Highly lattice-mismatched epitaxy for III-V/Si integration

Jia, Bowen

2018

Jia, B. (2018). Highly lattice-mismatched epitaxy for III-V/Si integration. Doctoral thesis, Nanyang Technological University, Singapore.

<http://hdl.handle.net/10356/73266>

<https://doi.org/10.32657/10356/73266>

**Highly Lattice-Mismatched Epitaxy
For
III-V/Si Integration**

Jia Bowen

School of Electrical & Electronic Engineering

A thesis submitted to the Nanyang Technological University
in partial fulfillment of the requirement for the degree of

Doctor of Philosophy

2018

Acknowledgements

Pursuing PhD degree is a long journey for every PhD candidate. At the end of the journey, I would like to express the depth of my gratitude to the people who have guided, inspired and helped me in the past few years.

First and foremost, I would like to gratefully acknowledge my supervisor Prof. Yoon Soon Fatt for his invaluable support and guidance. He provided me the excellent opportunity to investigate the interesting questions in the field of microelectronics and continuous support. His solid knowledge, critical thinking and passion in research also inspired and encouraged me during the entire PhD study.

I would like to express my gratitude to my TAC members—Prof. Zhang Qing and Assoc. Prof. Lew Wen Siang. They spend their valuable time on hearing my progress reports and gave me many valuable suggestions which were very important for this thesis.

The support from the members in our research team was indispensable for my research. Dr. Tan Kian Hua, Dr. Loke Wan Khai, and Dr. Satrio Wicaksono trained me how to use MBE and characterization equipment with sufficient patience. They provided generous help when I encountered difficulties. They also shared me lots of background knowledge and experience about material growth, characterization and device fabrication. Besides, I would like to express my gratitude to Dr. Lee Shuh Ying, Dr. Yeo Chiew Yong, Dr. Xu Zhe, Dr. Wang Cong, Mr. Li Daosheng and Mr. Leong Yurong Nelevin for their kindly assistance, insight suggestion and valuable discussions. This thesis would not have been finished without their constant support and strict training.

I would like to take this opportunity to thank my colleagues in NTU. I would like to give

Acknowledgments

special mention to Mr. Zheng Yi, a real geek in electronics, for guiding me in device fabrication and measurement; Dr. Tong Jingchao, Dr. Liang Guozheng and Dr. Hu Xiaonan for their kindly assistance in the measurement of photoresponsivity. I would like to thank Dr. Li Xiang, Zeng Yongquan, Li Xianqiang, Wu Dan, Kong Qinyu, Shi Yancheng and Gong Heqing for their friendship.

I would like to express my gratitude to collaborators from National University of Singapore (NUS), Prof. Yeo Yee Chia, Asst. Prof. Gong Xiao, Dr. Wang Wei, Dr. Dong Yuan, and Mr. Sachin for their support in technical resources and valuable discussions.

I would like to appreciate the technicians from the Characterization Laboratory and Nanyang Nanofabrication Centre (N²FC) of NTU, particularly Seet Lye Ping, Muhd Fauzi Bin Abudullah, Mohamad Shamsul Bin Mohamad, and Zulkiflee Bin Abdullah. Thanks for their efforts in managing the laboratories and for their training and assistance in my fabrication processes.

At the end, I would like to extend my deepest gratitude to my family for their love, support and encouragement. About 40 years ago, my grandfather and grandmother moved from Beijing to Hubei together with many engineers for an ambitious objective—building an automobile factory. Recently, the factory “Dongfeng Motor Corporation” has been one of the Fortune Global 500 companies. Their efforts and achievements always inspire and encourage me to overcome all difficulties in my research and life. My last and heartfelt gratitude to give my mother for her unconditional love and support that makes me feel confidence all along the journey of pursuing PhD degree.

List of Publications

Journal Articles :

1. **B.W. Jia**, K.H. Tan, W.K. Loke, S. Wicaksono, and S.F. Yoon, “Effect of in situ annealing on the structural and electrical properties and infrared photodetection of III-Sb on GaAs using interfacial misfit array”, *Applied Surface Science*, vol. 427, pp. 876-883, Jan. 2018.
2. **B.W. Jia**, K.H. Tan, W.K. Loke, S. Wicaksono, and S.F. Yoon, “Effects of surface reconstruction on the epitaxial growth of III-Sb on GaAs using interfacial misfit array”, *Applied Surface Science*, vol. 399, pp. 220-228, Mar. 2017.
3. **B.W. Jia**, K.H. Tan, W.K. Loke, S. Wicaksono, and S.F. Yoon, “Formation of periodic interfacial misfit dislocation array at the InSb/GaAs interface via surface anion exchange”, *Journal of Applied Physics*, vol. 120, pp. 035301, Jul. 2016.
4. **B.W. Jia**, K.H. Tan, W.K. Loke, S. Wicaksono, and S.F. Yoon, “Epitaxial growth of low threading dislocation density InSb on GaAs using self-assembled periodic interfacial misfit dislocations”, *Materials Letters*, vol. 158, pp. 258-261, Nov. 2015.
5. X. Gong, S. Yadav, K.H. Goh, K.H. Tan, A. Kumar, K.L. Low, **B.W. Jia**, S.F. Yoon, G. Liang, and Y.C. Yeo, “Enabling hetero-integration of III-V and Ge-based transistors on silicon with ultra-thin Buffers Formed by Interfacial Misfit Technique”, *ECS Transactions*, vol. 75, pp. 421-437, Aug. 2016.
6. K.H. Tan, **B.W. Jia**, W.K. Loke, S. Wicaksono, and S.F. Yoon, “Formation of interfacial misfit dislocation in GaSb/GaAs heteroepitaxy via anion exchange process”, *Journal of Crystal Growth*, vol. 427, pp. 80-86, Oct. 2015.
7. V.R. D'Costa, K.H. Tan, **B.W. Jia**, S.F. Yoon, and Y.C. Yeo, “Mid-infrared to ultraviolet optical properties of InSb grown on GaAs by molecular beam epitaxy”, *Journal of Applied Physics*, vol. 117, pp. 223106, Jun. 2015.

Conference Proceedings:

1. **B.W. Jia**, K.H. Tan, W.K. Loke, S. Wicaksono, and S.F. Yoon, “Hetero-integration

- of GaSb channel on silicon with ultra-thin buffer utilizing interfacial misfit dislocations”, in *2016 IEEE Silicon Nanoelectronics Workshop (SNW)*, pp. 158-159, June 2016.
2. S. Yadav, K.H. Tan, K.H. Goh, S. Subramanian, K.L. Low, N.Y. Chen, **B.W. Jia**, S.F. Yoon, G. Liang, X. Gong, and Y.C. Yeo, “First monolithic integration of Ge P-FETs and InAs N-FETs on silicon substrate: Sub-120 nm III-V buffer, sub-5 nm ultra-thin body, common raised S/D, and gate stack modules” in *2015 IEEE International Electron Devices Meeting (IEDM)*, pp. 2.3.1-2.3.4, Dec. 2015.

Submitted Articles:

1. **B.W. Jia**, K.H. Tan, W.K. Loke, S. Wicaksono, and S.F. Yoon, “Growth and characterization of InSb on (100) Si for mid-infrared application”, submitted to *Applied Surface Science*.
2. **B.W. Jia**, K.H. Tan, W.K. Loke, S. Wicaksono, and S.F. Yoon, “Growth and characterization of an InSb infrared photoconductor on Si via an AlSb/GaSb buffer”, submitted to *Journal of Crystal Growth*.
3. **B.W. Jia**, K.H. Tan, W.K. Loke, S. Wicaksono, K. H. Lee, and S.F. Yoon, “Monolithic Integration of InSb Photodetector on Silicon for Mid-Infrared Silicon Photonics”, submitted to *ACS Photonics*.
4. **B.W. Jia**, K.H. Tan, W.K. Loke, S. Wicaksono, and S.F. Yoon, “Mid-infrared InSb photodetector epitaxially grown on Si substrate via antimonide buffer”, submitted to *Optics Express*

Table of Contents

Acknowledgements.....	i
List of Publications	iii
Table of Contents	v
Summary	vii
List of Figures	x
List of Tables.....	xvi
List of Abbreviations.....	xvii
Chapter 1 Introduction	1
1.1 Background	1
1.2 Motivation and Objectives	7
1.3 Major Contributions	10
1.4 Thesis Organization.....	12
Chapter 2 Literature Review	15
2.1 Dislocation in Lattice Mismatched Heterostructure	15
2.2 Buffer Technologies	20
2.3 Interfacial Misfit Array	24
2.4 Infrared Photodetector.....	28
Chapter 3 Growth of InSb on GaAs Using Interfacial Misfit Array	39
3.1 Introduction	39
3.2 Experimental Details	41
3.3 XRD and AFM Measurement	43
3.4 TEM Observation.....	46
3.5 Formation of IMF Array at the InSb/GaAs Interface.....	53
3.6 Hall Measurement	55
3.7 Summary	60
Chapter 4 Growth Optimization of III-Sb on GaAs Using Interfacial Misfit Array.....	63
4.1 Effect of Pre-Growth Sb Reconstruction on GaAs Surface	63
4.1.1 Introduction	63
4.1.2 Experimental Details	65
4.1.3 XRD, AFM and TEM Measurement	67
4.1.4 X-ray RSM Simulation and Measurement	70

Table of Contents

4.1.5 Hall Measurement	77
4.1.6 Summary	83
4.2 Effect of In Situ Annealing.....	84
4.2.1 Introduction	84
4.2.2 Experimental Details	85
4.2.3 XRD, AFM and TEM Measurement	87
4.2.4 Hall Measurement	93
4.2.5 Photoresponsivity	97
4.2.6 Summary	100
Chapter 5 Integration of InSb Photodetector on Si <i>via</i> GaAs/Ge	101
5.1 Introduction	101
5.2 Experimental Details.....	102
5.3 Structural Characterization.....	104
5.4 Electrical Properties	108
5.5 Photoresponsivity	113
5.6 Summary	117
Chapter 6 Integration of InSb Photoconductor on Si <i>via</i> AlSb/GaSb	119
6.1 Introduction	119
6.2 Experimental Details.....	120
6.3 Structural Properties.....	124
6.4 Hall Measurement	134
6.5 Photoresponsivity.....	137
6.6 Summary	139
Chapter 7 Conclusions and Recommendations.....	141
7.1 Conclusions	141
7.2 Recommendations for Further Work.....	144
7.2.1 Epitaxy of InSb on GaAs.....	144
7.2.2 Epitaxy of InSb on Si <i>via</i> the AlSb/GaSb Buffer	145
7.2.3 InSb Photoconductor with a Barrier Layer.....	146
7.2.4 InSb p-i-n Photodetector on Si	147
Bibliography	149

Summary

Indium antimonide (InSb) is a competitive semiconductor for the applications in high electron mobility transistor and mid-infrared photodetector. Integration of InSb on GaAs and Si substrates is promising for higher performance of the devices and lower cost of fabrication, and becomes an important component of opto-electronic integrated circuits. Therefore, much effort has been made to investigate the novel heteroepitaxial approaches to overcome the high lattice mismatch (14.6% for InSb/GaAs and 19.3% for InSb/Si) and different crystal structures between InSb and Si.

The thesis focuses on the integration of InSb with Si using solid-state molecular beam epitaxy (MBE) system. First, interfacial misfit (IMF) array was intentionally formed at the InSb/GaAs interface to accommodate the lattice mismatch *via* surface anion exchange. The relationship between the growth temperature and the formation of IMF was investigated. Transmission electron microscope (TEM) images of the sample grown at 310 °C demonstrated an IMF array consisting of 90° misfit dislocations distributed uniformly along the interface and their separation (3.2 nm) agreed well with the calculated result. Growth temperature above 310 °C suppressed the surface anion exchange to impede the formation of IMF array. Below 310 °C, island coalescences resulted in the formation of threading dislocations.

Further optimization of growth process has been focused on two areas: 1) pre-growth Sb reconstructions; and 2) in situ thermal annealing. Sb atoms have three kinds of reconstruction on GaAs surface and III-Sb (GaSb and InSb) layers grew on GaAs with different pre-growth Sb reconstructions. Our simulation demonstrated that the x-ray reciprocal space map (RSM) can be employed to characterize the distribution of misfit

dislocations in the IMF array. The results of RSM showed pre-growth (2×8) Sb reconstruction promoted the formation of 90° misfit dislocations in an IMF array. The highest carrier mobility of III-Sb layers was achieved in the III-Sb layers grown on GaAs with pre-growth (2×8) Sb reconstruction. The optimized anneal process was 5 minutes at 590 °C for GaSb and 15 minutes at 420 °C for InSb, respectively. Compared with as-grown III-Sb layers, the lower density of threading dislocations, flatter surface and higher carrier mobility were observed in the annealed samples. We confirmed that the annealing also improved the photoresponsivity of GaSb and InSb photoconductors grown on GaAs.

To realize the integration of InSb on Si, GaAs buffer was grown on Ge-on-Si substrates with 6° offcut followed by InSb grown on GaAs *via* IMF array. An InSb p-i-n photodetector was developed based on this structure. An electron barrier layer was inserted into the conventional architecture of p-i-n photodetector to suppress the dark current. The electrical and optical properties of the photodetector were investigated. The photodetector owned a cutoff wavelength of about 5.3 μm at 80 K and the cutoff wavelength was found to increase with increasing temperature. The 80 K detectivity of the photodetector was $8.8 \times 10^9 \text{ cmHz}^{1/2} \text{W}^{-1}$ at 5.3 μm with the quantum efficiency of 16.3 %.

Another approach of the integration of InSb on Si through an AlSb/GaSb buffer was also investigated. In this buffer, GaSb was grown on Si using IMF array to avoid the formation of threading dislocations and InSb quantum dots (QDs) were grown on AlSb surface to decrease the density of microtwins. TEM images showed uniform IMF array existed at the GaSb/Si and QDs nucleated at the sites microtwins to terminate them. AlSb and GaSb were respectively applied on the top of the buffer layer to provide a

Summary

surface for the growth of InSb. InSb on AlSb surface demonstrated higher crystal quality and electron mobility than InSb on GaSb surface. Moreover, the photoresponsivity in InSb photoconductor on AlSb was higher than that on GaSb.

List of Figures

Figure 1.1 The energy gaps and lattice constants of IV and III-V group semiconductors. [2].....	1
Figure 1.2 Wavelength dependent absorption ranges for Si, Ge, InSb and common waveguide cladding materials. For IV group semiconductors, the light area is optical transparency (loss is less than 2 dB/cm), the dark orange area is high loss caused by the absorption of bandgap and the dark brown area is high loss caused by the absorption of the vibration of chemical bonds. For InSb, the edges of absorption at 77 K and 30 K are indicated using black triangles respectively. [3, 19]	4
Figure 2.1 Schematic diagrams of a lattice mismatched layer grown on a substrate: (a) a strained layer; (b) a relaxed layer where two misfit dislocations are indicated.	16
Figure 2.2 Three kinds of MD distribution in (100) grown semiconductor heterostructures: (a) pure 90° MDs; (b) 60° MDs with alternating orientations; (c) 60° MDs with the same orientations. [44]	17
Figure 2.3 Three schematic geometries of the Burgers vector of the different origins of threading dislocations: (a) mixed type threading dislocation from 60° MD; (b) edge type threading dislocation from 90° MD; (c) mixed type threading dislocation from 60° MD pair. The purple arrows indicate the Burgers vector of the 60° MD, the orange and red arrows represent the edge and screw components of the Burger vector. [51].....	18
Figure 2.4 Low magnification bright-field cross-sectional TEM images of InSb layers on (a) GaAs, (c) GaSb buffer, (e) InAlSb buffer, (g) step-graded GaSb+InAlSb. (b), (d), (f), and (h) are high magnification images corresponding to (a), (c), (e), and (g), respectively. All images were obtained near a two-beam condition with $g=220$ indicated by the black arrow in each image. The white arrows in (f) and (h) indicate the bending of threading dislocations at the interfaces between the InSb layers and buffers. [27]	20
Figure 2.5 InSb layer on GaAs <i>via</i> continuous graded buffer (CGB): (a) schematics of growth process of the sample; (b) cross-sectional TEM image across the sample. Eye-guide lines are inserted in the insets. [71]	22
Figure 2.6 Cross-sectional TEM images of GaSb on GaAs: (a) GaSb grown on AlSb buffer; (b) GaSb grown on the dislocation filter of AlSb/GaSb superlattice. [73]	23
Figure 2.7 Cross-sectional TEM images of GaSb grown on GaAs using IMF array: (a) low magnification, showing highly periodic IMF array at the GaSb/GaAs interface; (b) high magnification, indicating the spaces of misfit dislocations. [29]	24

Figure 2.8 The relationship between the calculated energy density and MD spacing. (a) The effect of misfit; (b) The effect of layer thickness. [45]	27
Figure 2.9 The schematic of a photoconductor with a planar architecture which are comprised of two metal contacts and a n-type semiconductor.....	30
Figure 2.10 The band diagrams of a PN junction showing (a) trap-assisted tunneling; (b) band to band tunneling.	34
Figure 2.11 The band diagrams of the nBn photodetector based on n-type semiconductor. (a) Electron barrier layer locates near the anode. (b) Electron barrier layer locates near the cathode.	36
Figure 2.12 The band diagrams of PN junction-based photodetector (a) without an electron barrier layer; (b) with an electron barrier layer at p-type side; (c) with an electron barrier layer at n-type side.	37
Figure 3.1 XRD scans of InSb grown on a (100) GaAs substrate from (004) reflection. The growth temperatures of InSb are indicated. (a) ω - 2θ scans; (b) ω scans.	43
Figure 3.2 $20\mu\text{m} \times 20\mu\text{m}$ AFM images of InSb grown on GaAs (100) substrate at (a) 350 °C; (b) 310 °C; (c) 270 °C; and (d) 210 °C.	45
Figure 3.3 Cross-sectional TEM images of InSb grown on (100) GaAs substrate at 310 °C. (a) A low magnification TEM image. The rich clusters are marked by a white circle. Threading dislocations are indicated by the white arrows; (b) TEM images of uniform interfacial misfit array; (c) TEM images showing the separation between two adjacent misfit dislocations.	46
Figure 3.4 High resolution TEM images of the InSb/GaAs interface in the InSb grown on GaAs at 310 °C. The black dashed arrows are a Burgers circuit around a 90° misfit dislocation with a Burger vector $b=a/2(1\bar{1}0)$. The box with a dotted line show the atomic sites between adjacent dislocations in InSb and GaAs, respectively.....	47
Figure 3.5 Cross-sectional TEM images of InSb grown on (100) GaAs substrate at 350 °C. (a) Bright field image of InSb/GaAs interface. A box indicates the area shown in (b); (b) high resolution image along $[1\bar{1}0]$ zone axis; (c) and (d) phase contrast images of selected areas C and D in (b) using Fourier mask filtering technique. At the interface, 60° misfit dislocations are indicated by one arrow and 90° misfit dislocations are indicated by two arrows.....	49
Figure 3.6 Cross-sectional TEM images of InSb grown on (100) GaAs substrate at 270 °C. (a) Bright field image of InSb/GaAs interface. A box indicates the area shown in (b); (b) high resolution image along $[1\bar{1}0]$ zone axis; (c) and (d) phase contrast images of selected areas C and D in (b) using Fourier mask filtering technique. At the interface, 60° misfit dislocations are indicated by one arrow and 90° misfit dislocations are indicated by two arrows.....	51
Figure 3.7 (a) A low magnification cross-sectional TEM image of InSb grown on (100)	

- GaAs substrate at 210 °C and one microtwin in InSb layer is indicates; (b) selected area electron diffraction (SAED) pattern taken from the InSb/GaAs interface. Spots belong to InSb and GaAs have been marked by white arrows. White dotted circles mark extra diffraction spots associated by mirotwins in InSb layer.52
- Figure 3.8 Dependence of (a) carrier mobility and (b) carrier concentration on growth temperatures for InSb grown on GaAs (100) substrate at 300 K and 77 K. .56
- Figure 4.1 (a) & (b) symmetric (004) ω - 2θ scans for GaSb and InSb on GaAs samples, respectively. The Sb reconstructions before growth are indicated; (c) & (d) Cross-sectional TEM of the interfacial misfit dislocations for GaSb and InSb grown on GaAs samples with (2 $\times$ 8) Sb reconstruction prior to growth, respectively.....67
- Figure 4.2 AFM images (20 $\mu\text{m} \times 20 \mu\text{m}$) of GaSb and InSb grown on (100) GaAs substrate. (a) - (d) 425 nm GaSb on GaAs with different Sb reconstructions before growth: (a) mixed (2 $\times$ 4)-(2 $\times$ 8); (b) (2 $\times$ 8); (c) mixed (2 $\times$ 8)-(1 $\times$ 3) and (d) (1 $\times$ 3). (e) - (f) 700 nm InSb on GaAs with different Sb reconstructions before growth: (a) (2 $\times$ 8) and (1 $\times$ 3).69
- Figure 4.3 Calculated (004) XRD RSMs for GaSb on GaAs: (a) pure 60° MD; (b) pure 90° MD; (c) scans along and perpendicular to the diffraction vectors for pure 60° MD and 90° MD (red curves) ; (d) Pure 90 ° MD with uniform separations ($\gamma=0.1$). 72
- Figure 4.4 Measured (004) RSMs of GaSb and InSb peaks of samples grown at different Sb surface reconstructions. (a)-(d) GaSb on GaAs with different Sb reconstructions before growth: (a) mixed (2 $\times$ 4)-(2 $\times$ 8); (b) (2 $\times$ 8); (c) mixed (2 $\times$ 8)-(1 $\times$ 3) and (d) (1 $\times$ 3), respectively. (e)-(f) InSb on GaAs with different Sb reconstructions before growth: (e) (2 $\times$ 8) and (f) (1 $\times$ 3). 74
- Figure 4.5 Dependence of hole mobility and concentration on Sb absorption temperatures for 425 nm GaSb grown on (100) GaAs substrate at (a) 300 K and (b) 77 K. The corresponding pre-growth Sb reconstructions are indicated. 77
- Figure 4.6 Temperature dependence of (a) electron mobility and (b) concentration of 700 nm InSb grown on (100) GaAs with (1 $\times$ 3) and (2 $\times$ 8) pre-growth Sb reconstruction before InSb growth, respectively. The inset in (a) shows electron mobility below 200 K..... 79
- Figure 4.7 Schematics of GaSb- and InSb-based photoconductors grown on (100) GaAs substrates.85
- Figure 4.8 Symmetric (004) XRD scan of III-Sb on GaAs. (a) & (b) symmetric (004) ω - 2θ scans for GaSb and InSb, respectively. The annealing times are indicated; (c) The annealing time dependent full-width-half-maximum (FWHM) from (004) ω scans of GaSb and InSb.87
- Figure 4.9 Annealing time dependent r. m. s. surface roughness for III-Sb layers on
-

GaAs. The insets are AFM images ($20\text{ }\mu\text{m} \times 20\text{ }\mu\text{m}$) for samples with different annealing times. (a) GaSb on GaAs; (b) InSb on GaAs.....	89
Figure 4.10 Cross-sectional TEM images and x-ray RSM of the GaSb grown on GaAs with annealing for 5 minutes. (a) Bright field image; (b) High resolution TEM image where the interfacial misfit array is indicated by black arrows; (c) Asymmetric (115) RSM where reciprocal lattice GaSb peak positions for fully relaxed and strained epitaxial layer on GaAs are indicated by dashed lines, respectively.....	91
Figure 4.11 Cross-sectional TEM images and x-ray RSM of the InSb grown on GaAs with annealing for 15 minutes. (a) Bright field image; (b) High resolution TEM image where the interfacial misfit array is indicated by black arrows; (c) Asymmetric (115) RSM where reciprocal lattice InSb peak positions for fully relaxed and strained epitaxial layer on GaAs are indicated by dashed lines, respectively.....	93
Figure 4.12 Temperature dependence of (a) electron mobility and (b) concentration of as-grown and annealed InSb on GaAs. The inset in (a) shows mobility below 180 K. The inset in (b) shows concentration between 150 K and 300 K. The solid black line indicates the intrinsic electron concentration calculated by Equation (4.6).....	95
Figure 4.13 Plot of DC relative responsivity vs wavelength for the as-grown and annealed (a) GaSb- and (b) InSb-based photoconductors at applied voltage of 0.5 V. The temperatures of measurement are 300 K for GaSb and 77 K for InSb. The inset in (a) is optical image of GaSb-based photoconductor. The inset in (b) is responsivity vs applied voltage for the annealed InSb-based photoconductor at 77 and 140 K.	98
Figure 5.1 The schematic of InSb photodetector on a Si substrate. (a) The heteroepitaxial structure of InSb grown on a 6° offcut Si substrate <i>via</i> the GaAs/Ge buffer. (b) The structure of InSb p-i-n photodetector with an $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ barrier layer.	102
Figure 5.2 (a) A $10\text{ }\mu\text{m} \times 10\text{ }\mu\text{m}$ AFM image of InSb layer grown on Si. (b) Optical microscopic image of the fabricated InSb p-i-n photodetector on Si.....	104
Figure 5.3 XRD characterization of InSb grown on Si. (a) (115) reciprocal space map (RSM), (b) (004) ω scan. All diffraction peaks are indicated.	105
Figure 5.4 (a) Bright-field TEM image for GaAs layer grown on the Ge-on-Si substrate. (b) Bright-field TEM image of InSb layer grown on GaAs where a highly periodic interfacial misfit (IMF) array locating at the InSb/GaAs interface. (c) High resolution TEM image of the IMF array shown in (b) and the separations (3.2 nm) of misfit dislocations are indicated.....	107
Figure 5.5 (a) Bias voltage dependent dark current density (J_{dark}) at different temperature. (b) and (c) The natural logarithm of $\ln(J_{\text{dark}})$ vs. $1/kT$ for the device presented in (a) and the solid lines are the Arrhenius fitting using Equation (5.1) from 300 K to 60 K (b) and from 100 K to 5 K (c).	109

- Figure 5.6 (a) and (b) The relationship between the perimeter to area (P/A) ratio and $1/R_0A$ for photodetectors with different sizes at 80 K(a) and 200 K(b). The solid lines are the fitting using Equation (5.2) and the fitted results are shown in (a) and (b), respectively. (c) 80 K dark current density for the photodetector with and without $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ barrier layer. 111
- Figure 5.7 (a) Spectral responsivity of the InSb photodetector as a function of wavelength at different temperatures. (b) The detail of the wavelengths near the absorption edge of InSb at different temperatures where the maximum value of each curve is normalized to 1..... 114
- Figure 5.8 (a) The absolute spectral responsivity, quantum efficiency, and (b) the calculated spectral detectivity of the InSb photodetector at 80 K..... 115
- Figure 6.1 Schematic of InSb-based photoconductors grown on (100) Si substrates. (b) STEM image of InSb on AlSb buffer. (c) STEM image of InSb on GaSb buffer. 122
- Figure 6.2 (a) The FWHM obtained from (004) ω scans of GaSb on Si with different growth conditions. The inset in (a) is ω - 2θ scans of GaSb on Si with and without AlSb prelayer. (b) The wavelength dependent optical absorption for GaSb on Si with and without AlSb prelayer. AFM image of GaSb on Si without AlSb prelayer (c) and with AlSb prelayer (d). 125
- Figure 6.3 High resolution TEM images of (a) GaSb grown directly on Si; (b) GaSb grown on Si with AlSb prelayer. 126
- Figure 6.4 X-ray (115) reciprocal space maps of (a) InSb on AlSb; (b) InSb on GaSb surface. AFM images of (a) InSb on AlSb; (b) InSb on GaSb surface. 128
- Figure 6.5 (a) & (b) Bright-field TEM images of InSb grown on (100) Si: (a) the AlSb/GaSb buffer containing InSb QDs; (b) InSb grown on AlSb. (c) high resolution TEM images of an InSb QD at the GaSb/AlSb interface. (d) and (e) diffraction patterns obtained using fast Fourier transform (FFT) in the area D and E which are indicated by white boxes. 130
- Figure 6.6 Bright-field TEM images for (a) InSb/AlSb interface, (b) InSb/GaSb interface. (c) and (d) phase contrast images of (a) and (b), respectively by using Fourier mask filtering technique. 132
- Figure 6.7 Temperature dependence of (a) electron mobility and (b) concentration of InSb on Si with different buffer structures. The inset in (a) shows mobility from 80 K to 220 K. Intrinsic concentration was calculated using Equation (2) and presented in (b). (c) and (d) The $\ln(\sigma)$ v.s. $1/T$ for (c) InSb on GaSb and (d) InSb on AlSb. σ is the dark conductivity obtained from Hall measurement. The red solid lines are the Arrhenius fitting using Equation (6.3) from 450 K to 230 K and the blue solid lines are the Arrhenius fitting using Equation (6.3) from 140 K to 77 K. 134
- Figure 6.8 Photocurrent for (a) InSb photoconductor grown on AlSb and (b) InSb photoconductor grown on GaSb at 77 K and 140 K, respectively. The inset is
-

List of Figures

microscopic optical image of the InSb photoconductor. (c) Temperature dependent photocurrent at 1.0 V from 77 K to 150 for InSb photoconductor grown on AlSb and GaSb, respectively. (d) The output spectrum of a black body maintained at 700 °C. 137

Figure 7.1 A proposed schematic of opto-electronic integrated circuit (OEIC) using InSb photodetectors on Si as optical signal receivers..... 148

List of Tables

Table 3.1 Overview of XRD, AFM and Hall resistance measurement results for InSb/GaAs heteroepitaxy at different growth temperatures.	43
Table 4.1 Pre-growth Sb reconstructions, lattice constants and relaxations for GaSb and InSb grown on GaAs with different Sb absorption temperatures before growth.	68
Table 4.2 Pre-growth Sb reconstructions and surface roughnesses for GaSb/GaAs and InSb/GaAs grown with different Sb absorption temperatures before growth.	69
Table 4.3 Pre-growth Sb reconstructions and ratios of FWHM in scans along (ΔQ_x) and perpendicular to (ΔQ_z) the diffraction vectors in RSMs for GaSb/GaAs and InSb/GaAs grown with different Sb absorption temperatures before growth.	75
Table 4.4 Lattice constant, relaxation, (004) rocking curve FWHM, and surface roughness for GaSb/GaAs and InSb/GaAs grown with different annealing times.	88
Table 4.5 300 K and 77 K carrier mobility and concentration of III-Sb on GaAs with different annealing durations.....	94
Table 6.1 The details of samples grown in this chapter	121
Table 6.2 Lattice constants, relaxation, FWHM from (004) ω scan, 300 K carrier mobility and concentration of InSb on AlSb and InSb on GaSb.....	127

List of Abbreviations

Abbreviation	Full Expression
APB	Anti-Phase Boundary
AFM	Atomic Force Microscopy
BEP	Beam Equivalent Pressure
CMOS	Complementary Metal Oxide Semiconductor
CGB	Continuous Graded Buffer
CVD	Chemical Vapor Deposition
D*	Detectivity
DC	Direct Current
EDX	Energy Dispersive X-ray
EPD	Etch Pit Density
FFT	Fast Fourier Transform
FPA	Focal-Plane Array
FTIR	Fourier Transform Infrared
FWHM	Full-Width-Half-Maximum
G-R	Generation Recombination
HBT	Heterojunction Bipolar Transistor
HEMT	High-Electron-Mobility Transistor
HR-TEM	High Resolution Transmission Electron Microscopy
IMF	Interfacial Misfit Array
I-V	Current-Voltage
OEIC	Opto-Electronic Integrated Circuit
QD	Quantum Dots
RHEED	Reflection High-Energy Electron Diffraction
RSM	Reciprocal Space Map
MBE	Molecular Beam Epitaxy
MD	Misfit Dislocation
MEE	Migration-Enhanced Epitaxy
MWIR	Mid-Wavelength Infrared
r.m.s.	Root Mean Square

List of Abbreviations

SAED	Selected Area Electron Diffraction
SGB	Step graded Buffer
SOI	Silicon-on-Insulator
SI	Semi-Insulating
SRH	Shockley-Read-Hall
SNR	Signal-Noise-Ratio
STM	Scanning Tunneling Microscopy
TD	Threading Dislocation
TDD	Threading Dislocation Density
TEM	Transmission Electron Microscopy
XRD	X-ray Diffraction

Chapter 1 Introduction

1.1 Background

In the last decades, antimony (Sb)-based semiconductors have been widely investigated due to their physical properties, such as high carrier (hole and electron) mobility and direct and narrow energy gap [1]. Sb-based devices have been proposed for use in mixed-signal circuits, low-noise high-frequency amplifiers, and infrared photonic devices to satisfy the fast-growing needs in communication, data processing, environmental monitoring, imaging, and biomedicine [2, 3]. Sb-based semiconductors include AlSb, GaSb, InSb, and their alloys. Indium arsenide (InAs) is always presented with GaSb and AlSb because it has nearly the same lattice constant. These three semiconductors are classified as the “6.1 Å family” in the literature [4]. Their physical properties are shown in Figure 1.1.

Figure 1.1 The energy gaps and lattice constants of IV and III-V group semiconductors. [2]

Compared with other Sb-based semiconductors, InSb has the smallest bandgap (0.17 eV at 300 K) and highest electron mobility (7,7000 cm²/Vs at 300 K). Therefore, it has great

potential in high frequency high electron mobility transistors (HEMTs), heterojunction bipolar transistors (HBTs), and photodetectors. The 300 K intrinsic concentration is $2.3 \times 10^{16} \text{ cm}^{-3}$ [5] and such high concentration results in high gate leakage current [2]. Introducing barrier layer or multi-gate techniques can mitigate the effect of high concentration on the gate leakage [6, 7]. The InSb-based HEMT was developed using an InSb channel and $\text{In}_{0.8}\text{Al}_{0.2}\text{Sb}$ barrier. Its cutoff frequency was up to 305 GHz at the drain-source voltage of 0.5 V [8]. Such high frequency and low operation voltage cannot be achieved with GaAs-and InP-based HEMTs [2]. A simulation suggested the cutoff frequency of InSb HEMT could exceed 1.1 THz, which identified it as an important device for rapid growing terahertz technology [9]. Therefore, InSb and other Sb-based semiconductors are considered the platform for the next generation HEMTs and HBTs [2].

A mid-wavelength infrared (MWIR) photodetector is another crucial application for InSb. Infrared radiation is an invisible electromagnetic radiation with a wavelength from 700 nm to 1 mm, which is divided into four parts: 1) near infrared (700 nm - 3 μm), 2) mid-wavelength infrared (3 μm - 8 μm), 3) long-wavelength infrared (8 μm – 15 μm), and 4) far infrared (beyond 15 μm). Infrared detection has been widely applied for optical communication, environment, deep space exploration, biomedicine, and the military [3]. A wide atmospheric window exists at 3 μm to 5 μm where infrared radiation is not absorbed by CO_2 and H_2O . Therefore, infrared radiation can transmit freely in the atmosphere [10]. Photodetectors for this range are always used in thermal imaging, environmental monitoring, and security monitoring [11].

Traditionally, HgCdTe (mercury cadmium telluride) is widely used in a commercial infrared photodetector due to its tunable bandgap, high electron mobility, and easy

doping techniques. Its cutoff frequency can be tunable from 1 μm to 30 μm through changing the Cd/Te ratio [12]. However, as an alloy, due to the high vapor pressure of Hg, instability and non-uniformity over a large area are two main disadvantages of HgCdTe [11]. As mentioned above, InSb is a direct energy gap semiconductor with a band gap of 0.17 eV at 300 K and 0.23 eV at 77 K. These features enable the InSb-based photodetector to cover the whole MWIR region, which is something no other III-V group semiconductor can do. As a binary compound semiconductor, the composition and chemical properties of InSb are very stable and the preparation techniques of the InSb wafer are also commercially available [13]. A detectivity of 10^{11} cm Hz/W at 77 K was achieved in an InSb bulk photodetector [14] and a 3-dB bandwidth of 8.5 GHz was achieved in an InSb-based photodetector on GaAs, which is designed for high frequency application [15].

Although InSb photodetectors have been developed, the integration of it on a (100) Si substrate has the exciting potential to decrease cost, improve performance, and expand application. The price of a 2-inch InSb wafer is more than US\$590, while it is about US\$100 for GaAs and US\$30 for a Si wafer of the same size. Moreover, the maximum size of a commercial InSb wafer is only 3 inches while 8 inches or larger are common sizes for Si. Beside the less cost of substrates, developing devices on a larger substrate is also approach to decrease the cost of fabrication. To realize a thermal imaging device such as a focal-plane array (FPA), both read out circuits and infrared photodetectors are required. Traditionally, circuits and photodetectors are fabricated on Si and InSb substrates separately, and the two wafers are bonded together using flip-chip technology [16]. This technology is limited by the small size of the InSb wafer and requires a complicated process. Therefore, monolithic integration of InSb photodetectors and Si circuits on a single wafer will result in device size minimization and cost reduction.

On the other hand, a semi-insulating (SI) wafer is very difficult to be obtained for InSb while it is easy for silicon-on-insulator (SOI). Using a SOI wafer as the substrate can improve the performance of InSb-based devices at high frequency because it minimizes the intrinsic parasitic capacitances in the devices [17].

Since the 1990s, silicon photonics have been regarded as an emerging technology for information technology. Due to some intrinsic shortages of Si and Ge, the integration of III-V compound semiconductors on Si substrates is promising for the realization of the opto-electronic integrated circuit (OEIC) [18]. Recently, the concept of “mid infrared silicon photonics” was proposed and some related devices were developed [19-21].

Figure 1.2 Wavelength dependent absorption ranges for Si, Ge, InSb and common waveguide cladding materials. For IV group semiconductors, the light area is optical transparency (loss is less than 2 dB/cm), the dark orange area is high loss caused by the absorption of bandgap and the dark brown area is high loss caused by the absorption of the vibration of chemical bonds. For InSb, the edges of absorption at 77 K and 30 K are indicated using black triangles respectively. [3, 19]

Figure 1.2 shows the infrared absorption ranges for common semiconductors and waveguide cladding materials. The absorption at the shorter wavelength is due to the bandgap of the material, while the absorption at the longer wavelength is due to the vibration of chemical bonds in the material [22]. The 2-5 μm region is a window for transmitting infrared light in the waveguide with low loss, and this region nearly coincides with the atmosphere window (3-5 μm), which enables a secure free-space communication. Due to its absorption edges (marked by black triangles in Figure 1.2), the InSb photodetector on Si has the potential to be a signal receiver in an OEIC.

