanowire is performed by using the SRIM program (The Stopping and Range of Ions in Matter), which is a free software developed by Ziegler. SRIM was used to calculate the mixing of B⁺ implanted into Co/Pd nanowires in this thesis [37]. The accurate calculation is able to be made highly efficient for complex targets with different energies and a various number of ions by the use of statistical algorithms [38, 39]. Depending on the required depth of ion implantation, the accurate calculation needs to set the parameters, including the ions species to be implanted, the ion energy, the angle of implantation and the number of the ions. The final distribution of the ions and all associated kinetic phenomena, namely target damage, sputtering, ionization, and phonon production can be obtained from the calculations [40]. The ion profile of Boron implanted into the Co/Pd multilayer with a Ta capping layer at energy 10 keV with 5×10^{15} ions at zero degrees is shown in Figure 3.11. From Figure 3.11, one can see the implantation profile is Gaussian distribution, and an implantation peak is displayed in the middle of the multilayer layer. The theoretical approaches give a basic

understanding of ion implantation and help in calculating the target ion energies to achieve the required implantation depth.

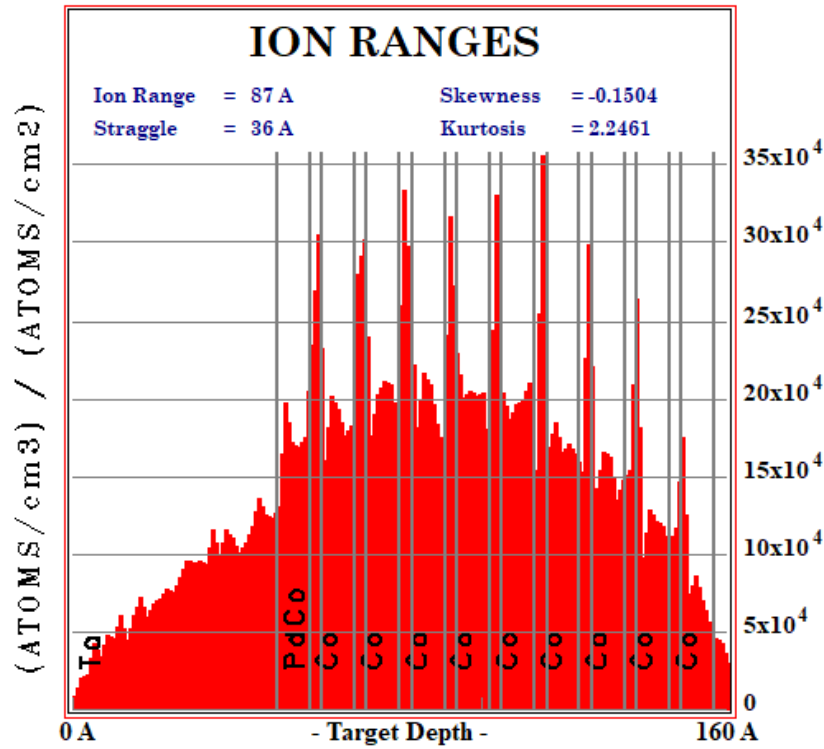


Figure 3.11. Ion profile of Boron implanted in the Co/Pd multilayer structure with the Ta capping layer.

3.7.3 Forming a tilted magnetization by using exchange coupling

It is well known that a stable state in magnetic systems is achieved by minimizing the magnetic energies. In a ferromagnetic material with PMA, the magnetization points in the out-of-plane direction, and the neighboring spins are parallel due to the exchange coupling constant to be $J > 0$ [41]. When a layer with no out-of-plane anisotropy is deposited, the magnetization lies in the in-plane direction, thus it shows the in-plane magnetic anisotropy (IMA). When the two layers are deposited next to each other, the magnetizations at the interface between them will change the directions due to the exchange interaction. The magnetization of the PMA layer will not perfectly point to the normal of the plane [42]. In chapter

6, I have investigated the effect on canting angles of the tilted magnetization for domain wall pinning in the PMA nanowire, by depositing a layer of the IMA material.

3.8 Conclusion

In this chapter, the details of various thin film depositions and sample characterization techniques have been presented. Various analytical techniques and micromagnetic simulation techniques have been described. The device fabrication techniques have also been described. Three approaches to form a synthetic magnetic texture for domain wall pinning have been discussed one by one.

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Chapter 4 Domain wall pinning by elemental diffusion

4.1 Introduction

The read-write operation in a domain wall device strongly depends on a precise position control, a clear boundary and specific chirality of a magnetic domain. In the case of a nanowire without artificial pinning sites, pinning happens randomly due to defects or changes in compositions. This randomness is not desired as one wants to control the position precisely. A controlled domain wall motion can be obtained by introducing synthetic pinning sites, which are fabricated by complicated lithography processes. One way of achieving this is by introducing notches in the ferromagnetic nanowire as shown in Figure 4.1(a). Figure 4.1(b) displays the top view of the domain wall device with notches pinning site. This technique has been widely studied by many groups [1-7]. With this technique, however, the pinning effect could be non-uniform due to the shape and size variations, indicating that better solutions are desired. Another problem of notches in the nanowire is the fact that they cause a reduction of the lateral wire dimensions, leading to increased local heating at the notch position. In addition, if the notches are made using the lowest limits of the lithography technique (the notch with a width of F), the width of domain wall nanowires need to be at least two or three times bigger than the notch width, as illustrated in Figure 4.1(b). Hence, it poses a limitation in the areal density of nanowires. Therefore, better techniques to pin the domain walls in narrow nanowires is desired [1, 5, 7, 8].

In this chapter, we study an efficient alternative method to precisely pin domain wall positions by local modification of the magnetic properties of the nanowire (Figure 4.1(c)). The local modification is achieved by heat-induced diffusion of a suitable element from a crossbar configuration into the magnetic

nanowire. The top view of this technology is shown in Figure 4.1(d). We noted that the width of ferromagnetic nanowires could be reduced to the same size (F) to the pinning sites. Firstly, we carried out simulations to investigate if the domain walls could be stabilized by local compositional variation. Then, we carried out magnetic property measurements at the thin film level to verify the effectiveness of tuning the properties of Permalloy ($\text{Ni}_{80}\text{Fe}_{20}$) by using metal diffusion. Ultimately, $\text{Ni}_{80}\text{Fe}_{20}$ (NiFe) nanowires were fabricated as domain wall devices, which were partially covered with a metallic capping layer to form a crossbar structure. The scanning electron micrograph (SEM) of the device used in this study is shown in Figure 4.1(e). The pinning process was observed by wide-field Kerr images which were performed on a nanowire without or with the metallic capping layer. Magnetoresistance (MR) measurements were carried as a function of an applied field (H), including the recording of minor $MR-H$ loops, to monitor the pinning strength.

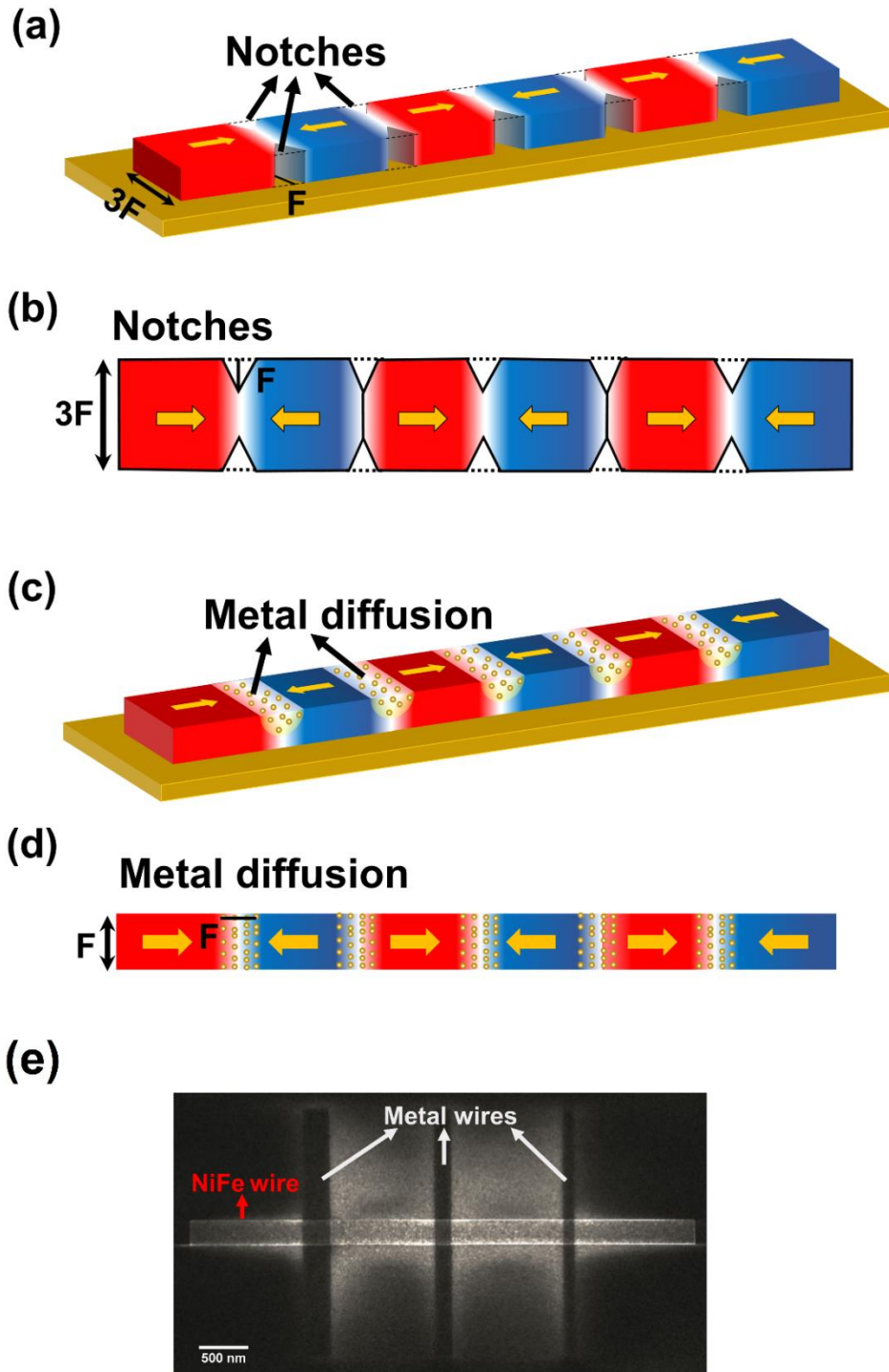


Figure 4.1. (a) Conventional pinning method, which utilizes physical notches to pin domain walls. (b) Top view of domain wall device where pinning sites were created by notches. (c) Proposed non-topographical method of using nanoscale modification of magnetic properties to pin domain walls. (d) Top view of domain wall device where pinning sites were created by compositional modification. (e) An SEM image of a NiFe nanowire with metal wires crossbars to modify the magnetic properties by metal diffusion.

4.2 Micromagnetic simulations

In the following section, I start with a theoretical analysis of the pinning probability of modifying the magnetic properties, using micromagnetic simulations [11]. For the micromagnetic simulation studies, the material parameters were chosen based on the literature data of NiFe nanowires (the saturation magnetization $M_s = 800 \text{ emu/cm}^3$, the exchange stiffness $A = 13 \times 10^{-7} \text{ erg/cm}$, the Gilbert damping constant $\alpha = 0.008$, and a zero uniaxial anisotropy energy) [12, 13]. For the magnetically modified region, representing the pinning site, the following parameters were used: $M_s = 400 \text{ emu/cm}^3$, $A = 6.5 \times 10^{-7} \text{ erg/cm}$ and $\alpha = 0.024$ [14-16]. The simulated nanowire has a dimension of $512 \text{ nm} \times 128 \text{ nm} \times 10 \text{ nm}$. And the cell size used in this simulation is $2 \text{ nm} \times 2 \text{ nm} \times 2 \text{ nm}$. Firstly, one transverse domain wall was introduced and then the magnetization was relaxed to form the lowest energy state as the initial state, as shown in Figure 4.2(a). Figures 4.2 (b-e) show the snapshot images of the propagating wall in the nanowire without the pinning site by sending current at a density $J = 1.9 \times 10^{12} \text{ A/m}^2$, which is driven by spin-transfer torque. After 3 ns, the domain wall has moved to the right and then disappeared at the right side of the nanowire. When the NiFe nanowire has a modified region with a length of 100 nm, as shown in Figure 4.2(f), the domain wall gets pinned at the modified region, as shown in Figures 4.2 (g-j). In less than 3 ns, the domain wall has moved to the modified region, driven by a DC with a density of $J = 1.9 \times 10^{12} \text{ A/m}^2$. Even after 10 ns, the wall remains pinned at the modified region. These simulation results show that the domain walls can be effectively pinned at compositionally modified regions, acting as pinning centres, which indicates that

the proposed scheme has the potential to pin domain walls in a controllable manner.

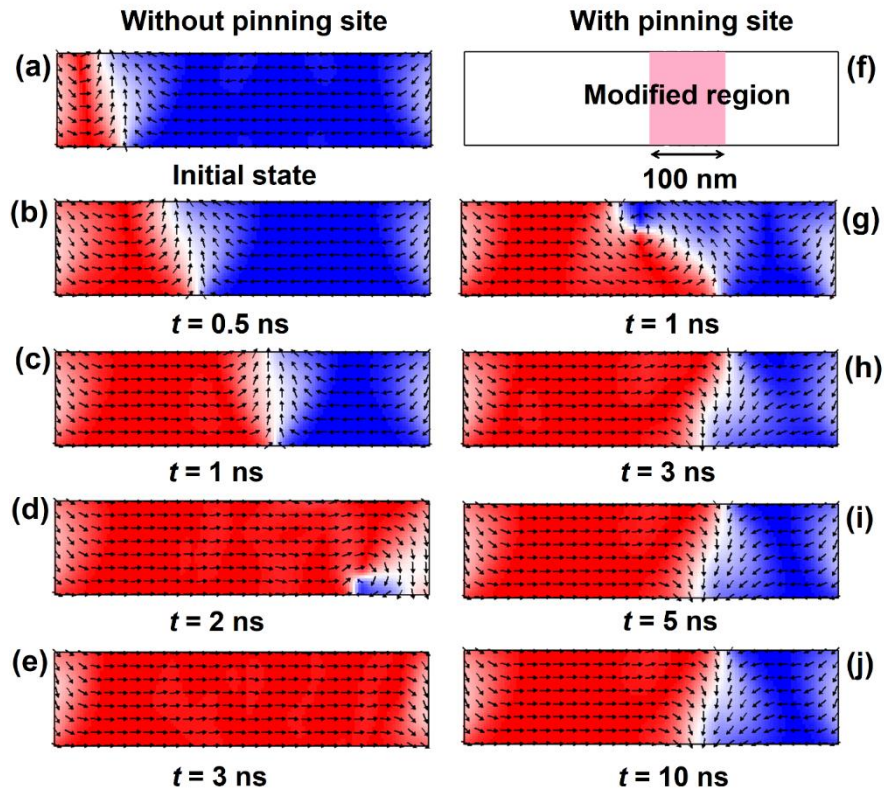


Figure 4.2. Simulated domain wall propagation along the (a-e) homogeneous and (f-j) locally modified nanowire with a length of 512 nm under the charge current applied with initial wall position (a). (f) The NiFe nanowire has a modified region with a length of 100 nm.

4.3 Tuning the properties of $\text{Ni}_{80}\text{Fe}_{20}$ thin film using metal diffusion

For carrying out the experimental investigations, $\text{Ni}_{80}\text{Fe}_{20}$ (NiFe) thin films with a thickness of 10 nm were sputtered using the AJA magnetron sputtering system and were capped with different metallic layers. The film stacks of the type are Si substrate/ SiO_2 / $\text{Ni}_{80}\text{Fe}_{20}$ (10 nm)/X (5 nm). Here, X refers to the metallic capping layers such as Ta, Cr, Cu, and Ru, which are commonly used in the memory industry. A $\text{Ni}_{80}\text{Fe}_{20}$ film without a capping layer was also prepared as a reference. After deposition, $\text{Ni}_{80}\text{Fe}_{20}$ films with capping layers were annealed for an hour at

different annealing temperatures, T_{an} , from 100 to 400 °C in vacuum to induce metal diffused into $\text{Ni}_{80}\text{Fe}_{20}$.

Depositing a pure $\text{Ni}_{80}\text{Fe}_{20}$ thin film with the excellent soft magnetic properties is the key initial step for further investigation. It was found that the properties of NiFe are more sensitive to the deposition conditions. Figures 4.3 (a) and (b) show the normalized magnetic hysteresis loops (M - H) of a pure $\text{Ni}_{80}\text{Fe}_{20}$ thin film along in-plane (IP) and out-of-plane (OOP) directions, which were grown using different sputtering systems. Figure 4.3(a) shows the properties of the film grown by an unclean sputter system at high Ar pressure of 10 mTorr. The thin film sample did not exhibit a clear in-plane magnetic anisotropy. In order to find out the reasons, the elemental analysis measurement was carried out by using Energy-dispersive X-ray spectroscopy (EDX). The spectrum was shown in Figure 4.4(b). The Cr and Cu peaks can be easily observed besides Ni and Fe. Since the deposition chamber was contaminated with the other materials, I switched to a cleaner AJA system and optimized various parameters such as deposition pressure, power, *etc.* The properties of the film, deposited at 2 mTorr, are shown in Figure 4.3(b). The M - H loop exhibits well defined in-plane anisotropy of $\text{Ni}_{80}\text{Fe}_{20}$ film, with a low coercivity of about 15 Oe. This experimental condition was used for the rest of the samples.

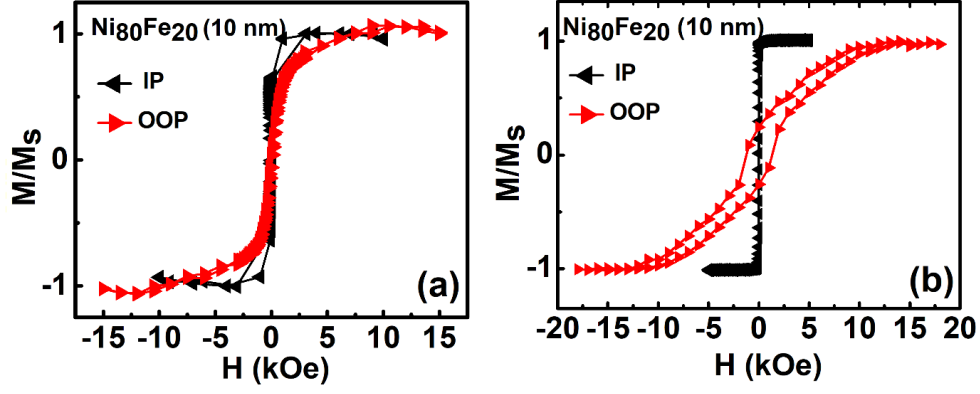


Figure 4.3. The normalized hysteresis loops of $\text{Ni}_{80}\text{Fe}_{20}$ thin films along in-plane (IP) direction and out-of-plane (OOP) directions, which were grown by (a) an unclean sputtering system and (b) the AJA sputtering system.

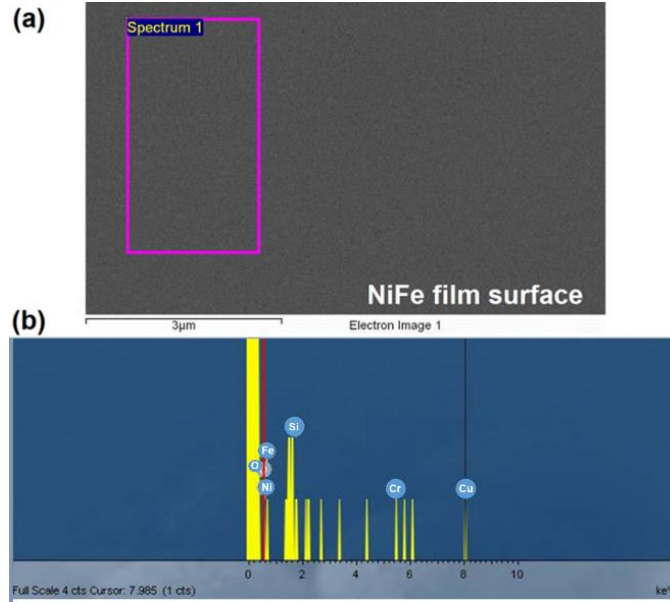


Figure 4.4. (a) The SEM image and (b) EDX analysis spectrum of NiFe thin films grown by the unclean sputtering system.

Broadband ferromagnetic resonance (FMR) measurements were carried out to study the magnetodynamics properties of films quantitatively. All the measurements were performed at room temperature. The full spectrum for $\text{Ni}_{80}\text{Fe}_{20}$ is presented in Figure 4.5(a), while the external field was swept parallel to the film plane at a rate of about of -2 Oe/s from 50 to 5500 Oe at several

selected frequencies. The resonance field (H_{res}) and inhomogeneities broadening (ΔH) increase for the microwave frequency from 3 GHz to 24 GHz. The spectrum can be accurately fit with a sum of symmetrical and antisymmetrical Lorentzian function [13]:

$$y = S \frac{\Delta H^2}{(H - H_{\text{res}})^2 + \Delta H^2} + A \frac{\Delta H (H - H_{\text{res}})}{(H - H_{\text{res}})^2 + \Delta H^2} \quad (4.1)$$

where S , A , ΔH , and H_{res} are the magnitude of the symmetric component, the magnitude of the antisymmetric component, the half linewidths, and the FMR resonance field. The well-fitting is shown in Figure 4.5(b) at the frequency of 12 GHz, where the half linewidths ΔH , and the FMR resonance field H_{res} are marked.

The field dependence of the extracted FMR resonance frequencies are shown in Figure 4.5(c), this experimental data can be fitted well to the Kittel equation [13]:

$$f = \frac{\gamma}{2\pi} [(H_{\text{res}} + H_k) \times (H_{\text{res}} + H_k + 4\pi M_s)]^{1/2} \quad (4.2)$$

where γ is the gyromagnetic ratio of a magnetic layer, H_k is the in-plane anisotropy field and H_{res} is the resonance field. Figure 4.5(e) shows the fitting parameters. From this fitting, the saturation magnetization M_s 726 ± 2 emu/cm³ of pure Ni₈₀Fe₂₀ is extracted and in-plane anisotropy field H_k is obtained to be very small and negligible as expected. The gyromagnetic value $\gamma/2\pi$ is about 2.97 MHz/Oe.

The saturation magnetization M_s of pure Ni₈₀Fe₂₀ was also measured by using VSM, which is 707 emu/cm³. There is a slight difference between the magnetization values due to the different measurement techniques. In the literature reported results, the M_s of Ni₈₀Fe₂₀ film is in the range of 600 to 1000 emu/cm³ [13, 17, 18], which is comparable with our measured results.

Figure 4.5(d) shows the frequency dependent of FMR spectral half linewidths ΔH for pure $\text{Ni}_{80}\text{Fe}_{20}$ film. The linear relationship can be fitted well using [13]:

$$\Delta H = \Delta H_0 + \frac{4\pi\alpha}{\gamma} f \quad (4.3)$$

where ΔH_0 is the inhomogeneous broadening, and α is the Gilbert damping constant, demonstrating the dynamic property of $\text{Ni}_{80}\text{Fe}_{20}$. Figure 4.5(f) shows the fitting parameters. The Gilbert damping constant (α) was found to be small, about $(77 \pm 2) \times 10^{-4}$, and was consistent with the values reported in the literature for $\text{Ni}_{80}\text{Fe}_{20}$ films [13, 19, 20].

