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**Bandgap Narrowing in Heavily Doped Base Regions of
Si/Si_{1-x}Ge_x/Si Heterojunction Bipolar Transistors**

by Brendon M. Manning, B.Eng

A thesis submitted to the Faculty of Graduate Studies and Research in partial
fulfillment of the requirements for the degree

Master of Engineering

Ottawa-Carleton Institute for Electrical Engineering
Department of Electronics
Faculty of Engineering
Carleton University
November 1993

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**“Bandgap Narrowing in Heavily Doped Base Regions of
Si/Si_{1-x}Ge_x/Si Heterojunction Bipolar Transistors”**

submitted by Brendon M. Manning in partial fulfillment of the requirements for the degree of Master of Engineering

Prof. N.G. Tarr
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Carleton University
November 1993

Abstract

Determination of the total bandgap narrowing in heavily doped p-type strained $\text{Si}_{1-x}\text{Ge}_x$ alloy layers forming the base region of heterojunction bipolar transistors (HBTs) is reported. Strain due to pseudomorphic growth of the alloy layers and effects associated with high doping levels were considered to be the potentially significant contributors to the reduction in bandgap energy. HBTs were fabricated from strained $\text{Si}_{1-x}\text{Ge}_x$ layers grown by molecular beam epitaxy. The bandgap was determined from the temperature dependence of the collector saturation current. The recent theory of Jain and Roulston for the bandgap narrowing resulting from heavy doping was applied to the $\text{Si}_{1-x}\text{Ge}_x/\text{Si}$ heterosystem, with allowance for strain effects. The theory was found to agree reasonably well with the measured bandgap narrowing once second order effects and process variations were accounted for. It was found that the additional bandgap narrowing due to heavy doping was significant and should be accounted for when designing and optimizing the vertical profile of $\text{Si}_{1-x}\text{Ge}_x/\text{Si}$ HBTs.

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Table of Contents

	<u>page</u>
1.0 Introduction	1
1.1 Introduction to Heterojunction Bipolar Transistors	1
1.2 Introduction to Silicon-Based Heterojunctions and $\text{Si}_{1-x}\text{Ge}_x$ Alloys	5
1.3 Introduction to Dopant Induced Bandgap Narrowing	10
1.4 Outline of Thesis	11
 2.0 Bandgap Narrowing in Heavily Doped $\text{Si}_{1-x}\text{Ge}_x$ Alloys	 12
2.1 Introduction	12
2.2 Theoretical Calculation of the Energy Bandgap	12
2.2.1 Bandgap of Strained $\text{Si}_{1-x}\text{Ge}_x$ on $\langle 001 \rangle$ Si Substrate	12
2.2.2 Dopant Dependent Bandgap Narrowing	14
2.3 Theoretical Basis for Bandgap Extraction from Collector Saturation Current	19
 3.0 $\text{Si}_{1-x}\text{Ge}_x$ Strained Layer Growth and HBT Fabrication	 25
3.1 Introduction	25
3.2 $\text{Si}_{1-x}\text{Ge}_x$ Strained Layer Growth and Characteristics	26
3.3 Process Development and Si/ $\text{Si}_{1-x}\text{Ge}_x$ /Si HBT Fabrication	29
3.3.1 Process Issues for HBT Fabrication	29
3.3.2 Isolation of $\text{Si}_{1-x}\text{Ge}_x$ HBTs	31
3.3.3 Passivation of $\text{Si}_{1-x}\text{Ge}_x$ Strained Layers	33
3.3.4 Contact Metallization for the HBTs Under Study	34
 4.0 $\text{Si}_{1-x}\text{Ge}_x$ HBT Characteristics and Bandgap Measurements	 36
4.1 Introduction	36
4.2 Room Temperature Characteristics of the $\text{Si}_{1-x}\text{Ge}_x$ HBTs	37
4.3 Determination of Bandgap Narrowing in the p^+ $\text{Si}_{1-x}\text{Ge}_x$ Base	45

	page
5.0 Discussion	48
5.1 Introduction	48
5.2 Issues Concerning the Theoretical Calculation of Bandgap Narrowing	48
5.3 Issues Concerning the Analysis of the Bandgap Narrowing Measurements	52
5.4 Processing Considerations	64
5.5 Comparison of Theory and Measurements to an Accepted Model	66
5.6 Summary of Discussion	69
 6.0 Conclusions and Recommendations	 71
6.1 Conclusions	71
6.2 Recommendations	72
 References	 74
Appendix A	A1
Appendix B	B1

List of Figures

	<u>page</u>
Figure 1.1 Bandgap and Lattice Constants of Common Semiconductor Materials[6]	6
Figure 2.1 Indirect Bandgap of Coherently Strained $\text{Si}_{1-x}\text{Ge}_x$ on $\langle 001 \rangle$ Si[14]	14
Figure 2.2 Dopant Induced Bandgap Narrowing in p-type $\text{Si}_{1-x}\text{Ge}_x$ Layers	18
Figure 2.3 Total Bandgap Narrowing in p-type $\text{Si}_{1-x}\text{Ge}_x$ Layers	19
Figure 2.4 Band Diagram for Analysis of Strained $\text{Si}_{1-x}\text{Ge}_x$ Base Layer	20
Figure 3.1 A Double Mesa Isolated HBT used for Bandgap Measurements	25
Figure 3.2 SIMS Profile of Sample 1034	28
Figure 3.3 Process Flow Used for HBT Fabrication	30
Figure 3.4 SEM Micrograph of Si/ $\text{Si}_{1-x}\text{Ge}_x$ Layers After EDP Etch	33
Figure 3.5 SEM Micrograph of a Completed HBT	35
Figure 4.1 Gummel Plot for HBT from Wafer 1034(8% Ge) ($V_{CB}=0V$)	39
Figure 4.2 Gummel Plot for HBT from Wafer 1035(15% Ge) ($V_{CB}=0V$)	40
Figure 4.3 Gummel Plot for HBT from Wafer 1036(20% Ge) ($V_{CB}=0V$)	41
Figure 4.4 Collector Current as a Function of Ge Content ($T=20^\circ\text{C}$)	42
Figure 4.5 Transfer Characteristics for an HBT from Wafer 1035 (15% Ge)	43
Figure 4.6 I_C versus V_{BE} with V_{CB} Stepped for an HBT from Wafer 1034 (8% Ge)	44
Figure 4.7 Plot of I_{CO}/T^4 versus $1000/T$ for 1034(8%), 1035(15%), and 1036(20%)	46
Figure 4.8 Bandgap Narrowing vs. Germanium Content	47
Figure 5.1 Effect of Assuming no Valence Subband Splitting on the Theoretical Calculation of Bandgap Narrowing	51

Figure 5.2 The Effect of the Fermi–Dirac Correction on Bandgap Narrowing	53
Figure 5.3 $\mathfrak{F}_{1/2}(\eta_V)/\exp(\eta_V)$ and $1/(1+ye^{\eta_V})$ versus Temperature at Various Doping Levels	55
Figure 5.4 $[\mathfrak{F}_{1/2}(\eta_V)/\exp(\eta_V)]/(T/300)^{m_1}$ versus Temperature at Various Doping Levels ($m_1=0.696$, $N_V=1.04 \times 10^{19} \text{ cm}^{-3}$)	57
Figure 5.5 Fermi–Dirac Correction to Bandgap Narrowing, Plotted Against Temperature for Various Doping Levels, N_A	59
Figure 5.6 Effect of Modified Fermi–Dirac Correction on Bandgap Narrowing	60
Figure 5.7 The Combined Effects of the Fermi–Dirac Correction and the Correction for Subband Splitting	63
Figure 5.8 Examination of an Assumed Deviation in Doping Level From the Nominal Value	65
Figure 5.9 Fermi–Dirac Correction versus Doping Level at 300 K	67
Figure 5.10 Comparison of the Theory of Jain and Roulston[36] with the Formula Presented by Slotboom[32]	68

Chapter 1

Introduction

1.1 Introduction to Heterojunction Bipolar Transistors

The performance of a silicon bipolar transistor is restricted by conflicting design requirements put on the base region. A narrow base region is desired in order to minimize the transit time of minority carriers across the base for improved frequency response. However, the minimum base width is limited by the requirements on collector-emitter breakdown voltage and base resistance. In order to maintain acceptable punchthrough voltage and base resistance, there is a corresponding increase in the doping level required in the base as the base width is reduced. In order to maintain gain, the doping level in the base of a conventional homojunction bipolar transistor must be much lower than that of the emitter to limit the back-injection of carriers and to prevent tunneling currents[1]. The maximum doping level in the emitter for the typical "poly emitter" bipolar transistor is set by the solid solubility of the dopant, and is typically on the order of 10^{20} cm^{-3} . This puts a physical restriction on the gain, breakdown voltage, and frequency response of the transistor, requiring trade-offs between them. In the 1950's, it was proposed to get around this restriction through the use of heterojunctions[2].

A heterojunction is created through the juxtaposition of different semiconductor materials, each having distinct material and electrical properties. By intentionally altering the properties of the semiconductor material in the different regions of a transistor, in particular the energy bandgap, considerable improvement

in device performance may be obtained. The use of heterojunctions in bipolar transistors has been primarily to obtain an emitter region having a wider bandgap than the base region. The emitter injection efficiency limited current gain of a bipolar transistor, h_{FE} , is exponentially related to the difference between the energy bandgap of the emitter and the energy bandgap of the base, ΔE_g [3] but is only linearly

$$h_{FE} = \frac{n_E}{p_B} \frac{v_e}{v_h} \frac{N_{C_H} N_{V_H}}{N_{C_E} N_{V_E}} e^{\frac{\Delta E_g}{kT}} \quad (1.1)$$

dependent on the ratio of majority carrier concentrations, n_E and p_B , and the ratio of electron and hole velocities injected across the base-emitter junction, v_e and v_h . The bandgap difference may be used to maintain a desired level of gain while allowing the ratio of emitter to base doping levels (ie. majority carrier concentrations), to be altered, thus decoupling the restriction relating base doping to transistor gain. The additional degree of freedom gained through the use of heterojunctions in a bipolar transistor is the primary advantage to their use. Note the use of different materials implies different densities of states in the conduction and valence bands of the emitter and base ($N_{C,V}$).

There are two approaches to the formation of heterojunction bipolar transistors (HBTs) resulting in the desired offset between the energy bandgap of the base and emitter. The first employs a material with a wide bandgap as the emitter of the device, forming the base and collector from the same material. The wide bandgap emitter results in an increased potential barrier at the base-emitter junction, reducing the back injection of minority carriers into the emitter from the base at a given bias. The current gain is increased over that of a similarly doped homojunction transistor by the reduction of this component of base current. To first order, the collector current for a given base-emitter bias is not affected.

The second method employs a narrow band gap material to form the base of the transistor. The narrow bandgap in the base results in an increase in collector current over the homojunction case through a reduction in the potential barrier to minority carrier injection. The base current, to first order, is unaffected. The amount of collector current enhancement of the HBT over a homojunction bipolar transistor (BJT) is given in equation 1.2[4]. This applies for the case of uniform doping and

$$\frac{I_{C_{HBT}}}{I_{C_{BJT}}} = \frac{W_{BJT}}{W_{HBT}} \frac{N_{A_{BJT}}}{N_{A_{HBT}}} \frac{D_{B_{HBT}}}{D_{B_{BJT}}} e^{\left(\frac{E_{B_{BJT}} - E_{B_{HBT}}}{kT}\right)} \quad (1.2)$$

composition profiles, with no band discontinuities to block current flow. Note that $E_{G_{HBT}}$ and $E_{G_{BJT}}$ refer to the bandgap in the base region of the respective transistors. The exponential dependence of this increase in current on the bandgap difference then allows for a reduction in the base resistance of the HBT over the BJT, through increased base doping, N_A , while maintaining the desired collector current level.

The ability to alter the material composition also allows for gradual compositional changes through a device structure. For example, by varying the bandgap across a given region of a device, "built-in" electric fields can be created. In the case of a "narrow bandgap" base npn HBT, grading the composition across the base such that the bandgap is narrowest at the base-collector junction and widest at the base-emitter junction, creates a drift field that assists minority carrier electron transport across the base. This results in a reduction in base transit time, τ_b . For devices where τ_b is a dominant delay component, the common-emitter unity current gain frequency, f_T , is increased, giving faster operation. In the case of homojunction devices, any "built-in" fields present are the result of doping gradients. Adjusting these fields to improve f_T impacts directly on base resistance and punchthrough

whereas for an HBT the doping profile is not adjusted so these parameters are not affected.