However, as shown in Figure 1.1, the lattice constant of Si is much smaller than that of InSb. Their structures are also different. Therefore, integration of InSb on Si is a big challenge. Generally, there are two paths to integrating lattice mismatched materials: 1) heteroepitaxial growth and 2) wafer bonding. Wafer bonding integrates two different materials through forming an oxidized layer between them. Some optoelectronic devices have been fabricated by this method [23]. Due to a high annealing temperature or plasma process during wafer bonding, mismatch in the thermal expansion coefficient still results in defects in the wafers. Moreover, wafer bonding requires the wafer of semiconductors, which limits its application in mass production, because the InSb wafer is small and very expensive. Heteroepitaxial growth can form different kinds of heterostructures through depositing material in sequence. However, it will result in a high density of defects for two dissimilar materials, such as the anti-phase boundary (APB) and threading dislocation (TD) [24]. These defects increase the dark current, which degrades the photodetector's performance [25]. For the heteroepitaxial growth of InSb on Si, Sb atoms prefer to bond with Si, while In atoms form metallic islands on the substrate [26]. Polycrystalline InSb can be grown on Si at low temperature; however there has been no report about a single crystalline InSb layer grown on Si directly. Using appropriate

buffers and a growth process, we are able to avoid InSb to contact with the Si surface and decrease defects to some extent [27]. Due to a high lattice mismatch and different structures between InSb and Si, the buffer structure and the growth process should be carefully designed. A buffer that is too thick is not favorable in device design and fabrication, because it results in devices that do not exist at the same level and difficulties in connecting them.

Forming an interfacial misfit (IMF) array at the GaSb/GaAs interface to grow a high quality GaSb layer on a (100) GaAs substrate has been reported and named the “IMF method” [28, 29]. Using IMF method, the strain energy caused by lattice mismatch is relieved by uniform IMF array at the interface rather than disordered threading dislocations. Simulation has suggested that the IMF method could be applied in all lattice mismatched material growth and the formation of an IMF array strongly depends on the interaction between deposited atoms and surface absorbed atoms on the substrate [30]. This method has also been applied in AlSb/Si and GaSb/Si [31, 32]. The GaSb or AlSb on a Si substrate provides a platform for all “6.1 Å family” compound semiconductors, and some novel electron and photonic devices can be developed based on these semiconductors. Recently, the experimental research about InSb on GaAs using this method has not been demonstrated. A detailed review about the IMF method is provided in Chapter 2.3.

Optimizing the photodetector’s architecture is another path to increase its performance. The performance of the photodetector strongly depends on its dark current. The unipolar barrier is an intentionally designed heterostructure in the diode, which can make only one carrier type (either electron or hole) flow unimpeded while the other carrier type is blocked [33]. Placing the barrier at an appropriate place in the diode can decrease the

dark current, which has been confirmed in some infrared photodetectors [34]. In-depth analysis about the origin of the dark current and the function of the unipolar barrier will be reviewed in Chapter 2.4.

1.2 Motivation and Objectives

InSb is a competitive material for application in the MWIR band. The integration of InSb on a (100) Si substrate and the development of devices on this structure has great potential to decrease fabrication cost, improve device performance, and become an important component in MWIR silicon photonics. Heteroepitaxial growth of high quality InSb layer on Si requires to bridge the gaps in lattice constants and crystal structures between InSb and Si. Recent research about this integration is in the initial stage, and some challenges exist:

1. The heteroepitaxial growth of InSb on GaAs or Si is an approach to realize the mass production of InSb devices on a GaAs or Si platform. However, a high quality InSb layer on Si and GaAs without a thick buffer is difficult to obtain. The IMF method is only used in “6.1 Å family” compound semiconductors on GaAs and Si, such as GaSb/GaAs, AlSb/Si, and GaSb/Si heteroepitaxy, and its effect on InSb on GaAs is still unknown. The relationship between the InSb layer’s properties and growth conditions also requires research to determine the optimized growth condition.
2. The mechanism of the formation of the IMF array should be further investigated. Although previous research pointed out that the formation of the IMF array is only driven by strain energy caused by lattice mismatch, this conclusion ignored the initial island-like growth during heteroepitaxy and the different microstructures of Sb

reconstructions on GaAs before growth. Deeply understanding its mechanism makes contribution to the optimization of the growth process.

3. The design and growth of a new buffer structure for the growth of InSb on Si is based on the IMF method. The IMF method can accommodate a large lattice mismatch without a thick buffer layer. Forming an IMF array at the interface can dramatically decrease the thickness of the buffer and increase the crystal quality of the InSb layer.
4. The report about an InSb-based device on GaAs and Si is rare. It is essential to develop an InSb device on GaAs and Si to demonstrate the potential of the heteroepitaxial structure of the InSb layer on GaAs and Si obtained in this thesis.

Driven by the above motivations, the five objectives of this thesis are:

1. Epitaxially grow InSb layers on (100) GaAs substrates using solid state molecular beam epitaxy (MBE) and characterize the samples' structural and electrical properties. The growth of InSb on GaAs is a prepositive work for the growth of InSb on Si, because a high-quality GaAs layer can be grown on Ge using migration-enhanced epitaxy (MEE) and a Ge-on-Si substrate has been fully developed [35, 36]. Therefore, the GaAs/Ge structure may be an appropriate buffer for the growth of InSb on Si. An IMF array will be formed intentionally to accommodate the lattice mismatch and avoid using a thick buffer layer. The effect of growth temperature on the formation of the IMF array at the InSb/GaAs interface and the effect of IMF array on carrier mobility and concentration are investigated.
2. Investigate the influence of Sb reconstruction on the GaAs surface before the growth of GaSb and InSb layers on GaAs. Sb atoms can form different reconstructions on the GaAs surface at different Sb absorption temperatures. Because absorbing Sb atoms

on the GaAs surface is an important step for the formation of the IMF array at the GaSb/GaAs interface, optimizing the pre-growth Sb reconstruction can further increase the uniformity of the IMF array at the GaSb/GaAs or InSb/GaAs interface. The effect of pre-growth Sb reconstruction on the structural and electrical properties of GaSb and InSb layers on GaAs are investigated.

3. Apply in situ annealing on GaSb and InSb layers on GaAs to further decrease the threading dislocation in the layers. In situ thermal annealing is a widely used method to promote the movements of threading dislocations to annihilate them. For the GaSb and InSb layer grown on GaAs, the process of annealing should be designed carefully to avoid the influence on the misfit dislocations on the IMF array. The effect of in situ annealing on the structural and electrical properties of GaSb and InSb layers on GaAs are investigated. Moreover, GaSb and InSb photoconductors on GaAs are grown and fabricated on GaAs to investigate the effect of in situ annealing on their responsivity.
4. Grow and fabricate InSb-based p-i-n photodetectors on Si substrates *via* GaAs and Ge buffers. The GaAs buffer is grown on a (100) Ge-buffered Si substrate with a 6° offcut acting as a buffer for the subsequent growth of the InSb device structure. The lattice mismatch between InSb and GaAs is accommodated using the IMF array. An InAlSb barrier layer is inserted into the architecture of photodetector to suppress the dark current. The structural, electrical, and optical properties of the InSb photodetector are characterized.
5. Grow and fabricate InSb photoconductors on a (100) Si substrate with a 0° offcut through an AlSb/GaSb buffer, which includes InSb quantum dots (QDs). Compared with the GaAs/Ge buffer structure, this buffer may cost less and avoid contamination because all growth is processed in molecular beam epitaxy (MBE) while the growth

of Ge on Si requires chemical vapor deposition (CVD). The lattice mismatch between InSb and AlSb is smaller than that between InSb and GaAs, which may decrease the density of dislocations. GaSb is first grown on Si using an interfacial misfit array followed by the growth of InSb QDs on AlSb to decrease the density of microtwins. InSb layers are respectively grown on AlSb and GaSb surface to compare the effect of two kinds of interfaces on the crystal quality of InSb layer. The structural and electrical properties of the InSb layer are characterized. InSb photoconductors on (100) Si are fabricated and measured to demonstrate the early application of this heteroepitaxial structure.

1.3 Major Contributions

1. The detailed mechanisms about the formation of the IMF array at the III-Sb/GaAs interface was investigated. The surface anion exchange process was identified as important for the formation of the IMF array. Sb anions replaced As ones on the GaAs surface to form a periodic misfit dislocation network. Therefore, ordered pre-growth (2×8) Sb reconstruction on GaAs increased the uniformity of the IMF array. Suppressing the density of islands formed at the initial stage of epitaxial growth could avoid the formation of threading dislocations, because the islands owned different misfit dislocation networks.
2. The optimal growth process of III-Sb layers on a (100) GaAs substrate was achieved. The key points in this process included: 1) Form (2×8) Sb reconstruction on GaAs before the growth of III-Sb layers. 2) The optimal growth temperature for the InSb on GaAs is 310 °C. 3) A 5 minutes in situ annealing for the GaSb buffer on GaAs at 590 °C. A 15 minutes in situ annealing for the InSb buffer on GaAs at 420 °C. All layers were totally relaxed and their surface roughness was approximately 1.0 nm.

The HR-TEM observation confirmed a highly uniform IMF array at both the GaSb/GaAs and InSb/GaAs interfaces and no obvious threading dislocation arising from the interfaces. The uniform IMF array relieved most of the strain energy caused by the lattice mismatch. At 300 K, a hole mobility of 503 cm²/Vs was achieved in a 425 nm GaSb layer, and an electron mobility of 29040 cm²/Vs was achieved in a 700 nm InSb layer. GaSb and InSb photoconductors on GaAs showed high gain and a steeper absorption edge at 77 K, which held great promise for application in infrared photodetection.

3. Totally relaxed InSb layers were grown on Si *via* the AlSb/GaSb buffer. The buffer structure was intentionally designed to accommodate lattice mismatch and suppress microtwins, which included a GaSb buffer on Si for the subsequent buffer growth, InSb quantum dots (QDs) for the suppression of microtwins, and an AlSb or GaSb top layer for the growth of the InSb layer. The optimal growth condition for GaSb on Si was achieved with the growth of 2 monolayers of AlSb prior to the growth of GaSb. The growth temperature of GaSb is 430 °C. Transmission electron microscopy (TEM) observation confirmed that the InSb QDs, which were intentionally formed at the AlSb surface, could suppress the microtwins by covering their surfaces. The IMF array was more uniform at the InSb/AlSb interface than at the InSb/GaSb interface. X-ray diffraction (XRD) and the Hall measurement indicated the crystal quality of the InSb layer on AlSb is higher than that on GaSb, and an electron mobility of 20202 cm²/Vs was achieved in InSb on AlSb at 300 K. InSb photoconductors on Si were developed based on these heterostructures, which could operate below 150 K and demonstrated standard photoconductive behavior. The photoresponsivity of the InSb photoconductor on AlSb is about 32% higher than that on GaSb under the applied voltage of 1.0 V at 77 K.

4. The abnormal temperature dependent carrier mobility and concentration were observed in InSb on GaAs and Si *via* the AlSb/GaSb buffer. The origin of this phenomenon is analysis associated with the presence of high density dislocations at the bottom of the InSb layer. The dislocations existing at or near the InSb/GaAs, InSb/AlSb, and InSb/GaSb interfaces contribute to a high density of extrinsic carriers, which results in the temperature dependent carrier mobility and concentration of the InSb layer being different from those of the InSb bulk.
5. An InSb p-i-n photodetector was grown on (100) Si with a 6° offcut *via* a GaAs/Ge buffer. The GaAs buffer was first grown on the lattice-matched Ge-on-Si substrate, then the InSb device layer was grown on a GaAs buffer *via* IMF array. Material characterization showed the device layer owned a smooth surface and the InSb layer was nearly 100% relaxed. Most of the lattice mismatch was accommodated *via* the IMF array at the InSb/GaAs interface. The dark current limiting mechanism was both generation-recombination and surface leakage from 300 K to 120 K and only surface leakage from 120 K to 40 K. The highest responsivity of 0.7 A/W and detectivity of $8.8 \times 10^9 \text{ cmHz}^{1/2} \text{W}^{-1}$ were achieved at 5.3 μm at 80 K.

1.4 Thesis Organization

The thesis contains seven main chapters:

Chapter 1 introduces the motivation, objectives, and major contributions of the research project presented in this thesis.

Chapter 2 presents the literature review about the theory of the dislocations in the lattice-mismatched heteroepitaxial growth and the progress of the integration of III-Sb (GaSb

and InSb) on GaAs and Si substrates. The background of the MWIR photodetector is also given.

Chapter 3 reports the growth and characterization of InSb grown on a (100) GaAs substrate using the IMF method. Under different growth temperatures, the structural and electrical properties of InSb layers are investigated using XRD, atomic force microscopy (AFM), TEM, and the Hall measurement.

Chapter 4 reports two approaches to improve the quality of III-Sb (GaSb and InSb) layers on (100) GaAs using an IMF array and demonstrates III-Sb photoconductors on GaAs. The first one is optimizing the Sb reconstruction on GaAs before III-Sb growth. The second one is applying in situ annealing on the as-grown III-Sb layers. The effect of the optimized strategies on the III-Sb layers is evaluated through comparing the structural and electrical properties of III-Sb layers. The III-Sb photoconductors on GaAs are grown using the optimized growth process and their photoresponsivities are evaluated.

Chapter 5 reports the heteroepitaxial growth and fabrication of an InSb p-i-n photodetector on a Si substrate using a GaAs/Ge buffer. The GaAs buffer was grown on a Ge-on-Si substrate through the migration-enhanced epitaxy (MEE) growth mode, and the photodetector structure was subsequently grown on GaAs using the IMF method, which was optimized in Chapters 3 and 4. The electrical and optical properties of the InSb photodetectors on Si were evaluated at different temperatures. The dark current limiting mechanism and the approaches to increase the detectivity of the InSb detector on Si are also discussed.

Chapter 6 reports the InSb photoconductor on a Si substrate using the GaSb/AlSb buffer. The structure and growth process of the buffer are optimized to achieve a high quality InSb layer. The structural and electrical properties of the InSb layer are characterized. Moreover, the photoconductivity results of the InSb photoconductor on Si are also investigated.

Chapter 7 summarizes the results accomplished in this thesis and suggests possible further works.

Chapter 2 Literature Review

2.1 Dislocation in Lattice Mismatched Heterostructure

The heterostructures consisted of two or more kinds of semiconductors with dissimilar crystal structures in a fundamental platform for state-of-the-art devices, such as high electron mobility transistors (HEMT), photodetectors, and lasers [37]. The lattice mismatch between different semiconductors results in threading dislocations. Threading dislocations are dislocations passing through the interior of a layer. In the epitaxial growth, threading dislocations are concomitantly generated with misfit dislocations (MDs) [38], and they may annihilate, bend, or multiply during the growth [39, 40].

Misfit dislocations originate from a mismatch between the lattice constants or thermal expansions of two different materials [23]. For a highly lattice mismatched systems ($>7\%$), such as GaSb/GaAs, InSb/GaAs, and InSb/Si, lattice mismatch dominates the generation of the misfit dislocation. The lattice mismatch (f) is defined as shown in Equation (2.1) [41].

$$f = \frac{a_0 - a_s}{a_s} \times 100\% \quad (2.1)$$

where a_0 is the lattice constant of the layer and a_s is that of the substrate. At the initial stage of growth, an elastic strain occurs at the layer as shown in Figure 2.1(a). Generally, there are two kinds of strain. When the lattice constant of the substrate is smaller than that of the layer, the strain is tensile in the vertical direction, as shown in Figure 2.1(a). On the contrary, the vertical strain is compressive when the lattice constant of the substrate is larger than that of the layer. The layer is coherent and no dislocation forms in this stage, which is called as “pseudomorphic growth”. The pseudomorphic growth

is always utilized to grow very thin lattice mismatched layers, such as the superlattice [42]. However, the strain energy is released by MDs once the layer's thickness exceeds a specific level called the "critical thickness (h_c)."

Most MDs form at the interface and the layer recovers to its intrinsic lattice constant, which is called the "relaxed layer," as shown in Figure 2.1(a). In this case, most of the lattice mismatch is accommodated by MDs and a MD network is formed at the interface.

Figure 2.1 Schematic diagrams of a lattice mismatched layer grown on a substrate: (a) a strained layer; (b) a relaxed layer where two misfit dislocations are indicated.

Most IV and III-V group semiconductors are diamond-like or zinc-blend structures, and there are normally two kinds of MDs: 60° and 90° MDs [43]. The 90° MD (or Lomer dislocation) is a pure edge dislocation which is sessile on the (110) plane. The 60° MD with a Burger vector inclined 45° to the interface is a mixed dislocation consisting of edge and screw dislocation, and it can glide on the (111) plane easily. Compared with 90° MDs, 60° MDs can only relieve about half of the strain energy [44, 45].

The theoretical predication of h_c is important to determine the strain layer's thickness. The calculation finished by Matthew *et al.* suggested that the h_c could be expressed as shown in Equation (2.2) [46].

$$h_c = \frac{b}{4\pi f(1+\nu)} \left[\ln \left(\frac{h_c}{b} \right) + 1 \right] \quad (2.2)$$

where b is the absolute value of the Burger vector of 90° MD, f is the lattice mismatch,

and ν is the Poisson ratio of the layer. Some experimental research showed that the real critical thickness was higher than the calculated one [47], however, a minimum thickness can still be obtained using this equation.

Figure 2.2 Three kinds of MD distribution in (100) grown semiconductor heterostructures: (a) pure 90° MDs; (b) 60° MDs with alternating orientations; (c) 60° MDs with the same orientations. [44]

Figure 2.2 shows three different kinds of MD distribution on the interface based on the assumptions that MDs accommodate all lattice mismatches and layers grown on a (100) substrate. Because 90° MDs can relieve more strain energy, their density on the interface is smaller than 60° MDs, as shown in Figure 2.2(a) and (b). For a relaxed GaSb layer grown on (100) GaAs, the spacing of 90° MDs is about 5.6 nm [28] while the spacing of 60° MDs is about 3.1 nm [44], which has been confirmed using a high-resolution transmission electron microscope (HR-TEM). If the majority of 60° MDs own the same orientation, a tilt will occur in the layer, which has been observed along the edge areas of GaSb islands grown on GaAs and is shown in Figure 2.2(c) [44]. The island-like growth mode is attributed to the formation of 60° MDs with the same orientation [44].

The selectivity of forming 60° or 90° MDs is a lack of a solid explanation. A conventional opinion suggested the formation of a 90° dislocation was more favorable

in a highly mismatched system, because of their high efficiency on relieving strain energy. Dregia *et al.* suggested that two gliding 60° MDs reacted to one 90° dislocation [48]. Some research suggested a 60° misfit dislocations were introduced at the edges of islands under the island-like growth mode [49, 50].

Figure 2.3 Three schematic geometries of the Burgers vector of the different origins of threading dislocations: (a) mixed type threading dislocation from 60° MD; (b) edge type threading dislocation from 90° MD; (c) mixed type threading dislocation from 60° MD pair. The purple arrows indicate the Burgers vector of the 60° MD, the orange and red arrows represent the edge and screw components of the Burger vector. [51]

The relationship between threading dislocation and misfit dislocation has been suggested and observed [38, 41]. Through systematical HR-TEM observation, Wang *et al.* summarized three possible processes for forming threading dislocations in the GaSb/GaAs system, as shown in Figure 2.3 [51]. Because its Burgers vector and itself lie in the (111) lattice plane, a 60° MD can easily glide along the (111) plane and the detailed geometry is demonstrated in Figure 2.2(a). A 60° MD owns a dislocation line along the $[1-10]$ direction and Burgers vectors of $a/2 [101]$. During layer growth, this dislocation propagates through gliding along (111) and forming a threading dislocation. Unlike a 60° MD, the 90° MD is sessile in the (001) lattice plane and does not glide in the (111) lattice plane. Figure 2.2(b) shows a 90° MD with a Burgers vector of $a/2[110]$

split into two 60° dislocations with Burgers vectors of $a/2$ [101] and $a/2$ [01-1]. Then these two 60° dislocations form two mixed type threading dislocations, respectively, as the first mechanism. The third mechanism is about 60° MD pair. As shown in Figure 2.2(c), a pair of 60° MDs with Burgers vectors of $a/2$ [101] and $a/2$ [10-1] and their total screw component is a [100]. During propagation, this MD pair separates into two independent 60° MDs to form mixed type threading dislocations. In sum, 60° MDs are the main source of the formation of threading dislocations. Thus, suppressing the formation of 60° MDs can effectively decrease threading dislocation density (TDD).

The characterization of threading dislocation is another challenging issue due to its nanoscale sizes. TEM can directly observe threading dislocations and the detailed atomic arrangement around the dislocation can be obtained by HR-TEM. TEM observation is very useful to obtain the detailed images of TDs and analyze their origins; however, it is difficult to obtain the macroscopic information of TDs by using TEM due to its limited measured area. The etch pit density (EPD) is another approach to observe threading dislocation directly [52]. Due to a higher etch rate in an area with dislocations, pits can form on the sample surface after applying an etch solution. These pits can be observed by an optical microscope. However, this method tends to underestimate the TDD for a sample with a TDD higher than $5 \times 10^6 \text{ cm}^{-2}$ [53].

X-ray diffraction (XRD) is a non-destructive method to characterize dislocations. In a single crystalline layer, threading dislocations destroy the periodic atomic arrangement to cause a small strain field around it [54, 55]. These phenomena contributes to broaden the full-width-half-maximum (FWHM) of an omega (ω) scan from the rocking curve: therefore, a higher FWHM implies a higher TDD in the sample [54]. Ayers *et al.* obtained the TDD in GaAs/Si by measuring ω scans on 10 different lattice planes [53].

Shalimov *et al.* utilized a ω scan to estimate the density and type of misfit dislocations in GaAs on Si [56]. Researchers also utilized XRD reciprocal space mapping (RSM) to investigate the types and arrangements of misfit dislocations in the semiconductor layers [57, 58].

2.2 Buffer Technologies

Figure 2.4 Low magnification bright-field cross-sectional TEM images of InSb layers on (a) GaAs, (c) GaSb buffer, (e) InAlSb buffer, (g) step-graded GaSb+InAlSb. (b), (d), (f), and (h) are high magnification images corresponding to (a), (c), (e), and (g), respectively. All images were obtained near a two-beam condition with $g=220$ indicated by the black arrow in each image. The white arrows in (f) and (h) indicate the bending of threading dislocations at the interfaces between the InSb layers and buffers. [27]

The growth of layers on the lattice-mismatched substrate inevitably results in threading dislocations, which were introduced in Chapter 2.1. Designing and growing appropriate buffer structures between two lattice mismatched materials can decrease the lattice mismatch to suppress the threading dislocation density (TDD). The lattice constant of the buffer material is between that of substrates and layers, and most of the dislocations are confined in the buffer layers [59]. For the epitaxial growth of InSb on GaAs, many kinds of material have been selected as a buffer layer, such as GaSb, AlSb, InAsSb, and

AlGaAsSb [27, 60, 61]. Figure 2.4 shows cross-sectional TEM images of InSb grown on GaAs directly and *via* different kinds of buffer layers.

In Figure 2.4(a), many threading dislocations arising from the InSb/GaAs interface are observed, and some of them pass through the InSb layer. In Figure 2.4(b), the threading dislocation annihilations reduce TDDs with increasing InSb layer thickness. This phenomenon can be described quantitatively using the scaling law where the threading dislocation density is inversely proportional to the layer thickness [62]. In Figure 2.4(c), the TDD in the GaSb layer is significantly lower than that in the InSb layer. A high magnification image of the GaSb/GaAs and InSb/GaAs interfaces in Figure 2.4(d) illustrates that the TDD arising from the InSb/GaSb interface is higher than that arising from the GaSb/GaAs interface, and even higher than that arising from the InSb/GaAs interface, as shown in Figure 2.5(b). This situation may be due to the different mechanisms about the formation or the reaction of threading dislocations for the InSb/GaSb interface [63]. InAlSb is a widely used buffer material with a lattice constant similar to InSb. Figure 2.4(e) and (f) shows that threading dislocations are contained in the InAlSb buffer and most of them bend on the InSb/InAlSb interface (indicated by white arrows). Figure 2.4(g) and (h) demonstrates that TDD does not decrease after inserting one more GaSb buffer layer. It is clear that containing threading dislocations in the buffer layer and bending them on the interface are two main mechanisms for reducing TDD in InSb layer by buffer technologies [27]. InSb grown at low temperature is also a buffer for the growth of InSb on GaAs. Davis, *et al.* grew 30 nm InSb buffer at 300 °C, then, continued to grow 5.5 μm InSb at 380 °C where a 77 K mobility of 35,000 cm^2/Vs was achieved [64].

For the growth of InSb on Si, GaAs and AlSb are the main buffer material because AlSb

and GaAs can be grown directly on Si [65, 66]. Chyi, *et al.* compared the growth of InSb on (100) Si substrate with and without 300 nm GaAs buffer and found a 300 K mobility of 3.2 μm InSb on Si with GaAs buffer was 23% higher than that of the sample without GaAs buffer [67]. Li, *et al.* utilized 300 AlSb buffer to grow 3 μm InSb on Si where a 300 K mobility of 55,000 cm^2/Vs was achieved [68]. Moreover, CaF_2 was used as buffer to grow 4.2 μm InSb on (100) Si and a 300 K mobility of 9,973 cm^2/Vs was achieved under the optimized growth conditions [69]. However, a rough surface and a high density of defects have been observed in the InSb layer on Si *via* GaAs or the AlSb buffer because the crystal quality of GaAs and AlSb is not high enough and the lattice mismatch between InSb and GaAs or AlSb is still obvious (InSb/GaAs: 14.61% and InSb/AlSb: 5.60%) [65, 70].

Figure 2.5 InSb layer on GaAs *via* continuous graded buffer (CGB): (a) schematics of growth process of the sample; (b) cross-sectional TEM image across the sample. Eye-guide lines are inserted in the insets. [71]

The lattice constant of InSb is significantly higher than that of other III-V group semiconductors. Therefore, a buffer composed of a single kind of material is inevitable to form high density dislocations on the interface between InSb and the buffer or the buffer and the substrate. A graded buffer is a structure with a composition changing from

bottom to top gradually. The lattice constant of the buffer layer's top pproximates that of the layer and the lattice constant of the bottom approximates that of the substrate.

Generally, the graded buffer can be divided into a step-graded buffer (SGB) and continuous graded buffer (CGB). In a SGB, the composition of the buffer layer is changed step by step. For example, Sato *et al.* grew 1 μm thick $\text{In}_x\text{Al}_{1-x}\text{Sb}$ SGB consisting of 9 different layers ($x=0.1$ to 0.9) [72]. In a CGB, the composition of buffer changes linearly. For instance, as shown in Figure 2.5(a), the $\text{In}_x\text{Al}_{1-x}\text{Sb}$ CGB growth process is divided into two stages [71]. At the first stage of $\text{In}_x\text{Al}_{1-x}\text{Sb}$ growth, the In flux, which is equivalent to the In growth rate, increases linearly from 0, while the Al flux, which is equivalent to the Al growth rate of $1.45 \text{ \AA/s}$, and the growth temperature are stable. Following this growth process, the In ratio increases in CGB. Due to the lower growth temperature of InSb rather than AlSb, the growth temperature decreases to 400°C gradually at the second stage. Meanwhile, the In flux is fixed and the Al flux decreases to 0 linearly to grow a pure InSb layer, which can guarantee that the top layer of the CGB is totally lattice-matched to InSb. Figure 2.5(b) shows lots of dislocations confined in about a $1.98 \mu\text{m}$ CGB layer and the $0.41 \mu\text{m}$ InSb layer is basically free of threading dislocations and microtwins.

Figure 2.6 Cross-sectional TEM images of GaSb on GaAs: (a) GaSb grown on AlSb buffer; (b) GaSb grown on the dislocation filter of AlSb/GaSb superlattice. [73]

A superlattice can be used as a dislocation filter to impede the gliding of threading dislocations [73, 74]. The theory developed by Frank *et al.* indicated that a strained layer grown on a lattice mismatched substrate resulted in a stress field [75]. Under the stress field, the gliding of dislocations is impeded on a (111) lattice plane causing threading dislocations bending to another one so that they annihilate one another or change their direction parallel to the interface [73]. Therefore, the strain buffer layer can impede the propagation of threading dislocations. Figure 2.6(a) illustrates the effect of the AlSb strain buffer where a high density of threading dislocations exists in the AlSb layer and only a few of them can propagate into the GaSb layer. Using a multi-strain layer (superlattice) can block rising threading dislocations, which is shown in Figure 2.6(b), where most threading dislocations propagate along the interface. Beside the AlSb/GaSb, $\text{AlGaAs}_x\text{Sb}_{1-x}/\text{AlGaAs}_y\text{Sb}_{1-y}$, and InAlSb/InSb superlattices have also been applied on a highly lattice mismatched system [73, 74, 76, 77].

2.3 Interfacial Misfit Array

Figure 2.7 Cross-sectional TEM images of GaSb grown on GaAs using IMF array: (a) low magnification, showing highly periodic IMF array at the GaSb/GaAs interface; (b) high magnification, indicating the spaces of misfit dislocations. [29]

Forming an interfacial misfit (IMF) array has become a new approach to integrate highly mismatched material on GaAs and Si [28, 31]. According to Chapter 2.1, compared with 60° misfit dislocations, 90° misfit dislocations can relieve more strain energy and make

little contribution to form threading dislocations. Therefore, a uniform IMF array consisting of pure 90° misfit dislocations is able to relieve all strain energy and avoid the formation of threading dislocations in the layer. Huang *et al.* successfully formed a periodic IMF array consisted of pure 90° misfit dislocations at the GaSb/GaAs interface [28, 29, 78]. The TDD in the 5 µm GaSb layer was reduced to $5.4 \times 10^5 \text{ cm}^{-2}$, while the TDD of 5 µm GaSb grown on GaAs using the conventional direct growth method was nearly 10^{10} cm^{-2} [78]. Figure 2.7(a) shows the cross-sectional TEM images of the GaSb/GaAs interface where uniform black dots distributed periodically at the interface are observed. The interface is flat and no threading dislocations are arising from the interface. The HR-TEM image shown in Figure 2.7(b) illustrates the details of the black dots. These dots are 90° misfit dislocations, and the separation between two adjacent dislocations is 5.6 nm. This separation (S) can be calculated using Equation (2.3) [30]:

$$S = \frac{b}{f} \quad (2.3)$$

where b is the Burgers vector of a 90° misfit dislocation, and f is the lattice mismatch. The calculated result is 5.51 nm, which agrees well with the measured one. This agreement indicates all strain energy induced by the lattice mismatch was relieved by a pure 90° interfacial misfit dislocation array.

The conditions and mechanisms of forming an IMF array is an important issue. Huang *et al.* and Bickel *et al.* revealed that a Sb-rich GaAs surface was essential to the formation of a highly periodic IMF array in GaSb grown on GaAs [29, 79]. The GaSb layer grown on an As-rich GaAs surface had a TDD four orders of magnitude higher than that grown on a Sb-rich GaAs surface. The mixture of 60° and 90° misfit dislocations was observed at the interface of GaSb grown on an As-terminated GaAs surface instead of a highly periodic IMF array. Jallipalli *et al.* established an atomistic

modeling to explain the mechanism of forming an IMF array in the GaSb/GaAs heterostructures [30]. At the initial stage of growth, one monolayer of Sb atoms are deposited on a Ga-terminated GaAs surface. Due to the difference between the covalent radius of the Sb atom (1.36 Å) and Ga atom (1.26 Å), slight distortions, such as stretching the length and bending the angles of the bonds, occur on Ga-Sb bonds. The accumulation of distortions in the bonds increases strain energy until Sb atoms spontaneously skip one Ga atom to relieve the accumulated energy. Frequently repeating this process results in a highly periodic IMF array at the interface of two highly lattice mismatched material. The simulated distribution of the IMF array agreed well with the plane-view TEM observations [30].

Rockett *et al.* established a model about the relationship between lattice mismatch, layer thickness, and MD spacing (assuming that all MDs are 90° with the same separations). Based on the model, the spacing dependent dislocation network energy under different mismatches and thicknesses are illustrated in Figure 2.8(a) and (b) [45]. In Figure 2.8(a), a minimum value exists in the curves with misfits of 5% and 10%, while it disappears in the curve with a misfit of 1%. The energy minimum is sharper with increasing misfit, indicating a uniform dislocation network is more easily formed at the more highly lattice mismatched growth. Figure 2.8(b) shows that the minimum of the dislocation network energy becomes sharper with increasing layer thickness. This result implies that the formation of a dislocation network is a dynamic process where random distributed MDs interact with each other to form a uniform MD network during the growth.

Figure 2.8 The relationship between the calculated energy density and MD spacing. (a) The effect of misfit; (b) The effect of layer thickness. [45]

An IMF array was also attempted to form between lattice-mismatched material and Si. Huang *et al.* grew GaSb on AlSb buffered 5° offcut Si substrates and TEM images showed that a uniform IMF array existed at the AlSb/Si interface. Besides the high lattice mismatch for the III-V compound semiconductor grown on Si, anti-phase boundary (APB) formation is an inherent issue due to a misalignment in group III and group V sublattices at Si surface steps. The Si substrate with a certain offcut degree can decrease the APB due to ordered two-atom-layer steps on the Si substrate's surface. High quality GaP and GaAs layers have been grown on offcut Si and Ge substrates, respectively [80, 81]. To avoid introducing extra defects for lattice mismatched heteroepitaxy, the degree of offcut should be selected carefully.

Huang *et al.* grew an AlSb layer on 3°, 5°, and 6° offcut Si substrates, respectively. Cross-sectional TEM images indicated all samples owned an IMF array at the AlSb/Si interfaces. However, only the 5° offcut Si suppressed the formation of the APB. The APB density in the AlSb layer on the 5° offcut Si was about 10^3 cm^{-2} , while the density in the other samples were above 10^7 cm^{-2} [31]. This result was attributed to the relationship between the length of two-atom-layer steps on the offcut Si substrate's surface and the spacing of dislocations at the IMF array. If the length and spacing was different, a high strain energy would accumulate at the edge of the step, which results in

the APB. Devenyi *et al.* compared the TEM images of GaAs, GaSb, InP, and AlSb layers on Si substrate with 0° and 4.7° offcut, respectively, and, suggested the suppression of APB is due to step-flow overgrowth caused by offcut substrate [82] [83].

Based on the IMF method, NIR photodetector [84], laser [85, 86], and tunnel diode [87] have been developed. An electrically pumped laser with a GaInSb/AlGaSb quantum well were grown on GaAs, which worked at $1.816\text{ }\mu\text{m}$ with a threshold current density of 1.265 kA/cm^2 at 300 K [86]. A GaSb/Al_{0.3}Ga_{0.7}Sb quantum well was integrated on a 5° offcut Si substrate [85], and a $1.54\text{ }\mu\text{m}$ laser with a threshold current density of 2 kA/cm^2 was achieved at 77 K. GaSb/InAs heterostructures [87] with a broken-gap energy band alignment and GaInAsSb photodetector [84] were also grown on GaAs using an IMF array. Recently, most devices grown on GaAs and Si using an IMF array are only based on the semiconductors with a lattice constant of about $6.1\text{ }\text{\AA}$. Therefore, there is still room to expand the IMF method on the growth of semiconductors for the MIR application with a higher lattice constant (InSb and HgCdTe).

2.4 Infrared Photodetector

An infrared detector is a sensor that converts infrared light into an electrical signal. Most photodetectors are classified into two groups: thermal detectors and photon detectors [88]. In thermal detectors, the incident infrared light is absorbed to change its temperature and the variation in temperature generates an electrical output *via* pyroelectricity or thermoelectric effect. In a photon detector, the infrared radiation is absorbed by the material and the photons excite electrons bound to lattice and impurity atoms or free electrons. This process can be simplified to electrons transition from the valence band to the conduction band and an electrical output signal can be observed [88]. A photon detector owns a good wavelength dependent response, high signal-to-noise,

and fast response. There are primary two device types: a photoconductor without a PN junction and a PN junction based photodetector. For the photodetector grown on dissimilar substrates, a high dark current associated with defects is a main disadvantage. Different materials and material systems are used in infrared photodetector. As mentioned in Chapter 1, HgCdTe and InSb are widely used in infrared photodetector. Beside them, InAs, InGaAs, InAsSb, HgZnTe, and HgMnTe are also used in infrared detector and their applied regions depend on their bandgaps [3]. Type-II superlattice (T2SL), such as InAs/GaSb, InAs/Ga_{1-x}In_xSb, and InAs/InAs_{1-x}Sb_x, are also competitive candidates for infrared detector. For T2SL, electrons mainly locate in InAs layer and holes mainly locate in GaSb layer which suppresses Auger combination and enhances carrier lifetime. Compared to narrow bandgap photodetector (such as InSb and HgCdTe), T2SL detector was predicated to have better electrical properties [89]. The bandgap of T2SL depends on the difference between the electron miniband and the first heavy hole state at the Brillouin zone center. Therefore, its detection region can cover both middle and long wavelength infrared regions [90]. Recently, the cutoff wavelength of InAs/GaSb based photodetector was up to 32 μm [91] and the integration of T2SL on GaAs and Si was also under study [92, 93].

2.4.1 The photoconductor

The photoconductor mainly comprises of a semiconductor and two metal contacts. Figure 2.9 demonstrates the basic principle of a photoconductor. The electrons are excited from the valence band to the conductive band under an incident light, and the increase in the electron-hole pairs decrease the device's resistance [94]. An applied voltage is essential to operate this device. For a photoconductor based on an intrinsic semiconductor, the photocurrent I_p can be defined as shown in Equation (2.4) [37]

$$I_p = q(\eta \frac{P_{opt}}{h\nu}) \frac{(\mu_n + \mu_p)\tau E}{L} \quad (2.4)$$

where q is the magnitude of a charged carriers (hole or electron), η is quantum efficiency (i.e. the ratio of the number of carriers collected by the photoconductor to the number of photons of incident light), P_{opt} is the power of incident light, h is the Planck constant, ν is the frequency of incident light, μ_n and μ_p are the mobility of the electron and hole in the photoconductor, τ is the carrier lifetime, E is the internal electrical field in the photoconductor (which depends on applied voltage), and L is the distance between two adjacent electrodes.