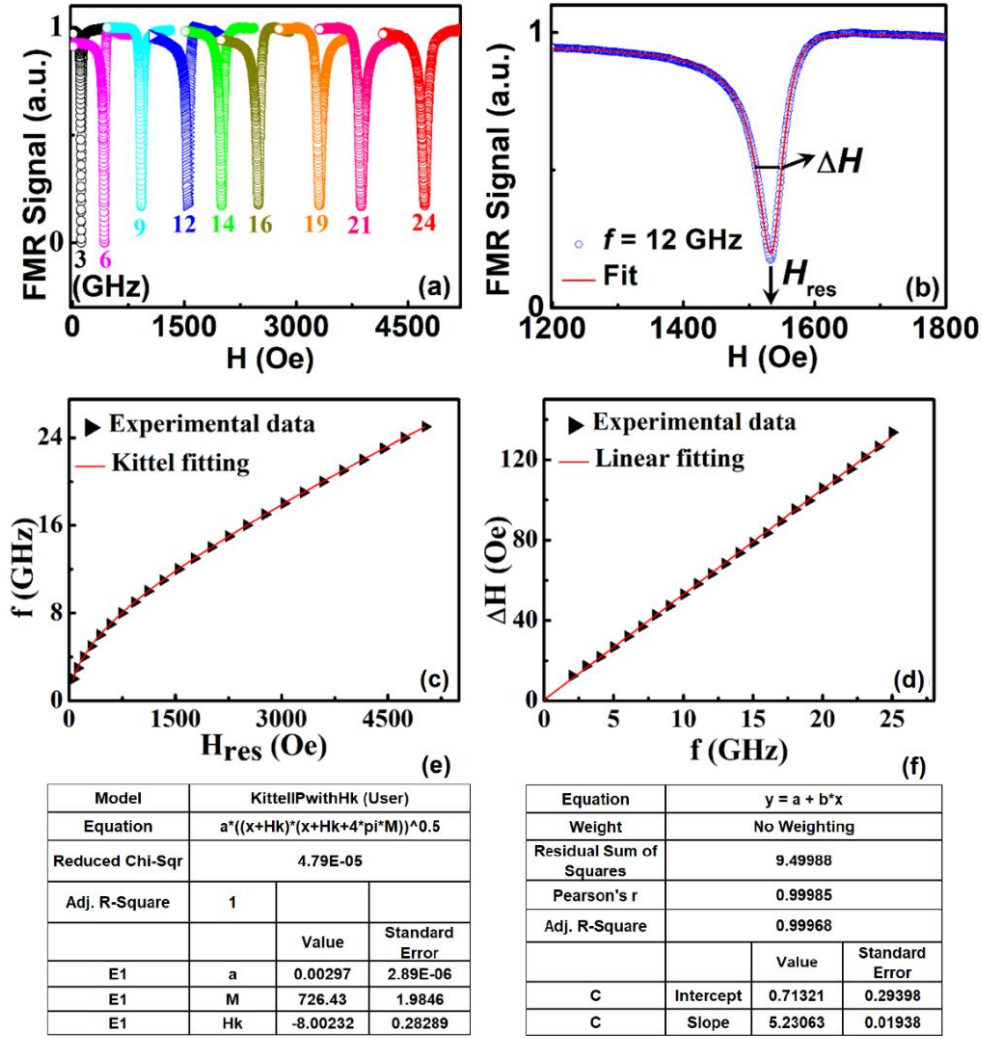


Figure 4.5. (a) The FMR signal for a $\text{Ni}_{80}\text{Fe}_{20}$ thin film with a thickness of 10 nm with the magnetic field applied parallel to the sample plane. (b) The spectrum fitted with a sum of symmetrical and antisymmetrical Lorentzian function at the frequency of 12 GHz. (c) Resonance frequency as a function of the resonance field and (d) FMR half linewidths as a function of the resonance frequency. (e) and (f) are the fitting parameters for (c) Kittel fitting and (d) linear fitting respectively.

Standard properties shown in Figure 4.5, verified the quality of the prepared $\text{Ni}_{80}\text{Fe}_{20}$ films. As a next step, the static and dynamic properties of $\text{Ni}_{80}\text{Fe}_{20}$ film samples without capping and with different metallic capping layers annealed at different T_{an} from 100 to 400 °C were studied by FMR spectral signal. Figures 4.6 (a) and (b) show the resonance frequency vs. the applied field and the

half linewidths vs. the resonance frequency of $\text{Ni}_{80}\text{Fe}_{20}$ film without capping layer annealed at different T_{an} . I fitted the Kittel equation to the data as shown in Figure 4.6(a) and extracted M_s of $\text{Ni}_{80}\text{Fe}_{20}$ film samples after annealing. M_s did not show much change after annealing. Even at higher $T_{an} = 400\text{ }^\circ\text{C}$, the value of M_s was around $704 \pm 10\text{ emu/cm}^3$. But the damping constant α , obtained by fitting the linewidth with the resonance frequency in Figure 4.6(b), increased after annealing, going up to $(122 \pm 2) \times 10^{-4}$ at $T_{an} = 400\text{ }^\circ\text{C}$ [17, 18]. The increase in damping is probably due to the diffusion of oxygen into NiFe from substrate Si/SiO₂ arising from an increase in the magnon scattering [13]. In order to understand the oxidized of NiFe thin film without capping layer after high-temperature annealing, I carried out the XPS measurement. The XPS spectrum showed that around 1~2 nm thick of NiFe adjacent to SiO₂/Si substrate was oxidized after 400 °C annealed. About 10% to 20% of the NiFe film (10 nm in total thickness) was oxidized, which is consistent with the ~8% decrease of M_s at 400 °C. The damping constant showed a remarkably high value of $(185 \pm 3) \times 10^{-4}$ at $T_{an} = 500\text{ }^\circ\text{C}$. It is well known that the depinning field increases with the damping constant [21-23]. As a result, a higher current is required to drive domain wall motion, which is not desired. Therefore, the maximum annealing temperature was fixed to be 400 °C in the subsequent experiments.

The magnetic properties of NiFe with the different capping layers, after annealed at different temperatures T_{an} , are characterized by measuring FMR spectral signal. Figures 4.6 (c) and (d) show the resonance frequency vs. the applied field, and the half linewidth vs. the resonance frequency of a $\text{Ni}_{80}\text{Fe}_{20}$ film with a Cr capping layer (NiFe/Cr) annealed at different T_{an} . The curve, shown in Figure 4.6(c), also can be fitted using the Kittel equation. After

annealing at 400 °C, the resonance field for the NiFe/Cr sample was found to be larger in comparison to that of reference NiFe sample without a capping layer. The half linewidth also became broader. The broader resonance field is due to the Cr diffusion, forming a non-uniform distribution along the NiFe film normal direction [13-15, 20-22]. Each concentration of the Ni₈₀Fe₂₀ films has its own peak at a specific magnetic field. In a system with a gradual change in concentration, the intrinsic FMR signals overlap with each other, resulting in the broadening distribution of the peak as well as the linewidth. In addition, different annealing temperatures lead to different degree of Cr diffusion. The broadest linewidth was observed in the film with the highest annealing temperature ($T_{an} = 400$ °C).

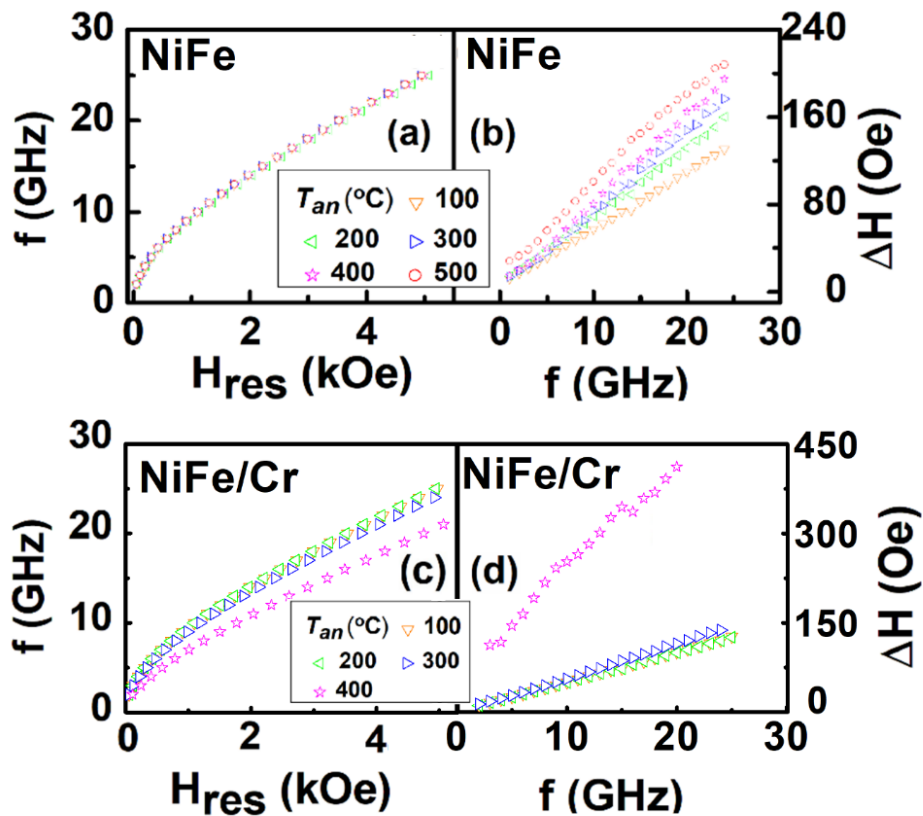


Figure 4.6. Ni₈₀Fe₂₀ without and with a Cr capping layer at different annealing temperatures (T_{an}), (a, c) resonance frequency as a function of a resonance field and (b, d) FMR half linewidths as a function of the resonance frequency.

The magnetic properties of NiFe without and with different capping layers, annealed at different temperatures T_{an} , are summarized in Figures 4.7 (a-d). These parameters were obtained from the dynamic measurements. Saturation magnetization M_s decreases with increasing T_{an} from 100 to 400 °C for all metallic capping layers as shown in Figure 4.7(a). The decrease has been the most significant in the case of Cr, where the effective magnetization M_s drops down to 350 ± 8 emu/cm³ at 400 °C. In the case of the Cu capping layer, M_s decreases from 720 ± 3 emu/cm³ to 624 ± 5 emu/cm³ with increasing T_{an} to 400 °C, while for Ta and Ru, the magnetization decreases at a much slower rate. The saturation field along the perpendicular direction is also found to increase steadily with T_{an} , beyond 200 °C.

Figures 4.7 (b-d) show the magneto-dynamic properties of pure Ni₈₀Fe₂₀ films and Ni₈₀Fe₂₀/X films, such as gyromagnetic ratio (Figure 4.7(b)), damping constant (Figure 4.7(c)) and inhomogeneous broadening (Figure 4.7(d)) at different T_{an} . The gyromagnetic ratio ($\gamma/2\pi$) remains at about 2.96 MHz/Oe for most of the samples. For the Cu and Cr capping layers, it increases to 3.01 and 3.04 MHz/Oe at $T_{an} = 400$ °C respectively. The damping constant of Ni₈₀Fe₂₀ increases with T_{an} for all capping layers. Among all the elements investigated, Cr shows the strongest effect and the Ta capping layer had the least effect. The damping constant of Ni₈₀Fe₂₀ is not influenced much for $T_{an} < 300$ °C. However, for Cr, an increase in α by 246% at $T_{an} = 400$ °C is observed. For the Cu capping layer, the damping constant of the annealed films increased to $(1198 \pm 3) \times 10^{-5}$ at $T_{an} = 400$ °C. The inhomogeneous broadening value is small for all capping layers at room temperature, and it shows the increasing trend for the Cr and Cu capping layers, increasing to 66 Oe and 28 Oe with T_{an} , respectively. However,

it does not show much change for the Ta and Ru capping layers. Anisotropy field was minimal as expected for all $\text{Ni}_{80}\text{Fe}_{20}/\text{X}$ films, and it does not show significant changes after annealing.

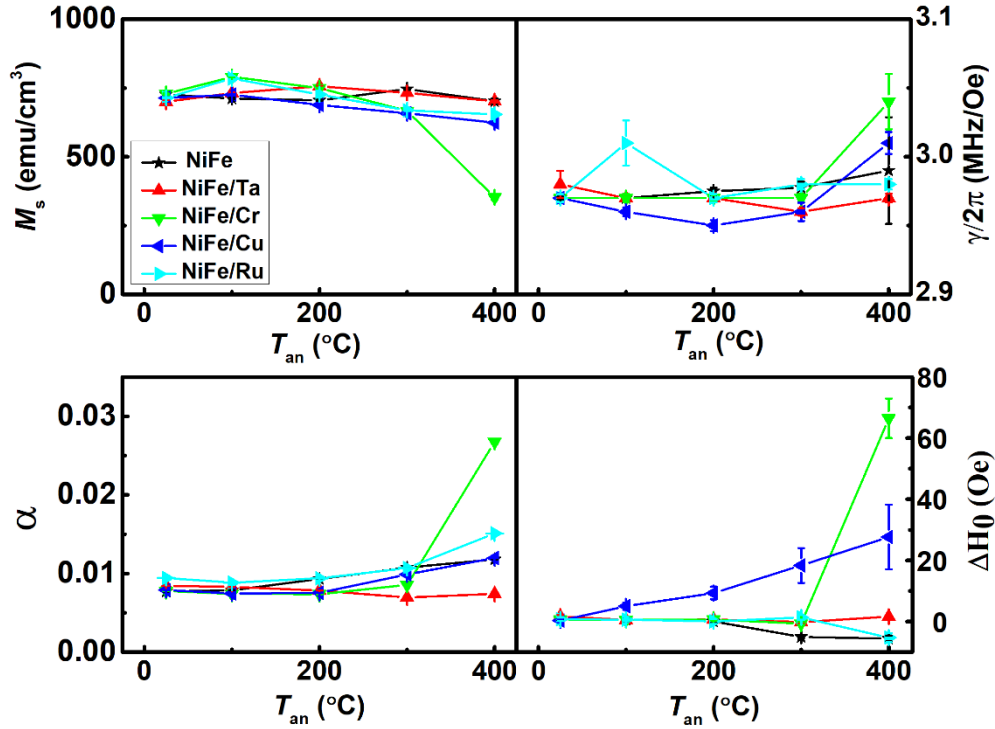


Figure 4.7. Magnetic properties of $\text{Ni}_{80}\text{Fe}_{20}$ without capping layer and with different capping layer $\text{Ni}_{80}\text{Fe}_{20}/\text{X}$ ($\text{X} = \text{Ta}, \text{Cr}, \text{Cu}, \text{Ru}$) at different annealing temperatures (T_{an}). (a) The effective saturation magnetization, (b) Gyromagnetic ratio, (c) the damping constant and (d) the inhomogeneous broadening are estimated.

Anisotropic magnetoresistance (AMR) was measured to further understand the magnetic texture. A current, I_{dc} of 80 μA , was applied by sweeping a magnetic field perpendicular to the current direction and measured the voltage (resistance) simultaneously, as shown in Figure 4.8. The inset of Figure 4.8 illustrates the typical magnetoresistance (MR) of a pure $\text{Ni}_{80}\text{Fe}_{20}$ thin film sample with a thickness of 10 nm, which depends on the applied magnetic field. The current direction is orthogonal to the applied magnetic field direction. When the magnetization was saturated at the high magnetic field, which is along the transverse direction to the current flow, the MR is low. Whereas, the MR is high

when the magnetization direction is parallel/antiparallel to the current direction, corresponding an AMR ratio of $\sim 1\%$ for pure $\text{Ni}_{80}\text{Fe}_{20}$ sample. For $\text{Ni}_{80}\text{Fe}_{20}$ samples with a metallic capping layer, AMR ratios reduce with increasing T_{an} , particularly in the case with the Cr capping layer, which is attributed to a decrease in magnetization [13, 24-26].

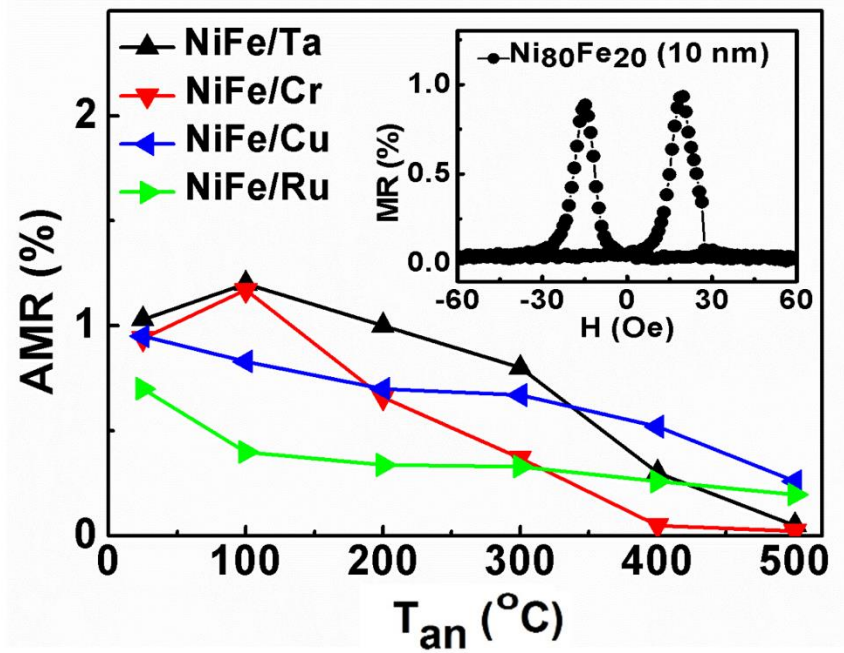


Figure 4.8. AMR ratios for $\text{Ni}_{80}\text{Fe}_{20}$ with different capping layers at different T_{an} . Inset of the figure shows the magnetoresistance for pure $\text{Ni}_{80}\text{Fe}_{20}$ film with applied magnetic field sweep.

The investigations on the thin film level indicate that the metal diffusion causes changes in the properties of a $\text{Ni}_{80}\text{Fe}_{20}$ film and among several elements, Cr has the strongest effect both in static (M_s) and dynamic properties (α). It is worthwhile to compare the properties with those reported in the literature [20-22, 27, 28]. Ruiz-Calaforra *et al.* have studied the effect of metal capping layers adjacent to the NiFe layer. Some of their samples (9-11 nm thickness of NiFe) are similar to our samples at room temperature. In their samples, the presence of a Pt layer enhances the damping constant significantly due to spin pumping. In

comparison, Ru enhances the damping constant to a larger extent than Cr. A similar trend in this study was also observed at room temperature. However, absolute values and the changes in the damping constant are smaller in our samples due to less noise in our FMR data and probably the nature of the deposition process of our samples. In our samples, Cr capping layers did not cause a significant change in the damping constant at room temperature. However, it causes a substantial increase in damping constant after annealing. From this comparison, the increase in damping constant is mainly attributed to the annealing-induced mixing.

Faulkner *et al.* have studied the influence of Cr doping by ion implantation in NiFe. They also observe a similar increase in damping constant α , a decrease in the saturation magnetization M_s and the anisotropy energy K_u . It is well known that the addition of Cr causes a decrease of magnetization in ferromagnetic systems. Cr spins are believed to antiferromagnetically couple to the spins of host atoms and as a result, my samples also show a decrease of M_s .

In order to further understand the diffusion of Cr in a Ni₈₀Fe₂₀ film, X-ray photoelectron spectroscopy (XPS) was carried out. Figures 4.9 (a-e) show the Cr diffusion effect with the different T_{an} . It can be seen that the Cr diffusion is less for $T_{an} < 300$ °C, but shoots up for $T_{an} > 300$ °C. Figure 4.9(f) summarizes the concentration of Cr, closer to the bottom of the Ni₈₀Fe₂₀ layer, for different T_{an} . The changes in the saturation magnetization M_s , the damping constant α , and the other properties can be explained to be arising as the result of Cr diffusion, which is significantly stronger at higher temperatures.

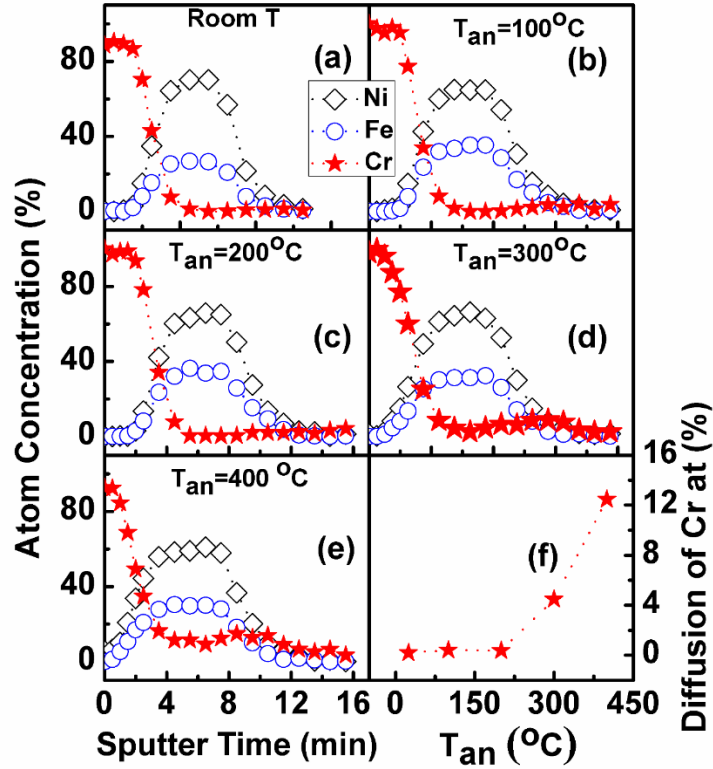


Figure 4.9. (a-e) XPS analysis for Cr diffusion effect vs. T_{an} from 100 to 400 °C and (f) summarizes the concentration of Cr diffusion into the $Ni_{80}Fe_{20}$ layer with the T_{an} taking the diffusion value of Cr at 7 minutes sputter etching time.

From the XPS results, as shown in Figure 4.9, it can be noted that the concentration of Cr is higher at the top layer and that it gradually decreases along the vertical direction from the top to bottom. The distribution of the concentration along the depth direction is non-uniform. To further confirm whether the elements diffuse into NiFe uniformly along the lateral direction and form solid solutions or clusters, I also carried out elemental analysis of the interfaces by using transmission electron microscopy (TEM) equipment. Figure 4.10 shows the TEM cross-section image of NiFe with the Cr capping layer annealed at 400 °C and the corresponding energy dispersive X-ray (EDX) spectroscopy analysis images. The results indicate that the diffusion of Cr into the NiFe layer by annealing is uniform along the lateral direction and hence a solid solution.

Different elements may behave differently, and their behavior may, to some extent, be predicted from the phase diagrams [25, 29, 30]. It is well known that Cu does not mix well with Fe and hence, the different results should be expected. Figure 4.11 shows the EDX analysis of the elemental concentration of Si (substrate), Ni, Fe, and Cu. It can be seen that the Cu atoms do not diffuse so uniformly like Cr atoms.