Further benefits in frequency response are derived from the reduction in emitter charge storage due to the presence of the narrow bandgap in the base (as anticipated by the increase in collector current). The reduction in emitter charge storage, τ_e , is seen to be exponentially dependent on the bandgap reduction at the edge of the base-emitter depletion region, $\Delta E_{g \text{ BEjunction}}$, and is further reduced by grading the bandgap across the neutral base. This allows a reduction in the intrinsic base resistance of the HBT (R_{bi}) through increased doping, but still maintaining low charge storage in the emitter. For the case of an HBT with uniform doping in the base and a linearly graded composition across the neutral base, with the grading being greater than the thermal voltage (ie. $\Delta E_{g \text{ BC} - \text{BE}} \gg kT$), the reductions in τ_b and τ_e are given by equations 1.3 and 1.4 respectively[4].

$$\frac{\tau_{b \text{ HBT}}}{\tau_{b \text{ BJT}}} \cong \frac{2kT}{E_{g \text{ BCjunction}} - E_{g \text{ BEjunction}}} \left[1 - \frac{kT}{E_{g \text{ BCjunction}} - E_{g \text{ BEjunction}}} \left[1 - e^{-\left(\frac{E_{g \text{ BCjunction}} - E_{g \text{ BEjunction}}}{kT} \right)} \right] \right] \quad (1.3)$$

$$\frac{\tau_{e \text{ HBT}}}{\tau_{e \text{ BJT}}} \cong \frac{R_{bi \text{ BJT}}}{R_{bi \text{ HBT}}} e^{-\frac{E_{g \text{ BEjunction}}}{kT}} \left[\frac{1 - e^{-\left(\frac{E_{g \text{ BCjunction}} - E_{g \text{ BEjunction}}}{kT} \right)}}{\frac{E_{g \text{ BCjunction}} - E_{g \text{ BEjunction}}}{kT}} \right] \quad (1.4)$$

The use of heterojunctions then not only gives the additional degree of freedom for decoupling gain and base doping but also allows some tradeoffs as to the composition across the neutral regions. These are then used in device design for reducing the overall device delay for a desired breakdown voltage or minimum gain desired. However the advantages to using heterojunction transistors come at the expense of increased difficulties in their fabrication.

1.2 Introduction to Silicon-Based Heterojunctions and $\text{Si}_{1-x}\text{Ge}_x$ Alloys

For the formation of high quality heterojunctions for use in HBTs, the two materials involved must have the desired difference in bandgap while having a similar lattice constant in order to grow a junction of suitable quality[4]. Figure 1 shows the lattice constant and electrical bandgap of common semiconductor materials. The "group III-V" semiconductor materials, such as gallium, aluminum, arsenic, and phosphorus, have a wide range of alloys where the conditions on lattice constant and bandgap are met. Correspondingly, HBTs made from these materials have been widely reported and are nearing commercial use[5].

Despite the advances of the "III-V" HBTs, for reasons of economics, a heterojunction technology compatible with conventional silicon processing is desired[6]. As shown in figure 1.1, there is a significant lattice mismatch between the "group IV" semiconductors, such as silicon, carbon, and germanium. Any mismatch between the lattices of the two materials results in the formation of misfit dislocations at the heterointerface. In the case of HBTs, one heterointerface lies at the base-emitter junction. Should the dislocations at this junction become electrically active, in particular those having a threading component, an increase in the recombination component of the base current arises, resulting in a decrease in gain and a degradation of performance[7]. Should dislocations formed at the base-collector junction become electrically active, the result is an increase in reverse leakage currents[7].

Considering the two approaches for forming HBTs mentioned, the wide bandgap emitter approach appears most attractive from a process integration viewpoint[12]. Several materials have been tried as candidates for HBT applications on silicon substrates. These include gallium phosphide (GaP)[8], β -SiC[9], and amorphous silicon[10,11]. These however have had difficulties both in process

Fig. 1.1 Bandgap and Lattice Constants of Common Semiconductor Materials[6]

integration and in performance. GaP is difficult to grow epitaxially on $\langle 001 \rangle$ Si due to its highly polar nature[3,8,12]. β -SiC typically requires high temperature processing at a late stage in the device formation and the resulting diffusion of the base implant is a major drawback[12]. Amorphous or microcrystalline SiC films have also been attempted, but along with amorphous silicon, have performance

problems associated with poor interface quality and high resistance (both bulk and contact resistances)[9,12]. Amorphous silicon has a further drawback in that it tends to recrystallize to polysilicon at temperatures used during metallization (eg. during silicide formation, etc.)[3,10–12]. In summary, all these approaches have integration issues to be resolved and as yet none have shown significant improvement over the industry standard poly-emitter BJT. Thus wide bandgap emitter HBTs have not seen widespread application.

Bulk alloys of silicon and germanium, $\text{Si}_{1-x}\text{Ge}_x$, have been found to be readily formed. The two elements are from “group IV” in the periodic table, are similar chemically and are miscible over the entire range of composition. The alloy layers retain a diamond lattice structure, with the lattice constant, a , being dependent on the composition of the layer. For low mole fractions, x , of germanium this relationship is described by Vegard’s rule, given in equation 1.5.

$$a_{\text{SiGe}} = a_{\text{Si}} + x(a_{\text{Ge}} - a_{\text{Si}}) \quad (1.5)$$

The bandgap of these alloys was determined in the 1950’s through optical absorption measurements[13]. It was found that increasing the amount of germanium in the alloy layer reduced the energy bandgap from that of silicon toward that of germanium. This leads to applications of this material as the base region of an HBT. This application is limited, however, as the amount of germanium needed for a significant bandgap reduction results in a considerable misfit between the lattice constant of the alloy and the silicon substrate. (Note that silicon and germanium have a 4.17% mismatch in their lattice constants.) However, recent advances in strained layer epitaxial growth techniques allow for the pseudomorphic growth of these alloys.

Strained layer epitaxy is only possible when the materials are chemically similar and under conditions where the growth is two-dimensional or “layer by layer.” Under these conditions the first layer of growth will lattice match to the substrate since for the first layer the strain energy is nearly always less than the energy required to form a dislocation. The following layers will then continue to match to the layer beneath them, resulting in a lattice matched heterointerface and an epitaxial layer grown with a corresponding tetragonal distortion of the lattice. This circumvents the problem of the lattice mismatch. Note that for strained $\text{Si}_{1-x}\text{Ge}_x$ on Si, the compressive strain in the layer caused by the lattice distortion results in a significant additional reduction in the bandgap[14].

This layer by layer growth will continue to happen until the strain energy associated with the lattice distortion surpasses the energy required for dislocation formation and there is sufficient thermal activation for a dislocation to form. The formation of dislocations occurs at the heterointerface (misfits) and relaxes the strain in the layer. For any heterosystem, the point where this relaxation occurs is a function of strain (alloy composition), layer thickness, and activation (thermal exposure). This leads to the concepts of the “critical thickness” for a stable layer and the maximum thermal budget a metastable layer can be exposed to prior to strain relaxation. Note a stable layer is thin enough such that it will not relax under thermal exposure (ie. its strain energy is less than the energy for dislocation formation) and a metastable layer exceeds the critical thickness for stability but has not relaxed due to insufficient activation. Theories for predicting stability are still under development for the $\text{Si}_{1-x}\text{Ge}_x/\text{Si}$ heterosystem[15]. Dislocations at the heterointerface have been seen to have a negligible impact on the ideality of the current-voltage characteristics of $\text{Si}_{1-x}\text{Ge}_x/\text{Si}$ diodes until large numbers are present[7]. Speculation is that the dislocations must have threading components and be decorated in order to be electrically active. However, the impact of the resulting

reduction in strain has a very significant impact on the bandgap of the layer. The amount of bandgap narrowing seen in the layer is reduced, decreasing the benefit to device performance expected through the use of the heterojunction. Thus misfit dislocations and the resulting strain relaxation are to be avoided in HBT applications, so the critical thickness and maximum thermal budget set upper limits for growing and processing alloy layers of a given composition.

Strained $\text{Si}_{1-x}\text{Ge}_x$ alloy layers have been grown using epitaxial growth techniques such as molecular beam epitaxy (MBE)[16], rapid thermal chemical vapour deposition (RTCVD)[17], and ultrahigh vacuum chemical vapour deposition (UHV/CVD)[18]. These low temperature, low thermal budget growth techniques are necessary to ensure the preservation of the lattice match and compressive strain and to provide control of the layer growth in order to achieve abrupt changes in both composition and doping, allowing the alloys to be used in HBTs.

As anticipated, HBTs using $\text{Si}_{1-x}\text{Ge}_x$ alloy layers as the base region material have been demonstrated. The first of these were mesa isolated devices with the layers fabricated using MBE. Although nonideal base currents were present, the desired increases in collector current were observed[19,20]. In subsequent devices the enhanced performance of $\text{Si}/\text{Si}_{1-x}\text{Ge}_x/\text{Si}$ HBTs has been observed, with gains of over 5000[21] and cutoff frequencies of 75 GHz[22] at room temperature being reported. In the latter case, "built-in" drift fields resulting from a grading from 0% Ge at the base-emitter junction to 7% Ge at the base-collector junction were used. These devices have been demonstrated in digital gates having sub-30 ps gate delays[23]. More recently, HBTs have been reported exhibiting not only high cutoff frequencies but a high maximum frequency of oscillation, f_{max} , as well (50 GHz)[24,25]. This result is encouraging in that the performance of the material system has been

demonstrated in analog microwave communications applications previously restricted to the "III-V" semiconductor devices[25].

Note there have been several other types of devices fabricated using $\text{Si}_{1-x}\text{Ge}_x/\text{Si}$ or $\text{Si}_{1-x}\text{Ge}_x/\text{Si}_{1-y}\text{Ge}_y$ heterojunctions in order to exploit the different band structures achievable. These include MODFETs and MOSFETs where enhanced channel mobilities may be achieved by means of carrier confinement and modulation doping[26,27]. Also, the energy band structure has been exploited to form resonant tunneling diodes and transistors[28,29]. Long wavelength infrared detectors have demonstrated using $\text{Si}/\text{Si}_{1-x}\text{Ge}_x$ superlattices, having a photoresponse in the range from 6–12 μm [30]. Of these other devices, only the MOSFET and MODFET applications have generated interest similar to that surrounding the HBTs.

1.3 Introduction to Dopant Induced Bandgap Narrowing

As outlined previously, one of the key benefits of using heterojunctions is the ability to decouple the restrictions on base doping and transistor gain, allowing for increased doping in the base. This reduces the base resistance, and may improve f_{max} and serve to reduce the noise of the device[3,12,31]. Heavy doping, however, is known to reduce the bandgap of silicon and most other semiconductors[32–36]. This is true for both n-type and p-type doping and presents a problem when the band reduction due to heavy doping is in the emitter, as is the case for silicon homojunction BJTs[34]. This reduces the device gain from the anticipated value, as the bandgap difference now favours increased back injection of carriers from the base to the emitter. In the case of HBTs, additional bandgap narrowing from heavy doping in the base would serve to increase the gain of the transistor, allowing for a different optimization of gain versus composition grading to reduce the delay intrinsic to the transistor.

There are several physical mechanisms that tend to narrow the energy gap, including impurity band widening, majority carrier screening, band tailing, and dopant induced strain in the lattice[34,35]. Strain resulting from heavy doping with phosphorus or boron in silicon is tensile in nature. This appears to work against the effect of germanium incorporation, which induces a compressive strain in the layer. The presence of the dopant may tend to reduce the total strain present, giving a reduction in the total strain present and hence a reduction in the amount of bandgap narrowing seen. Due to the very high doping levels used in HBT applications[24], knowledge of the impact of doping on the band structure of $\text{Si}_{1-x}\text{Ge}_x$ becomes important.

1.4 Outline of Thesis

The impact of heavy doping on the bandgap of $\text{Si}_{1-x}\text{Ge}_x$ has been theoretically predicted in a model applicable to the common semiconductor families[26]. The work presented in this thesis provides the first experimental confirmation of this theory as applied to $\text{Si}_{1-x}\text{Ge}_x$ alloy layers used in heterojunction bipolar transistors. The theoretical calculation of the bandgap in silicon germanium alloys is outlined in the next chapter. Included is a discussion of the conventional technique for measuring the energy bandgap, and the subsequent extension of the technique to the base region of an HBT in the case of the strain induced splitting of the normally degenerate subbands in the valence band. $\text{Si}_{1-x}\text{Ge}_x$ strained layer growth and the fabrication of the experimental HBT samples studied here are discussed in Chapter 3. Chapter 4 deals with the measurement of the base bandgap using the HBTs, and compares the measured bandgap narrowing with theory. Chapter 5 presents a discussion of the results. Conclusions and suggestions for future work are presented in Chapter 6.

Chapter 2

Bandgap Narrowing in Heavily Doped $\text{Si}_{1-x}\text{Ge}_x$ Alloys

2.1 Introduction

As mentioned in the previous chapter, there are two mechanisms that tend to reduce the bandgap of heavily doped strained $\text{Si}_{1-x}\text{Ge}_x$ alloy layers typically used for HBT applications, the strain associated with the pseudomorphic growth and the effect of the heavy doping levels used. The total bandgap narrowing then consists of a combination and interaction of the two effects. The calculation of each of these components forms the basis of the following section, where the application to those layers of use in HBTs is considered. The final portion of the chapter then examines the theoretical basis for the determination of the bandgap of the base region of a $\text{Si}/\text{Si}_{1-x}\text{Ge}_x/\text{Si}$ HBT from the measurement of the temperature dependence of the collector saturation current.

2.2 Theoretical Calculation of the Energy Bandgap

2.2.1 Bandgap of Strained $\text{Si}_{1-x}\text{Ge}_x$ on $\langle 001 \rangle$ Si Substrate

For application to heterojunction bipolar transistors, the narrow bandgap $\text{Si}_{1-x}\text{Ge}_x$ alloy layer forms the base region of the device, with the substrate forming the collector and the emitter being either a subsequent epitaxial layer or a deposited polycrystalline silicon layer. This arrangement has the majority of the bandgap difference in the valence band ($\Delta E_v \gg \Delta E_c$ or $\Delta E_v \simeq \Delta E_g$), a situation desirable for npn HBTs[4].

