ANTIMONIDE-BASED FIELD-EFFECT TRANSISTORS AND HETEROJUNCTION BipOLAR TRANSISTORS GROWN BY MOLECULAR BEAM EPITAXY

BY

CHI-CHIH LIAO

DISSERTATION

Submitted in partial fulfillment of the requirements for the degree of Doctor of Philosophy in Electrical and Computer Engineering in the Graduate College of the University of Illinois at Urbana-Champaign, 2011

Urbana, Illinois

Doctoral Committee:

Adjunct Professor Keh-Yung Cheng, Chair
Adjunct Professor Kuang-Chien Hsieh
Professor Milton Feng
Professor Jianming Jin
ABSTRACT

For the development of novel high-speed devices, the epitaxial growth of antimonide-based compounds and devices, including field effect transistors (FETs) and hetero-junction bipolar transistors (HBTs), was explored using gas-source molecular beam epitaxy (MBE). The first and second parts of the dissertation detail the growth of InAsSb and InGaSb as the channel materials for n- and p-type FETs, respectively. Both compounds were grown metamorphically on InP substrates with a composite AlSb/AlAs$_{0.5}$Sb$_{0.5}$ buffer layer, which was proved to be effective in enhancing the epitaxial quality. By optimizing the growth conditions, the intrinsic carrier mobilities of n-type InAsSb and p-type pseudomorphic InGaSb quantum wells could reach 18000 and 600 cm$^2$/V-s at room temperature, respectively. InAsSb FET showed a high transconductance of 350 mS/mm, which indicated the high potential in the high-speed applications. The third part of the dissertation describes the modification of the emitter-base junction of ultra-fast type-II GaAsSb-based HBTs in order to eliminate the carrier blocking and enhance the current gain. InAlP was used to replace the InP emitter and form a type-I emitter-base junction. Results for large devices show that this modification could improve DC current gain from 80 to 120. The results indicate that type-I/II InAlP/GaAsSb HBTs are promising to achieve better radio-frequency (RF) performance with higher current driving capability.
ACKNOWLEDGMENTS

I would like to thank my adviser with all my heart, Professor Keh-Yung Cheng, for his guidance, not only in the academic research, but also in other aspects of personal life. Also, I appreciate the assistance and experience sharing from group members, including Dr. Bing-Ruey Wu, Dr. Chiun-Lung Tsai, Dr. Kevin Meneou, and Chien-Chia Cheng. It would have been impossible to finish my research without their kind help. Finally, I would like to thank my parents, elder brother and sister, and my girlfriend for their unwavering support.

The research presented in this thesis is partly supported by the DARPA University Photonic Research Center for Hyper-Uniform Nanophotonic Technologies Program (Grant No. HR0011-04-1-0034, HUNT Center) and the MARCO MSD Center.
# TABLE OF CONTENTS

1. Introduction .............................................................................................................. 1  
   1.1 Motivation ........................................................................................................... 1  
   1.2 InAs$_x$Sb$_{1-x}$ Compound for N-Type FET ................................................... 3  
   1.3 In$_x$Ga$_{1-x}$Sb Compound for P-Type FET ....................................................... 7  
   1.4 Current Gain Improvement of GaAsSb HBT ..................................................... 10  

2. Growth of InAs$_{0.8}$Sb$_{0.2}$ Compound for N-Type FET ................................... 13  
   2.1 GSMBE System ................................................................................................ 13  
   2.2 Sample Growth .................................................................................................. 16  
   2.3 Results and Discussions ..................................................................................... 19  

3. Growth of In$_x$Ga$_{1-x}$Sb Compound for P-Type FET ..................................... 33  
   3.1 Valence Bands and Effective Mass of Strained In$_x$Ga$_{1-x}$Sb ....................... 33  
   3.2 Growth of Strained In$_x$Ga$_{1-x}$Sb Quantum Well ........................................ 36  
   3.3 Results and Discussions ..................................................................................... 38  

4. Type-I/II DHBT With High Current Gain ............................................................... 46  
   4.1 Introduction ........................................................................................................ 46  
   4.2 Type-I/II InAlP/GaAsSb DHBT ........................................................................ 50  

5. Conclusions and Future Works ................................................................................. 58  
   5.1 Summary of Works ............................................................................................ 58  
   5.2 Future Works ...................................................................................................... 59  

References .................................................................................................................. 62
Chapter 1 INTRODUCTION

1.1 Motivation

The invention of metal-oxide-semiconductor field-effect transistor (MOSFET) technologies enabled the development of integrated circuits and the semiconductor industry. In order to achieve higher operation speed and circuit density, scaling MOSFETs to smaller dimensions is the most straightforward approach and requires no excessive design changes. In the past decades, the scalability of MOSFETs has shrunk the minimum feature size from 8 µm to 22 nm, which requires a physical gate length smaller than 9 nm, and improves the device performance dramatically [1]. However, it is foreseeable that there will be an end to the downscaling because of the limitation of material properties. For example, as the gate dielectric thickness decreases with the shrinking device dimensions, gate leakage will increase because of tunneling [2]. Also, as the gate length continues to decrease, source/drain and band-to-band tunneling become larger and increase the power dissipation [2]. Moreover, in nanoscale transistors, the carrier mobility will degrade due to the large vertical electric fields [3]. Figure 1.1 shows the electron mobility of silicon versus the effective electric field in the channel, which increases as the technology node advances. The electron mobility has decreased from 400 to 120 cm^2/V-s with the device dimension scaling down from 0.8 µm to 0.13 µm. Therefore, in order to sustain the performance improvement beyond the 22-nm technology node, it is probably necessary to introduce new materials or modify the device structure.
Among several emerging nano-devices, including carbon nanotube (CNT) and graphene, III-V compound semiconductors and devices based on these materials are most attractive. First, these materials can be fabricated into devices using “top-down” lithographic and etching methods, which are well established and have great potential to be integrated into current Si technology. Second, and most important of all, some III-V compound semiconductors possess much higher carrier mobility than silicon/germanium, especially electron mobility. Figure 1.2 compares electron and hole mobility of elemental and III-V semiconductors [4]. Apparently, compound semiconductors have a great advantage over silicon/germanium in electron mobility. Among all, InAs and InSb have the highest intrinsic electron mobility. At room temperature, the electron mobility of InAs can reach 30,000 cm$^2$/V-s [5], while that of InSb is as high as 78,000 cm$^2$/V-s [6] at a carrier concentration of ~ $10^{16}$ cm$^{-3}$. InAs and InSb are atop all the other...
compound semiconductors in electron mobility, and their mobilities are more than ten times higher than that of silicon/germanium. Therefore, the InAs$_x$Sb$_{1-x}$ ternary alloy, which combines compounds with highest electron motilities, could be adaptive to structure design and epitaxial growth and is an excellent candidate for high-speed electronic devices.

![Fig. 1.2 Room-temperature carrier mobility versus minimum energy band gap for elemental and III-V compound semiconductors (modified after [4]).](image)

Fig. 1.2 Room-temperature carrier mobility versus minimum energy band gap for elemental and III-V compound semiconductors (modified after [4]).

### 1.2 InAs$_x$Sb$_{1-x}$ Compound for N-Type FET

Figure 1.3 shows the relation between electron mobility and the composition of InAs$_x$Sb$_{1-x}$ compound, both theoretically and experimentally [7,8]. From theoretical calculations, there is a peak of electron mobility around InAs$_{0.2}$Sb$_{0.8}$, and no alloy scattering, which would drop the mobility below either InAs or InSb, is observed. Although InAs$_x$Sb$_{1-x}$ compounds have high
electron mobility and could be a good candidate for high-speed electronic devices, the utilization of these materials is limited by some problems. The biggest concern for the growth of these materials is the lack of lattice-matched and semi-insulating substrates with high quality and attainability. Although GaSb and InAs substrates are commercially available and have little lattice mismatch, their high electrical conductivity severely limits the usage of these substrates. Therefore, it is common to grow the compounds of a lattice constant around or beyond 6.1 Å metamorphically on GaAs or Si substrates. For example, InAs has been successfully grown on
GaAs substrates, with a thick buffer layer of more than 2 µm to overcome the large lattice mismatch between GaAs and InAs. With this method, modulation-doped InAs high electron mobility transistor (HEMT) structures have been grown and InAs HEMTs have been demonstrated [9]. However, the required thick buffer layers need an extended growth time, and their functions of reducing defect density vary for different growth techniques.

Compared with GaAs, InP has a 50% smaller lattice mismatch with 6.1 Å materials and is expected to be capable of achieving better results with a thinner buffer layer. Second, the InP has a higher thermal conductivity than GaAs, which provides more protection against overheating to the Sb-compound layers. Although InP has these advantages over GaAs, very few studies about growing 6.1 Å materials on InP have been reported so far. Therefore, it is worthwhile to investigate the growth of high quality InAs\(_{x}\)Sb\(_{1-x}\) layers on InP substrates, using relatively thin buffer layers.

The compound to start with is InAs\(_{0.8}\)Sb\(_{0.2}\), due to the following reasons. First, as shown in Figure 1.4, InAs\(_{0.8}\)Sb\(_{0.2}\), unlike InAs, is lattice-matched to AlSb. A lattice-matched channel/buffer layer epitaxy could be fabricated into the surface-channel FET. Second, with the addition of a small amount of antimony, InAs\(_{0.8}\)Sb\(_{0.2}\) might have higher electron mobility than InAs, although alloy scattering could limit the mobility enhancement. Third, it is easy to change the composition of ternary InAs\(_{x}\)Sb\(_{1-x}\) to introduce tensile strain, which is proved to be effective in increasing the electron mobility [3]. Nevertheless, the electron mobility of bulk InAs\(_{0.8}\)Sb\(_{0.2}\) has been rarely reported. The typical electron mobility at room temperature of bulk InAs grown on GaAs by molecular beam epitaxy (MBE) is about 16,000 cm\(^2\)/V-s with a donor concentration of 9 x 10\(^{16}\) cm\(^{-3}\) [10]. For the InAs\(_{0.8}\)Sb\(_{0.2}\) compound, the results of a HEMT device with InAs\(_{0.8}\)Sb\(_{0.2}\) digital alloy as the channel can also give us a hint about how high the electron mobility of bulk InAs\(_{0.8}\)Sb\(_{0.2}\) can be [11]. The structure was grown on GaAs substrates with
2.1 µm AlSb buffer layer, and the 10 nm channel was composed of 10 pairs of 4 monolayers (ML) InAs/1 ML InSb superlattice. The room-temperature electron mobility and sheet carrier density are 14,000 cm²/V-s and 1.4 x 10¹² cm⁻², respectively. However, there are some concerns about the digital alloy approach. First, the theoretical critical thickness of InSb on InAs is less than 1 ML (calculated using Matthews and Blakeslee’s method [12]). Second, because antimony tends to float at the growth front when an antimonide compound is grown, the switching between arsenic and antimonide compounds requires attention to avoid intermittent compounds existing in between the superlattice layers [13]. Therefore, it is worthwhile to investigate growing the InAsₓSb₁₋ₓ layer directly to avoid these issues.

![Bandgap energy and lattice constant of various III-V semiconductors at room temperature](image)

Fig. 1.4 Bandgap energy and lattice constant of various III-V semiconductors at room temperature [14].
1.3 In$_x$Ga$_{1-x}$Sb Compound for P-Type FET

For III-V compound semiconductors to be applied in logic circuits, the biggest concern is the imbalance of hole and electron mobility. To achieve the same current driving ability as n-type FET, p-type FET would require a larger device area because of lower carrier mobility. As shown in Figure 1.2, although many III-V compound semiconductors possess much higher electron mobility than silicon, they have little advantage in hole mobility. For example, In$_x$Ga$_{1-x}$As compounds are known to have high electron mobility, but they are not good candidates for p-type FET. InAs (400 cm$^2$/V-s) and GaAs (400 cm$^2$/V-s) have almost equivalent intrinsic hole mobility as silicon. In fact, Ge (2200 cm$^2$/V-s) has the highest intrinsic hole mobility among elemental and compound semiconductors. The rapid development of Si/Ge MOSFET could easily erase the advantage of III-V materials in electron mobility. Therefore, it is important to enhance the hole mobility in III-V semiconductors to balance p- and n-FETs made from them.