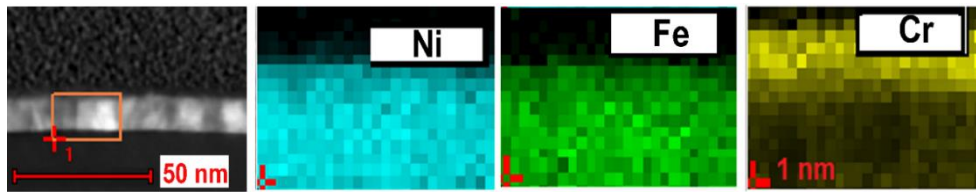


Figure 4.10. TEM cross configuration and the EDX analysis pattern result in the square area. From left to right, the images show the Ni, Fe, and Cr the quantitative analysis.

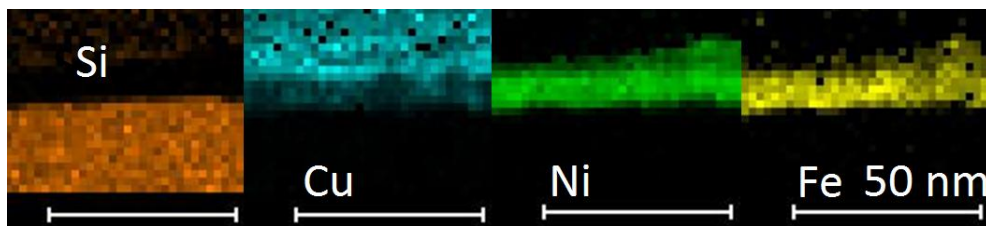


Figure 4.11. Elemental map of Si, Cu, Ni and Fe in the Si substrate/NiFe/Cu structures.

Diffusion of Cr proves to be more effective in comparison to the other elements such as Ta, Cu, and Ru in changing the magnetic and magneto-dynamic properties. In order to prove that the local modification is effective in stabilizing the domain walls, NiFe wire devices with Cr crossbars have been investigated here after.

4.4 Domain wall pinning effect for a $\text{Ni}_{80}\text{Fe}_{20}$ device with Cr crossbar pinning sites

NiFe magnetic wires (1 μm wide and 100 μm long) were fabricated by electron-beam lithography (EBL), followed by argon ion milling. During the development process, the NiFe films deposited directly on the Si substrate were formed to be peeled off. Therefore, for the fabrication of these devices, a Ta seed layer was used to prevent the diffusion of oxygen from the Si substrate and to improve the adhesion of NiFe with the substrate. After the NiFe wires were fabricated, Cr cross lines were fabricated by EBL, followed by 5 nm thick Cr deposition by using the AJA magnetron sputtering. The NiFe magnetic wire was annealed in a vacuum chamber at 400 $^{\circ}\text{C}$ atmospheres before all the electrode fabrication. A top view micrograph of the magnetic wire is shown in Figure 4.12(b). The inset in Figure 4.12(b) is the line-scan of an Atomic Force Microscopy (AFM) image of the NiFe wire with the Cr wires, and the height profile shows that the Cr is on top of the NiFe wire. The NiFe magnetic wire without the Cr pinning cross lines also was prepared as a reference sample using EBL, as shown in Figure 4.12(a).

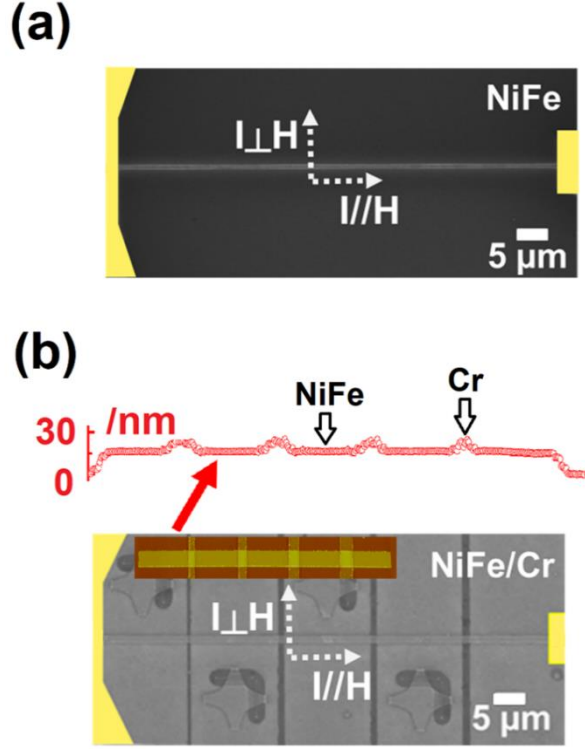


Figure 4.12. Schematic representation of the NiFe magnetic wire (a) without the Cr crossbars and (b) with Cr crossbars. The insert figure is the AFM image of the NiFe wire with the Cr crossbars.

I. Optical Measurements

In order to study the pinning effect introduced by Cr-modified crossbars, firstly the optical measurement was carried out by performing Kerr imaging on the microwires at different applied magnetic fields. Figure 4.13(a) shows images which were taken at different fields for wires with a width of 4 μm. It can be noticed that the domain wall propagation is stochastic. The domain wall shows rapid propagation between 13 and 16 Oe. Figure 4.13(b) shows the domain patterns in 1 μm width wires without crossbars and Figure 4.13(c) shows the images for wires with crossbars. It can be noticed from Figure 4.13(b) that the domain wall propagation is random in the case of wires without the Cr crossbars. For samples with the Cr crossbars, the domains appear to nucleate at the pinning

sites and then propagate. This is a typical problem associated with field-dependent domain wall motion investigations where the whole wire is subject to a field. Nevertheless, hysteresis loops of the two wires were measured by extracting the intensity of the wire region and they are shown in Figures 4.13 (d) and (e). In all the measurements shown in Figure 4.13, the field was applied along the length of the wire. A noticeable increase in the coercivity is observed in samples with the Cr crossbars, indicating an additional energy barrier for domain wall propagation. However, no direct evidence is found in Figures 4.13 (b-c) to verify a clear sign for domain wall pinning due to the Cr crossbar.

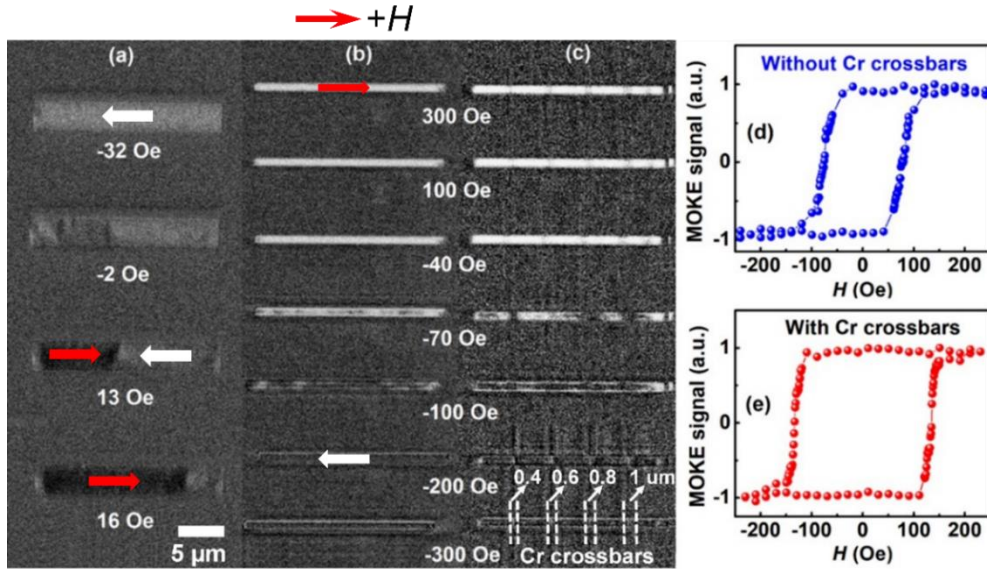


Figure 4.13. High-resolution Kerr images of domain evaluation in an external magnetic field on (a) NiFe wire (4 μm width), 1 μm width (b) without metal wire and (c) with Cr wire imaged with a 100× oil immersion lens. The dashed lines in (c) indicate the Cr crossbars. The red and white arrows correspond to different magnetization orientations along right and left directions, respectively. In-plane magneto-optical loops measured on NiFe wires (d) without capping metal crossbar and (e) with Cr crossbar.

II. Electrical measurements

In order to obtain an insightful understanding of the domain wall motion, the electrical measurements were carried out by detecting the magnetoresistance

(MR) properties of the devices. A DC (I) of 30 μA was passed through the nanowires. At the same time, a nanovoltmeter was used to pick up the voltage. Figure 4.14(a) shows a top view micrograph of the NiFe magnetic wire. Figure 4.14(b) shows the dependence of the MR on the angle θ between the direction of the magnetization and the current orientation along NiFe wire. The red line is a fit to the well-known cosine dependence $R = R_{\perp} + (R_{//} - R_{\perp}) \cos^2 \theta$, where $R_{//}$ and $R_{\perp}$ denote the resistance when the field direction is parallel ($\theta = 0$) and perpendicular ($\theta = \pi/2$) to the current orientation, respectively [25]. The table shows the fitting parameters. The anisotropic magnetoresistance (AMR) is then determined by the relation $\Delta R/R = (R_{//} - R_{\perp})/R_{\perp}$. When the current (I) direction is parallel to the magnetic field, one can observe that the magnetoresistance is small. For $I \perp H$, a change in the resistance value of about 1.5% is observed. Figure 4.14(c) shows the MR change with the sweeping magnetic field from -600 Oe to 600 Oe, which is applied along the NiFe wire and perpendicular to the NiFe wire direction, respectively. The MR loops show typical properties in AMR measurements. These results are typical properties of $\text{Ni}_{80}\text{Fe}_{20}$, and these results validate the fabrication process of the nanowires [13, 31].

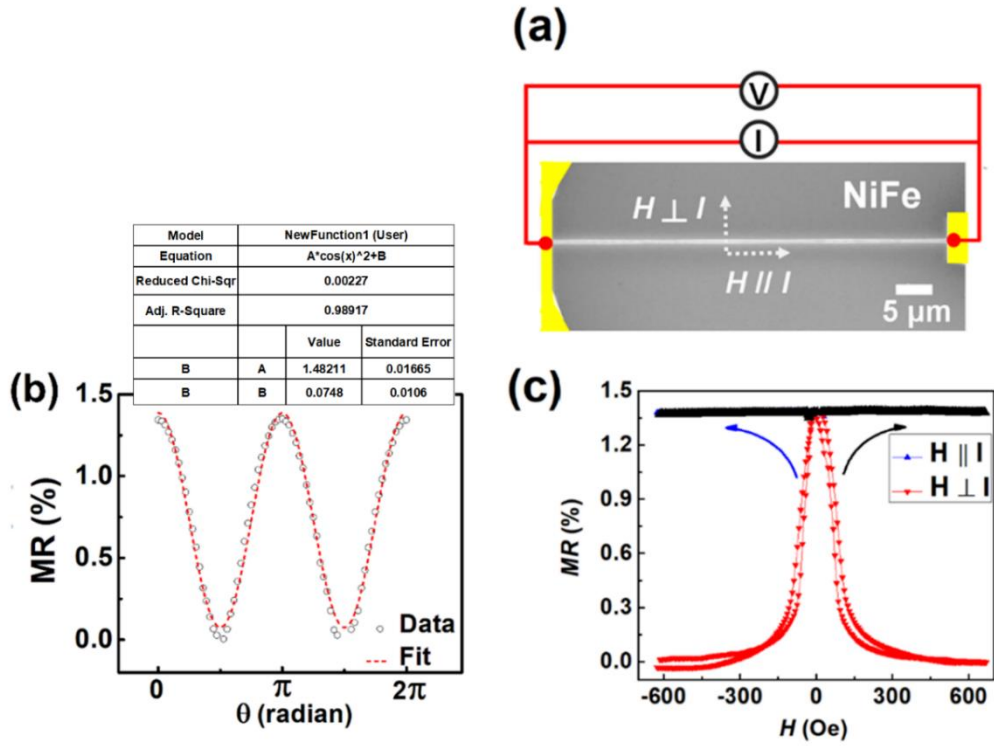


Figure 4.14. (a) Optical image with schematics of the NiFe wires and the electrodes. (b) The MR ratio of the NiFe is a fit for the well-known cosine (θ) dependence. The angle θ is between a current and a magnetic field. (c) The MR ratio of NiFe measured by applying an in-plane magnetic field along ($\theta = 0$) and perpendicular ($\theta = \pi/2$) to the NiFe wire direction.

Figure 4.15(a) shows a top view micrograph of the NiFe magnetic wire with the Cr crossbars, and Figures 4.15 (b) and (c) show the dependence of MR on the angle θ and the MR change with a magnetic field. The behavior of the curve for $H \perp I$ ($\theta = \pi/2$) is similar to that of $\text{Ni}_{80}\text{Fe}_{20}$ without the Cr crossbars. The magnitude of change in $R-H$ is about 0.8%, which is lower than thus indicating the effect of the Cr diffusion. However, we can notice a striking difference in $R-H$ for the case $H \parallel I$ ($\theta = 0$), where a dip at fields of about 100 Oe is seen. This dip is not a common feature of NiFe, and hence one can safely say that this effect is brought out by the Cr crossbars. It is expected that the pinning centers lead to the presence of an energy barrier at every pinning site. If there is

an energy barrier associated with the pinning sites (Cr-diffused regions), then the $MR-H$ curves should exhibit an irreversible behavior. To categorically observe such an irreversible behavior, minor $MR-H$ curves were measured. The magnetic field was swept starting from a positive saturated state (600 Oe) to the negative field (named as reversal field (H_{Re})), such as -50 Oe, -70 Oe, -90 Oe and -110 Oe, and then returned to the saturated state at 600 Oe.

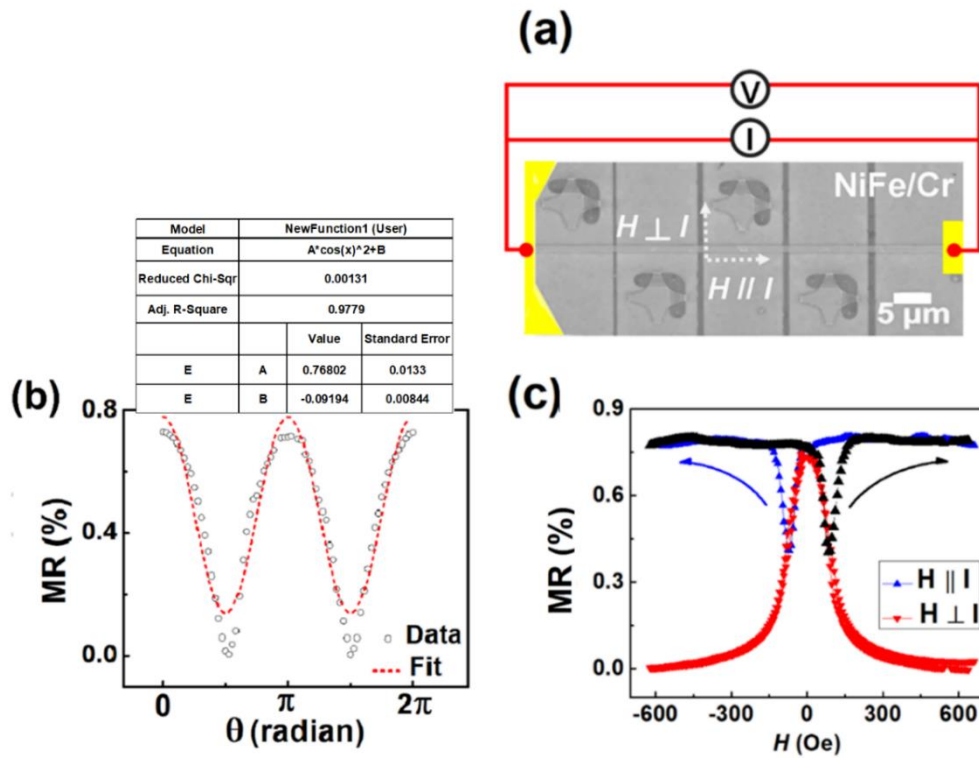


Figure 4.15. (a) Optical image with schematics of the NiFe wires and the electrodes with Cr crossbars. (b) The MR ratio of NiFe is a fit for the well-known cosine (θ) dependence. The angle θ is between a current and a magnetic field. (c) Change of MR in NiFe wires with Cr crossbars by applying an in-plane magnetic field along wire ($\theta = 0$) and perpendicular ($\theta = \pi/2$) to the NiFe wire directions.

Figures 4.16 (a-d) show the reversal curves for different H_{Re} , where a kind of hysteresis in the $MR-H$ curves is observed. Figure 4.16(e) shows the area occupied by the curves, which increased with an increase in H_{Re} . Here, areas were calculated based on the integrated function in the Origin software [32], and the

error bar was obtained by calculating the difference in the occupied area between the experimental raw curve and the smoothed curve. An $MR-H$ curve, similar to Figure 4.15(c) has been reported in the study of Mohanan *et al.* [33] where the resistance of $\text{Ni}_{80}\text{Fe}_{20}$ nanowire drops due to domain wall injected by a junction pad and pinned by notches. The depinning field was about 80 Oe in their study [33]. However, they did not report a minor $MR-H$ loop as carried out in this study. Similar to the area in a $B-H$ loop, the areas of these curves indicate the energy required to overcome the pinning strength of the domain walls [34, 35]. The points marked in the figures correspond to different domain configurations, as shown in the inset. These results indicate that the Cr diffused regions indeed act as pinning centers [36].

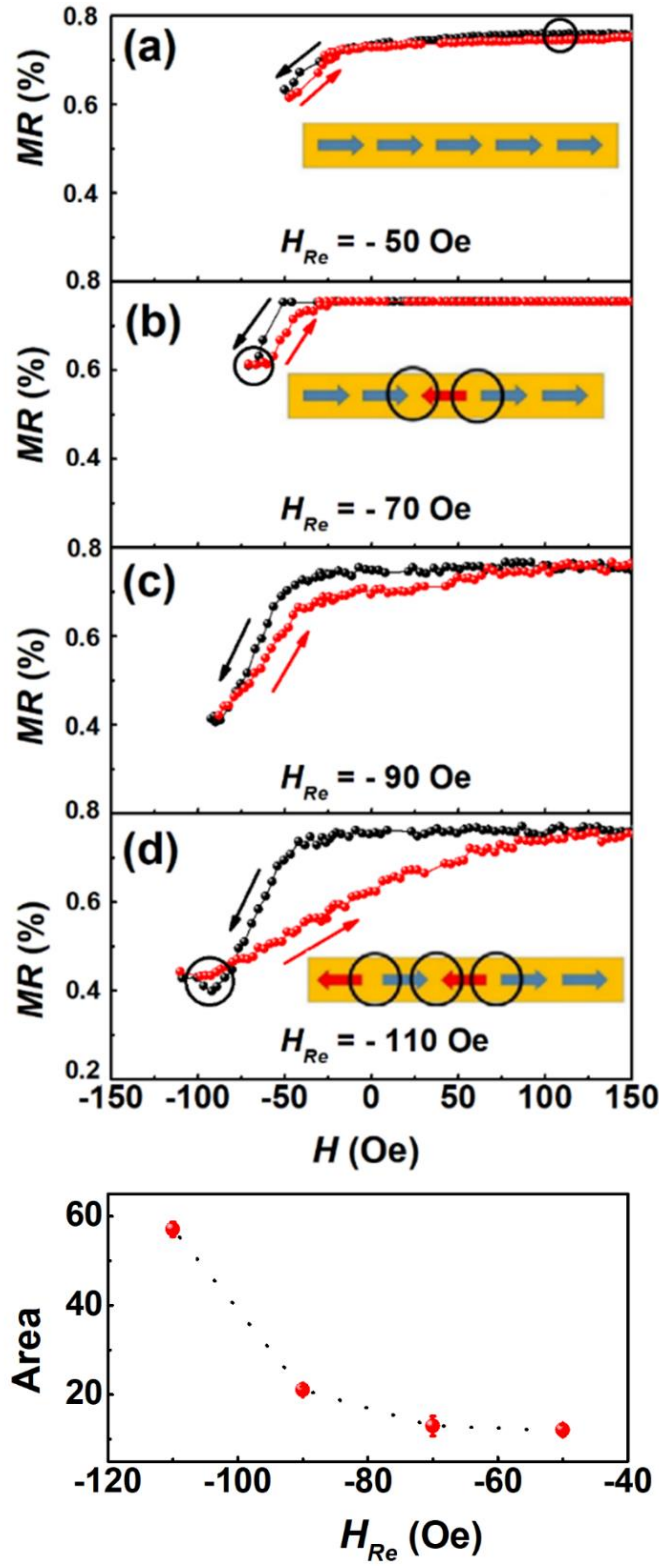


Figure. 4.16. Minor loop measurement of the $MR-H$ curves, starting from a saturation field, to a reversal field H_{Re} and back to saturating field. (a-d) are the minor resistance-field hysteresis loops for different reversal fields, H_{Re} . The inset shows the points marked in the loops that correspond to different domain configurations as illustrated by the arrows. The area occupied by the curves for different reversal fields is plotted in (e).

III. Micromagnetic simulations

The experimental investigations on the pinning effect for $\text{Ni}_{80}\text{Fe}_{20}$ device was discussed so far. In order to further understand the pinning effect with different properties of the pinning sites, I carried out micromagnetic simulations [11]. It can be noticed from the $\text{Ni}_{80}\text{Fe}_{20}$ thin film study that the damping constant (α) increased to 0.024 from 0.008 after the non-magnetic metal Cr diffusion, while the saturation magnetization (M_s) of $\text{Ni}_{80}\text{Fe}_{20}$ decreased by 50%. The parameters used for the modified regions were decided by following this trend, which is summarized in Table 4.1. The exchange stiffness (A) decreases following the changing trend of M_s . M_{s0} , A_0 and α_0 are properties parameters for the unmodified region. The saturation magnetization (M_s) and exchange stiffness (A) decrease as expected by 10%, 20%, 30%, 40% and 50%. The damping constant increased by 10%, 50%, 100%, 150% and 200%. For example, the saturation magnetization (M_s) decreases to 720 from 800 emu/cm^3 and the exchange stiffness (A) decreases to 11.7×10^{-7} from 13×10^{-7} erg/cm by 10%, corresponding to that the damping constant (α) increased to 0.0088 in the modified region, The simulated nanowire has a dimension of 1024 nm $\times$ 128 nm $\times$ 10 nm and the cell size used is 2 nm $\times$ 2 nm $\times$ 2 nm. The length of the pinning region ΔX is varied from 50 to 200 nm. The critical depinning current densities are plotted with α , as shown in Figure 4.17.

	Unmodified region		Properties of modified region				
			Run 1	Run 2	Run 3	Run 4	Run 5
M_{s0} (emu/cm ³)	800	M_s (emu/cm ³)	720	640	560	480	400
A_0 (erg/cm)	13E-7	A/A_0	0.9	0.8	0.7	0.6	0.5
α_0	0.008	α/α_0	1.1	1.5	2	2.5	3

Table 4.1. The parameters used for the modified regions and the unmodified region. The effect of the width of the modified region on the depinning current density was also investigated (see Figure 4.17).