The pseudomorphic growth of $\text{Si}_{1-x}\text{Ge}_x$ on cubic $\langle 001 \rangle$ Si results in the normally cubic alloy layer undergoing a biaxial compression in the plane of the interface (ie. parallel to the heterointerface) and an extension of the lattice in a direction perpendicular to the interface. This tetragonal distortion is homogeneous in the alloy layer and is described by elasticity theory. The implication of this is that the compositional components of the tensor describing the strain in the lattice are readily determined from the lattice constants of the alloy (from Vegard's rule) and the substrate and from the Poisson's ratio of the alloy layer, taken to be an interpolation between that of silicon and germanium. The strain tensor along with the corresponding deformation potentials was then used by People to determine the indirect bandgap of $\text{Si}_{1-x}\text{Ge}_x$ on cubic $\langle 001 \rangle$ Si [14].

In the calculation of the bandgap of the strained layer, four contributions need to be considered [14]. These include the bandgap of the unstrained (bulk) alloy, uniform shifts of the bandgap resulting from dilation of the lattice (ie. the hydrostatic contribution), splittings of the indirect conduction band edges at their minima, and splittings of the degenerate valence band edges. The indirect bandgap of the strained alloy layers is then the sum of these contributions. The results of People's analysis are shown in figure 2.1 [14]. As is readily apparent, there is a considerable reduction in the bandgap of the strained alloy from that of the unstrained alloy.

Despite the complexity of People's calculation, the bandgap of strained $\text{Si}_{1-x}\text{Ge}_x$ alloy layers has been shown by People and Bean to be well described by the linear relation of equation 2.1 for mole fractions of germanium less than

$$\Delta E_V = 0.74x \quad (2.1)$$

approximately 30% [37]. It is this relation that is used to determine the bandgap of alloy layers customarily, due to its simplicity.

Figure 2.1 Indirect Bandgap of Coherently Strained $\text{Si}_{1-x}\text{Ge}_x$ on $\langle 001 \rangle$ Si[14]

2.2.2 Dopant Dependent Bandgap Narrowing

Recently, Jain and Roulston presented a model for bandgap narrowing in heavily doped semiconductors, including $\text{Si}_{1-x}\text{Ge}_x$ strained layers[36]. The model results in a simple expression for calculating the bandgap narrowing. The expression requires knowledge of only the Rydberg energy and the effective Bohr radius associated with the dopant atom, which are determined from the effective mass of the carriers, the dielectric constant, and the doping level of the semiconductor. It is, however, only applicable to the heavy doping regime where doping levels higher than $1 \times 10^{18} \text{ cm}^{-3}$ are under consideration (with the assumption of an uncompensated semiconductor).

The theory considers four contributions to the bandgap narrowing. These include the shift of the majority band edge due to exchange interactions or exchange energy ($\Delta E_{x(maj)}$), the shift of the minority band edge due to carrier-carrier or electron-hole interactions, also known as correlation energy ($\Delta E_{cor(mino)}$), and the shifts of the majority and minority band edges due to carrier-impurity interactions ($\Delta E_{i(maj)}$ and $\Delta E_{i(mino)}$). Note all these shifts are into the bandgap resulting in their numerical sum being equal to the change in bandgap[36].

$$\Delta E_g = \Delta E_{x(maj)} + \Delta E_{cor(mino)} + \Delta E_{i(maj)} + \Delta E_{i(mino)} \quad (2.2)$$

In determining the exchange energy, the effect of screening must be taken into account. Rather than using many body methods to determine a correction factor based on density in each case, a value yielding approximately the correct effect over the doping range in question was used for simplicity[36]. It had been previously found that in the heavy doping regime the effect was only weakly dependent on doping[36, and references within].

Furthermore, when calculating exchange energy of p-type semiconductors, the presence of N_b equivalent valleys in the valence band and the interaction between heavy and light holes must also be accounted for. The reduction in exchange energy due to light and heavy hole interactions is represented by a correction factor, Λ , taken to be 0.75[36]. In order to explain discrepancies between luminescence data and conventional theory for the position of the fermi level in p-semiconductors without affecting the calculation of exchange energy or hence bandgap narrowing, the number of equivalent valleys in the valence band, N_b , was set to 2 instead of the conventional 1, but a lighter effective mass for the carriers was used. Note that for strained p-type $Si_{1-x}Ge_x$ this value is suggested to be set to 1.5 in order to account

for the strain induced separation of the normally degenerate valence subbands[36]. The exchange energy component of the bandgap narrowing is calculated using equation 2.3.

$$\frac{\Delta E_{x(max)}}{R} = 1.83 \frac{\Lambda}{N_b^{1/3}} \frac{1}{r_s} \quad (2.3)$$

The other terms are straightforward in application, though approximations are made to simplify and generalize the application of the theory[36].

$$\frac{\Delta F_{ex(mino)}}{R} = \frac{0.95}{r_s^{3/4}} \quad (2.4)$$

$$\frac{\Delta E_{R(max)}}{R} = \frac{1.57}{N_b r_s^{3/2}} \quad (2.5)$$

$$\frac{\Delta E_{R(mino)}}{R} = \frac{R_{(mino)}}{R} \frac{1.57}{N_b r_s^{3/2}} \quad (2.6)$$

In the above equations R is the Rydberg energy, $R_{(mino)}$ is the minority carrier Rydberg energy, and r_s is a dimensionless parameter associated with the doping concentration[36].

$$R = \frac{13.6 m_d}{\epsilon^2} \quad \text{eV} \quad (2.7)$$

$$r_s = \frac{r_d}{a_B} \quad (2.8)$$

$$1 = \frac{4 \pi N r_d^3}{3} \quad (2.9)$$

$$a_B = \frac{0.53\epsilon}{m_d} \times 10^{-8} \quad \text{cm}^{-3} \quad (2.10)$$

In the above equations r_d represents half the average distance between impurity atoms, a_B is the effective Bohr radius, N is the doping concentration, ϵ is the dielectric constant, and m_d is the effective density of states mass in the majority band, normalized to the free electron mass[36].

Putting these equations into equation 2.2, the result for bandgap narrowing, ΔE_g , is determined. Note that in this form equation 2.11 is applicable to all group IV and III-V semiconductors.

$$\frac{\Delta E_g}{R} = 1.83 \frac{A}{N_b^{1/3}} \frac{1}{r_s} + \frac{0.95}{r_s^{3/4}} + \left[1 + \frac{R_{(minor)}}{R} \right] \frac{1.57}{N_b r_s^{3/2}} \quad (2.11)$$

Figure 2.2 shows the application of equation 2.11 to p-type strained $\text{Si}_{1-x}\text{Ge}_x$. Here the mole fraction of germanium is kept below 0.3, so the conduction band is assumed Si-like (ie. $m_c=0.33$). The alloy or compositional dependence of the hole effective mass, m_h , and dielectric constant, ϵ , were accounted for using linear interpolations. Note that strain effects are assumed accounted for by the N_b factor.

$$m_h(x) = 0.59 (1 - 0.41x) \quad (2.12)$$

$$\epsilon(x) = 13.6 (1 + 0.35x) \quad (2.13)$$

In order to determine the total bandgap narrowing in heavily boron doped $\text{Si}_{1-x}\text{Ge}_x$ alloys, the result of equation 2.1 must be added to that of equation 2.11. A straight forward addition ignores any impact of the dopant induced reduction in the hydrostatic component of the strain present in pseudomorphic $\text{Si}_{1-x}\text{Ge}_x$ alloy layers[35]. The validity of neglecting this is to be determined, but it is expected to

Figure 2.2. Dopant Induced Bandgap Narrowing in p-type $\text{Si}_{1-x}\text{Ge}_x$ Layers
(from equation 2.11)

solid line 0% Ge, -.- 8% Ge, ... 15% Ge, --- 20% Ge

be worse as doping levels increase. Note that for boron doping levels 5 to 10 times lower than the Ge concentration the strain caused by Ge incorporation will be entirely compensated[35]. Figure 2.3 shows the combined (ie. superimposed) effects of the strain induced bandgap narrowing along with the dopant induced bandgap narrowing.

Figure 2.3. Total Bandgap Narrowing in p-type $\text{Si}_{1-x}\text{Ge}_x$ Layers

solid line 0% Ge, -- 8% Ge, ... 15% Ge, --- 20% Ge

2.3 Theoretical Basis for Bandgap Extraction from Collector Saturation Current

The conditions relevant to the measurements considered here include a thin, heavily doped $\text{Si}_{1-x}\text{Ge}_x$ base region, uniform in both dopant concentration and alloy composition, with a lightly doped emitter and collector. Assuming the acceptors are completely ionized and that the energy bandgap narrowing is the same under low level injection conditions as it is in equilibrium, then the analysis presented by

Neugroschel[34] for a heavily doped base may be applied. With the additional assumption of the electron quasi-fermi energy, E_{Fn} , being constant across the base-emitter depletion region (effectively assuming there is no discontinuity in the conduction band edge), then the bandgap, E_g , is given by equation 2.14 as illustrated in figure 2.4, in the case of strain-induced splitting of normally degenerate conduction and valence subbands. Note that for this analysis, the subscripts 1 denote the lowest conduction subband or the highest valence subband and the subscripts 2 denote the next higher conduction subband or the next lower valence subband.

Figure 2.4 Band Diagram for Analysis of Strained $\text{Si}_{1-x}\text{Ge}_x$ Base Layer

$$E_g = (E_{Fp} - E_{V1}) + qV + (E_{C1} - E_{Fn}) \quad (2.14)$$

For no subband splitting, such that $E_{C1} = E_{C2}$, $E_{V1} = E_{V2}$ in equation 2.14, the carrier densities at the base side of the base-emitter depletion region, $p,n(0)$, (x -coordinate = 0) are given by equations 2.15 and 2.16.

$$n(0) = N_C e^{\frac{-(E_C - E_{Fn})}{kT}} \quad (2.15)$$

$$p(0) = N_V \mathcal{F}_{1/2}(\eta_V) \quad (2.16)$$

$$\eta_V = \frac{E_V - E_F}{kT} \quad (2.17)$$

Assuming $p(0) = N_A$, combining equations 2.14, 2.15, and 2.16 gives 2.18.

$$n(0) = \frac{N_C N_V}{N_A} \frac{\mathcal{F}_{1/2}(\eta_V)}{e^{\eta_V}} e^{\frac{-E_g}{kT}} e^{\frac{qV}{kT}} \quad (2.18)$$

Here k is Boltzman's constant, T is absolute temperature, $N_{C,V}$ are the densities of states for the conduction and valence bands, V is applied potential, q is electronic charge, and $\mathcal{F}_{1/2}(\eta_V)$ is the fermi integral of order 1/2. This factor is required to account for the need to apply Fermi-Dirac statistics in place of Maxwell-Boltzman statistics at these doping levels[34].

In order to allow for subband splitting, it is assumed that carriers lie only in the two lowest conduction subbands and the two highest valence subbands, each of which are assumed to have a different density of states. With this assumption, equations 2.15 and 2.16 become 2.19 and 2.20.

$$n(0) = N_{C1} e^{\frac{-(E_{C1} - E_{Fn})}{kT}} + N_{C2} e^{\frac{-(E_{C2} - E_{Fn})}{kT}} \quad (2.19)$$

$$p(0) = \left[1 + \frac{N_{V2}}{N_{V1}} \frac{\mathcal{F}_{1/2}(\eta_{V2})}{\mathcal{F}_{1/2}(\eta_{V1})} \frac{e^{\eta_{V1}}}{e^{\eta_{V2}}} e^{\frac{-(E_{V1} - E_{V2})}{kT}} \right] \frac{N_{V1} \mathcal{F}_{1/2}(\eta_{V1})}{e^{\eta_{V1}}} e^{\frac{-(E_{V1} - E_{Fn})}{kT}} \quad (2.20)$$

Again assuming $p(0) = N_A$, combining equations 2.14, 2.19, and 2.20 gives equation 2.21.

$$n(0) = \frac{(N_C N_V)_{eff}}{N_A} \frac{\mathfrak{F}_{1/2}(\eta_{V1})}{e^{\eta_{V1}}} e^{\frac{-E_g}{kT}} e^{\frac{qV}{kT}} \quad (2.21)$$

where

$$(N_C N_V)_{eff} = N_{C1} N_{V1} f_{split} \quad (2.22)$$

and

$$f_{split} = \left[1 + \frac{N_{C2}}{N_{C1}} e^{\frac{-(E_{C2}-E_{C1})}{kT}} \right] \left[1 + \frac{N_{V2}}{N_{V1}} \frac{\mathfrak{F}_{1/2}(\eta_{V2})}{\mathfrak{F}_{1/2}(\eta_{V1})} \frac{e^{\eta_{V1}}}{e^{\eta_{V2}}} e^{\frac{-(E_{V1}-E_{V2})}{kT}} \right] \quad (2.23)$$

$$= 1 + a e^{\frac{-\Delta E_C}{kT}} + b e^{\frac{-\Delta E_V}{kT}} + a b e^{\frac{-(\Delta E_C + \Delta E_V)}{kT}} \quad (2.24)$$

where $\Delta E_C = E_{C2} - E_{C1}$, $\Delta E_V = E_{V1} - E_{V2}$, and $a = N_{C2}/N_{C1}$. The expression defining b is given in equation 2.25.

$$b = \left[\frac{N_{V2}}{N_{V1}} \frac{\mathfrak{F}_{1/2}(\eta_{V2})}{\mathfrak{F}_{1/2}(\eta_{V1})} \frac{e^{\eta_{V1}}}{e^{\eta_{V2}}} \right] \quad (2.25)$$

Recall the collector current density of a uniform base npn bipolar transistor is related to $n(0)$ by equation 2.26.

