Epitaxy and characterization of SiGeC layers grown by reduced pressure chemical vapor deposition

Licentiate Thesis
by
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Stockholm, Sweden
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Epitaxy and characterization of SiGeC layers grown by reduced pressure chemical vapor deposition

A dissertation submitted to the Royal Institute of Technology, Stockholm, Sweden, in partial fulfillment of the requirements for the degree of Teknologie Licentiat.

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Abstract

Heteroepitaxial SiGeC layers have attracted immense attention as a material for high frequency devices during recent years. The unique properties of integrating carbon in SiGe are the additional freedom for strain and bandgap engineering as well as allowing more aggressive device design due to the potential for increased thermal budget during processing. This work presents different issues on epitaxial growth, defect density, dopant incorporation and electrical properties of SiGeC epitaxial layers, intended for various device applications.

Non-selective and selective epitaxial growth of Si$_{1-x-y}$Ge$_x$C$_y$ (0 \leq x \leq 0.30, 0 \leq y \leq 0.02) layers have been optimized by using high-resolution x-ray reciprocal lattice mapping. The incorporation of carbon into the SiGe matrix was shown to be strongly sensitive to the growth parameters. As a consequence, a much smaller epitaxial process window compared to SiGe epitaxy was obtained. Different solutions to decrease the substrate pattern dependency (loading effect) of SiGeC growth have also been proposed. The key point in these methods is based on reduction of surface migration of the adsorbed species on the oxide. In non-selective epitaxy, this was achieved by introducing a thin silicon polycrystalline seed layer on the oxide. The thickness of this seed layer had a crucial role on both the global and local loading effect, and on the epitaxial quality. Meanwhile, in selective epitaxy, polycrystalline stripes introduced around the oxide openings act as migration barriers and reduce the loading effect effectively. Chemical mechanical polishing (CMP) was performed to remove the polycrystalline stripes on the oxide.

Incorporation and electrical properties of boron-doped Si$_{1-x-y}$Ge$_x$C$_y$ layers (x=0.23 and 0.28 with y=0 and 0.005) with a boron concentration in the range of 3 \times 10^{18}-1 \times 10^{21}$ atoms/cm$^3$ have also been investigated. In SiGeC layers, the active boron concentration was obtained from the strain compensation. It was also found that the boron atoms have a tendency to locate at substitutional sites more preferentially compared to carbon. These findings led to an estimation of the Hall scattering factor of the SiGeC layers, which showed good agreement with theoretical calculations.

Keywords: Silicon germanium carbon (SiGeC), Epitaxy, Chemical vapor deposition (CVD), Loading effect, High resolution x-ray diffraction (HRXRD), Hall measurements, Atomic force microscopy (AFM).
List of appended papers

**Paper A**

*Growth of high quality epitaxial SiGeC layers by using chemical vapour deposition*

**J. Hållstedt**, E. Suvar, P. O. Å. Persson, L. Hultman, Y. -B. Wang och H. H. Radamson


**Paper B**

*Methods to reduce the loading effect in selective and non-selective epitaxial growth of SiGeC layers*

**J. Hållstedt**, E. Suvar, C. Menon, M. Östling och H. H. Radamson


**Paper C**

*Strain and electrical characterisation of boron-doped SiGeC layers grown by chemical vapor deposition*


**Other contributions by the author not included in the thesis:**

*The effect of carbon and germanium on phase transformation of nickel on Si$_{1-x-y}$Ge$_x$C$_y$ epitaxial layers*

**J. Hållstedt**, M. Blomqvist, P. O. Å. Persson, L. Hultman och H. H. Radamson


*Formation of As- or P-doped polycrystalline Si layers grown by RPCVD for emitter application in SiGeC-based Heterojunction Bipolar Transistors*

E. Suvar, E. Haralson, **J. Hållstedt**, H. H. Radamson och M. Östling

*Physica Scripta* (2004), accepted for publ.
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On a more personal level I would like to send my gratitude to every “Bergsman” here in Kista. Moreover, thanks to everyone else who has supported me in my struggle towards my Licentiate degree. Finally, I would like to dedicate this thesis to Therese.