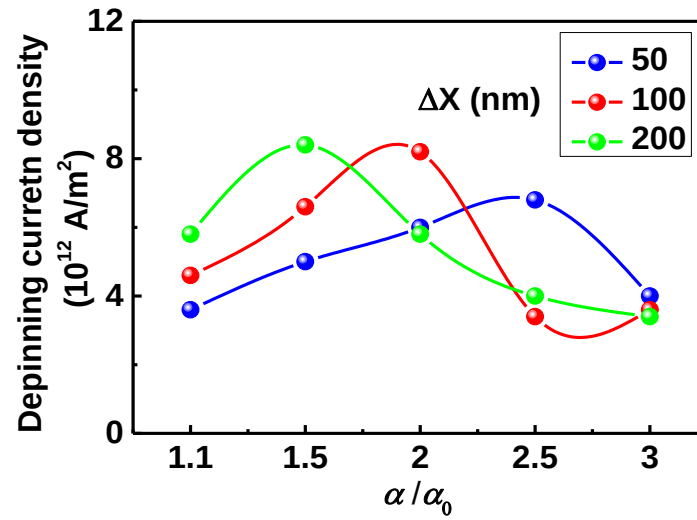


Figure 4.17. The effect of damping constant (α) on the depinning current density for various values of ΔX , the length of the modified region.

The depinning current does not increase linearly with damping constant but shows a peak in each curve. The depinning current depends directly on damping constant (α) and the saturation magnetization (M_s). Therefore, when the M_s decreases and α increases, a peak is seen. When the length of the pinning site is less than 50 nm, domain wall could not be pinned efficiently. These results indicate that a desired critical pinning current density can be achieved by choosing proper parameters [37].

4.5 Conclusion

A new domain wall pinning method was proposed using local compositional modifications and was examined this method theoretically and experimentally. Micromagnetic simulation predicts the ability of the local modification to increase pinning on desired positions. Investigations at a thin-film level indicate that the diffusion caused by annealing is useful to form domain wall pinning sites. Diffusion of Cr proves to be more effective in comparison to the other elements such as Ta, Cu, and Ru in changing the magnetic and magnetodynamic properties. Domain wall devices were fabricated and investigated using resistance measurements as a gauge of domain wall pinning. Compared to Permalloy wires without Cr-diffused pinning sites, Permalloy wires with Cr-diffused pinning sites exhibited different properties. *R-H* curves show that the Cr crossbars help to pin the domain walls. Although the annealing effect on the magnetic properties of NiFe and NiFe coated with Cr is investigated experimentally in this study, the compositional modification can also be carried out using techniques such as ion-implantation [38, 39]. For much higher density operation, materials with a perpendicular magnetic anisotropy may be used.

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Chapter 5 Domain wall pinning by ion implantation

5.1 Introduction

In the previous chapter, the pinning of domain wall in a NiFe nanowire was reported by tuning magnetic properties in a local region using metal diffusion from crossbars. The pinning field was found to be about 100 Oe. Although NiFe was selected for the case of testing the proposed concept, NiFe is not a suitable candidate for a domain wall memory application. Due to its low magnetic anisotropy ($\sim 10^4$ erg/cm³), a large domain wall size and poor thermal stability are some of the problems of NiFe [1]. The domain wall width of NiFe is about 100 nm, which dramatically limits the data density. A thermal stability, which could store data for about 300 days has been reported, whereas the requirement of such devices is 10 years [2]. Therefore, we investigated the magnetic material based on Co/Pd multilayers, which shows strong perpendicular magnetic anisotropy (PMA). In these materials, the domain wall size is about 10 to 40 nm and they exhibit potentially higher thermal stability [3].

In the past two decades, materials with perpendicular magnetic anisotropy (PMA) have attracted much attention due to their high magnetic anisotropy [3, 5-8]. The materials with high anisotropy possess higher thermal stability, which makes them appropriate for information storage and the other various technological applications [3, 9]. Recently, different materials have been proposed and studied for increasing the perpendicular anisotropy to use them for HDD, MRAM and DW memory [7, 9-12]. There are many facts which give rise to PMA in a material such as magnetocrystalline anisotropy, shape anisotropy, interfacial anisotropy, and magneto-elastic anisotropy. Out of several systems, ferromagnetic multilayers have been investigated extensively by the research

community, due to the various possibilities that can be explored easily. In this ferromagnetic multilayers system, transition metals, such as Co, Ni and Fe are grown with alternating layers of heavy non-magnetic metals such as Pt, Pd, and Au, *etc.* [13-16] These multilayers attain PMA mainly because of the interfacial anisotropy arising due to the lattice mismatch between the magnetic and non-magnetic layers. At the interface, strong hybridization between the 3d and 5d orbitals of the ferromagnetic and non-magnetic metals takes place, which increases the spin-orbit interaction of the ferromagnet.

In this chapter, firstly, the experimental details related to this study are introduced. Secondly, the metal diffusion using thermal annealing is investigated to tune the properties of Co/Pd multilayers. Then, an alternative approach is discussed to tune the properties of Co/Pd based on ion-implantation. The micromagnetic simulation was performed to show the pinning efficiently. Co/Pd microwires with pinning sites formed by ion-implantation were fabricated and investigated.

5.2 Experimental details

The Co/Pd multilayers used for experiments were deposited on thermally oxidized Si substrates by DC sputtering. The thickness of Co and Pd were fixed to 0.3 nm and 0.8 nm after optimization, respectively, which can achieve a relatively high perpendicular magnetic anisotropy. The stack is shown in Figure 5.1(a), where N (= 5, 10 and 15) indicates the repeating cycles in each film. A seed layer of Ta (5 nm)/Cu (5 nm)/Pd (3 nm) was used to improve the crystallographic growth of the multilayers and achieve the perpendicular magnetic anisotropy. The magnetic properties of each film were characterized by

AGFM in the out-of-plane (OOP) direction. The characterized magnetic hysteresis loops (M - H) are shown in Figure 5.1(b). The rectangular loops for $(\text{Co/Pd})_{5,10,15}$ bilayers indicate the presence of strong PMA in each film. While for Co/Pd multilayers with small $N = 5$, the sharp magnetization switching, and larger domains were observed. In contrast, Co/Pd multilayers with a larger $N = 15$ have a long tail in the hysteresis loop due to an increase in the magnetostatic energy and drastic reduction in domain width.

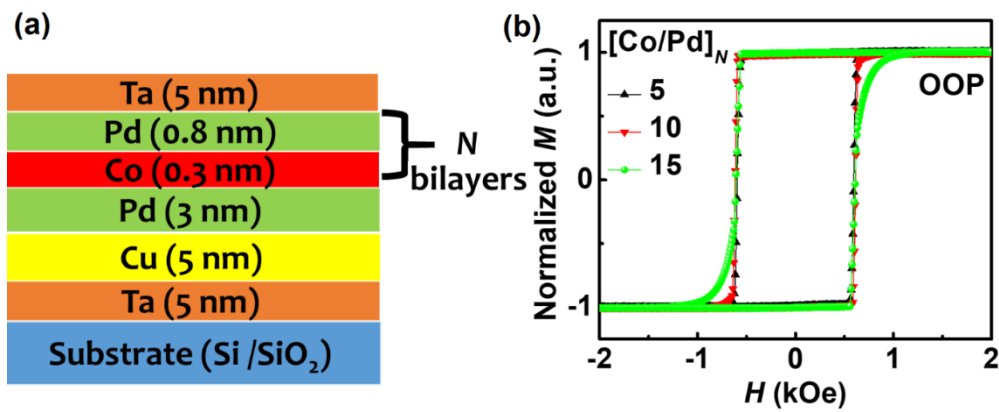


Figure 5.1. (a) Schematic illustration of a Co/Pd multilayer stack structure and (b) Out-of-plane (OOP) hysteresis loops of a Co/Pd multilayers with different bilayers of N .

5.3 Magnetic properties of Co/Pd modified by Cr diffusion

As a first-step in the investigations on Co/Pd multilayers, the effect of Cr diffusion was investigated for its effectiveness in domain wall pinning. Co/Pd with 15 bilayers was selected as an example to study the effect of tuning its properties by using Cr diffusion. M - H loops for $(\text{Co/Pd})_{15}$ along OOP and IP directions as shown in Figure 5.2. The anisotropy magnetic field is about 20 kOe. $(\text{Co/Pd})_{15}$ multilayer samples with a Cr capping layer of 5 nm and without a Cr capping layer were prepared. Both were annealed in a vacuum atmosphere at different temperatures.

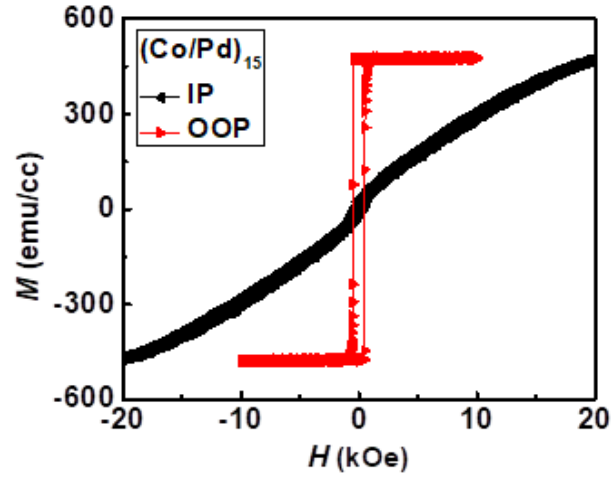


Figure 5.2. In-plane (IP) and out-of-plane (OOP) hysteresis loops of $(\text{Co/Pd})_{15}$.

Figure 5.3 shows the magnetic properties for $(\text{Co/Pd})_{15}$ without Cr diffusion and with Cr diffusion at different annealing temperatures (T_{an}), as shown in Figures 5.3 (a) and (b), respectively. In this investigation, the annealing time was kept at 60 mins, similar to the value used in the investigations on NiFe. After annealing, the magnetic properties of each film were characterized by AGFM. It can be seen that a higher T_{an} causes more effective changes in the properties of Co/Pd multilayers samples. However, the saturation magnetization (M_s) of both $(\text{Co/Pd})_{15}$ samples, without Cr and with Cr, show a reduction after annealing. For $(\text{Co/Pd})_{15}/\text{Cr}$, this reduction in M_s is larger. The hysteresis loop of $(\text{Co/Pd})_{15}$ along the OOP direction becomes narrow and tilt, and the loops along the IP direction becomes easier to be saturated with increasing T_{an} . The magnetic anisotropy was changed from OOP to IP after $T_{an} = 400^\circ\text{C}$, even for the $(\text{Co/Pd})_{15}$ sample without Cr diffusion. Such a deterioration in the magnetic properties in samples without a Cr layer is not desired to make a pinning site.

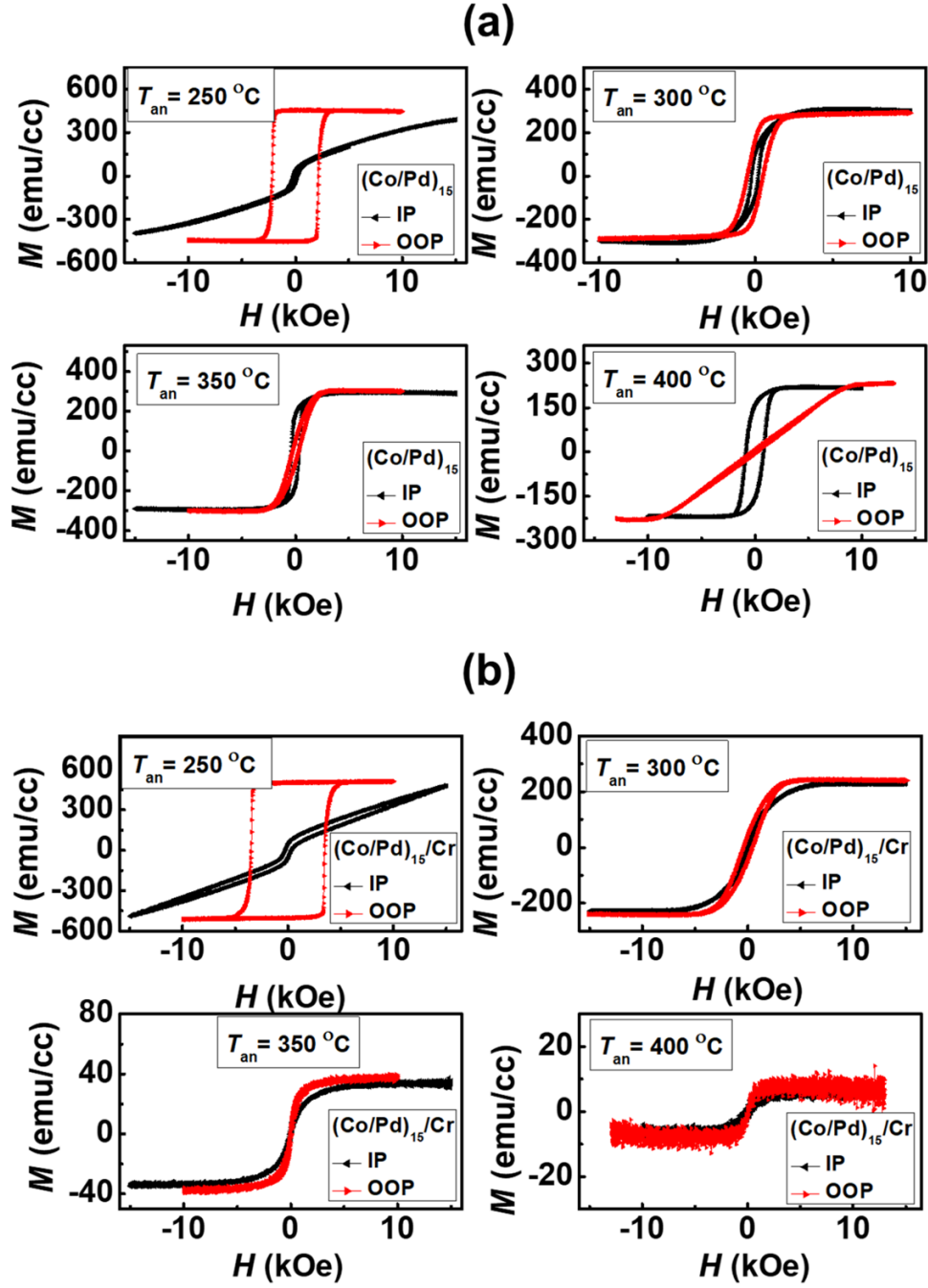


Figure 5.3. Hysteresis loops along the in-plane (IP) and out-of-plane (OOP) directions for the $(\text{Co/Pd})_{15}$ samples (a) without a Cr capping layer and (b) with a Cr capping layer after annealing for 60 mins at different temperatures T_{an} .

The loss in PMA for $(\text{Co/Pd})_{15}$ is understood to be due to the mixing of Co and Pd at the interfaces [17]. As a next step to solve this problem, the annealing time was reduced to 20 mins, while maintaining the same annealing temperature

T_{an} , in order to maintain PMA for $(\text{Co/Pd})_{15}$ without a Cr layer. Figure 5.4(a) shows the magnetic properties for $(\text{Co/Pd})_{15}$ without Cr diffusion annealed at different temperatures T_{an} , for annealing duration of 20 mins. PMA is still present even at $T_{an} = 400$ °C. However, the saturation magnetization M_s decreased from 470 to 240 emu/cm³ and the field (H_k) at which the in-plane magnetization saturates decreased from about 20 to 5 kOe. These results indicate an interfacial mixing of Co and Pd atoms and a reduction in PMA. $(\text{Co/Pd})_{15}/\text{Cr}$ shows an apparent loss of PMA above $T_{an} = 400$ °C, as shown in Figure 5.4(b). The saturation magnetization M_s reduced to 160 emu/cm³.

To further understand the changes in anisotropy with Cr diffusion, the domain structure was measured using magnetic force microscopy (MFM). The results are shown in Figure 5.5, which were taken over a 5 µm by 5 µm area of each thin film. The top and bottom rows show the domain structure for Co/Pd multilayer without a Cr capping layer and with a Cr capping layer, for annealing temperatures T_{an} ranging from 100 to 400 °C, respectively. In the MFM images, the yellow and brown regions represent the magnetization pointing in up and down directions. When T_{an} is less than 200 °C, both samples show a clear stripe domain. No much difference in the size of the domain has been observed for those samples. The size of the domain for Co/Pd multilayer without Cr capping layer is about 1.5 µm and its size for the sample with Cr capping is about 1.2 µm. When T_{an} goes to 250 °C, the size of the domain shrinks to around 300 nm, and small dots have been observed in both two kinds of samples. Compared to Co/Pd multilayers without a Cr capping layer, the stripe domain signal becomes weaker for Co/Pd multilayers with a Cr capping layer, which indicates the reduction in PMA after Cr diffusion. For higher annealing temperature, $T_{an} = 400$ °C, PMA is

completely lost for $(\text{Co/Pd})_{15}/\text{Cr}$ with Cr diffusion. The saturation magnetization M_s change is summarized for $(\text{Co/Pd})_{15}$ without and with Cr diffusion at different T_{an} , with the annealing time 60 mins and 20 mins, in Figures 5.6 (a) and (b) respectively.

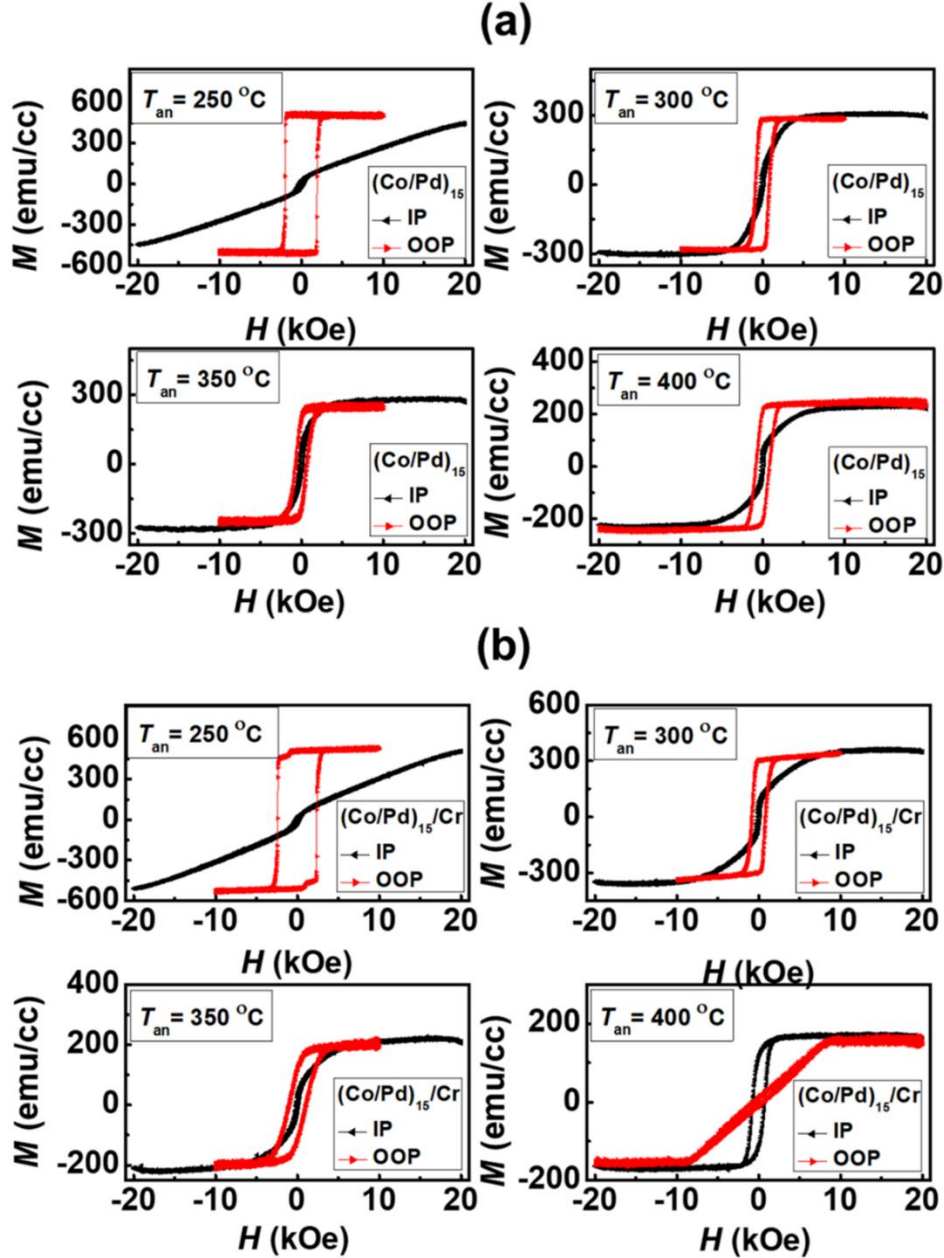


Figure 5.4. Hysteresis loops along the in-plane and out-of-plane directions for $(\text{Co/Pd})_{15}$ samples (a) without a Cr capping layer and (b) with a Cr capping layer after annealing for 20 mins at different temperatures T_{an} .

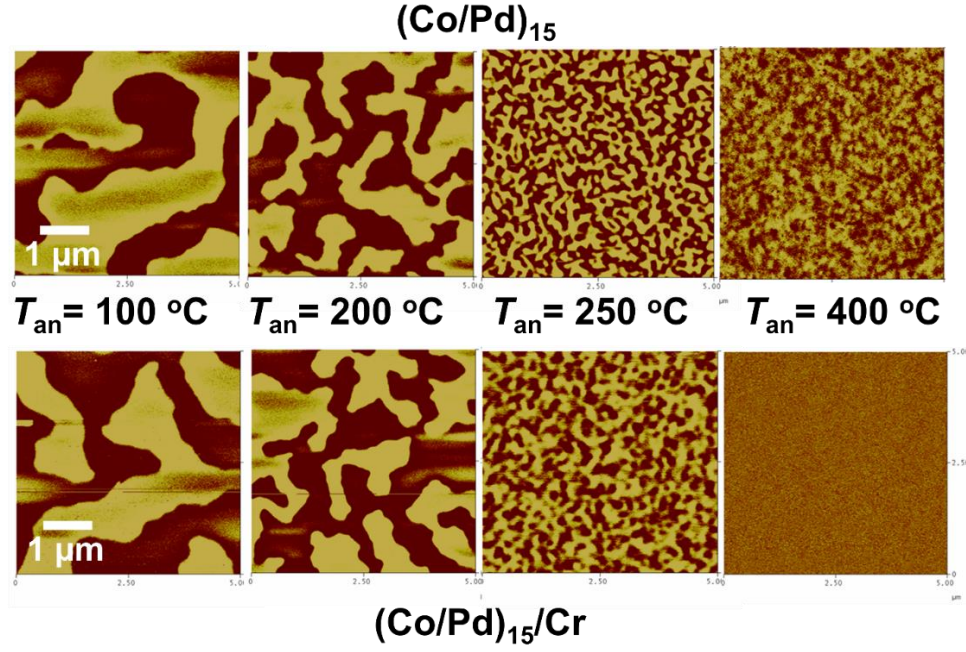


Figure 5.5. Domain structures for $(\text{Co/Pd})_{15}$ and $\text{Co/Pd}_{15}/\text{Cr}$ after annealing at different T_{an} .