Figure 2.9 The schematic of a photoconductor with a planar architecture which are comprised of two metal contacts and a n-type semiconductor.

Clearly, the device's responsivity (I_p/P_{opt}) increases with increasing applied voltage. Therefore, a high gain (G) can be achieved in a photoconductor. The gain is expressed as shown in Equation (2.5) [37]

$$G = \tau \left(\frac{1}{t_m} + \frac{1}{t_{rp}} \right) \quad (2.5)$$

where τ is the carrier lifetime and t_m and t_{rp} are the transit time from two adjacent electrodes for electrons and holes, respectively. Increasing the material quality to increase τ and carrier mobility, decreasing the separation of electrodes, and increasing the applied voltage are the main approaches to improve gain. The sweep-out effect of

the minority carriers limits the gain [95, 96], and the highest value was reported to be 10^6 . [97]. For any kind of photodetector, the signal-to-noise ratio (SNR) is an important parameter to evaluate its performance. SNR is defined as:

$$\text{SNR} = \frac{I_p}{I_{\text{dark}}} \quad (2.6)$$

where I_p is the photocurrent and I_{dark} is the dark current. The dark current is the current flowing in the photodetector, which is not generated by the photons of incident light. In a photoconductor, the metal-semiconductor contactors are ohmic and no barrier impedes the flow of majority carriers between two contacts [98]. The carriers are excited by light flow to two contactors together with the carriers excited by other mechanisms, such as thermal activation and applied voltage. Therefore, a photocurrent is generated associated with the dark current. Although a low SNR associated with a high dark current limits the photoconductor's performance, it is still applied in the areas where the intensity of the incident light is strong enough or an ultra-high gain is required. Moreover, a photoconductor is always used to investigate the potential optoelectronic application of novel material due to its simple structure [99, 100].

2.4.2 The p-i-n photodetector and unipolar barrier

The PN junction-based photodetector normally consists of one PN junction. When the p-type and n-type semiconductor are joined together, the majority carriers (holes in the p-type region and electrons in the n-type region) diffuse into the other side of the junction. The diffusion process results in a built-in electrical field. The direction of the electrical field is opposed to that of the carriers' diffusion. Therefore, only immobile negative and positive charges exist, and all mobile chargers leave at the region between p-type and n-type regions. This region is named the depletion region or spacing charge region. Under incident light, the carriers are generated by photons at any part of the

device. In the depletion region, the photogenerated carriers are accelerated by the built-in electrical field and form a photocurrent. Increasing the width of depletion can improve the quantum efficiency and high frequency response of the photodetector [37]. The PN junction detector can be worked under 0 voltage (photovoltaic mode) and reverse bias voltage (its direction is the same as the built-in electrical field). To further improve the photodetector's performance, an intrinsic semiconductor layer is inserted between the n- and p-type regions, which is named the p-i-n photodetector.

As introduced in Chapter 2.4.1, suppressing the dark current is an important approach to developing a photodetector with high performance. Commercial MIR photodetectors require cooling to decrease noise because of the high dark current at room temperature [13, 101]. Additional cooling equipment increases the cost in both the operation and manufacture of the MIR photodetector system. Therefore, investigating the methods to decrease a dark current in MIR photodetectors can simplify fabrication, decrease cost, and extend application fields [11, 34, 102]. The PN junction in the p-i-n photodetector impedes the majority carriers flow to two contacts, therefore, its dark current is much less. However, other sources can still result in a dark current. The origins of the dark current density (J_{dark}) in a p-i-n photodetector can be classified into five components [33, 103, 104] as shown in Equation (2.7):

$$J_{dark} = J_D + J_{SRH} + J_B + J_T + J_{surface} \quad (2.7)$$

where J_D is the diffusion current density, J_{SRH} is the current density caused by the Shockley-Read-Hall (SRH) generation-recombination (G-R) processes, $J_{surface}$ is the surface leakage current density, J_B is the current density caused by band-to-band tunneling, and J_T is the trap-assisted tunneling current density.

The diffusion current density (J_D) is the most fundamental source for the dark current

density (J_{dark}). Most of the J_D is generated at the neutral regions (p-type and n-type) of the photodetector and mainly includes three origins. The first origin is thermally generated carriers, which can be described as shown in Equation (2.8) [104].

$$J_D = n_i^2 \sqrt{q k_B T} \left(\frac{1}{N_D} \sqrt{\frac{\mu_h}{\tau_h}} + \frac{1}{N_A} \sqrt{\frac{\mu_e}{\tau_e}} \right) (e^{\frac{qV}{k_B T}} - 1) \quad (2.8)$$

where n_i is the intrinsic carrier concentration of the material used in the diode, q is the electron charge, V is applied reverse bias, k_B is the Boltzmann constant, N_A and N_D are the doping concentration of the p- and n-type neutral regions, respectively, and μ_h , τ_h , μ_e , and τ_e are the carrier mobility and lifetime of the electrons and holes. The J_D caused by thermally generated carriers is a main source of the J_{dark} in the narrow bandgap semiconductor-based photodetector, and the cooling device can suppress it because it strongly depends on temperature. Radiative generation is the second origin where photons generated by the thermal process are reabsorbed by the semiconductor itself to generate carriers. Its effect is not remarkable unless the device operates at a very low temperature and short wavelength [105]. The third origin is Auger generation, which depends on a semiconductor's intrinsic properties. Among many types of Auger processes, the effect of "Auger 1" is the most remarkable in a narrow band gap semiconductor-based photodetector, which includes two transitions: 1) one electron is excited from the heavy-hole band to the conduction band and leaves one heavy hole; and 2) one electron transits from a higher energy state to a lower energy state in the conduction band [106]. Improving the material's quality and optimizing the architecture of the device are unable to avoid the Auger generation: therefore, it is always considered as a fundamental limit for the performance of a photodetector. If the dark current in a photodetector mainly originates from Auger generation, the material and architecture of the photodetector is sufficiently optimized [107].

The J_{SRH} , J_B , and J_T mainly occur at the depletion region, which are related to the PN junction. The J_{SRH} originates from thermally generated carriers *via* SRH defects. The SRH defects locate within the bandgap to provide transitional energy states for thermal generation. Although the J_{SRH} can generate at any part of the photodetector, its effect on the device's performance is more remarkable in the depletion region rather than in the neutral region [108]. The J_{SRH} also strongly depends on temperature, and improving the semiconductor's quality can also suppress it [109].

Figure 2.10 The band diagrams of a PN junction showing (a) trap-assisted tunneling; (b) band to band tunneling.

The J_T and J_B are both related to the tunneling effect between the conduction and valence bands and the difference between them is whether the defect involves the process. Figure 2.10(a) illustrates the band diagram about trap-assisted tunneling, which involves two types. For the first type, an electron tunnels from the valence band in the p-type side to the conduction band in the n-type side *via* a defect trap state located within the band gap. This process occurs at the PN junction; therefore, the energy of the electron does not change. The other type is a mixture of SRH generation-recombination processes and trap-assisted tunneling where an electron is excited to the defect trap state and then tunnels to the conduction band [110]. Improving the quality of the semiconductor results in a lower density of the defect trap; thereby, reducing the J_T . The J_B is not a defect-induced dark current density. However, reverse bias is essential for the J_B . The electrons

are accelerated under reverse bias to tunnel through the bandgap at the PN junction, as shown in Figure 2.10(b). Operating the device at low reverse bias or 0 V can suppress the J_B [105].

A high density of dangling chemical bonds on the surface of semiconductors replace the infinite and periodical lattice in bulk. This feature creates a bending of bands at the surface associated with an accumulation of charges. For a narrow bandgap semiconductor, the bending of bands results in conduction bands located below the Fermi level. Therefore, a shunt path is formed at the surface where electrons flow freely [111]. The J_{surface} is increased with increasing reverse bias and temperature, and its effect is remarkable in the photodetector with a high perimeter-to-area ratio [109]. Conventionally, surface passivation can suppress the J_{dark} to some extent [112].

Based on the previous discussion, the J_{dark} depends on semiconductor quality, temperature, and reverse bias during device operation. Recently, applying an electron barrier layer on traditional p-i-n photodetector architecture has been proposed to suppress the dark current. The electron barrier layer is constituted by one or more kinds of semiconductors with a larger bandgap than the semiconductor used in the photodetector. The electron barrier layer should have a large band offset in the conduction band to block electrons, while it should have a nearly zero band offset in the valence band so that the holes can flow unimpededly [113]. Therefore, the electron barrier layer was named the “unipolar barrier.” Maimon, *et al.* first inserted an AlAsSb electron barrier layer in the conventional n-type InAs photoconductor with a vertical architecture to form nBn type photodetector [114]. This nBn detector was able to operate under conditions of background limited performance which was better than the InAs photoconductor. The electron/hole barrier or similar concept were applied on many other

kinds of detectors. For example, Klipstein, *et al.* proposed XBn barrier photodetector where X was contact layer, B was barrier layer and n was absorption layer [115]. The material bandgaps and doping levels for these three layers could be tailored to achieve a high performance. An InAsSb-based XB_nn was grown on GaAs substrate to achieve a low 150 K dark current density of 10^{-7} A/cm² [116]. Ting, *et al.* proposed the complementary barrier infrared detector (CBIRD) structure which included both electron and hole barrier layer and a CBIRD with InAs/GaSb superlattice absorption layer achieved a 87 K black body detectivity of 1.1×10^{11} cmHz^{1/2}W⁻¹ under 0.2 V bias [117]. The barrier layer was also applied on the p-i-n photodetector such as InSb and InAs detector [118, 119].

Figure 2.11 The band diagrams of the nBn photodetector based on n-type semiconductor. (a) Electron barrier layer locates near the anode. (b) Electron barrier layer locates near the cathode.

Properly locating the barrier in the photoconductor is important, and there are two choices: near anode or near cathode. As shown in Figure 2.11, the band diagrams of the nBn type photodetector illustrates the effect of different barrier positions. For both positions, majority carriers are blocked by the barrier and cannot flow from the anode (V-) to the cathode (V+). When the barrier is located near the anode (Figure 2.11(a)), most of the photon generated carriers (electrons) are not blocked, whereas most of the photon generated carriers do not flow into the cathode when the barrier is located near the cathode (Figure 2.11(b)) [120]. Obviously, locating the barrier near the cathode severely suppresses the performance of the photoconductor.

Figure 2.12 The band diagrams of PN junction-based photodetector (a) without an electron barrier layer; (b) with an electron barrier layer at p-type side; (c) with an electron barrier layer at n-type side.

Figure 2.12 shows the band diagrams about the PN junction-based photodetector with and without an electron barrier layer. The incident light is absorbed in the p-type region [33]. In Figure 2.12(a), no barrier is located in the device. Thus, the photocurrent and all kinds of J_{dark} can flow freely in the device. Photo-generated electrons flow in the n-type region to the n-type side contact, while photo-generated holes flow across the PN junction to the p-type side contact.

After placing a barrier at the p-type side as shown in Figure 2.12(b), the path for the J_{surface} is blocked because the J_{surface} is constituted by electrons flowing at the surface. The J_D , which is thermally generated at the p-type side, is also blocked. However, a large part of the J_{dark} is unable to be blocked because the J_T , J_B , and J_{SRH} are generated in the depletion region. Therefore, placing a barrier at the n-type side can further suppress the dark current as shown in Figure 2.12(c). The J_{surface} , J_T , J_B , and J_{SRH} can be blocked by this barrier. Moreover, the J_D which is thermally generated below the barrier, is also

blocked. Some sources of the J_{dark} remain unimpeded, such as the J_{D} generated by Auger generation and the J_{D} thermally generated above the barrier by radiative generation.

The design of the photodetector where the electron barrier layer locates at the n-type side was realized in an InAs-based p-i-n photodetector on a native substrate with a 100 nm barrier layer of InAs_{0.18}Sb_{0.82}. Any surface passivation was not applied on the devices [33]. The barrier layer was located at the p-type side and the n-type side to form a p-type side barrier and an n side barrier photodetector, respectively. Both photodetectors blocked the J_{surface} , and only the n-type side barrier one blocked the other junction related to the J_{dark} , which agreed well with the design. For the n-type side barrier photodetector, its R_0A increased by six orders compared with a conventional p-i-n photodetector where R_0 is the resistance at zero bias and A is the area of the photodetector. An InSb-based photodetector with a unipolar barrier on GaAs was reported where the barrier layer was located at the p side [119]. Due to the lack of a lattice-matched wide bandgap semiconductor with InSb, the material and thickness of the barrier layer in an InSb-based unipolar photodetector should be carefully determined to form a strain layer without misfit dislocations. An improper architecture or growth process will introduce extra dislocations, which increases the J_{dark} and suppresses the device's performance.

Chapter 3 Growth of InSb on GaAs Using Interfacial Misfit Array

3.1 Introduction

Due to its high electron mobility ($77,000 \text{ cm}^2/\text{V}\cdot\text{s}$ at 300 K) and narrow band gap (0.17 eV at 300 K), InSb is an ideal candidate for the channel material in high electron mobility transistors (HEMT) [121] and for photon absorbers in mid-infrared photonic applications [3]. Unlike GaAs, a semi-insulating InSb substrate is not commercially available. Integrating InSb on GaAs will not only provide a semi-insulating substrate, but also enable the co-existence of InSb-based and GaAs-based devices on a single platform. Moreover, it is an essential step for the growth of InSb on Si using a GaAs/Ge buffer to realize the monolithic integration of InSb on Si as introduced in Chapter 1.

It is extremely challenging to integrate InSb on a GaAs platform due to the considerable lattice mismatch between InSb and GaAs (14.6%). The conventional direct growth method of InSb on GaAs resulted in a high threading dislocation density (TDD) and other defects [122]. Lattice strain could be accommodated by a thick ($>1\mu\text{m}$) buffer layer grown on a GaAs substrate [74]. However, a thick buffer layer may not be favorable for integration with other GaAs-based devices, as the resulting non-planar device arrangements require lengthy interconnect metal contacts.

As introduced in Chapter 2, utilizing a uniform interfacial misfit (IMF) array consisting of pure 90° misfit dislocations is a new approach for lattice mismatched growth where all lattice mismatch is accommodated at the interface and a thick buffer layer is no longer required. A strain induced by a lattice mismatch $>7\%$ favors the formation of 90° misfit

dislocations compared to 60° misfit dislocations [45]. The lattice mismatch between InSb and GaAs is 14.6%. Till now, utilizing an IMF array to grow a high quality InSb layer on GaAs has not been realized. Moreover, researches on the growth conditions and mechanisms of InSb on GaAs using the IMF method are required to have an in-depth understanding of this method.

In this chapter, four samples with an InSb layer grown on a GaAs substrate were grown using a molecular beam epitaxy (MBE) system at different growth temperatures (350°C , 310°C , 270°C , and 220°C) to investigate the IMF formation at the InSb/GaAs interface. In contrast to previous studies [28, 78] where a monolayer of Sb atoms was formed by depositing Sb atoms on a Ga-rich GaAs surface, this study utilized an As-for-Sb anion exchange process to form a monolayer of Sb atoms on the GaAs surface. The advantage of this process is that it occurs at a lower temperature than the former process, which relies on the formation of a Ga-rich GaAs surface. The formation of a Ga-rich GaAs surface requires a substrate temperature of 560°C or higher, which may lead to a rough GaAs surface [123]. The structural and electrical properties of the InSb layer were investigated using high-resolution x-ray diffraction (HR-XRD), TEM, AFM, and a Hall effect measurement system. A well-suited growth temperature for an InSb layer with the lowest TDD is worked out. Furthermore, the relationship between IMF formation and growth temperatures is analyzed. The formation of 90° and 60° misfit dislocations at different temperatures is also discussed.

3.2 Experimental Details

700 nm InSb layers were grown on (100) GaAs substrates with various growth temperatures. All experiments were carried out in solid source MBE equipped with arsenic (As) and antimony (Sb) valved cracker sources, which can provide As₂ and Sb₂. The substrate temperature was measured by a tungsten/rhenium alloy thermocouple, which was placed at the back of the molybdenum substrate holder. In situ reflection high-energy electron diffraction (RHEED) was used to monitor the surface reconstruction during growth. The oxide desorption temperature of the GaAs substrate (580 °C) was used to calibrate the thermocouple reading. The detailed growth procedures were listed below:

1. GaAs buffer growth. The onset of oxide desorption was detected by in-situ reflection high-energy electron diffraction (RHEED) observation. After the oxide desorption, 200 nm GaAs was grown on the substrate in 12 minutes at a growth rate of 17 nm/min to smoothen the surface. The RHEED pattern exhibited a clear (2×4) As-terminated surface reconstruction after the buffer growth.
2. Reduction of the As background pressure. After the growth of GaAs buffer, the As valve was closed. The substrate temperature was lowered to 540 °C and maintained at this temperature for 10 minutes to reduce the As background pressure. The RHEED pattern remained the same throughout this period.
3. Absorption of the Sb atoms on GaAs surface. The substrate temperature was lowered to the respective growth temperature for each of the four samples (350 °C, 310 °C, 270 °C, and 220 °C) and the conditioned substrate was exposed to Sb flux with a beam equivalent pressure (BEP) of about 1.2×10^{-6} Torr. The As atoms at the surface

were replaced gradually by Sb atoms *via* the anion exchange process during this Sb supply step. After 4 minutes of Sb supply, a streaky (1×3) Sb-rich surface reconstruction was observed with RHEED.

4. InSb layer growth. Following the occurrence of this (1×3) reconstruction, 700 nm InSb layers were grown at set temperatures with a Sb/In beam equivalent pressure ratio of 4 and a growth rate of 900 nm/hrs. During the initial stage of growth, the streaky (1×3) pattern rapidly changed to a bright spotty pattern. This observation indicated that an island-like growth mode occurred at this stage. Subsequently, the RHEED patterns slowly recovered to a streaky (1×3) pattern. The recovery time increased with the decreasing substrate temperature. The longest time is about 2 minutes, which was observed at 220 °C. In the following, the RHEED patterns remained streaky (1×3), which indicated all samples kept a layer-by-layer growth mode until the end of their growth.

The crystal qualities and relaxations of the InSb layers were evaluated by HR-XRD with a monochromator to obtain high resolution, and the surface morphologies were characterized using AFM. Cross-sectional TEM was used to observe the microstructures of the layers. Indium was deposited on the InSb surface as metal contact, and the carrier mobility and concentration values were evaluated by Hall/Van der Pauw measurements.

3.3 XRD and AFM Measurement

Table 3.1 Overview of XRD, AFM and Hall resistance measurement results for InSb/GaAs heteroepitaxy at different growth temperatures.

Sample No.	Growth Temperature (°C)	Strain Relaxation (%)	(004) ω scan FWHM (arcsec)	r.m.s. Roughness (nm)	300 K carrier concentration (10^{16} cm^{-3})	77 K carrier concentration (10^{16} cm^{-3})
1	350	98.7	454	0.9	1.852	1.056
2	310	98.9	415	1.1	1.647	0.849
3	270	99.1	462	1.3	2.278	1.435
4	220	99.9	650	8.8	3.374	2.074

Figure 3.1 XRD scans of InSb grown on a (100) GaAs substrate from (004) reflection. The growth temperatures of InSb are indicated. (a) ω - 2θ scans; (b) ω scans.

Table 3.1 lists the degree of strain relaxation of the InSb layers, the full width at half maximum (FWHM) of the x-ray diffraction (XRD) rocking curves, the root mean square (r.m.s.) roughness in the AFM measurements and carrier concentrations for each sample. X-ray diffraction symmetric (004) ω - 2θ scans of InSb layers on GaAs with different growth temperatures are shown in Figure 3.1(a), and the position of GaAs peaks were normalized to 0°. The simulated results of 100% relaxed InSb on GaAs are shown at the bottom. All InSb peaks from InSb on GaAs samples own the same position as the

simulated one. Therefore, all samples are nearly 100% relaxed. No additional peaks were observed between InSb and GaAs peaks, which indicated no buffer layer formed between InSb and GaAs.

To measure the precise relaxation in the InSb layer, an asymmetric (115) ω - 2θ scan was applied on samples (not presented). Vertical and parallel lattice constants ($a_{\perp}$ and $a_{\parallel}$) were extracted from the symmetric and asymmetric XRD scans (Table 3.1). The degree of strain relaxation (R) can be calculated using Equation (3.1),

$$R = \frac{a_{\parallel} - a_s}{a_0 - a_s} \times 100\% \quad (3.1)$$

where $a_{\parallel}$ is the measured parallel lattice constant of InSb layer, a_0 is the lattice constant of bulk InSb, and a_s is the lattice constant of the substrate material. Using this equation, relaxations in the InSb layers were calculated and are also listed in Table 3.1. The degree of strain relaxations increased with decreasing growth temperature and are all above 98.5%.

Figure 3.1(b) shows symmetric (004) ω scans of InSb grown on GaAs at different growth temperatures. As the presence of threading dislocations in a layer could result in significant broadening of the symmetric ω scan [54, 124], the FWHMs obtained from the (004) ω scans can be used to qualitatively compare the TDD in the samples [125, 126]. From the FWHM comparison in Table 3.1, it is evident that the changes in growth temperature can influence the TDD of the samples. The (004) rocking curve FWHM analysis suggests that TDD is the lowest in the InSb sample grown at 310 °C.

Figure 3.2 shows the AFM images for the four samples with InSb layers grown at (a) 350 °C, (b) 310 °C, (c) 270 °C, and (d) 220 °C, respectively. The r.m.s. roughness for each sample is listed in Table 3.1. Samples 1-3 show a smooth surface with a r.m.s.

roughness below 1.3 nm. However, the sample with an InSb layer grown at 210 °C (Sample 4 in Figure 3.2(d)) exhibits a high r.m.s. roughness of 8.8 nm and a high density of steps was observed on the surface of the layer. This comparatively high r.m.s. roughness can be attributed to limited adatom diffusivity at a very low temperature [127]. Moreover, craters are observed on the surface of Sample 4. These craters exhibited random size distribution and shape. This was in contrast to the previously reported pinhole defects shown in a highly mismatched layer where the pinhole defects had been caused by threading dislocations and low atomic migration rate [128]. These pinhole defects have a uniform size and are hexagonal in shape. The origin of the craters shown in Sample 4 will be discussed later in conjunction with microstructural information in Chapter 3.4.

Figure 3.2 20 μm $\times$ 20 μm AFM images of InSb grown on GaAs (100) substrate at (a) 350 °C; (b) 310 °C; (c) 270 °C; and (d) 210 °C.

3.4 TEM Observation

Figure 3.3 Cross-sectional TEM images of InSb grown on (100) GaAs substrate at 310 °C. (a) A low magnification TEM image. The rich clusters are marked by a white circle. Threading dislocations are indicated by the white arrows; (b) TEM images of uniform interfacial misfit array; (c) TEM images showing the separation between two adjacent misfit dislocations.

Transmission electron microscopy (TEM) was used to characterize the microstructural properties of four InSb layers on GaAs. Bright field TEM images of the sample grown at 310 °C are shown in Figure 3.3(a)-(c). Figure 3.3(a) shows a clean InSb/GaAs interface with only a few dislocations rising from the interface. I observed only two

threading dislocations (marked by white arrows) that reached the surface of the 720 nm-thick InSb layer over a length of 1.7 μm across the surface. Hence, the TDD of the InSb layer is estimated to be $1.38 \times 10^8 \text{ cm}^{-2}$.

In addition to the TDs, black spots were observed in the InSb layer as shown in Figure 3.3(a). Energy dispersive x-ray (EDX) analysis of these spots revealed that they are In-rich clusters, where the In to Sb atomic ratio is 13% higher compared to the ratio at the area outside the cluster. These In-rich clusters could be due to excessive indium flux. Alternatively, an insufficient Sb/In ratio was used in the growth of the InSb layer. Figure 3.3(b) shows a high resolution TEM image taken at the InSb/GaAs interface where periodic IMF dislocations can be observed. A higher magnification TEM image for these periodic dislocations is shown in Figure 3.3(c). The dislocations were distributed uniformly across the InSb/GaAs interface, and the distance between two adjacent dislocations is measured as about 3.2 nm.

Figure 3.4 High resolution TEM images of the InSb/GaAs interface in the InSb grown on GaAs at 310 °C. The black dashed arrows are a Burgers circuit around a 90° misfit dislocation with a Burger vector $b=a/2(1\bar{1}0)$. The box with a dotted line show the atomic sites between adjacent dislocations in InSb and GaAs, respectively.

The lattice constant of InSb and GaAs are 6.479 Å and 5.653 Å, respectively. Thus, it is energetically favorable for every seventh Sb atom in the InSb layer to skip one Ga atom

in the GaAs layer to decrease the lattice strain energy of the Ga-Sb bonds at the InSb/GaAs interface [30]. This phenomenon was observed in the TEM image as shown in the white boxes in Figure 3.4. The separation (S) between two adjacent 90° dislocations can be calculated using Equation (3.2) [30]:

$$S = \frac{b}{f} \quad (3.2)$$

where b is the Burgers vector of misfit dislocation, and f is the lattice mismatch. The details of the 90° misfit dislocation with the Burgers vector of $\frac{a}{2}[11\bar{0}]$ (marked by the Burgers circuit in black dashed arrows) is shown in Figure 3.4. Substituting a $|b|$ value of $4.58 \text{ \AA}$ and a lattice mismatch of 14.6% into Equation (3.2) gives the separation distance between two adjacent dislocations of 3.19 nm. This calculated value is consistent with the 3.2 nm distance of separation observed in Figure 3.3(c).

Unlike Sample 2 (grown at 310°C), Figure 3.5(a) shows threading dislocations arising from InSb/GaAs interface at the InSb/GaAs interface in Sample 1 (grown at 350°C). To investigate the origin of the threading dislocations, a high-resolution TEM image (shown in Figure 3.5(b)) was taken at the InSb/GaAs interface where the threading dislocations arise (marked by a white dotted box in Figure 3.5(a)). Clear and distinct contrasts in the InSb layer and GaAs layer indicate an abrupt InSb/GaAs interface. A discontinuous V-shaped region (two bifurcate lines arising from interface and marked by two opposed white arrows) are found at the left of Figure 3.5(b). A previous study suggested that these dots indicate 90° misfit dislocation nucleation areas [29].

Figure 3.5 Cross-sectional TEM images of InSb grown on (100) GaAs substrate at 350 °C. (a) Bright field image of InSb/GaAs interface. A box indicates the area shown in (b); (b) high resolution image along $[1\bar{1}0]$ zone axis; (c) and (d) phase contrast images of selected areas C and D in (b) using Fourier mask filtering technique. At the interface, 60° misfit dislocations are indicated by one arrow and 90° misfit dislocations are indicated by two arrows.

Since the details of the interfacial atomic arrangement in the image are too noisy to confirm the type of dislocations, a Fast Fourier mask filter was applied on the two selected areas (marked C and D) in Figure 3.5(b) to reduce the noise and investigate the atomic arrangement around the InSb/GaAs interface. The results are shown in Figure

3.5(c) and (d). Figure 3.5(c) shows the interface at the V-shaped regions where two 90° and two 60° MDs are found. A 90° MD is associated with two extra lattice planes symmetrically located at the interface while a 60° MD is only associated with one extra lattice plane. The distance between the two adjacent 60° MDs is 5.3 nm. Since 60° misfit dislocations only relax half the strain compared to the 90° misfit dislocations the separation distance between two adjacent 60° dislocations should be half of the distance between two 90° dislocations [44] if the strain is to be relaxed fully. Therefore, 60° dislocations with a separation distance of 5.3 nm on an InSb/GaAs interface are insufficient to fully relax the strain energy. Since the degree of relaxation in Sample 1 is 98.7%, the residual strain must be relaxed through other mechanisms such as the formation of threading dislocations [129]. Figure 3.5(d) shows a uniform 90° misfit dislocation array with a separation of 3.2 nm in Area D. There are no other visible defects around that area and the strain is fully relaxed.

Figure 3.6(a) shows a bright-field TEM image of Sample 3 (grown at 270°C) at the InSb/GaAs interface. Arc-shaped features with different image contrasts are observed at the interface and threading dislocations are found in the InSb layer. Figure 3.6(b) shows a high-resolution TEM image taken at the edge of an arc-shaped feature (indicated by the dotted box). In that area, one dislocation line, which is nucleated at the InSb/GaAs interface, is found.

Figure 3.6 Cross-sectional TEM images of InSb grown on (100) GaAs substrate at 270 °C. (a) Bright field image of InSb/GaAs interface. A box indicates the area shown in (b); (b) high resolution image along $[1\bar{1}0]$ zone axis; (c) and (d) phase contrast images of selected areas C and D in (b) using Fourier mask filtering technique. At the interface, 60° misfit dislocations are indicated by one arrow and 90° misfit dislocations are indicated by two arrows.

By applying a Fast Fourier filter to the TEM image, the arrangements of the atoms in Areas C and D are presented in Figure 3.6(c) and (d). In Figure 3.6(c), three 90° MDs and one 60° MD are observed. The separations for these dislocations are not uniform. From left to right, the space between two adjacent dislocations is 3.8 nm, 4.7 nm, and 2 nm, respectively. With a 60° MD and wider distances between two adjacent dislocations, this IMF array was not capable of fully relaxing the entire lattice strain. Thus, the formation of threading dislocations to relax the remaining strain was promoted. In contrast to the dislocations in Figure 3(c), uniformly distributed 90° MDs with a

separation of 3.2 nm, which can relieve the entire lattice strain, are observed in Area D as shown in Figure 3(d). No threading dislocations are observed in that area.

Figure 3.7 (a) A low magnification cross-sectional TEM image of InSb grown on (100) GaAs substrate at 210 °C and one microtwin in InSb layer is indicates; (b) selected area electron diffraction (SAED) pattern taken from the InSb/GaAs interface. Spots belong to InSb and GaAs have been marked by white arrows. White dotted circles mark extra diffraction spots associated by mirotwins in InSb layer.

Figure 3.7(a) contains a bright-field cross-sectional TEM image taken from Sample 4. In contrast to Figures 3.4(a), 3.5(a), and 3.6(a) obtained from Samples 1, 2, and 3, respectively, various defects were observed in this TEM image obtained from Sample 4. In addition to the threading dislocations, a microtwin defect is observed in the heteroepitaxial InSb layer, which is marked by a white arrow in Figure 3.7(a). In Figure 3.7(b), a selected area electron diffraction (SAED) taken at the InSb/GaAs interface of Sample 4 illustrates two sets of separate diffraction spots, which are associated with two different lattice constants from InSb and GaAs, respectively. The SAED also shows additional microtwin spots (circled by a white dotted line). These spots confirm the presence of microtwin defects, which have different crystal orientations than the remaining InSb layer and could be the cause of the formation of craters at the surface of the layer (these are observed in the TEM image in Figure 4(e) and the AFM image in

Figure 3.2(d)). In a strongly mismatched heteroepitaxy, the formation of microtwins defects is normally linked to the introduction of a stacking fault during the island growth phase [130].

3.5 Formation of IMF Array at the InSb/GaAs Interface

The results from XRD, AFM, and TEM reveal the influence of the growth temperature on IMF array formation in the heteroepitaxy of InSb on GaAs. A growth temperature of 310 °C (Sample 2) resulted in the lowest observed TDD in the layer, as suggested by the trend from the TEM observation in conjunction with the FWHM of the XRD (004) ω scan. In Samples 1 and 3, 60° dislocations and a shorter distance between two adjacent dislocations disrupt the uniformly distributed 90° IMF array. The existence of 60° misfit dislocations is commonly associated with the formation of threading dislocations in an antimonide compound semiconductor on GaAs [51].

A previous TEM study revealed that 90° dislocations start to nucleate at the interface during the phase of island growth [49]. Each island has its own 90° dislocation network. When two islands coalesce, their dislocation networks can merge to form a larger network. Usually, a small shift between two networks can be accommodated by an elastic deformation. However, in the case of coalescence between two islands, where the shift between two networks is roughly half of the dislocation separation (S), a 60° dislocation will be formed [50]. Although a quantitative relationship between the density of islands and TDD is difficult to establish, a higher InSb island density can be qualitatively associated with a higher rate of island coalescence, and thus leads to a higher TDD. A study on the heteroepitaxy of GaSb on GaAs revealed that GaSb grown at a low temperature exhibited higher island nucleation site density compared to GaSb grown at a high temperature [131]. Thus, the higher TDD in Sample 3 could be due to

the higher island coalescence, as Sample 3 was grown at a temperature lower than the growth temperature of Sample 2.

The island coalescence mechanism does not explain the high TDD found in Sample 1, which was grown at a higher temperature than Sample 2. Huang *et al.* showed that a Ga-rich GaAs surface before the Sb supply is essential to form a uniform 90° interfacial misfit dislocation array. In their report, using the contrast test, one 5 μm GaSb layer was grown on a Ga-rich GaAs surface and another 5 μm GaSb layer was grown on an As-terminated GaAs surface. A plan-view TEM revealed TDDs of about $5 \times 10^5 \text{ cm}^{-2}$ for the sample grown on Ga-rich GaAs and of about 10^9 cm^{-2} for the sample grown on As-terminated GaAs [29]. A theoretical calculation [30] predicted that Ga-Sb bonds, which resulted from the deposition of a monolayer of Sb atoms on the Ga-rich GaAs surface, are stretched and bent due to the different covalent radius between the Sb and As atoms. The stretching and bending of these Ga-Sb bonds result in lattice strain. The minimization process of this strain drives the formation of a periodic IMF array [30]. The existence of Ga-As bonds changes the distribution of strain and obstructs the formation of 90° MDs.

In this study, the formation of Ga-Sb bonds was done through the As-for-Sb anion exchange process at the GaAs surface instead of through the deposition of Sb on Ga-rich surface. This anion exchange process can be expressed as:

The change of enthalpy for this reaction is 47.6 kJ/mol at 350 °C, which yields an equilibrium constant of 0.138 [132]. Under a Sb flux with a BEP of 1.2×10^{-6} Torr, less than 5% of the As on the GaAs surface can be replaced by Sb atoms. This indicates an inefficient As-for-Sb anion exchange process, which may lead to residual Ga-As bonds

at the surface. However, the situation differs for low substrate temperatures. Arsenic released from the substrate forms arsenic tetrameric (As_4) on the surface of the substrate when the substrate temperature is below 350 °C [133]. The As-for-Sb exchange reaction with As_4 has an equilibrium constant of up to 148 [133]. Under the same Sb flux (BEP of 1.2×10^{-6} Torr), the As-for-Sb conversion is nearly 100% on the GaAs surface, which indicates that As atoms can be replaced efficiently by Sb atoms. Complete As-for-Sb anion exchange on the As-terminated GaAs surface has the same effect as depositing As on a Ga-rich GaAs surface as mentioned in the introduction. Strain caused by Ga-Sb bonds stretching and bending plays an important role in the formation of a uniform IMF array, while Ga-As bonds are unable to result in lattice strain because the substrate is GaAs, which impedes the formation of the IMF array. The higher substrate temperature during the Sb absorption step in Sample 1 (grown at 350 °C) resulted in an inefficient As-for-Sb anion exchange process, which did not occur on the other three samples grown at 310 °C or lower. The higher growth temperature in Sample 1 has led to a higher density of Ga-As bonds on the surface, and it has suppressed the formation of a periodic IMF array in the InSb layer. Therefore, depositing Sb atoms on the GaAs surface at lower temperature and then growing a InSb layer at the optimal temperature is expected to further increase the layer's quality.

3.6 Hall Measurement

In this study, most of the lattice mismatch between InSb and GaAs is accommodated through the IMF array. However, threading dislocations still exist in the InSb layer. For InSb on GaAs, a strong relationship between threading dislocations and carrier mobility was reported and the highest carrier mobility was achieved in the sample with a lowest threading dislocation density [27]. Sato *et al.* prepared InSb layers with different

thicknesses on GaAs substrates and found that carrier mobility in the InSb layer was limited by threading dislocations [134]. This limitation is prominent for a layer thickness below 1 μm [135]. Therefore, it is essential to measure the carrier mobility and concentration of the InSb layer to further investigate the effect of different growth temperatures.

Figure 3.8 Dependence of (a) carrier mobility and (b) carrier concentration on growth temperatures for InSb grown on GaAs (100) substrate at 300 K and 77 K.

Figure 3.8(a) and (b) illustrates growth temperature-dependent carrier mobility and concentration for an unintentional n-type doped InSb grown on GaAs measured at 300 K and 77 K. The detailed 300 K mobility and concentration are listed in Table 3.1. The

influence of the GaAs substrate can be negligible due to its resistivity of $10^7 \Omega\text{cm}$. At 300 K, the highest carrier mobility value ($33,840 \text{ cm}^2/\text{Vs}$) was observed in Sample 2. This is consistent with XRD and TEM results. However, all samples exhibited decreasing mobility as the temperature was lowered. Furthermore, this decrease of carrier mobility with a decrease of temperature from 300 K to 77 K is unusual. Typically, higher carrier mobility at low temperature measurement is expected due to the suppression of phonon scattering at low temperature, and this behavior is consistently reported in bulk InSb [136]. A reduced carrier mobility was reported [137] in a n-type bulk InSb, which was measured at a very low temperature (20 K-80 K). The carrier mobility increased with an increasing measurement temperature and reached the maximum mobility value around 60 K. This phenomenon was attributed to the scattering of ionized impurity [138]. Below 80 K, the effect of phonon scattering on carrier mobility was very weak and can be ignored. In this low temperature condition, the carrier mobility was significant limited by the Coulomb interaction between the ionized impurities and carriers. At a higher measurement temperature, the increasing thermal velocities of carriers weaken the influence of ionized impurities, and the value of carrier mobility increased in proportion to $T^{3/2}$, where T was the temperature based on the Brooks-Herring formula [139]. However, it should be noted that the influence of ionized impurities on the carrier mobility in the n-type bulk InSb material was only significant in the $< 80\text{K}$ measurements. Carrier mobility in the n-type InSb bulk decreased with an increasing measurement temperature from 60 K to 600 K due to the effect of phonon scattering [136]. Therefore, the effect of ionized impurities could not explain the reduced carrier mobility observed in the InSb layers on GaAs measured at 77 K.