$$J_C = \frac{kT\mu_n}{W_B} \frac{(N_C N_V)_{eff}}{N_A} \frac{\mathfrak{F}_{1/2}(\eta_{V1})}{e^{\eta_{V1}}} e^{\frac{-E_g}{kT}} e^{\frac{qV}{kT}} \quad (2.26)$$

$$= J_{CO} e^{\frac{qV}{kT}} \quad (2.27)$$

Here μ_n is the minority carrier mobility in the base and W_B is the base width. The collector saturation current density, J_{CO} , is seen to be exponentially related to the bandgap. Eg.

$$J_{CO} = \frac{kT\mu_n (N_C N_V)_{eff}}{W_B N_A} \frac{\mathfrak{F}_{1/2}(\eta_V)}{e^{\eta_V}} e^{\frac{-E_g}{kT}} \quad (2.28)$$

Equation 2.28 can then be rewritten in the following form:

$$J_{CO} \frac{W_B N_A}{kT\mu_n N_C N_V} = \frac{f_{split} \mathfrak{F}_{1/2}(\eta_V)}{e^{\eta_V}} e^{\frac{-E_g}{kT}} \quad (2.29)$$

Two assumptions are made at this point. The first assumes at this doping level and over the temperature range considered, μ_n is independent of temperature. This assumption is true for silicon, but there is no data for $\text{Si}_{1-x}\text{Ge}_x$ [38]. The second assumption is that $N_{C1, V1}$ vary as $T^{3/2}$ as in silicon. Again the validity for $\text{Si}_{1-x}\text{Ge}_x$ is not known. With these two assumptions, the term $W_B N_A / kT\mu_n N_C N_V$ is seen to have a T^{-4} dependence. Recalling that near room temperature the temperature dependence of the bandgap can be described by $E_g = E_{go} - \alpha T$, where E_{go} is the bandgap extrapolated to 0 K, equation 2.30 is obtained.

$$\ln\left(\frac{J_{CO}}{T^4}\right) = \ln f_{split} + \ln\left(\frac{\mathfrak{F}_{1/2}(\eta_V)}{e^{\eta_V}}\right) - \frac{E_{go}}{k} \frac{1}{T} + C \quad (2.30)$$

If the temperature dependence of $\ln f_{split}$ and $\ln (\mathfrak{F}_{1/2}(\eta_V)/\exp(\eta_V))$ is taken to be zero, then a plot of $\ln (J_{CO}/T^4)$ versus $1/T$ will yield E_{go} from its slope, either directly assuming that bandgap narrowing, ΔE_g , is independent of temperature, or through numerical solution. Otherwise E_{go} will be given by equation 2.31.

$$E_{go} = \Delta E_{split} + \Delta E_{FD} - k \frac{d \ln\left(\frac{J_{CO}}{T^4}\right)}{d(1/T)} \quad (2.31)$$

$$\Delta E_{split} = k \frac{d \ln f_{split}}{d(1/T)} = \frac{k}{f_{split}} \frac{df_{split}}{d(1/T)} < k \frac{df_{split}}{d(1/T)} \quad (2.32)$$

$$\Delta E_{FD} = k \frac{d \ln \left(\frac{\mathcal{F}_{1/2}(\eta_V)}{e^{\eta_V}} \right)}{d(1/T)} \quad (2.33)$$

The bandgap narrowing is then determined by subtracting E_{go} from 1.206eV, the extrapolated value for the bandgap of silicon.

The assumptions of the temperature dependence of μ_n and $N_{C,V}$ for $Si_{1-x}Ge_x$ are two unknowns in the theory. If there is a deviation from the assumed dependence, the temperature dependence of J_{CO} will be different. That is, the T^{-4} will become a T^{-m} , where the exact exponent, m , may not be known. The error introduced into the bandgap narrowing by having an incorrect exponent can be shown to be given by equation 2.34[34]. (Please note the equation given in [34] was found to be in error. The derivation of equation 2.34 is given in Appendix A.)

$$\Delta E_g = \Delta m \ k \ln \left(\frac{T_2}{T_1} \right) \left(\frac{1}{T_2} - \frac{1}{T_1} \right)^{-1} \quad (2.34)$$

Here Δm is the change in the exponent as required by the actual temperature dependence, k is Boltzman's constant, and $T_{1,2}$ are the lower and upper absolute temperatures used to calculate the slope. For the temperature range investigated, 10°C to 120°C, changing m by 0.25 changes the bandgap narrowing by 7.15 meV. Changing m by 1 changes the bandgap narrowing by 28.6 meV. This is one potential source of error encountered using this technique for determining the bandgap of the strained alloy layers.

Chapter 3

$\text{Si}_{1-x}\text{Ge}_x$ Strained Layer Growth and HBT Fabrication

3.1 Introduction

The purpose of this experiment was to determine the total bandgap narrowing occurring in the base of a $\text{Si}_{1-x}\text{Ge}_x$ HBT where the doping levels in the strained $\text{Si}_{1-x}\text{Ge}_x$ layer were sufficient for significant bandgap narrowing to arise from heavy doping effects. The process parameter under investigation was the germanium content of the layer. It was decided that a simple process flow resulting in a double mesa isolated structure, such as that shown in figure 3.1, would suffice for these measurements. The strained $\text{Si}_{1-x}\text{Ge}_x$ layers were grown at the Institute for Microstructural Sciences at the National Research Council of Canada. Their fabrication into HBTs occurred at Carleton University's semiconductor fabrication facility. The following section describes the layer growths and characterization of the experimental wafers used. That is followed by a section detailing the key process steps used to fabricate HBTs from the as grown layers.

Figure 3.1 A Double Mesa Isolated HBT used for Bandgap Measurements

3.2 Si_{1-x}Ge_x Strained Layer Growth and Characteristics

The layers used to form the HBTs were blanket deposited layers, grown by molecular beam epitaxy (MBE) in a VG Semicon V80 MBE system. The details of this system are given elsewhere[39,40]. This system is capable of achieving the ultra high vacuum conditions (base pressures $< 10^{-10}$ Torr) required for low temperature epitaxy. Silicon and germanium atoms required for the epitaxial growth are e-beam evaporated from solid sources, with *in situ* monitoring of the flux of each species. Doping is achieved by evaporation of boron for p-type layers and by low energy arsenic implantation for n-type regions. Note that doping is done during the layer growth in both cases, not in a subsequent step. Doping by low energy implantation has a major drawback in that the beam currents achievable with this system are low, requiring very slow growth rates for regions where high As doping is desired. Even when the growth rate is slowed, the maximum doping achieved is not very high, requiring an additional doping stage for n⁺ regions (eg. where ohmic contact to n-type silicon is required).

The starting material was 100 mm diameter 0.01 Ω cm, Sb doped silicon substrates of <001> orientation. Heavy n⁺ doping in the substrate was necessary as the devices examined made use of the substrate as a common collector.

Cleaning of the wafers prior to the epitaxial growth consisted of a 50 min *ex situ* treatment in UV-ozone to remove any hydrocarbon contamination present, followed by an *in situ* oxide desorption where the wafers were subjected to a Si flux equivalent to a 0.01 nm/s growth rate for 10 min at 900°C in order to evaporate any oxide from the wafer surface. This technique is good for blanket depositions (ie. uniform depositions on unpatterned wafers), but leaves a residual boron spike at the initial growth interface having a non-negligible concentration[41]. This spike must be compensated with n-type doping to remove any parasitic junctions.

The n^- collector, the p^+ $\text{Si}_{1-x}\text{Ge}_x$ strained layer base, and the n^- emitter region were grown sequentially following the preclean, without breaking vacuum. The collector layer was grown at 650°C with a target thickness of 600nm and an arsenic doping level nominally $1 \times 10^{17} \text{cm}^{-3}$. The base layer was grown next at a temperature of 500°C. Nominal doping level and thickness were $2 \times 10^{19} \text{cm}^{-3}$ of boron and 50nm respectively. The germanium content of the layer was kept uniform across the neutral base for each sample, but was varied from 8% to 20% concentration between samples. Note that no spacer layers were used between the base and collector but the germanium content was graded from 0% in the collector to the desired value in the base across the base-collector depletion region to prevent the formation of parasitic barriers to current flow in the conduction band[4,42–44]. Also, the germanium flux was initialized prior to the boron flux in order to contain the base (ie. the boron profile) within the narrow bandgap region (ie. the germanium profile), again to avoid parasitic barriers forming in the conduction band[42,43]. A nominally 600 nm thick, lightly doped ($1 \times 10^{17} \text{cm}^{-3}$ As) emitter was grown, again at 500°C.

The actual doping and germanium profiles achieved were measured by secondary ion mass spectroscopy (SIMS). The germanium content was also measured by x-ray diffraction (XRD). The thickness of the emitter and base were determined by cross-section transmission electron spectroscopy (XTEM). The results are summarized in Table 3.1. Figure 3.2 shows the SIMS profile of sample 1034.

Table 3.1 Doping and Composition of Wafers used for HBT fabrication

Wafer	% Ge	Base Doping (cm^{-3})	Emitter and Collector Doping (cm^{-3})	Emitter Thickness (nm)	Base Thickness (nm)	Collector Thickness (nm)
1034	8	2×10^{19}	1×10^{17}	589	50	476
1035	15	2×10^{19}	5×10^{16}	590	47	600
1036	20	2×10^{19}	8×10^{16}	600	47	591

Figure 3.2 SIMS Profile of Sample 1034

3.3 Process Development and Si/Si_{1-x}Ge_x/Si HBT Fabrication

3.3.1 Process Issues for HBT Fabrication

There are several issues that must be considered when developing a process flow for the formation of HBTs using Si_{1-x}Ge_x strained layers. These include the technique to be used to electrically isolate adjacent devices, the means by which the exposed Si and Si_{1-x}Ge_x surfaces are to be passivated, and how electrical contact is to be made to the layers.

There are also factors preventing the application of standard silicon processing to the Si_{1-x}Ge_x/Si heterosystem. The restriction on the total thermal budget the layers can be exposed to while maintaining their strain is of primary concern. In the case of metastable layers such as those used here, this restriction removes most capacity for any high temperature processing to be done after the layers are in place. The only high temperature processing that can be done must be of short duration (eg. rapid thermal anneal). Since here the strained layers were deposited prior to any processing, the entire process flow must be at low temperature. The other issue of note is the oxidation behaviour of Si_{1-x}Ge_x. It has been reported that these strained layers oxidize in a manner such that SiO₂ is formed, but as the layer is consumed during oxidation, the germanium segregates out of the oxide into the layer. This effectively increases the germanium content of the layer, leading to increased strain relaxation[45,46]. As thermal oxidation is the passivation technique of choice to silicon processing, an alternative means must be found.

With these concerns in mind, a fabrication sequence was developed. The interested reader is referred to Appendix B for the full process flow. A summary of the process flow is shown in figure 3.3. The following subsections detail how these concerns were addressed in the fabrication of the HBTs used in this study.

Figure 3.3 Process Flow Used for HBT Fabrication

3.3.2 Isolation of $\text{Si}_{1-x}\text{Ge}_x$ HBTs

As was outlined in the introduction to the chapter, the devices under study here were isolated using a double mesa approach. Conventional silicon processing relies on oxidation and pn junction diodes for electrical isolation between neighbouring devices. The thermal budget associated with a field oxidation cycle, such as used in a LOCOS isolation, prohibits its use for isolating HBTs formed from the blanket deposited layers used here. A single mesa approach requires a high dose boron implant in order to contact the extrinsic base[43,47,48]. This must be followed by an anneal to repair implant damage and to activate the dopant. This adds to the thermal budget seen by the layers. Furthermore, since the implant must reach the strained $\text{Si}_{1-x}\text{Ge}_x$ layer, there is likely to be damage to the lattice in this layer. This may provide sites for the nucleation of dislocations. The impact of this on device characteristics is not known. The intrinsic base is not directly involved, except for the case where the conditions of the implant and anneal are such to create dislocations which during subsequent processing grow into the intrinsic base, relaxing the strain there. It is unlikely that an anneal to repair the damage would return the extrinsic base to its previously strained state, but may instead increase the number of dislocations present and enhance strain relaxation. Thus the conditions of implant and anneal used to make single mesa structures[43,47,48] tend not to activate all the dopant or remove all the damage in the extrinsic base[49], with the hope that the intrinsic base is not affected. The double mesa approach leaves the extrinsic base as an exposed surface, which may have implications for device characteristics (ie. increased base currents due to improperly passivated surfaces), but there are no high temperature steps required.

The emitter mesas were patterned using a wet chemical etch consisting of 132 ml of ethylene diamine, 21.12 g of pyrocatechol, and 17.6ml of water (EDP) at a temperature of 115°C, while the base mesas were patterned using a standard

CF₄/O₂ plasma etch. The EDP etch was used because the etch rate of p⁺ doped silicon is greatly reduced from that of n⁻ doped silicon[50]. The presence of germanium in the layer enhances this stopping characteristic even further[51]. The requirement of a good etch stop is essential as the Si_{1-x}Ge_x alloy layers used are nominally only 50nm thick. If this were not the case, any time spent overetching to ensure the removal of the entire Si emitter material over the extrinsic base could result in etching through the alloy layer in local places causing the base and collector to be short circuited once contacts are made, or at least a substantial thinning of the alloy, leading to increased extrinsic base resistance and premature punchthrough of the base-collector junction. The EDP was found to stop abruptly on the strained p-doped Si_{1-x}Ge_x alloy layers, even after overetching for more than a minute.