In order to improve the hole mobility in III-V semiconductors, strain-induced hole mobility enhancement used in Si MOSFETs could be also applied [3]. The introduction of either compressive or tensile strain in Si or SiGe MOSFET has been proved to be effective in improving hole mobility. The enhancement factor can be as large as 1.5 or even larger over unstrained silicon, depending on the composition of the channel and applied strain. There have been extensive studies, both theoretical and experimental, on how the strain improves the hole mobility [15]. The main concept of strain-induced mobility improvement is to increase the warping of valence bands, especially the heavy-hole one, and split heavy-hole and light-hole bands away, as shown in Figure 1.5 [3]. In either the tensile or compressive strain case, the effective mass around the lowest energy for hole population reduces because the top-most valence band curve becomes light-hole-like. The curvature of the top-most valence band curve

7
increases and effective hole mass reduces; therefore hole mobility increases. Besides the mobility enhancement by effective mass reduction, the reduction of scattering could also contribute to the enhancement [15].

The same phenomenon of warping and splitting of valence band curves is also expected in III-V semiconductors. It is shown by simulation how the valence band of In$_x$Ga$_{1-x}$As would change under tensile and compressive strains [16]. Along in-plane directions, both tensile and compressive strains would separate heavy-hole and light-hole curves. The separation increases the probability that at lower energy, hole population would possess a lighter mass. In this case of In$_x$Ga$_{1-x}$As, the curvature of the lowest energy in the valence band would be larger in compressive strain than in tensile strain, which means compressive strain can enhance hole mobility more. Therefore, the hole mobility of compound semiconductors could be enhanced with strain, which can be introduced by simply changing the composition during epitaxy.

![Diagram of valence band structure](image)

Fig. 1.5 Valence band structure for longitudinal in-plane direction. (a) Unstrained and (b) strained Si [3].
Among III-V compound semiconductors, GaSb (850 cm\(^2\)/V-s) and InSb (800 cm\(^2\)/V-s) are the only compounds with much higher hole mobility than Si. Therefore, In\(_x\)Ga\(_{1-x}\)Sb compound is a good candidate as the channel of p-type FETs. When forming compounds, GaSb and InSb are totally miscible and the physical properties of the In\(_x\)Ga\(_{1-x}\)Sb alloy vary monotonically with increasing indium composition without suffering severe alloy scattering at room temperature [17]. Recently, a room-temperature hole mobility about 1000 cm\(^2\)/V-s at a sheet carrier concentration around \(1.1 \pm 0.2 \times 10^{12}\) cm\(^{-2}\) has been demonstrated in strained In\(_x\)Ga\(_{1-x}\)Sb/Al\(_x\)Ga\(_{1-x}\)Sb modulation-doped quantum well (QW) FETs grown on GaAs substrates using beryllium (Be) as the p-type dopant [18]. The hole mobility increases with increasing indium content, showing a peak value when InSb composition is 0.4. Similar hole mobility improvement, with either In\(_x\)Ga\(_{1-x}\)Sb [19] or GaSb [20], was also reported recently. These results demonstrated that III-V complementary logic circuits with balanced n- and p-channel FETs similar to Si/Ge CMOS are possible.

However, Be, which is widely utilized for p-type doping in MBE grown III-V materials, has the tendency of out-diffusing to adjacent layers, especially at elevated temperatures. This characteristic raises the concern that Be may diffuse into the conduction channel since the distance between the Be-doped layer and the channel is generally less than 20 nm. Diffusion of Be into the conduction channel would cause impurity scattering and decrease the hole mobility. The out-diffusion of Be has been reported elsewhere in InSb:Be, which shows a broad secondary ion mass spectroscopy (SIMS) doping profile even using a low growth temperature of 340 °C [21]. In that report, the doping concentration in InSb at 20 nm below the 30 nm-thick Be-doped InSb is ten times larger than the background carrier concentration. The growth temperatures for the In\(_x\)Ga\(_{1-x}\)Sb channel and the Al\(_x\)Ga\(_{1-x}\)Sb buffer layer are usually much higher than 340 °C, which increases the possibility of Be diffusion. In contrast, carbon has a much smaller diffusion
coefficient than Be. For example, the diffusion coefficient of carbon is one fifth that of Be in GaAs at 800 °C [22]. Therefore, to avoid this problem, we choose carbon as the p-type dopant since it is much more stable than Be and would minimize the impurity scattering.

1.4 Current Gain Improvement of GaAsSb HBT

InP-based heterojunction bipolar transistor (HBT) is the only known transistor technology that can deliver tera-hertz (THz) bandwidth with output breakdown voltage (BV$_{CEO}$) greater than 1 V [23]. These figures of merit make InP-based HBT the leading candidate for higher power transistor amplifiers. Recent developments have proven type-II GaAsSb/InP double heterojunction bipolar transistors (DHBT), grown by either MBE [24,25] or metal-organic chemical vapor deposition (MOCVD) [26], to possess characteristics comparable to InGaAs/InP type-I HBT [27]. The cutoff frequency of sub-micron HBT devices could be as high as 620 GHz [24], while the output breakdown voltage is higher than 5 V. A type-II base/collector junction can eliminate the current blocking effect, and a wide-bandgap collector can reduce impact ionization and increase the breakdown voltage. However, the same conduction band alignment at the emitter/base junction forms an energy barrier, which would block electrons injected from the InP emitter. Therefore, the base current from recombination at the emitter/base junction would increase. Since current gain ($\beta$) is defined as $I_C/I_B$, increased base current would lower the current gain. Also, the energy barrier could defer the electron transport and maybe even degrade the high-frequency response. Therefore, it is important to modify the emitter/base junction to improve the current gain and operation frequency of a GaAsSb/InP DHBT.
1.4.1 Conduction Band Discontinuity of GaAs$_{0.5}$Sb$_{0.5}$/InP

Although the band alignment between GaAsSb and InP is type-II, the conduction band discontinuity ($\Delta E_c$) between undoped GaAsSb and InP is small and the reported values vary. As shown in Figure 1.6, $\Delta E_c$ at room-temperature and equilibrium is about 0.08 eV, which is calculated with model-solid theory and parameters of GaAs and GaSb [28]. This value could be underestimated and may be larger in real structures. Temperature-dependent current-voltage measurements of MBE-grown lattice-matched GaAsSb/InGaAs single-barrier diode show that $\Delta E_c$ is $0.47 \pm 0.02$ eV [29]. Because $\Delta E_c$ between InP and InGaAs is 0.2 eV [30], experimentally determined $\Delta E_c$ between GaAsSb and InP is roughly 0.27 eV. Also, from type-II room-temperature photoluminescence measurements of low-pressure metalorganic vapor phase

![Fig. 1.6 Band energies of GaAs$_{0.5}$Sb$_{0.5}$/InP heterojunction. The conduction band discontinuity between these two compounds is approximately 0.08 eV. The table on the right lists the material parameters used for the calculation.](image)

<table>
<thead>
<tr>
<th>Material Parameters</th>
<th>GaAs</th>
<th>GaSb</th>
</tr>
</thead>
<tbody>
<tr>
<td>Lattice constant (Å)</td>
<td>5.6533</td>
<td>6.0959</td>
</tr>
<tr>
<td>$E_{v,av}$ (eV)</td>
<td>-6.92</td>
<td>-6.25</td>
</tr>
<tr>
<td>$\Delta$ (eV)</td>
<td>0.34</td>
<td>0.82</td>
</tr>
<tr>
<td>$E_v$ (eV)</td>
<td>-6.807</td>
<td>-5.977</td>
</tr>
<tr>
<td>$a_v$ (eV)</td>
<td>1.16</td>
<td>0.79</td>
</tr>
</tbody>
</table>
epitaxy (LP MOVPE)-grown GaAsSb/InP heterostructures, the $\Delta E_c$ between InP and GaAsSb was found to be about 0.18 eV [31]. These experimental results indicate that $\Delta E_c$ at a type-II GaAsSb/InP heterojunction could be larger than the theoretically estimated value.

To eliminate the energy barrier at the E/B junction, several approaches could be used. Adjusting the base composition of ternary GaAs$_x$Sb$_{1-x}$ (higher antimonide composition) or quaternary In$_x$Ga$_{1-x}$As$_y$Sb$_{1-y}$ alloy could lower the conduction band energy of the base layer. But this approach would either induce strain in the base (ternary GaAs$_x$Sb$_{1-x}$) or increase the base resistance (quaternary In$_x$Ga$_{1-x}$As$_y$Sb$_{1-y}$). An easier way is to replace InP with larger-bandgap compounds, which could raise the conduction band edge on the emitter side and reduce the conduction band barrier between the emitter and the base. The change of the emitter/base junction to type-I band alignment could eliminate the current blocking and improve the performance of GaAsSb-based DHBTs.
Chapter 2 GROWTH OF INAS$_{0.8}$SB$_{0.2}$ COMPOUND FOR N-TYPE FET

2.1 GSMBE System

The sample growth is carried out in a modified Perkin-Elmer gas source molecular beam epitaxy (MBE) system, as shown in Figure 2.1. The samples are introduced through the load-lock chamber, which is pumped with a cryopump (Helix, OB-100). Then the samples are transferred into the transfer chamber, which is pumped by an ion pump to keep the pressure lower than $1 \times 10^{-9}$ torr. Before loading into the growth chamber, samples are pre-baked (up to 250 °C and 300 °C for InP and GaAs, respectively) on the heating stage in the transfer chamber to get rid of residual contaminants. The growth chamber is pumped by a turbo pump (2200 liter/sec, Leybold MAG W2000) and a cryopump (9000 liter/sec, Helix OB-10). The dual pumps can keep the background pressure in the range of mid to low $1 \times 10^{-9}$ torr, with effusion cells heated up to temperatures for the growth and liquid nitrogen flowing in the cryopanel and the source shroud.

Several gauges and instruments are installed to assist growth. An ion gauge is located behind the sample growth position to calibrate the beam equivalent pressure (BEP) of group III elements and antimony. A quadrupole mass spectrometer (QMS) is installed with the probe located near the growth position. This instrument allows for the analysis of residual gases in the growth chamber and cracked species of group V elements. In addition, a reflective high energy electron-beam diffraction (RHEED) system (Staib Inc. EK-15-R) is installed, glancing on the sample surface with a small angle. It provides surface reconstruction patterns during the growth. To monitor the sample temperature, besides a thermocouple located behind the sample, a pyrometer is installed, aiming at the sample surface through a quartz view port. Its temperature reading is calibrated using the desorption temperature of GaAs (580 °C).

The group III elements, including gallium (Ga), indium (In) and aluminum (Al), are supplied using effusion cells, pointing at the sample growth position. The on/off of group III flux is
controlled using pressure-actuated shutters in front of effusion cells. Arsenic (As) and phosphor (P), on the other hand, are provided using thermally cracked high-purity arsine (AsH₃) and phosphine (PH₃). AsH₃ and PH₃ are injected through thermal crackers with temperatures higher than 950°C, and cracked into dimers, As₂ and P₂, respectively, and hydrogen (H₂). Because of the high vapor pressure of hydrogen, the growth chamber pressure typically increases to mid to high 1E-6 when AsH₃ and PH₃ are injected. The flow rate are controlled by mass flow controllers (Brooks Instruments). A vent-run configuration is employed for the gas line to ensure an abrupt switching of group V elements. When arsine and phosphine are not injected into the chamber, the gas lines are pumped with a turbo pump (Pfeiffer TPU 510C) to avoid gas pile-up and overshoot of the growth chamber pressure when the gases are injected. During the standby period, high-purity hydrogen constantly purges the gas lines to prevent contamination.