Another possible explanation is related to the effect of dislocations in InSb layers. A lower carrier mobility at 77 K has been reported in some InSb/GaAs heteroepitaxy

grown below 2 μm thick [140, 141]. Although the mechanism of this deviation is still not fully understood, the reports attributed it to the influence of dislocations in InSb layers. For all samples, the carrier concentrations at 300 K were higher than $2 \times 10^{16} \text{ cm}^{-3}$, which is the intrinsic carrier concentration of InSb at 300 K [142]. Moreover, all samples showed carrier concentrations between 1×10^{16} to $3 \times 10^{16} \text{ cm}^{-3}$ at 77 K, which is at least two orders of magnitude higher than the intrinsic carrier concentration of InSb ($\sim 10^{14} \text{ cm}^{-3}$) [142]. The carrier concentration also depends on the TDD in the InSb layers. Moreover, a higher dislocation density, as suggested by the FWHM of the XRD (004) ω scans, seems to result in a higher carrier concentration. This phenomenon suggests the presence of extrinsic carriers in the unintentional doped InSb layers at both 300 K and 77 K, and these extrinsic carriers may be related to defects (threading dislocations and microtwins) in the InSb layers. The background unintentional doping concentration of the MBE system used in this study was in the range of 10^{14} cm^{-3} and the extrinsic carriers observed in the samples were above 10^{16} cm^{-3} . Therefore, background unintentional doping makes little contribution to extrinsic carriers in the InSb layer.

Because IMF arrays and threading dislocations are in the region close to the InSb/GaAs interface, this region is named the “dislocation region” in this study. Although the accurate thickness and clear boundary of this region is difficult to be extracted from the available experimental data, TEM images in Figures 3.5–3.6 clearly show that high density dislocations exist near the interface. Previous research demonstrated that dislocation density decreased rapidly with increasing thickness in heteroepitaxy [40]. Beside dislocations, microtwins are also found in Sample 4 and the influence of microtwins on the carrier mobility is similar to the influence of dislocations [143]. In the dislocation region, most of carriers are extrinsic and excited from dislocations [140]. These carriers are inevitably scattered by dislocations, leading to low carrier mobility.

The region above the dislocation region where the dislocations density is low is named as the “bulk region” in this study. The carriers (electrons) in this region are intrinsic and directly excited from the valence band of the InSb. Due to the minimum of the dislocation-induced scattering, the carrier mobility in this region is relatively higher compared to the mobility in the dislocation region.

At 300 K, the carrier concentration in each sample is slightly higher compared to the intrinsic carrier concentration of the bulk InSb material. A small fraction of carriers could be extrinsic ones located in the dislocation region, while most carriers are intrinsic ones located in the bulk region. At 77 K, the carrier concentrations in the samples were at least two orders higher than the intrinsic carrier concentration of InSb at 77 K, which indicated the dominance of extrinsic carriers for electric current. Accordingly, most of the conductance originated from the dislocation region. Therefore, the anomalous decrease in the carrier mobility at 77 K could be attributed to the presence of the dislocation region and different conduction path of carriers at different temperatures.

The extent of decrease in carrier mobility from 300 K to 77 K varied between different samples as shown in Figure 3.8(a). The extent of decrease in the carrier mobility of Samples 3 and 4 are much smaller compared to that of Samples 1 and 2. A similar observation was reported for undoped and Sn-doped InSb layers grown on a GaAs substrate [140]. While the undoped InSb layer showed a large decrease in carrier mobility in the 77 K Hall resistance measurement, the Sn doped InSb showed only a small decrease value in carrier mobility in the low temperature Hall resistance measurement. In Sn doped InSb, extrinsic carriers exist in both the bulk region and dislocation region due to the donor impurity of Sn [140]. At low temperature, extrinsic carriers were not confined in the dislocation region and can be found in the bulk region

as well, resulting in higher carrier mobility compared to low temperature carrier mobility in the undoped InSb layer on the GaAs substrate.

Samples 3 and 4 were grown at lower temperatures compared to the optimal growth temperature of InSb. The optimal growth temperature of InSb is above 300 °C [144], and the two samples were grown at 270 °C and 220 °C. When compound semiconductors were grown below their optimal growth temperature, native point defects (anti-site, vacancy, and interstitial defects) were known to occur readily [145]. Höglund *et al.* calculated that these defects act as n-type donors in an InSb layer using density-functional theory [146]. Due to the low growth temperature of Samples 3 and 4, it is possible that native point defects contributed to the extrinsic carrier in the InSb layer of both samples. At 77 K, these defect-generated extrinsic carriers were responsible for the carrier transport in the bulk region, leading to a smaller decrease in the carrier mobility of Samples 3 and 4 compared to that of Samples 1 and 2, where an electric current at low temperatures occurred primarily through the dislocation region.

3.7 Summary

The growth temperature dependence of the formation of IMF dislocations array in the highly mismatched InSb/GaAs heteroepitaxy system was systematically investigated. In this study, the formation of Ga-Sb bonds, which is an essential step to form a uniform IMF array, was achieved through an As-for-Sb anion exchange process. The sample grown at 310 °C (Sample 2) exhibited the lowest surface roughness, smallest FWHM in the rocking curve, a uniformly distributed IMF array, and the lowest threading dislocation density. Considering the MBE system condition, the growth temperature window for the formation of uniform IMF array was 290 °C~330 °C. Growth temperature beyond the region introduced 60° misfit dislocations interrupting the

uniformity of the IMF array resulting in the development of threading dislocations. The possible origins of these 60° MDs have also been analyzed. The samples showed higher carrier mobility at 300 K than at 77 K. This observation was explained by the presence of a “dislocation region” and a “bulk region” in the InSb layer. The dislocation region is defined as the region where the carrier transport is dominated by the misfit dislocation-induced carrier scattering.

Chapter 4 Growth Optimization of III-Sb on GaAs Using Interfacial Misfit Array

4.1 Effect of Pre-Growth Sb Reconstruction on GaAs Surface

4.1.1 Introduction

The relationship between the formation of interfacial misfit (IMF) array and growth temperature at the InSb/GaAs interface was investigated in Chapter 3. InSb grown on GaAs at 310 °C was found to exhibit the lowest threading dislocation density and highest electron mobility. The results indicate that the uniformity of the IMF array is important to accommodate lattice mismatch and avoid the formation of threading dislocation. Hence, improving the uniformity of the IMF array at the III-Sb (GaSb and InSb)/GaAs interface can further increase the layers' quality. An ideal uniform IMF array requires the type of misfit dislocation (MD) at 90° and the separation between two adjacent MDs agrees with the calculated one using Equation 3.1.

The pre-growth reconstruction impacts atomic incorporation during growth, which is important for heteroepitaxial growth [147]. Previous studies have suggested that a Sb-terminated GaAs surface promoted the formation of 90° MDs compared with an As-terminated GaAs surface in both GaSb layers [29, 79] and quantum dots [148] on GaAs. In Chapter 3, the Sb reconstruction on GaAs before InSb growth is (1×3), and it is worth noting that Sb atoms can be configured in different surface reconstructions on GaAs surface [149]. Different surface reconstructions, such as (1×3), (2×8), and (2×4) Sb-terminated reconstructions, were observed at different Sb absorption temperatures. Using scanning tunneling microscopy (STM), a long-range ordered arrangement of Sb

atoms was found in the (2×8) reconstruction, while the arrangement in the (1×3) reconstruction was long-range disordered [150]. Studies of heteroepitaxial GaSb and Sb₂Te₃ on (111) Si substrates suggested that controlling Sb reconstruction on the Si surface is an effective method to increase the quality of GaSb and Sb₂Te₃, respectively [151, 152]. Therefore, it is essential to identify the effect of Sb reconstruction on GaAs, before growth, on the formation of the IMF array in III-Sb-on-GaAs heteroepitaxy.

Cross-sectional transmission electron microscope (TEM) is conventionally utilized to observe the types and distributions of MDs. However, due to its limited observation range of only a few micrometers, TEM is unsuitable for measuring threading dislocations, which are not uniformly distributed with a density $< 10^8 \text{ cm}^{-2}$. It is noted that misfit dislocations result in a non-uniform strain field, which broadens the layer peaks in an x-ray diffraction (XRD) measurement i.e. FWHM of x-ray peak [58]. In an XRD reciprocal space map (RSM), the shape of a layer's peak depends on the types of MDs (60° or 90°) and their spatial correlations (uniform or non-uniform spacing) [153]. Therefore, XRD RSM is a suitable approach to characterize threading dislocations, 60° MDs, and 90° MDs in the macroscopic observation range.

In this study, III-Sb layers were grown on (100) GaAs substrates under different Sb surface reconstructions by changing the substrate temperature (390 °C, 450 °C, 500 °C, and 560 °C) at which the Sb was adsorbed before growth. The structures and electrical properties of the III-Sb layers were characterized using atomic force microscope (AFM), high resolution XRD, cross-sectional TEM, and the Hall measurement. Simulation demonstrated the RSMs of GaSb on GaAs with the interfaces of pure 60° and 90° MDs, respectively. Based on the results of the simulation, measured XRD RSMs were used to

evaluate the MDs at the III-Sb/GaAs interface. The effect of Sb reconstruction before III-Sb growth was discussed with the results of the XRD RSMs and Hall measurement.

4.1.2 Experimental Details

GaSb and InSb layers were grown on (100) GaAs substrates using a solid-source MBE system with valved cracker As and Sb sources. After oxide desorption at 580 °C, a 100-nm-thick GaAs layer was grown to achieve a smooth GaAs surface. Subsequently, the substrate temperature was lowered to 540 °C to desorb excess As on the surface. After the desorption of the excess As, the layer exhibited a clear (2×4) reconstruction in reflection high-energy electron diffraction (RHEED): indicating an As-terminated surface. The four detailed growth procedures are:

1. The surface of the GaAs buffer was exposed to Sb₂ flux with a beam-equivalent flux of about 1.2×10^{-6} Torr to form different Sb reconstructions on GaAs surfaces through controlling the temperature of the substrate from 390 °C to 580 °C with a 10 °C step. The Sb reconstructions were monitored using RHEED, and the RHEED pattern was recorded after 4 minutes of exposure under Sb₂ flux to stabilize the pattern. Two RHEED reconstruction transitions were found. A clear (1×3) pattern was observed at 390 °C, while the (1×3) pattern began to change to a fuzzy (2×8) at 450 °C. A clear (2×8) pattern with bright and streaky lines was observed at 500 °C. At 560 °C, the (2×8) pattern gradually changed to a (2×4) pattern. A further increased temperature (600 °C) produced a clear (2×4) surface reconstruction pattern. However, such high temperature roughes the GaAs surface, therefore, it's impossible to grow III-Sb layers on GaAs with the pre-growth Sb reconstruction of (2×4). To identify the effect of pre-growth Sb reconstructions, the GaAs substrate absorbed Sb₂ at 390 °C,

450 °C, 500 °C, and 560 °C, respectively, for five minutes to form different Sb reconstructions for the subsequent growth of III-Sb layers on GaAs.

2. After the Sb absorption process, the Sb valve was closed and the substrate temperature was lowered to the growth temperature by 30 °C/min. During the cooling process, the RHEED pattern remained unchanged. It was necessary to close the Sb valve during temperature cooling because the RHEED pattern was observed to gradually change from (2×8) to (1×3) Sb reconstruction if the Sb flux was supplied at a temperature below 450 °C. Therefore, the Sb valve was closed during the cooling process to maintain the (2×8) Sb reconstructions obtained at higher temperature. To avoid the influence of different growth temperatures, the growth temperature of the GaSb layers was fixed at 390 °C and the growth temperature of InSb layers was fixed at 310 °C.
3. For the growth of the GaSb layer on GaAs, following the temperature stabilization, a 425 nm GaSb layer was grown at 390 °C with a Sb/Ga beam equivalent pressure ratio of 4 and a growth rate of 10.4 nm/min. At the initial growth stage, the streaky RHEED pattern rapidly changed to a bright spotty pattern, which indicated an island-like growth mode occurring at this stage. Subsequently, the RHEED pattern slowly recovered to a streaky (1×3) streaky pattern indicating a layer-by-layer growth mode. The recovery time was about 10 seconds for all samples. The (1×3) pattern remained until the end of growth.
4. The growth of the InSb layer on GaAs is similar to the growth of GaSb on GaAs. A 700 nm InSb layer was grown at 310 °C with a Sb/In beam equivalent pressure ratio of 3 and a growth rate of 12 nm/min. The RHEED pattern also changed to a spotty pattern at the initial stage of growth. Then it recovered to a (1×3) streaky pattern until the end of growth. The recovery time was about 30 seconds for all samples.

The XRD ω - 2θ scans and RSMs of the samples were obtained using an XRD with beam collimator and analyzer crystals to achieve a sufficiently high resolution. The IMF arrays in all samples were examined using cross-sectional TEM, and the surface morphology of the samples was measured by AFM. The carrier mobilities and concentrations of the samples were evaluated by Hall/Van der Pauw measurements.

4.1.3 XRD, AFM and TEM Measurement

Figure 4.1 (a) & (b) symmetric (004) ω - 2θ scans for GaSb and InSb on GaAs samples, respectively. The Sb reconstructions before growth are indicated; (c) & (d) Cross-sectional TEM of the interfacial misfit dislocations for GaSb and InSb grown on GaAs samples with (2×8) Sb reconstruction prior to growth, respectively.

Figure 4.1(a) and (b) shows the symmetric (004) ω - 2θ scans from GaSb and InSb on GaAs, which were grown with different Sb surface reconstructions by changing the Sb absorption temperatures. The corresponding Sb reconstructions are indicated in the

images. Only layer peaks and substrate peaks were observed. No XRD signal was observed between the layer peaks and the substrate peaks, which indicated that the III-Sb/GaAs interfaces were abrupt and there was no intermediate buffer layer between them. The vertical and parallel lattice constants ($a_{\perp}$ and $a_{\parallel}$) were extracted from the symmetric (004) and asymmetric (115) ω - 2θ scans (not presented) and are listed in Table 4.1.

Table 4.1 Pre-growth Sb reconstructions, lattice constants and relaxations for GaSb and InSb grown on GaAs with different Sb absorption temperatures before growth.

Structure	GaSb on GaAs				InSb on GaAs	
Sb absorption temperature (°C)	390	450	500	560	390	500
Sb reconstruction	(1×3)	Mixed (1×3)-(2×8)	(2×8)	Mixed (2×4)-(2×8)	(1×3)	(2×8)
$a_{\perp}$ (Å)	6.104	6.098	6.099	6.100	6.480	6.485
$a_{\parallel}$ (Å)	6.092	6.095	6.094	6.093	6.478	6.476
Relaxation (%)	99.1	99.7	99.6	99.4	99.9	99.6

Using the parallel lattice constant, the degrees of relaxation in the layers were calculated and are listed in Table 4.1. All GaSb and InSb layers presented in this study are fully relaxed. In Figure 4.1(c) and (d), the cross-sectional TEM images of III-Sb on GaAs under the Sb absorption temperature of 500 °C, where the pre-growth Sb surface reconstruction was (2×8), are presented. Uniform IMF arrays were observed in both samples and are indicated by black arrows. In GaSb/GaAs, the distance between two adjacent MDs was measured to be 5.6 nm. In InSb/GaAs, the distance was 3.2 nm in InSb/GaAs. These results were consistent with previous reports on GaSb and InSb grown on GaAs using an IMF array [29, 154]. IMF arrays are successfully formed at the interfaces.

Table 4.2 Pre-growth Sb reconstructions and surface roughnesses for GaSb/GaAs and InSb/GaAs grown with different Sb absorption temperatures before growth.

Structure	GaSb on GaAs				InSb on GaAs	
Sb absorption temperature (°C)	390	450	500	560	390	500
Sb reconstruction	(1×3)	Mixed (1×3)-(2×8)	(2×8)	Mixed (2×4)-(2×8)	(1×3)	(2×8)
r.m.s. Roughness (nm)	1.3	1.3	1.5	1.5	1.1	1.0

Figure 4.2 AFM images ($20\ \mu\text{m} \times 20\ \mu\text{m}$) of GaSb and InSb grown on (100) GaAs substrate. (a) - (d) 425 nm GaSb on GaAs with different Sb reconstructions before growth: (a) mixed (2×4)-(2×8); (b) (2×8); (c) mixed (2×8)-(1×3) and (d) (1×3). (e) - (f) 700 nm InSb on GaAs with different Sb reconstructions before growth: (a) (2×8) and (1×3).

Figure 4.2 shows the AFM images for four GaSb layers and two InSb layers grown on GaAs with different pre-growth Sb reconstructions. The root-mean-square (r.m.s.) surface roughness for each sample are listed in Table 4.2. All samples exhibited a roughness below 1.5 nm over a $20\ \mu\text{m} \times 20\ \mu\text{m}$ scan area, and no obvious cavity or pin hole was observed in the AFM images. The AFM measurement also showed that

different pre-growth Sb reconstructions had a negligible effect on the r.m.s. surface roughness.

4.1.4 X-ray RSM Simulation and Measurement

X-ray measurement ω - 2θ scans and AFM scans of III-Sb layers suggest all samples are nearly 100% relaxed and own flat surfaces. The effect of pre-growth Sb reconstructions requires further characterization using x-ray RSMs. Based on the theory developed by Kaganer *et al.* [58] and Shalimov, *et al.* [56], in a layer with a high density of defects, the relationship between the distribution of x-ray intensity and positions in symmetric (004) reciprocal space can be described as shown in Equation (4.1):

$$I(q_x, q_z) = \pi \int_0^d \frac{dz}{\sqrt{w_x w_z}} \exp\left[-\frac{(q_x - q_{x0})^2}{4w_x(z)} - \frac{(q_z - q_{z0})^2}{4w_z(z)}\right] \quad (4.1)$$

where d is the layer's thickness; (q_{x0}, q_{z0}) is the coordinate at the center of the peak, which is equal to (0,0) in the calculations; and w_x and w_z are defect-related functions that depend on both defects' density and types. It is complicated to calculate w_x and w_z in a given statistical distribution of defects in a layer. However, the calculation can be simplified with two assumptions: 1) defects only influence the distribution of x-ray intensity and 2) MD is the only form of defect in the layer. Using these two assumptions, w_x and w_z can be simplified as Equations (4.2) and (4.3):

$$w_x = \frac{\gamma \rho Q_z^2}{8\pi d} (b_x^2 + 5b_z^2) \quad (4.2)$$

$$w_z = \frac{\gamma \rho Q_z^2}{4\pi d} (b_x^2 + b_z^2) \left[\frac{v}{1-v}\right]^2 \quad (4.3)$$

where v is the layer's Poisson ratio and Q_z is the ideal reciprocal-space vector perpendicular to the diffraction direction. Q_z depends on the wavelength of incident x-ray and the lattice constants of materials. b_x and b_z are the Burgers vectors of the MD in

the interface and perpendicular to it [56]. A 90° MD is a pure edge dislocation with a Burgers vector of $\frac{a}{2}[1\bar{1}0]$, where a is the lattice constant of GaSb or InSb. Therefore, b_x is $\frac{a}{2}[1\bar{1}0]$ and b_z is 0 for a 90° MD. For a 60° MD with mixed edge and screw dislocations, b_x is $\frac{a}{4}[1\bar{1}0]$ and b_z is $\frac{a}{2}[001]$ [51]. ρ is the linear density of interfacial MDs. If all strain energy is relieved by the MDs, the linear density can be calculated using Equation (4.4) [30]

$$\rho = \frac{f}{|b_x|} \quad (4.4)$$

where f is the lattice mismatch. $1/\rho$ is the distance between adjacent dislocations. If all MDs are 90°, the distances are 5.51 nm for GaSb/GaAs and 3.14 nm for InSb/GaAs, which is closely matched with the measured distances shown in Figure 1(c) and (d). b_x in 90° MD is twice as b_x in 60° MD. Therefore, higher proportion of 60° MDs in an interfacial MDs results in an increase in linear density of interfacial MDs. When MDs are arranged periodically, i.e., distances between adjacent dislocations are kept constant throughout the interface, the non-uniform strain field decreases, leading to a narrower peak in the XRD measurement. A spatial correlation factor γ is used to reflect the arrangement of the MDs. When all dislocations are distributed randomly without spatial correlation, $\gamma = 1$, while $\gamma < 1$ means that the MDs have an ordered distribution. The model of XRD RSM is based on the scattering of x-ray induced by MDs. A uniformly distributed MDs network weakens the scattering of x-ray. Therefore, γ was included into the analysis to account for the effect of MDs distribution.

Equations (4.1)–(4.4) were used to calculate the RSMs of relaxed GaSb layers on GaAs, in which the strain energy caused by the lattice mismatch was totally relieved by either pure 60° MDs or pure 90° MDs at the GaSb/GaAs interface. These are presented in Figure 4.3(a) and (b), respectively. The x-ray intensity in the calculations is normalized.

The scans along (Q_z) and perpendicular to (Q_x) and the diffraction vectors extracted from Figure 4.3(a) and (b) are presented in Figure 4.3(c). It is evident that 60° MDs broaden the GaSb peak in both the Q_x and Q_z directions more than 90° MDs. Peak broadening in the Q_x direction is larger than that in the Q_z direction. Because Equation (4.1) is a function of a Gaussian distribution, the full-width half-maximum (FWHM) of the Q_x scan (ΔQ_x) and Q_z scan (ΔQ_z) can be calculated using the values of w_x and w_z . $\Delta Q_x/\Delta Q_z$ is 2.726 for pure 60° MDs and 1.492 for pure 90° MDs. A higher 60° MD density in an IMF array results in a higher $\Delta Q_x/\Delta Q_z$ ratio. Therefore, the type of MD can be characterized using the $\Delta Q_x/\Delta Q_z$ ratio from the XRD RSM.

Figure 4.3 Calculated (004) XRD RSMs for GaSb on GaAs: (a) pure 60° MD; (b) pure 90° MD; (c) scans along and perpendicular to the diffraction vectors for pure 60° MD and 90° MD (red curves) ; (d) Pure 90° MD with uniform separations ($\gamma=0.1$).

Figure 4.3(d) shows the calculated RSM for pure 90° MDs assuming a spatial correlation where $\gamma = 0.1$. Because Equation (4.1) is a two-dimensional Gaussian

distribution, $2\sigma_x$ and $2\sigma_z$ are standard deviations for the scans along (Q_z scan) and perpendicular to (Q_x scan) the diffraction vectors, respectively. The ΔQ_x and ΔQ_z are proportional to w_x and w_z due to ΔQ_x and ΔQ_z being the FWHM of the Q_x and Q_z scans. Moreover, Equations (4.2) and (4.3) indicate that w_x and w_z are proportional to γ . Therefore, the values of ΔQ_x and ΔQ_z decrease with a decreasing γ . However, the value of $\Delta Q_x/\Delta Q_z$ is not affected by gamma, because this ratio depends on w_x/w_z , where gamma was cancelled out. A lower γ indicates more uniform MD distribution, which weakens the scattering of x-ray and results in a smaller FWHM (ΔQ_x and ΔQ_z). The extreme case is $\gamma=0$, where both ΔQ_x and ΔQ_z are zero and the elliptic RSM becomes a dot. This is an ideal case for heterostructures where the entire strain energy caused by the lattice mismatch is relieved by uniform 90° MDs and no single threading dislocation arises from the interface. $\gamma = 0.1$ is used in the calculated results shown in Figure 4.3(d), because its result is close to the measured ones which are shown below. The only difference between Figure 4.3(b) and (d) is the value of γ . Therefore, decreases are found in the absolute value of ΔQ_x and ΔQ_z in Figure 4.3(d), while the value of the $\Delta Q_x/\Delta Q_z$ ratio is identical.

Reyner, *et al.* [153] prepared GaSb with different thicknesses (0.25 μm , 0.5 μm , 1 μm , and 1.5 μm) on GaAs. They found the $\Delta Q_x/\Delta Q_z$ ratio was thickness-dependent. The ratio decreased and gradually approached the calculated value with an increase in the layer thickness. The calculated values of $\Delta Q_x/\Delta Q_z$ for pure 60° MDs and pure 90° MDs, which are 2.726 and 1.492, respectively, are only correct in a very thick layer. In this study, both the GaSb and InSb layers are thin (425 nm for GaSb and 700 nm for InSb). Therefore, the values of the Q_x/Q_z ratio listed in Table 4.3 are higher compared to the calculated value. $\Delta Q_x/\Delta Q_z$ can be used to evaluate the type of MDs qualitatively if all the III-Sb layers grown on GaAs have an identical thickness. In short, a higher value

of $\Delta Q_x/\Delta Q_z$ indicates a higher ratio of 60° MDs to 90° MDs in an IMF array. Smaller absolute values of ΔQ_x or ΔQ_z indicate a higher degree of uniformity in the distribution of MDs.

Figure 4.4 Measured (004) RSMs of GaSb and InSb peaks of samples grown at different Sb surface reconstructions. (a)-(d) GaSb on GaAs with different Sb reconstructions before growth: (a) mixed $(2\times 4)-(2\times 8)$; (b) (2×8) ; (c) mixed $(2\times 8)-(1\times 3)$ and (d) (1×3) , respectively. (e)-(f) InSb on GaAs with different Sb reconstructions before growth: (e) (2×8) and (f) (1×3) .

Table 4.3 Pre-growth Sb reconstructions and ratios of FWHM in scans along (ΔQ_x) and perpendicular to (ΔQ_z) the diffraction vectors in RSMs for GaSb/GaAs and InSb/GaAs grown with different Sb absorption temperatures before growth.

Structure	GaSb on GaAs				InSb on GaAs	
Sb absorption temperature (°C)	390	450	500	560	390	500
Sb reconstruction	(1×3)	Mixed (1×3)-(2×8)	(2×8)	Mixed (2×4)-(2×8)	(1×3)	(2×8)
$\Delta Q_x/\Delta Q_z$	3.248	3.026	2.703	3.898	3.181	2.856

The high-resolution RSM images of GaSb/GaAs under different Sb surface reconstructions, which were induced by different absorption temperatures, are illustrated in Figure 4.4(a)–(d) with normalized x-ray intensity. The shape of the GaSb peaks changed with different Sb surface reconstructions. The absolute values of ΔQ_x and ΔQ_z in all RSMs are closest to those in the calculated RSM, in which the spatial correlation γ is about 0.1. This indicates that the arrangement of MDs is periodic in all samples. In addition, TEM observation confirms that the separation between two adjacent MDs is identical at the III-Sb/GaAs interfaces. The $\Delta Q_x/\Delta Q_z$ ratios of the samples were extracted from the measured RSMs and are listed in Table 4.3.

As mentioned above, the $\Delta Q_x/\Delta Q_z$ ratios decrease with an increasing layer thickness for the thin layers. As shown in Table 3, the values of the $\Delta Q_x/\Delta Q_z$ ratios are significantly higher compared to the calculated ratio. This is expected due to the thin layers used in the measurement. The smallest $\Delta Q_x/\Delta Q_z$ ratio in any GaSb/GaAs sample was found in the sample grown with a clear pre-growth (2×8) Sb reconstruction, where the temperature was 500 °C. Other Sb surface reconstructions resulted in elongated GaSb peaks and higher $\Delta Q_x/\Delta Q_z$ ratios. Similar results were also observed in InSb-on-GaAs

samples under different Sb surface reconstructions, as shown in Figure 4.4(e) and (f). The results of modeling suggest that a higher 90° to 60° MD ratio in the IMF array led to a smaller $\Delta Q_x/\Delta Q_z$ ratio. Therefore, the measured $\Delta Q_x/\Delta Q_z$ ratios in the III-Sb-on-GaAs samples suggest that the (2×8) Sb surface reconstruction prior to III-Sb growth promoted the formation of 90° MDs over 60° MDs in IMF arrays at the III-Sb/GaAs interface. The fuzzy (2×8) reconstructions observed at 450 °C and 560 °C arose from the transitions from (1×3) to (2×8) and (2×8) to (2×4), and the samples with these reconstructions showed higher $\Delta Q_x/\Delta Q_z$ ratios than did the samples with clear (2×8) reconstruction.

The effect of Sb reconstruction on the type of MD formation in an IMF array at the III-Sb/GaAs interface can be explained by the microstructure of the reconstruction. A previous study [30] on IMF arrays suggested that MDs were formed *via* periodic skipping of Ga–Sb or In–Sb bonds once the first monolayer Sb atoms bonded with Ga or In atoms during growth. Therefore, the distribution of MDs depends on the Sb reconstructions on the GaAs surface. Core-level and valence-band photoemission and STM images showed (2×8) reconstructions consisting of ordered long chains of absorbed Sb atoms [155]. The proposed structures of (2×8) Sb reconstruction were examined using both density functional calculations and STM, and two stable structures are determined [156]. For (1×3) reconstructions, short chains of absorbed Sb atoms with anisotropic orientations were observed in which the long-range order was absent [157]. The formation of MDs on (1×3) reconstructions results in many small dislocation circuits (the areas where MDs arrange in an ordered fashion) with different orientations. Misalignments among different dislocation circuits result in 60° MDs, and thus increase the TDD [50]. This phenomenon can be avoided when MDs form on the long-range ordered (2×8) reconstruction. The (2×4) Sb reconstruction on GaAs also has an ordered

structure [155]. However, this structure is not suitable for heteroepitaxy because the substrate temperature required to obtain (2×4) roughens the GaAs surface.

4.1.5 Hall Measurement

Figure 4.5 Dependence of hole mobility and concentration on Sb absorption temperatures for 425 nm GaSb grown on (100) GaAs substrate at (a) 300 K and (b) 77 K. The corresponding pre-growth Sb reconstructions are indicated.

Figure 4.5(a) and (b) shows the hole mobilities and hole carrier concentrations for 425 nm-thick unintentionally doped GaSb layers grown on GaAs with different Sb absorption temperatures before growth at 77 K and 300 K. Their corresponding Sb reconstructions are indicated in Figure 4.5(a). The influence of the GaAs substrate could be negligible due to its resistivity of $10^7 \Omega\text{m}$. All GaSb layers exhibited unintentional p-

type doping, which was mainly attributed to native defects such as Ga anti-site and Ga vacancy [158, 159]. As shown in Figure 4.5(a) and (b), all samples demonstrated a higher hole mobility at 77 K than at 300 K due to weak phonon scattering at low temperatures [160]. The highest hole mobilities at both 77 K and 300 K were exhibited with the GaSb layer grown with a pre-growth (2×8) Sb reconstruction. Hole carrier concentrations in all samples increased as the measurement temperature increased from 77 K to 300 K. The GaSb layer grown on GaAs with a pre-growth (2×8) Sb reconstruction exhibited the lowest hole concentration.

A previous study on undoped GaSb suggested that threading dislocations could act as acceptors leading to an increase in hole concentration [161] and a decrease in carrier mobility due to ionic impurity scattering effect [162]. Using XRD RSMs, I have shown that GaSb grown on GaAs with a pre-growth (2×8) Sb reconstruction exhibited the lowest ratio of 60° MDs to 90° MDs compared with samples with other pre-growth reconstructions. Thus, it is expected that the TDD in the sample is lower compared to other samples. Although ionized impurity and native defects can also suppress hole mobility and increase concentration [158], their effect on the mobility and concentration of GaSb layers can be neglected because all samples were grown under the same process except for pre-growth Sb reconstructions. Therefore, the results of hole mobility measurements on the GaSb layers were consistent with the results of XRD RSMs, which suggest that GaSb grown on GaAs with a pre-growth (2×8) Sb reconstruction promoted the formation of 90° MDs on an IMF array at the GaSb/GaAs interface.

Unintentionally doped InSb layers grown on the GaAs substrate are n-type. Figure 4.6(a) and (b) shows temperature dependent electron mobility and concentration in InSb layers grown with pre-growth (1×3) and (2×8) reconstructions, and the inset in Figure 4.6(a)

demonstrates a change of mobility in InSb layers from 77 K to 200 K.

Figure 4.6 Temperature dependence of (a) electron mobility and (b) concentration of 700 nm InSb grown on (100) GaAs with (1×3) and (2×8) pre-growth Sb reconstruction before InSb growth, respectively. The inset in (a) shows electron mobility below 200 K.

The samples exhibit an anomalous relationship between temperature and electron mobility, where mobility increased with an increase in temperature from 77 K to 350 K and decreased with an increase in temperature above 350 K. The characteristics of temperature dependent electron mobility and concentration could be categorized by four temperature ranges. Below 150 K, the mobility was gradually increased and the electron

concentration was nearly unchanged with an increase in temperature. From 150 K to 220 K, the mobility began to increase rapidly and a decrease in the electron concentration was observed. From 220 K to 350 K, a rapid increase in both mobility and electron concentration were observed. Above 350 K, the electron mobility decreased and the electron concentration continued to increase with an increase in temperature, which is a common behavior of semiconductors. The behaviors of mobility and concentration in response to the change in temperature can be explained by the effect of dislocations near the InSb/GaAs interface [140, 163].

The electron concentration of the InSb layer grown on GaAs at 77 K ($\sim 2.2 \times 10^{16} \text{ cm}^{-3}$) was two orders higher compared to that of the bulk InSb ($\sim 10^{14} \text{ cm}^{-3}$) [142], which indicated that most of the electrons at 77 K were extrinsic. The TEM image in Figure 4.1(d) and the previous report [154] showed the presence of high density misfit and threading dislocations near the InSb/GaAs interface. This area is referred to as the “dislocation region” in this study, while the region above with low density dislocations is called the “bulk region.” Dislocation acted as a donor in InSb to provide extrinsic electrons [164]. The influence of native point defects on concentration [145] could be neglected because both samples were grown using an identical growth temperature. Because of the small electron effective mass in the InSb and large Bohr radius of the donor impurities, the donors arising from dislocations contributed to the formation of an impurity energy band just below the conduction band in the dislocation region of the InSb layer [165, 166]. Therefore, both conduction and impurity bands existed in the dislocation region and only the conduction band existed in the bulk region. A model consisting of these two bands [165] can be used to describe the temperature dependent mobility and concentration in the InSb layer grown on GaAs. The mobility and concentration of electrons transported in this band are referred as μ_{CB} and n_{CB} ,

respectively. The impurity band only exists in the dislocation region where the mobility and concentration of electrons are referred to as μ_{IB} and n_{IB} . The measured mobility is a weighted average of μ_{CB} and μ_{IB} . The measured concentration is a sum of n_{CB} and n_{IB} , respectively. With increasing measured temperature, μ_{CB} decreases and n_{CB} increases, which has been demonstrated in n-type bulk InSb [142].

Since n_{CB} in the InSb layer decreased with a decrease in measurement temperature, its value dropped to below 10^{16} cm^{-3} compared to the measured electron concentration (above 10^{16} cm^{-3}) [142] for temperatures below 230 K. Therefore, the measured electron mobility and concentration from 77 K to 230 K mainly reflected the properties of the extrinsic carriers in the “dislocation region” [141] (i.e., the properties of the impurity band). Both theoretical and experimental results suggest the Fermi level was higher than the impurity band when the donor concentration was higher than 10^{16} cm^{-3} [166]. Therefore, the electron mobility and concentration in the impurity band demonstrated the behavior of a degenerate semiconductor. The calculation about the impurity band in InSb indicated that the electron concentration was nearly unchanged at a measurement temperature below 150 K and decreased as the temperature increased above 150 K [165]. This matched well with the measured concentration, which is shown in Figure 4.6(b), for the measurement temperature from 77 K to 230 K. Within this temperature range (77 K to 230 K), the mobility of electrons was mainly limited by a dislocation-related and impurity-related scattering process [167] [168]. The increase in carrier mobility from 77 K to 150 K could be attributed to the increase in the thermal velocities of electrons [169]. From 150 K to 230 K, the decrease in electron concentration weakened the effect of the impurity-related scattering process and led to a rapid increase in mobility. Within this temperature range, a higher dislocation density in the InSb layer increased the electron concentration and decreased mobility. As shown in the inset in Figure 4.6(a) and (b),

InSb grown with a pre-growth (2×8) Sb reconstruction exhibited higher carrier mobility and lower concentration compared to InSb grown with a pre-growth (1×3) Sb reconstruction for measurement temperature from 77 K to 230 K. Therefore, the density of interfacial MDs was expected to be lower in the InSb grown with a pre-growth (2×8) Sb reconstruction compared to that grown with a pre-growth (1×3) Sb reconstruction. A lower density of MDs was associated with a higher 90° MD-to-60° MD ratio. This result implied that the pre-growth (2×8) Sb reconstruction promoted the formation of 90° MD at the InSb/GaAs interface.

Above 230 K, the increase in the measured concentration was attributed to n_{CB} . This trend remained for a measurement temperature up to 390 K. n_{CB} is related to the intrinsic property of InSb itself rather than dislocation density. Therefore, the two InSb samples showed nearly identical electron concentrations above 270 K. An increase in n_{CB} resulted in a higher proportion of electrons transporting in the conduction band. Compared with the impurity band, which only existed in the dislocation region, the conduction band existed in both regions. The electrons transporting in the bulk region were not affected by dislocations and exhibited a higher mobility. Higher n_{CB} in the measured concentration increased the measured mobility, which explained the increasing mobility with increasing temperature. Above 230 K, InSb grown with a pre-growth (2×8) Sb reconstruction showed a higher mobility compared to InSb grown with a pre-growth (1×3) Sb reconstruction. Considering that the scattering process of impurity and phonon were identical in both samples, the difference in mobility could be attributed to the effect of the dislocation-related scattering process.

As mentioned above, 60° MDs can slip in a (111) lattice plane to form a threading dislocation, while 90° MDs cannot. Thus, pre-growth (2×8) Sb reconstruction promoted

the formation of 90° MDs compared to other reconstructions. In sum, the results of the Hall measurements were consistent with the results of the XRD RSMs, which implied that the pre-growth (2×8) Sb reconstruction promoted the formation of 90° MD in an IMF array.