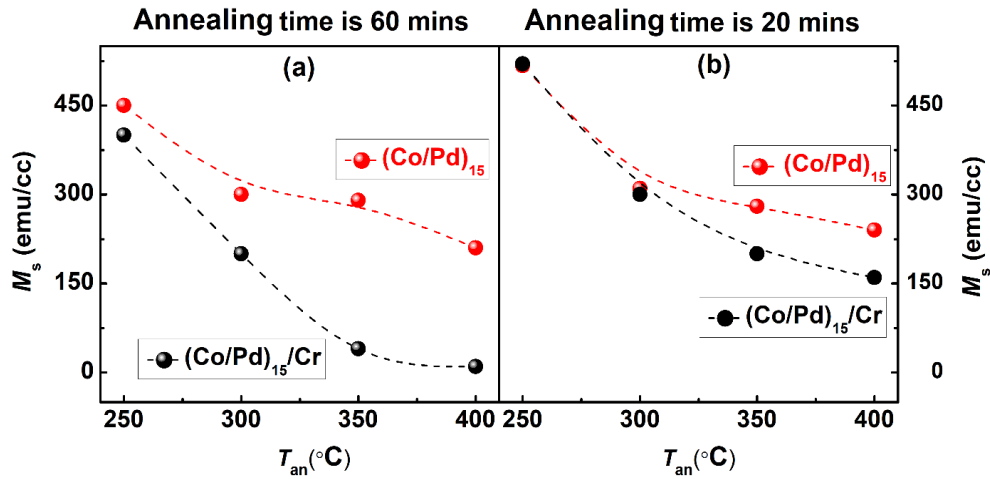


Figure 5.6. Saturation magnetization changes with different T_{an} for $(\text{Co/Pd})_{15}$ and $(\text{Co/Pd})_{15}/\text{Cr}$ annealed for (a) 60 mins and (b) 20 mins.

5.4 Magnetic properties of Co/Pd modified by ion implantation.

Although Cr diffusion can change the magnetization orientation from out-of-plane to in-plane under thermal annealing, the anisotropy of Co/Pd without Cr diffusion is also reduced due to the intermixing between Co and Pd atoms. In addition, the saturation magnetization is also reduced by 80%. These results are

not acceptable for making domain wall memory. Therefore, the introduction of non-magnetic ions was investigated by implantation into Co/Pd multilayers to change the magnetization configuration. Ion-implantation is an advantageous method to tune material properties for mass manufacturing [18]. This method offers a variety of elements for selection, which is easy to be adapted when ion-implanted through the mask.

Firstly, ion-implantation was tested on $(\text{Co/Pd})_N$ multilayer samples with different N . All the films were then diced and divided into two different groups to be compared the magnetic properties with and without ion-implantation. A 10 keV B^+ implantation was carried out at a fluence of 5×10^{15} ions/cm² to achieve the compositional modification on one group of the films. The magnetic properties of the two group samples were characterized by AGFM along the IP and OOP directions. The domain structure was mapped by using the polar Kerr imaging microscopy after AC demagnetization with applying an OOP field.

The hysteresis loops for $(\text{Co/Pd})_N$ multilayer samples without and with implantation are shown in Figure 5.7. Changes of the OOP hysteresis loops were observed from the two group samples, as shown in Figures 5.7 (a2), (b2) and (c2). The hysteresis loops of B^+ implanted $(\text{Co/Pd})_{5,10}$ samples, which vary from that of the non-implanted samples, indicate that the implanted $(\text{Co/Pd})_{5,10}$ samples have no PMA and that the magnetizations lie in-plane. The same results can be confirmed from the domain images. All the unimplanted samples show micron-sized stripe domains with a sharp contrast, which are typical films with PMA, as shown in Figures 5.8 (a1), (b1) and (c1), but after B^+ implantation, the stripe domains disappear, as shown in Figures 5.8 (a2) and (b2). The PMA of the implanted sample with $N = 15$ decreases, as compared to the unimplanted ones,

as shown in Figure 5.7(c2). The stripe domains in the $N = 15$ sample becomes weaker after implantation, as compared in Figures 5.8 (c1) and (c2). In general, the anisotropy decreases for the B^+ implanted samples. The decline in PMA can be attributed to the change of Co/Pd interfaces caused by the implantation, as PMA arises from Co/Pd interfaces [18-22].

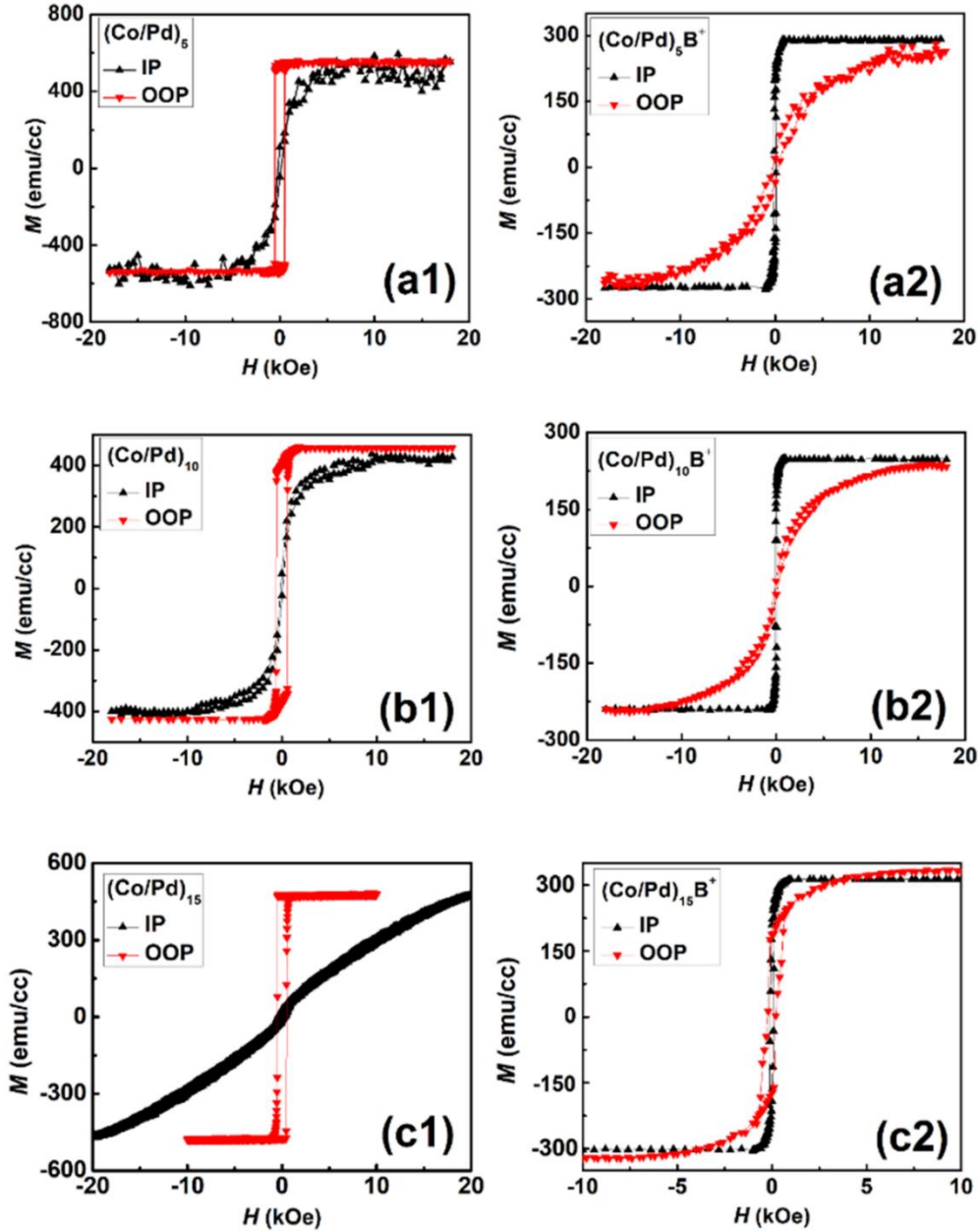


Figure 5.7. Hysteresis loops for the $(\text{Co/Pd})_N$ multilayer sample with different repeats, $N = 5, 10$ and 15 , along the in-plane (IP) and out-of-plane (OOP) directions (a1), (b1) and (c1) without and (a2), (b2) and (c2) with B^+ implantation.

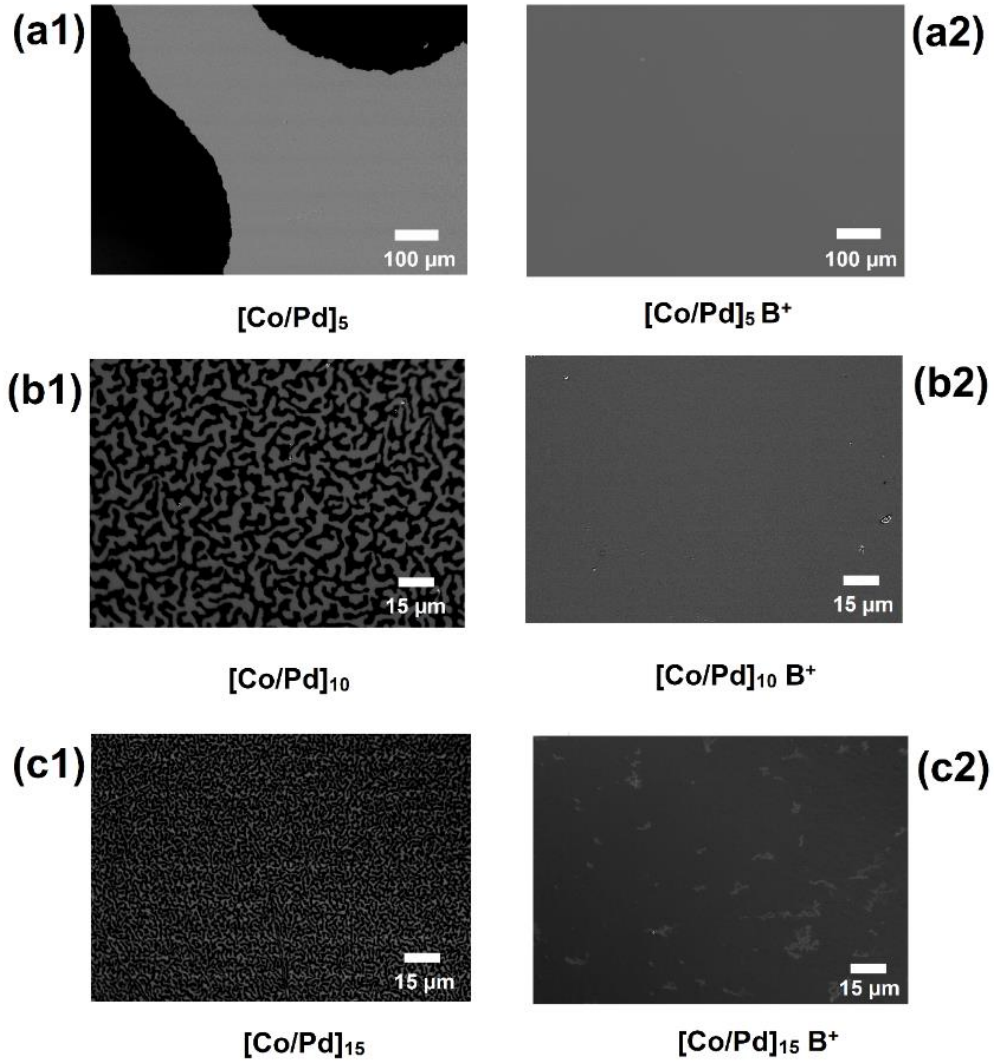


Figure 5.8. Domain images of the $(\text{Co/Pd})_N$ multilayers with different N (a1), (b1) and (c1) without implantation and (a2), (b2) and (c2) with implantation. The black and grey regions represent magnetization in two different orientations (into the paper and out-of-paper directions, for example).

X-Ray Diffraction (XRD) measurements were carried out on two groups of samples to validate my hypothesis. The measurement results are shown in Figure 5.9. For the unimplanted group, the samples show a peak at 40.8, 41.0 and 41.1 degrees for $N = 5, 10$ and 15 , respectively. The peaks follow the widely reported behavior that the peak position depends on the relative thickness of the Co and Pd layers in Co/Pd multilayer systems [23, 24]. Since pure Pd (111) exhibits a peak at around 40.8 degrees and pure Co (111) exhibits a peak at

around 44.4 degrees, the Co/Pd multilayers exhibit a peak in between these two angles, depending on the relative thickness of Co and Pd. The peak position also varies with the number of bilayers. In the case of $N = 15$, there are more Co layers in comparison to the sample with $N = 5$ and hence the peak shifts towards the higher angles (closer toward the Co (001) peak). For the B^+ implanted group, the peak shifts towards lower 2θ values, as compared to the unimplanted group. This shift to lower 2θ values indicates that the distance between lattices has increased in all the implanted samples. The shift magnitude is 0.59, 0.45 and 0.32 degrees for $N = 5, 10$ and 15 , which indicates that it has an inverse dependence on the N value. For the B^+ implanted samples, the intermixing between Co and Pd atoms also contributes to the shift, resulting in a further shift towards lower 2θ values [20].

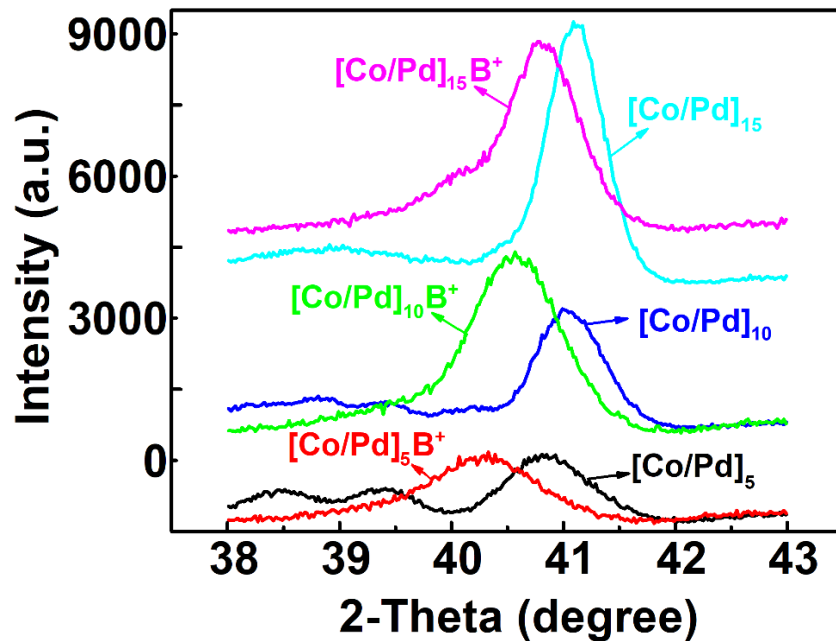


Figure 5.9. XRD peaks of samples with different values of N , without and with B^+ implantation.

In order to confirm the intermixing of Co and Pd caused by the implantation, Transport of Ions in Matter (TRIM) simulations were carried out. The TRIM

simulation focused on the distribution of Co and Pd atoms displaced from their initial positions [21, 25]. The diffusion between Co and Pd atoms is clear for $N = 5$ and 10 as shown in Figure 5.10. For $N = 15$, the diffusion between Co and Pd is not significant. This inter-diffusion corroborates with the reduction in the magnetic anisotropy and the XRD results.

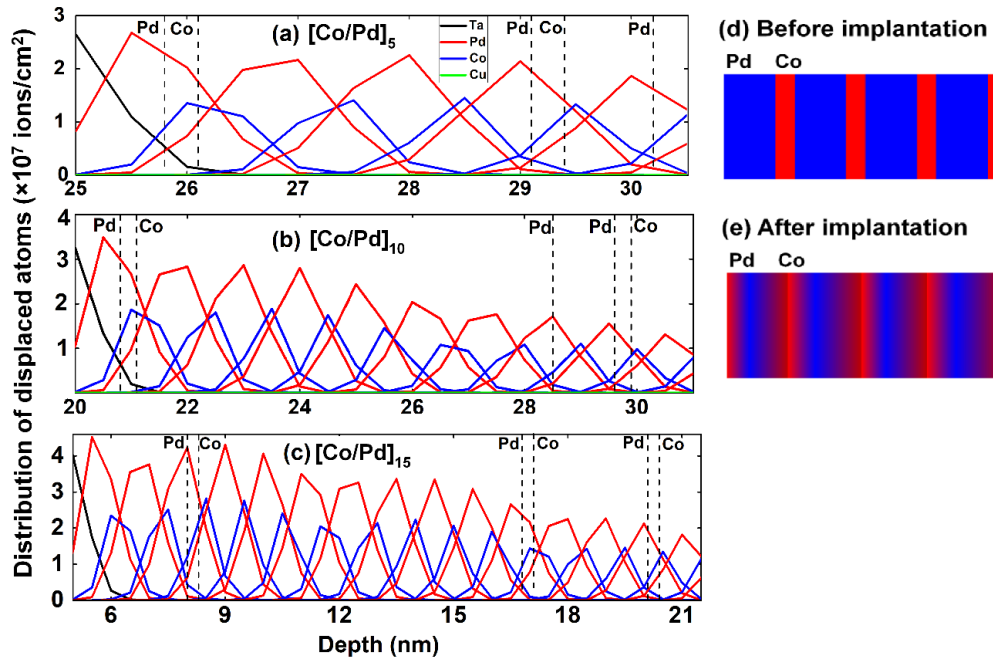


Figure 5.10. The distribution of displaced atoms (obtained from TRIM simulations) of Co and Pd with B^+ implantation (a) for $N = 5$, (b) for $N = 10$ and (c) for $N = 15$ bilayers. The dashed lines indicate the middle portion of the Co and Pd layers before ion implantation. The Co and Pd layers were assumed to be sharp before implantation, as shown in illustration (d). After implantation, the atoms are diffused as illustrated in (e).

5.5 Micromagnetic simulations

Cr diffusion and B^+ implantation both are effective in changing the magnetic properties of a Co/Pd multilayer, such as reducing M_s and K_u . Micromagnetic simulations were carried out by using Mumax3 to understand the possibility of DW pinning in synthetic magnetic texture regions, which formed in low M_s and K_u [26].

Two types of nanowire devices of Co/Pd multilayer structures (Type A: with the length of 128 nm, the width of 16 nm and the thickness of 2 nm, Type B: length of 5120 nm, the width of 1280 nm and thickness of 10 nm) were studied in the simulation. The material properties of Co/Pd multilayers were chosen to be at the unimplanted areas, viz., $M_s = 400 \text{ emu/cm}^3$ and magnetic uniaxial anisotropy energy $K_u = 5 \times 10^6 \text{ erg/cm}^3$ and defined the areas where compositional modification was carried out to have a relatively lower saturation magnetization $M_s = 200 \text{ emu/cm}^3$ and a lower magnetic anisotropy energy $K_u = 1 \times 10^6 \text{ erg/cm}^3$. All through simulation, a discretization of $2 \text{ nm} \times 2 \text{ nm} \times 2 \text{ nm}$ is used which is sufficiently smaller than the exchange length of 4 nm [27, 28]. Figure 5.11 shows the simulation results of type A structures.

Figures 5.11 (a1-a3) show DW propagation in nanowires without domain wall pinning sites. In these wires, M_s and K_u are uniform throughout the length, representing a wire with no implantation-induced local modification. When a current was sent with a density of $5 \times 10^{10} \text{ A/m}^2$, the domain walls propagated to the end of nanowire without pinning. The domain wall velocity was found to be 8 m/s [29]. Figures 5.11 (b1-b3) illustrate the scenario for the nanowires with modified regions. It can be noticed that DW stopped at the modified region and became wider. When the spin-polarized current was increased to $2 \times 10^{11} \text{ A/m}^2$, DW was depinned and moved to the end of the nanowire. Therefore, the simulated results indicate that the proposed scheme has the potential to pin DWs in a controllable manner. More detailed descriptions of simulation studies are presented later.

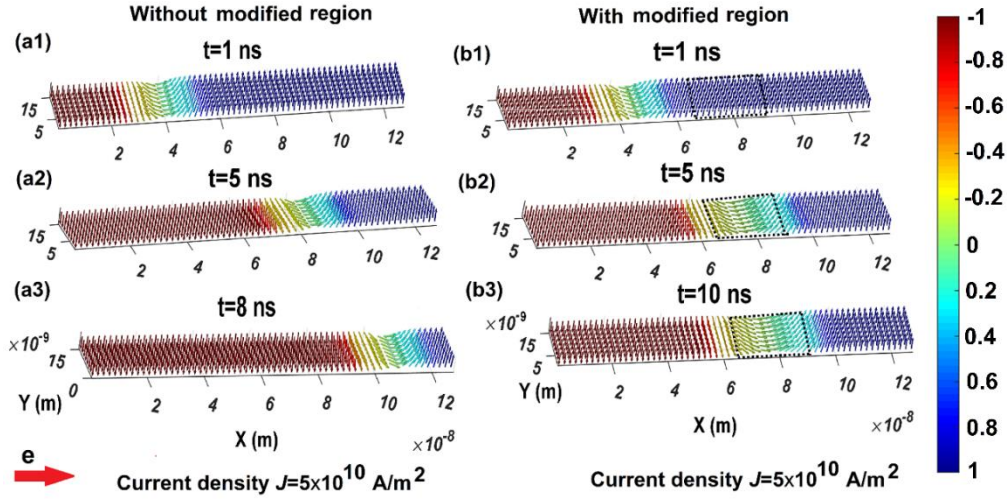


Figure 5.11. Micromagnetic simulation studies of DW propagation in two magnetic nanowires (a) without pinning sites and (b) with periodic pinning sites which have different magnetic properties from the other regions. (a1-a3) depict the domain wall positions at different times in a nanowire without pinning sites. (b1-b3) depict the domain wall positions at different times in a nanowire where local modification of magnetic properties was introduced to create pinning sites. A spin current with a current density of $5 \times 10^{10} \text{ A/m}^2$ was used to move DWs [145, 197]. The dashed boxes show the modified area. “-1” and “1” present the magnetization pointing down and up directions, as shown in the color map.

5.6 Domain wall pinning in $(\text{Co/Pd})_{10}$ devices

In order to demonstrate such modified areas to act as pinning centers in experiments, a $(\text{Co/Pd})_{10}$ wire device with localized B^+ implantation was fabricated by a combination of e-beam lithography and ion-implantation experiments. E-beam lithography was used to form trenches. Ion-implantation was carried out to modify the wire by sending B^+ ions through the trenches.

The Kerr imaging system was used to observe the DW nucleation and motion. Initially, the nanowire device was magnetized to saturation to eliminate DWs. Kerr image shows grey regions, (Figure 5.12(a)), which indicates that there is no DW in the device after saturation. Later, a pulsed current with a density of

$1 \times 10^{12} \text{ A/m}^2$ and a width of 1ms was sent through the injection line to give rise to an Oersted field to nucleate DWs in the wire devices [30, 31]. As shown in a Kerr image in Figure 5.12(b), two magnetic domains which are in black were generated beside the cross-bar. (The reason for the small domain at the right side is not apparent). The reversal magnetic field was increased gradually. At an external magnetic field of 900 Oe, DW moved to the pinning centre, as evidenced by the expansion of the black region to the position where the B^+ implantation is located. The black region does not expand further when the applied field was increased to 910 Oe, which indicates that the DW is pinned at this position. This fact implies that the pinning center formed by B^+ implantation in this region is effective. When the external magnetic field was increased to 920 Oe, the black region moved across the implantation position and spread to the end of the device. Hence, it can be concluded that the implantation region acts as a pinning center to stabilize the DWs, and the depinning field is close to 920 Oe.