Fig. 2.1 Illustration of the modified Perkin-Elmer gas-source molecular beam epitaxy.
For antimony source, a solid antimony cracker cell (Veeco PE-200-Sb) with an adjustable trombone valve, as shown in Figure 2.2, is used. This cracker has two independent temperature zones. One is for the Sb bulk charge evaporator and set at 560 °C to provide the Sb vapor pressure, while the other is at the crack zone and set at 900 °C to crack the polyatomic Sb flux into atomic or small molecular Sb. The Sb flux can be fine-tuned by an automated position controller, which adjusts the pyrolytic boron nitride (PBN) trombone valve in the front of the bulk evaporator. The Sb flux can be completely switched on/off by the shutter in front of the cell.

Fig. 2.2 Schematic of Veeco 200 cm³ antimony valved cracker (modified after [32]).

Regarding the dopants, carbon (C) and silicon (Si) are used for p-type and n-type dopants, respectively. Vapors from temperature-controlled carbon tetrabromide (CBr₄) and silicon tetrabromide (SiBr₄) bubblers are used as the source. The concentrations of C and Si are modulated in different ways. CBr₄ is injected through an ultra-high vacuum leak valve, and a pressure controller (MKS) installed in the CBr₄ line precisely controls the pressure. On the other hand, the
flow of SiBr\textsubscript{4} is controlled by a mass flow controller to change the doping concentration. Both dopants are injected into the growth chamber through the same thermal cracker held at 150°C, and the same run-vent configuration as in the hydride gas line is also applied to produce abrupt doping profiles.

After the growth, samples are characterized with several analysis tools. High-resolution x-ray diffraction (HRXRD, Bede D1) with a highly monochromatic Cu Kα1 x-ray beam conditioned by four bounces through Si <022> channel cut crystals is used to measure the rocking curves and characterize compound compositions and film quality. Atomic force microscopy (AFM, Digital Instruments Dimension 3000) with tapping mode and scanning electron-beam microscopy (SEM, Hitachi S-4800) are used to observe the surface roughness and morphology. To measure the resistivity, carrier concentration and mobility, a Hall measurement system (Accent HL5500) with a fixed magnetic field of 0.5 Tesla is used.

2.2 Sample Growth

The samples were grown on Fe-doped semi-insulating InP substrates. Two-inch InP substrates were cut into quarters. The samples were loaded in two ways. The first one was to mount the sample on a heated (280 °C) molybdenum (Mo) puck with high-purity indium. The second method, the non-bonding method, was to drop the sample inside the open slot of the thin Mo holder backed with a PBN plate. The PBN plate could enhance the thermal conduction and distribution. The second method was better for faster temperature ramping and stable temperature control, which is important for the exploration of the temperature effect during epitaxy.

After loading, the growth started with a thermal cleaning of the InP surface. InP substrates were first ramped up at a rate of 100 °C/min to 300 °C, and 5 sccm PH\textsubscript{3} would inject at 300 °C to provide the phosphor overpressure and suppress the phosphor dissociation of the InP substrate.
The substrates were then ramped at a rate of 70 °C /3 min up to 510 °C. Because of the emissivity difference between test and prime grade InP substrates and the mounting methods, InP surface oxides would be completely desorbed at different temperatures, using the readings of the thermocouple. Usually for samples loaded with the non-bonding method, which was used to grow all the samples, the oxides could be desorbed with the thermal couple reading lower than 500 °C. When the InP surface was cleaned, RHEED would show a (2x8) streaky pattern. The substrate temperature would then decrease by 35 °C to start the growth. For a general reference, the desorption temperature of InP (T\textsubscript{d}) was set as 480 °C.

The structure for the growth of bulk InAs\textsubscript{0.8}Sb\textsubscript{0.2} is shown in Figure 2.3. After the desorption of InP substrate, a 100 nm InP buffer layer was first grown at a substrate temperature of T\textsubscript{d} - 35. After the 100 nm InP buffer layer, a 100 nm AlAs\textsubscript{0.5}Sb\textsubscript{0.5} lattice-matched to InP was grown, and following the AlAs\textsubscript{0.5}Sb\textsubscript{0.5} layer was the AlSb layer. Both AlAs\textsubscript{0.5}Sb\textsubscript{0.5} and AlSb were grown at T\textsubscript{d} + 20. At the end, the growth was concluded with an unintentionally doped InAs\textsubscript{0.8}Sb\textsubscript{0.2} active layer. The flow rates of AsH\textsubscript{3} and Sb flux for the growth of InAs\textsubscript{0.8}Sb\textsubscript{0.2} were determined using HRXRD.

<table>
<thead>
<tr>
<th>Layer</th>
<th>Thickness</th>
</tr>
</thead>
<tbody>
<tr>
<td>InAs\textsubscript{0.8}Sb\textsubscript{0.2} (200 nm)</td>
<td></td>
</tr>
<tr>
<td>AlSb (800 nm)</td>
<td></td>
</tr>
<tr>
<td>AlAs\textsubscript{0.5}Sb\textsubscript{0.5} (100 nm)</td>
<td></td>
</tr>
<tr>
<td>InP (100 nm)</td>
<td></td>
</tr>
<tr>
<td>InP:Fe</td>
<td></td>
</tr>
</tbody>
</table>

Fig. 2.3 Layer structures for the growth of bulk InAs\textsubscript{0.8}Sb\textsubscript{0.2}.
2.2.1 Buffer Layer

The lattice mismatch between InP (lattice constant = 5.8686 Å) and AlSb (lattice constant = 6.1355 Å) is about 4.7%. The critical layer thickness is less than one monolayer (calculated using Matthews and Blakeslee’s model [12]). Therefore, it is difficult to grow high-quality InAs$_{0.8}$Sb$_{0.2}$ directly on InP and a buffer layer is required. Here we used an AlSb/AlAs$_{0.5}$Sb$_{0.5}$ composite buffer layer scheme. Usually, growing one compound on another with such a large lattice mismatch would result in a change of growth mode from two-dimensional (2D, planar) to three-dimensional (3D, island). That means a spotty RHEED pattern would be observed when initializing AlSb growth directly on InP, as shown in Figure 2.4(a). The 3D growth morphology will introduce extra defects and, thus, degrade the transport property of the top layer (here it is InAs$_{0.8}$Sb$_{0.2}$). However, if AlSb is grown on AlAs$_{0.5}$Sb$_{0.5}$, the RHEED pattern does not change to spotty and remains streaky (1x3) during the transition, as shown in Figure 2.4(b). This indicates that two-dimensional mode, and hence a smooth growth front, are maintained. The smooth growth front is preferred for the hetero-epitaxial growth, and the AlSb/AlAs$_{0.5}$Sb$_{0.5}$ bi-layer buffer layer structure could improve the electron mobility of the top layer.

![RHEED patterns](image)

Fig. 2.4 RHEED patterns of AlSb grown on (a) InP and (b) AlAs$_{0.5}$Sb$_{0.5}$. (a) is along [110] while (b) is along [110].
The phenomenon of the RHEED pattern remaining streaky during AlSb/AlAs$_{0.5}$Sb$_{0.5}$ growth transition could be explained by the surfactant effect of antimony. It was determined that antimony acts as a surfactant during epitaxial growth of lattice-mismatched hetero-layers to suppress the 3D growth and enhance a planar growth front [33,34]. During the growth of antimony compounds, the antimony atoms have the tendency to segregate from the bulk and form a thin layer floating on the growth front. When growing the AlSb/AlAs$_{0.5}$Sb$_{0.5}$ buffer layer structure, the common antimony cation of both layers acts as a surfactant and promotes a 2D growth front. This assessment is supported by the smooth streaky RHEED pattern observed during the transition between these two layers. To verify the surfactant effect of antimony compounds, we also tried to grow AlSb on GaAs$_{0.51}$Sb$_{0.49}$, which is lattice-matched to InP. As a result, during the growth transition from GaAs$_{0.51}$Sb$_{0.49}$ to AlSb, the RHEED also remained streaky (1x3) and did not show a spotty pattern, the same as AlAs$_{0.5}$Sb$_{0.5}$ to AlSb. The electron mobility of InAs$_{0.8}$Sb$_{0.2}$ on top of this kind of buffer layer demonstrates a similar value as that grown on AlSb/AlAs$_{0.5}$Sb$_{0.5}$.

2.3 Results and Discussions

2.3.1 HRXRD and XTEM

As shown in Figure 2.5, the HRXRD rocking curve of the as-grown InAs$_{0.8}$Sb$_{0.2}$ structure has two major peaks. The strongest peak, located at zero arc-sec, is attributed to the InP substrate and the AlAs$_{0.5}$Sb$_{0.5}$ buffer layer. The peak located near -5600 arc-sec is attributed the 800 nm AlSb buffer layer and 200 nm InAs$_{0.8}$Sb$_{0.2}$. The relaxation of the AlSb buffer layer is 98%, which is determined by measuring the sample at two glancing angles, (004) and (115). The highly relaxed AlSb indicates that almost no strain existed for the growth of InAs$_{0.8}$Sb$_{0.2}$. 
Fig. 2.5 HRXRD rocking curve of the InAs\textsubscript{0.8}Sb\textsubscript{0.2} structure.

Figure 2.6 shows the cross-sectional transmission electron micrograph (XTEM) of the InAs\textsubscript{0.8}Sb\textsubscript{0.2}/AlSb/AlAs\textsubscript{0.5}Sb\textsubscript{0.5}/InP structure. Although the 2D growth mode was observed at the AlSb/AlAs\textsubscript{0.5}Sb\textsubscript{0.5} interface, the large lattice mismatch between AlSb/AlAs\textsubscript{0.5}Sb\textsubscript{0.5} and InP must be accommodated through the formation of dislocations. Figure 2.6(a), which demonstrates the AlSb/AlAs\textsubscript{0.5}Sb\textsubscript{0.5} interface, shows a high density (>10\textsuperscript{10} cm\textsuperscript{-2}) of threading dislocations. However, after the growth of 800 nm of AlSb buffer layer, the threading dislocation density reduced dramatically and only very few are visible in the images. Figure 2.6(b) shows the XTEM image of the AlSb/InAsSb interface, where one dislocation line is visible in the image, which corresponds to a defect density of ~10\textsuperscript{7} cm\textsuperscript{-2}. The decreased defect density indicates the high quality of the InAs\textsubscript{0.8}Sb\textsubscript{0.2} epilayer.
Fig. 2.6 Cross-sectional transmission electron micrographs (XTEMs) of the (a) AlSb/AlAs$_{0.5}$Sb$_{0.5}$/InP and (b) InAs$_{0.8}$Sb$_{0.2}$/AlSb interface.
2.3.2 Growth Optimization

The relation between the buffer layer thickness and the electron mobility InAs$_{0.8}$Sb$_{0.2}$ was explored. The thickness of InAs$_{0.8}$Sb$_{0.2}$ was fixed at 200 nm and the thickness of AlSb was varied from 0 to 1000 nm. To demonstrate the effect of AlSb/AlAs$_{0.5}$Sb$_{0.5}$ composite buffer layer, InAs$_{0.8}$Sb$_{0.2}$ samples with AlSb buffer layer grown directly on InP were also prepared for comparison. The growth temperature of InAs$_{0.8}$Sb$_{0.2}$ was $T_d$ - 35. Figure 2.7 shows the room-temperature electron mobility of 200 nm-thick InAs$_{0.8}$Sb$_{0.2}$ versus the thickness of different buffer layers. As expected, the mobility increased with increasing buffer layer thickness for samples with composite and single buffer layer. This is because the dislocation density decreases as the buffer layer thickness increases, which could be observed in the XTEM results. As the figure shows, it is apparent that structures with the composite buffer layer always demonstrate a higher electron mobility than those with single AlSb buffer. When the thickness of AlSb increases from 200 to 800 nm, the InAs$_{0.8}$Sb$_{0.2}$ grown on the composite buffer layer has the highest room-temperature electron mobility of 12,000 cm$^2$/V-s. With the same buffer layer thickness, the electron mobility of the InAs$_{0.8}$Sb$_{0.2}$ grown on the single AlSb buffer layer is 10,000 cm$^2$/V-s. The improvement of the electron mobility of the thin InAsSb layer indicates the effectiveness of inserting an AlAsSb buffer layer in between the InP substrate and the AlSb buffer layer.
Fig. 2.7 Room-temperature electron mobility of 200 nm-thick InAs$_{0.8}$Sb$_{0.2}$ versus the thickness of different buffer layers. The carrier concentration is in the range of 1-3 x $10^{16}$ cm$^{-3}$ for all points. Circle points represent structures with AlSb/AlAsSb buffer layers, while square points represent those with AlSb-only buffer layer.