4.1.6 Summary

The effect of Sb surface reconstructions on IMF arrays in GaSb/GaAs and InSb/GaAs heteroepitaxy was investigated. The RHEED observations showed that the Sb surface reconstruction exhibited two transitions when the absorption temperature was increased from 390 °C to 580 °C. GaSb and InSb were grown on a GaAs surface with different pre-growth Sb reconstructions. All of the GaSb and InSb layers had nearly 100% relaxation, as measured by (004) and (115) ω - 2θ scans. The surface morphology of GaSb and InSb layers is flat with a r.m.s. roughness below 1.5 nm. The results of modeling the GaSb/GaAs heteroepitaxy suggested that the shape of the RSM is a reliable indicator of the type of MDs in an IMF array. The RSMs obtained from the high-resolution XRD of all samples indicated a periodic arrangement of MDs. The 90° MD-to-60° MD ratio was affected by the Sb surface reconstructions. The highest 90° MD-to-60° MD ratio was achieved in a sample with a (2×8) Sb reconstruction before the III-Sb layer growth, and this phenomenon was related to the long-range ordered structure in the (2×8) Sb surface reconstruction. The Hall measurement indicated that GaSb and InSb grown with a pre-growth (2×8) Sb reconstruction exhibited a higher carrier mobility at both 300 K and 77 K compared to other samples. Both XRD RSMs and the Hall measurement suggested that pre-growth (2×8) Sb reconstruction on GaAs promote the formation of 90° MDs in the IMF array at the III-Sb/GaAs interface.

4.2 Effect of In Situ Annealing

4.2.1 Introduction

In Chapter 4.1, the growth process of III-Sb on GaAs using an IMF array is optimized by using a (2×8) pre-growth Sb reconstruction on GaAs. However, the RSMs of the III-Sb layers cannot confirm that no 60° MDs exist in the IMF array and 60° MDs are the main source for threading dislocations in the heterostructures. Moreover, the 300 K carrier mobility in those III-Sb layers is nearly half that in bulk III-Sb. Further suppression of threading dislocation is still required, especially for application in HEMTs and photodetectors. Dislocations reduce the carrier mobility, which is essential in HEMTs [170]. They also increase the dark current of photodetectors, which leads to deteriorating detection capability [171]. Therefore, techniques used to reduce threading dislocation, such as thermal annealing, are required.

Thermal annealing has been widely used in heteroepitaxy, such as Ge/Si [171, 172], GaAs/Si [173], HgCdTe/Si [174], and GaP/GaAs [175], to suppress threading dislocation and improve the device performance. The high temperature annealing process could result in annihilation and coalescence of threading dislocations [174]. An appropriate annealing process should be investigated to reduce the density of threading dislocation without disrupting the stability of the IMF array at the III-Sb/GaAs interface. Previous reports of annealing on GaSb/GaAs focused on the reduction of interfacial states rather than on improving the quality of the layer [176]. Moreover, no III-Sb/GaAs photodetectors that use an IMF array have been reported. Therefore, the effects of in situ annealing on III-Sb grown on GaAs using an IMF array and its effect on the performance of photodetectors were investigated.

In this study, GaSb and InSb layers were first grown on GaAs using an IMF array by MBE. In situ annealing was carried out after the growth of the 125 nm GaSb and 100 nm InSb layers, which are referred to as “buffer layers”. After annealing, a photoconductor structure was grown. This arrangement minimizes the effects of the annealing process on the device layer, which may consist of superlattice and doping layers. The structural and electrical properties of annealed and as-grown III-Sb/GaAs samples were investigated, and photoconductors based on III-Sb/GaAs were fabricated.

4.2.2 Experimental Details

Figure 4.7 Schematics of GaSb- and InSb-based photoconductors grown on (100) GaAs substrates.

Figure 4.7 is a schematic diagram of the photoconductors, which consisted of top interdigitate electrodes and the heterostructures of III-Sb on GaAs. All growth processes were carried out using a solid-source MBE system equipped with arsenic (As) and antimony (Sb) valved crackers to provide As₂ and Sb₂. After oxide desorption at 580 °C, a 100-nm-thick GaAs layer was grown on the GaAs substrate to achieve a smooth GaAs surface. The excess absorbed As on the GaAs surface was then removed at 540 °C by terminating the As flux supply for 10 minutes. As shown in the figure, the III-Sb buffer layers were first grown on the GaAs surface using the IMF method with a pre-growth (2×8) Sb reconstruction. GaSb buffer (125 nm) was grown at 390 °C, and InSb buffer

(100 nm) was grown at 310 °C. The detailed growth process is described in Chapter 4.1.2.

Before in situ annealing, the temperature of the GaSb or InSb layer was increased under Sb overpressure until the RHEED patterns disappeared. The temperature 10 °C below the temperature at which the RHEED pattern vanished was determined as the annealing temperature for GaSb or InSb on GaAs, respectively. In situ annealing was carried out at 590 °C for GaSb and 420 °C for InSb for different durations (5, 10, 15, and 30 minutes) in the MBE chamber and exposed to Sb₂ flux with a beam-equivalent flux of about 1.2×10^{-6} Torr. After in situ annealing, 300 nm of GaSb and 600 nm of InSb were grown at 390 °C and 310 °C, respectively. The growth rate and the III/V ratio is the same as those for GaSb and InSb buffers. To ensure the effect of annealing, a 425 nm GaSb layer and a 700 nm InSb layer were grown on the GaAs without annealing as a control sample.

After MBE growth, the x-ray ω -2 θ scan, ω scan, and reciprocal space map (RSM) were used to analyze the structural properties of the samples. The surface morphology was measured by atomic-force microscopy (AFM), and the interfacial microstructures were examined by cross-sectional TEM. The Hall/Van der Pauw measurement was used to evaluate carrier mobilities and concentrations. Interdigitated Au/Ti (80 nm/30 nm) electrodes were formed on the surface of samples using the standard photolithography, electron beam evaporation, and metal lift-off fabrication techniques. Near-infrared photoresponsivity was measured with a xenon lamp in conjunction with a monochromator. A tunable laser and a 700 °C black body radiation source were used for mid-infrared photoresponsivity measurement. The InSb photoconductors were mounted on a liquid-nitrogen-cooled optical cryostat for measurements below 300 K.

4.2.3 XRD, AFM and TEM Measurement

Figure 4.8 Symmetric (004) XRD scan of III-Sb on GaAs. (a) & (b) symmetric (004) ω - 2θ scans for GaSb and InSb, respectively. The annealing times are indicated; (c) The annealing time dependent full-width-half-maximum (FWHM) from (004) ω scans of GaSb and InSb.

Symmetric (004) ω - 2θ scans of the GaSb and InSb on GaAs with different annealing durations are shown in Figure 4.8(a) and (b). The position of the GaAs peaks in the samples is normalized to 0°, and the annealing durations of the samples are indicated above the curves. The diffraction peaks of GaSb and GaAs can be observed in Figure 4.8(a). No other obvious peak signal is observed between these two peaks, which indicates the absence of an intermediate layer between the III-Sb layers and the substrates. The GaSb peaks of all the samples are located at the same position, and

different annealing durations do not change the position of the GaSb peaks. Similar results are also found in the symmetric (004) XRD measurement of InSb on the GaAs samples in Figure 4.8(b).

The perpendicular and parallel lattice constants ($a_{\perp}$ and $a_{\parallel}$) of the III-Sb layers, which were extracted from the symmetric (004) and asymmetric (115) ω - 2θ scans (not presented here), are listed in Table 4.4. The degrees of relaxation in the GaSb and InSb layers are calculated using parallel lattice constants and are also listed in Table 4.4. The GaSb and InSb layers in all the samples are nearly 100% relaxed. The influence of the annealing duration on the degree of relaxation is not obvious.

Table 4.4 Lattice constant, relaxation, (004) rocking curve FWHM, and surface roughness for GaSb/GaAs and InSb/GaAs grown with different annealing times.

Structure	Annealing time (minute)	$a_{\perp}$ (Å)	$a_{\parallel}$ (Å)	Relaxation (%)	(004) rocking curve FWHM (arcsec)	Root mean square roughness (nm)
GaSb on GaAs	0	6.101	6.093	99.4	498	1.6
	5	6.100	6.094	99.6	471	0.9
	10	6.103	6.092	99.1	474	1.1
	15	6.104	6.091	98.9	484	0.9
	30	6.099	6.094	99.6	489	1.0
InSb on GaAs	0	6.484	6.471	99.0	540	1.8
	5	6.480	6.475	99.5	528	1.1
	10	6.487	6.472	99.1	515	0.9
	15	6.482	6.476	99.6	507	1.0
	30	6.480	6.478	99.8	526	1.0

The presence of threading dislocations broadens the full-width-half-maximum (FWHM) of the symmetric ω scan (rocking curve) [177]. Therefore, the FWHM of a (004) ω scan was measured to qualitatively compare the TDD of the samples. Figure 4.8(c) shows the relationship between the FWHM of the (004) ω scans and the annealing duration in the III-Sb layers. The values of FWHM are also listed in Table 4.4. The annealing process

effectively reduced the FWHM of the III-Sb peaks of the samples. The optimal annealing durations were 5 minutes and 15 minutes for the GaSb and InSb layers, respectively. The quality of the III-Sb layer deteriorated when the samples were annealed for longer than the optimal duration. One explanation is that the long annealing process could have changed the IMF array at the III-Sb/GaAs interfaces, resulting in the formation of a non-uniform IMF array and threading dislocations [30, 163].

Figure 4.9 Annealing time dependent r. m. s. surface roughness for III-Sb layers on GaAs. The insets are AFM images ($20 \mu\text{m} \times 20 \mu\text{m}$) for samples with different annealing times. (a) GaSb on GaAs; (b) InSb on GaAs.

Figure 4.9(a) and (b) shows the annealing duration-dependent r.m.s. surface roughness of the III-Sb layers. The insets show the AFM surface images of the samples. The r.m.s. roughness values of the samples are listed in Table 4.4. All of the GaSb/GaAs samples showed a smooth surface, and the annealing process slightly decreased the surface roughness, as shown in Figure 4.9(a). The annealed GaSb layers exhibited a r.m.s. roughness of approximately 1.0 nm. A longer annealing duration did not further decrease the roughness.

As shown in Figure 4.9(b), the surface roughness of the InSb/GaAs samples illustrated characteristics similar to those of the GaSb/GaAs samples. The annealed InSb layers exhibited a r.m.s. roughness of approximately 1.0 nm. The roughness of the as-grown sample was 1.8 nm. No obvious defects or steps were observed in the AFM images of InSb/GaAs. The annealing process decreased the surface roughness of the III-Sb layers. However, the effects of a different annealing duration on the surface roughness were not obvious. This result could be attributed to the diffusion of surface atoms during the annealing process [178].

A previous study suggested that annealing could result in interfacial diffusion and rearrangement of IMF dislocations [179]. In this study, III-Sb grown on GaAs using an IMF array exhibited a high density of misfit dislocations, which were uniformly distributed along the interface to relieve the strain energy induced by the considerable lattice mismatch. Changes in the type of separation between the misfit dislocations could promote the formation of threading dislocations [163]. Therefore, x-ray RSMs of the samples were measured to investigate the possibility of interfacial diffusion. The diffusion at the III-Sb/interface formed a mixture of III-Sb and GaAs, which was probably a metamorphic buffer in the heteroepitaxy. The metamorphic buffer could have

been tilted to the orientation of the GaAs substrate, which could not be detected by (004) ω - 2θ scanning. In contrast, the tilted buffer could be presented in an asymmetric RSM in a wafer-scale [180]. Furthermore, cross-sectional TEM was applied to observe the microstructures of IMF dislocations.

Figure 4.10 Cross-sectional TEM images and x-ray RSM of the GaSb grown on GaAs with annealing for 5 minutes. (a) Bright field image; (b) High resolution TEM image where the interfacial misfit array is indicated by black arrows; (c) Asymmetric (115) RSM where reciprocal lattice GaSb peak positions for fully relaxed and strained epitaxial layer on GaAs are indicated by dashed lines, respectively.

Figure 4.10(a) and (b) shows the TEM images of GaSb on GaAs with 5 minutes of annealing at different magnifications. A sharp GaSb/GaAs interface is observed in Figure 4.10(a), and black dots are distributed along this interface. If the entire strain energy was relieved solely by the IMF array, then the separation (S) of the adjacent dislocations can be calculated using the formula in Equation (4.5) [30]:

$$S = \frac{b}{f} \quad (4.5)$$

where b is the Burger vector of a 90° misfit dislocation and f is the lattice mismatch. The calculated separations are 5.51 nm for the GaSb/GaAs interface and 3.14 nm for the

InSb/GaAs interface. A high-resolution TEM image at the interface is shown in Figure 4.10(b), and black spots are distributed along the interface. Previous research confirmed that one black spot was a misfit dislocation in III-Sb on GaAs [28, 154] caused by the strain field induced by chemical bonds bending in misfit dislocations [181]. The separations between two adjacent dislocations were measured at 5.5 nm, which agreed well with the calculated value. These TEM observations were consistent with the observations of as-grown GaSb/GaAs demonstrated in a previous report [182].

Figure 4.10(c) shows the (115) RSM of the same sample. The diagonal and vertical dashed lines indicate the reciprocal lattice peaks for fully relaxed and fully strained GaSb layers, respectively. The GaSb layer peak of the annealed sample is located at the diagonal dashed line, which indicates that the layer was fully relaxed. Figure 4.10(c) shows only two distinguishing GaSb and Si substrate peaks in the RSM. This clearly implies that no metamorphic buffer structure induced by interfacial diffusion was formed in the annealed sample.

The TEM images of the InSb/GaAs interface after 15 minutes of annealing are shown in Figure 4.11(a) and (b). The InSb/GaAs interface is sharp and includes black spots, which indicates the presence of MDs uniformly distributed along the interface [154]. Compared with the misfit dislocations observed at the GaSb/GaAs interface, the two adjacent dislocations in InSb/GaAs are closer together (3.2 nm) due to the greater lattice mismatch. This separation agrees with the calculated value mentioned earlier.

Figure 4.11(c) shows the (115) RSM of the same sample. It indicates that the InSb layer was 100% relaxed. No metamorphic buffer structure was observed because only the peaks of InSb and GaAs are observed. The results shown in Figures 4.11 and 4.12 indicate that the IMF arrays of annealed III-Sb on GaAs samples were stable with misfit

dislocations uniformly distributed along the interfaces. The distances between the dislocations did not change, and no interfacial diffusion was seen. Compared with as-grown III-Sb layers on GaAs, the annealed samples remained 100% relaxed, which further supported the observation of a stable IMF array after annealing because the rearrangement in the IMF dislocation inevitably influenced the relaxation of the layers [183].

Figure 4.11 Cross-sectional TEM images and x-ray RSM of the InSb grown on GaAs with annealing for 15 minutes. (a) Bright field image; (b) High resolution TEM image where the interfacial misfit array is indicated by black arrows; (c) Asymmetric (115) RSM where reciprocal lattice InSb peak positions for fully relaxed and strained epitaxial layer on GaAs are indicated by dashed lines, respectively.

4.2.4 Hall Measurement

GaSb and InSb materials are importantly applied as a channel material for HEMTs because of their high hole (GaSb) and electron (InSb) mobility. A high density of threading dislocation suppresses the carrier mobility of a semiconductor layer [184]. Therefore, the effect of annealing on the TDD can be qualitatively compared by measuring the carrier mobility of III-Sb layers. In this study, both the GaSb and InSb layers were unintentionally doped. The influence of the GaAs substrate was negligible

because of its semi-insulating nature. The carrier mobilities and concentrations of the III-Sb layers at 77 and 300 K are listed in Table 4.5.

Table 4.5 300 K and 77 K carrier mobility and concentration of III-Sb on GaAs with different annealing durations.

Material	Annealing time (min)	300 K carrier mobility (cm ² /Vs)	300 K carrier concentration (cm ⁻³)	77 K carrier mobility (cm ² /Vs)	77 K carrier concentration (cm ⁻³)
GaSb*	0	475	5.38×10^{16}	2,782	5.58×10^{15}
	5	503	5.05×10^{16}	2,880	4.87×10^{15}
InSb [#]	0	24320	2.61×10^{16}	2,541	2.02×10^{16}
	15	29040	2.27×10^{16}	3,608	1.21×10^{16}

* The GaSb layers grown on GaAs in this study are unintentionally p-type.

The InSb layers grown on GaAs in this study are unintentionally n-type.

The 77 K mobilities in the GaSb layers were higher than the 300 K mobilities due to weaker phonon scattering. In contrast, the carrier concentration at 77 K was lower than that at 300 K due to the higher thermal activation at higher temperatures. The annealed GaSb layer exhibited a higher hole mobility and lower concentration than the as-grown sample at 300 K. The threading dislocations in unintentionally doped GaSb acted as acceptors to increase the hole concentration [161]. Moreover, they reduced the hole mobility because of the ionic impurity scattering effect [162]. Therefore, the higher mobility in the annealed GaSb indicated a reduction in the threading dislocation density of the GaSb layer, which was consistent with XRD ω scans. Unintentional p-type doping in the GaSb layer originated from multiple sources: 1) Ga anti-site, 2) Ga vacancy, and 3) threading dislocation-induced acceptors [159]. A previous study of annealed GaSb indicated unchanged Ga anti-sites and reduction of Ga vacancies after annealing at 580 °C under N₂ [185]. Ga vacancy (acceptor) transformed to Sb anti-site (donor) when the sample was annealed under Sb-rich conditions [186]. Thus, the reduction in hole

concentration after annealing could be attributed to the decrease in the threading dislocations and Ga vacancies.

Figure 4.12 Temperature dependence of (a) electron mobility and (b) concentration of as-grown and annealed InSb on GaAs. The inset in (a) shows mobility below 180 K. The inset in (b) shows concentration between 150 K and 300 K. The solid black line indicates the intrinsic electron concentration calculated by Equation (4.6).

Annealed InSb also exhibited a higher mobility and lower concentration than as-grown InSb, which indicated a reduction in the defect density in the annealed InSb layer. The annealed 700 nm InSb layer showed a 300 K mobility of 29,040 cm²/Vs which was higher than 750 nm InSb grown on GaAs using two-step growth process where 300 K mobility was about 20,000 cm²/Vs [187].

Previous studies also demonstrated that the annealing process significantly decreased the TDD in the layer [174, 188]. The carrier mobility and concentration in the InSb layer on GaAs exhibited abnormalities in which both the mobility and concentration were lower at 77 K than at 300 K. Figure 4.12(a) and (b) shows the temperature-dependent electron mobility and concentration in the as-grown and annealed InSb layers. The inset in Figure 4.12(a) depicts the mobility below 180 K, and the inset in Figure 4.12(b) illustrates the concentration from 150 to 300 K. The temperature-dependent intrinsic electron concentration (n_i) in bulk InSb above 200 K can be described using the formula in Equation (4.6) [5]:

$$n_i = 5.76 \times 10^{14} T^{1.5} \exp\left(-\frac{0.129}{k_B T}\right) \quad (4.6)$$

where k_B is the Boltzmann constant, T is the temperature, and $k_B T$ is given in eV. The temperature-dependent intrinsic concentration curve is shown in Figure 4.12(b).

The measured concentrations agreed well with the calculated value at temperatures above 250 K, but deviated from the calculated value at temperatures below 250 K. At 300 K, the annealed InSb layer on GaAs has a concentration of $2.27 \times 10^{16} \text{ cm}^{-3}$ which is close to the intrinsic concentration of InSb. Therefore, the previous optimized approaches [6, 7] can be used to suppress gate leakage if the InSb layers on GaAs is applied in InSb HEMTs. The intrinsic concentration was calculated to be around 10^{14}

cm^{-3} in bulk InSb at temperatures less than 230 K (much lower than the measured concentration). For both samples, the highest mobility was achieved at 340 K. Above 340 K, the phonon scattering effect became the dominant mechanism, leading to decreasing mobility with increasing temperature. The carrier mobility was nearly unchanged at temperatures below 170 K and increased rapidly as the temperature increased from 170 to 340 K. In contrast, the concentration was nearly constant below 170 K, decreased from 170 to 250 K, and then increased rapidly above 250 K.

The origins of the temperature dependent carrier mobility and concentration in the InSb layers grown on GaAs have been explained in Chapter 4.1.5. The InSb layer can be considered as a combination of a low-defect density InSb bulk layer (the bulk region) and a degenerated InSb buffer layer (the dislocation region) due to the high density of dislocation near the InSb/GaAs interface.

4.2.5 Photoresponsivity

Another important application of GaSb and InSb materials is infrared photodetection. The 300 K wavelength-dependent photoresponse of the as-grown and annealed GaSb photoconductors at the DC bias voltage of 0.5 V are shown in Figure 4.13(a). The inset is an optical microscopic image of a GaSb-based photoconductor, in which an interdigitate electrode consists of 11 strip-shaped electrodes with 15.5 μm gaps between them. The annealed and as-grown photoconductors demonstrated similar spectra response characteristics from 1.2 to 2.3 μm , after which the annealed sample exhibited greater responsivity. Both exhibited a sharp decline in responsivity at the cutoff wavelength of approximately 1.7 μm . The magnitude of responsivity is an indication of the material's minority carrier lifetime. Therefore, a high responsivity in the annealed device indicated an improvement in material quality.

Figure 4.13 Plot of DC relative responsivity vs wavelength for the as-grown and annealed (a) GaSb- and (b) InSb-based photoconductors at applied voltage of 0.5 V. The temperatures of measurement are 300 K for GaSb and 77 K for InSb. The inset in (a) is optical image of GaSb-based photoconductor. The inset in (b) is responsivity vs applied voltage for the annealed InSb-based photoconductor at 77 and 140 K.

The cutoff wavelength is near the 300 K band gap energy of GaSb (0.724 eV). Compared with the as-grown photoconductor, the downward slope is steeper in the annealed photoconductor. The electrical fields of the charged defects broaden the absorption edge of the semiconductors [189]. Thus, a steep decline in spectral responsivity indicates a lower density of defects in the annealed sample.

Figure 4.13(b) shows the 77 K photoresponsivity of the as-grown and annealed InSb photoconductors at a DC bias voltage of 0.5 V. The measurements were carried out with a tunable quantum cascade laser. Its wavelength was tunable from 4.3 to 6.6 μm with a resolution of 0.1 μm . Like the GaSb photoconductor, the annealed InSb photoconductor showed greater responsivity than the as-grown sample. The highest responsivity was achieved around 5.3 μm , which was consistent with the 77 K band gap energy of InSb (0.23 eV). The downward slope in responsivity was also steeper for the annealed sample than the as-grown sample, indicating an improvement in the material quality in the annealed InSb layer.

The 77 K, 120 K, and 160 K responsivities of the annealed InSb photoconductor were also measured at DC applied voltages from 0 to 4 V under a black body radiation source. The results are shown in the inset of Figure 4.13(b). The photocurrent increases linearly with the increase in applied voltage because the photoconductive gain is proportional to the transit time of the carriers that is enhanced in the electrical field [98]. A noticeable increase in the photocurrent is observed with a decrease in temperature due to a decrease in the number of thermally activated carriers and a longer carrier lifetime [190]. The photocurrent signal became undetectable above 220 K due to a high dark current.

4.2.6 Summary

In conclusion, GaSb and InSb were successfully grown on (100) GaAs substrates using an IMF array. The samples were annealed at different durations. The annealed samples demonstrated a lower surface roughness and defect density than the as-grown samples. The optimal annealing duration for GaSb and InSb were 5 minutes and 15 minutes, respectively. Transmission electron microscopy observation and x-ray RSM indicated a stable IMF array and the absence of interfacial diffusion or rearrangement at the III-Sb/GaAs interfaces after the annealing process. The annealed samples exhibited greater carrier mobilities. The annealed III-Sb photoconductors consistently demonstrated greater photoresponsivity than the as-grown samples.

Chapter 5 Integration of InSb Photodetector on Si *via* GaAs/Ge

5.1 Introduction

As mentioned in Chapter 1, integration of the InSb photodetector on Si is the ultimate objective in this thesis, as this approach can decrease cost and realize the coexistence of the InSb devices with other devices in the same wafer. To overcome the high lattice mismatch (20.14%) and different crystal structures between InSb and Si, an appropriate buffer is essential. Previous studies have developed high-quality Ge on Si and GaAs on Ge. Therefore, a GaAs-on-Si substrate is feasible. Moreover, Ge and GaAs are widely used in electronic and photonic devices which enable more kinds of devices to integrate with InSb on Si. Yadav *et al.* integrated InAs n-FET and Ge p-FET on a GaAs buffered Ge-on-Si substrate through IMF array and achieved a high drive current I_{ON} [191]. The growth process of InSb on GaAs using IMF array has been optimized in Chapters 3 and 4, and a high quality InSb layer on GaAs has been achieved. Therefore, the growth of an InSb layer on a GaAs buffered Ge-on-Si substrate is a potential approach to realize a high performance InSb photodetector on Si.

Previous studies about the InSb photodetector on GaAs suggested that an InAlSb electron barrier layer effectively decreased the dark current and realized the room temperature operation [119]. As introduced in Chapter 2.4.3, the barrier layer allows one kind of carriers (electrons or holes) to flow in the device and block the other one to suppress the dark current. In this chapter, an $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ electron barrier is grown in the intrinsic layer (i-layer). The bandgap of $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ is 0.54 eV at 77 K and 0.37 eV at 300 K [192], which is higher than that of InSb. Moreover, a nearly zero band offset

in the valence band is between $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ and InSb [119]. Its effect on the electrical properties of the photodetector is evaluated through comparing it with a detector without a barrier layer.

In this chapter, InSb p-i-n photodetectors with electron barrier layers are grown and fabricated on 6° offcut Ge-on-Si substrates. For the growth of an InSb layer on the GaAs buffer, all optimized conditions obtained in Chapters 3 and 4 are used. X-ray diffraction, AFM, and TEM are used to characterize the structural properties of InSb on Si *via* the Ge/GaAs buffer. Temperature dependent electrical properties and photoresponsivity in the mid-infrared range of the devices are studied.

5.2 Experimental Details

Figure 5.1 The schematic of InSb photodetector on a Si substrate. (a) The heteroepitaxial structure of InSb grown on a 6° offcut Si substrate *via* the GaAs/Ge buffer. (b) The structure of InSb p-i-n photodetector with an $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ barrier layer.

The entire epitaxial structure of the InSb photodetector on Si is shown in Figure 5.1(a), which includes a Si substrate, Ge and GaAs buffers, and an InSb device layer. The Ge-on-Si substrate was prepared through depositing 450 nm Ge on a (100) Si substrate with a 6° offcut towards $\langle 111 \rangle$ by chemical vapor deposition (CVD). The detailed growth process is given elsewhere [193]. The growth of III-V compound semiconductor layers

was carried out using a solid state molecular beam epitaxy (MBE) system with valved cracker As and Sb sources to provide As₂ and Sb₂. The Ge-on-Si substrate was first annealed at 650 °C for 20 minutes to form double atomic steps on the surface, which is essential for the suppression of the anti-phase boundary (APB) in GaAs. After annealing, the substrate was cooled down to 250 °C for the growth of a 10 nm GaAs buffer using the migration enhanced epitaxy (MEE) technique. Then the temperature of the substrate was raised to 330 °C for the growth of a 40 nm GaAs buffer with the growth rate of 100 nm/hrs under As overpressure. Finally, the temperature of the substrate was raised to 580 °C for the growth of a 150 nm GaAs buffer with the growth rate of 1000 nm/hrs under As overpressure.

The InSb device layer was grown on a GaAs buffer using the IMF method and the detailed information of the device layer is shown in Figure 5.1(b). A 100 nm InSb buffer layer without intentional doping was first grown on the GaAs surface at 310 °C using the IMF method. Prior to the growth, a (2×8) Sb surface reconstruction was formed on the GaAs surface through anion exchange. The detailed growth process is shown in Chapter 3 and Chapter 4.1. After the growth of 100 nm InSb buffer, a 15-minute thermal annealing was applied on the InSb buffer at 490 °C under Sb overpressure. The details of the annealing are shown in Chapter 4.2.

Be and Si dopants were used respectively in the p InSb layer and n InSb layer. The doping level for each layer is shown in Figure 5.1(b). The i-layer was not intentionally doped with a n-type doping level of about $2 \times 10^{16} \text{ cm}^{-3}$. The In_{0.82}Al_{0.18}Sb layer was grown on an InSb layer at 310 °C, and the flux gauge in the MBE system was used to calibrate the flux of Al and In to guarantee the composition of the In_{0.82}Al_{0.18}Sb layer.

The InSb photodetector mesa was formed using a wet etch process with a citric acid and H₂O₂ (the ratio of citric acid and H₂O₂=33:2 by weight) etchant solution to etch the InSb device layer. The etch rate is about 50 nm/min at room temperature. The targeted etch depth was about 2400 nm to expose the p⁺ InSb layer, and the etching depth was monitored using a surface profiler. Mesas with different sizes were fabricated. Metal contacts were formed at the n⁺ and p⁺ InSb layers, respectively, using the standard metal lift-off technique of electron beam-evaporated Au (150 nm)/Ti (30 nm), as shown in Figure 5.1(b). Due to the small bandgap of InSb, no thermal annealing was required for the metal contacts. No surface passivation or antireflection coating was applied on these devices. The structural properties of the InSb layer on Si were characterized using AFM, XRD, and TEM. The temperature dependent current-voltage (I-V) curves were obtained using a semiconductor parameter analyzer. The photodetectors were mounted in a liquid nitrogen cooled optical cryostage for the measurement of photoresponsivity at different temperatures and the light sources for the measurement were Fourier Transform infrared spectroscopy (FTIR), which can provide wavelengths from 1.33 μm to 27.15 μm .

5.3 Structural Characterization

Figure 5.2 (a) A 10 μm ×10 μm AFM image of InSb layer grown on Si. (b) Optical microscopic image of the fabricated InSb p-i-n photodetector on Si.

Figure 5.2(a) is a $10\ \mu\text{m} \times 10\ \mu\text{m}$ AFM image to demonstrate the surface of the InSb layer on Si *via* the Ge/GaAs buffer. The r.m.s. roughness of the surface was 1.6 nm, and no obvious defect or step was observed in the AFM image. The low surface roughness avoided the occurrence of a diffuse reflection at the surface of the photodetector to deteriorate its performance. Figure 5.2(b) contains the optical microscopic image of the InSb photodetector with different sizes where a low-defect and flat surface is shown. The mesa of the detector included a square contact area and a circular exposed area. A single annular electrode was located around the circular exposed area to collect the generated carriers. The detectors shown in Figure 5.2(b) have mesa areas of $0.0145\ \text{mm}^2$, $0.0216\ \text{mm}^2$, $0.0787\ \text{mm}^2$, and $0.172\ \text{mm}^2$ (from right to left). The larger detectors (which are not shown) have a mesa area of $0.3\ \text{mm}^2$ and $0.388\ \text{mm}^2$.

Figure 5.3 XRD characterization of InSb grown on Si. (a) (115) reciprocal space map (RSM), (b) (004) ω scan. All diffraction peaks are indicated.

Figure 5.3(a) shows the (004) ω scan of the epitaxial structure of InSb on Si. The peaks of InSb, $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$, GaAs/Ge, and Si are indicated from left to right. The vertical lattice constants of the layers were calculated using the ω separation of the peaks. The peaks of GaAs and Ge overlapped to form a single peak due to their approximate lattice constants (GaAs: $5.653\ \text{\AA}$ and Ge: $5.658\ \text{\AA}$). The peak of InAlSb located near the peak

of InSb and its weak intensity was due to its only 20-nm-thickness. The composition of the InAlSb layer can be investigated using x-ray scan. The vertical lattice constant of InAlSb was measured to be 6.353 Å, which was smaller than the unstrained lattice constant of $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ (6.417 Å). Assuming the composition of InAlSb layer is $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ and this layer was totally tensile strained on InSb at the parallel direction, the compressive vertical lattice constant of $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ was calculated to be 6.351 Å [194, 195]. The strain in the entire heteroepitaxial structure was investigated further using x-ray RSM.

Figure 5.3(b) shows an x-ray reciprocal space map (RSM) from an asymmetric (115) lattice plane, and the parallel lattice constant of all layers can be calculated using Q_x . The diagonal and vertical dashed lines indicate the reciprocal lattice peaks for fully relaxed and fully strained InSb layers, respectively. The peaks of GaAs and Ge overlapped completely, which indicate the minimal lattice mismatch between GaAs and Ge. Therefore, the GaAs layer on the Ge-on-Si substrate is a “virtual” substrate for the subsequent growth of the InSb layer. The peak of InSb located at the diagonal dashed line indicates the fully relaxed InSb layer, which agrees with the previous results about InSb on GaAs demonstrated in Chapters 3 and 4.

Using the data from Figure 5.3(a) and (b), the vertical lattice constant of InSb was calculated as 6.482 Å, and the parallel lattice constant was calculated as 6.471 Å. Therefore, the relaxation of the InSb layer was 99.4%. InAlSb peak deviated from the dashed line and owns the same value of Q_x with the InSb peak which indicated the parallel lattice constant of InAlSb was equal to that of InSb. Therefore, the InAlSb is tensile-strained. The strained InAlSb avoided the formation of MDs at the InSb/ $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ interface due to no lattice misfit. In the last paragraph, the calculated

vertical lattice constant of the InAlSb agreed well with the measured one based on the assumptions that 1) the composition of the InAlSb was $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ and 2) it was totally tensile strained at the parallel direction. X-ray RSM confirmed the layer is totally tensile-strained. Thus, the composition of the InAlSb layer is confirmed to be $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$.

Figure 5.4 (a) Bright-field TEM image for GaAs layer grown on the Ge-on-Si substrate. (b) Bright-field TEM image of InSb layer grown on GaAs where a highly periodic interfacial misfit (IMF) array locating at the InSb/GaAs interface. (c) High resolution TEM image of the IMF array shown in (b) and the separations (3.2 nm) of misfit dislocations are indicated.

The microstructures of the InSb layer and GaAs/Ge buffer were observed using cross-sectional TEM. Figure 5.4(a) shows the GaAs and Ge layers on Si. Although the lattice mismatch between Ge and Si was about 4.18%, only some defects were observed at the Ge/Si interface and no threading dislocation is observed in this image due to the optimal growth process [193]. Some black dots were observed at the Ge/GaAs interface. Considering that the lattice mismatch between GaAs and Ge could be neglected, these dots were attributed to the surface contamination rather than to dislocations because the Ge/Si substrates were exposed to the atmosphere during the transfer from CVD to MBE. Because the Ge-on-Si substrate was 6° offcut and a thermal annealing was applied on the substrate before the growth of GaAs to form double atomic steps on the Ge surface,

no APB was observed in the GaAs layer. Accordingly, the GaAs buffer was a suitable platform for the subsequent growth of InSb on GaAs using the IMF method.

Figure 5.4(b) shows the InSb/GaAs interface where a high density of black dots are observed distributed uniformly along the interface. No threading dislocation arising from the interface is observed in this image. Figure 5.4(c) is a high resolution TEM image that shows that these black dots are 90° MDs with the separation of 3.2 nm. These results were consistent with the research about the growth of InSb grown on GaAs using the IMF array in Chapters 3 and 4, which indicated that all lattice mismatches were accommodated by a highly periodical distribution of 90° MDs at the InSb/GaAs interface. The detailed analysis about the IMF array at the InSb/GaAs interface is presented in Chapters 3 and 4. The structural characterization suggested the entire heteroepitaxial structure agreed well with the design, the IMF array was successfully formed at the InSb/GaAs, and the quality of InSb layer was suitable for the further development of the InSb photodetector.

5.4 Electrical Properties

Figure 5.5(a) shows the dark current density *v.s.* bias voltage at different temperatures from 5 K to 300 K for the detector with a mesa area of 0.0787 mm². The dark current density increased with increasing temperature where the current density in the reverse bias region reduced by about nine times from 300 K to 5 K. At reverse bias $V=-50$ mV, the 300 K dark current density was 3.4 A/cm², which was smaller than the reported InSb photodetector on GaAs with the dark current density of 27 A/cm² [16]. The 80 K dark current density at 50 mV was 0.085 A/cm², which was comparable to the reported InSb photodetector on GaAs *via* 100 nm GaSb buffer with the dark current density of 0.11 A/cm² [15]. However, it was much higher than the dark current density ($\sim 10^{-7}$

A/cm²) of the InSb photodetector on InSb [196]. The dark current can be attributed to the absent of surface passivation in the InSb detector on Si and threading dislocations in InSb layer. The InSb detector was grown on a GaAs buffer on Ge/Si substrate and the threading dislocation density in the InSb grown on GaAs was reported to be 1.38×10^8 cm⁻³ in Chapter 3.4 and the effect of threading dislocation on the dark current is remarkable when its density is higher than 10^6 cm⁻² [197].

Figure 5.5 (a) Bias voltage dependent dark current density (J_{dark}) at different temperature. (b) and (c) The natural logarithm of $\ln(J_{\text{dark}})$ vs. $1/kT$ for the device presented in (a) and the solid lines are the Arrhenius fitting using Equation (5.1) from 300 K to 60 K (b) and from 100 K to 5 K (c).

To investigate the dark current limiting mechanism in this device, the temperature dependent dark current density (J_{dark}) under reverse bias $V = -50$ mV was fitted using the Arrhenius plot, which is expressed as shown in Equation (5.1):

$$J_{\text{dark}} = J_0 \exp\left(-\frac{E_a}{k_B T}\right) \quad (5.1)$$

where J_0 is the pre-exponential factor, E_a is the activation energy, k_B is the Boltzmann constant, and T is temperature. The fitted results are shown in Figure 5.5(b) and (c). Based on different activation energies, the temperature dependent J_{dark} was divided into three regions. In Region I from 300 K to 140 K, as shown in Figure 5.5(b), the activation energy was 61 ± 0.7 meV, which was about one-third of the bandgap of intrinsic InSb (0.17 eV at 300 K). Within this temperature range, the activation energy was significantly smaller than the bandgap of the semiconductor. It was always attributed to the generation-recombination (G-R) mechanism where the thermally generated carriers were excited from the valence band to the conduction band *via* the defects existing in the two bands [198]. Yet the activation energy of the G-R mechanism should be just half the bandgap of the semiconductor. Therefore, other factors should be considered, such as surface leakage [113].