It may be argued that the geometrical modification around the small cross-bar due to ion-implantation is a possible cause of pinning. However, this effect should be negligible, as the depinning field is much larger (~ 920 Oe) than that observed in the case of geometrical modification, such as notches (~ 100 Oe) [32-34].

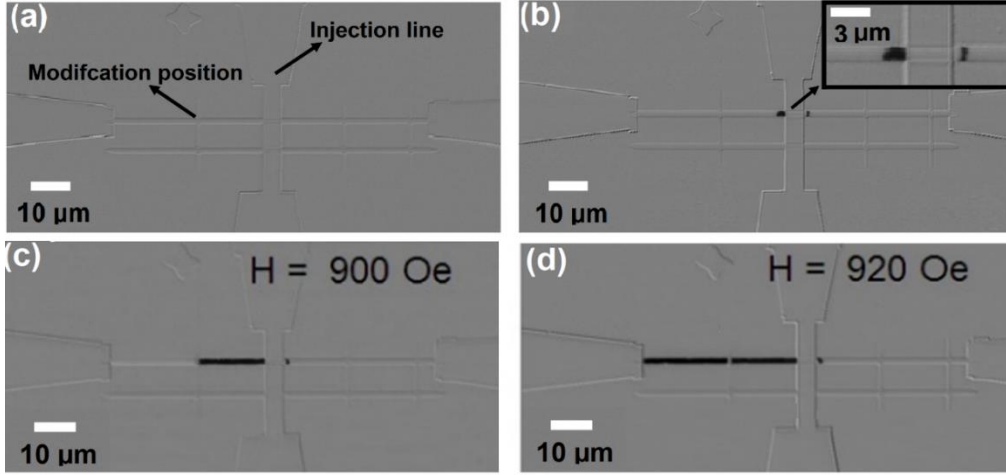


Figure 5.12. (a) Kerr images which show the nanowire device with the saturation state. (b) DW nucleation at the cross-bar and (c and d) the DW propagation by applying different magnetic fields along the perpendicular to the plane direction.

As the next step, we studied the pinning properties of the nanowires by controlling B^+ ion implantation depth systematically. For this study, an additional Ta layer was deposited on top of the Co/Pd multilayers. The thickness of this Ta layer would determine the depth at which B^+ ions are implanted. The devices with multiple pinning sites were fabricated as follows: Firstly, the $(Co/Pd)_{10}$ wires were fabricated by e-beam lithography (EBL), followed by ion-milling. For the formation of pinning sites, an additional lithography step was carried out. Through the trenches of the resist, a Ta overcoat layer with different thickness, such as 6 nm, 3 nm, and 0 nm was deposited, to control the depth of the implanted B^+ ions.

Figure 5.13(a) shows a schematic of the domain wall microwire. The red and blue colors indicate up and down magnetized domains respectively. Yellow regions show the regions implanted with B^+ ions. The lengths of these regions are 400 nm, 600 nm, 800 nm and 1 μ m from left to right. The circular region at the left side with a diameter of 5 μ m for injecting domain wall. Figures 5.13 (b) and (c) show the hysteresis loops of the microwires measured using a polar Kerr

microscope. The widths of the wires are 1 μm and 2 μm respectively, and the length is the same as 15 μm . In both cases, the sample without B^+ implantation (which is named as Ref) shows a steep hysteresis loop, indicating that the magnetization reverses by a swift motion of a domain wall in a narrow range of an applied magnetic field. However, for samples implanted with B^+ , the behavior is different.

For shallow-implantation (SI) sample (for Ta layer thickness of $t_{\text{Ta}} = 6 \text{ nm}$), the reversal shows two steps. In this sample, the implantation is expected to happen only at the top surface of the nanowire. As a result, there is weak pinning which leads to some small steps in the magnetization reversal. For medium-implantation (MI) sample (with $t_{\text{Ta}} = 3 \text{ nm}$) and deep-implantation (DI) (with $t_{\text{Ta}} = 0 \text{ nm}$) sample, the magnetization shows multiple steps, indicating that the domain wall propagation is hindered by the pinning sites. The loops also become wider for the MI and DI samples, indicating a stronger domain wall pinning.

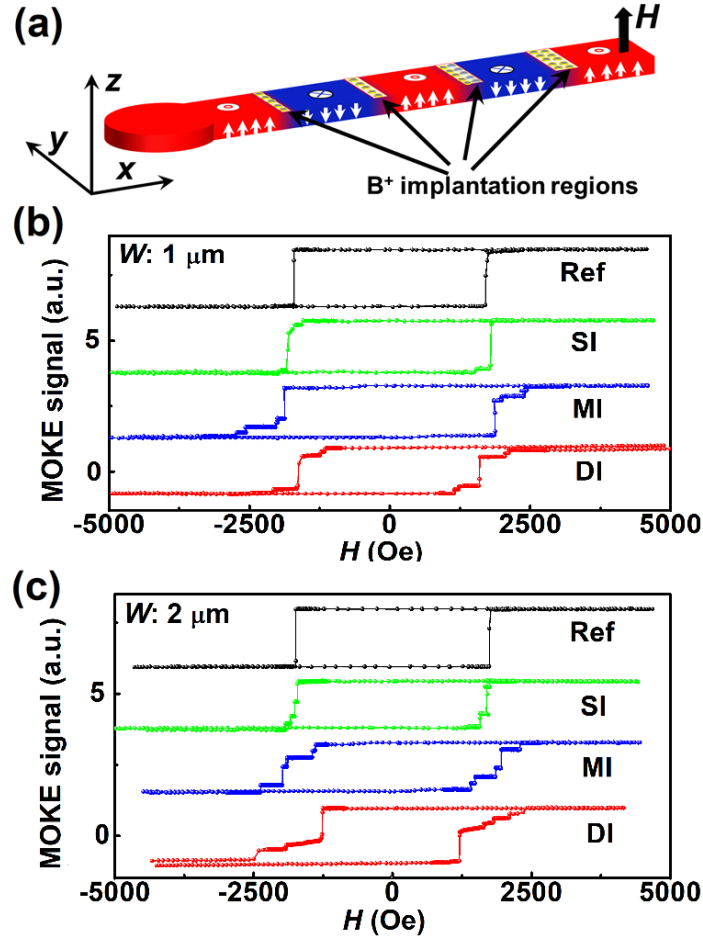


Figure 5.13. (a) Schematic view of the domain wall microwire. The yellow regions are implantation area, which length is 400 nm, 600 nm, 800 nm and 1 μm from left to right. Hysteresis loops of the microwires with different widths of (b) $w = 1 \mu\text{m}$ and (c) $w = 2 \mu\text{m}$ with different depth of B^+ implantation.

Figure 5.14 shows the trend of switching field distribution (SFD) as a function of t_{Ta} for devices with two different widths W . SFD was measured by calculating the ratio $\Delta H/H_c$, where ΔH is the difference between a nucleation field and a saturating field and H_c is the coercivity [35, 36]. A nucleation field is defined as the field at which the magnetization $M = 0.95M_s$ and a saturating field is the field at which $M = -0.95M_s$. In both cases ($w = 1 \mu\text{m}$ and $2 \mu\text{m}$), the reference samples show the lowest SFD. As the ion-implantation depth is increased, SFD increases. DI sample with $t_{\text{Ta}} = 0 \text{ nm}$ shows effective pinning at

all sites, leading to a larger SFD. These results indicate that the pinning strength increases with the implantation depth. This provides tunability to control the pinning strength.

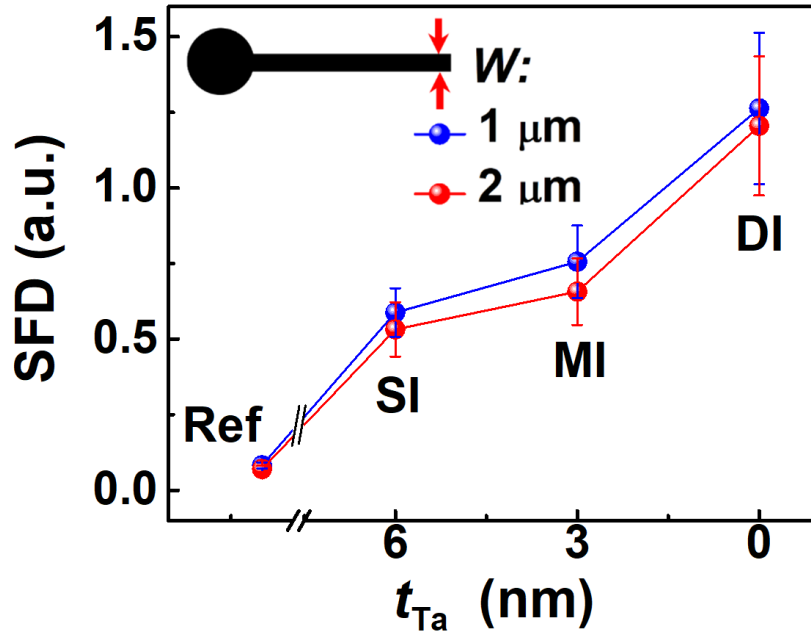


Figure. 5.14. Switching field distribution (SFD) as a function of overcoat layer Ta thickness t_{Ta} for microwires with widths of $w = 1 \mu m$ and $2 \mu m$. The inset is the schematic of the microwire with a circular region at the left side.

To visualize the domain wall propagation and to understand the pinning effect, the images of the domain patterns were captured at different values of the applied magnetic field. Figure 5.15 shows the trend of reversed domains as a function of the applied magnetic field, for devices with $1 \mu m$ width and $15 \mu m$ length. The circular region fabricated at the left side with a diameter of $5 \mu m$ at the left side is designed intentionally to nucleate and introduce DW. In this experiment, a 4.5 kOe magnetic field was applied to magnetize the sample in one direction and an image was captured to serve as a reference. All the domain images shown in Figure 5.15, which were captured at different magnetic fields, were obtained by subtracting the reference data from the signal at a particular

field. It can be noticed that the sample without implantation (Ref) shows a grey colored domain (representing a positive magnetization in the up direction) until a reversal field of about -1854 Oe was applied. When the magnitude of the reversal field was increased to -1858 Oe, the magnetization of the whole wire reversed. The presence of a single black-colored domain, representing a negative magnetization in the down direction, is an indication of a swift magnetization reversal caused by domain wall propagation throughout the whole device without pinning. It is such a magnetization reversal that led to the observation of a sharp hysteresis loop in Figure 5.13(b) for the reference sample.

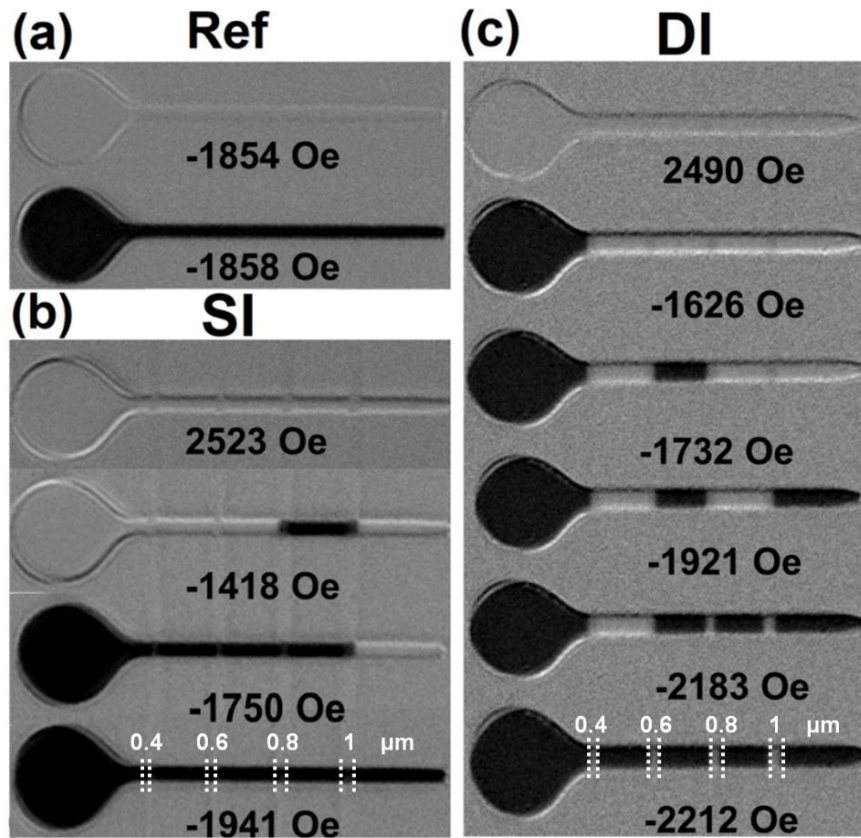


Figure. 5.15. Differential Kerr images at the different magnetic fields for sample without (a) implantation (Ref), SI and DI samples with a Ta overcoat layer with a thickness of (b) $t_{\text{Ta}} = 6$ nm and (c) $t_{\text{Ta}} = 0$ nm. The grey color represents the domain with a positive magnetization and the black color the domain with a negative magnetization. The devices with 1 μm width and 15 μm length.

For SI sample, however, the magnetization reversal occurs in a narrow range of field around -1418 Oe. Since the Co/Pd multilayers have an anisotropy field of over 10 kOe, this reversal at low fields is caused by nucleation of a small region where the reversal field is much lower than the anisotropy field. Such a reversal is reported to be caused by defects that arise due to many reasons, one of them being local non-uniformity in the layer thickness of Co and Pd [37]. However, it is interesting to note that this reversed domain did not propagate uncontrollably as in the reference sample but is stopped at the two edges where B^+ implantation was carried out. This result indicates that the regions where B^+ was implanted can act as pinning centres. When the reversal field was increased to -1750 Oe, the left region of the device reversed its magnetization. A sufficiently strong field of -1941 Oe was required to reverse the entire device. Such magnetization reversal in steps resulted in the observation of a wider and stepped hysteresis loop and a larger SFD for this sample as shown in Figures 5.13(b) and 5.14, respectively.

In Figure 5.15(d), the magnetization reversal in the DI sample is shown, which has the deepest implantation of B^+ . In this sample, each region separated by pinning sites reverses individually, indicating that the pinning strength is high in this sample. Every region, implanted with B^+ ions, hinders the domain wall propagation and offers effective pinning. The above-described behavior explains why the hysteresis loop in Figure 5.13(b) exhibits many steps and a larger SFD, as observed in Figure 5.14. These results indicate that B^+ implantation through a resist mask is a useful technique to achieve effective domain wall pinning.

In order to have a better understanding of the DW properties, the resistance of the device was measured at several magnetic fields. Figure 5.16(a) shows the

schematic of the experimental setup. A DC of 700 μA was sent through the device, corresponding to a current density of $7 \times 10^{10} \text{ A/m}^2$, and the magnetoresistance along the wire was measured while sweeping the external magnetic field.

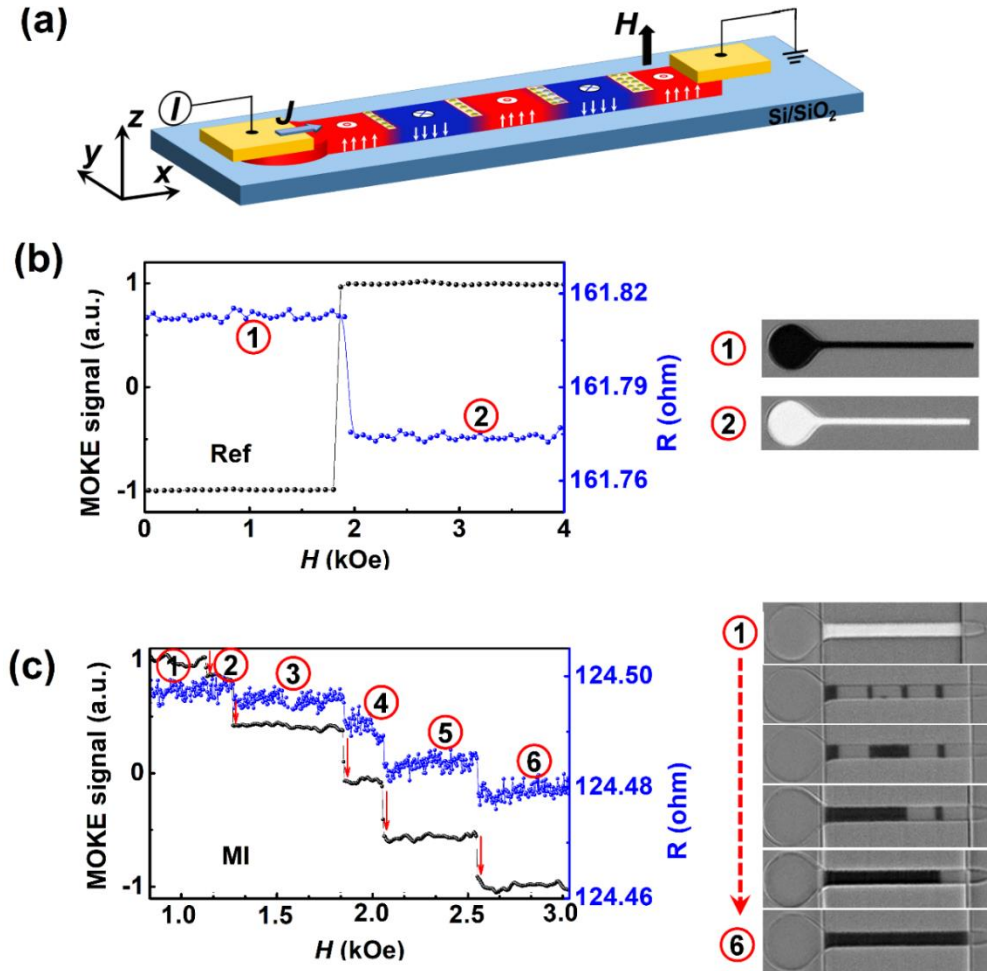


Figure 5.16. (a) Schematic of the setup used for magnetoresistance measurements. (b) and (c) show the resistance change with respect to the magnetic field, as well as the MOKE signal change for the Ref sample and the MI sample.

Figure 5.16(b) shows the variation of MOKE signals and resistance as a function of the magnetic field for the Ref sample. It can be noticed that the magnetization of the whole device reverses together and hence two resistive states are observed. However, the MI sample in Figure 5.16(c) shows multiple resistive states. The snapshots of the domain configuration at the corresponding

magnetic field show clear switching of individual magnetic domains. A slight positive slope in the resistance as a function of time is an indication of Joule heating. It can be noticed that there is a distinct change in the resistance besides the positive slope, which confirms that the change in resistance is due to the different domain states and not due to Joule heating. I propose that this device showing multiple resistive states, based on a domain wall can mimic the functionality of both neurons and synapses for neuromorphic computing. More energy efficient, spin-torque current-induced domain wall motion can be used to switch such a multilevel resistive state. Moreover, more layers could be added and a reading mechanism based on tunnel magnetoresistance (TMR) be implemented to increase the voltage swing and the difference between voltage levels [38].

In order to understand how effective the pinning would be in the case of spin-torque induced DW motion, further micromagnetic simulations were carried out. For this study, the type B structures were used, which resemble the experimental devices closely. Figure 5.17 shows the values of anisotropy constant (K_u) and the saturation magnetization (M_s) of the implanted regions. When a current was sent through the device, the domain walls were moved and pinned in the regions with a lower K_u or M_s . When the current density was increased further, the domain walls become depinned and moved further. The current density associated with depinning J_{dep} is shown in the table and in the graph. It can be noticed from the table that the depinning current density increases when both K_u and M_s decreases (diagonal direction of the table).

In the case of ion-implantation, both K_u and M_s are expected to decrease. So, this trend is clear indication that the ion-implantation induced local

modification is a good technique to pin the domain walls. The graph shows the trend of J_{dep} as a function of $\text{Mod}(H_{kw}-H_{kp})$ where H_{kw} is the anisotropy field of the wire and H_{kp} is the anisotropy field of the pinning region, obtained from the relation $H_k = 2K_{eff}/M_s$ [39]. In general, an increase in the value of J_{dep} associated with an increase in the difference between H_{kw} and H_{kp} is noticed. These results further confirm that the ion implantation induced changes in local magnetic properties can help in forming pinning centres.

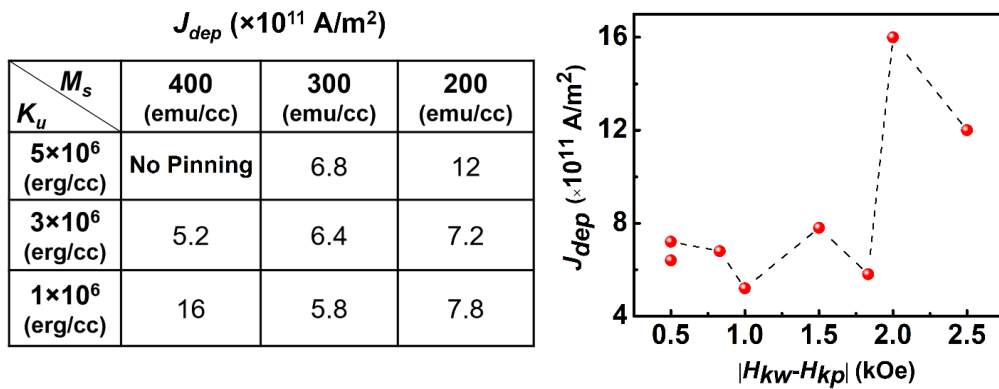


Figure 5.17. (a) Tabulation of depinning current density J_{dep} for pinning regions with various values of K_u and M_s . (b) Plot of depinning current density J_{dep} for pinning regions with various values of K_u and M_s .

5.7 Conclusion

In this chapter, the modification of magnetic properties in Co/Pd multilayer samples using Cr diffusion and B^+ implantation was investigated. Thermal annealing causes a significant interlayer mixing, making this technique unsuitable for Co/Pd multilayer system. Ion implantation was investigated in detail and is concluded to be a promising method to reduce the anisotropy of the Co/Pd multilayers and change the magnetization from the perpendicular to in-plane direction. The pinning effect has been validated by simulations and experiments [40, 41].

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Chapter 6 Domain wall pinning by exchange coupling

6.1 Introduction

Two different methods of domain wall pinning were discussed in the previous chapters. In this chapter, I propose and demonstrate another domain wall pinning method by introducing local exchange coupling between two layers with perpendicular magnetic anisotropy (PMA) and in-plane magnetic anisotropy (IMA) respectively [1]. The pinning sites can be formed by tilting the magnetization direction of the PMA layer locally. The ferromagnetic nanowire with PMA carries data bits, and magnetic nanowire with IMA is orthogonally deposited to the PMA nanowire. The magnetization of the PMA layer will be tilted due to the exchange interaction between two layers, as shown in Figure 6.1.