Julius Hållstedt, Stockholm, March 2004
<table>
<thead>
<tr>
<th>Acronyms</th>
<th>Description</th>
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<tr>
<td>AFM</td>
<td>Atomic Force Microscopy</td>
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<td>As</td>
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<td>AsH₃</td>
<td>Arsine</td>
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<td>B</td>
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<td>B₂H₆</td>
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<td>BJT</td>
<td>Bipolar Junction Transistor</td>
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<td>CMOS</td>
<td>Complementary MOS</td>
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<td>CMP</td>
<td>Chemical Mechanical Polishing</td>
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<td>CVD</td>
<td>Chemical Vapor Deposition</td>
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<td>C</td>
<td>Carbon</td>
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<td>CTE</td>
<td>Coefficient of Thermal Expansion</td>
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<td>Bandgap</td>
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<td>HBT</td>
<td>Heterojunction Bipolar Transistor</td>
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<td>High-Resolution X-Ray Diffraction</td>
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<td>HRRLM</td>
<td>High-Resolution Reciprocal Lattice Mapping</td>
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<td>MB</td>
<td>Matthew and Blakeslee theory</td>
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<td>MBE</td>
<td>Molecular Beam Epitaxy</td>
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<td>MFC</td>
<td>Mass Flow Controller</td>
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<td>Metal Oxide Semiconductor</td>
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<td>NSEG</td>
<td>Non-Selective Epitaxial Growth</td>
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<td>Selective implanted collector</td>
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<td>Transient enhanced diffusion</td>
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<td>VLSI</td>
<td>Very Large Scale Integration</td>
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<td>X-Ray Diffraction</td>
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Chapter 1

Introduction

1.1 Background and motivation

Since the invention of the first transistor the evolution of microelectronics has been truly remarkable. In the mid-sixties the famous Moore’s law was proposed [1] which predicted an exponential development of the number of transistor on a chip. This law was later extended to a number of other relations e.g. speed of devices. This rapid technological progress is the result of successful research in a wide range of different areas where the material perspective has become increasingly important.

Silicon (Si) materials are the dominant choice for mass production in microelectronics due to mature and well developed processing technology which lead to low process costs in combination with high rate of device integration. In the high speed vicinity, III-V heterostructure devices showed extremely impressive results compared to the best performance in Si technology [2]. However, high cost and incompatibility with the Si platform are the main drawbacks for this technology.

Heterostructures in Si technology, i.e. SiGe, have attracted considerable attention in the recent decades for high frequency applications [3]. The possibility of having bandgap engineering by tuning the Ge-amount in SiGe can be easily applied for different device applications [4-6]. The research on Si/SiGe heterostructures is extended to the late seventies when the struggle of SiGe epitaxial growth was initiated by molecular beam epitaxy (MBE) technique [7]. A decade later the first SiGe heterojunction bipolar transistors (HBT) were manufactured [8], which soon showed cut-off frequencies beyond several 10 GHz. Almost ten years later the first commercial SiGe-based circuits were available [9]. At this point the technology had matured and relatively cheap industrial chemical vapor deposition (CVD) reactors for large wafers were available [10].