In Region II from 120 K to 40 K, the activation energy decreased to about 3.0 meV, which was much less than the bandgap of intrinsic InSb. Previous research about the photodetector based on the narrow bandgap semiconductor suggested this small activation energy was attributed to surface leakage [199]. The effect of surface leakage will be investigated later by utilizing the detectors with different sizes. The activation energy in Region III (below 40 K) only included two data points measuring at 20 K and 5 K, respectively. Although the activation energy in this region could not be estimated due to limited data, the dark current density was weak temperature dependent, which indicated a different dark current limiting mechanism from Region I and II. The research about the GeSn photodetector suggested that the dark current density at an ultra-low temperature was affected by carrier freeze out [98]. However, the carrier freeze out

strongly depended on temperature [200], which was different from the result in Region III. For a narrow bandgap semiconductor-based photodetector, the effect of the tunneling mechanism could not be ignored. The tunneling mechanism, including band-to-band tunneling and trap-assisted tunneling, was nearly temperature independent [105].

The research regarding the HgCdTe photovoltaic photodetector suggested the tunneling mechanism dominated at a low temperature [201]. Therefore, the limiting mechanism in Region III should be the tunneling mechanism. This analysis indicated that the surface leakage mechanism might exist at both high and low temperature. The relationship between mesa area products at zero bias voltage (R_0A) and the perimeter-to-area (P/A) ratio of the photodetector was used to further analyze the surface leakage.

Figure 5.6 (a) and (b) The relationship between the perimeter to area (P/A) ratio and $1/R_0A$ for photodetectors with different sizes at 80 K(a) and 200 K(b). The solid lines are the fitting using Equation (5.2) and the fitted results are shown in (a) and (b), respectively. (c) 80 K dark current density for the photodetector with and without $In_{0.82}Al_{0.18}Sb$ barrier layer.

Figure 5.6(a) and (b) shows the inverse of R_0A ($[R_0A]^{-1}$) as a function of P/A at 80 K and 200 K. The measurement of the I-V curve was taken on the detectors with a mesa size from 0.0145 mm^2 ($P/A=298 \text{ cm}^{-1}$) to 0.228 mm^2 ($P/A=40 \text{ cm}^{-1}$) to obtain R_0 . At both temperatures, the $[R_0A]^{-1}$ of the devices increased with the increasing P/A , which indicated the influence of surface leakage. However, the device with the largest area ($P/A=40 \text{ cm}^{-1}$) did not agree with the linear relationship between P/A and $[R_0A]^{-1}$. An explanation for this result was due to the current localization effect for the detector with a large exposed area [16]. As shown in Figure 5.2(b), a single annular electrode around the exposed area was used to collect the current. The thermally generated carriers towards the center of the exposed area were not collected by the electrode. Consequently, the measured dark current did not cover all thermally generated carriers. A lower measured dark current density results in a lower $[R_0A]^{-1}$; therefore, the $[R_0A]^{-1}$ scaling saturated for the detector with a large exposed area.

Notably, the current localization effect also suppressed the collection of photon-generated carriers when the detector was exposed to light [184]. Therefore, the responsivity of the detector with a P/A of 40 cm^{-1} should be lower than that of the other four devices with smaller exposed areas, and it was not suitable for the measurement of responsivity.

To investigate the contributions of surface leakage and bulk to R_0A , the $[R_0A]^{-1}$ of the other four devices with smaller mesa areas were fitted using the formula in Equation (5.2):

$$\frac{1}{R_0A} = \frac{1}{R_0A_{\text{bulk}}} + \frac{1}{r_{\text{surface}}} \frac{P}{A} \quad (5.2)$$

where R_0A_{bulk} is the bulk contribution, r_{surface} is the surface resistivity, and P/A is the perimeter-to-area ratio of the detector. The fitted results (black solid lines and

parameters) at 80 K and 200 K are shown in Figure 5.6(a) and (b), respectively. Obviously, the slope of the line is $1/r_{\text{surface}}$ and the intercept is $1/R_0A_{\text{bulk}}$. At 80 K, the surface resistivity, r_{surface} , was calculated to be $88.26 \Omega\text{cm}$ and the R_0A_{bulk} was $93.28 \Omega\text{cm}^2$. At 200 K, the r_{surface} was $10.89 \Omega\text{cm}$ and the R_0A_{bulk} was $0.088 \Omega\text{cm}^2$. Therefore, the contribution of the surface leakage current to the dark current is remarkable at 80 K, and it also existed at 200 K. This result agreed with that the dark current limiting mechanism from 120 K and 40 K was surface leakage. Moreover, the effect of surface leakage on R_0A also explained that the activation energy of the temperature dependent dark current above 120 K is one-third of the bandgap of intrinsic InSb because the limiting mechanism involved both the G-R mechanism and surface leakage. Therefore, the suppression of surface leakage could decrease the dark current density of the detector.

Applying surface passivation on the detector was believed to be an effective method to decrease surface leakage [112]. To identify the effect of the $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ electron barrier layer on dark current density, a sample without this barrier layer was prepared. The growth and fabrication process of this control device (without the barrier layer) was the same as that for the device with the barrier layer. Figure 5.6(c) shows the comparison of the 80 K I-V curves of the InSb detector with and without the barrier layer. The leakage current (the current under reverse bias) was suppressed by the barrier layer by about 50%. This effect is comparable to the InSb with the same barrier layer grown on GaAs where the barrier layer suppressed the leakage current by 70% [119].

5.5 Photoresponsivity

Figure 5.7(a) shows the spectral responsivity of the InSb photodetector with the detector area of 0.0285 mm^2 from 80 K to 200 K with the step of 20 K. The measurement was carried out using a Fourier Transform Infrared (FTIR) spectrometer with a KBr beam

splitter. The built-in photodetector was replaced by one of the fabricated InSb photodetectors, and the output signal was used to measure the light intensity emitting from the MWIR source in FITR. All measurement was carried out under atmospheric pressure and the fabricated InSb detectors were mounted in a liquid nitrogen cooled optical cryostage with an ZnSe window for measurement at different temperatures.

Figure 5.7 (a) Spectral responsivity of the InSb photodetector as a function of wavelength at different temperatures. (b) The detail of the wavelengths near the absorption edge of InSb at different temperatures where the maximum value of each curve is normalized to 1.

The range of measurement is from 2.5 μm to 6.6 μm and the device operated at 0 V (i.e. photovoltaic mode) and the drop of signal at 4.2 μm was due to the absorption of CO₂ in the atmosphere. The weak responsivity at shorter wavelengths was due to the filtering effect of n-type doped layer where most of infrared light with shorter wavelengths is absorbed by n-typed doped layer and does not contribute to photocurrent [202]. The relative responsivities decreased dramatically with increasing temperature. At 80 K, the highest responsivity was achieved around 5.3 μm, which was close to the 80 K energy bandgap of InSb (0.23 eV). The cutoff wavelength increased with increasing temperature due to the bandgap narrowing with increasing temperature. This phenomenon is presented in detail in Figure 5.7(b), where the highest value of relative

responsivities at different temperatures are normalized to 1. The cutoff wavelength shifted from 5.7 μm at 80 K to 6.3 μm at 200 K due to the change in the bandgap of InSb, and the cutoff wavelength was defined to be where the responsivity decreases to 50% of the highest responsivity. At higher temperature, the responsivity was undetectable due to the increasing noise signal originating from the increasing dark current density.

Figure 5.8 (a) The absolute spectral responsivity, quantum efficiency, and (b) the calculated spectral detectivity of the InSb photodetector at 80 K.

To obtain the absolute value of responsivity, a commercial InSb photodetector with a known active area and responsivity at 5.3 μm was used to calibrate the responsivity of the fabricated InSb photodetector on Si. The commercial InSb detector was first cooled down to 77 K by liquid nitrogen, then, the build-in detector of FTIR was replaced by it. The commercial InSb photodetector was operated at 0 V, and its output current at 5.3 μm was measured. Based on the known active area and responsivity at 5.3 μm of the commercial InSb photodetector, the output power of FTIR at 5.3 μm was calculated, thus, the absolute value of responsivity at 80 K of InSb photodetector on Si was obtained and shown in Figure 5.8(a). The highest responsivity was about 0.7 A/W around 5.3 μm . Figure 5.8(b) shows the spectral quantum efficiency at 80 K, and the highest quantum efficiency is 16.3%.

Finally, the 80 K detectivity (D^*) of this device was estimated using Equation (5.3):

$$D^* = R_p \left(\frac{R_0 A}{4k_B T} \right)^{1/2} \quad (5.3)$$

where k_B is Boltzmann's constant and T is the device temperature (80 K). R_p and $R_0 A$ were obtained in the previous discussion. The calculated detectivity at 80 K is shown in Figure 5.8(b). The maximum D^* was $8.8 \times 10^9 \text{ cmHz}^{1/2} \text{W}^{-1}$. This detectivity was about one order lower than that of the InSb photodetector grown on the InSb substrate [203]. As an alternative candidate for the infrared detection, the complementary barrier infrared detector with 3.9 μm InAs/GaSb absorption layer on native substrate [117] demonstrated a 77 K responsivity of 1.1 A/W which was comparable to this InSb detector on Si considering the difference in the absorption layer thickness. However, the $R_0 A$ of the complementary barrier infrared detector was above $10^4 \Omega\text{cm}^2$ which was much higher than the $R_0 A$ of $0.71 \Omega\text{cm}^2$ for the InSb detector on Si.

Based on the previous characterization, this limited detectivity can be improved using four approaches:

1. Optimizing the barrier layer. An important reason for the low D^* is a low $R_0 A$, which is caused by a high dark current. Applying the barrier layer on a InSb detector suppresses the dark current by 50%. It is worth noting that applying a 100 nm InAs_{0.18}Sb_{0.82} barrier layer on an InAs detector on InAs increased the $R_0 A$ by six orders [33]. Therefore, the effect of the barrier layer can be further increased. For example, a 70 nm In_{0.8}Al_{0.2}Sb layer was used as barrier layer in an nBn InSb photoconductor [204]. Also, attempting a new barrier layer with a larger bandgap material and higher thickness may further impede the electron transporting in the

detector. Moreover, optimizing the growth process of the barrier layer to avoid extra defects is also important.

2. Improving the overall material quality of the InSb device layer. Due to the growth of InSb on Si, most of the dark current appears to originate from threading dislocations in the InSb device layer. As presented in Chapter 3, the imperfect IMF array results in threading dislocations. Therefore, further optimizing the uniformity of the IMF array is essential.
3. Applying surface passivation on the detector. The contribution of surface leakage to the dark current exists in the detector above 20 K, and it was also the main dark current limiting the mechanism below 120 K. Investigating an appropriate process to passivate the surface of the detector to suppress surface leakage can further increase its R_0A and detectivity, especially at 80 K.
4. Increasing the thickness of the i-layer. Beside low R_0A , the insufficient collection of incident light is another reason for the low detectivity. The highest quantum efficiency is only 17%. The absorption coefficient of InSb is about 2000 cm^{-1} [13]. Therefore, the total absorption of $(1-1/e)$ of $5.3\text{ }\mu\text{m}$ light requires a layer thickness of $5\text{ }\mu\text{m}$. Accordingly, the thickness of the i-layer can be further increased to increase the quantum efficiency.

5.6 Summary

In conclusion, InSb p-i-n photodetectors with an electron barrier layer of $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ are integrated successfully on a Si substrate *via* a GaAs/Ge buffer. Material characterization using AFM, XRD, and TEM indicate the InSb layer has a smooth surface, nearly 100% relaxation, a low-defect GaAs and Ge layer, and an IMF array at

the InSb/GaAs interface. The analysis of temperature dependent I-V curves suggests the dark limiting mechanism is attributed to both the G-R mechanism and surface leakage above 140 K and only surface leakage from 140 K to 20 K. Below 20 K, the tunneling was the main mechanism. Compared with a photodetector without a barrier layer, the dark current was 50% lower in the device with a barrier layer at 80 K. The photodetector had a cutoff wavelength of 5.7 μm at 80 K and 6.3 μm at 200 K. At 80 K, the highest responsivity of 0.7 A/W, the quantum efficiency of 16.3% and the detectivity of $8.8 \times 10^9 \text{ cmHz}^{1/2}\text{W}^{-1}$ were achieved at 5.3 μm .

Chapter 6 Integration of InSb Photoconductor on Si *via* AlSb/GaSb

6.1 Introduction

In Chapter 5, the structure and performance of an InSb-based p-i-n photodetector on Si *via* the GaAs/Ge buffers was demonstrated. The growth of Ge on Si was not finished using solid state MBE, which results in an extra cost. The surface contamination could not be avoided during the wafer transportation. Previous research showed that the GaSb layer was grown on a (100) Si substrate using MBE and that interfacial misfit dislocations accommodate the lattice mismatch [31]. An optical pumped GaSb/AlGaSb multi-quantum-well laser working at 77 K was developed based on this structure [32]. The detailed growth conditions and mechanisms for GaSb on Si are not fully understood and require further discussion. Due to a relative small lattice mismatch (6.28%) between GaSb and InSb, GaSb or other “6.1 Å family” compound semiconductors (AlSb and InAs) are suitable platforms for the growth of InSb. Moreover, “6.1 Å family” compound semiconductors are also important materials for electron and photonic devices. Therefore, utilizing them as a buffer is one approach to realize the monolithic integration of InSb devices and other kinds of devices on Si.

In this chapter, an AlSb/GaSb buffer structure is proposed and successfully grown using the IMF method, which allows the fabrication of an InSb photoconductor on (100) Si. The formation of an IMF array significantly decreases the buffer thickness and threading dislocation density in the InSb layer.

The new buffer structure included two IMF arrays. The first was located at the GaSb/Si

interface. GaSb was grown on a Si substrate instead of on the more common AlSb substrate, as AlSb on Si has been reported to have a high surface roughness [205], which can impair the subsequent growth. In addition to the threading dislocations induced by the large lattice mismatch, an anti-phase boundary (APB) is the other dominant defect in InSb on Si. When a polar III-V semiconductor compound was grown on nonpolar Si, the formation of a monoatomic step on the Si surface caused part of the atoms of the III-V compound to be arranged in opposite order from those in an ideal lattice [206]. Ko *et al.* suggested that InSb quantum dots (QDs) on an AlSb surface could act as a defect filter to decrease the APB density [170]. Therefore, following the growth of 120 nm GaSb and 80 nm AlSb, InSb QDs were grown on the AlSb surface. A GaSb layer was sequentially grown to cover the InSb QDs. The lattice mismatch between InSb and GaSb was accommodated through the second IMF array, which was separately formed at two interfaces-InSb/AlSb and InSb/GaSb. To investigate the effect of two different interfaces on the quality of the InSb layer, two samples were grown on AlSb and GaSb. Referred to as “InSb on AlSb” and “InSb on GaSb,” their structural, electrical, and optical properties were compared.

6.2 Experimental Details

The samples were grown using a solid-state molecular beam epitaxy (MBE) system equipped with antimony (Sb) valved crackers, which provided a Sb₂ flux. The entire growth process was monitored using *in-situ* reflection high-energy electron diffraction (RHEED). Before growth, the (100) Si substrates were first cleaned with an HCl/H₂O₂ solution to remove residual contamination. Then they were etched with hydrofluoric acid (HF) to remove oxides on the Si surface. The substrates were transferred to an MBE chamber and heated to 950 °C for 30 minutes to remove residual oxides. The RHEED

showed a clear (2×2) pattern, which verified a pure (100) Si surface. The growth temperature was lowered to allow the growth of GaSb under Sb overpressure with a beam-equivalent flux of about 1.2×10^{-6} Torr.

Table 6.1 The details of samples grown in this chapter

No.	Sample	Growth temperature (°C)	Description
1	GaSb on Si	380	Direct growth
2	GaSb on Si	430	Direct growth
4	GaSb on Si	380	With 2 ML AlSb
5	GaSb on Si	430	With 2 ML AlSb
6	GaSb on Si	480	With 2 ML AlSb
7	GaSb on Si	520	With 2 ML AlSb
8	InSb on AlSb	345	InSb grown on AlSb surface
9	InSb on GaSb	345	InSb grown on GaSb surface

In this study, the heteroepitaxial growth included two parts: 1) Optimize the growth conditions of GaSb on Si; 2) Optimize the AlSb/GaSb buffer structure for the growth of InSb. The details of samples are listed in Table 6.1.

To optimize the growth conditions of GaSb on Si, two sets of samples were grown at various growth temperatures from 380 °C to 520 °C. In the first set of samples, 120 nm GaSb was grown directly on Si. In the second set, two monolayers (ML) of AlSb were grown on Si before the growth of 120 nm GaSb. The growth rates of GaSb and AlSb were kept at 550 nm/hr in both sets of samples. In the direct growth of GaSb on Si, the RHEED pattern was spotty at the initial stage of growth. Then it gradually changed to a (3×3) pattern, which was consistent with a previous report [207]. For the growth of GaSb with an AlSb prelayer, the RHEED pattern remained spotty during the deposition of the 2 ML AlSb, which indicated the formation of AlSb islands rather than a continuous AlSb

layer on Si. After the deposition of AlSb, growth was interrupted for 5 minutes under Sb overpressure to enhance the nucleation of AlSb islands on Si. The RHEED pattern evolved to a (3×3) pattern during the subsequent growth of GaSb. The structural and optical properties of the GaSb layers were characterized and compared using x-ray diffraction (XRD), transmission electron microscopy (TEM), atomic force microscopy (AFM), and a near-infrared spectrophotometer system to determine the optimal growth condition for the GaSb layer. The characterization and discussion about the growth of GaSb on Si will be shown in Chapter 6.3. The optimal growth condition is determined to grow GaSb on Si at 430 °C with 2 ML AlSb which is used for the subsequent growth of the AlSb/GaSb buffer.

Figure 6.1 Schematic of InSb-based photoconductors grown on (100) Si substrates. (b) STEM image of InSb on AlSb buffer. (c) STEM image of InSb on GaSb buffer.

After the growth of 120 nm GaSb on Si under the optimal growth conditions, an 80 nm AlSb layer was grown at 430 °C at a growth rate of 550 nm/hr. Subsequently, InSb QDs were grown on the AlSb surface at 345 °C at a growth rate of 0.6 ML/s. After depositing 3 ML InSb, growth was interrupted for 30 seconds to enhance the surface migration of atoms to form QDs. The RHEED pattern became spotty once the growth of InSb QDs began, which indicated the formation of InSb islands on AlSb. Two different structures were grown on the QDs to optimize the material quality of the top InSb layer, as shown in Figure 6.1(a).

In the first structure, a 60 nm GaSb layer was grown on the QDs, followed by 40 nm AlSb. Then the top layer of InSb was grown on the AlSb layer. In the second structure, a 100 nm GaSb layer was grown on the QDs. The top InSb layer was then grown on the GaSb layer. GaSb and AlSb were both grown at 345 °C, and the RHEED pattern recovered to (3×3) during the growth of GaSb. The 700 nm top InSb layers were also grown at 345 °C at a growth rate of 730 nm/hr. At the initial stage of InSb growth, the RHEED patterns became spotty for about 20 s, and then they recovered to streaky (1×3) lines.

The structures of the samples were characterized using a scanning transmission electron microscope (STEM), as shown in Figure 6.1(b) and (c), where the bottom Si substrate, middle AlSb/GaSb buffer, and top InSb are clearly shown and their thicknesses agree with the design. Striations in Figure 6.1(b) and contrast variations in Figure 6.1(c) are observed which should be due to the imperfect focused ion beam process during the TEM sample preparation [208]. The poor resolution of the InSb/GaSb interface, as shown in Figure 6.1(c), is due to the difference between the approximate combined atomic numbers of InSb (100) and GaSb (82). The samples were also structurally

characterized using XRD. InSb photoconductors on Si were fabricated by a standard metal lift-off technique using electron beam-evaporated Au/Ti (80 nm/30 nm) to form interdigitated electrodes on the InSb surface, as shown in Figure 6.1(a). The mid-infrared photoresponsivity was measured using a 700 °C black body as the light source. The photoconductors were mounted on a liquid-nitrogen-cooled optical cryostat for measurement below 300 K and a lock-in amplifier was used as a dark current filter.

6.3 Structural Properties

A high-quality sample of GaSb on Si is important for the subsequent growth of the buffer and InSb. Therefore, to determine the effect of the growth conditions, 120 nm GaSb layers were first grown on Si under different conditions and characterized by XRD. Figure 6.2(a) shows the growth-temperature-dependent full-width-at-half-maximum (FWHM) from XRD (004) ω scans, which can be used to qualitatively compare the crystal qualities. The samples grown with an AlSb prelayer exhibited smaller FWHM values than those without an AlSb prelayer. The optimal growth temperature for GaSb on Si was 430 °C for growth both with and without the AlSb prelayer. For the GaSb layer grown directly on Si, the RHEED pattern became dim at growth temperatures above 430 °C, which indicates a very rough GaSb surface caused by the re-evaporation of Sb at a high temperature. The inset in Figure 6.2(a) displays the ω - 2θ scans for GaSb on Si grown at 430 °C with and without an AlSb prelayer. Both samples are nearly 100% relaxed.

Figure 6.2 (a) The FWHM obtained from (004) ω scans of GaSb on Si with different growth conditions. The inset in (a) is ω - 2θ scans of GaSb on Si with and without AlSb prelayer. (b) The wavelength dependent optical absorption for GaSb on Si with and without AlSb prelayer. AFM image of GaSb on Si without AlSb prelayer (c) and with AlSb prelayer (d).

The 300 K optical absorption spectrum of GaSb was measured from 1.3 to 2.5 μm to further compare growth with and without an AlSb prelayer, as shown in Figure 6.2(b). The optical absorption intensity of both samples decreased around 1.7 μm , which agreed well with the 300 K bandgap energy of GaSb (0.724 eV). The absorption spectrum of GaSb grown on Si with an AlSb prelayer showed a steeper decline at the band edge compared with that of GaSb grown without an AlSb prelayer. The absorption edge in a semiconductor can be broadened by the presence of charged defects [189]. Thus, the steeper decline indicated a lower defect density in the GaSb grown with an AlSb prelayer. Figure 6.2(c) and (d) shows the AFM images for both samples. The GaSb grown without an AlSb prelayer exhibited a very flat surface with no visible defects and a roughness

(r.m.s.) of 0.7 nm. For the GaSb grown with an AlSb prelayer, a rougher surface was presented with a roughness of 1.6 nm. A high density of surface undulation was observed in the latter case, which can be attributed to the formation of AlSb islands at the initial stage of growth.

Figure 6.3 High resolution TEM images of (a) GaSb grown directly on Si; (b) GaSb grown on Si with AlSb prelayer.

The microstructure of the GaSb/Si interface was investigated using HR-TEM (Figure 6.3(a) and (b)). As shown in Figure 6.3(a), the interface between the top black area (GaSb) and the bottom white area (Si) was not sharp, and an approximately 2-nm-thick mixture layer was observed between the two phases, which indicated interfacial diffusion. Using energy-dispersive x-ray spectrometry (EDXS), Vajargah *et al.* observed interfacial diffusion in GaSb grown directly on Si and suggested that this occurred *via* the diffusion of metallic Ga atoms into Si [209]. Bond-bending around interfacial misfit dislocations has been shown to lead to the formation of white dots at the interface [181]. These dots indicate the presence of IMF [31]. No white dots can be observed in Figure 6.3(a). However, clear white dots, which indicate the presence of an IMF array, are observed in Figure 6.3(b): showing GaSb grown on Si with a 2 ML AlSb prelayer. The separation (S) of two adjacent misfit dislocations in a highly uniform IMF array can be calculated using Equation (6.1) under the assumption that the IMF array completely

accommodates the lattice mismatch [28]:

$$S = \frac{b}{f} \quad (6.1)$$

where b is the Burgers vector of a misfit dislocation in the layer and f is the lattice mismatch between the layer and substrate. Note that the interface presented in Figure 6.3(b) is a mixture of both the GaSb/Si and AlSb/Si interfaces. The calculated separation of misfit dislocations in the IMF array at the GaSb/Si and AlSb/Si interfaces is 3.34 nm and 3.52 nm, respectively. This 0.2 nm difference between the separations is close to the resolution limit of TEM [210]. Therefore, the method is unable to distinguish between the misfit dislocations at GaSb/Si and those at AlSb/Si. The separation of misfit dislocations in Figure 6.3(b) was measured to be around 3.4 nm, which agreed well with the calculated value, and the dislocations were uniformly distributed. Previous studies suggested that the AlSb islands on Si provided the nucleation sites for the subsequent growth of GaSb, which was important for the formation of a uniform IMF array [209]. In summary, GaSb grown on Si at 430 °C with a 2 ML AlSb prelayer exhibited the highest crystal quality in all the samples. Subsequently, the growth of a GaSb buffer on Si was performed under this optimal growth condition.

Figure 6.4(a) and (b) shows the x-ray (115) reciprocal space maps (RSMs) of InSb on AlSb and on GaSb surfaces, respectively. Diagonal dashed lines indicate the fully relaxed reciprocal lattice peaks. In both samples, four peaks were found in each RSM. These peaks represent (from top to bottom) the Si, GaSb, AlSb, and InSb layers. Using the reciprocal lattice vectors Q_x and Q_z of the InSb peak in the RSM, the vertical ($a_{\perp}$) and parallel ($a_{\parallel}$) lattice constants of the InSb layer were calculated. The calculated results are listed in Table 6.2.

Figure 6.4 X-ray (115) reciprocal space maps of (a) InSb on AlSb; (b) InSb on GaSb surface.

AFM images of (a) InSb on AlSb; (b) InSb on GaSb surface.

Table 6.2 Lattice constants, relaxation, FWHM from (004) ω scan, 300 K carrier mobility and concentration of InSb on AlSb and InSb on GaSb

Structure	$a_{\perp}$ (Å)	$a_{\parallel}$ (Å)	Relaxation (%)	(004) ω scan FWHM (arcsec)	300 K carrier mobility (cm ² /Vs)	300 K carrier concentration (cm ⁻³)
InSb on AlSb	6.491	6.474	99.5	810.8	20202	3.76×10^{16}
InSb on GaSb	6.499	6.471	99.2	904.0	17700	4.25×10^{16}

The InSb layers in both samples were nearly 100% relaxed. The positions of both the GaSb and AlSb peaks deviated from the dashed line. These two peaks were actually aligned along a vertical line, which was parallel to the Q_z axis. From this line, both peaks

were calculated to have an identical $a_{||}$ of about 6.117 Å. Also, the AlSb peak deviated to the right of the dashed line, while the GaSb peak deviated to the left. The AlSb layer was compressively strained, while GaSb was slightly tensile-strained. The only observable difference between Figure 6.4(a) and (b) is the intensity of the AlSb peaks: the peak in Figure 6.4(a) is higher due to the different thicknesses of the AlSb layers in the two samples.

Figure 6.4 (c) and (d) shows the AFM images of InSb on AlSb and on GaSb surfaces, respectively. The r.m.s roughness of the InSb on AlSb and InSb on GaSb are 8.2 nm and 18.5 nm, respectively. The high roughness is attributed to island-like growth mode in the initial stage of InSb growth. Snyder, *et al.* suggested the rough surface was a way to relieve the strain energy caused by lattice mismatch [211]. The difference in the roughness of InSb on AlSb and InSb on GaSb will be discussed associated with the TEM observation.

The effectiveness of the InSb QDs in defect filtering was assessed using cross-section TEM (Figure 6.5(a)–(e)). Figure 6.5(a) and (b) shows low-magnification TEM images of InSb on AlSb. Figure 6.5(a) shows the entire AlSb/GaSb buffer structure, including the top layer of AlSb, while Figure 6.5(b) shows the InSb layer grown on AlSb with no obvious dislocations arising from the InSb/AlSb interface. In Figure 6.5(a), a high density of defects arising from the GaSb/Si interface can be seen. These defects propagated through the first GaSb and AlSb layers.

Figure 6.5 (a) & (b) Bright-field TEM images of InSb grown on (100) Si: (a) the AlSb/GaSb buffer containing InSb QDs; (b) InSb grown on AlSb. (c) high resolution TEM images of an InSb QD at the GaSb/AlSb interface. (d) and (e) diffraction patterns obtained using fast Fourier transform (FFT) in the area D and E which are indicated by white boxes.

InSb QDs with diameters of ~ 20 nm and heights of ~ 10 nm were observed at the surface of the first AlSb layer. The distribution of InSb QDs was irregular because of their preference to nucleate at defect sites. A detailed visualization of the InSb QDs is presented in Figure 6.5(c), where a defect with a different crystal structure from the

surrounding material is terminated by an InSb QD. Fast Fourier transformation of the TEM image was carried out in area D (located above an InSb QD), and in area E (which contained defects). The diffraction patterns for each area are shown in Figure 6.5(d) and (e). Both areas present standard diffraction patterns of the zinc blende structure. However, in Figure 6.5(e), extra diffraction points with weak intensity are observed between the peaks of the standard pattern. It is known that microtwin defects can tilt the crystal orientation with respect to the orientation of the surrounding material. This tilt introduces extra diffraction points, which serve as a clear indicator of the presence of microtwins [212]. As microtwins have different crystal orientations from the rest of the AlSb layer, they result in steps at the AlSb surface. Previous research suggested that microtwins could be terminated by the preferential nucleation of InSb QDs on AlSb at these steps [170]. In Figure 6.5(a), the GaSb layer above the InSb QDs is nearly defect-free due to the defect-filtering function of the InSb QDs.

Two InSb layers were grown on the AlSb and GaSb surfaces, respectively, as mentioned earlier. The InSb/AlSb interface is shown in the upper part of Figure 6.5(a) and the lower part of Figure 6.5(b), both of which show a high density of defects at the interface. When samples are grown using IMF arrays, a high density of MDs is generally formed at the interface. If all the MDs are oriented at 90° and their separation is uniform, the strain induced by the lattice mismatch can be entirely accommodated by the IMF array [30].

Threading dislocations originate from a non-uniform IMF array and the presence of 60° misfit dislocations [51]. As listed in Table 6.2, the FWHM in the (004) ω scan of InSb on AlSb is narrower than that of InSb on GaSb. This indicates that the density of threading dislocations is lower in the InSb on the AlSb sample, as the presence of threading dislocations broadens the (004) ω -scan FWHM [177]. Figure 6.5 compares

the interfacial dislocations at the InSb/AlSb and InSb/GaSb interfaces. Cross-sectional TEM images of the InSb/AlSb and InSb/GaSb interfaces are shown in Figure 6.5(a) and 6.5(c), respectively.

Figure 6.6 Bright-field TEM images for (a) InSb/AlSb interface, (b) InSb/GaSb interface. (c) and (d) phase contrast images of (a) and (b), respectively by using Fourier mask filtering technique.

Although black interfaces can be observed in both images, no further structural information can be obtained by visual inspection. To enable a meaningful analysis of the misfit dislocations at the interfaces, a Fourier mask filtering technique was applied to Figure 6.6(a) and (c) to obtain the atomic arrangements at the interfaces shown in Figure 6.6(b) and 6.6(d), respectively. A 90° misfit dislocation is associated with two

extra planes, while a 60° misfit dislocation is associated with one extra plane. The extra planes in Figure 6.6(b) and (d) are indicated by arrows. Figure 6.6(b) exhibits uniformly distributed 90° misfit dislocations at the InSb/AlSb interface. In contrast, one 60° MD and three 90° MDs are observed at the InSb/GaSb interface. The 60° misfit dislocations can glide on the (111) lattice planes to form threading dislocations, leading to a wider (004) ω -scan FWHM of InSb on GaSb compared with that of InSb on AlSb. The 60° MD relieved less strain energy than 90° MD [44, 45]. Considering the similar relaxations of InSb layers on AlSb and on GaSb, other mechanisms are involved to relieve the strain energy caused by lattice mismatch. A rough surface results in additional surface energy to relieve strain energy [211]. Therefore, the surface roughness of InSb on GaSb is higher than that of InSb on AlSb to compensate the weaker capacity of 60° MD in the relief of the strain energy caused by lattice mismatch. The presence of 60° dislocations in one sample but not the other may be due to the difference in growth modes between InSb on AlSb and InSb on GaSb.

An island-like (Volmer–Weber) growth mode, which requires a large lattice mismatch and large difference in surface energy between the layers [213], is essential for forming IMFs [182]. At the stage of island growth, pure 90° misfit dislocations form at the interface through periodic skipping of chemical bonds [30]. The islands coalesce into a continuous layer during subsequent growth. The parallel lattice constants ($a_{\parallel}$) of AlSb and GaSb were found to be identical in this study. The surface energy of InSb is known to be 17.3% lower than that of AlSb and 5.8% lower than that of GaSb [214]. This suggests that the growth mode of InSb on GaSb may be different from that of InSb on AlSb. Previous studies of InSb QDs grown on GaSb indicated that the growth of InSb on GaSb occurred *via* the Stranski–Krastanow mode [27, 215], which starts with the growth of a two-dimensional strained layer and switches to that of a relaxed island after

exceeding a critical thickness. Under the Stranski–Krastanow mode, misfit dislocations mainly form at the interface between the edge of the island and the strained layer [216, 217].

6.4 Hall Measurement

Figure 6.7 Temperature dependence of (a) electron mobility and (b) concentration of InSb on Si with different buffer structures. The inset in (a) shows mobility from 80 K to 220 K. Intrinsic concentration was calculated using Equation (2) and presented in (b). (c) and (d) The $\text{Ln}(\sigma)$ v.s. $1/T$ for (c) InSb on GaSb and (d) InSb on AlSb. σ is the dark conductivity obtained from Hall measurement. The red solid lines are the Arrhenius fitting using Equation (6.3) from 450 K to 230 K and the blue solid lines are the Arrhenius fitting using Equation (6.3) from 140 K to 77 K.

The carrier mobility in a semiconductor depends on the density of threading dislocations within it and is therefore considered as a criterion to evaluate the quality of the layer. Moreover, in a photoconductor, high carrier mobility results in high gain [37]. In this

study, the InSb layers on both AlSb and GaSb were unintentionally n-type doped. Figure 6.7(a) and (b) shows the temperature-dependent carrier mobility and concentration of InSb layers. The electron mobilities at 300 K are also listed in Table 6.2. The temperature-dependent intrinsic electron concentration (n_i) in bulk InSb above 200 K can be described using Equation (6.2) [5]:

$$n_i = 5.76 \times 10^{14} T^{1.5} \exp\left(-\frac{0.129}{k_B T}\right) \quad (6.2)$$

where k_B is the Boltzmann constant and T is the temperature. It's worth to note that the electron mobility of InSb on AlSb is 20,202 cm²/Vs and its concentration is 3.76×10¹⁶ cm⁻³ which enable it to be applied in InSb HEMT on Si. The high electron mobility is important for the channel material and the approximate concentration with the intrinsic one (as shown in Figure 6.7) avoids higher gate leakage. The temperature dependence of the mobility and concentration differ from those of other semiconductors, where the mobility decreases while the concentration increases with the increasing temperature. This difference is attributed to the dislocations near the interface (of both misfit and threading type) [163, 165, 218].

The concentrations measured above 200 K are clearly higher than the calculated n_i , indicating that extrinsic electrons participate in current transport in the InSb layer. The extrinsic electrons originate from dislocations near the InSb/AlSb and InSb/GaSb interfaces [164] and result in the formation of impurity bands in the region near the interface due to the small effective mass of the electrons and the large Bohr radius of the donor in InSb [166]. The impurity bands are located below the Fermi level of InSb, which is characteristic of degenerated semiconductors [219]. Therefore, the InSb layer on Si in this study may be a heterostructure in which the region near the interface has the properties of a degenerated semiconductor while the bulk region has the properties

of a normal semiconductor. The inhomogeneity of the InSb layer on Si results in the complicated temperature-dependent mobility and concentration, especially at a low temperature. A detailed explanation of this effect is shown in Chapter 4.1.5. At 300 K, the mobility of InSb on AlSb is higher than that of InSb on GaSb. As mentioned previously, the 60° MDs forming at the interface can glide to form threading dislocations, which act as defect-related scattering centers to suppress the mobility of carriers [168]. As the growth conditions for both InSb layers are the same, the higher mobility in InSb on AlSb indicates a lower TDD, which is consistent with the previous XRD and TEM results.

The electron mobility (μ) and concentration (n) shown in Figure 6.7 (a) and (b) can be used to calculate the dark conductivity (σ) using $\sigma=nq\mu$ where q is elementary charge. Figure 6.7 (c) and (d) show the temperature dependent σ from 77 K to 450 K for the InSb on GaSb (c) and the InSb on AlSb (d). The natural logarithm of σ ($\ln(\sigma)$) *v.s.* $1/T$ was fitted using the Arrhenius plot, which is expressed as shown in Equation (6.3):

$$\sigma = C \exp\left(-\frac{E_a}{k_B T}\right) \quad (6.3)$$

where C is the pre-exponential factor, E_a is the activation energy, k_B is the Boltzmann constant, and T is temperature. From 450 K to 230 K, the E_a for the InSb on AlSb and the InSb on GaSb was both about 0.13 eV. The 300 K and 450 K bandgap of InSb was 0.17 eV and 0.14 eV, respectively [220], therefore, this E_a was mainly attributed to the bandgap of intrinsic InSb. From 140 K to 77 K, the E_a was below 7 meV indicating another mechanism for the dark conductivity. Previous research about semiconductors suggested this low E_a was due to donor, acceptor or defects [98, 221, 222]. As mentioned above, the effect of interfacial misfit dislocations at the bottom of the InSb layer on the carrier mobility and concentration was remarkable at low temperature, therefore, this

low E_a could be attributed to the interfacial misfit dislocations at the InSb/AlSb or InSb/GaSb interface.

6.5 Photoresponsivity

Figure 6.8 Photocurrent for (a) InSb photoconductor grown on AlSb and (b) InSb photoconductor grown on GaSb at 77 K and 140 K, respectively. The inset is microscopic optical image of the InSb photoconductor. (c) Temperature dependent photocurrent at 1.0 V from 77 K to 150 for InSb photoconductor grown on AlSb and GaSb, respectively. (d) The output spectrum of a black body maintained at 700 °C.