In this chapter, micromagnetic simulations were initially carried out to understand the pinning effect of the tilted magnetization to investigate whether it can stabilize the domain wall or not. Next, the coupling effect was investigated experimentally by depositing Co/Ni multilayer and characterizing the PMA properties of the Co/Ni multilayer. An IMA Co layer was chosen with a thickness of 3 nm. Angle dependent VSM measurements were performed to understand the Co/Ni multilayer magnetization direction when the IMA Co layer was deposited on its top. Between the Co/Ni multilayer with PMA and the thick Co layer with IMA, a Pt spacer layer was inserted with various thickness in order to tune the exchange coupling strength. The tilted angle was mimicked as a function of the strength of exchange coupling. Last part, DW pinning was studied in a Co/Ni nanowire with Co crossbar pinning sites.

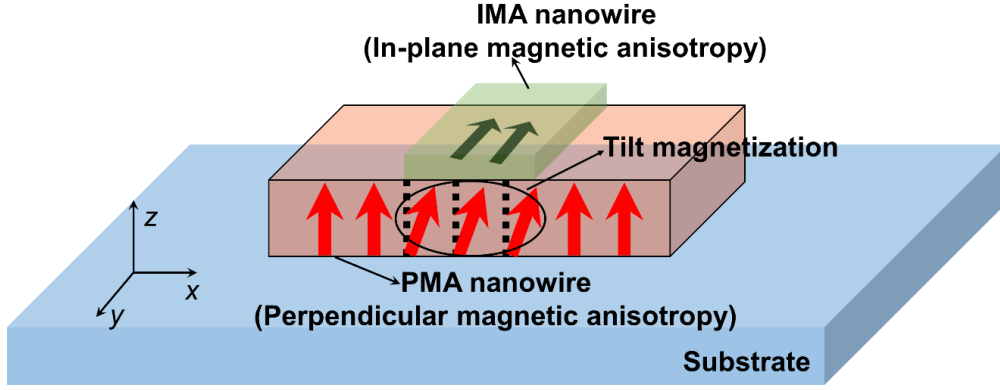


Figure 6.1. Schematic of localized tilting of magnetization via exchange interaction for pinning domain walls.

6.2 Micromagnetic simulations

The micromagnetic simulations were performed by using Mumax3 to demonstrate that a tilted magnetization can stabilize DWs in PMA nanowires [2]. The parameters used for a PMA layer are saturation magnetization $M_{s0} = 600$ emu/cm³ and the magnetic anisotropy energy $K_{u0} = 8 \times 10^6$ erg/cm³ [3, 4]. The parameters used for an IMA layer: saturation magnetization $M_{s1} = 800$ emu/cm³, and the magnetic anisotropy energy $K_{u1} = -2\pi M_s^2$ [5]. The tilted magnetization at a certain position was set, with the effective saturation magnetization $M_{s10} = 667$ emu/cm³ and the magnetic anisotropy energy $K_{u10} = 5.5 \times 10^6$ erg/cm³, which are calculated from equations (6.1) and (6.2). The thickness for the PMA layer t_0 is assumed to be 6 nm and the thickness for IMA layer t_1 is assumed to be 3 nm.

$$M_{s2} = \frac{M_{s0}t_0 + M_{s1}t_1}{t_0 + t_1} \quad (6.1)$$

$$K_{u2} = \frac{\sqrt{K_{u0}^2 t_0^2 + K_{u1}^2 t_1^2}}{t_0 + t_1} \quad (6.2)$$

Firstly, the DW dynamics was studied in the PMA nanowire without tilted magnetization. Figures 6.2 (a) and (b) show the DW position against time in a homogeneous nanowire driven by a magnetic field $H_z = 300$ Oe and an electric current with a density of $J = 2 \times 10^{12}$ A/m². DWs showed a constant velocity of 240 m/s and 254 m/s with respect to a magnetic field and an electric current.

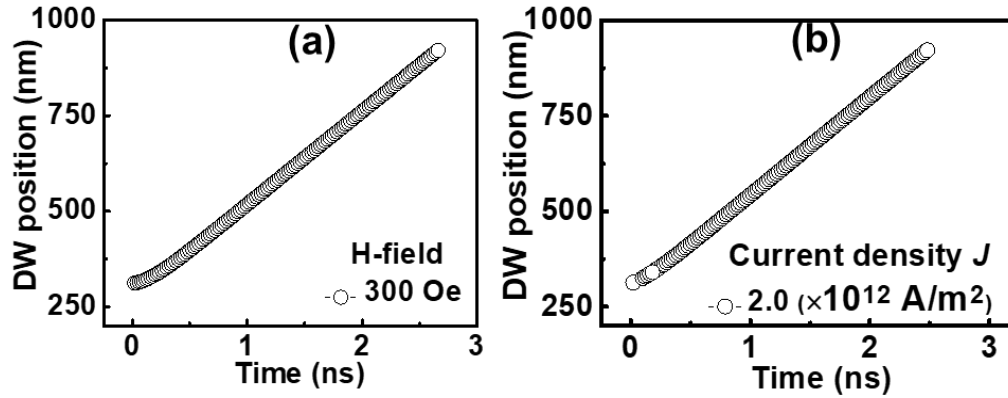


Figure 6.2. DW position with time driven by (a) a magnetic field and (b) an electric current. The velocity is 240 m/s at a magnetic field $H = 300$ Oe and 254 m/s at a current density $J = 2 \times 10^{12}$ A/m².

Figure 6.3(a) shows the nanowire with a tilted magnetization (M), with a length of 1024 nm and a width of 128 nm, and a thickness of 2 nm. The tilted magnetization area has a dimension of 200 nm in length and 128 nm in width, as indicated with the green colored box. The tilt angle was set by θ , expressed as $(0, \sin(\frac{\pi}{180} \theta), \cos(\frac{\pi}{180} \theta))$. DW shows steady motion before reaching the pinning site. At the position with tilted M , it was pinned.

The DW position is plotted as a function of time for different magnetic fields (H) and electric currents density (J), in Figures 6.3 (b) and (c). The magnetization tilt angle is 5° at the pinning position, expressed as $(0, \sin(\frac{\pi}{36}), \cos(\frac{\pi}{36}))$. When the magnetic field H is below 840 Oe, DW is moved up to and pinned at the position with tilted M . The velocity, which is determined

from the slope of the steady motion, increases with increasing magnetic field. When the field is increased to 850 Oe, DW is depinned and traveled to the end of the nanowire. A similar trend is observed for current-driven DW motion. The depinning current is determined to be at $3.64 \times 10^{12} \text{ A/m}^2$.

Figures 6.3 (d) and (e) summarize the depinning field and the depinning current (J_{de}) for different tilt angles, from 5° to 25° . Higher tilted angles of the PMA layer were not investigated as the exchange interaction would not energetically favor such a configuration [6, 7]. The depinning field and current do not consistently increase with the tilt angles. There exists an optimal tilt angle for maximum pinning which is dependent on how the DW was driven by a magnetic field or an electric current.

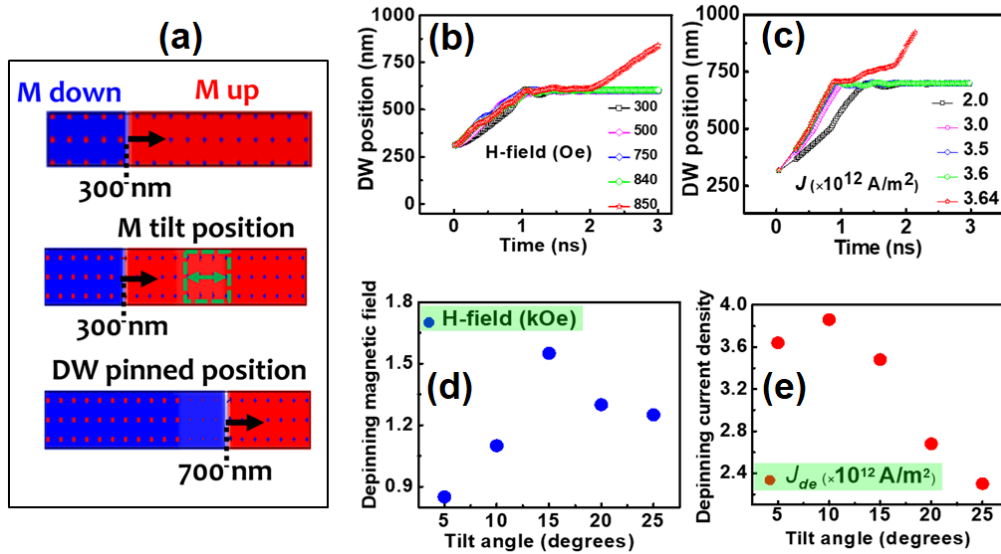


Figure 6.3. (a) Schematic of DW pinning at a tilted magnetization position. The DW position in the PMA nanowire with the tilt magnetization pinning site against time driven by different (b) magnetic fields and (c) electric currents. Depinning (d) magnetic fields and (e) electric currents density change with the different tilting angles.

Since the dynamics of a domain wall is sensitive to the damping coefficient, simulations were carried out to study the depinning field and current density with respect to different damping constants, as shown in Figure 6.4. The depinning

field and current density both are increasing with increasing damping constant in the presence of a fixed tilted magnetization.

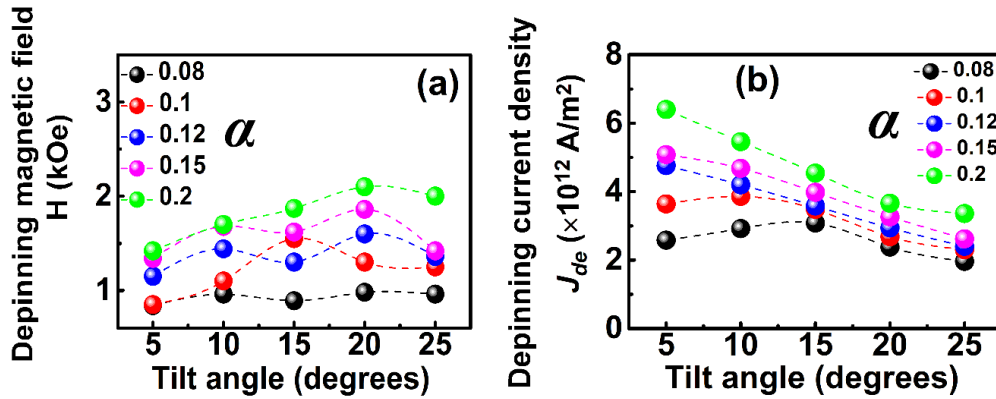


Figure 6.4. (a) Depinning field and (b) depinning current density with different damping constants.

6.3 Magnetization tilt for a Co/Ni multilayer with perpendicular magnetic anisotropy

The simulation results predict the possibility of DW pinning with localized tilting of magnetization. To achieve the tilted magnetization experimentally, a Co/Ni multilayer was grown with perpendicular magnetic anisotropy. The stack structure is shown in Figure 6.5(a), which is abbreviated as (Co/Ni)₂. Ta (5 nm) was used as the seed layer for good adhesion. The underlayer Pt (5 nm) was used to induce out-of-plane anisotropy for the Co/Ni multilayers. The thickness of Co and Ni was maintained to be 0.25 nm and 0.5 nm, respectively. A capping layer made of Pt was deposited, to protect the magnetic film from oxygenation, and also for tuning the exchange coupling strength with the in-plane orientation layer. Figure 6.5(b) shows the hysteresis loops for the Co/Ni multilayer in different directions. The hysteresis loop shows almost a perfect squareness shape along the out-of-plane (OOP) direction, which is labeled as 90. Along in-plane (IP) direction, labeled as 0, the loop shows a straight line. Both these results indicate

that the material has an out-of-plane magnetic anisotropy. The thicker Co layer shows good in-plane magnetic anisotropy. The measured stack and the hysteresis loops for the Co film along IP and OOP directions are shown in Figures 6.5 (c) and (d). The loop measured in the IP direction shows almost a perfect square shape, and in the OOP direction, it shows a straight line, which indicates the material has in-plane magnetic anisotropy.

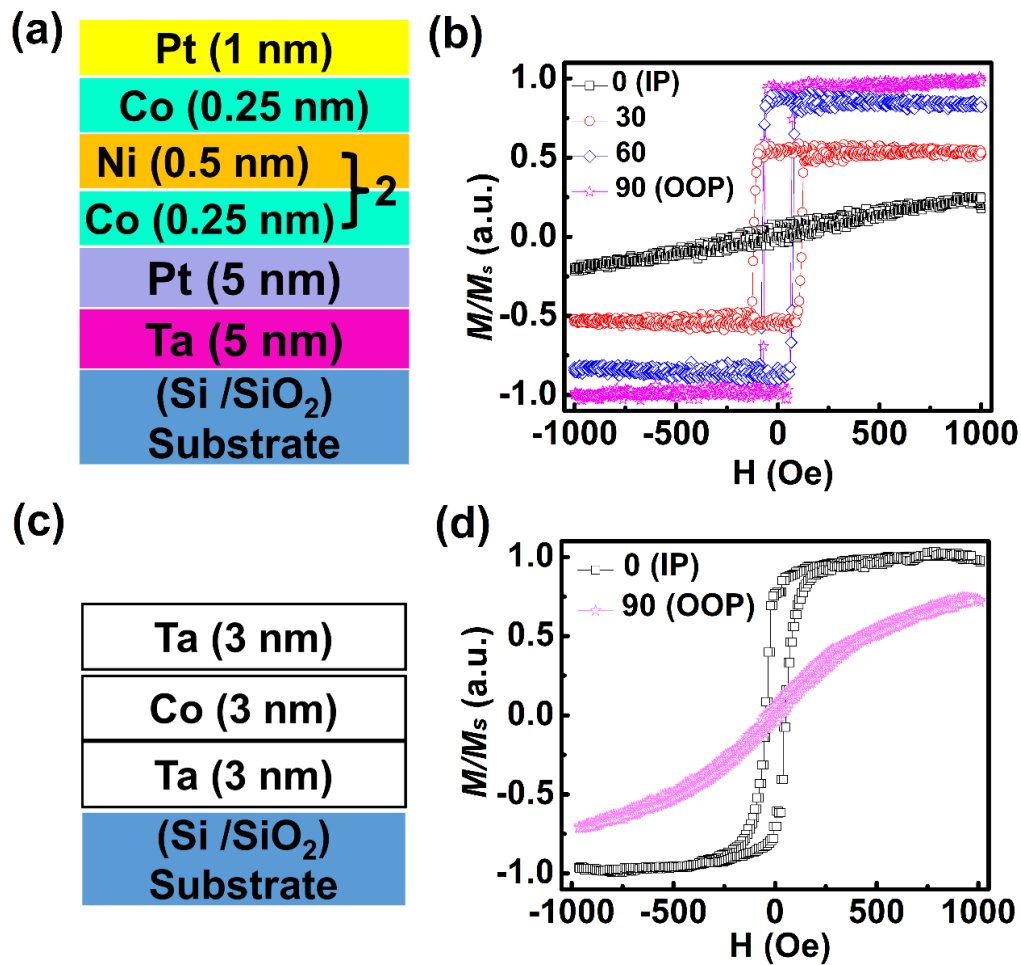


Figure 6.5. (a) and (c) The stack of the (Co/Ni)₂ multilayer and the Co layer. (b) The hysteresis loops for the (Co/Ni)₂ multilayer sample measured in applying a magnetic field along different directions from in-plane (IP), *i. e.* 0 degree, to out-of-plane (OOP), *i. e.* 90 degrees. (d) The hysteresis loops for the Co sample along IP and OOP directions.

(Co/Ni)₂ samples with the Pt thickness of 1nm and 2 nm both show excellent PMA properties, with an anisotropy field over 4.5 kOe. To study the

effect of magnetization tilting, 3 nm thick Co layer was deposited on the top of the $(\text{Co/Ni})_2$. A 5 nm thick Ta layer was grown on the top of the Co layer for avoiding oxidation. The stacks are shown in Figures 6.6 (a) and (b), which have different Pt spacer layer thickness as of 1 nm and 2 nm. The characterization of their magnetic properties was carried out by VSM. It is important to note that the thickness of the Pt spacer determines the exchange coupling and hence the extent of tilting of magnetization in the PMA layer. In the thin film level, the thickness of the Pt layer was optimized, and it was observed that the magnetization of the $(\text{Co/Ni})_2$ multilayer could be tilted for the Pt thickness of 1 nm, as shown in Figures 6.6 (a1) and (a2) (image (a2) is the zoom-in image of (a1)). However, when the Pt layer is 2 nm thick, the $(\text{Co/Ni})_2$ multilayers and the Co layer are magnetically decoupled, as shown in Figures 6.6 (b1) and (b2).

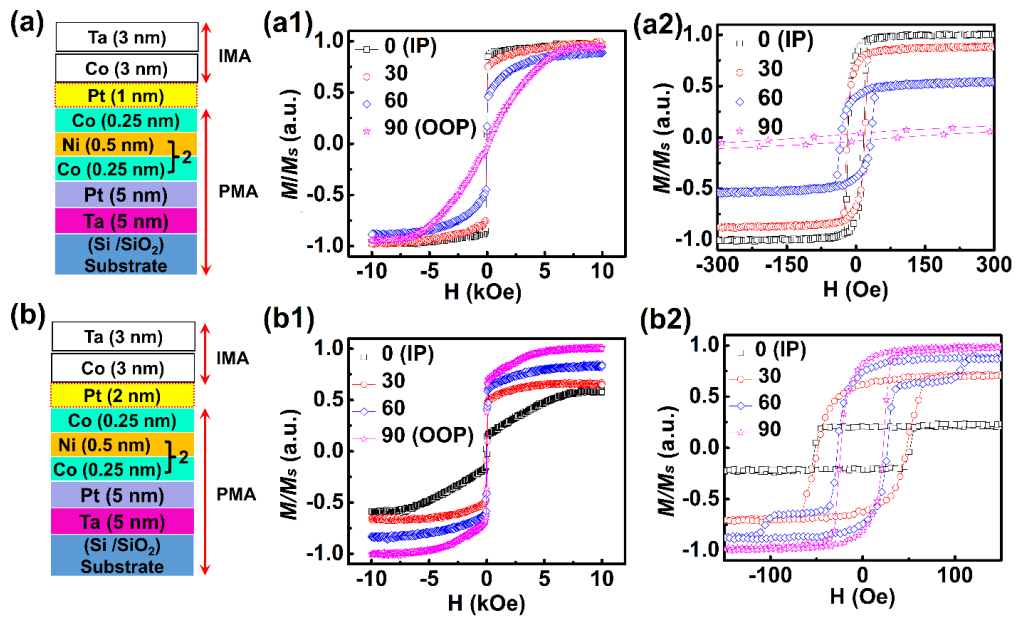


Figure 6.6. (a) and (b) The diagram of the $(\text{Co/Ni})_2$ multilayers with the IMA Co layer, and the Pt spacer layer to separate them is 1 nm and 2 nm. (a1) and (a2) The hysteresis loops for the $(\text{Co/Ni})_2$ multilayer sample with the Co layer, where the Pt spacer layer is 1 nm along different directions. (b1) and (b2) The hysteresis loops for the $(\text{Co/Ni})_2$ multilayer sample with the Co layer, where the Pt spacer layer is 2 nm along different applied field directions.

6.4 Estimation of the tilt angle using simulations

When the Pt spacer layer is 1 nm, the magnetic coupling between the (Co/Ni)₂ layer and the Co layer was observed. In order to estimate the tilt angle of the magnetization of the PMA layer, the normalized remanent magnetization (M_r/M_s) change was calculated with applied field angles, as shown in Figure 6.7. The value of M_r/M_s for the (Co/Ni)₂ PMA sample shows the highest along the OOP direction and the lowest along the IP direction, which confirms the excellent PMA properties of the (Co/Ni)₂ sample. For the (Co/Ni)₂ PMA coupled with the Co IMA layer, the value of M_r/M_s shows the highest along the IP direction and the lowest along the OOP direction. It was expected to have the maximum value at some particular angles, where the magnetization of the PMA layer would be tilted. However, the magnetization measured by VSM mostly corresponds to that of the Co IMA layer, whose thickness is larger than the PMA layer. Hence, the tilt angle could not be measured by VSM.

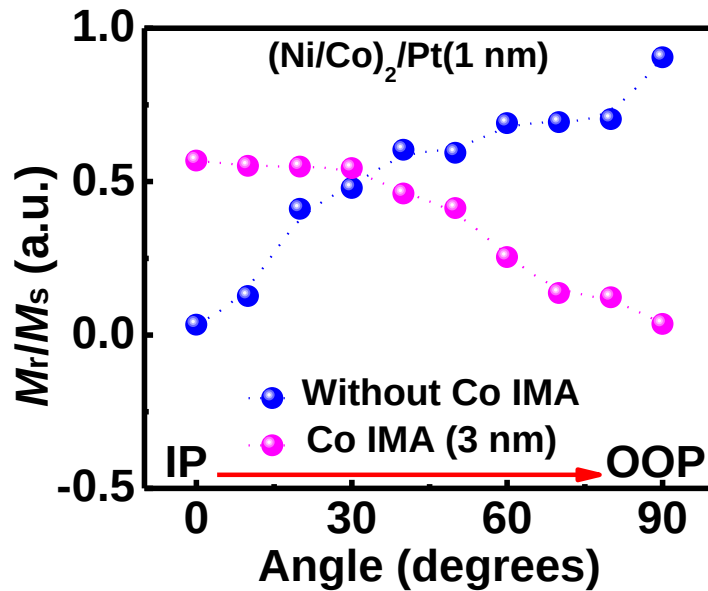


Figure 6.7. The normalized remanent magnetization (M_r/M_s) change with applied field angles.

Since it is difficult to estimate the tilt angle at the pinning site experimentally, the tilt angle was simulated by micromagnetic simulations using Mumax3 [2]. The cell size used is $1 \text{ nm} \times 1 \text{ nm} \times 1 \text{ nm}$. The material parameters of the IMA layer are: $M_{s1} = 1000 \text{ emu/cm}^3$, $K_{u1} = 2 \times 10^6 \text{ erg/cm}^3$ and $A_{ex1} = 10 \times 10^{-7} \text{ erg/cm}$, which are chosen based on those of a Co layer [6]. The magnetization of the IMA Co layer lies in the (0, 1, 0) direction and the thickness is 3 nm. The material parameters for the PMA layer are: $M_{s0} = 655 \text{ emu/cm}^3$, $K_{u0} = 3.74 \times 10^6 \text{ erg/cm}^3$ and $A_{ex0} = 15 \times 10^{-7} \text{ erg/cm}$ [6], which are chosen based on those of a Co/Ni multilayer reported in the literature [6]. The thickness of the PMA layer is 2 nm. The exchange coupling strength between the PMA and IMA layers, A_{ex10} , is defined as $A_{ex10} = S * A_{ex-average}$, where S is the scale factor, indicating the coupling strength and $A_{ex-average}$ is the average exchange coupling between the PMA and IMA layers. Figures 6.8 (a-e) show the captured magnetization images in the y-z plane with various scale factor S . With increasing S factor, the tilt of the magnetization of the PMA layer towards the y-direction is enhanced. When $S = 0.04$, the magnetization of the PMA layer is tilted to the in-plane direction, as shown in Figure 6.8(e).