EDP is an anisotropic etch, etching the <111> directions much more slowly than the <001>. This helps reduce the undercut of the mask, if the mask is aligned to the substrate properly[50]. The only drawback of the EDP etch is it requires a silicon dioxide (SiO₂) mask rather than photoresist (ie. extra process steps are required). EDP etches SiO₂ at a rate of 0.3nm/min[50]. This, however, requires the wafers undergo a flash dip in 10% hydrofluoric acid (HF) prior to etching in EDP in order to remove the native oxide present in the etch window. This native oxide is sufficient to prevent etching for a substantial amount of time.

The result of a 1min, 20sec. EDP etch of the silicon cap on a Si_{1-x}Ge_x strained layer is shown in figure 3.4. Note the SEM micrograph reveals a smooth bottom of the etch window. The SiO₂ mask overhanging the etch window is also apparent, as the etch proceeds rapidly both horizontally and vertically until <111> directions or the p⁺ strained alloy layer are encountered.

Figure 3.4 SEM Micrograph of Si/Si_{1-x}Ge_x Layers After EDP Etch

3.3.3 Passivation of Si_{1-x}Ge_x Strained Layers

In order to avoid the limitations of the maximum thermal budget, the passivation of the exposed surfaces was performed simply by using a deposited oxide layer. It was hoped that having this in place during the later thermal cycles would provide some passivation of the exposed surfaces and junctions. This is by no means the best form of passivation in terms of reducing the density of surface states[52], however it a technique used by several other groups[42,47]. An alternative approach

was to rapid thermal anneal the sample, then deposit the CVD oxide[53]. Note the annealing ambient was not reported in [53], but it is presumed to have been an oxidizing ambient. This would result in a thin thermal oxide at the surface for passivation (the anneal cycle reported was 10 sec at 800°C) with a thicker deposited oxide on top for protection. This results in a compromise between low thermal budget and a passivation oxide of reasonable quality[53]. In the absence of these techniques however, a deposited oxide is the best solution. Note that the issue of germanium segregation is avoided using a deposited oxide.

3.3.4 Contact Metallization for the HBTs Under Study

The heavy doping levels in the $\text{Si}_{1-x}\text{Ge}_x$ strained alloy layer make contacting the base a straightforward issue. The light doping levels in the emitter and collector layers used here cause problems however. The doping level of the n-type layers must be high in order to obtain good ohmic contact to them. Typically this is done by a low energy, high dose implant into the contact window. Again the need to anneal resulting damage arises. The alternative approach used was to deposit an *in situ* phosphorus doped amorphous silicon (a-Si) film, with metal contacts subsequently made to this film. The deposition was carried out at 560°C by low pressure chemical vapour deposition (LPCVD). The resulting film was 50nm thick and the phosphorus doping level has been estimated by SIMS to be $4 \times 10^{20} \text{cm}^{-3}$.

In order to lower the sheet resistance of the n^+ a-Si film, a 5sec. rapid thermal anneal was carried out at 850°C in a nitrogen ambient. This thermal cycle was the maximum thermal cycle the wafers were exposed to during the entire fabrication run. It is believed to be within the stability criteria reported to date, so strain present in the as grown layers should not be relaxed. Note the resistivity of the now recrystallized polysilicon film was measured to be approximately $3 \text{ m}\Omega \cdot \text{cm}$, which is considerably higher than desired in a silicon BJT process.

Contacts were made to the $\text{Si}_{1-x}\text{Ge}_x$ base and the n^+ polysilicon film using 50nm of thermal evaporated titanium and a subsequent 500nm of aluminum. A SEM micrograph of a completed HBT is shown in figure 3.5. Note only base and emitter contacts are shown as for this device a backside contact was used.

Figure 3.5 SEM Micrograph of a Completed HBT

Chapter 4

Si_{1-x}Ge_x HBT Characteristics and Bandgap Measurements

4.1 Introduction

The results of the measurement of the dc characteristics at room temperature and the bandgap narrowing in the base of the Si_{1-x}Ge_x HBTs are given in the following sections. The measurements were performed on an HP4145 Semiconductor Parameter Analyzer. For the determination of bandgap narrowing, the wafers were mounted on a temperature controlled chuck with measurements taken over the temperature range 10°C to 120°C. Several devices from each wafer were analyzed, with devices having emitter stripe dimensions 20 μm by 110 μm being those typically used.

The saturation current density, J_{CO} , was determined by the extrapolation of the ideal region of the Gummel plot. The device operating temperature was determined from the slope of the same portion of the Gummel plot. In order to ensure accuracy in the measurements, a least squares fitting routine was used to obtain both values from the measured data. Excellent coefficients of fit were obtained. The bandgap narrowing was then determined from the dependence of J_{CO} on temperature, as discussed in Chapter 2.

The results of the room temperature measurements are presented in the following section while the bandgap narrowing results are given in the subsequent section. Discussion of the results and their implication is left to the next chapter.

4.2 Room Temperature Characteristics of the $\text{Si}_{1-x}\text{Ge}_x$ HBTs

The characteristics of the $\text{Si}_{1-x}\text{Ge}_x$ HBTs from the wafers listed in table 3.1 were measured with two major issues examined in detail. The I-V characteristics were examined to determine if a relationship existed between gain (h_{FE}) and collector current density (J_{CO}), and the germanium content of the layer. The HBTs were also tested to determine if any parasitic barriers were present in the conduction band[4.42-44]. This was done to determine if the bandgap measurements could be considered valid as these barriers tend to reduce the amount of bandgap narrowing seen in the electrical measurements.

Shown in figure 4.1 to 4.3 are the room temperature Gummel plots for HBTs typical to wafers 1034, 1035, and 1036 respectively. Note the poor base currents shown in all the Gummel plots. This nonideal base current was typical of devices from all the wafers. Also, the base current was seen to increase with increasing germanium concentration, which was not expected. (The base current should be determined by back injection into the emitter, and hence be independent of the Ge content in the base.) This current resulted in the poor gains seen for these transistors, with the maximum value being approximately 25. These values did not vary significantly for increasing Ge content, a result that was not expected. That is, the expected enhancement of gain, h_{FE} , with incorporation of germanium (predicted by equation 1.1) did not occur for these HBTs. It is now thought that the nonideal base current seen in the $\text{Si}_{1-x}\text{Ge}_x$ HBTs was due to contamination of the germanium source used in the MBE growth.

Despite the poor base currents, the collector currents were seen to be ideal over many decades of current in that the relation $J_C \propto J_{CO}e^{(qV/kT)}$ given in equation 2.27 was obeyed. The range of currents was typically $1 \mu\text{A cm}^{-2} < J_C < 1 \text{ A cm}^{-2}$ with high current performance restricted by the large parasitic resistance associated with the collector, which gives quasi-saturation. It should be noted that the collector

currents were seen to increase with increasing Ge content as anticipated by equations 2.1 and 2.26. This is shown explicitly in figure 4.4. This dependence forms the basis of the bandgap measurements of the following section.

The issue of potential barriers existing at the base–collector junction impeding carrier flow has been mentioned previously in the context of processing the wafers[also 4,42–44]. The common method for observing their presence is through offset voltages in the transfer curve of the HBT in question (ie. no collector current flows though there is a finite collector–emitter voltage and base current). The potential barriers result in a lower collector current when the base–collector bias does not eliminate them. Thus they are observed as an increase in collector current on a Gummel plot when the base–collector bias is increased. The results shown in figure 4.5 demonstrate the offset voltages are on the order of the step voltage used on the HP4145 for an HBT from wafer 1035. The same was found true for HBTs on wafers 1034 and 1036. Figure 4.6 shows the collector current versus base–emitter bias as per a Gummel plot, with the base–collector bias stepped, for an HBT on wafer 1034. Closer examination of the data showed little change in collector current in the high current regime with changes in base–collector bias. These results indicate an absence of any parasitic potential barriers in the conduction band. Note the high leakage current component of collector current appearing at low bias.

Thus the HBTs were seen to have ideal collector currents and no parasitic barriers to current flow, allowing the measurement of the bandgap narrowing to proceed.

Figure 4.1 Gummel Plot for HBT from Wafer 1034(8% Ge) ($V_{CB}=0V$)

Figure 4.2 Gummel Plot for HBT from Wafer 1035(15% Ge) ($V_{CB}=0V$)

Figure 4.3 Gummel Plot for HBT from Wafer 1036(20% Ge) ($V_{CB}=0V$)

Figure 4.4 Collector Current as a Function of Ge Content ($T=20^{\circ}\text{C}$) ($V_{CB}=0\text{V}$)

Figure 4.5 Transfer Characteristics for an HBT from Wafer 1035 (15% Ge)

Figure 4.6 I_C versus V_{BE} with V_{CB} Stepped for an HBT from Wafer 1034(8% Ge)

4.3 Determination of Bandgap Narrowing in the $p^+ \text{Si}_{1-x}\text{Ge}_x$ Base

I_{CO}/T^4 versus $1000/T$ is shown for all three wafer samples in figure 4.7 (note $I_{CO} = J_{CO}\text{Area}$). The value of E_{GO} was determined from a least squares fit to these data, using the relation described in equation 2.30. The bandgap narrowing is taken to be the difference between this value and 1.206 eV, the value of the bandgap of silicon extrapolated to 0 K [54]. Note that in the initial analysis of the data to extract the bandgap narrowing, the terms ΔE_{plif} and ΔE_{FD} as given in equations 2.30 to 2.34 were neglected. The implication of this on the results is left for discussion in the following chapter.

The theoretical value of the bandgap was calculated using equations 2.11 and 2.1, using the value of N_b equal to 1.5 as suggested in [36]. The theoretical bandgap narrowing along with the experimental data is shown in figure 4.8. Note the good agreement between the experimental data and the theory. There is a significant increase in bandgap narrowing observed due to the heavy doping as compared to the "undoped" case. Note that a doping level of $2 \times 10^{19} \text{ cm}^{-3}$ as for the fabricated devices gives approximately 90 meV of bandgap reduction, greater than the bandgap reduction due to 10% germanium incorporation. The errors associated with the measurements and their interpretation, along with consideration of the assumptions of the theory are discussed in the next chapter, but the first observation is the data and the theory are in fairly close agreement for the devices considered here under the given assumptions.

Figure 4.7 Plot of I_{CO}/T^4 versus $1000/T$ for 1034(8%), 1035(15%), and 1036(20%)

Figure 4.8 Bandgap Narrowing vs. Germanium Content

broken lines: theoretical calculation
solid line: experimental data

Chapter 5

Discussion

5.1 Introduction

The results given in the previous chapter were determined using several assumptions which warrant closer examination in some cases. Note that some of these were made in the absence of any data. In particular, the assumptions of the $\text{Si}_{1-x}\text{Ge}_x$ strained layer having the same temperature dependence of minority electron mobility in the base, μ_n , and the same temperature dependence of density of states in the valence band, N_v , as silicon are arbitrary. These are made on the basis that the band structure of bulk $\text{Si}_{1-x}\text{Ge}_x$ alloy is Si-like at low alloy concentrations. In the absence of data for strained layers, the same assumptions were made here. This may result in a considerable change in the determined value for bandgap narrowing.

The other issues that need addressing include the parameters used in the theoretical calculation of the bandgap narrowing, the neglecting of two terms in equation 2.30 during the initial analysis of the data, process variations that may impact the results (eg. cross wafer doping non-uniformity), and a standard of comparison for the results and theory to an existing bandgap narrowing model. These issues are examined in the following sections.

5.2 Issues Concerning the Theoretical Calculation of Bandgap Narrowing

There are two major assumptions in the calculation of the bandgap narrowing as outlined in section 2.2. The assumption that equations 2.1 and 2.11 could be superimposed giving valid results for bandgap narrowing in heavily doped $\text{Si}_{1-x}\text{Ge}_x$,

neglecting strain compensation due to the presence of boron is the first of these. The other is the use of a single parameter, N_b , to account for the effect of strain on the electron and hole effective masses, the dielectric constant, and the effect of splitting of the normally degenerate valence subbands on the interaction between light and heavy holes. (Note the strain effect on the bandgap narrowing is accounted for in equation 2.1.)

The assumption of superposition is validated, to first order, by the results shown in figure 4.8. Simply put, the extra bandgap narrowing from heavy doping must be added to that from strain in order to explain the data seen. Again the point needs to be made that at higher doping levels the strain from the boron incorporation may compensate the strain from the germanium incorporation completely. Note this strain reduction may already be accounted for here, as the Ge content was determined from XRD, a technique that measures the strain in the layer (ie. the lattice distortion) and correlates it to the presence of germanium. The nominal Ge concentrations were 10%, 20%, and 30% from the Ge flux and growth rate, whereas the measured concentrations were 8%, 15%, and 20%. However, the difference between the nominal and the measured germanium content was not constant from sample to sample. Since the doping level was constant ($N_A = 2 \times 10^{19} \text{ cm}^{-3}$ for all samples) implying a constant strain due to dopant incorporation for all samples, it is not believed this difference between the expected and achieved Ge concentrations was due to a systematic reduction in strain due to B incorporation. Rather, it is more likely a result of the difficulty of the layer growth.

The use of N_b as a fitting parameter stems from the fact that the valence subbands are no longer degenerate in the presence of strain. The factor was arbitrarily chosen to be 1.5, as suggested in [36], to account for the separation of the subbands. However, their proximity could not disallow their interaction. That is, the subbands are separated, but not enough so their interactions may be neglected.