Carbon (C) in Si technology has been considered as a contamination element during many years due to the creation of numerous defects [11]. In the early nineties, however, the addition of C to Si and SiGe was proposed and realized through different techniques [12]. The small size of C atoms enables the opportunity to create a tensile strain in Si-based materials creating a band off-set in the conduction band [13, 14]. Another application of C in SiGe is the influence on dopant out-diffusion, especially on boron (B) [15]. In this field implementation of C in SiGe layers has increased the temperature budget of the device process and paved the way for more aggressive device design. This is beneficial e.g. for selective implanted collector (SIC) layers. This process results in a transient enhanced diffusion (TED) of B in the SiGe base layer due to an overwhelming number of generated point defects [16, 17]. The out-diffusion of B to the adjacent layers degrades the high
The frequency behavior of the transistors [18]. The main benefit of integrating C in SiGe layers is the reduction of TED [19]. The other benefit is an increase of thermal stability of SiGeC layers due to strain compensation [20]. On the other hand, the diffusion of other dopants e.g. phosphorous (P) or arsenic (As) in the HBT structures can be affected and more attention about preserving the dopant profile is needed [21]. New designs of HBTs involving SiGeC technology have demonstrated cut-off frequencies of 350 GHz which is approaching to the values provided by III-V technology [22].

1.2 Aim of thesis

The purpose of this study is to provide knowledge about implementing carbon in Si or SiGe matrix from aspects of device integration. The growth and characterization of doped and un-doped Si$_{1-x-y}$Ge$_x$C$_y$ layers deposited by the Chemical Vapor Deposition (CVD) technique is presented. This thesis concludes the impact of the growth parameters, pattern dependency of the epitaxy process, incorporation of boron and carbon in SiGe and the electrical behavior of these layers.
Chapter 2

Si/SiGeC as a material system

2.1 Chemical and structural properties

Si, Ge and C are all group IV element materials that crystallize with covalent bonds in the diamond lattice, which is constituted of two face centred cubic (fcc) matrices displaced by a quarter of the space diagonal. The lattice parameter for Si, Ge and C is 5.43, 5.66 and 3.56 Å, respectively [23]. Si and Ge can be alloyed over the whole compositional range showing no intermediate phases [24], whereas C alloying is more complicated. By looking at the binary phase diagram of Si and C, it is clear that the only stable phase for carbon above the equilibrium solubility (i.e. $10^{17}$ cm$^{-3}$ at the melting point) is stoichiometric silicon-carbide [25-26], see figure 2.1. However, previous studies have reported that up to 5% C can be incorporated in Si$_{1-x}$C$_x$ thin films grown on Si substrates [27-28]. This is several orders of magnitude larger than the equilibrium solubility value. Therefore, all layers considered in this thesis are in a metastable position and special care has to be taken in order to avoid silicon-carbide formation during growth or post annealing treatments. The solubility of carbon in Ge is even lower than in Si around $10^8$ cm$^{-3}$ at the melting point. Moreover, since the Ge/C phase diagram show no stable intermediate phases, any formation of GeC crystallites can be out ruled [29].
2.2 Si/SiGeC lattice mismatched systems

Strain and mismatch definition

When the layer has different lattice constants compared to the substrate, a mismatch system is created. It is a well known fact that Si based materials alter the lattice parameter when atoms with different sizes are incorporated substitutionally [30]. This creates the possibility to generate layers with lattice constant both larger and smaller than the Si substrate. Depending on the composition both compressive and tensile strain can be obtained if the bonding between layer and substrate is at least partially coherent. A compressive strain occurs in the growth of $\text{Si}_{1-x}\text{Ge}_x$ on Si due to larger lattice constant of $\text{Si}_{1-x}\text{Ge}_x$ than Si. This causes an elongation of the lattice constant along the growth direction (out-of-plane). On the contrary, in a $\text{Si}/\text{Si}_{1-y}\text{C}_y$ system a tensile strain is obtained due to smaller lattice constant of $\text{Si}_{1-y}\text{C}_y$ than Si. This causes a shrinkage of the lattice parameter in the growth direction. By combining these two effects it is clear that a totally strain compensated system can be formed. A visualization of these effects can be seen in figure 2.2. The relaxation behaviour of these systems are also schematically illustrated. A number of parameters need to be defined in order to characterise a strained epitaxial layer [31]. The coordinate system is defined in figure 2.2 where $z$ denotes the out-of-plane direction and $x$ and $y$ denotes the in-plane direction.