Etching pit density (EPD) was examined on samples grown on different buffer layers to provide a complimentary comparison. The samples were etched with hydrogen chloride (HCl) for 10 seconds to reveal the etching pits. The etching rate, referring to [35], would be slightly less than 1 nm/s. The pit density was counted under an optical microscope with differential interference contrast (Nomarski) imaging. Because the wet etching may not reveal all the defects completely, the pit density may be lower than the defect density estimated by XTEM. As shown in Figure 2.8, InAs$_{0.8}$Sb$_{0.2}$ samples grown on the AlSb/AlAs$_{0.5}$Sb$_{0.5}$ composite buffer layer demonstrated lower EPD than those grown on the single AlSb buffer layer. When AlSb is 1000 nm thick, the EPD of the sample with composite buffer layers is about one third that of the sample with single buffer.
layer. The result indicates that AlSb/AlAs$_{0.5}$Sb$_{0.5}$ composite buffer layer could lead to a better epitaxial quality.

![Graph](image)

**Fig. 2.8** Etching pit density of 200 nm-thick InAs$_{0.8}$Sb$_{0.2}$ versus the thickness of different buffer layers. Circle points represent structures with AlSb/AlAsSb buffer layers, while square points represent those with AlSb-only buffer layer.

The growth temperature of InAs$_{0.8}$Sb$_{0.2}$ was varied to see how it affects the electron mobility. Unlike AlSb, InAs-rich compounds usually required a lower growth temperature to achieve better quality [36]. The layer structure shown in Figure 2.3 was used and the growth temperature of InAs$_{0.8}$Sb$_{0.2}$, as shown in Figure 2.9, was varied from 445 °C (T$_d$ -35) to 330 °C (T$_d$ -150). The substrate temperature was lowered only at the final stage of AlSb growth and at a ramping rate of 40 °C/min. The temperature ramping was set to be as fast as possible to maintain a uniform growth of the AlSb buffer. Figure 2.9 demonstrates that the room-temperature electron mobility of InAs$_{0.8}$Sb$_{0.2}$ increased from ~13,000 to 14,500 cm$^2$/V-s as the growth temperature decreased from
445 °C to 360 °C. The data suggests that the optimized growth temperature for InAs$_{0.8}$Sb$_{0.2}$ was around 360 °C, or 120 °C below $T_d$. Further decreasing the growth temperature down to 330 °C did not lead to a better result, and therefore 360 °C was selected for the following experiments.

![Graph](image)

**Fig. 2.9** Room-temperature electron mobility of InAs$_{0.8}$Sb$_{0.2}$ grown at different temperatures. The buffer layer is fixed with 100 nm AlAs$_{0.5}$Sb$_{0.5}$ plus 800 nm AlSb. The carrier concentration is in the range of 1-3 x $10^{16}$ cm$^{-3}$ for all points.

### 2.3.3 Interface vs. Mobility

Using the optimized growth temperature for InAs$_{0.8}$Sb$_{0.2}$, the AlSb buffer layer thickness was further increased to achieve higher electron mobility. As shown in Figure 2.10, several InAs$_{0.8}$Sb$_{0.2}$ samples were grown at 360 °C using the layer structure shown in Figure 2.3, with different AlSb layer thickness. The result shows that the electron mobility seems to saturate when the thickness of AlSb increases beyond 800 nm. The electron mobility does not improve further as the buffer layer
thickness increases to 1600 nm. Several possible reasons might contribute to the electron mobility saturation in InAs$_{0.8}$Sb$_{0.2}$. The first one could be the finite impurity scattering, which comes from the unintentionally doped background impurities with a concentration near mid-to-high $10^{16}$ cm$^{-3}$. Second, other un-optimized growth conditions of InAs$_{0.8}$Sb$_{0.2}$, in addition to buffer layer thickness, could be the bottleneck for improving electron mobility. Also, the interface roughness between AlSb and InAs$_{0.8}$Sb$_{0.2}$ may be not optimized, which would lead to interface scattering [37].

Fig. 2.10 Room-temperature electron mobility of InAs$_{0.8}$Sb$_{0.2}$ with different buffer layer thicknesses (100 nm AlAsSb plus AlSb of various thicknesses). The thickness and growth temperature of InAsSb are fixed at 200 nm and $T_d$ - 120. The carrier concentration is in the range of 1-3 x $10^{16}$ cm$^{-3}$ for all points.

To clarify whether the InAs$_{0.8}$Sb$_{0.2}$/AlSb interface limits the electron mobility, several samples with 800 nm-thick AlSb but InAs$_{0.8}$Sb$_{0.2}$ with different thickness were grown. The results are shown in Figure 2.11. From the figure, it is clear that as the thickness of InAs$_{0.8}$Sb$_{0.2}$
increases from 200 to 1400 nm, the electron mobility increases from 14,500 to 18,000 cm²/V-s, which corresponds to an improvement of 24%. It is reasonable to assume that as the layer thickness increases, the carrier mobility is less affected by the scattering occurring at the InAs₀.₈Sb₀.₂/AlSb interface. With the AlSb buffer fixed at 800 nm, the electron mobility of InAs₀.₈Sb₀.₂ reaches ~18,000 cm²/V-s as its thickness increases to 1600 nm. However, as previously addressed and shown in Figure 2.10, increasing AlSb buffer from 800 to 1600 nm while InAs₀.₈Sb₀.₂ is maintained at 200 nm would lead to little improvement in the electron mobility of InAs₀.₈Sb₀.₂. This result indicates InAs₀.₈Sb₀.₂/AlSb interface would affect the electron mobility of InAs₀.₈Sb₀.₂.

Fig. 2.11 Room-temperature electron mobility of InAs₀.₈Sb₀.₂ with different thicknesses. The buffer layer is fixed with 100 nm AlAs₀.₅Sb₀.₅ plus 800 nm AlSb. The dark circle represents the electron mobility of a 200 nm-thick InAs₀.₈Sb₀.₂ grown on 1600 nm AlSb/100 nm AlAs₀.₅Sb₀.₅.
To evaluate the interface effects, samples with different interfaces were grown to test. First we adjusted the switching sequence of groups III and V at the interface of AlSb and InAs$_{0.8}$Sb$_{0.2}$. The general switching sequence used in previous experiments (listed as reference in Table 2.1) was to keep antimony on after aluminum (Al) was switched off (InSb-like). Then indium and arsenic (As) were switched on to grow InAs$_{0.8}$Sb$_{0.2}$. For the InAs-like interface, in contrast, both Al and Sb were off and then As was switched on for one second followed by opening the indium shutter for three seconds to form an InAs-like interface. The thickness of InAs was roughly one mono-layer (ML). Then Sb was turned on to grow InAs$_{0.8}$Sb$_{0.2}$. Other factors, such as layer thickness and growth temperatures, were kept the same for both structures. Although the switching was adjusted, the Hall measurement results, as shown in Table 2.1, suggested the InAs-like interface does not provide a better interface and the electron mobility of InAs$_{0.8}$Sb$_{0.2}$ was not improved.

<table>
<thead>
<tr>
<th>Interface</th>
<th>Electron mobility of InAs$<em>{0.8}$Sb$</em>{0.2}$ (cm$^2$/V-s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Reference</td>
<td>14500</td>
</tr>
<tr>
<td>InAs-like</td>
<td>12300</td>
</tr>
<tr>
<td>25 MLs GaSb</td>
<td>16100</td>
</tr>
<tr>
<td>50 MLs GaSb</td>
<td>16300</td>
</tr>
</tbody>
</table>

The other interfacial effect studied was the use of GaSb as an interfacial layer due to the small lattice mismatch between AlSb and GaSb. Since Ga adatoms have higher mobility than Al
adatoms on the sample surface, GaSb can provide a smoother growth front, or interface, for the following InAs$_{0.8}$Sb$_{0.2}$ layer. A smoother interface can reduce the scattering and improve the carrier mobility. GaSb with two different layer thicknesses, 25 and 50 MLs, were inserted between AlSb and InAs$_{0.8}$Sb$_{0.2}$. The Hall measurement results, as shown in Table 2.1, suggest GaSb can effectively improve the electron mobility of InAs$_{0.8}$Sb$_{0.2}$ from 14500 to 16100 cm$^2$/V-s when a 25 monolayer-thick GaSb is inserted between AlSb and InAs$_{0.8}$Sb$_{0.2}$. However, the mobility does not improve much when the thickness of GaSb increases from 25 to 50 MLs. The surface roughness of AlSb and GaSb (2.5 nm)-capped AlSb was also measured using AFM to see the difference. As a result, GaSb-capped AlSb effectively reduced the surface roughness from 0.6 nm of AlSb to 0.35 nm. The smoother surface could contribute to the electron mobility improvement of InAs$_{0.8}$Sb$_{0.2}$.

One concern of using GaSb to improve the interface is that the intrinsic GaSb is not semi-insulating as AlSb. Nonetheless, Hall measurement shows that the sheet resistance of 800 nm-thick GaSb grown on InP is about 4800 $\Omega$/sq, which is much higher than that of InAs$_{0.8}$Sb$_{0.2}$ (130 $\Omega$/sq). If the layer thickness of GaSb can be maintained to a few MLs, the parallel leakage currents can be greatly suppressed. Moreover, unintentionally doped GaSb shows p-type conductivity due to native lattice defects [38], and is opposite to the conductivity of the unintentionally doped InAs$_{0.8}$Sb$_{0.2}$ channel, which is n-type. The InAs$_{0.8}$Sb$_{0.2}$-GaSb would form a reverse-biased p-n junction and the leakage currents will be negligibly low.

2.3.4 HEMT Device Test

Once the high electron mobility was achieved in InAs$_{0.8}$Sb$_{0.2}$, a HEMT structure, as shown in Figure 2.12, was grown. Its energy band diagram, simulated with Simwindows, is shown in
Figure 2.13. The conduction band discontinuity between the InAs$_{0.8}$Sb$_{0.2}$ channel and the AlSb buffer is larger than 1 eV, which is large enough to confine electrons in the channel.

InAs:Si (40 nm)
AlSb (15 nm)
InAs:Si (1.2 nm)
AlSb (10 nm)
InAsSb (15 nm)
AlSb (800 nm)
AlAs$_{0.5}$Sb$_{0.5}$ (100 nm)
InP:Si

Fig. 2.12 InAs$_{0.8}$Sb$_{0.2}$ HEMT structure.

Fig. 2.13 Band diagram of the InAs$_{0.8}$Sb$_{0.2}$ HEMT structure shown in Figure 2.12.
The interface improvement, addressed in the last section, was not used in this HEMT test structure. A two-inch InP substrate was used to grow the structure, following the same procedure described in the previous section. After the growth of 8000Å AlSb/1000Å AlAs$_{0.5}$Sb$_{0.5}$ composite buffer layer, a 15 nm-thick InAs$_{0.8}$Sb$_{0.2}$ channel was grown at 360 °C. On top of the channel was a 10 nm AlSb spacer, followed by a thin (10 Å) silicon-doped InAs layer acting as the delta doped layer. AlSb:Si or GaSb:Si could not be used as the delta-doping because Si shows p-type conductivity in both AlSb and GaSb [39]. On top of the delta-doped layer was a 25 nm AlSb, and the device was finally capped with a heavily silicon doped InAs contact layer. The electron mobility of the as-grown device is about 12500 cm$^2$/V-s, with a sheet carrier density of about 10$^{12}$ cm$^{-2}$. As discussed in the last section, the low electron mobility is due to the InAs$_{0.8}$Sb$_{0.2}$/AlSb interface, which plays an important role because the InAs$_{0.8}$Sb$_{0.2}$ channel is thin. The device was then fabricated with standard procedures with a gate dimension of 6 µm × 100 µm. E-beam-evaporated AuGe (75 nm)/Ni (15 nm)/Au (150 nm) was used as metal contact for all three terminals.