This value was used in the initial calculations, as shown in figure 4.8. Shown in figure 5.1 is the sum of equation 2.1 and 2.11, with N_b set to 1.5 and to 2.0. Included for comparison is the experimentally measured data, as determined in the previous chapter. Note the extraordinarily good fit of the data to the theoretical case where the subband splitting is neglected (ie. $N_b=2.0$). This raises the question as to whether the use of $N_b=1.5$ was appropriate. Recall that in the analysis of the bandgap measurements, the effect of the subband splitting on the measurements was neglected, along with the effect of degeneracy. The fit of the $N_b=2.0$ case to the data as it appears in figure 4.8 (and figure 5.1) may then be considered a match between two quantities that are both not analyzed completely. An alternative viewpoint is the heavy doping in the base region has "recreated" the degeneracy of the strain split valence subbands. In this case $N_b=2.0$ would be the appropriate factor to use. However, this view does not account for corrections required due to the application of Fermi–Dirac statistics at the heavy doping levels used. This effect has been seen to result in an increase in bandgap narrowing over that predicted by Maxwell–Boltzman statistics[34] and at this juncture has not been applied to the data. An increase in bandgap narrowing would tend to point toward a reduced N_b being the appropriate factor to use in the calculation (such as 1.5). Note the use of an interpolation for N_b may be even more appropriate. The subband splitting begins in the case of degenerate subbands, where $N_b=2.0$ would be considered valid, (ie. there is no subband splitting when there is no strain) and increases with strain. This point has not been investigated here. For the remainder of the discussion, a value of $N_b=1.5$ is applied.

Thus the point of interest now focuses on the terms resulting from the need for Fermi–Dirac statistics and from the effect of subband splitting on the density of states in the valence band. These terms were previously neglected in the analysis, but form the basis for discussion in the following section.

Figure 5.1 Effect of Assuming no Valence Subband Splitting on the Theoretical Calculation of Bandgap Narrowing

5.3 Issues Concerning the Analysis of the Bandgap Narrowing Measurements

During the initial analysis, the values of the terms for ΔE_{split} and ΔE_{FD} in equation 2.31 were both neglected. This section discusses the validity of neglecting these terms. Their relevance to the overall result is discussed in the context of the magnitude of the experimental scatter of the data recorded.

Consider first the term for the additional bandgap narrowing resulting from the application of Fermi–Dirac statistics, ΔE_{FD} as given in detail below.

$$\Delta E_{\text{FD}} = k \frac{d \ln \left(\frac{\mathcal{F}_{1/2}(\eta_V)}{e^{\eta_V}} \right)}{d(1/T)} \quad (5.1)$$

From Blakemore[55],

$$\mathcal{F}_{1/2}(\eta_V) \approx \frac{e^{\eta_V}}{1 + \gamma e^{\eta_V}} \quad (5.2)$$

where $\gamma = 0.27$ is a constant, and η_V is defined as before (equation 2.17).

This gives

$$\Delta E_{\text{FD}} = k \frac{d \ln \left(\frac{1}{1 + \gamma e^{\eta_V}} \right)}{d(1/T)} \quad (5.3)$$

Differentiating equation 5.3, it can be shown that

$$\Delta E_{\text{FD}} = -\gamma \mathcal{F}_{1/2}(\eta_V) (E_V - E_F) \quad (5.4)$$

where

$$\mathcal{F}_{1/2}(\eta_V) = \frac{p}{N_V} \approx \frac{N_A}{N_V} \quad (5.5)$$

For $N_A = 2 \times 10^{19} \text{ cm}^{-3}$, and assuming $N_V = 1.04 \times 10^{19} \text{ cm}^{-3}$ as in silicon, then $\mathcal{F}_{1/2}(\eta_V) \simeq 2$. From Blakemore, $\eta_V \simeq 1.4$, giving $E_V - E_F \simeq 36 \text{ meV}$. Substituting in equation 5.4 gives $\Delta E_{\text{FD}} \simeq -19 \text{ meV}$. That is, the value of E_{go} determined by equation 2.30 would be overestimated by 19 meV by neglecting ΔE_{FD} , and the bandgap narrowing is increased by 19 meV when it is included. This is shown in figure 5.2 where the 19 meV has been added directly to the previous data, which is also shown for comparison.

Figure 5.2 The Effect of the Fermi-Dirac Correction on Bandgap Narrowing

broken lines: theoretical calculation
solid line: experimental data

Figure 5.3 shows the value of $\mathcal{F}_{1/2}(\eta_V)/\exp(\eta_V)$ versus temperature, plotted using both a numerical table look-up for $\mathcal{F}_{1/2}(\eta_V)$ (solid lines) and the approximation of equation 5.2 (dotted lines). The similarity between the two curves at $N_A = 2 \times 10^{19} \text{ cm}^{-3}$ shows the validity of this approximation over the range of η_V used in this case. Note the assumptions of N_V and its temperature dependence being the same as for silicon were made, as equation 5.5 was used to determine $\mathcal{F}_{1/2}(\eta_V)$. For both curves the value of η_V was interpolated using a table look-up and the relationship[55]

$$\begin{aligned} \eta &= \eta_0 + \Delta\eta \\ &= \eta_0 + \frac{\mathcal{F}_{1/2}(\eta) - \mathcal{F}_{1/2}(\eta_0)}{\mathcal{F}_{-1/2}(\eta_0)} - \frac{\mathcal{F}_{-3/2}(\eta_0) [\mathcal{F}_{1/2}(\eta) - \mathcal{F}_{1/2}(\eta_0)]^2}{2 [\mathcal{F}_{-1/2}(\eta_0)]^3} \end{aligned}$$

It was assumed the value of N_V is the same in $\text{Si}_{1-x}\text{Ge}_x$ strained layers as in Si. In actuality, $N_V|\text{Si}_{1-x}\text{Ge}_x$ is expected to be somewhat lower than $N_V|\text{Si}$, as the effective mass for holes is reported to be lower[42], resulting in a further increase in the bandgap narrowing, since the value for ΔE_{FD} varies inversely with N_V (equations 5.4, 5.5). However there is evidence that the fermi level lies closer to the valence band in heavily doped p-type silicon than would be predicted using the Si value of N_V [36]. This implies the value of $E_V - E_F$ would be lower than the 36 meV used here. This results in a reduction in the bandgap narrowing seen, implying less error in neglecting this term.

An alternative means of determining the impact of the Fermi-Dirac correction, ΔE_{FD} , on the bandgap narrowing is to determine the temperature dependence of $\mathcal{F}_{1/2}(\eta_V)/\exp(\eta_V)$ over the temperature range of interest. This is shown at various doping levels in figure 5.3, again with the assumption of the same temperature dependence of N_V in strained $\text{Si}_{1-x}\text{Ge}_x$ as in silicon, as was assumed for the bandgap extraction from the measurements. This dependence is then used to adjust the temperature dependence of the prefactor in the bandgap extraction,

Figure 5.3 $\mathcal{F}_{1/2}(\eta_V)/\exp(\eta_V)$ and $1/(1+ye^{\eta_V})$ versus Temperature at Various Doping Levels

$\mathcal{F}_{1/2}(\eta_V)/\exp(\eta_V)$ solid lines
 $1/(1+ye^{\eta_V})$ dashed lines

(currently assumed to be T^4) to give the correct value for E_{go} . This adjustment to the bandgap narrowing is ΔE_{FD} . The derivation of this method is as follows.

Recall, from equation 2.28,

$$J_{CO} = CT^4 \frac{\mathcal{F}_{1/2}(\eta_V)}{e^{\eta_V}} e^{\frac{-E_g}{kT}} \quad (5.6)$$

where C is a temperature independent constant. Assume $\mathcal{F}_{1/2}(\eta_V)/\exp(\eta_V)$ has a temperature dependence that can be described over the temperature range of interest by T^{m1} . Then equation 5.6 can be written in the form of equation 5.7.

$$J_{CO} = CT^4 \left[\frac{\mathcal{F}_{1/2}(\eta_V)}{e^{\eta_V}} \right]_{T=300K} \left(\frac{T}{300} \right)^{m1} e^{\frac{-E_g}{kT}} \quad (5.7)$$

This gives,

$$\left[\frac{\mathcal{F}_{1/2}(\eta_V)}{e^{\eta_V}} \right] / \left[\frac{T}{300} \right]^{m1} = \left[\frac{\mathcal{F}_{1/2}(\eta_V)}{e^{\eta_V}} \right]_{T=300K} \quad (5.8)$$

The right side of equation 5.8 is a constant, so a plot of $[\mathcal{F}_{1/2}(\eta_V)/\exp(\eta_V)]/(T/300)^{m1}$ versus temperature will have zero slope (ie. will be a horizontal line) when the correct value of $m1$ is used. (That is, when the temperature dependence of the numerator of the left side of equation 5.8 has been cancelled by the temperature dependence of the denominator.) This is shown in figure 5.4 where, over the temperature range of interest, 283K to 393K, the value of $[\mathcal{F}_{1/2}(\eta_V)/\exp(\eta_V)]/(T/300)^{m1}$ is approximately constant when $m1 \approx 0.7$, at a doping level of $2 \times 10^{19} \text{ cm}^{-3}$. Thus the temperature dependence of equation 5.7 becomes $T^{4+m1} = T^{4.7}$.

Recall equation 2.34 describing the impact of a change in the temperature dependence of the prefactor on the extracted value for bandgap, repeated as equation 5.9 for reference.

Figure 5.4 $[\mathcal{F}_{1/2}(\eta_V)/\exp(\eta_V)]/(T/300)^{m_1}$ versus Temperature at Various Doping Levels ($m_1=0.696$, $N_V=1.04 \times 10^{19} \text{ cm}^{-3}$)

$$\Delta E_g = \Delta m \cdot k \ln \left(\frac{T_2}{T_1} \right) \left(\frac{1}{T_2} - \frac{1}{T_1} \right)^{-1} \quad (5.9)$$

Here $\Delta m = m_1 = 0.7$, $\Delta E_g = \Delta E_{FD}$, and $T_{1,2}$ are 283K and 393K respectively. The resulting value for $\Delta E_{FD} = -20.03$ meV. This implies an additional 20.03 meV of bandgap narrowing are a result of the Fermi-Dirac correction required for this doping level.

By approximating equation 5.1 as a difference,

$$\begin{aligned} \Delta E_{FD} &\approx k \frac{\Delta \ln \left(\frac{F_{1.2} \eta_{1.2}}{c^{\eta_{1.2}}} \right)}{\Delta 1/T} \\ &\approx -22.4 \text{ meV} \end{aligned}$$

This numerical method gives the Fermi-Dirac correction, ΔE_{FD} , of 22.4 meV at $N_A = 2 \times 10^{19} \text{ cm}^{-3}$ and $T = 300\text{K}$. Note that the temperature steps were 5 °K, in the calculation of ΔE_{FD} (ie. $\Delta 1/T = 0.2$). This is shown graphically in figure 5.5. Note the sensitivity of the correction to the doping level. The value of the correction becomes very large as doping levels increase, approaching 100 meV at a doping level of $1 \times 10^{20} \text{ cm}^{-3}$. Figure 5.6 shows the superposition of the Fermi-Dirac correction determined through numerical differentiation on the data. The results are similar to those shown previously (figure 5.2). The potential cause of the apparent underestimation of the theoretical calculation of the bandgap narrowing from that measured experimentally is left for subsequent sections.

Consider now the other contribution to bandgap narrowing neglected during the initial analysis, that due to the strain induced splitting of the normally degenerate conduction and valence subbands, ΔE_{spln} . This term is expected to increase the total bandgap narrowing observed.

Figure 5.5 Fermi-Dirac Correction to Bandgap Narrowing, Plotted Against Temperature for Various Doping Levels, N_A

Figure 5.6 Effect of Modified Fermi-Dirac Correction on Bandgap Narrowing

broken lines: theoretical calculation
solid line: experimental data

Recall from section 2.3,

$$\Delta E_{split} = k \frac{d \ln f_{split}}{d(1/T)} = \frac{k}{f_{split}} \frac{df_{split}}{d(1/T)} < k \frac{df_{split}}{d(1/T)} \quad (5.10)$$

$$f_{split} = 1 + ae^{\frac{-\Delta E_C}{kT}} + be^{\frac{-\Delta E_V}{kT}} + abc \frac{-(\Delta E_C + \Delta E_V)}{kT} \quad (5.11)$$

where a,b, ΔE_C , and ΔE_V are as defined in section 2.2. Thus,

$$\frac{df_{split}}{d(1/T)} = -\frac{a\Delta E_C}{k} e^{\frac{-\Delta E_C}{kT}} - \frac{b\Delta E_V}{k} e^{\frac{-\Delta E_V}{kT}} - \frac{ab(\Delta E_C + \Delta E_V)}{k} e^{\frac{-(\Delta E_C + \Delta E_V)}{kT}} \quad (5.12)$$

Consider the first term in equation 5.12, $-(a/k)\Delta E_C e^{-\Delta E_C/kT}$. When ΔE_C is small, the term will be negligible as there will be little subband splitting. Likewise, when ΔE_C is large, $e^{-\Delta E_C/kT}$ will be small and again the term will be negligible. This case implies the subbands are split far apart and the split off subband is thinly populated. To find the value for ΔE_C where this term is at its maximum, set

$$\frac{d\left(\frac{a\Delta E_C}{k} e^{\frac{-\Delta E_C}{kT}}\right)}{d\Delta E_C} = 0 \quad (5.13)$$

This gives,

$$e^{\frac{-\Delta E_C}{kT}} \left(1 - \frac{\Delta E_C}{kT}\right) = 0 \quad (5.14)$$

This term is maximum at $\Delta E_C = kT$. So for the maximum ΔE_{split} due to the first term,

$$\begin{aligned} \Delta E_{split, \Delta E_C, \max} &= -a\Delta E_C e^{\frac{-\Delta E_C}{kT}} \\ &= -akTe^{-1} \\ &= -9.5 \text{ meV} \end{aligned} \quad (5.15)$$

assuming $a \approx 1$, and $T=300K$.