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**Figure 2.1 Phase diagram of Si and C [25].**
**Figure 2.2** Schematic presentation of lattice distortion in Si/SiGeC systems.
The lattice mismatch is defined by the following relation:

\[ f = \frac{a_{\text{lay}}}{a_{\text{sub}}} - 1 \]  

(2.1)

Similarly the in-plane and out-of-plane components of the lattice mismatch are defined as following:

\[ f_z = \frac{a_{\text{lay}}}{a_{\text{sub}}} - 1 \quad f_{x,y} = \frac{a_{\text{lay}}^{x,y}}{a_{\text{sub}}} - 1 \]  

(2.2)

For a totally strained layer in-plane lattice parameter is the same as the substrate which results in \( f_{x,y} = 0 \). Then the strain (\( \varepsilon \)) in the layer is described by the relaxed and strained lattice parameter perpendicular and parallel to the surface through the following definition:

\[ \varepsilon_z = \frac{a_{\text{lay}}}{a_{\text{layer}}} - 1 \quad \varepsilon_{x,y} = \frac{a_{\text{lay}}^{x,y}}{a_{\text{layer}}} - 1 \]  

(2.3)

The relation between the relaxed and strained layer is determined by the elastic compliances through the poisson ratio (\( \nu \)) which for an isotropic material as well as for [001] geometry is described by [32]:

\[ \frac{\varepsilon_{x,y}}{\varepsilon_{x,y} - \varepsilon_z} = \frac{1-\nu}{1+\nu} \]  

(2.4)

Now it is possible to express the total mismatch as a function of the out-of plane and in-plane contribution \( f_z \) and \( f_{x,y} \):

\[ f = \frac{1-\nu}{1+\nu} (f_z - f_{x,y}) + f_{x,y} \]  

(2.5)

For most materials the poisson ratio is in the range \( \nu = 0.3 \pm 0.1 \). It has been shown that for SiGe grown on Si substrates \( \nu = 0.278 \). C incorporation in SiGe layers is assumed to have only a minor effect on this poisson ratio value [33]. By measuring the mismatch components by e.g. HRXRD (described in appendix A) and using the formulas above the strained and relaxed lattice parameters may be obtained. The relaxation amount is then defined as:

\[ R = \frac{f_{x,y}}{f} \]  

(2.6)
Relaxation of SiGeC layers

Strained materials have by definition an elastic energy hidden in the lattice, which can be released totally, or partially when this energy is too large, i.e. when the layer thickness exceeds the so-called critical thickness. The relaxation of energy may result in the creation of misfit dislocations, which degrade the electrical properties. One of the early conceptually and simple models that predicted the critical thickness for strained layers was the Matthew-Blakeslee (MB) theory [34], see figure 2.3. This theory is based on the equilibrium between two forces: the force that is due to the lattice mismatch and the tension force that is created by the dislocation. Theoretically equilibrium critical thickness ($t_{\text{equ}}$) for different Ge amount can thus be obtained. However, the experimental data demonstrated that thicker strained SiGe layers ($t_{\text{exp}}$) than $t_{\text{equ}}$ could be grown without any relaxation. The failure of MB theory is due to the ignorance of a number of aspects, e.g. the interaction, multiplication and propagation of misfit dislocations. More complicated non-equilibrium models could describe the relaxation behavior more accurately [37].

It should be emphasized here that the onset of relaxation is gradual and significant relaxation is only obtained for films with thickness greatly exceeding the critical thickness. SiGe layers with thickness between $t_{\text{equ}}$ and $t_{\text{exp}}$ are defined as metastable (see figure 2.3). Incorporation of small atoms, e.g. carbon or boron, compensate the strain in SiGe matrix and increases the critical thickness of the epitaxial layers. Recent results have also shown that the growth of SiGe layers on patterned substrates behave differently. Thick SiGe layers can be grown selectively without any relaxation, meanwhile, the non-selectively grown layers show a poor stability. The part of selectively grown SiGe layers close to the oxide edges are more relaxed compared to the central part. This causes a strain distribution in SiGe layers in oxide openings which extends the critical thickness. On the contrary, in the non-selective case the polycrystalline surrounding part induces defects into the crystalline SiGe layer. Thus, figure 2.3 is inapplicable for the SiGe layers grown on patterned substrates [38].