The photocurrents of the InSb photoconductors were measured under a black body maintained at 700 °C as a function of DC applied voltage, as shown in Figure 6.8(a) and (b). The output spectrum and power density of the black body depended on its temperature. For the photoconductors, the power of incident light mainly depended on the output power density of the black body, the shape of top electrode, and the distance between the device and the black body. In this experiment, these factors were kept consistent in the measurement of different InSb photoconductors. Therefore, the

photoresponsivity was proportional to the measured photocurrent. An optical microscopic image of one of the photoconductors is shown in the inset of Figure 6.8(a), where the separation of two adjacent electrodes is about 13.3 μm . The output spectrum of the black body maintained at 700 $^{\circ}\text{C}$ is shown in Figure 6.8 (d) and its output focuses on the 3~6 μm region. The absorption coefficient of the InSb layer on GaAs at 3 μm , 4 μm , and 5 μm is 7500 cm^{-1} , 5000 cm^{-1} , and 2900 cm^{-1} , respectively [223]. If the (1-1/e) of incident light is absorbed, the absorption layer thickness should be 1.33 μm , 2 μm , and 3.45 μm for the wavelength of 3 μm , 4 μm , and 5 μm , respectively. Due to the InSb layer thickness of 700 nm in this study, it's reasonable to consider that photo-generated carriers exist in the whole InSb layer. Therefore, the measured photocurrents in different InSb photoconductors can be used to compare their difference in the material quality. The photocurrent increased linearly with the increasing voltage at both 77 and 150 K which was photoconductive gain. This gain was attributed to a low transit time under a high applied voltage [37]. The theoretical responsivity (R) of a photoconductor can be expressed as shown in Equation (6.4) [224]

$$R = \frac{q\lambda}{hc} \frac{\eta\tau\mu_e V}{L^2} \left(1 + \frac{1}{b}\right) \quad (6.4)$$

where q is the electron charge, λ is the wavelength, μ_e is electron mobility, η is the quantum efficiency, b is the electron-to-hole mobility ratio which depends on the effective mass of carriers in InSb, τ is minority carrier lifetime, L is the distance between two electrodes, and V is the applied voltage. In this study, η was a constant for the InSb photoconductors on AlSb and GaSb because η depended on the reflectance of InSb surface, layer thickness, and the absorption coefficient of InSb. Obviously, a high applied voltage resulted in a high responsivity (or photoconductive gain).

Figure 6.8(c) shows the temperature-dependent photocurrent for the InSb photoconductors on AlSb and GaSb, respectively. With the increased temperature from 77 K to 150 K, the measured responsivity was observed to decrease. According to Equation (6.4), the possible reasons of the decrease in the responsivity in this study were the decrease in the μ_e and τ . The Hall measurement indicated the μ_e increased from 77 K to 300 K. Therefore, the responsivity decreases at a higher temperature can be attributed to a lower carrier lifetime. The carrier lifetime was reported to decrease with the increasing temperature in the n-type InSb [225].

The InSb photoconductor on AlSb demonstrated a higher responsivity than that on GaSb under all measurement conditions (applied voltage and temperature). This observation is consistent with the higher TDD in the InSb on the GaSb sample discussed earlier. Threading dislocations can act as recombination centers and result in a low carrier lifetime [226]. Therefore, further suppressing the density of threading dislocations is necessary to improve the performance of InSb photoconductors on Si.

6.6 Summary

We demonstrated the growth of InSb through an AlSb/GaSb buffer on (100) Si substrates. A GaSb layer of thickness 120 nm was first grown on Si at 430 °C with an AlSb prelayer to accommodate the lattice mismatch between GaSb and Si. An AlSb layer with a thickness of 80 nm was subsequently grown, and InSb QDs were finally grown on the AlSb surface to suppress microtwinning. The success of this approach was confirmed using TEM. For the top layer of the buffer, two structures were compared. One was 100 nm GaSb, which provided a GaSb surface for the growth of InSb. The other comprised of 60 nm GaSb topped with 40 nm AlSb to provide an AlSb surface. X-ray diffraction, TEM, and Hall measurements indicated that the InSb layer on AlSb

had a narrower FWHM of ω -scanning, a more uniform interfacial misfit array, and higher mobility. The InSb photoconductors were demonstrated to operate even below 150 K, while the InSb photodetector on AlSb showed a higher responsivity than that on GaSb. Thus, the buffer structure and growth conditions proposed in this work provide an approach to integrating InSb-based devices on Si.

Chapter 7 Conclusions and Recommendations

7.1 Conclusions

In this thesis, I have developed a technique to integrate III-Sb (GaSb and InSb) on GaAs and Si substrates, and I have shown its application on the integration of an III-Sb infrared device on GaAs and Si. The heteroepitaxial structures for the integration were grown using a solid-state molecular beam epitaxy system in conjunction with a valved As and Sb cracker source to provide As₂ and Sb₂ during growth. The growth process was intentionally designed based on an interfacial misfit (IMF) array to accommodate the high lattice mismatch between the layer and the substrate. Different optimized strategies were used to increase the uniformity of the IMF array to further increase the qualities of the layers. By carefully controlling the growth conditions and further understanding the mechanism of the IMF array, high quality III-Sb layers were grown on GaAs and Si.

InSb layers were grown on GaAs at different temperatures and their structural properties were characterized using XRD, AFM, and TEM in Chapter 3. A highly uniform IMF array was only observed in the InSb grown on GaAs at 310 °C. In this thesis, the formation of a monolayer of Sb atoms on GaAs utilized an As-for-Sb anion exchange under Sb overpressure instead of depositing Sb atoms on the Ga-rich GaAs surface, which was presented in previous researches. This anion exchange was enhanced at a lower temperature. Consequently, a high growth temperature resulted in residual As atoms on the GaAs surface, which decreased the uniformity of the IMF array. A low growth temperature resulted in a high density of InSb islands at the initial stage of growth, and the coalescence of islands introduced extra 60° misfit dislocations to

decrease the uniformity of the IMF array. The Hall measurement also demonstrated that the 300 K and 77 K carrier mobilities of the InSb layer grown at 320 °C were higher than those of the other samples.

Two optimized strategies were demonstrated in Chapter 4 to further increase the crystal quality of the III-Sb layer grown using an IMF array. The TEM observation in Chapter 3 showed the 60° misfit dislocations. Undulated separations of the misfit dislocations resulted in the threading dislocation. The first strategy is controlling the pre-growth Sb reconstructions on GaAs to improve the uniformity of the IMF array. From low to high Sb absorption temperature, Sb could show three different reconstructions – (2×4), (2×8), and (1×3). III-Sb layers were grown on GaAs using the IMF array with different pre-growth Sb reconstructions. The simulation based on the functions of x-ray distribution in heteroepitaxial structures demonstrated that x-ray RSM is a sensitive method to characterize the uniformity of IMF array. The results of x-ray RSM indicated the highest uniformity was achieved in the III-Sb layers grown with pre-grown Sb (2×8) reconstruction. The Hall measurement demonstrated these layers owned higher 77 K and 300 K carrier mobilities than the other samples, which indicated high crystal quality. The effect of different Sb reconstructions was attributed to the long-range ordered structure in Sb (2×8) reconstruction.

The other optimized strategy is in situ thermal annealing, which was widely used to decrease the TDD in heteroepitaxy. In this study, the annealing was only carried out on the III-Sb buffer on GaAs while the III-Sb layer above the buffer was free of annealing, which could avoid the annealing destroying the device structure. The best annealing conditions were 5 minutes at 590 °C for GaSb and 15 minutes at 420 °C for InSb. The annealing did not introduce new defects and decrease threading dislocation density in

the III-Sb layers. The annealed III-Sb layers owned higher carrier mobility at both 77 K and 300 K. GaSb and InSb photoconductors on GaAs were also presented, and the annealing improved the photoresponsivity of the photoconductors.

Chapter 5 demonstrated an InSb p-i-n photodetector on Si where a GaAs/Ge buffer was used to decrease the lattice mismatch. A 200 nm GaAs buffer was first grown on a Ge-on-Si substrate. Then the InSb photodetector was grown on GaAs using the IMF array. The photodetector including a 20 nm $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ electron barrier to suppress the dark current. X-ray diffraction and TEM observation indicated that the InSb layer was nearly 100% relaxed and the lattice mismatch was accommodated at the Ge/Si and InSb/GaAs interfaces. High-resolution TEM observation demonstrated a uniform IMF array forming at the InSb/GaAs interface with the separation of 3.2 nm, which agreed well with the results demonstrated in Chapters 3 and 4. The activation energy of the dark current was 95 meV above 160 K and 3 meV below 120 K, and the effect of surface leakage on the dark current existed at both 80 K and 200 K. Therefore, the dark current limiting mechanism was surface leakage below 120 K, and both surface leakage and the G-R mechanism form at 300 K to 120 K. The photodetector owned a cutoff wavelength of about 5.3 μm at 80 K, and the cutoff wavelength increased with increasing temperature. The 80 K detectivity of the photodetector was $8.8 \times 10^9 \text{ cmHz}^{1/2}\text{W}^{-1}$ at 5.3 μm where the quantum efficiency is 16.3 %.

Another integration approach to integrate InSb on Si was proposed and realized in Chapter 6 where an AlSb/GaSb buffer was used. A GaSb buffer was first grown on a (100) Si substrate without offset using the IMF array to accommodate lattice mismatch. The best growth condition was 430 °C with a 2 ML AlSb prelayer. Three ML InSb QDs were grown in the buffer to decrease the density of microtwins. Transmission electron

microscopy observation confirmed that the existence of the IMF array at the GaSb/Si interface and InSb QDs could block the microtwins. InSb layers were grown on AlSb and GaSb, respectively, and both layers were nearly 100% relaxed. A higher electron mobility existed in the InSb on AlSb than on GaSb. The 300 K electron mobility of InSb on AlSb was 20202 cm²/Vs and the concentration was 3.76×10¹⁶ cm⁻³. An InSb photoconductor on Si was developed and the photoresponsivity of the InSb photoconductor on AlSb was higher than that on GaSb.

7.2 Recommendations for Further Work

The heteroepitaxial growth of the InSb layer on the Si substrate is an attractive approach to integrate the InSb device on Si in the wafer scale. In this thesis, two different buffer structures are used: 1) GaAs/Ge buffer and 2) AlSb/GaSb buffer. An InSb photoconductor on GaAs and Si and a p-i-n photodetector on Si are fabricated. To further improve the performance of the InSb device on Si, it is essential to further research the growth of InSb on GaAs and Si and optimize the architecture of the InSb photodetector.

7.2.1 Epitaxy of InSb on GaAs

For the GaAs/Ge buffer, the main kind of defect in the InSb is the threading dislocation, which originates from the imperfect IMF array at the InSb/GaAs interface. Increasing the uniformity of the IMF array is important to obtain an InSb layer with high quality on a GaAs buffer or substrate. In Chapter 4, the ordered (2×8) Sb reconstruction on the GaAs surface was identified to enhance the uniformity of the IMF array. However, a more detailed mechanism should be investigated. Actually, the (2×8) Sb reconstruction on GaAs is a complicated structure including three ML Sb atoms. Therefore, the rearrangement of Sb atoms should occur when Sb atoms bond with Ga or In atoms to

form GaSb or InSb molecules on the GaAs surface. Monitoring this phenomenon using scanning tunneling microscope would be useful to increase the uniformity of the IMF array at the III-Sb/GaAs interfaces.

Annealing is another approach to decrease the TDD in III-Sb layers on GaAs through promoting the annihilation of threading dislocations. The carrier mobilities of annealed III-Sb layers are half that of III-Sb bulk. Therefore, there is room for further increasing their crystal quality. Besides the in situ thermal annealing demonstrated in Chapter 4, other annealing methods, such as rapid thermal annealing and cycling thermal annealing, can be used to decrease the TDD. In further, several other annealing methods could be engaged to further decrease TDD in the III-Sb layers. Moreover, a recent research suggested that Be-doping on GaSb grown on GaAs using the IMF method decreased the TDD by nearly one order [227]. This result was attributed to the Be-related hardness mechanism. Investigating different doping sources (Si, Be, and Te) and methods (δ doping) may further decrease the TDD in III-Sb layers on GaAs.

7.2.2 Epitaxy of InSb on Si *via* the AlSb/GaSb Buffer

The thesis demonstrates the investigation about the growth of InSb on Si *via* the AlSb/GaSb buffer in Chapter 6. Two buffer structures are demonstrated with different top layers (AlSb and GaSb). InSb grown on AlSb is shown to be better than InSb grown on GaSb. However, the different mechanisms of InSb nucleating at the AlSb and GaSb surfaces require further detailed research. Attempting a different growth temperature, growth rate, and V/III ratio to grow InSb on AlSb and GaSb and using TEM annular dark-field imaging technology, x-ray photoelectron spectroscopy (XPS), and Auger electron spectroscopy to observe their interfacial microstructures may be useful to

further understand the different nucleation mechanisms and obtain an InSb layer with higher crystal quality.

The growth of GaSb on Si with an AlSb prelayer is an important step for the growth of the buffer, which also acts as a platform for materials with a lattice constant of around 6.1 Å. Although uniform IMF array at the GaSb/Si interface is observed using high resolution TEM, many defects are observed in the TEM image with low magnification. Optimizing the thickness and temperature of the AlSb prelayer and attempting to use an AlSb/GaSb superlattice to suppress defects are expected to further increase the crystal quality of GaSb.

7.2.3 InSb Photoconductor with a Barrier Layer

The InSb photoconductors on GaAs and Si are shown in Chapters 4 and 6, respectively. Both devices demonstrate the classic behavior of photoconductive gain below 160 K and operate at the mid-infrared range. At higher temperature, the rapid increasing dark current inundates the photocurrent and makes it undetectable. For the photoconductor based on narrow bandgap semiconductor, one of main source for the dark current is thermally generated carriers. In heteroepitaxy, defectes, especially dislocations, is another source for the dark current. To block the dark current, a new architecture of photoconductor, such as nBn and pBp, should be used where a barrier layer is inserted in a vertical photoconductor. As introduced in Chapter 2.4.2, the barrier layer can effectively block the dark current between two electrodes and enable the device to operate at higher temperature. Due to the lack of a wide bandgap binary semiconductor with the same lattice constant as InSb, investigating an appropriate ternary or quaternary semiconductor is important. Moreover, it is necessary to utilize XPS to investigate the heterointerfacial energy band alignment between InSb and the barrier layer to guarantee

that the barrier is unipolar where only electrons or holes can flow in the device unimpededly.

7.2.4 InSb p-i-n Photodetector on Si

In Chapter 5, an InSb photodetector with an electron barrier layer is integrated on Si through the Ge/Si substrate where the highest responsivity of 0.7 A/W and a detectivity of $8.8 \times 10^9 \text{ cmHz}^{1/2}\text{W}^{-1}$ are achieved at 5.3 μm at 80 K. With increasing operation temperature, the photoresponsivity of the InSb photodetector decreases significantly. Previous research suggested the electron barrier layer enables the InSb photodetector on GaAs to operate at room temperature [16, 119], which indicates there is still room to improve the performance of the InSb photodetector on Si. To overcome this problem, besides further optimizing the growth of InSb on GaAs, three other possible approaches should be attempted:

1. Surface passivation. The analysis of P/A dependent R_0A demonstrates that the surface leakage current is a source of dark current at both 80 K and 200 K. Therefore, applying surface passivation may decrease the dark current.
2. Optimizing the growth of $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$. More material characterization, such as TEM, XPS, and Auger electron spectroscopy, should be used to evaluate the quality of $\text{In}_{0.82}\text{Al}_{0.18}\text{Sb}$ and its interfaces with InSb under different growth conditions to obtain the optimized condition.
3. Increasing the thickness of the i-layer to increase quantum efficiency. The i-layer of the InSb photodetector for the absorption of light is only 2 μm (do not consider the effect of the electron barrier layer) and this layer thickness should be up to 5 μm to

absorb $(1-1/e)$ of light intensity at $5.3\ \mu\text{m}$ due to the absorption coefficient of InSb at $5.3\ \mu\text{m}$ being about $2000\ \text{cm}^{-1}$.

Figure 7.1 A proposed schematic of opto-electronic integrated circuit (OEIC) using InSb photodetectors on Si as optical signal receivers.

As proposed in Chapter 1, integrating InSb on Si makes a contribution to realize an opto-electronic integrated circuit (OEIC) operating at a mid-infrared wavelength. Figure 7.1 shows the possible structure of the OEIC including an InSb photodetector including Si CMOS, laser, modulator-on-chip waveguide, and InSb photodetector. The OEIC can utilize mid-infrared light to communicate internally and emit or receive mid-infrared for the application of detecting, imaging, and free-space communication. Components in the OEIC may consist of different materials. To minimize the kinds of material in an OEIC, it would be interesting to investigate novel buffer structures for the integration of InSb on Si, such as InAs- and InP-based buffers.

Bibliography

- [1] I. Vurgaftman, J. R. Meyer, and L. R. Ram-Mohan, "Band parameters for III-V compound semiconductors and their alloys," *Journal of Applied Physics*, vol. 89, pp. 5815-5875, 2001.
- [2] B. R. Bennett, R. Magno, J. B. Boos, W. Kruppa, and M. G. Ancona, "Antimonide-based compound semiconductors for electronic devices: A review," *Solid-State Electronics*, vol. 49, pp. 1875-1895, 2005.
- [3] A. Rogalski, "Infrared detectors: status and trends," *Progress in Quantum Electronics*, vol. 27, pp. 59-210, 2003.
- [4] H. Kroemer, "The 6.1 A family (InAs, GaSb, AlSb) and its heterostructures: a selective review," *Physica E: Low-dimensional Systems and Nanostructures*, vol. 20, pp. 196-203, 2004.
- [5] R. Cunningham and J. Gruber, "Intrinsic Concentration and Heavy-Hole Mass in InSb," *Journal of Applied Physics*, vol. 41, pp. 1804-1809, 1970.
- [6] S. Datta, T. Ashley, J. Brask, L. Buckle, M. Doczy, M. Emeny, D. Hayes, K. Hilton, R. Jefferies, and T. Martin, "85nm gate length enhancement and depletion mode InSb quantum well transistors for ultra high speed and very low power digital logic applications," in *Electron Devices Meeting, 2005. IEDM Technical Digest. IEEE International*, 2005, pp. 763-766.
- [7] N. Collaert, A. De Keersgieter, A. Dixit, I. Ferain, L. S. Lai, D. Lenoble, A. Mercha, A. Nackaerts, B. Pawlak, and R. Rooyackers, "Multi-gate devices for the 32nm technology node and beyond," *Solid-State Electronics*, vol. 52, pp. 1291-1296, 2008.
- [8] T. Ashley, L. Buckle, S. Datta, M. Emeny, D. Hayes, K. Hilton, R. Jefferies, T. Martin, T. Phillips, and D. Wallis, "Heterogeneous InSb quantum well transistors on silicon for ultra-high speed, low power logic applications," *Electronics Letters*, vol. 43, 2007.
- [9] M. Tonouchi, "Cutting-edge terahertz technology," *Nature photonics*, vol. 1, pp. 97-105, 2007.
- [10] J. Hodgkinson and R. P. Tatam, "Optical gas sensing: a review," *Measurement Science and Technology*, vol. 24, p. 012004, 2012.
- [11] M. Kopytko and A. Rogalski, "HgCdTe barrier infrared detectors," *Progress in Quantum Electronics*, vol. 47, pp. 1-18, 2016.
- [12] A. Rogalski, "HgCdTe infrared detector material: history, status and outlook," *Reports on Progress in Physics*, vol. 68, p. 2267, 2005.
- [13] M. Razeghi, "Overview of antimonide based III-V semiconductor epitaxial layers and their applications at the center for quantum devices," *European Physical Journal-Applied Physics*, vol. 23, pp. 149-205, 2003.
- [14] A. Rogalski, "Quantum well photoconductors in infrared detector technology," *Journal of Applied Physics*, vol. 93, pp. 4355-4391, 2003.

- [15] I. Kimukin, N. Biyikli, T. Kartaloglu, O. Aytur, and E. Ozbay, "High-speed InSb photodetectors on GaAs for mid-IR applications," *IEEE Journal of selected topics in quantum electronics*, vol. 10, pp. 766-770, 2004.
- [16] V. Pusino, C. Xie, A. Khalid, M. J. Steer, M. Sorel, I. G. Thayne, and D. R. Cumming, "InSb Photodiodes for Monolithic Active Focal Plane Arrays on GaAs Substrates," *IEEE Transactions on Electron Devices*, vol. 63, pp. 3135-3142, 2016.
- [17] P. V. Schwartz, C. W. Liu, and J. C. Sturm, "Semiinsulating Crystalline Silicon Formed by Oxygen Doping during Low-Temperature Chemical Vapor-Deposition," *Applied Physics Letters*, vol. 62, pp. 1102-1104, 1993.
- [18] G. Roelkens, L. Liu, D. Liang, R. Jones, A. Fang, B. Koch, and J. Bowers, "III-V/silicon photonics for on-chip and intra-chip optical interconnects," *Laser & Photonics Reviews*, vol. 4, pp. 751-779, 2010.
- [19] R. Soref, "Mid-infrared photonics in silicon and germanium," *Nature photonics*, vol. 4, pp. 495-497, 2010.
- [20] Y. C. Chang, V. Paeder, L. Hvozdar, J.-M. Hartmann, and H. P. Herzig, "Low-loss germanium strip waveguides on silicon for the mid-infrared," *Optics letters*, vol. 37, pp. 2883-2885, 2012.
- [21] A. G. Griffith, R. K. Lau, J. Cardenas, Y. Okawachi, A. Mohanty, R. Fain, Y. H. D. Lee, M. Yu, C. T. Phare, and C. B. Poitras, "Silicon-chip mid-infrared frequency comb generation," *Nature Communications*, vol. 6, 2015.
- [22] W. Kaiser, P. H. Keck, and C. Lange, "Infrared absorption and oxygen content in silicon and germanium," *Physical Review*, vol. 101, p. 1264, 1956.
- [23] H. Zimmermann, *Integrated silicon optoelectronics*. Berlin ; New York: Springer, 2000.
- [24] R. Beanland, D. Dunstan, and P. Goodhew, "Plastic relaxation and relaxed buffer layers for semiconductor epitaxy," *Advances in Physics*, vol. 45, pp. 87-146, 1996.
- [25] J. Michel, J. Liu, and L. C. Kimerling, "High-performance Ge-on-Si photodetectors," *Nature Photonics*, vol. 4, pp. 527-534, 2010.
- [26] G. Franklin, D. Rich, H. Hong, T. Miller, and T. C. Chiang, "Interface formation and growth of InSb on Si (100)," *Physical Review B*, vol. 45, p. 3426, 1992.
- [27] X. Weng, N. G. Rudawski, P. T. Wang, R. S. Goldman, D. L. Partin, and J. Heremans, "Effects of buffer layers on the structural and electronic properties of InSb films," *Journal of Applied Physics*, vol. 97, 2005.
- [28] S. H. Huang, G. Balakrishnan, A. Khoshakhlagh, A. Jallipalli, L. R. Dawson, and D. L. Huffaker, "Strain relief by periodic misfit arrays for low defect density GaSb on GaAs," *Applied Physics Letters*, vol. 88, 2006.
- [29] S. H. Huang, G. Balakrishnan, and D. L. Huffaker, "Growth Mode and Defect Evaluation of GaSb on GaAs Substrate: A Transmission Electron Microscopy Study," *Journal of Nanoscience and Nanotechnology*, vol. 11, pp. 5108-5113, 2011.
- [30] A. Jallipalli, G. Balakrishnan, S. H. Huang, A. Khoshakhlagh, L. R. Dawson,

- and D. L. Huffaker, "Atomistic modeling of strain distribution in self-assembled interfacial misfit dislocation (IMF) arrays in highly mismatched III-V semiconductor materials," *Journal of Crystal Growth*, vol. 303, pp. 449-455, 2007.
- [31] S. H. Huang, G. Balakrishnan, A. Khoshakhlagh, L. R. Dawson, and D. L. Huffaker, "Simultaneous interfacial misfit array formation and antiphase domain suppression on miscut silicon substrate," *Applied Physics Letters*, vol. 93, 2008.
- [32] A. Jallipalli, M. Kutty, G. Balakrishnan, J. Tatebayashi, N. Nuntawong, S. Huang, L. Dawson, D. Huffaker, Z. Mi, and P. Bhattacharya, "1.54 μm GaSb/AlGaSb multi-quantum-well monolithic laser at 77 K grown on miscut Si substrate using interfacial misfit arrays," *Electronics Letters*, vol. 43, pp. 1198-1199, 2007.
- [33] G. Savich, J. Pedrazzani, D. Sidor, S. Maimon, and G. Wicks, "Dark current filtering in unipolar barrier infrared detectors," *Applied Physics Letters*, vol. 99, p. 121112, 2011.
- [34] P. Martyniuk, M. Kopytko, and A. Rogalski, "Barrier infrared detectors," *Opto-Electronics Review*, vol. 22, pp. 127-146, 2014.
- [35] H. Tanoto, S. F. Yoon, W. K. Loke, E. A. Fitzgerald, C. Dohrman, B. Narayanan, M. T Doan, and C. H. Tung, "Growth of GaAs on vicinal Ge surface using low-temperature migration-enhanced epitaxy," *Journal of Vacuum Science & Technology B*, vol. 24, pp. 152-156, 2006.
- [36] K. H. Lee, A. Jandl, Y. H. Tan, E. A. Fitzgerald, and C. S. Tan, "Growth and characterization of germanium epitaxial film on silicon (001) with germane precursor in metal organic chemical vapour deposition (MOCVD) chamber," *AIP Advances*, vol. 3, p. 092123, 2013.
- [37] S. M. Sze and K. K. Ng, *Physics of semiconductor devices*: John wiley & sons, 2006.
- [38] A. E. Romanov, W. Pompe, S. Mathis, G. E. Beltz, and J. S. Speck, "Threading dislocation reduction in strained layers," *Journal of Applied Physics*, vol. 85, pp. 182-192, 1999.
- [39] A. E. Romanov, W. Pompe, G. Beltz, and J. S. Speck, "Modeling of threading dislocation density reduction in heteroepitaxial layers .2. Effective dislocation kinetics," *Physica Status Solidi B-Basic Research*, vol. 199, pp. 33-49, 1997.
- [40] A. E. Romanov, W. Pompe, G. Beltz, and J. S. Speck, "Modeling of threading dislocation density reduction in heteroepitaxial layers .1. Geometry and crystallography," *Physica Status Solidi B-Basic Research*, vol. 198, pp. 599-613, 1996.
- [41] E. A. Fitzgerald, "Dislocations in Strained-Layer Epitaxy - Theory, Experiment, and Applications," *Materials Science Reports*, vol. 7, pp. 91-142, 1991.
- [42] J. Bean, L. Feldman, A. Fiory, S. t. Nakahara, and I. Robinson, "Ge x Si $_{1-x}$ /Si strained-layer superlattice grown by molecular beam epitaxy," *Journal of Vacuum Science & Technology A: Vacuum, Surfaces, and Films*, vol. 2, pp. 436-440, 1984.
- [43] J. Narayan and S. Sharan, "Mechanism of Formation of 60-Degrees and 90-Degrees Misfit Dislocations in Semiconductor Heterostructures," *Materials*

- Science and Engineering B-Solid State Materials for Advanced Technology*, vol. 10, pp. 261-267, 1991.
- [44] J. M. Kang and S. K. Min, "Asymmetric Tilt Interface Induced by 60-Degrees Misfit Dislocation Arrays in Gasb Gaas(001)," *Applied Physics Letters*, vol. 65, pp. 2954-2956, 1994.
- [45] A. Rockett and C. J. Kiely, "Energetics of Misfit-Dislocation and Threading-Dislocation Arrays in Heteroepitaxial Films," *Physical Review B*, vol. 44, pp. 1154-1162, 1991.
- [46] J. Matthews and A. Blakeslee, "Defects in epitaxial multilayers: I. Misfit dislocations," *Journal of Crystal Growth*, vol. 27, pp. 118-125, 1974.
- [47] E. Fitzgerald, "Dislocations in strained-layer epitaxy: theory, experiment, and applications," *Materials science reports*, vol. 7, pp. 87-142, 1991.
- [48] S. A. Dregia and J. P. Hirth, "A Rebound Mechanism for Lomer Dislocation Formation in Strained Layer Structures," *Journal of Applied Physics*, vol. 69, pp. 2169-2175, 1991.
- [49] W. Qian, M. Skowronski, R. Kaspi, M. DeGraef, and V. P. Dravid, "Nucleation of misfit and threading dislocations during epitaxial growth of GaSb on GaAs(001) substrates," *Journal of Applied Physics*, vol. 81, pp. 7268-7272, 1997.
- [50] J. M. Kang, M. Nouaoura, L. Lassabatere, and A. Rocher, "Accommodation of Lattice Mismatch and Threading of Dislocations in Gasb Films Grown at Different Temperatures on Gaas(001)," *Journal of Crystal Growth*, vol. 143, pp. 115-123, 1994.
- [51] Y. Wang, P. Ruterana, S. Kret, S. El Kazzi, L. Desplanque, and X. Wallart, "The source of the threading dislocation in GaSb/GaAs hetero-structures and their propagation mechanism," *Applied Physics Letters*, vol. 102, 2013.
- [52] L. Zhang, Y. L. Shao, Y. Z. Wu, X. P. Hao, X. F. Chen, S. A. Qu, and X. G. Xu, "Characterization of dislocation etch pits in HVPE-grown GaN using different wet chemical etching methods," *Journal of Alloys and Compounds*, vol. 504, pp. 186-191, 2010.
- [53] J. E. Ayers, "The Measurement of Threading Dislocation Densities in Semiconductor Crystals by X-Ray Diffraction," *Journal of Crystal Growth*, vol. 135, pp. 71-77, 1994.
- [54] M. A. Moram and M. E. Vickers, "X-ray diffraction of III-nitrides," *Reports on Progress in Physics*, vol. 72, 2009.
- [55] P. B. Rago and J. E. Ayers, "Dynamical X-ray Diffraction from In (x) Ga1-x As Heterostructures with Dislocations," *Journal of Electronic Materials*, vol. 42, pp. 2450-2458, 2013.
- [56] A. Shalimov, J. Bak-Misiuk, V. M. Kaganer, M. Calamiotou, and A. Georgakilas, "Strain nonuniformity in GaAs heteroepitaxial films on Si(001) studied by x-ray diffraction," *Journal of Applied Physics*, vol. 101, 2007.
- [57] B. Liu, R. Zhang, Z. L. Xie, H. Lu, Q. J. Liu, Z. Zhang, Y. Li, X. Q. Xiu, P. Chen, P. Han, S. L. Gu, Y. Shi, Y. D. Zheng, and W. J. Schaff, "Microstructure and dislocation of epitaxial InN films revealed by high resolution x-ray diffraction (vol 103, art no 023504, 2008)," *Journal of Applied Physics*, vol. 104, 2008.
-

- [58] V. M. Kaganer, R. Kohler, M. Schmidbauer, R. Opitz, and B. Jenichen, "X-ray diffraction peaks due to misfit dislocations in heteroepitaxial structures," *Physical Review B*, vol. 55, pp. 1793-1810, 1997.
- [59] W. Hoke, R. Leoni, C. Whelan, T. Kennedy, A. Torabi, P. Marsh, Y. Zhang, C. Xu, and K. Hsieh, "Material properties and performance of metamorphic optoelectronic integrated circuits grown by molecular beam epitaxy on GaAs substrates," *Journal of Vacuum Science & Technology B: Microelectronics and Nanometer Structures Processing, Measurement, and Phenomena*, vol. 22, pp. 1554-1557, 2004.
- [60] T. L. Tran, F. Hatami, W. T. Maselink, V. P. Kunets, and G. J. Salamo, "Comparison of MBE Growth of InSb on Si (001) and GaAs (001)," *Journal of Electronic Materials*, vol. 37, pp. 1799-1805, 2008.
- [61] R. M. Biefeld and J. D. Phillips, "Growth of InSb on GaAs using InAlSb buffer layers," *Journal of Crystal Growth*, vol. 209, pp. 567-571, 2000.
- [62] J. S. Speck, M. A. Brewer, G. Beltz, A. E. Romanov, and W. Pompe, "Scaling laws for the reduction of threading dislocation densities in homogeneous buffer layers," *Journal of Applied Physics*, vol. 80, pp. 3808-3816, 1996.
- [63] N. Bertru, O. Brandt, M. Wassermeier, and K. Ploog, "Growth mode, strain relief, and segregation of (Ga,In)Sb on GaSb(001) grown by molecular beam epitaxy," *Applied Physics Letters*, vol. 68, pp. 31-33, 1996.
- [64] J. L. Davis and P. E. Thompson, "Molecular beam epitaxy growth of InSb films on GaAs," *Applied Physics Letters*, vol. 54, pp. 2235-2237, 1989.
- [65] M. Mori, N. Akae, K. Uotani, N. Fujimoto, T. Tambo, and C. Tatsuyama, "Heteroepitaxial growth of InSb films on a Si (0 0 1) substrate via AlSb buffer layer," *Applied Surface Science*, vol. 216, pp. 569-574, 2003.
- [66] E. Michel, J. Xu, J. Kim, I. Ferguson, and M. Razeghi, "InSb infrared photodetectors on Si substrates grown by molecular beam epitaxy," *IEEE Photonics Technology Letters*, vol. 8, pp. 673-675, 1996.
- [67] J. I. Chyi, D. Biswas, S. Iyer, N. Kumar, H. Morkoc, R. Bean, K. Zanio, H. Y. Lee, and H. Chen, "Molecular beam epitaxial growth and characterization of InSb on Si," *Applied physics letters*, vol. 54, pp. 1016-1018, 1989.
- [68] L. Li, Y. Hsu, and W. Wang, "Molecular-beam epitaxial growth of InSb on GaAs and Si for infrared detector applications," *Journal of Vacuum Science & Technology B: Microelectronics and Nanometer Structures Processing, Measurement, and Phenomena*, vol. 11, pp. 872-874, 1993.
- [69] W. Liu, J. Winesett, W. Ma, X. Zhang, M. Santos, X. Fang, and P. McCann, "Molecular beam epitaxy of InSb on Si substrates using fluoride buffer layers," *Journal of applied physics*, vol. 81, pp. 1708-1714, 1997.
- [70] Y. Kim, J. Lee, Y. Noh, M. Kim, Y. Kwon, J. Oh, and R. Gronsky, "Effect of two-step growth on the heteroepitaxial growth of InSb thin film on Si (001) substrate: a transmission electron microscopy study," *Applied physics letters*, vol. 89, p. 031919, 2006.
- [71] S. H. Shin, J. D. Song, J. Y. Lim, H. C. Koo, and T. G. Kim, "Structural and electrical properties of high-quality 0.41 μm -thick InSb films grown on GaAs

- (100) substrate with $\text{In}_x\text{Al}_{1-x}\text{Sb}$ continuously graded buffer," *Materials Research Bulletin*, vol. 47, pp. 2927-2930, 2012.
- [72] T. Sato, M. Akabori, and S. Yamada, "High-quality highly mismatched InSb films grown on GaAs substrate via thick AlSb and $\text{In}_x\text{Al}_{1-x}\text{Sb}$ step-graded buffers," *Physica E-Low-Dimensional Systems & Nanostructures*, vol. 21, pp. 615-619, 2004.
- [73] W. Qian, M. Skowronski, and R. Kaspi, "Dislocation density reduction in GaSb films grown on GaAs substrates by molecular beam epitaxy," *Journal of the Electrochemical Society*, vol. 144, pp. 1430-1434, 1997.
- [74] T. D. Mishima, M. Edirisooriya, N. Goel, and M. B. Santos, "Dislocation filtering by $\text{Al}_x\text{In}_{1-x}\text{Sb}/\text{Al}_y\text{In}_{1-y}\text{Sb}$ interfaces for InSb-based devices grown on GaAs (001) substrates," *Applied Physics Letters*, vol. 88, 2006.
- [75] F. C. Frank and J. H. Vandermerwe, "One-Dimensional Dislocations .1. Static Theory," *Proceedings of the Royal Society of London Series a-Mathematical and Physical Sciences*, vol. 198, pp. 205-216, 1949.
- [76] V. Tokranov, P. Nagaiah, M. Yakimov, R. J. Matyi, and S. Oktyabrsky, "AlGaAsSb superlattice buffer layer for p-channel GaSb quantum well on GaAs substrate," *Journal of Crystal Growth*, vol. 323, pp. 35-38, 2011.
- [77] T. Soga, H. Nishikawa, T. Jimbo, and M. Umeno, "Very Low Dislocation Density GaAs on Si Using Superlattices Grown by MOCVD," *Journal of Crystal Growth*, vol. 107, pp. 479-482, 1991.
- [78] S. H. Huang, G. Balakrishnan, and D. L. Huffaker, "Interfacial misfit array formation for GaSb growth on GaAs," *Journal of Applied Physics*, vol. 105, 2009.
- [79] J. E. Bickel and J. M. Millunchick, "The impact of the initial surface reconstruction on heteroepitaxial film growth and defect formation," *Physica Scripta*, vol. 89, 2014.
- [80] T. J. Grassman, M. R. Brenner, S. Rajagopalan, R. Unocic, R. Dehoff, M. Mills, H. Fraser, and S. A. Ringel, "Control and elimination of nucleation-related defects in GaP/Si(001) heteroepitaxy," *Applied Physics Letters*, vol. 94, 2009.
- [81] R. M. Sieg, S. A. Ringel, S. M. Ting, E. A. Fitzgerald, and R. N. Sacks, "Anti-phase domain-free growth of GaAs on offcut (001) Ge wafers by molecular beam epitaxy with suppressed Ge outdiffusion," *Journal of Electronic Materials*, vol. 27, pp. 900-907, 1998.
- [82] G. A. Devenyi, S. Y. Woo, S. Ghanad-Tavakoli, R. A. Hughes, R. N. Kleiman, G. A. Botton, and J. S. Preston, "The role of vicinal silicon surfaces in the formation of epitaxial twins during the growth of III-V thin films," *Journal of Applied Physics*, vol. 110, 2011.
- [83] S. F. Fang, K. Adomi, S. Iyer, H. Morkoc, H. Zabel, C. Choi, and N. Otsuka, "Gallium-Arsenide and Other Compound Semiconductors on Silicon," *Journal of Applied Physics*, vol. 68, pp. R31-R58, 1990.
- [84] K. C. Nunna, S. L. Tan, C. J. Reyner, A. R. J. Marshall, B. L. Liang, A. Jallipalli, J. P. R. David, and D. L. Huffaker, "Short-Wave Infrared GaInAsSb Photodiodes Grown on GaAs Substrate by Interfacial Misfit Array Technique," *Ieee*