The current-voltage (I-V) characteristics were measured with the HP-4155A semiconductor parameter analyzer. Figure 2.14(a) shows the I-V characteristics of a depletion mode HEMT under different bias conditions. Although the device can be effectively controlled by the gate bias, the I-V curves show a weak saturation when increasing drain-to-source voltage ($V_{DS}$) and the device cannot be completely turned off, indicating that there are leakage paths. Although the AlSb buffer layer was confirmed to be highly resistive, other leakage paths, such as surface leakage, could exist. This issue will be addressed in Chapter 4. Figure 2.14(b) shows the transconductance ($g_m$) and drain current ($I_{DS}$) characteristics of the device. The maximum $g_m$ of the device is about 350 mS/mm when it is biased at $V_{GS} = 0.1$V and $V_{DS} = 1$V. The corresponding drain current density is ~200 mA/mm. This result is quite favorable in comparison with reported InAsSb HEMT results.
using a submicron gate length of 0.1 µm [40]. Further improvement of the process and the interfaces of the HEMT structure, as the last section suggests, should improve the electron mobility as well as the device performance.

![Figure 2.14](image)

**Fig. 2.14** (a) I-V curves at different gate biases \(V_{GS}\) and (b) drain current \(I_{DS}\) and transconductance \(g_m\) vs. gate bias of InAs\(_{0.8}\)Sb\(_{0.2}\) HEMT.
Chapter 3 GROWTH OF IN<sub>x</sub>GA<sub>1-x</sub>Sb COMPOUND FOR P-TYPE FET

3.1 Valence Bands and Effective Mass of Strained In<sub>x</sub>Ga<sub>1-x</sub>Sb

Based on the results of Chapter 2, a p-type FET was also developed, using strained In<sub>x</sub>Ga<sub>1-x</sub>Sb compounds addressed in Chapter 1. To grow strained In<sub>x</sub>Ga<sub>1-x</sub>Sb compounds, the critical thickness and the strain effect on the valence band structures need to be estimated. Figure 3.1 shows the calculated compressive strain and critical thickness of In<sub>x</sub>Ga<sub>1-x</sub>Sb on AlSb with different indium compositions, using Matthews and Blakeslee’s method [12]. Because the lattice constant of GaSb (6.0959 Å) is smaller than AlSb (6.1355 Å), In<sub>x</sub>Ga<sub>1-x</sub>Sb is lattice-matched to AlSb when x is 0.1. As shown in Figure 3.1, when the ratio of InSb (lattice constant = 6.4794 Å) is 0.3, the corresponding compressive strain and critical thickness are 1.2 % and 5 nm, respectively. Therefore, when AlSb is used as the buffer layer and the In<sub>x</sub>Ga<sub>1-x</sub>Sb QW channel thickness is set to be 10 nm, the highest ratio of InSb would be around 0.3.

![Fig. 3.1 Compressive strain and critical thickness of In<sub>x</sub>Ga<sub>1-x</sub>Sb on AlSb at different compositions. The critical thickness is calculated using Matthews and Blakeslee’s method [12]. Dark squares represent the critical thickness, while dark circles represent the strain.](image-url)
Besides the critical thickness, calculations were also carried out to see how the heavy-hole (HH) and light-hole (LH) bands would change with composition. The solutions of 4 x 4 k*p equations were utilized for the calculation of HH and LH bands [41]:

\[
E_{HH}(k_x) = -P_e \left( \frac{\hbar^2 \gamma_1}{2m_0} k_x^2 + \left[ (Q_e + \frac{\hbar^2 \gamma_2}{2m_0} k_x^2)^2 + 3 \left( \frac{\hbar^2 \gamma_2}{2m_0} \right)^2 k_x^4 \right]^{\frac{1}{2}} \right) \tag{3.1}
\]

\[
E_{LH}(k_x) = -P_e \left( \frac{\hbar^2 \gamma_1}{2m_0} k_x^2 - \left[ (Q_e + \frac{\hbar^2 \gamma_2}{2m_0} k_x^2)^2 + 3 \left( \frac{\hbar^2 \gamma_2}{2m_0} \right)^2 k_x^4 \right]^{\frac{1}{2}} \right) \tag{3.2}
\]

\[
P_e = -a_v (\varepsilon_{yy} + \varepsilon_{yy} + \varepsilon_{zz}) ; \quad Q_e = -\frac{b}{2} (\varepsilon_{yy} + \varepsilon_{yy} - 2\varepsilon_{zz}) \tag{3.3}
\]

\[
\varepsilon_{xx} = \varepsilon_{yy} = \frac{a_0 - a}{a} ; \quad \varepsilon_{zz} = \frac{-2C_{12}}{C_{11}} \tag{3.4}
\]

\(a\) : epilayer lattice constant  
\(a_0\) : substrate lattice constant  
\(m_0\) : effective mass  
\(a_v, b, C_{11}, C_{12}, \gamma_1, \gamma_2\) : material parameters

The parameters used in the calculation are from [42], and for In\(_x\)Ga\(_{1-x}\)Sb, its parameters are interpolated from those of GaSb and InSb. Spin-orbit bands are neglected because they are hundreds of mV below HH and LH bands, and the interaction with HH and LH is negligible. As Figure 3.2(a) shows, the splitting between HH and LH bands increases with InSb compositions and hence the compressive strain increases. At the same time, the curvature of the parabolic curves at the zone center increases for the heavy-hole band but decreases for the light-hole band. The effective hole mass at the zone center is estimated by taking the second derivative of the
Fig. 3.2 (a) Heavy-hole and light-hole band In$_x$Ga$_{1-x}$Sb on AlSb at different gallium compositions. Solid lines represent heavy-hole bands, while dashed lines represent light-hole bands. (b) Effective mass of HH and LH band at the zone center versus strains.
band curves. As shown in Figure 3.2(b), the effective hole mass first decreases for HH and increases for LH as indium composition increases from 0.1 to 0.2. Further increasing the indium composition would decrease the effective hole mass of both HH and LH. The HH band would have a lighter mass than the LH band at zero strain. Therefore, it is necessary to apply a compressive strain larger than 0.5%, which means the indium composition has to be larger than 0.2, to observe the mobility enhancement in a strained $\text{In}_x\text{Ga}_{1-x}\text{Sb}$ QW.

### 3.2 Growth of Strained $\text{In}_x\text{Ga}_{1-x}\text{Sb}$ Quantum Well

Like the growth procedures described in Chapter 2, the epitaxy of strained $\text{In}_x\text{Ga}_{1-x}\text{Sb}$ structures was also carried out in the gas-source MBE system. The layer structure of the modulation-doped $\text{In}_{0.3}\text{Ga}_{0.7}\text{Sb}$ QW structure is shown in Figure 3.3. Figure 3.4 shows the band diagram of the structure, which is simulated with Simwindow. From the simulation result, the valence band discontinuity is larger than 200 meV, which is enough for the carrier confinement. The indium composition in InGaSb QW was set to be 0.3, and the layer thickness was 9 nm, which is within the critical layer thickness shown in Figure 3.1 to avoid defect generation. The valence band discontinuity between the $\text{In}_{0.3}\text{Ga}_{0.7}\text{Sb}$ QW and the AlSb barrier is about 500 meV, which is large enough for the hole confinement. The buffer layers up to the $\text{In}_{0.3}\text{Ga}_{0.7}\text{Sb}$ channel were grown by following the same growth procedures addressed in Chapter 2. After the AlSb buffer layer grown at 500 °C, the substrate temperature was lowered to 430 °C to grow the pseudomorphic $\text{In}_{0.3}\text{Ga}_{0.7}\text{Sb}$ channel layer. The growth was finalized with a 25 nm AlSb barrier layer and 5 nm InAs contact layer. The delta-doping layer was inserted inside the AlSb barrier layer and located at 15 nm above the channel. Unlike the Be-doping structures reported in Sb-alloy structures [18], where a thin region (~ 2 nm) of the barrier layer is doped, the carbon-doping utilized here is a delta-doping with zero thickness. Carbon delta-doping was
achieved by interrupting the group III element (aluminum) while the flow of CBr\textsubscript{4} was on. After the growth, the lattice constants of the structure were determined from the rocking curves measured with HRXRD. To perform Hall measurements, the ohmic contacts were made with In:Zn alloy on the InAs contact layer, and the excess InAs materials outside the contact regions were then removed with an etching solution consisting of citric acid and hydrogen peroxide.

![Fig. 3.3 Modulation-doped p-channel In\textsubscript{0.3}Ga\textsubscript{0.7}Sb strained QW structure.](image)

![Fig. 3.4 Band diagram of the modulation-doped p-channel In\textsubscript{0.3}Ga\textsubscript{0.7}Sb QW structure.](image)
3.3 Results and Discussions

3.3.1 Calibration of In$_{0.3}$Ga$_{0.7}$Sb QW

For the HRXRD measurement, the capping InAs contact layer was removed prior to the measurement to simplify the result. As shown in Figure 3.5, the HRXRD rocking curve of a pseudomorphic In$_{0.3}$Ga$_{0.7}$Sb QW structure shows multiple peaks. Because most of the layers are the same as the n-type InAs$_{0.8}$Sb$_{0.2}$ structure, the rocking curve shows peaks similar to those in Figure 2.5. The strongest peak, located at zero arc-sec, is attributed to the InP substrate and AlAs$_{0.5}$Sb$_{0.5}$ buffer layer, which is lattice-matched to InP. The peak located near -5600 arc-sec is the 800 nm AlSb buffer layer. The minor peak at around -8400 arc-sec belongs to the strained In$_x$Ga$_{1-x}$Sb QW. The analysis of the peak position with respect to the AlSb peak showed the composition of InSb in the In$_x$Ga$_{1-x}$Sb QW is 0.3, assuming the strain is not relaxed.

Fig. 3.5 High-resolution X-ray rocking curves of strained In$_x$Ga$_{1-x}$Sb/AlSb quantum well structure with x = 0.3. Peaks are attributed to the InP substrate plus AlAs$_{0.5}$Sb$_{0.5}$, AlSb buffer and In$_{0.3}$Ga$_{0.7}$Sb quantum well.
To set up the baseline of the intrinsic hole mobility and sheet carrier concentration of strained In$_{0.3}$Ga$_{0.7}$Sb QW structures, structures without carbon delta-doping were first grown. The growth temperature of the In$_{0.3}$Ga$_{0.7}$Sb QW was fixed at 430 °C, and the V/III BEP ratio was adjusted to optimize the intrinsic characteristics. A 5 nm-thick In$_{0.3}$Ga$_{0.7}$Sb with different V/III beam equivalent pressure (BEP) ratios was grown on AlSb, which has a root-mean-square (RMS) roughness of ~ 0.6 nm. As shown in Figure 3.6, if the V/III BEP ratio was too low, which means too little antimony, the surface would be filled with Ga droplets. In contrast, if the V/III BEP ratio was too high, which means there are too many antimony surface atoms, the group III adatoms would not be able to move freely and lead to a deteriorated surface. Therefore, in order to achieve good film quality, the V/III BEP ratio needs to be maintained in a certain range.

![AFM Image](image)

<table>
<thead>
<tr>
<th>V/III Pressure ratio</th>
<th>2.2</th>
<th>6.5</th>
<th>12</th>
</tr>
</thead>
<tbody>
<tr>
<td>AFM $R_q$ (nm)</td>
<td>4.2</td>
<td>0.35</td>
<td>15.7</td>
</tr>
</tbody>
</table>

Fig. 3.6. Surface morphology of 5 nm-thick In$_{0.3}$Ga$_{0.7}$Sb grown on AlSb with different V/III beam equivalent pressure ratios. Images shows that too little antimony would lead to droplets, while too much antimony would inhibit the migration of group III adatoms and form clusters.