Another important point that has to be mentioned here is that the relaxation of thin SiGe layers may appear through surface roughening, not formation of the misfit dislocations [39]. This roughening that often appears as a sinusoidal perturbation enables the lattice planes to relax towards the roughness peaks. The roughness results in an increase in surface energy, which balances the relaxation.

![Figure 2.3](image-url)  
*Figure 2.3 Critical thickness values predicted by MB theory [34] and experimental data for SiGe layers grown at 550 °C [35] and 750 °C [36].*
Calculation of composition in SiGeC layers

As has been pointed out above, strained pseudomorphic layers will experience different lattice parameters in different directions depending on the amount of alloying components. Therefore by measuring the lattice parameter the strain can be calculated and thus also the composition. For a binary alloy e.g. Si_{1-x}Ge_x it is fairly easy to calculate the composition using Vegard’s law or some other polynomial expression [40]:

$$a(\text{Si}_{1-x}\text{Ge}_x) = a(\text{Si})(1-x) + a(\text{Ge})(x) - 0.02393x + 0.01957x^2 + 0.00436x^3, \text{ for all } x \quad (2.7)$$

In this relation the first linear terms correspond to Vegard’s law and the other are higher order correction terms. For a ternary SiGeC compound a strain compensation effect will occur due to the C atoms. This strain compensation amount can be converted to the substitutional C concentration. It has to be mentioned here that the Ge amount in SiGeC layers is not affected by the presence of C as confirmed by our results (see figure 4.2) and other groups [33, 41, 42]. For a ternary compound Vegard’s law yields, a Ge:C ratio of 8.2, see equation 2.8. Every C atom will therefore in this assumption compensate 8.2 Ge atoms:

$$a(x, y) = (1-x-y)a_{\text{Si}} + xa_{\text{Ge}} + ya_{\text{C}} \quad (2.8)$$

where \(a(x,y)\) is the lattice parameter of the alloy and \(a_{\text{Si}}, a_{\text{Ge}}\) and \(a_{\text{C}}\) are the bulk lattice parameters of the subscripted components. In a similar way this method is applicable for estimating the substitutional B concentration in SiGe layers [43-44].

2.3 Thermal stability

Strained SiGe films can be grown both stable (in MB equilibrium region) and metastable as described previously. SiGe layers in MB region do not relax (diffusion may occur) whereas the films in the metastable zone relax e.g. by post-annealing treatments. Si_{1-y}C_y or Si_{1-x-y}Ge_xC_y layers with carbon concentration exceeding the solid solubility \(\sim 10^{17} \text{ cm}^{-3}\) are always metastable. These films are thermally stable up to \(\sim 800 \, ^\circ\text{C}\) for long time annealing or oxidation as shown in figure 2.4 where the C out-diffusion is significant. When rapid thermal annealing treatment (RTA) is used the temperature window is expanded to almost 1000 \( ^\circ\text{C}\). Further increased annealing temperatures will beside out-diffusion of C also lead to the formation of silicon-carbide precipitations.
2.4 Electrical properties of SiGeC layers

Bandstructure

Si and Ge have a bandgap of 1.11 and 0.66 eV, respectively. Both Si and Ge are indirect semiconductors meaning that the valence band maximum is located at the \( \Gamma \) point and conduction band edge at some other \( k \) value. For Si the conduction band has a six-fold degeneracy with ellipsoidal minima located at around 0.85\( \chi \) along the \( \Gamma - \chi \) direction, denoted \( \Delta \). Ge has the conduction band edge located at the L point. For the diamond structure there exist eight L points but only four are independent since the opposite points are connected to each other [23]. The valence band of Si and Ge are similar with degenerated light (lh) and heavy hole (hh) band at the \( \Gamma \) point, whereas, the spin orbit (so) band has a 44 meV or 290 meV split from the valence band edge for Si and Ge respectively. The band diagram of Si and Ge are shown in figure 2.5 [46].