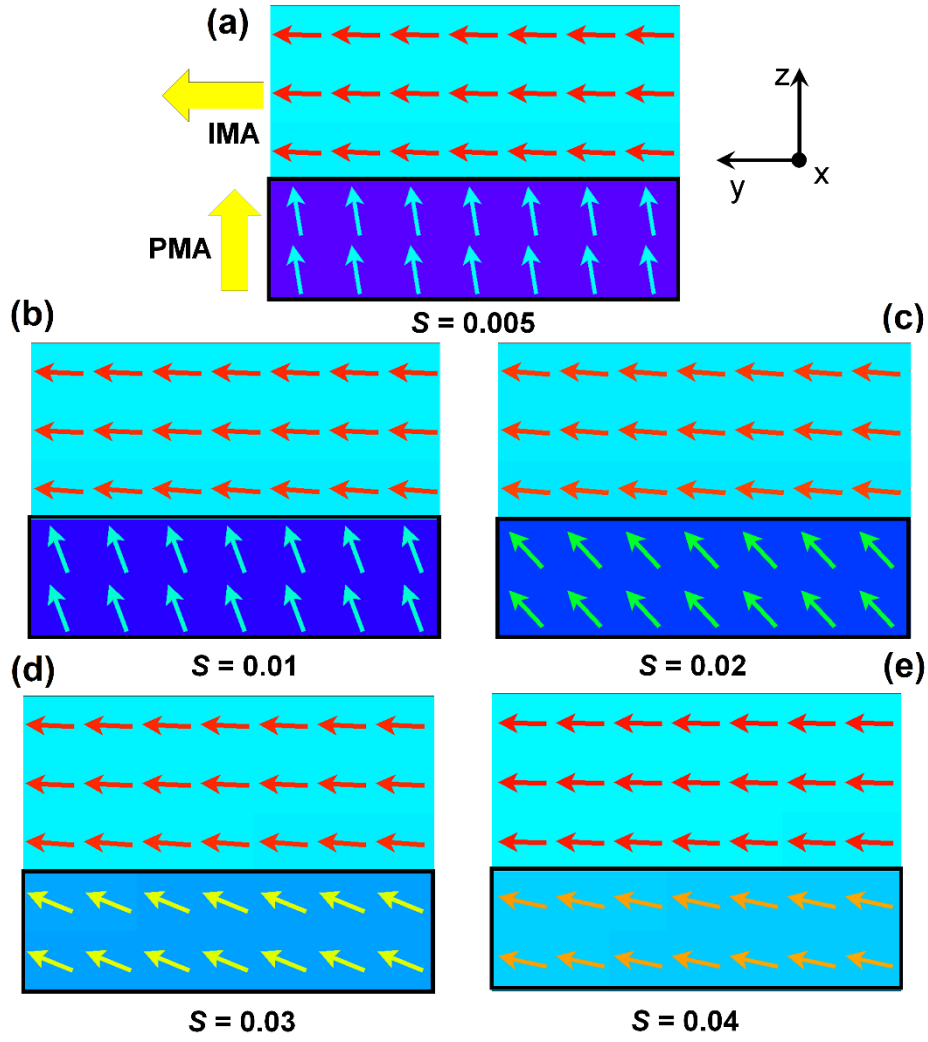


Figure 6.8. The magnetization configurations with different exchange coupling strength, which is defined by the scale factor S from 0.005 to 0.04. The bottom layer highlighted by the black box is the PMA layer and the top layer is the IMA layer.

The magnetization tilt angle θ is calculated at the remanence using the following equation:

$$\theta = \cos^{-1}(m_{\text{rem}}/m_{\text{sat}}) \quad (6.3)$$

where m_{rem} is the normalized remanence ($M_{\text{rem}}/M_{\text{sat}}$) and m_{sat} is the normalized saturation, equal to 1 in this simulation [5]. Figure 6.9(a) shows the hysteresis loops with different scale factor values. For the PMA layer without the Co IMA layer, the loop shows a perfect square shape. With the scale factor increasing, the switching field is reduced, and remanent magnetization is also reduced. Figure

6.9(b) is the zoom-in image for the hysteresis loops. The magnetization (M) of the PMA layer is extracted along with different directions and the component is plotted with scale factors. The magnetization along the x-direction (m_x) is almost zero. While the magnetization along the y-direction (m_y) is increasing with increasing scale factor S . To the contrary, the magnetization along the z-direction (m_z) is decreasing with increasing S , as shown in Figure 6.9(c). These results indicate the magnetization of the PMA layer is tilted to the y-direction with larger exchange coupling.

Figure 6.9(d) shows the tilted angles for various scale factors S . θ increases linearly in the range S less than 0.03 and then θ goes to 60 degrees. When $S = 0.04$, the magnetization of the PMA layer is fully tilted to the in-plane direction as discussed earlier, which is shown in Figure 6.8(e).

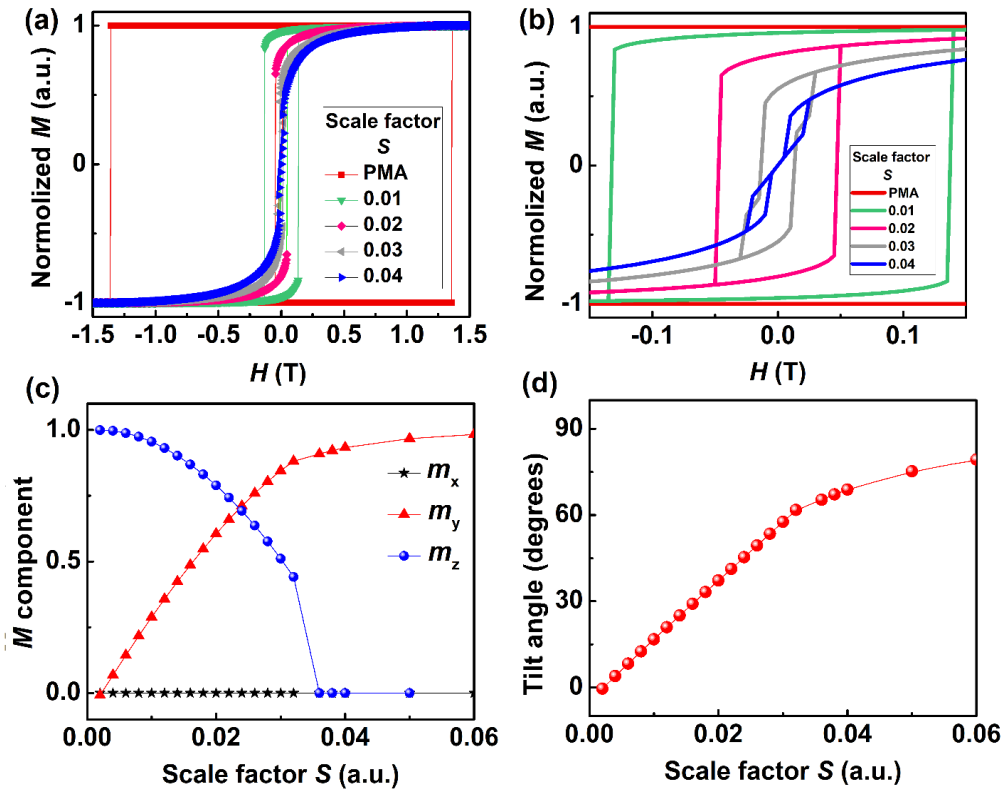


Figure 6.9. (a-b) The normalized hysteresis loops of the top PMA layer. (c) The magnetization (M) components and (d) the tilt angles calculated with various exchange coupling A_{ex10} . The exchange coupling A_{ex10} is defined as $A_{\text{ex10}} = S * A_{\text{ex-average}}$.

6.5 Domain wall pinning by tilting (Co/Ni)₂ magnetization

To test the pinning at the device level, (Co/Ni)₂ nanowires with a width of 1 μm and a length of 100 μm were fabricated with a Pt as a capping layer. Co crossbars with a width of 800 nm were fabricated to change the magnetization locally. A large pad was created at the left side of the wires for DW injection. The sample was saturated at a field of 5 kOe along the OOP direction to achieve a uniform domain state. Then a pulsed magnetic field of a certain magnitude and a duration of 1 ms was applied in the opposite direction to create a reversed domain. The domain was reversed and the DW was observed clearly using the MOKE microscope. The field at which the reverse domain is formed is different for samples with different Pt thickness. For the sample with the Pt thickness of 2 nm, the field required to nucleate a domain is -150 Oe and for the sample with the Pt thickness of 1 nm, it is -180 Oe.

After nucleating a reverse domain in the pads, the pulsed magnetic field was applied with increasing amplitude but keeping the same duration time. As can be seen in Figure 6.10(b), DW can be pinned at a field of -215 Oe and then depinned at a higher applied field *i.e.* -222 Oe, for the Pt thickness of 1 nm. On the contrary, when the Pt spacer layer thickness is 2 nm (as shown in Figure 6.10(a)), DW shows continuous motion *i.e.* DW does not get pinned. These results indicate that it is necessary to achieve optimum coupling to achieve domain wall pinning.

For the Pt thickness of 1 nm, the Co crossbars have the same dimensions everywhere. As a result, they have the same pinning field strength at every cross point. Therefore, for an applied magnetic field stronger than the pinning field strength, the domain walls are not pinned at the other crossbars. A similar result

was also seen by Al Bahri *et al.*, where the domain wall was pinned in staggered nanowire fabricated by two simple wires partially overlapping at the edges [7]. The pinning strength is able to be adjusted by changing the depth of step and length of the step. When the pinning sites have the same dimensions in depth and length, the domain wall just got pinned at the first pinning site and moved to the end after depinned at the first staggered position [7, 8].

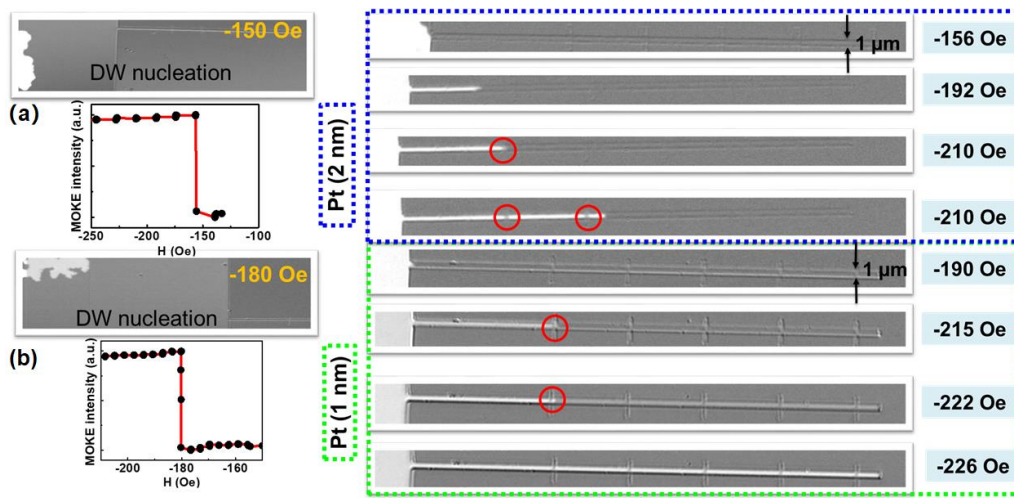


Figure 6.10. Domain wall was introduced by applying a magnetic pulse field. Then DW shows (a) pinning for the Pt thickness of 1 nm and (b) no pinning for the Pt thickness of 2 nm upon the application of the reverse magnetic pulse field.

Furthermore, current-driven DW motion was investigated in the device with Pt = 1 nm. The device was saturated firstly by applying 5 kOe along the OOP direction. Then applying a pulsed magnetic field of -100 Oe along the opposite direction, DW was injected at the large pad, which size is $20 \mu\text{m} \times 20 \mu\text{m}$. Then a series of pulsed currents, with a current density J , was passed through the wire in the direction indicated by the arrow J . Domain wall was found to move along the electron direction but get pinned at the pinning sites. Figure 6.11 shows the DW position for the different current densities, as noted on the left side of Figure 6.11. It can be observed that DW is pinned at every pinning site and

depinning with increasing current density. These results indicate that the proposed method also has the potential to pin the domain walls.

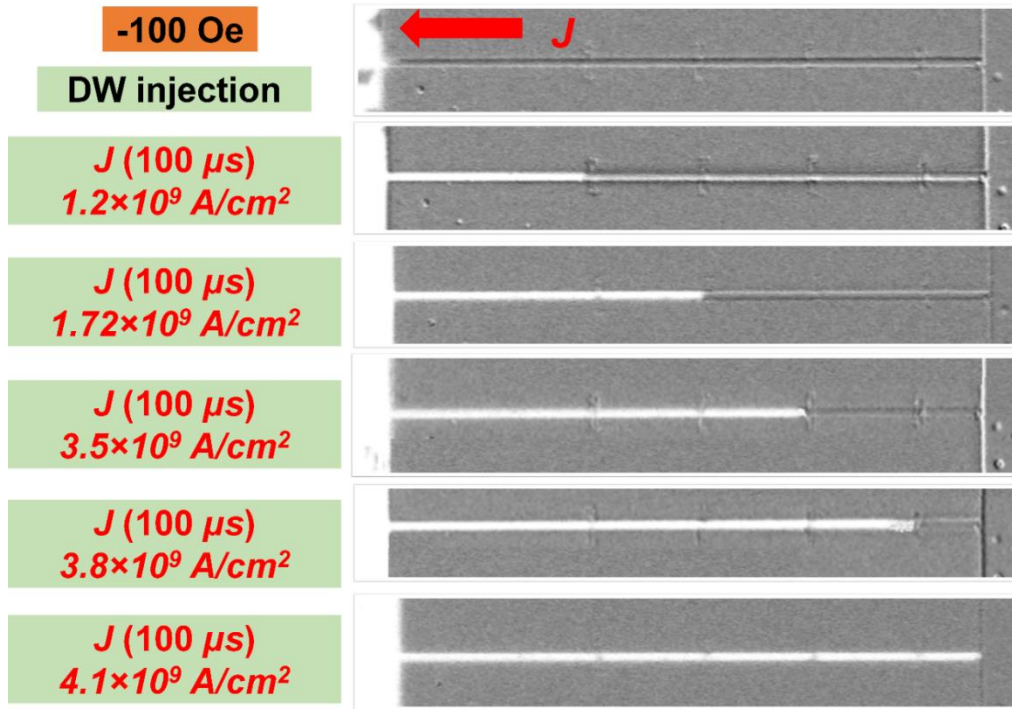


Figure 6.11. Domain wall pinning and depinning driven by current. The current density was noted at the left side.

6.6 Conclusion

In this chapter, I proposed that the pinning of domain wall can be achieved by tilt magnetization of the ferromagnetic nanowire and investigated the samples in a thin film and device form. The simulation results support the possibility of pinning domain wall using localized tilting of magnetization and that the depinning current increases with increasing tilt angle. In the experimental study, I showed that the tilted magnetization of the (Co/Ni)₂ PMA layer can be achieved by depositing a magnetic layer of Co with in-plane anisotropy. The space layer between PMA and IMA layer should be less than 1 nm to tune the exchange interaction strength. In the (Co/Ni)₂ devices, the pinning of a domain wall was

observed in the $(\text{Co/Ni})_2$ wire at the position of the crossbar, where its magnetization was tilted by a Co IMA layer due to the exchange coupling.

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Chapter 7 Summary and future work

7.1 Conclusions of this thesis

The spintronics based domain wall motion offers the possibility of 3-dimensional domain wall memory which combines the advantages of the high capacity in HDD as well as the low energy consumption and ultrafast operation in MRAM. However, some challenges still need to be addressed for commercial production. The suitable layout and the stable design of the 3-dimensional structure are still difficult to be figured out by now [1]. The more efficient method to drive domain wall motion is still required to achieve high-performance domain wall devices. Even though the velocity was reported to be as high as 750 m/s in the synthetic antiferromagnetic system and 800 m/s in the ferrimagnetic system, the current density required to drive current density is high $\sim 10^{12}$ A/m² [2-5]. Another important issue for the practical implementation of domain wall devices is the stochastic behavior of domain wall motion. In this thesis, I have investigated several novel methods of pinning the domain wall motion.

In the first study to control domain wall propagation, as investigated by T. Ono [6], they fabricated narrower regions in the nanowire to help to pin the domain wall position. Later creating notches in the nanowire to a pin domain wall was widely studied. In chapter 2, the studies about notches were reviewed in the ferromagnetic nanowire to pin a domain wall. It should be noted that the pinning is not uniform due to the variation in the property of the notches. The narrow region also causes Joule heating at the notches due to the increased resistance in the notch position. Furthermore, it requires the width of the nanowire larger than the notch width, limiting the areal density of domain wall memory [7]. In 2016, the staggered nanowire was proposed as the other method to pin the domain wall

[8]. The staggered nanowire is fabricated using two simple wires and partially overlapping at the edges. The pinning strength can be adjusted by changing the depth of a step and length of a step. All these methods, known as the geometrical methods, require a more complex lithography process and a high-resolution lithography technique.

In this thesis, I investigated three approaches to form domain wall pinning sites in different magnetized systems, which are based on forming synthetic magnetically textured nanostructures, named as the non-geometrical method.

(I) Localized metal-diffused $\text{Ni}_{80}\text{Fe}_{20}$ system. In this investigation, NiFe thin films were deposited with different metallic capping layers such as Ta, Ru, Cr, and Cu. Then the thin film samples were annealed in the vacuum chamber for an hour at different annealing temperatures T_{an} and their magnetic and structural properties were investigated. The results showed that Cr is more effective in changing the magnetic properties of NiFe. The uniform diffusion for Cr atoms into the NiFe layer forms the solid solution as observed by TEM. $\text{Ni}_{80}\text{Fe}_{20}$ domain wall devices were investigated using Kerr microscopy and magnetoresistance measurements as a gauge of domain wall pinning. Compared to homogeneous $\text{Ni}_{80}\text{Fe}_{20}$ wires without Cr-diffused pinning sites, $\text{Ni}_{80}\text{Fe}_{20}$ wires with Cr-diffused pinning sites, which exhibit modified properties with smaller magnetization and larger damping constant, show effective pinning. A novel R-H minor loop was measured from the magnetoresistance measurements, which also indicates that the Cr crossbars are effective in pinning the domain walls. The pinning field is about 100 Oe [9, 10].

(II) High-anisotropy Co/Pd multilayers system. Co/Pd multilayers were investigated due to their high magnetic anisotropy, which could be beneficial in

achieving the small size of the domain wall for a higher density. The properties of these multilayers can be easily tailored by changing the ratio of the thicknesses between ferromagnetic and non-magnetic layers, as well as their bilayers repeats. The method involving localized Cr diffusion into Co/Pd under high-temperature annealing was found to be unsuitable for domain wall pinning. It was found that the annealing modifies the properties of Co/Pd multilayers due to the intermixing of Co and Pd atoms. As an alternative method, localized B⁺ implantation through a mask was investigated. B⁺ implantation was found to be more effective in changing the easy direction of magnetization in the Co/Pd multilayers from out-of-plane to in-plane. The properties of the Co/Pd multilayer can be changed gradually by controlling the depth of B⁺ implanted. Moreover, the depth of B⁺ implantation can be tuned by depositing an additional Ta layer with different thickness to reduce the implanted energy into the Co/Pd layer. The pinning field was about 900 Oe. The micromagnetic analysis also confirmed the pinning effect by modification of composition and the corresponding depinning effect [11].

(III) Localized magnetization tilting in a Co/Ni multilayer system. In the PMA system, the pinning sites can also be formed by tilting its magnetization locally. The micromagnetic simulations showed the potential pinning effect by the tilted magnetization. The depinning current was found to increase with increasing tilt angle. The magnetization of the Co/Ni multilayer with a tilting angle, which initially points the out-of-plane direction, can be achieved by depositing an IMA Co layer with the optimized space layer between the PMA and IMA layers. When the Pt spacer layer is thin (thickness less than 1 nm), the magnetization of the Co/Ni multilayer is tilted by the Co IMA layer due to a strong exchange coupling. While the space layer is thick with a thickness above

2 nm, they are decoupled. The magnetization of these two layers switches independently. The tilted angle was estimated using micromagnetic simulations. The pinning and depinning of a domain wall in the Co/Ni devices, driven by an electric current, was observed by using polar Kerr microscopy in the samples with a thinner spacer Pt layer [12].

In this thesis, three approaches have been investigated to control domain wall propagation. The studies were carried out in different magnetic systems. Thermal annealing is easy to be conducted and various elements can be chosen to tune the magnetic properties of the nanowire to form pinning sites. However, this method requires high-temperature annealing, and it could destroy the magnetic property of certain films such as multilayer samples. Hence, this technique is not suitable for a multilayer sample. Ion-implantation is a good way to avoid this intermixing problem. Ion implantation is effective to locally change the properties of the nanowire to form pinning sites. The other issue emerges, that when the nanowire thickness is less than 2 nm, it is very hard to control the dose value of ions, as well as implanted energy. In this case, tilting the magnetization of the nanowire locally by using exchange coupling is a better method. However, this method requires that the anisotropy of the sample is not very strong. A suitable method has to be chosen based on the requirement of the final objective. Moreover, the investigation on controlling of domain wall motion is a growing field, and various proposed pinning approaches should be explored for the various requirements.

7.2 Suggestions for future work

Domain wall motion is ultrafast in the synthetic antiferromagnetic system, due to the exchange coupling torque. Moreover, the speed is up to several kilometers per second in ferrimagnetic materials. Those materials have almost zero net magnetization. Thus the reading process is a big challenge due to the weak stray field, and the stability of the domain wall is another issue for the ferrimagnetic system. The future work should explore new systems, which can provide a high velocity of domain wall motion and easy to be read out. The new configurations for reading sensors other than GMR and TMR have to be explored.

In the current study, we have explored pinning sites in the range of hundreds of nm. For future memory, uniform pinning of domain wall at sub-100 nanometer dimension needs to be explored in future work. Ion-implantation is a promising method that can be combined with self-assemble technology. Self-assembly was reported for fabricating the pinning trench with a width below 20 nm. After successfully patterning the trench, ion-implantation can be carried out to change the nanowire properties to form uniform pinning sites [13-15].

In an analogous domain wall tracked in the nanowire for the memory system, skyrmion based racetrack memory is also considered for future storage applications. While the smallest thermally stable domain in domain wall memory could be about 15 nm in length, skyrmions are stable at dimensions as small as 4 nm [16-19]. Therefore, skyrmions based racetrack memory could store information at much higher densities. Magnetic skyrmions are particle-like spin textures which are topologically protected. The chiral Dzyaloshinskii-Moriya interactions (DMIs) plays a vital role in generating skyrmions in magnetic systems. The critical current is very low for driving skyrmion motion, and it is

energy efficiency comparing to domain wall memory. However, more investigation of the stability and dynamics of skyrmions at room temperature without an external magnetic field is still needed in the future.

In recent years, there is a growing trend of research on neuromorphic computing. Synthetic neurons and synaptic elements are being researched. The domain wall devices studied in this thesis could be explored for synaptic elements. Multilevel resistance, resembling the learning process of synapses in the brain, could be achieved in domain wall devices [20-22]. For this purpose, the domain wall devices should be made in a magnetic tunnel junction form. For low power consumption, the domain walls could be moved using spin-orbit torque. In conclusion, the domain wall memory devices offer exciting possibilities and this is an interesting area of research.

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