The intrinsic hole mobility and sheet carrier concentration of samples grown under different
V/III BEP ratios are plotted in Figure 3.7. The hole mobility of the strained In$_{0.3}$Ga$_{0.7}$Sb QW generally decreases from above 350 cm$^2$/V-s to below 300 cm$^2$/V-s as the V/III BEP ratio is increased. On the other hand, the sheet carrier concentration decreases with the decreasing V/III BEP ratio, opposite to the trend of the hole mobility. The same variations of hole mobility and sheet carrier concentration with V/III BEP ratio have also been reported in GaSb grown by MBE [42]. The dependence of the sheet carrier concentration in In$_{0.3}$Ga$_{0.7}$Sb on the V/III BEP ratio is similar to GaSb. By adjusting the V/III BEP ratio, the sheet carrier concentration could be lowered to about $8\times10^{11}$ cm$^{-2}$, and the hole mobility at this concentration is about 360 cm$^2$/V-s.

Fig. 3.7. Intrinsic sheet carrier concentration and hole mobility vs. V/III beam equivalent pressure ratio used to grow the unintentionally doped In$_{0.3}$Ga$_{0.7}$Sb quantum well, i.e. the quantum well structures without carbon delta-doping. Filled squares represent the hole mobility while empty squares represent the sheet carrier concentration.

However, despite the addition of indium into the QW, the pseudomorphic In$_{0.3}$Ga$_{0.7}$Sb QW
structure does not show an expected higher mobility than the GaSb QW structure. The reference GaSb QW structure was grown and measured using the same layer structure as shown in Figure 3.3, except the channel material was a 15 nm-thick GaSb. The intrinsic hole mobility and sheet carrier concentration of the GaSb QW structure are $360 \text{ cm}^2/\text{V-s}$ and $1.3 \times 10^{12} \text{ cm}^{-2}$, respectively. The hole mobility of the unstrained GaSb channel is the same as that of the strained In$_{0.3}$Ga$_{0.7}$Sb channel at a similar sheet carrier concentration, which means there is no hole mobility improvement despite the incorporation of indium and compressive strain into the GaSb channel. The results indicate that other factors might have limited the hole mobility of the strained In$_{0.3}$Ga$_{0.7}$Sb channel. For example, heterojunction interface roughness between AlSb and In$_{0.3}$Ga$_{0.7}$Sb could lead to interface scattering and affect the carrier transport in the channel.

### 3.3.2 Buffer Layer Modification

To clarify the sources of the extra scatterings, samples with the same structure as shown in Figure 3.3 were grown, except the AlSb buffer layer was replaced with Al$_{0.7}$Ga$_{0.3}$Sb to improve the surface smoothness of the buffer layer. Devices with an Al$_{0.7}$Ga$_{0.3}$Sb buffer layer and 9 nm strained In$_{0.3}$Ga$_{0.7}$Sb QW channel were grown, and the V/III flux ratio during the growth of the channel was varied between 3.5 and 5.5. The results of intrinsic hole mobility and sheet carrier concentration for samples grown at different V/III flux ratios were shown in Figure 3.8. The sheet carrier concentration was higher for the structures grown, at similar V/III flux ratios, on the Al$_{0.7}$Ga$_{0.3}$Sb buffer layer. The lowest sheet carrier concentration achieved, for the sample grown with a V/III flux ratio of 4.4, was $3.6 \times 10^{12} \text{ cm}^{-2}$. In the sample with the lowest sheet carrier concentration, the hole mobility was as high as $420 \text{ cm}^2/\text{V-s}$ and better than the GaSb QW structure, which has about one third the sheet carrier density of the In$_{0.3}$Ga$_{0.7}$Sb QW structure. This shows that utilizing the Al$_{0.7}$Ga$_{0.3}$Sb buffer layer in the modulation-doped pseudomorphic
In$_{0.3}$Ga$_{0.7}$Sb heterostructure did lead to higher hole mobility at similar or even higher sheet carrier concentration. The improvement of the hole mobility might be attributed to the smoother interface by adding Ga into the AlSb buffer layer.

Fig. 3.8 Sheet carrier concentration vs. hole mobility of FETs with strained In$_{0.3}$Ga$_{0.7}$Sb and GaSb quantum well channels grown on AlSb and Al$_{0.7}$Ga$_{0.3}$Sb buffer layers. Different V/III flux ratios were used for the growth of the channel layer. Filled triangles represent strained In$_{0.3}$Ga$_{0.7}$Sb quantum well channel on AlSb buffer, empty triangles represent GaSb quantum well channel on AlSb buffer and filled squares represent strained In$_{0.3}$Ga$_{0.7}$Sb quantum well channel on Al$_{0.7}$Ga$_{0.3}$Sb buffer.

In order to figure out the cause of increased intrinsic carrier concentration, the resistivity of the Al$_{0.7}$Ga$_{0.3}$Sb buffer layer was measured. As a result, the resistivity of Al$_{0.7}$Ga$_{0.3}$Sb was approximately 1 $\Omega$-cm, which was much smaller than that of AlSb. Therefore, the parallel
conduction of the Al\textsubscript{0.7}Ga\textsubscript{0.7}Sb buffer layer interfered with the Hall measurement and increased the measured carrier concentration. In order to reduce the parallel conduction in the Al\textsubscript{0.7}Ga\textsubscript{0.7}Sb buffer layer, a structure as described in Figure 3.3 was grown except that a 10 nm Al\textsubscript{0.7}Ga\textsubscript{0.3}Sb layer was inserted between the In\textsubscript{0.3}Ga\textsubscript{0.7}Sb channel and the AlSb buffer layer. By adding Ga into AlSb, it is expected that the improved Al\textsubscript{0.7}Ga\textsubscript{0.3}Sb surface morphology may lead to a better interface, the same as the result addressed in Chapter 2. The RMS surface roughness of 800 nm-thick GaSb and AlSb grown on InP substrates was measured to be roughly 0.4 and 0.6 nm, respectively. Therefore, the thin Al\textsubscript{0.7}Ga\textsubscript{0.3}Sb layer is expected to provide a smoother heterojunction interface than AlSb to reduce the scattering. However, unlike the case of n-type In\textsubscript{0.3}Ga\textsubscript{0.7}Sb, Al\textsubscript{0.7}Ga\textsubscript{0.3}Sb instead of pure GaSb is used to minimize the parallel conduction to the In\textsubscript{0.3}Ga\textsubscript{0.7}Sb channel. From the previous experiment results, the thickness of the Al\textsubscript{0.7}Ga\textsubscript{0.3}Sb layer has to be small since it is not semi-insulating as pure AlSb. The V/III BEP ratio for the growth of In\textsubscript{0.3}Ga\textsubscript{0.7}Sb QWs was fixed at 4.4 and the growth temperature was 430 °C.

The hole mobility and sheet carrier concentration of QW structures under different delta-doping conditions are displayed in Figure 3.9. The data at the carrier concentration of 7.5x10\textsuperscript{11} cm\textsuperscript{-2} corresponds to an undoped structure without carbon delta-doping and shows a hole mobility of 600 cm\textsuperscript{2}/V-s. The hole mobility enhancement is observed by comparing with the GaSb QW structure addressed above. It is evident that the hole mobility is improved by incorporating indium and compressive strain into the In\textsubscript{0.3}Ga\textsubscript{0.7}Sb QW. The insertion of an Al\textsubscript{0.7}Ga\textsubscript{0.3}Sb layer between the AlSb buffer and the In\textsubscript{0.3}Ga\textsubscript{0.7}Sb QW improves the hole mobility by 1.5 times at the same carrier concentration over the strained In\textsubscript{0.3}Ga\textsubscript{0.7}Sb QW structure grown directly on AlSb. The hole mobility is close to, but not as high as, that reported in [18], which showed a hole mobility of \( \sim 800 \text{ cm}^2/\text{V-s} \) at a carrier concentration of 1.1x10\textsuperscript{12} cm\textsuperscript{-2}. One possible reason is that the growth conditions of the In\textsubscript{0.3}Ga\textsubscript{0.7}Sb channel, such as the growth
temperature and growth rate, were not optimized. Another possible reason could be the parallel conduction of the thin Al$_{0.7}$Ga$_{0.3}$Sb layer because of its finite resistivity. Its composition and thickness need further optimization to minimize the parallel conduction.

![Graph](image)

Fig. 3.9 Sheet carrier concentration vs. hole mobility of strained In$_{0.3}$Ga$_{0.7}$Sb quantum wells. Filled squares represent results of In$_{0.3}$Ga$_{0.7}$Sb quantum wells grown on 10 nm Al$_{0.7}$Ga$_{0.3}$Sb/800 nm AlSb buffer layers under different carbon delta-doping conditions. The filled triangle represents the result of a In$_{0.3}$Ga$_{0.7}$Sb quantum well grown on AlSb buffer without carbon delta-doping, which has similar mobility to the GaSb QW structure.

### 3.3.3 Carbon Delta-Doping

The effect of carbon delta-doping was investigated using the structure discussed above, with a 10 nm-thick Al$_{0.7}$Ga$_{0.3}$Sb layer between the In$_{0.3}$Ga$_{0.7}$Sb QW and the AlSb buffer. The carbon delta-doping layer was located at 15 nm above the channel layer, and the doping concentration
was modulated by controlling the flow time of CBr$_4$. The flow of CBr$_4$ was controlled by a pressure controller and the pressure was set at 0.075 torr. The corresponding hole mobility and sheet carrier concentration are shown in Figure 3.9. The results show that carbon delta-doping is effective, increasing the sheet carrier concentration from $7.5 \times 10^{11}$ cm$^{-2}$ to about $1 \times 10^{12}$ cm$^{-2}$, with one second of CBr$_4$ flow. The hole mobility decreases from 600 cm$^2$/V-s to 540 cm$^2$/V-s with increasing sheet carrier concentration. When the CBr$_4$ flow time is increased to three seconds, the carrier concentration increases to $4.1 \times 10^{12}$ cm$^{-2}$, and the hole mobility decreases to 400 cm$^2$/V-s. The results suggest that the doping concentration of the In$_{0.3}$Ga$_{0.7}$Sb channel can be easily controlled using CBr$_4$ as the delta-doping source.
Chapter 4 TYPE-I/II DHBT WITH HIGH CURRENT GAIN

4.1 Introduction

As discussed in Section 1.5, modifying the emitter-base junction into type-I band alignment could eliminate the current blocking and hence improve the current gain. It has been demonstrated that the current gain could be effectively improved by using a larger-bandgap compound, such as In$_x$Al$_y$Ga$_{1-x-y}$As [43], [44-46] or In$_x$Ga$_{1-x}$P [47], as the emitter. The current gain could be greatly improved from below 30 to 150 when the aluminum composition in In$_x$Al$_{1-x}$P was increased from 0 to 0.25, with a 20 nm uniform carbon-doped GaAsSb base (base resistance 1800 $\Omega$/sq) [46].

Among these materials, the advantage of phosphide compounds over arsenide compounds is the process compatibility with the original InP emitter. An arsenic compound as the emitter could decrease etching selectivity over the GaAsSb base, which is critical especially for the DHBT with thin base layer. For the selection between In$_x$Al$_{1-x}$P and In$_x$Ga$_{1-x}$P, because In$_x$Ga$_{1-x}$P has smaller bandgap than In$_x$Al$_{1-x}$P at the same indium ratio, to achieve the same conduction band energy, indium composition has to be lower in In$_x$Ga$_{1-x}$P than in In$_x$Al$_{1-x}$P. This would lead to a higher strain in the pseudomorphic In$_x$Ga$_{1-x}$P layer. Also, because the capacity of the gallium effusion cell in the GSMBE system is much larger than that of the aluminum effusion cell, it is difficult to ramp the gallium effusion cell temperature quickly to achieve uniform compositional grading. Therefore, In$_x$Al$_{1-x}$P is utilized as the emitter in the following experiments.