**Figure 2.4** X-ray rocking curves illustrating the thermal stability of Si\(_{0.994} Ge_{0.006} \) layers with a) 10s RTA and b) long-time oxidation treatment [45].

**Figure 2.5** Band diagram of a) Si and b) Ge, figure from [46].
A composition of Si and Ge will have a bandgap between the elemental values. Previous reports have stated that the SiGe alloy stays Si-like up to approximately 85% Ge [47]. Compressive strained layers will induce further modifications on the band structure. In the conduction band the six fold $\Delta 6$ valley degeneracy is replaced by a lowering of the in-plane $\Delta 4$ valleys whereas the perpendicular $\Delta 2$ valleys are raised. In this case the increased filling of the $\Delta 4$ valleys will result in a lower electron mass perpendicular to the growth direction. In the valence band the strain will cause a splitting of the degenerated hh, lh and so band with the hh band defining the valence band edge and so shifted downwards. Figure 2.6 a) illustrates these effects schematically. An empirical result of the bandgap value for a strained SiGe on Si (100) substrate layer is given by eq. 2.9 [48, 49]:

$$E_g = 1.107 - 0.896x - 0.396x^2, \quad x < 0.3$$  \hspace{1cm} (2.9)

Incorporation of carbon in Si matrix will induce tensile strain which surprisingly decreases the bandgap of Si$_{1-y}$C$_y$ alloy according to relation 2.10 [50]:

$$E_g = 1.17 - (6.5 \pm 0.3)y$$  \hspace{1cm} (2.10)

The strain splits the degenerated $\Delta 6$ valleys with a lowering of the $\Delta 2$ valleys in the growth direction, which provides a repopulation of electrons to these valleys. This might in fact increase the mobility of in-plane electrons since the effective mass in this direction is the small transversal mass. In the valence band the degenerated lh and hh hole bands are splitted with the lh band defining the valence band edge. This is illustrated in figure 2.6 b). The exact effect of carbon in fully or partially strain compensated SiGeC layers is not well known. Mainly the splitting is believed to decrease as the strain is compensated [51-52].
Fig. 2.6 Upper part of figure illustrates schematically the surfaces of constant energy in $k$-space for the conduction band in case of different strain. The lower part shows the degeneracy of conduction and valence band for different induced strain in Si structures [51-52].
Chapter 3

Chemical vapor deposition

3.1 Epitaxy

The word “epitaxy” is of Greek origin meaning “arranged upon” and is defined as the growth of a single crystalline layer arranged as the single crystalline bulk substrate. The mechanism behind epitaxy is that atoms can find a lower energy state as condensed matter aligned to the substrate. Defects will distort the system which will lead to an increase in the internal energy of the system, thus enabling the growth of extremely high quality epitaxial layers nearly defect free within the limits of the substrate quality. There are some major advantages of epitaxial growth compared to other techniques e.g. implantation, diffusion or bonding. An abrupt doping profile with low defect density can be obtained. Moreover, epitaxy can be performed at low temperatures, which solves many out-diffusion problems.