The same can be shown for the other terms in ΔE_{split} , $\Delta E_{split, \Delta E_V, \max}$ and $\Delta E_{split, (\Delta E_V + \Delta E_C) \max}$. By examining the data of People[14] it is seen that the valence

and conduction subbands are not split to $\Delta E_{C,V,C+V} = kT$ at the same germanium content. That is,

$$\Delta E_V \approx kT \text{ @ } x \approx 15\%$$

$$\Delta E_C \approx kT \text{ @ } x \approx 5\%$$

and their sum, $\Delta E_C + \Delta E_V \approx kT$ for x between these limits.

Due to the exponential nature of the terms involved, it is not expected that ΔE_{split} will exceed ≈ 10 meV by very much, despite the addition of the three terms. That is, one term may approach 10 meV, but the other two will be small. Remember also there is a division by f_{split} in the actual term for ΔE_{split} , the terms considered do not account for this and hence are considered maximum values.

The value of ΔE_{split} at a germanium content of 20% is given as an example (as per wafer 1036). From People[14], $\Delta E_C \approx 120$ meV, $\Delta E_V \approx 40$ meV, and assuming $a,b \approx 1$, this gives

$$\begin{aligned} \Delta E_{split} &= \frac{k}{f_{split}} \frac{df_{split}}{d(1/T)} \\ &= \frac{-\Delta E_C e^{\frac{-\Delta E_C}{kT}} - \Delta E_V e^{\frac{-\Delta E_V}{kT}} - (\Delta E_C + \Delta E_V) e^{\frac{-(\Delta E_C + \Delta E_V)}{kT}}}{1 + e^{\frac{-\Delta E_C}{kT}} + e^{\frac{-\Delta E_V}{kT}} + e^{\frac{-(\Delta E_C + \Delta E_V)}{kT}}} \\ &= -8.160 \text{ meV} \end{aligned}$$

Thus the value of ΔE_{split} in undoped $Si_{1-x}Ge_x$ strained layers is -8.16 meV at $x=20\%$. This value is superimposed on the data from the bandgap measurements, including the degeneracy effects calculated previously, and is shown in figure 5.7. Note this results in an increase in the discrepancy between the measured data and the theoretical calculation using $N_b=1.5$.

It is important to note that these two corrections are based on the assumptions concerning the temperature dependence of the density of states, N_V , and the values of a,b being approximately unity. However, further analysis to extract better values

Figure 5.7 The Combined Effects of the Fermi-Dirac Correction and the Correction for Subband Splitting

broken lines: theoretical calculation
solid line: experimental data

for these corrections becomes clouded by the accuracy of the data. The scatter of the data is 10 – 20 meV typically, values similar to those reported in the literature[33]. Both ΔE_{FD} and ΔE_{split} individually have values on this order at these doping levels. This then limits the utility of further refinement of these parameters to attempt to reduce the perceived errors in calculation or in analysis, as these parameters are already within the limits of the experimental scatter.

5.4 Processing Considerations

Two of the main processing concerns, strain relaxation and the existence of parasitic barriers to current flow, have already been addressed in previous chapters. Strain relaxation was avoided by keeping the thermal budget seen by the wafers low, both during layer growth and during HBT fabrication. The results of the room temperature measurements on the HBTs led to the conclusion that no parasitic barriers were present. The other main issue is the uniformity of the dopant and germanium profiles across a wafer and from wafer to wafer. MBE growth is typically very good for uniformity across the wafer. However, alterations from the nominal boron doping of $2 \times 10^{19} \text{ cm}^{-3}$ could theoretically be responsible for much of the scatter in the data for each wafer. A higher than nominal doping level would result in an increased amount of bandgap narrowing measured.

Figure 5.8 shows illustrates the effect of an assumed variation from the nominal doping level of $2 \times 10^{19} \text{ cm}^{-3}$. The data corrected with the 22.4 meV determined for ΔE_{FD} and the 8.2 meV determined for ΔE_{split} lies above but near the $3 \times 10^{19} \text{ cm}^{-3}$ doping line, while the uncorrected data fit well to a boron doping of $1.5 \times 10^{19} \text{ cm}^{-3}$. Thus the discrepancies between the measurements and the calculations could be due simply to a deviation in the desired doping level. The SIMS profile of wafer 1034, shown in figure 3.2, shows a doping level higher than the nominal $2 \times 10^{19} \text{ cm}^{-3}$. This was the case for all the wafer samples. The data as

Figure 5.8 Examination of an Assumed Deviation in Doping Level From the Nominal Value

corrected lie within experimental scatter of the higher doping level, but the corrections were made assuming the nominal doping level. At a doping level of $3 \times 10^{19} \text{ cm}^{-3}$ the Fermi–Dirac correction is expected to be higher than the 22.4 meV used, as shown in figure 5.9. This implies the theory and measurements are still only approximately in agreement.

5.5 Comparison of Theory and Measurements to an Accepted Model

For heavy doping in semiconductors, the bandgap narrowing has sometimes been calculated by the empirical formula of Slotboom (equation 5.16) [32,36].

$$\Delta E_g = K_1 \sqrt{\ln \frac{N}{N_0} + \left(\ln \frac{N}{N_0} \right)^2 + 0.5} \quad (5.16)$$

Here, N is the doping level, and for p-type silicon and $\text{Si}_{1-x}\text{Ge}_x$, $K_1 = 9.0 \text{ meV}$ and $N_0 = 1 \times 10^{17} \text{ cm}^{-3}$ [36].

This equation is plotted, with equation 2.1 superimposed to allow for strain due to germanium incorporation, in figure 5.10. The results of the bandgap measurements done in this thesis are presented also for comparison. The two theoretical calculations agree well once the values of ΔE_{split} and ΔE_{FD} appropriate to the doping level used are added to equation 5.16. It was noted in [36] that equation 5.16 gives the apparent bandgap narrowing seen by the device, not the actual bandgap narrowing.

The results of this comparison conclude the theory presented in [36] is accurate to within experimental fit when compared to the previously published empirical results, as long as both the effect of subband splitting and the use of Fermi–Dirac statistics are accounted for. This implies the discrepancies between the measured bandgap narrowing and the calculations lies in the analysis. The

Figure 5.9 Fermi-Dirac Correction versus Doping Level at 300 K

Figure 5.10 Comparison of the Theory of Jain and Roulston[36] with the Formula Presented by Slotboom[32]

unresolved points include the use of the correct doping level and the assumptions about the value of the physical parameters and their temperature dependences.

5.6 Summary of Discussion

The results of analyzing the data as per Neugroschel[34] without accounting for the need for Fermi-Dirac statistics or for the effect of subband splitting were found to match well with the theory in the case where the subband splitting was also neglected during the theoretical calculation (ie. $N_b=2.6$, as in figure 5.1)

Accounting for Fermi-Dirac statistics at the nominal doping level of $2 \times 10^{19} \text{ cm}^{-3}$ provided an increase in bandgap narrowing of approximately 20 meV. Applying this to the data resulted in fair agreement with the theoretical calculation for the case of the inclusion of subband splitting effects, with the theory underestimating the bandgap narrowing actually seen. Note that no correction to the data for the presence of subband splitting was included (see figures 5.2 and 5.6). The determination of the Fermi-Dirac correction factor was based on the assumption of the strained $\text{Si}_{1-x}\text{Ge}_x$ having the same density of states as silicon, which is most likely not true, particularly if there is significant subband splitting in the strained alloy layer[42]. Also, the temperature dependences of the densities of states for Si and strained $\text{Si}_{1-x}\text{Ge}_x$ were assumed to be the same.

The correction to the data required to account for the splitting of the degenerate subbands was determined to have a maximum value of approximately 10 meV and a value of 8.2 meV for 20% Ge incorporation, based on the calculations of People[14]. This lead to a greater discrepancy between the measured data and the theoretical calculations.

The empirical calculation of reference [36] was found to agree with the empirical results normally used to describe bandgap narrowing[32], when the

empirical results were adjusted to allow for the subband splitting and the use of Fermi-Dirac statistics. That is, the theory and empirical data were found to agree when the two were treated identically. The empirical formula for silicon and the theoretical calculation were both altered to mimic the $\text{Si}_{1-x}\text{Ge}_x$ strained alloy layers. This leads to the conclusion that there is an inconsistency either in the measurements or in the application of the theory.

The measurements were done on samples whose actual doping levels in the alloy layers were higher than expected, the result of this being that the calculations will underestimate the bandgap narrowing seen. When a higher doping level was used in the theoretical calculation, the agreement was better, as shown in figure 5.8. However, the value of the Fermi-Dirac correction applied to the data in this comparison was lower than should have been used, implying the theory was still slightly underestimating the bandgap narrowing.

Summarizing, the theory and the experimental data were consistent with each other. That is, they were found to be in reasonable agreement before and after allowing for the effects of subband splitting. Note that subband splitting is known to occur for lightly doped $\text{Si}_{1-x}\text{Ge}_x$ [14,42,43], but the presence of this effect has not been confirmed at high doping levels. The doping level in the samples was higher than expected, leading to the appearance of a poor fit of theory to data. This was resolved, but the theory still seems to slightly underestimate the data.

The preceding discussion focuses on experimental data and the closeness of a line of best fit through it compared to the theoretical calculation. One should note the data has a 10 to 20 meV scatter associated with it, which is on the order of the correction terms needed. Note this experimental scatter is typical for measurements of this type [33]. For this reason, further refinements of the correction factors were not attempted.

Chapter 6

Conclusions and Recommendations

6.1 Conclusions

As stated earlier, the purpose of this study was to determine the total bandgap narrowing occurring in the heavily doped p-type base region of Si/Si_{1-x}Ge_x/Si HBTs. For this purpose, a theoretical model for the dopant dependant bandgap narrowing presented in the literature[36] was applied to this material system. In particular, the presence of strain induced bandgap narrowing in the alloy layers was accounted for. The experimental technique normally used for determining bandgap narrowing in bipolar transistors was adjusted for application to the HBTs used here. This required accounting for the presence of strain induced splitting of the conduction and valence subbands. A process flow was developed for the fabrication of the HBTs such that the strain in the Si_{1-x}Ge_x layer was maintained. The HBTs were characterized, and measurement of the bandgap of the strained layer was performed. The results of these measurements were analyzed in order to determine the validity of the assumptions made in both the theoretical calculation and the extraction of the bandgap from the transistor I-V characteristics.

The results of the study lead to the following general conclusions. First, the theoretical calculation and the experimentally extracted data are in reasonable agreement when the same physical effects are accounted for in each. Secondly, the bandgap narrowing arising due to the incorporation of high levels of boron in the strained Si_{1-x}Ge_x layer is significant. This effect must be accounted for during transistor design, as the additional 90 meV bandgap reduction due to $2 \times 10^{19} \text{ cm}^{-3}$

boron incorporation will result in an increase in current gain by a factor of approximately 32 (see equation 1.1). The important conclusion is therefore the model as presented can be used as a useful tool in order to obtain an optimized Si/Si_{1-x}Ge_x/Si HBT structure, allowing for the benefits of both heavy doping and germanium incorporation to be utilized.

6.2 Recommendations

In order to obtain the fit to the data, the factors N_b , ΔE_{FD} , and ΔE_{spin} can each be applied in a seemingly arbitrary manner, though each has some basis for selection. The parameters end up being used to fit data to theory or theory to data as the case may be, leading to ambiguities as to the correct manner for their application. For example, the use of $N_b=1.5$ to account for all the strain effects is rather empirical. It may be that this value is too high, and a better fit between theory and corrected measurements may be obtained if a slightly smaller number, or some form of interpolation, were used instead. Strain dependent effective masses of the carriers should be used to account for some of the strain effects, rather than the "number of equivalent valleys," N_b . (Recall the effective masses used in the theoretical calculation were those of silicon for electrons and a linear interpolation between that of silicon and germanium for the holes.) However, data regarding the effective masses in strained Si_{1-x}Ge_x is presently unavailable, as is data concerning the densities of states for the various valence and conduction bands and subbands.

The presence or absence of subband splitting in the presence of heavy doping has not been entirely resolved, as the agreement between theory and data is similar for both cases at the doping level investigated. If heavy doping does indeed remove or reduce the effect of the splitting of degenerate subbands, then examining the bandgap of less heavily doped samples along with some unstrained (ie. silicon only) samples may reveal this.

The importance of the use of Fermi-Dirac statistics has been demonstrated. The doping level in question gave a large value for ΔE_{FD} , based arbitrarily on the density of states for silicon. The effect should be more readily observed at higher doping levels ($>1 \times 10^{20} \text{ cm}^{-3}$) than it was here. By using different doping levels, some idea of the correct density of states to use should become apparent. Note the density of states is expected to be dependent on the germanium content of the sample[42]. Thus a larger sample set, especially including a variation in doping levels as well as control samples having no germanium, should be investigated. Fitting the theory to a larger matrix of samples would aid in resolving the ambiguities present. This is the primary recommendation resulting from this study of bandgap narrowing.