4.1.1 Conduction Band Energy of In$_{1-x}$Al$_x$P

Before using In$_{1-x}$Al$_x$P to replace InP as the emitter, it is necessary to determine the required aluminum composition in the In$_{1-x}$Al$_x$P compound. First, to eliminate conduction band discontinuity between In$_{1-x}$Al$_x$P and GaAs$_{0.5}$Sb$_{0.5}$, the conduction band energy of In$_{1-x}$Al$_x$P was
calculated. Model-solid theory and parameters of AlP and InP shown in Table 4.1 were used for the calculations. Strain effect on band-edge energies was also included in the calculation to enhance the accuracy [41]. As shown in Figure 4.1, the conduction band edge increment (ΔEc) of In1-xAlxP in the range of 0<x<0.2 is almost linear. If the target value is 180 meV, the empirical conduction band discontinuity between GaAsSb and InP addressed in the last section, the required aluminum ratio in In1-xAlxP would be ~ 0.1.

Table 4.1 Material parameters for the calculation of conduction band edge of In1-xAlxP.

<table>
<thead>
<tr>
<th></th>
<th>Lattice constant (Å)</th>
<th>Ex,av (eV)</th>
<th>D (eV)</th>
<th>Ev (eV)</th>
<th>a,v (eV)</th>
<th>a,c (eV)</th>
<th>C_{11} (10^{11} \text{ dyn/cm}^2)</th>
<th>C_{12} (10^{11} \text{ dyn/cm}^2)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AlP</td>
<td>5.4670</td>
<td>-8.09</td>
<td>0.06</td>
<td>-8.070</td>
<td>3.15</td>
<td>-5.54</td>
<td>13.3</td>
<td>6.3</td>
</tr>
<tr>
<td>InP</td>
<td>5.8687</td>
<td>-7.04</td>
<td>0.11</td>
<td>-7.003</td>
<td>1.27</td>
<td>-5.04</td>
<td>10.11</td>
<td>5.61</td>
</tr>
</tbody>
</table>

Fig. 4.1 Conduction band edge increment of In_{1-x}Al_{x}P over InP at different aluminum ratio x.
Since In$_{1-x}$Al$_x$P always has a smaller lattice constant than InP, the growth of In$_{1-x}$Al$_x$P on InP would be pseudomorphic. The critical thickness of In$_{1-x}$Al$_x$P needs to be calculated in order to design a suitable layer thickness. The Matthews and Blakeslee model was used for the calculation [12]. Figure 4.2 shows the critical thickness and tensile strain of In$_{1-x}$Al$_x$P grown on InP with different aluminum composition x. The critical thickness of In$_{0.9}$Al$_{0.1}$P on InP is to be 10 nm. To reduce the tensile strain and increase allowable emitter layer thickness, the aluminum composition was reduced gradually in the growth direction. If the composition of the emitter is In$_{0.9}$Al$_{0.1}$P at the E/B junction and then linearly graded to InP, the average tensile strain in this graded emitter is about 0.35%, which corresponds to a critical thickness of 25 nm.

Fig. 4.2 Critical thickness and tensile strain of In$_{1-x}$Al$_x$P grown on InP with different aluminum ratio x.
4.1.2 Ternary Graded GaAs$_x$Sb$_{1-x}$ Base

Grading composition in the base toward the emitter with larger bandgap energy in the growth direction could establish a built-in electric field in the base [48]. Assuming the base composition is linearly graded, the electric field ($\varepsilon_B$) is uniform and defined as

$$E_{LH}(k_x) = -P_e \frac{\hbar^2 \gamma_i}{2m_0} k_x^2 - \left[ \left( \frac{h^2 \gamma_2}{2m_0} k_x^2 \right)^2 + 3 \left( \frac{h^2 \gamma_2}{2m_0} k_x^4 \right) \right]^{\frac{1}{2}}$$  (4.1)

$E_{g,\text{base}}$ is the bandgap energy difference between the base layer composition at the emitter end and the collector end, $q$ is electron charge, and $X_B$ is the base layer thickness. This field, pointing from the collector toward the emitter end, generates an acceleration action upon electrons flowing through the base layer. It reduces the base transit time and hence enhances the current gain, which is defined as

$$\beta = \frac{\tau_n \text{ (minority carrier lifetime)}}{\tau_b \text{ (base transit time)}}$$  (4.2)

Grading from In$_{0.2}$Ga$_{0.8}$As$_{0.7}$Sb$_{0.3}$ to GaAs$_{0.5}$Sb$_{0.5}$ in the growth direction was utilized to achieve bandgap grading [25]. However, it was rather difficult for the growth control of quaternary grading, which needs to adjust at least two effusion cells during the growth if optimized growth conditions are required. Also, In$_{0.2}$Ga$_{0.8}$As$_{0.7}$Sb$_{0.3}$ compound has a lower hole mobility than GaAs$_{0.5}$Sb$_{0.5}$. At the same doping concentration of approximately $7 \times 10^{19}$ cm$^{-3}$, the hole mobility of In$_{0.2}$Ga$_{0.8}$As$_{0.7}$Sb$_{0.3}$ (24 cm$^2$/V-s) is 30% lower than that of GaAs$_{0.5}$Sb$_{0.5}$ (34 cm$^2$/V-s), measured from calibration samples grown by the GSMBE. Therefore, incorporation of In$_x$Ga$_{1-x}$As$_y$Sb$_{1-y}$ in the graded base would inevitably increase the base resistance. The same issue also happens to the graded Al$_x$Ga$_{1-x}$As$_y$Sb$_{1-y}$-GaAs$_{0.5}$Sb$_{0.5}$ base, because the incorporation of aluminum would lower the carrier mobility. In order to make the base resistance as low as
possible, strain-balanced ternary graded GaAs$_x$Sb$_{1-x}$ may be a better solution, although it is not easy to control two group V elements at the same time during GSMBE.

In the following experiments, the base layer was graded from GaAs$_{0.4}$Sb$_{0.6}$ on the collector side to GaAs$_{0.6}$Sb$_{0.4}$ on the emitter side. The bandgap energy difference is approximately 135-140 meV between GaAs$_{0.6}$Sb$_{0.4}$ and GaAs$_{0.4}$Sb$_{0.6}$ [49,50], and the corresponding electric field is $5.4 - 5.6 \times 10^4$ V/cm, assuming the electric field is uniform and the base layer thickness is 25 nm.

4.2 Type-I/II InAlP/GaAsSb DHBT

4.2.1 DHBT Structures

The growth parameters of the GaAs$_{0.6}$Sb$_{0.4}$ and GaAs$_{0.4}$Sb$_{0.6}$, which correspond to the base compounds at the emitter/base and base/collector junction, respectively, were calibrated using a strained multiple quantum well (MQW) structure shown in Figure 4.3. The difference between these two structures was the antimony flux used for the growth of GaAs$_{0.6}$Sb$_{0.4}$ and GaAs$_{0.4}$Sb$_{0.6}$. The adjustment of the antimony flux was based on the results of [51] to achieve designed compositions. The HRXRD curves of both structures are shown in Figure 4.4. Because of the lattice mismatch to the InP, the fringes of the GaAs$_{0.6}$Sb$_{0.4}$ MQW curve lie on the tensile side, while those of the GaAs$_{0.4}$Sb$_{0.6}$ MQW lie on the compressive side. From the theoretical fitting using RADS Mercury software, the $x$ value of each MQW is approximately 0.41 and 0.63 for GaAs$_{0.6}$Sb$_{0.4}$ and GaAs$_{0.4}$Sb$_{0.6}$, respectively. The antimony flux settings for these compounds could be utilized as a reference for the growth of graded GaAs$_x$Sb$_{1-x}$ base layers in DHBT structures.
Layer | Thickness (Å)
--- | ---
InP | 500
GaAs\(_x\)Sb\(_{1-x}\) | 100
InP | 100
InP buffer | 1000

Fig. 4.3 Multiple quantum well structure used for the calibration of GaAs\(_x\)Sb\(_{1-x}\) (\(x = 0.4\) or 0.6).

Fig. 4.4 HRXRD curves of GaAs\(_{0.6}\)Sb\(_{0.4}\) and GaAs\(_{0.4}\)Sb\(_{0.6}\) multiple quantum well structures shown in Figure 4.3.

After the calibration of the base layer composition, a few type-I/II InAlP/GaAsSb DHBT structures were grown. The DHBT growth was also carried out in GSMBE with the same growth procedures described in Section 2.2. Growth temperatures were \(T_d = 35\) for all layers except the graded GaAsSb base layer, which was grown at \(T_d = 55\). The graded GaAsSb base was realized...
by adjusting the valve position and hence the flux of the antimony source, while the arsine flow was kept the same. Although it is necessary to adjust both arsine flow and antimony flux to achieve optimized growth of GaAs<sub>x</sub>Sb<sub>1-x</sub> at different compositions, arsine flow was fixed during the growth because the arsine flow could not respond and stabilize immediately when the setting changed.

The DHBT layer structures (GS1486 & 1494) are shown in Table 4.2. The difference between GS1486 and 1494 is in the thickness of base and collector, which offers a trade-off between <i>f_T</i> and <i>f_{MAX}</i>. Thicker base and collector would increase carrier transit time through these layers and hence decrease <i>f_T</i>. On the other hand, thicker base and collector layers lower the base resistance (R<sub>b</sub>) and base-collector capacitance C<sub>bc</sub>, and increase <i>f_{MAX}</i>. The reference graded (In)GaAsSb base device structure shown in Table 4.3 has high frequency parameters <i>f_T</i> and <i>f_{MAX}</i> of 620 and 180 GHz, respectively [51]. It is listed as a comparison to show the potential of newly grown type-I/II DHBTs.

<table>
<thead>
<tr>
<th>Layer</th>
<th>GS1486 Thickness (Å)</th>
<th>GS1494 Thickness (Å)</th>
<th>Doping concentration (cm&lt;sup&gt;-3&lt;/sup&gt;)</th>
</tr>
</thead>
<tbody>
<tr>
<td>InGaAs:Si</td>
<td>400</td>
<td>650</td>
<td>N = 3E19</td>
</tr>
<tr>
<td>InP:Si (n+)</td>
<td>300</td>
<td>350</td>
<td>N = 1E19</td>
</tr>
<tr>
<td>InP:Si (n-)</td>
<td>75</td>
<td>175</td>
<td>N = 3E18</td>
</tr>
<tr>
<td>In&lt;sub&gt;0.9&lt;/sub&gt;Al&lt;sub&gt;0.1&lt;/sub&gt;P:Si (n--)</td>
<td>175</td>
<td>175</td>
<td>N = 7E17</td>
</tr>
<tr>
<td>GaAs&lt;sub&gt;x&lt;/sub&gt;Sb&lt;sub&gt;1-x&lt;/sub&gt;:C (x = 0.4→0.6)</td>
<td>180</td>
<td>300</td>
<td>P = 7E19</td>
</tr>
<tr>
<td>InP</td>
<td>1200</td>
<td>1200</td>
<td>UID (~5E16)</td>
</tr>
<tr>
<td>InGaAs:Si</td>
<td>650</td>
<td>650</td>
<td>N = 8E19</td>
</tr>
<tr>
<td>InP:Si</td>
<td>3500</td>
<td>3500</td>
<td>N</td>
</tr>
</tbody>
</table>
Table 4.3 Reference structures (GS620, type-II DHBT).