There are many techniques available to manufacture high quality epitaxial layers. Many of these techniques can only be used in research applications due to low production rate and vulnerable process steps. Among these research-oriented techniques are Molecular Beam Epitaxy (MBE) and Pulsed Laser Deposition (PLD) which are very good for development of new materials. For industrial applications, however, chemical vapor deposition (CVD) has come to play the most important role. CVD uses chemical vapors that are stable in room-temperature. Therefore, no deposition occurs on surrounding room-temperature surfaces. However, on the heated substrate the vapors decompose and react to create a film. CVD and especially reduced pressure CVD (RPCVD) provide a good trade-off between quality of the grown layer and deposition rate. Moreover, with this technique it is possible to simultaneously etch and deposit. By appropriate adjustment of the growth parameters deposition will only occur on selected Si areas on patterned substrates. This operation is referred to as selective epitaxial growth (SEG) and will be discussed in greater detail below. The normal operation is thereby defined as non-selective epitaxial growth (NSEG). These issues are in fact the dominating reasons for the advantages of RPCVD technique [31].

3.2 ASM Epsilon CVD reactor

The samples in this study were grown in an ASM epsilon 2000 single wafer RPCVD reactor (see figure 3.1). The system consists of four main parts: nitrogen purged load-locks, wafer handling section, process quartz chamber and gas exhaust cleaning section. A Bernoulli arm transfers the samples from the load-lock to the susceptor in the process chamber. The susceptor area is heated by a series of halogen tungsten lamps from both the top- and downside which, enable growth and rapid thermal adjustments in the range 550-1200 °C. The
pressure in the chamber may range from atmospheric pressure down to 15 Torr. The connected process gases are the following: SiH₄ and SiCl₂H₂ as the Si sources, 10% Ge in H₂ as Ge source and 1% SiH₃CH₃ in H₂ as C source. The dopant gases are B₂H₆, AsH₃ and PH₃ as B, As and P sources diluted in H₂ to 1% concentration. In order to achieve extremely low partial pressures of the dopant sources, i.e. to create low doping, dilution lines are used to decrease the partial pressure of the dopant gases. In this way partial pressures in the range ~10⁻⁶ – 10⁻¹ mtorr may be obtained. H₂ and N₂ are used as carrier and purge gases in the reactor. In this reactor, HCl is used for etching the quartz chamber or as a process gas in selective epitaxy. Pneumatic valves are used to control the on/off switching of the gas pipe. Both HCl and H₂ have an in-line purifier unit in order to reduce O₂ contamination in the epitaxial layers. The exhaust gases from the process chamber will be neutralized in a scrubber and later extinguished in a burner (not shown in the picture). All wafers are chemically cleaned (see next session) prior to loading to Epsilon reactor and they are purged for 30 min in the load-lock in order to avoid the O₂ contamination transfer to the process chamber [53].

\[ \text{Figure 3.1 Outline of ASM Epsilon 2000 CVD reactor from [53].} \]
3.3 Substrate cleaning

An epitaxial growth process is extremely sensitive to surface contamination, thus a reliable cleaning method is crucial in order to obtain a high quality epitaxial layer. In this study all substrates were chemically ex-situ cleaned. First an oxidizing step was performed for at least 5 min in a mixture of H₂SO₄ and H₂O₂ (2.5:1) at 120 °C. Thereafter, the formed oxide was stripped by a solution containing 50% HF, C₃H₇OH and de-ionized (DI) H₂O (1:1:100) for 100 sec. at room temperature. After each step the wafers were rinsed for five minutes in N₂ bubbled DI H₂O. Finally the wafers were spin dried and inserted into the load locks. The in-situ cleaning consists of a 2 min bake at 1050 °C in H₂ for blanket wafers and 5-20 min bake at 900-950 °C for patterned substrates.