- Photonics Technology Letters*, vol. 24, pp. 218-220, 2012.
- [85] A. Jallipalli, M. N. Kutty, G. Balakrishnan, J. Tatebayashi, N. Nuntawong, S. H. Huang, L. R. Dawson, D. L. Huffaker, Z. Mi, and P. Bhattacharya, "1.54 μm GaSb/AlGaSb multi-quantum-well monolithic laser at 77 K grown on miscut Si substrate using interfacial misfit arrays," *Electronics Letters*, vol. 43, pp. 1198-1199, 2007.
 - [86] J. Tatebayashi, A. Jallipalli, M. N. Kutty, S. H. Huang, G. Balakrishnan, L. R. Dawson, and D. L. Huffaker, "Room-temperature lasing at 1.82 μm of GaInSb/AlGaSb quantum wells grown on GaAs substrates using an interfacial misfit array," *Applied Physics Letters*, vol. 91, 2007.
 - [87] J. S. Liu, M. B. Clavel, R. Pandey, S. Datta, M. Meeker, G. A. Khodaparast, and M. K. Hudait, "Growth and characterization of metamorphic InAs/GaSb tunnel heterojunction on GaAs by molecular beam epitaxy," *Journal of Applied Physics*, vol. 119, 2016.
 - [88] A. Rogalski, "History of infrared detectors," *Opto-Electronics Review*, vol. 20, pp. 279-308, 2012.
 - [89] C. Mailhiot and D. Smith, "Long-wavelength infrared detectors based on strained InAs-Ga_{1-x}In_xSb type-II superlattices," *Journal of Vacuum Science & Technology A: Vacuum, Surfaces, and Films*, vol. 7, pp. 445-449, 1989.
 - [90] A. Rogalski, J. Antoszewski, and L. Faraone, "Third-generation infrared photodetector arrays," *Journal of applied physics*, vol. 105, p. 4, 2009.
 - [91] Y. Wei, A. Gin, M. Razeghi, and G. J. Brown, "Type II InAs/GaSb superlattice photovoltaic detectors with cutoff wavelength approaching 32 μm ," *Applied Physics Letters*, vol. 81, pp. 3675-3677, 2002.
 - [92] S. A. Pour, B. Nguyen, S. Bogdanov, E. Huang, and M. Razeghi, "Demonstration of high performance long wavelength infrared type II InAs/GaSb superlattice photodiode grown on GaAs substrate," *Applied Physics Letters*, vol. 95, p. 173505, 2009.
 - [93] C. G. Burguete, D. Guo, P. Jurczak, F. Cui, M. Tang, W. Chen, Z. Deng, Y. Chen, M. Gutiérrez, and B. Chen, "Direct growth of InAs/GaSb type II superlattice photodiodes on silicon substrates," *IET Optoelectronics*, 2017.
 - [94] D. A. Neaman, *Semiconductor physics and devices*: Irwin Chicago, 1992.
 - [95] M. Johnson, "Sweep-Out Effects in Hg_{1-x}Cd_xTe Photoconductors," *Journal of Applied Physics*, vol. 43, pp. 3090-3093, 1972.
 - [96] B. R. Conley, J. Margetis, W. Du, H. Tran, A. Mosleh, S. A. Ghetmiri, J. Tolle, G. Sun, R. Soref, and B. Li, "Si based GeSn photoconductors with a 1.63 A/W peak responsivity and a 2.4 μm long-wavelength cutoff," *Applied Physics Letters*, vol. 105, p. 221117, 2014.
 - [97] Y. Yu, Y. Jiang, K. Zheng, Z. Zhu, X. Lan, Y. Zhang, Y. Zhang, and X. Xuan, "Ultralow-voltage and high gain photoconductor based on ZnS: Ga nanoribbons for the detection of low-intensity ultraviolet light," *Journal of Materials Chemistry C*, vol. 2, pp. 3583-3588, 2014.
 - [98] B. R. Conley, A. Mosleh, S. A. Ghetmiri, W. Du, R. A. Soref, G. Sun, J. Margetis, J. Tolle, H. A. Naseem, and S.-Q. Yu, "Temperature dependent spectral response
-

- and detectivity of GeSn photoconductors on silicon for short wave infrared detection," *Optics express*, vol. 22, pp. 15639-15652, 2014.
- [99] Y. Lee, J. Kwon, E. Hwang, C. H. Ra, W. J. Yoo, J. H. Ahn, J. H. Park, and J. H. Cho, "High-performance perovskite-graphene hybrid photodetector," *Advanced materials*, vol. 27, pp. 41-46, 2015.
- [100] O. Lopez-Sanchez, D. Lembke, M. Kayci, A. Radenovic, and A. Kis, "Ultrasensitive photodetectors based on monolayer MoS₂," *Nature nanotechnology*, vol. 8, pp. 497-501, 2013.
- [101] A. Rogalski, P. Martyniuk, and M. Kopytko, "Challenges of small-pixel infrared detectors: a review," *Reports on Progress in Physics*, vol. 79, 2016.
- [102] C. Downs and T. E. Vandervelde, "Progress in Infrared Photodetectors Since 2000," *Sensors*, vol. 13, pp. 5054-5098, 2013.
- [103] S. Singh, V. Gopal, R. Bhan, and V. Kumar, "An analysis of the dynamic resistance variation as a function of reverse bias voltage in a HgCdTe diode," *Semiconductor science and technology*, vol. 15, p. 752, 2000.
- [104] X. Li, D. Jiang, Y. Zhang, D. Wang, Q. Yu, T. Liu, H. Ma, and L. Zhao, "Investigation of dark current mechanisms on type-II InAs/GaSb superlattice very long wavelength infrared detectors," *Journal of Physics D: Applied Physics*, vol. 49, p. 165105, 2016.
- [105] A. Rogalski, K. Adamiec, and J. Rutkowski, *Narrow-gap semiconductor photodiodes* vol. 77: SPIE Press, 2000.
- [106] A. Beattie and P. Landsberg, "Auger effect in semiconductors," in *Proceedings of the Royal Society of London A: Mathematical, Physical and Engineering Sciences*, 1959, pp. 16-29.
- [107] W. Tennant, "'Rule 07' revisited: Still a good heuristic predictor of p/n HgCdTe photodiode performance?," *Journal of Electronic Materials*, vol. 39, pp. 1030-1035, 2010.
- [108] C. Canedy, E. Aifer, J. Warner, I. Vurgaftman, E. Jackson, J. Tischler, S. Powell, K. Olver, J. Meyer, and W. Tennant, "Controlling dark current in type-II superlattice photodiodes," *Infrared Physics & Technology*, vol. 52, pp. 326-334, 2009.
- [109] X. C. Li, D. W. Jiang, Y. Zhang, D. B. Wang, Q. J. Yu, T. Liu, H. H. Ma, and L. C. Zhao, "Investigation of dark current mechanisms on type-II InAs/GaSb superlattice very long wavelength infrared detectors," *Journal of Physics D: Applied Physics*, vol. 49, 2016.
- [110] J. Y. Wong, "Effect of trap tunneling on the performance of long-wavelength Hg_{1-x}Cd_xTe photodiodes," *IEEE Transactions on Electron Devices*, vol. 27, pp. 48-57, 1980.
- [111] P.-Y. Delaunay, A. Hood, B. M. Nguyen, D. Hoffman, Y. Wei, and M. Razeghi, "Passivation of type-II In As/Ga Sb double heterostructure," *Applied Physics Letters*, vol. 91, p. 091112, 2007.
- [112] I. Bloom and Y. Nemirovsky, "Surface passivation of backside-illuminated indium antimonide focal plane array," *IEEE transactions on electron devices*, vol. 40, pp. 309-314, 1993.
-

- [113] A. M. Itsuno, J. D. Phillips, and S. Velicu, "Mid-wave infrared HgCdTe nBn photodetector," *Applied Physics Letters*, vol. 100, p. 161102, 2012.
- [114] S. Maimon and G. Wicks, "nBn detector, an infrared detector with reduced dark current and higher operating temperature," *Applied Physics Letters*, vol. 89, p. 151109, 2006.
- [115] P. Klipstein, "XBn barrier photodetectors for high sensitivity and high operating temperature infrared sensors," in *Proc. SPIE*, 2008, p. 69402U.
- [116] E. Weiss, O. Klin, S. Grossmann, N. Snapi, I. Lukomsky, D. Aronov, M. Yassen, E. Berkowicz, A. Glozman, and P. Klipstein, "InAsSb-based XB n n bariodes grown by molecular beam epitaxy on GaAs," *Journal of Crystal Growth*, vol. 339, pp. 31-35, 2012.
- [117] D. Z.-Y. Ting, C. J. Hill, A. Soibel, S. A. Keo, J. M. Mumolo, J. Nguyen, and S. D. Gunapala, "A high-performance long wavelength superlattice complementary barrier infrared detector," *Applied Physics Letters*, vol. 95, p. 023508, 2009.
- [118] G. R. Savich, J. R. Pedrazzani, D. E. Sidor, S. Maimon, and G. W. Wicks, "Dark current filtering in unipolar barrier infrared detectors," *Applied Physics Letters*, vol. 99, 2011.
- [119] K. Ueno, E. G. Camargo, T. Katsumata, H. Goto, N. Kuze, Y. Kangawa, and K. Kakimoto, "InSb Mid-Infrared Photon Detector for Room-Temperature Operation," *Japanese Journal of Applied Physics*, vol. 52, 2013.
- [120] A. Rogalski and P. Martyniuk, "Mid-wavelength infrared nBn for HOT detectors," *Journal of Electronic Materials*, vol. 43, pp. 2963-2969, 2014.
- [121] S. Datta, T. Ashley, J. Brask, L. Buckle, M. Doczy, M. Emeny, D. Hayes, K. Hilton, R. Jefferies, T. Martin, T. J. Phillips, D. Wallis, P. Wilding, and R. Chau, "85nm gate length enhancement and depletion mode InSb quantum well transistors for ultra high speed and very low power digital logic applications," *Ieee International Electron Devices Meeting 2005, Technical Digest*, pp. 783-786, 2005.
- [122] J. R. Soderstrom, M. M. Cumming, J. Y. Yao, and T. G. Andersson, "Molecular-Beam Epitaxy Growth and Characterization of Insb Layers on Gaas Substrates," *Semiconductor Science and Technology*, vol. 7, pp. 337-343, 1992.
- [123] V. P. LaBella, M. R. Krause, Z. Ding, and P. M. Thibado, "Arsenic-rich GaAs(001) surface structure," *Surface Science Reports*, vol. 60, pp. 1-53, 2005.
- [124] T. A. Lafford, B. K. Tanner, and P. J. Parbrook, "Direct measurement of twist mosaic in GaN epitaxial films as a function of growth temperature," *Journal of Physics D-Applied Physics*, vol. 36, pp. A245-A248, 2003.
- [125] S. K. Ghandhi and J. E. Ayers, "Strain and Misorientation in Gaas Grown on Si(001) by Organometallic Epitaxy," *Applied Physics Letters*, vol. 53, pp. 1204-1206, 1988.
- [126] B. Yarlagadda, A. Rodriguez, P. Li, R. Velampati, J. F. Ocampo, E. N. Suarez, P. B. Rago, D. Shah, J. E. Ayers, and F. C. Jain, "X-ray characterization of dislocation density asymmetries in heteroepitaxial semiconductors," *Applied Physics Letters*, vol. 92, 2008.
- [127] N. E. Lee, D. G. Cahill, and J. E. Greene, "Evolution of surface roughness in

- epitaxial Si_{0.7}Ge_{0.3}(001) as a function of growth temperature (200-600 degrees C) and Si(001) substrate miscut," *Journal of Applied Physics*, vol. 80, pp. 2199-2210, 1996.
- [128] H. M. Wang, T. W. Fan, J. Wu, Y. P. Zeng, J. R. Dong, and M. Y. Kong, "Effects of growth temperature on highly mismatched InAs grown on GaAs substrates by MBE," *Journal of Crystal Growth*, vol. 186, pp. 38-42, 1998.
- [129] V. Lebedev, V. Cimalla, J. Pezoldt, M. Himmerlich, S. Krischok, J. A. Schaefer, O. Ambacher, F. M. Morales, J. G. Lozano, and D. Gonzalez, "Effect of dislocations on electrical and electron transport properties of InN thin films. I. Strain relief and formation of a dislocation network," *Journal of Applied Physics*, vol. 100, 2006.
- [130] H. Suzuki, D. Ito, A. Fukuyama, and T. Ikari, "Reduction of rotational twin formation by indium pre-evaporation in epitaxially grown GaAs films on Si (111) substrate," *Journal of Crystal Growth*, vol. 380, pp. 148-152, 2013.
- [131] T. Wang and A. Forchel, "Growth of self-organized GaSb islands on a GaAs surface by molecular beam epitaxy," *Journal of Applied Physics*, vol. 85, pp. 2591-2594, 1999.
- [132] M. Losurdo, P. Capezzuto, G. Bruno, A. S. Brown, T. Brown, and G. May, "Fundamental reactions controlling anion exchange during mixed anion heterojunction formation: Chemistry of As-for-Sb and Sb-for-As exchange reactions," *Journal of Applied Physics*, vol. 100, 2006.
- [133] Y. Q. Wang, Z. L. Wang, T. Brown, A. Brown, and G. May, "Thermodynamic analysis of anion exchange during heteroepitaxy," *Journal of Crystal Growth*, vol. 242, pp. 5-14, 2002.
- [134] T. Sato, T. Suzuki, S. Tomiya, and S. Yamada, "Dislocation-limited electron transport in InSb grown on GaAs(001)," *Physica B-Condensed Matter*, vol. 376, pp. 579-582, 2006.
- [135] G. Singh, E. Michel, C. Jelen, S. Slivken, J. Xu, P. Bove, I. Ferguson, and M. Razeghi, "Molecular-Beam Epitaxial-Growth of High-Quality InSb for P-I-N Photodetectors," *Journal of Vacuum Science & Technology B*, vol. 13, pp. 782-785, 1995.
- [136] D. L. Rode, "Electron Transport in InSb, InAs and InP," *Physical Review B*, vol. 3, pp. 3287-&, 1971.
- [137] Y. M. Galperin, E. M. Gershenzon, I. L. Drichko, and L. B. Litvakgorskaya, "Low-Temperature Transport Phenomena in Compensated N-Type InSb (Review)," *Soviet Physics Semiconductors-Ussr*, vol. 24, pp. 1-15, 1990.
- [138] L. M. Falicov and M. Cuevas, "Mobility of Electrons in Compensated Semiconductors .2. Theory," *Physical Review*, vol. 164, pp. 1025-&, 1967.
- [139] D. Chattopadhyay and H. J. Queisser, "Electron-Scattering by Ionized Impurities in Semiconductors," *Reviews of Modern Physics*, vol. 53, pp. 745-768, 1981.
- [140] A. Okamoto and I. Shibusaki, "Transport properties of Sn-doped InSb thin films and applications to Hall element," *Journal of Crystal Growth*, vol. 251, pp. 560-564, 2003.
- [141] C. Besikci, Y. H. Choi, R. Sudharsanan, and M. Razeghi, "Anomalous Hall-
-

- Effect in InSb Layers Grown by Metalorganic Chemical-Vapor Deposition on GaAs Substrates," *Journal of Applied Physics*, vol. 73, pp. 5009-5013, 1993.
- [142] M. Oszwaldowski and M. Zimpel, "Temperature-Dependence of Intrinsic Carrier Concentration and Density of States Effective Mass of Heavy Holes in InSb," *Journal of Physics and Chemistry of Solids*, vol. 49, pp. 1179-1185, 1988.
- [143] T. D. Mishima, M. Edirisooriya, and M. B. Santos, "Reduction of microtwin defects for high-electron-mobility InSb quantum wells," *Applied Physics Letters*, vol. 91, 2007.
- [144] H. T. Pham, S. F. Yoon, K. P. Chen, and D. Boning, "Characterization of carbon-doped InSb diode grown by molecular beam epitaxy," *Journal of Physics D-Applied Physics*, vol. 41, 2008.
- [145] X. Liu, A. Prasad, J. Nishio, E. R. Weber, Z. Lilientalweber, and W. Walukiewicz, "Native Point-Defects in Low-Temperature-Grown GaAs," *Applied Physics Letters*, vol. 67, pp. 279-281, 1995.
- [146] A. Hoglund, C. W. M. Castleton, M. Gothelid, B. Johansson, and S. Mirbt, "Point defects on the (110) surfaces of InP, InAs, and InSb: A comparison with bulk," *Physical Review B*, vol. 74, 2006.
- [147] S. Froyen and A. Zunger, "Surface segregation and ordering in III-V semiconductor alloys," *Physical Review B*, vol. 53, pp. 4570-4579, 1996.
- [148] A. J. Martin, T. W. Saucer, K. Sun, S. J. Kim, G. Ran, G. V. Rodriguez, X. Q. Pan, V. Sih, and J. Millunchick, "Analysis of defect-free GaSb/GaAs(001) quantum dots grown on the Sb-terminated (2x8) surface," *Journal of Vacuum Science & Technology B*, vol. 30, 2012.
- [149] F. Maeda, Y. Watanabe, and M. Oshima, "Sb-Induced Surface Reconstruction on GaAs(001)," *Physical Review B*, vol. 48, pp. 14733-14736, 1993.
- [150] P. Moriarty, P. H. Beton, Y. R. Ma, and M. Henini, "Adsorbed and substituted Sb dimers on GaAs(001)," *Physical Review B*, vol. 53, pp. 16148-16151, 1996.
- [151] J. E. Boschker, J. Momand, V. Bragaglia, R. N. Wang, K. Perumal, A. Giussani, B. J. Kooi, H. Riechert, and R. Calarco, "Surface Reconstruction-Induced Coincidence Lattice Formation Between Two-Dimensionally Bonded Materials and a Three-Dimensionally Bonded Substrate," *Nano Letters*, vol. 14, pp. 3534-3538, 2014.
- [152] Y. Sun, B. Zheng, X. Wu, L. Yuan, J. Wu, H. Guo, K. Huang, and S. Feng, "Surface reconstruction: An effective method for the growth of mismatched materials," *Applied Surface Science*, vol. 360, pp. 547-552, 2016.
- [153] C. J. Reyner, J. Wang, K. Nunna, A. Lin, B. L. Liang, M. S. Goorsky, and D. L. Huffaker, "Characterization of GaSb/GaAs interfacial misfit arrays using x-ray diffraction," *Applied Physics Letters*, vol. 99, 2011.
- [154] B. W. Jia, K. H. Tan, W. K. Loke, S. Wicaksono, and S. F. Yoon, "Epitaxial growth of low threading dislocation density InSb on GaAs using self-assembled periodic interfacial misfit dislocations," *Materials Letters*, vol. 158, pp. 258-261, 2015.
- [155] P. Laukkanen, R. E. Perala, R. L. Vaara, I. J. Vayrynen, M. Kuzmin, and J. Sadowski, "Electronic and structural analysis of Sb-induced GaAs(100)(2x4)
-

- and (2x8) surfaces," *Physical Review B*, vol. 69, 2004.
- [156] J. E. Bickel, N. A. Modine, and J. M. Millunchick, "Determining the GaSb/GaAs-(2x8) reconstruction," *Surface Science*, vol. 603, pp. 2945-2949, 2009.
- [157] L. J. Whitman, B. R. Bennett, E. M. Kneedler, B. T. Jonker, and B. V. Shanabrook, "The structure of Sb-terminated GaAs(001) surfaces," *Surface Science*, vol. 436, pp. L707-L714, 1999.
- [158] M. Ichimura, K. Higuchi, Y. Hattori, T. Wada, and N. Kitamura, "Native Defects in the $\text{Al}_x\text{Ga}_{1-x}\text{Sb}$ Alloy Semiconductor," *Journal of Applied Physics*, vol. 68, pp. 6153-6158, 1990.
- [159] Y. B. Li, Y. Zhang, Y. W. Zhang, B. Q. Wang, Z. P. Zhu, and Y. P. Zeng, "Molecular beam epitaxial growth and characterization of GaSb layers on GaAs (001) substrates," *Applied Surface Science*, vol. 258, pp. 6571-6575, 2012.
- [160] P. S. Dutta, V. Prasad, H. L. Bhat, and V. Kumar, "Carrier compensation and scattering mechanisms in p-GaSb," *Journal of Applied Physics*, vol. 80, pp. 2847-2853, 1996.
- [161] H. Matsuura, K. Morita, K. Nishikawa, T. Mizukoshi, M. Segawa, and W. Susaki, "Acceptor densities and acceptor levels in undoped GaSb determined by free carrier concentration spectroscopy," *Japanese Journal of Applied Physics Part 1-Regular Papers Short Notes & Review Papers*, vol. 41, pp. 496-500, 2002.
- [162] W. Zhou, X. Li, S. J. Xia, J. Yang, W. Tang, and K. M. Lau, "High Hole Mobility of GaSb Relaxed Epilayer Grown on GaAs Substrate by MOCVD through Interfacial Misfit Dislocations Array," *Journal of Materials Science & Technology*, vol. 28, pp. 132-136, 2012.
- [163] B. W. Jia, K. H. Tan, W. K. Loke, S. Wicaksono, and S. F. Yoon, "Formation of periodic interfacial misfit dislocation array at the InSb/GaAs interface via surface anion exchange," *Journal of Applied Physics*, vol. 120, p. 035301, 2016.
- [164] Y. A. Osipyan and Y. I. Fedyaev, "Donor Action of Dislocations in InSb," *Jetp Letters-Ussr*, vol. 9, pp. 21-&, 1969.
- [165] J. J. Harris, T. Zhang, W. R. Branford, S. K. Clowes, M. Debnath, A. Bennett, C. Roberts, and L. F. Cohen, "The role of impurity band conduction in the low temperature characteristics of thin InSb films grown by molecular beam epitaxy," *Semiconductor Science and Technology*, vol. 19, pp. 1406-1410, 2004.
- [166] F. Stern and R. M. Talley, "Impurity Band in Semiconductors with Small Effective Mass," *Physical Review*, vol. 100, pp. 1638-1643, 1955.
- [167] E. Conwell and V. F. Weisskopf, "Theory of Impurity Scattering in Semiconductors," *Physical Review*, vol. 77, pp. 388-390, 1950.
- [168] M. Oszwaldowski, "Scattering of Electrons by Charged Dislocations in InSb-Type Semiconductors," *Journal of Physics and Chemistry of Solids*, vol. 46, pp. 791-796, 1985.
- [169] F. Seitz and D. L. Dexter, "Effects of Dislocations on Mobilities in Semiconductors," *Physical Review*, vol. 87, pp. 189-190, 1952.
- [170] K.-M. Ko, J.-H. Seo, D.-E. Kim, S.-T. Lee, Y.-K. Noh, M.-D. Kim, and J.-E. Oh,

- "The growth of a low defect InAs HEMT structure on Si by using an AlGaSb buffer layer containing InSb quantum dots for dislocation termination," *Nanotechnology*, vol. 20, p. 225201, 2009.
- [171] H.-C. Luan, D. R. Lim, K. K. Lee, K. M. Chen, J. G. Sandland, K. Wada, and L. C. Kimerling, "High-quality Ge epilayers on Si with low threading-dislocation densities," *Applied physics letters*, vol. 75, pp. 2909-2911, 1999.
- [172] L. Colace, G. Masini, G. Assanto, H. C. Luan, K. Wada, and L. Kimerling, "Efficient high-speed near-infrared Ge photodetectors integrated on Si substrates," *Applied Physics Letters*, vol. 76, pp. 1231-1233, 2000.
- [173] J. W. Lee, H. Shichijo, H. L. Tsai, and R. J. Matyi, "Defect Reduction by Thermal Annealing of GaAs-Layers Grown by Molecular-Beam Epitaxy on Si Substrates," *Applied Physics Letters*, vol. 50, pp. 31-33, 1987.
- [174] T. Sasaki and N. Oda, "Dislocation Reduction in HgCdTe on GaAs by Thermal Annealing," *Journal of Applied Physics*, vol. 78, pp. 3121-3124, 1995.
- [175] N. Hayafuji, T. Kimura, N. Yoshida, N. Kaneno, M. Tsugami, K. Mizuguchi, T. Murotani, and S. Ibuki, "Improvement of Inp Crystal Quality on GaAs Substrates by Thermal Cyclic Annealing," *Japanese Journal of Applied Physics Part 2-Letters*, vol. 28, pp. L1721-L1724, 1989.
- [176] M. Aziz, A. Mesli, J. F. Felix, D. Jameel, N. Al Saqri, D. Taylor, and M. Henini, "Effect of post-growth annealing treatment on interfacial misfit GaSb/GaAs heterostructures," *Journal of Crystal Growth*, vol. 424, pp. 5-10, 2015.
- [177] B. Heying, X. Wu, S. Keller, Y. Li, D. Kapolnek, B. Keller, S. P. DenBaars, and J. Speck, "Role of threading dislocation structure on the x-ray diffraction peak widths in epitaxial GaN films," *Applied Physics Letters*, vol. 68, pp. 643-645, 1996.
- [178] H. J. Gao and W. D. Nix, "Surface roughening of heteroepitaxial thin films," *Annual Review of Materials Science*, vol. 29, pp. 173-209, 1999.
- [179] H. L. Tsai and J. W. Lee, "Defect Structures at the GaAs/Si Interface after Annealing," *Applied Physics Letters*, vol. 51, pp. 130-132, 1987.
- [180] D. Lee, M. Park, Z. Tang, H. Luo, R. Beresford, and C. Wie, "Characterization of metamorphic $\text{In}_x\text{Al}_{1-x}\text{As}$ /GaAs buffer layers using reciprocal space mapping," *Journal of applied physics*, vol. 101, p. 063523, 2007.
- [181] M. De Giorgi, A. Taurino, A. Passaseo, M. Catalano, and R. Cingolani, "Interpretation of phase and strain contrast of TEM images of $\text{In}_x\text{Ga}_{1-x}\text{As}$ /GaAs quantum dots," *Physical Review B*, vol. 63, p. 245302, 2001.
- [182] K. H. Tan, B. W. Jia, W. K. Loke, S. Wicaksono, and S. F. Yoon, "Formation of interfacial misfit dislocation in GaSb/GaAs heteroepitaxy via anion exchange process," *Journal of Crystal Growth*, vol. 427, pp. 80-86, 2015.
- [183] L. Dong, J. Schnitker, R. W. Smith, and D. J. Srolovitz, "Stress relaxation and misfit dislocation nucleation in the growth of misfitting films: A molecular dynamics simulation study," *Journal of applied physics*, vol. 83, pp. 217-227, 1998.
- [184] N. G. Weimann, L. F. Eastman, D. Doppalapudi, H. M. Ng, and T. D. Moustakas, "Scattering of electrons at threading dislocations in GaN," *Journal of Applied*
-

- Physics*, vol. 83, pp. 3656-3659, 1998.
- [185] W. Hu, Z. Wang, B. Su, Y. Dai, S. Wang, and Y. Zhao, "Gallium antisite defect and residual acceptors in undoped GaSb," *Physics Letters A*, vol. 332, pp. 286-290, 2004.
- [186] H. Bracht, S. Nicols, W. Walukiewicz, and J. P. Silveira, "Large disparity between gallium and antimony self-diffusion in gallium antimonide," *Nature*, vol. 408, p. 69, 2000.
- [187] M. Debnath, T. Zhang, C. Roberts, L. Cohen, and R. Stradling, "High-mobility InSb thin films on GaAs (001) substrate grown by the two-step growth process," *Journal of crystal growth*, vol. 267, pp. 17-21, 2004.
- [188] H. C. Luan, D. R. Lim, K. K. Lee, K. M. Chen, J. G. Sandland, K. Wada, and L. C. Kimerling, "High-quality Ge epilayers on Si with low threading-dislocation densities," *Applied Physics Letters*, vol. 75, pp. 2909-2911, 1999.
- [189] D. Redfield, "Effect of defect fields on the optical absorption edge," *Physical Review*, vol. 130, p. 916, 1963.
- [190] Z.-M. Liao, Y. Lu, J. Xu, J.-M. Zhang, and D.-P. Yu, "Temperature dependence of photoconductivity and persistent photoconductivity of single ZnO nanowires," *Applied Physics A: Materials Science & Processing*, vol. 95, pp. 363-366, 2009.
- [191] S. Yadav, K.-H. Tan, K. H. Goh, S. Subramanian, K. L. Low, N. Chen, B. Jia, S.-F. Yoon, G. Liang, and X. Gong, "First monolithic integration of Ge P-FETs and InAs N-FETs on silicon substrate: Sub-120 nm III-V buffer, sub-5 nm ultra-thin body, common raised S/D, and gate stack modules," in *Electron Devices Meeting (IEDM), 2015 IEEE International*, 2015, pp. 2.3. 1-2.3. 4.
- [192] G. Chen, W. Sun, and Y. Lv, "Empirical expression for the composition and temperature dependence of the energy gap in InAlSb," *Infrared Physics & Technology*, vol. 81, pp. 262-265, 2017.
- [193] K. H. Lee, S. Bao, B. Wang, C. Wang, S. F. Yoon, J. Michel, E. A. Fitzgerald, and C. S. Tan, "Reduction of threading dislocation density in Ge/Si using a heavily As-doped Ge seed layer," *AIP Advances*, vol. 6, p. 025028, 2016.
- [194] D. Bolef and M. Menes, "Elastic Constants of Single-Crystal Aluminum Antimonide," *Journal of Applied Physics*, vol. 31, pp. 1426-1427, 1960.
- [195] K. Ozasa, M. Yuri, S. Tanaka, and H. Matsunami, "Effect of misfit strain on physical properties of InGaP grown by metalorganic molecular-beam epitaxy," *Journal of applied physics*, vol. 68, pp. 107-111, 1990.
- [196] G. Lee, P. Thompson, J. Davis, J. Omaggio, and W. Schmidt, "Characterization of molecular beam epitaxially grown InSb layers and diode structures," *Solid-state electronics*, vol. 36, pp. 387-389, 1993.
- [197] S. M. Johnson, D. Rhiger, J. Rosbeck, J. Peterson, S. Taylor, and M. Boyd, "Effect of dislocations on the electrical and optical properties of long-wavelength infrared HgCdTe photovoltaic detectors," *Journal of Vacuum Science & Technology B: Microelectronics and Nanometer Structures Processing, Measurement, and Phenomena*, vol. 10, pp. 1499-1506, 1992.
- [198] C. E. Jones, V. Nair, and D. L. Polla, "Generation-recombination centers in p-

- type $\text{Hg}_{1-x}\text{Cd}_x\text{Te}$," *Applied Physics Letters*, vol. 39, pp. 248-250, 1981.
- [199] M. Erkus, O. Senel, and U. Serincan, "Structural, optical and electrical characterization of $\text{InAs}_{0.83}\text{Sb}_{0.17}$ p- π -n photodetector grown on GaAs substrate," *Thin Solid Films*, vol. 616, pp. 141-144, 2016.
- [200] C. Elliott, I. Melngailis, T. Harman, and A. Foyt, "Carrier freeze-out and acceptor energies in p-type $\text{Hg}_{1-x}\text{Cd}_x\text{Te}$," *Journal of Physics and Chemistry of Solids*, vol. 33, pp. 1527IN3-1531, 1972.
- [201] W. Hu, X. Chen, F. Yin, Z. Quan, Z. Ye, X. Hu, Z. Li, and W. Lu, "Analysis of temperature dependence of dark current mechanisms for long-wavelength HgCdTe photovoltaic infrared detectors," *Journal of Applied Physics*, vol. 105, p. 104502, 2009.
- [202] J. Huang, W. Ma, Y. Zhang, Y. Cao, K. Liu, W. Huang, and S. Lu, "Narrow-Band Type II Superlattice Photodetector With Detection Wavelength Shorter Than $2\sim\mu\text{m}$," *IEEE Photonics Technology Letters*, vol. 27, pp. 2276-2279, 2015.
- [203] A. Foyt, W. Lindley, and J. Donnelly, "n-p Junction photodetectors in InSb fabricated by proton bombardment," *Applied Physics Letters*, vol. 16, pp. 335-337, 1970.
- [204] A. Evirgen, J. Abautret, J. Perez, H. Aït-Kaci, P. Christol, J. Fleury, H. Sik, A. Nedelcu, R. Cluzel, and A. Cordat, "InSb photodetectors with PIN and nBn designs," in *SPIE OPTO*, 2013, pp. 899313-899313-11.
- [205] G. Balakrishnan, S. Huang, L. Dawson, Y. C. Xin, P. Conlin, and D. Huffaker, "Growth mechanisms of highly mismatched AlSb on a Si substrate," *Applied Physics Letters*, vol. 86, p. 034105, 2005.
- [206] I. Németh, B. Kunert, W. Stolz, and K. Volz, "Heteroepitaxy of GaP on Si: Correlation of morphology, anti-phase-domain structure and MOVPE growth conditions," *Journal of Crystal Growth*, vol. 310, pp. 1595-1601, 2008.
- [207] J. Tatebayashi, A. Jallipalli, M. N. Kuttu, S. Huang, K. Nunna, G. Balakrishnan, L. R. Dawson, and D. L. Huffaker, "Monolithically integrated III-Sb-based laser diodes grown on miscut Si substrates," *IEEE Journal of Selected Topics in Quantum Electronics*, vol. 15, pp. 716-723, 2009.
- [208] L. A. Giannuzzi and F. A. Stevie, "A review of focused ion beam milling techniques for TEM specimen preparation," *Micron*, vol. 30, pp. 197-204, 1999.
- [209] S. H. Vajargah, S. Ghanad-Tavakoli, J. Preston, R. Kleiman, and G. Botton, "Growth mechanisms of GaSb heteroepitaxial films on Si with an AlSb buffer layer," *Journal of Applied Physics*, vol. 114, p. 113101, 2013.
- [210] D. B. Williams and C. B. Carter, "The transmission electron microscope," in *Transmission electron microscopy*, ed: Springer, 1996, pp. 3-17.
- [211] C. Snyder, B. Orr, D. Kessler, and L. Sander, "Effect of strain on surface morphology in highly strained InGaAs films," *Physical Review Letters*, vol. 66, p. 3032, 1991.
- [212] T. Mishima, J. Keay, N. Goel, M. Ball, S. Chung, M. Johnson, and M. Santos, "Effect of micro-twin defects on InSb quantum wells," *Journal of Vacuum Science & Technology B: Microelectronics and Nanometer Structures*
-

- Processing, Measurement, and Phenomena*, vol. 23, pp. 1171-1173, 2005.
- [213] M. Copel, M. Reuter, E. Kaxiras, and R. Tromp, "Surfactants in epitaxial growth," *Physical Review Letters*, vol. 63, p. 632, 1989.
- [214] W. Liu, W. Zheng, and Q. Jiang, "First-principles study of the surface energy and work function of III-V semiconductor compounds," *Physical Review B*, vol. 75, p. 235322, 2007.
- [215] J. Bomphrey, M. Ashwin, and T. Jones, "The formation of high number density InSb quantum dots, resulting from direct InSb/GaSb (001) heteroepitaxy," *Journal of Crystal Growth*, vol. 420, pp. 1-5, 2015.
- [216] Y. Chen, X. Lin, Z. Liliental-Weber, J. Washburn, J. Klem, and J. Tsao, "Dislocation formation mechanism in strained $\text{In}_x\text{Ga}_{1-x}\text{As}$ islands grown on GaAs (001) substrates," *Applied physics letters*, vol. 68, pp. 111-113, 1996.
- [217] Y. Chen and J. Washburn, "Structural transition in large-lattice-mismatch heteroepitaxy," *Physical review letters*, vol. 77, p. 4046, 1996.
- [218] B. W. Jia, K. H. Tan, W. K. Loke, S. Wicaksono, and S. F. Yoon, "Effects of surface reconstruction on the epitaxial growth of III-Sb on GaAs using interfacial misfit array," *Applied Surface Science*, vol. 399, pp. 220-228, 2017.
- [219] J. Harris, T. Zhang, W. Branford, S. Clowes, M. Debnath, A. Bennett, C. Roberts, and L. Cohen, "The role of impurity band conduction in the low temperature characteristics of thin InSb films grown by molecular beam epitaxy," *Semiconductor science and technology*, vol. 19, p. 1406, 2004.
- [220] S. Zollner, S. Gopalan, and M. Cardona, "The temperature dependence of the band gaps in InP, InAs, InSb, and GaSb," *Solid state communications*, vol. 77, pp. 485-488, 1991.
- [221] E. Putley, "Electrical conduction in n-type InSb between 2 K and 300 K," *Proceedings of the Physical Society*, vol. 73, p. 280, 1959.
- [222] M. Zandian, A. Chen, D. Edwall, J. Pasko, and J. Arias, "p-type arsenic doping of $\text{Hg}_{1-x}\text{Cd}_x\text{Te}$ by molecular beam epitaxy," *Applied physics letters*, vol. 71, pp. 2815-2817, 1997.
- [223] M. Razeghi, "Overview of antimonide based III-V semiconductor epitaxial layers and their applications at the center for quantum devices," *The European Physical Journal-Applied Physics*, vol. 23, pp. 149-205, 2003.
- [224] J. Kim, D. Wu, J. Wojkowski, J. Piotrowski, J. Xu, and M. Razeghi, "Long-wavelength InAsSb photoconductors operated at near room temperatures (200–300 K)," *Applied physics letters*, vol. 68, pp. 99-101, 1996.
- [225] R. Laff and H. Fan, "Carrier lifetime in indium antimonide," *Physical Review*, vol. 121, p. 53, 1961.
- [226] S. Shin, J. Arias, M. Zandian, J. Pasko, and R. DeWames, "Effect of the dislocation density on minority-carrier lifetime in molecular beam epitaxial HgCdTe ," *Applied physics letters*, vol. 59, pp. 2718-2720, 1991.
- [227] M. Gutiérrez, D. Araujo, P. Jurczak, J. Wu, and H. Liu, "Solid solution strengthening in GaSb/GaAs: A mode to reduce the TD density through Be-doping," *Applied Physics Letters*, vol. 110, p. 092103, 2017.