<table>
<thead>
<tr>
<th>Layer</th>
<th>Thickness (Å)</th>
<th>Doping concentration</th>
</tr>
</thead>
<tbody>
<tr>
<td>InGaAs:Si</td>
<td>400</td>
<td>N = 3E19</td>
</tr>
<tr>
<td>InP:Si (n+)</td>
<td>200</td>
<td>N = 1E19</td>
</tr>
<tr>
<td>InP:Si (n-)</td>
<td>200</td>
<td>N = 3E18</td>
</tr>
<tr>
<td>(In)GaAsSb:C</td>
<td>200</td>
<td>P = 7E19</td>
</tr>
<tr>
<td>InP</td>
<td>650</td>
<td>UID (~2E16)</td>
</tr>
<tr>
<td>InGaAs:Si</td>
<td>250</td>
<td>N = 8E19</td>
</tr>
<tr>
<td>InP:Si</td>
<td>3500</td>
<td>N+</td>
</tr>
</tbody>
</table>

4.2.2 DC Characteristics

Large devices were fabricated with a standard mesa process as detailed in [52], with the emitter mesa of 50x50 µm². The DC performance, as shown in Figure 4.5, was measured with a HP-4145B semiconductor parameter analyzer. Figure 4.5(a) shows the common-emitter I-V family curves of the large device results of GS1486. The family curves show a breakdown slightly larger than 2 V, which could be higher in a sub-micron device because of the current crowding in large devices [48]. The ideality factors of base and collector current shown in Figure 4.5(b) are 2.0 and 1.03, respectively. The high base current ideality factor could indicate the base/emitter junction was not abrupt, which could be due to the insufficient cryogenic pumping of the growth chamber. Nonetheless, from the Gummel plot, the device demonstrated a high current gain, which is larger than 120 at the base voltage of 0.8 V.
Fig. 4.5 (a) Common-emitter I-V family curves and (b) Gummel plot of GS1486 with an emitter mesa of 50x50 µm.
To compare the device performance of type-I/II and type-II DHBTs, some DC characteristics are summarized in Table 4.4. Since the base thickness is not identical for all DHBT devices, current gain/base resistance ratio ($\beta/R_{sb}$) was used to compare the performance between HBTs with different base thicknesses. The ideal HBT requires higher current gain and lower base resistance to achieve higher $f_{\text{MAX}}$, which would hence lead to a higher $\beta/R_{sb}$.

Table 4.4 DC characteristics summary emitter size = 50x50 µm.

<table>
<thead>
<tr>
<th>Device</th>
<th>Type-I/II DHBT</th>
<th>Type-II DHBT</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>GS1486</td>
<td>GS1494</td>
</tr>
<tr>
<td>Base (Å)</td>
<td>180</td>
<td>300</td>
</tr>
<tr>
<td>$\beta$</td>
<td>120</td>
<td>45</td>
</tr>
<tr>
<td>Base sheet resistance $R_{sb}$ (Ω/sq)</td>
<td>2700</td>
<td>1000</td>
</tr>
<tr>
<td>Normalized $R_{sb}$ (base thickness = 200 Å)</td>
<td>2430</td>
<td>1500</td>
</tr>
<tr>
<td>$\beta/R_{sb}$</td>
<td>0.045</td>
<td>0.045</td>
</tr>
<tr>
<td>Ideality factor $\eta$ (I$_B$)</td>
<td>2.0</td>
<td>1.6</td>
</tr>
<tr>
<td>Ideality factor $\eta$ (I$_C$)</td>
<td>1.03</td>
<td>1.02</td>
</tr>
</tbody>
</table>

From Table 4.4 it is clear that the DHBT structures with type-I E/B junction have about 50% higher $\beta/R_{sb}$ than the reference device. This indicates that the type-I/II DHBT could achieve higher current gain with similar or lower base resistance compared with the type-II DHBT, the reference structure. Also, from the normalized base resistance, it is also evident that the ternary graded GaAsSb base could lead to lower base resistance than quaternary graded InGaAsSb base, while providing a higher current gain. Although the RF performance is not directly related with these DC characteristics, it still shows the potential of the type-I/II InAlP/GaAsSb DHBT for higher frequency operation.
4.2.3 RF Characteristics

To evaluate the RF performance of type-I/II DHBT devices, another sample with the same structure as that of GS1486, shown in Table 4.2, was grown and fabricated into sub-micron devices. The processes were done with self-aligned e-beam lithography by Donald Cheng of Professor Milton Feng’s group [52]. The emitter, base metal, collector mesa and collector metal were defined with electron beam lithography, which could minimize extrinsic base-collector junction capacitance and extrinsic collector resistance.

The RF measurements of the sub-micron type-I/II DHBTs were conducted in an HP 8510c network analyzer to measure the S-parameters, from 0.5 to 50 GHz. By extrapolating -20 dB/decade lines on the current gain ($h_{21}$) and Mason’s unilateral power gain (U) to 0 dB, the RF performances of the DHBT, $f_T$ and $f_{MAX}$, could be extracted respectively. The emitter size of the measured sub-micron device is 0.48x8 $\mu m^2$. As shown in Figure 4.6, the extrapolation shows that the $f_T$ and $f_{MAX}$ are 350 and 210 GHz, respectively. These numbers cannot compare with those of the reference graded (In)GaAsSb base type-II DHBTs. Although the excellent DC characteristics of type-I/II DHBTs look promising, they do not reflect on the RF performances. One possible reason is the growth control of the ternary graded GaAsSb base layer. Although the valve position of the antimony effusion cell and hence the antimony flux were decreased during the epitaxial growth of the base layer, it might not achieve uniform compositional grading and the designed built-in electric field in the base layer. Therefore, it may not provide the expected acceleration for the electron flows through the base layer, which was found to be important to improve the RF performance over 500 GHz [51].
Fig. 4.6 Current gain $h_{21}$, Mason’s unilateral gain $U$ and extrapolations for $0.48 \times 8 \, \mu m^2$ type-I/II DHBT. Inset shows the extrapolated cutoff frequencies as a function of the frequency, where the -20 dB/decade extrapolation was made.
Chapter 5 CONCLUSIONS AND FUTURE WORKS

5.1 Summary of Works

5.1.1 Antimonide-Based n- and p-Channel FETs

III/V compound semiconductors have great potential for high-speed applications in the future, due to their high intrinsic carrier mobility. In this work, n- and p-channel FETs were successfully grown metamorphically on InP substrates using gas source molecular beam epitaxy. InAs$_{0.8}$Sb$_{0.2}$ and strained In$_{0.3}$Ga$_{0.7}$Sb QW were utilized as the n- and p-channel material, respectively. An AlSb/AlAs$_{0.5}$Sb$_{0.5}$ composite buffer layer was grown on the InP substrates prior to the channel growth, which was found to effectively improve the film quality and carrier mobility of the channel.

For the n-type FET, the electron mobility of InAs$_{0.8}$Sb$_{0.2}$ could be improved to 18000 cm$^2$/V-s by tuning the growth temperature, layer thickness and the interface between the channel and the buffer layer. HEMT devices utilizing InAs$_{0.8}$Sb$_{0.2}$ as the channel were grown and fabricated, which showed a high $g_m$ of 350 mS/mm at the gate length of 6 µm. These results indicate the great potential for the high-speed application.

For the p-type FET, the hole mobility in pseudomorphic In$_{0.3}$Ga$_{0.7}$Sb QW structures was studied. Similar to the n-type InAs$_{0.8}$Sb$_{0.2}$, the hole mobility of the In$_{0.3}$Ga$_{0.7}$Sb QW channel could be enhanced to 600 cm$^2$/V-s at a hole concentration of 7.5×10$^{11}$ cm$^{-2}$ by improving the interface morphology between the channel and the buffer layer. P-type delta-doping is found to be effective using CBr$_4$ as the doping source, and controlling the flow time of CBr$_4$ can modulate the doping concentration easily.

5.1.2 Type-I/II DHBTs With High Current Gain

For GaAsSb-based DHBTs, type-I/II DHBTs with ternary graded base and In$_{1-x}$Al$_x$P as the
emitter to form type-I emitter/base junction could achieve high DC current gain while maintaining low base resistance. The current gain/base sheet resistance ratio ($\beta/R_{sb}$) of type-I/II DHBTs is 50% higher than that of the reference graded (In)GaAsSb base type-II DHBTs. The RF performance of type-I/II DHBTs, however, does not match the excellent DC characteristics and requires further studies. The measured $f_T$ and $f_{MAX}$ are 350 and 210 GHz, respectively.

5.2 Future Works

5.2.1 Antimonide-Based n- and p-Channel FETs

Currently, the major issue for the fabrication of n-type In$_{0.8}$Sb$_{0.2}$ and p-type In$_{0.3}$Ga$_{0.7}$Sb FETs is to process the buffer layer, AlSb. Both Al and Sb are extremely chemically active and pure AlSb would oxidize very easily even under a short exposure to the air. Figure 5.1 shows SEM images of an InGaSb FET after the isolation etch. The structure was etched with HF-based solution down to the AlSb buffer layer. As a result, the AlSb buffer was exposed to the air and oxidized. For the fabrication of RF devices, isolation etching is required, so the buffer layer has to be able to undergo lithography without generating defects and leakage paths. A few methods may help to solve this issue. First, insert an etching stop layer in the AlSb buffer layer to protect the exposed surface after wet etching. However, finding a compound with good etching selectivity over AlSb is not easy since very few etching solutions could etch AlSb fast but others slowly. Second, dry etching the whole buffer layer away and down to the semi-insulating InP substrate could be another approach. AlSb may get less oxidized during the dry etching, compared with wet etching. Also, using ternary AlGaSb or AlAsSb instead of pure AlSb as the buffer may help to inhibit the oxidization.
In addition, using a buffer layer with lower fraction of AlSb or compounds other than AlSb-based materials could reduce the difficulties of the process. However, for p-type FET, the selection of compounds as the buffer layer is more limited than for n-type FET because of the low mobility of the p-type channel. Compounds with a high fraction of InAs, which has a lattice constant close to AlSb, would not be suitable because of the high electrical conductivity of InAs.

For an n-type InAs$_x$Sb$_{1-x}$ FET, increasing the channel layer thickness could be helpful, as suggested by the results of Chapter 2. Since the electron gas would be located in the channel/top barrier layer, a thicker channel could keep electrons away from the InAs$_x$Sb$_{1-x}$/AlSb interface and hence the interface scattering. Also, increasing the fraction of InSb could also help since InSb has a higher electron mobility and the alloy scattering is not prominent when the fraction of InSb < 0.5, as shown in Fig.1.3. However, the channel layer thickness needs to be thin since there is strain in the channel.

For a p-type FET, using pure GaSb as the channel with different buffer layers could help to improve the hole mobility since it will eliminate the alloy scattering. If GaSb is used, AlAs$_x$Sb$_{1-x}$ could be used as the buffer. For example, to generate the same amount of compressive strain in GaSb/AlAs$_x$Sb$_{1-x}$ as in In$_x$Ga$_{1-x}$Sb/AlSb, $x$ would be ~ 0.25 for AlAs$_x$Sb$_{1-x}$, which greatly reduces
the fraction of AlSb in the buffer layer.

5.2.2 Type-I/II DHBTs

To further improve the DC and RF performance of type-I/II DHBTs, several approaches could be explored. The growth control of the ternary base addressed at the end of Chapter 4 is one of these. If the ternary graded GaAsSb is difficult to achieve in the GSMBE system, it is necessary to use a quaternary graded base, including (Al)GaAsSb and (In)GaAsSb. Also, the aluminum composition in the In$_{1-x}$Al$_x$P emitter could increase to enhance the current gain, which is important for driving capability.

The biggest issue with improving the operation frequency of a GaAsSb-based DHBT beyond THz, especially the $f_{\text{MAX}}$, is its high base resistance. GaAsSb suffers strong alloy scattering, and the Hall mobility as a function of $x$ in heavily carbon-doped ($10^{19}$ - $10^{20}$ cm$^{-3}$) GaAs$_x$Sb$_{1-x}$ (grown by MOVPE on InP) shows strong alloy scattering [53]. To avoid alloy scattering, a GaAs/GaSb super-lattice could be a method to achieve high hole mobility. However, the interfaces generated between GaAs/GaSb quantum wells could enhance the scattering when electrons travel through the base and increase the recombination rate. In addition, strain-induced hole mobility improvement could be another way to reduce the base resistance. This could be achieved by using a GaAs$_x$Sb$_{1-x}$ with $x$ smaller than 0.5 as the layer. This method is valid because the base thickness has to be thin for the high-speed application, which allows the incorporation of large strain in the base. The hole mobility enhancement is in the lateral direction, which could help improve the carrier transport from the base metal to the active region and reduce the base resistance. This approach could have another advantage in lowering the conduction band energy of the base to achieve type-I conduction band alignment at the E/B junction.
REFERENCES


1982.