By going through this exercise, it may be possible to determine some values for some of the parameters used to describe carrier motion which are presently unknown for strained $\text{Si}_{1-x}\text{Ge}_x$ alloys of either doping type. These include the density of states of various bands and subbands along with their respective temperature dependences. From this comes the effective masses and mobility of the carriers. Note that before accurate values may be determined, the scatter of the data will have to be improved. The scatter of the data was seen in this study to be on the order of or larger than some of the effects being inserted as correction factors. This would preclude drawing conclusions based on the values of these factors, and any attempt to do so could result in inaccuracies or misinterpretation.

The determination of the bandgap, along with the other parameters, leads to modelling the HBTs in order to predict their performance. The modelling of HBTs having $\text{Si}_{1-x}\text{Ge}_x$ base regions has been attempted[56, for example]. The absence of accurate values for these parameters has limited the optimization of the doping and composition profiles required to optimize the performance of the HBTs, in particular the frequency response. Note that the additional bandgap narrowing observed in this study due to the doping levels alone will result in a different

optimization of the device structure than if it were not present, as mentioned in the previous section.

This additional reduction in bandgap of approximately 90 meV at a doping level of $2 \times 10^{19} \text{ cm}^{-3}$ is a greater bandgap reduction than for an incorporation of 10% Ge into the lattice. This has direct implications to the design and fabrication of HBTs. The tradeoffs for base resistance, f_T , f_{max} , and punchthrough voltage are affected as now there is another source of bandgap reduction, allowing the germanium profile to be somewhat different. The additional bandgap narrowing due to dopant incorporation allows for less germanium, giving greater thermal stability of the layer. However, care must be taken that the strain resulting from the dopant incorporation does not compensate the strain due to the presence of germanium.

Thus further work in the area should be pursued in order to lead to a greater understanding of the semiconductor material as needed for application to HBTs or other device structures.

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Appendix A

Derivation of Equation 2.34

Recall the collector saturation current density can be described by equation A-1 (equation 2.38),

$$J_{CO} = \frac{kT\mu_n (N_C N_V)_{eff}}{W_B N_A} \frac{\mathcal{F}_{1/2}(\eta_V)}{e^{\eta_V}} e^{\frac{-E_g}{kT}} \quad (A-1)$$

Regrouping the temperature independent terms into a constant, C , recalling $(N_C N_V)_{eff} = N_{C1} N_{V1} f_{phl}$, and assuming a power dependence for $N_{C1, V1}$, f_{phl} , and the Fermi-Dirac correction, $\mathcal{F}_{1/2}(\eta_V)/\exp(\eta_V)$, on temperature is valid over the range of temperatures used, equation A-1 can be rewritten as

$$J_{CO} = C T^m e^{\frac{-E_g}{kT}} \quad (A-2)$$

Rearranging and taking the logarithm, equation A-3 is obtained. Note that $\log_{10}$ is used, as this is the value commonly plotted. Conversion to natural logarithms is straightforward.

$$\log_{10} \left(\frac{J_{CO}}{T^m} \right) = \log_{10} e \left(\frac{-E_g}{kT} \right) + \log_{10} C \quad (A-3)$$

Assuming the relationship $E_g = E_{g0} - \alpha T$ is valid near room temperature, as in chapter 2 and taking derivative with respect to $1/T$, equation A-4 is obtained.

$$\frac{d \log_{10} \left(\frac{J_{CO}}{T^m} \right)}{d(1/T)} = - \log_{10} e \left(\frac{-E_{g0}}{k} \right) \quad (A-4)$$

Approximating the differentiation with a difference, gives

$$E_{g0} = \frac{-k}{\log_{10} e} \left[\frac{\log_{10} \left(\frac{J_{CO}}{T^m} \right)_2 - \log_{10} \left(\frac{J_{CO}}{T^m} \right)_1}{1/T_2 - 1/T_1} \right] \quad (A-5)$$

where 1,2 are the boundaries where the approximation is valid. Rearranging,

$$E_{g0} = \frac{-k}{\log_{10} e} \left[\frac{\log_{10} \left(\frac{J_{c02}}{J_{c01}} \right) - m \log_{10} \left(\frac{T_2}{T_1} \right)}{1/T_2 - 1/T_1} \right] \quad (\text{A-6})$$

The desired relationship between the change in E_{g0} and a change in the exponent, m , is then obtained by taking the derivative of A-6 with respect to m .

$$\frac{dE_{g0}}{dm} = \frac{-k}{\log_{10} e} \left[\frac{-\log_{10} \left(\frac{T_2}{T_1} \right)}{1/T_2 - 1/T_1} \right] \quad (\text{A-7})$$

Approximating the derivative as a difference once more,

$$\frac{dE_{g0}}{dm} = \frac{\Delta E_{g0}}{\Delta m} = \frac{E_{g02} - E_{g01}}{m_2 - m_1} \quad (\text{A-8})$$

$$\Delta E_{g0} = \Delta m \frac{k}{\log_{10} e} \left[\frac{\log_{10} \left(\frac{T_2}{T_1} \right)}{1/T_2 - 1/T_1} \right] \quad (\text{A-9})$$

which is the desired result. Following the analysis instead using natural logarithms gives,

$$\Delta E_{g0} = \Delta m k \ln \left(\frac{T_2}{T_1} \right) \left(\frac{1}{T_2} - \frac{1}{T_1} \right)^{-1} \quad (\text{A-10})$$

which is equation 2.34 as before.

Appendix B

Process Flow for HBT Fabrication

1. Substrate: silicon, 1/4 – 4” diameter wafer with SiGe Base layer

1. Cut 1/4 – 4” wafer such that it fits in the LOTOX boat.
2. Number all test wafers
3. RCA clean all wafers

2. Lotox Deposition (undoped) Temp. = 405 °C

Notes:

1. Use recipe number: 1
2. Include bare silicon test wafer for deposited oxide thickness evaluation
3. Deposition rate approx. 17.5 nm/min
 - SiH₄ flow : 25.0%
 - O₂ flow : 27.5%
 - Deposition time: 5.5 min.

target thickness: 0.10 µm

3. Photolithography – Emitter Mesa Mask level #1

Notes:

Make sure mask is accurately aligned to wafer flat

Mask: CU-078-10

- Positive photoresist (PROTECT BACKS)
 - Flood coat and spin all wafers
 - Prebake
 - Expose wafer to mask CU-078-10 (5.5 sec.)
 - Develop
 - Inspect at 100x (minimum)
 - Postbake
 - Etch in Siloxide Etchant @ 20 °C (hydrophobic)
 - Expected etch time: 30 sec.
- Actual etch time: 1 min.
- Strip P.R. in Caro's acid

4. Silicon Etch

Notes:

1. Use EDP etching solution

- Ethylene diamine - 132 ml
- Pyrocatechol - 21.12 g
- Deionized water - 17.6 ml

2. Dip wafer in 10% HF for 5 sec. to remove native oxide just prior to etch in EDP

- Heat EDP solution to 115 °C
- Etch wafer in EDP for 1 min 10 sec

Expected silicon etch depth: 0.6 μm

- Rinse in D.I.
- Strip remaining SiO_2 in HF

5. Photolithography - Base Mesa

Mask level #2

Notes:

1. Bake wafers at 200 °C for 1 hr. prior to coating with resist

Mask: CU-078-02

- Positive photoresist
- Flood coat and spin all wafers
- Prebake
- Expose wafer to mask CU-078-02
- Develop
- Inspect at 100x (minimum)
- Postbake
- Plasma etch Mesa structure in the plasma etcher using $\text{CF}_4/5\%\text{O}_2$ at 100W and 100 mTorr
- Expected etch depth: 0.10 to 0.20 μm
- Expected etch time: 1 to 2 min.

Actual etch time: 40 sec.

- Strip P.R. in Caro's acid

6. Lotox Deposition

(undoped)

Temp. = 405 °C

Notes:

1. RCA clean wafers prior to LOTOX deposition
2. Use recipe number: 1
3. Include bare silicon test wafer for deposited oxide thickness evaluation
4. Deposition rate app.ox. 17.5 nm/min

- SiH₄ flow : 25.0%
- O₂ flow : 27.5%
- Deposition time: 16 min.

target thickness: 0.30 µm

7. Photolithography - Emitter Window

Mask level #3

Mask: CU-078-03

- Negative photoresist
- Flood coat and spin all wafers
- Prebake
- Expose wafer to mask CU-078-03
- Develop
- Inspect at 100x (minimum)
- Protect backs
- Postbake
- Etch Emitter window in Siloxide Etchant @ 20 °C (hydrophobic)
- Expected etch time: 1 min. 40 sec.

Actual etch time: 2 min. 15 sec.

- Strip P.R. in Caro's acid

8. Emitter Amorphous Silicon Deposition

Notes:

1. RCA clean (but no HF) wafers prior to run
2. Use recipe number: 2
3. Phosphorus doped LPCVD amorphous silicon
4. Deposit the amorphous silicon immediately after a 5 min. dilute (100:1) D.I. : HF etch
5. Include 0.1 µm gate oxide test wafer

- 6. SiH_4 : PH_3 flow ratio: 160:1
- 7. include test wafer for diode evaluation

- SiH_4 MFC flow setting: 35.0%
- PH_3 MFC flow setting: 26.9%
- Deposition time: 90 min.

target thickness: 0.60 μm

9. Photolithography - Emitter Amorphous Si

Mask level #4

Notes:

Prebake wafers at 200 °C for 1 hour unless wafers taken directly from LPCVD system

Mask: CU-078-04

- Positive photoresist
- Flood coat and spin all wafers
- Prebake
- Expose wafer to mask CU-078-04
- Develop
- Inspect at 100x (minimum)
- Postbake
- Etch silicon emitter in wet poly etching solution $\text{HF}:\text{HNO}_3:\text{H}_3\text{PO}_4$ D.I.
- Expected etch time: 10 to 11 sec.

Actual etch time: 12 sec.

- Strip P.R. in Caro's acid
- Rinse in D.I. for a minimum of 10 min.
- Inspect

10. Emitter Anneal - RTA

Notes:

1. RCA clean wafers (no HF)
 2. Anneal wafers immediately after a 5 min. etch in 100:1 D.I.:HF solution
- Anneal in RTA annealer for 5 sec. in N_2
 - Temp = 850 °C

11. Metallization -**Back Contact****E-beam Aluminum**

Notes:

1. Etch back of wafers in HF until hydrophobic
2. Load pellets of aluminum in e-beam crucible
3. Use process #1, film #1

- Load wafers onto planets
- Load planets into vacuum system
- Evacuate system to 1×10^{-6} torr
- Evaporate 1.0 μm of aluminum on backs

12. Metal Anneal -**Hydrogen**

Temp. = 400 °C

Gas

- Load wafers into quartz boat
- 5 min. preheat in carrier
- 5 min. purge in carrier
- 20 sec. PUSH to center
- 20 min. anneal
- 20 sec. PULL from center
- 5 min. cool in carrier
- 10 min. cool in carrier
- Unload

N₂H₂H₂H₂H₂H₂N₂**13. Photolithography -****Base Contact Windows**

Mask level #5

Notes:

Prebake wafers at 200 °C for 1 hour

Mask: CU-078-07

- Negative photoresist (PROTECT BACKS)
- Flood coat and spin all wafers
- Prebake
- Expose wafer to mask CU-078-07
- Develop
- Inspect at 100x (minimum)
- Postbake

2 OF 2

PM-1 3 1/2" x 4" PHOTOGRAPHIC MICROCOPY TARGET
NBS 1010a ANSI/ISO #2 EQUIVALENT

PRECISIONSM RESOLUTION TARGETS

- Etch Contact windows in siloxide etchant
- Expected etch time: 2 min. 20 sec.

Actual etch time: 2 min. 15 sec.

- Strip P.R. in Micro Strip (J-100)
- Rinse in D.I. for 30 min.
- Inspect

14. Metallization - Front Thermal Ti/Al

Notes:

1. Etch wafers in BD etchant until scribe channels go hydrophobic (approx. 40 sec.) then D.I. rinse
 2. Load 2 Tungsten filaments with Titanium wire in the Electrode #1 position
 3. Load 2 Tungsten filaments with 2 pellets of Aluminum in each filament in the Electrode #2 position
 4. Use Process #4, Film #4 for Titanium evaporation
 5. Use Process #2, Film #2 for Aluminum evaporation
 6. The Titanium and Aluminum evaporations are done without breaking vacuum between evaporations
- Load wafers onto planets
 - Load planets into vacuum system
 - Evacuate system to 1×10^{-6} torr
 - Evaporate 0.05 μm of Titanium on fronts
 - Switch to electrode #2 and evaporate 0.5 μm of Aluminum

15. Photolithography - Metal Mask level #6

Notes:

Prebake wafers at 200 °C for 1 hour unless wafers taken directly from vacuum system

Mask: CU-078-09

- Positive photoresist (PROTECT BACKS)
- Flood coat and spin all wafers
- Prebake
- Expose wafer to mask CU-078-09

- Develop
- Inspect at 100x (minimum)
- Postbake
- Etch aluminum in H_3PO_4 at 60 °C
- Expected etch time: 1 min. 40 sec.

Actual etch time: 1 min. 35 sec.

- Etch titanium in $\text{H}_2\text{O} - \text{NH}_4\text{OH} - \text{H}_2\text{O}_2$ (5:1:1) at 75 °C
- Expected etch time: 15 sec.

Actual etch time: 25 sec.

- Strip P.R. in Micro Strip 2001 @ 110 °C for 10 min.
- Rinse in D.I. for a minimum of 30 min.
- Inspect at 100x (minimum)

END

0 8-1 1-94

FIN