3.4 SiGeC epitaxy

Chemical reactions

Si deposition occurs in RPCVD by using hydrogen or chlorine based silicon sources. It would be valuable to briefly discuss the steps in the chemical reaction:

Silane-based epitaxy

\[
\begin{align*}
\text{SiH}_4(g) + 2 _- &\leftrightarrow H + \text{SiH}_3 & (3.1) \\
\text{SiH}_3 + _- &\leftrightarrow H + \text{SiH}_2 & (3.2) \\
2\text{SiH}_2 &\leftrightarrow \text{H}_2(g) + 2\text{SiH} & (3.3) \\
2\text{SiH} &\leftrightarrow \text{H}_2(g) + 2\text{Si}(_\text{film}) + 2_- & (3.4)
\end{align*}
\]

Dichlorsilane-based epitaxy

\[
\begin{align*}
\text{SiH}_2\text{Cl}_2 + 5_- &\leftrightarrow \text{Si} + 2\text{H} + 2\text{Cl} & (3.5) \\
\text{SiH}_2\text{Cl}_2(g) &\leftrightarrow \text{SiCl}_2 + \text{H}_2(g) & (3.6) \\
\text{SiCl}_2 + 3_- &\leftrightarrow \text{Si}(_\text{film}) + 2\text{Cl} & (3.7) \\
\text{H} + \text{Cl} &\leftrightarrow \text{HCl}(g) + 2_- & (3.8) \\
2\text{Cl} + \text{Si} &\leftrightarrow \text{SiCl}_2 + 3_- & (3.9)
\end{align*}
\]

where “ _- ” denotes an incorporation site on the surface [33, 54]. In these reactions, the species SiH₂ and SiCl₂ are very reactive. In general, the growth rate is limited by hydrogen desorption at low temperatures since this process is thermally activated and limited by the supply of reactants at high temperatures [11]. Germane (GeH₄) and/or methylsilane (SiH₃CH₃) are then added to create Si₁₋ₓGeₓ, Si₁₋ₓCₓ or Si₁₋ₓ₋ₐGeₓCₓ layers. The germane radicals are created analogously to the silane based reactions, meanwhile the main radical for C incorporation (silylene) is created through the following reactions [33]:

\[
\begin{align*}
\text{SiH}_2 + \text{SiH}_3\text{CH}_3(g) &\leftrightarrow \text{SiH}_4(g) + \text{SiCH}_4 & (3.10) \\
\text{SiCl}_2 + \text{SiH}_3\text{CH}_3(g) &\leftrightarrow \text{SiH}_2\text{Cl}_2(g) + \text{SiCH}_4 & (3.11)
\end{align*}
\]

In these reactions SiH₂ or SiCl₂ is needed and a SiH₄ or SiH₂Cl₂ molecular is generated. SiCH₄ is thus very sensitive to the fraction of SiH₃CH₃ or SiH₂/SiCl₂. A high partial pressure of methylsilane will result in C species that cannot be frozen into substitutional sites and nano-crystallites may be formed acting as the contamination [55]. Furthermore, if the SiH₂ or SiCl₂ amount is too low the reactions 3.10 and 3.11 will be distorted creating defects in the layer.
Growth kinetics

As we have discussed before from the binary Si/C phase diagram $\text{Si}_{1-y}\text{C}_y$, films are always meta-stable. This raises several issues in the growth of layers containing C. First of all, the growth parameters must be set so that the deposition condition is far away from thermodynamic equilibrium where silicon carbide is formed (i.e. at low temperatures). Once deposited though, the $\text{Si}_{1-y}\text{C}_y/\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ films appear to be able to keep crystallinity and composition up to significantly higher temperatures than the deposition temperature, see chapter 2.3. Furthermore, due to the small size of the C atom, interstitials are easily formed. Although, the bulk solubility of C in Si is extremely low, during thin film growth it is the surface solubility that plays the most important role in C incorporation. It could be up to four orders of magnitude larger than the bulk value [56]. The increased solubility results from the fact that a surface offers additional degrees of freedom for the C atoms. This renders the possibility to incorporate a larger amount of C substitutionally. As the films grow thicker the C adatoms become rapidly buried on subsurface sites, and are quickly immobilized. The activation energy needed to leave the lattice is quite high (3.04 eV) [55], which makes the films rather stable. This solubility is also increased weakly with $\text{GeH}_4$ partial pressure and then saturates as illustrated in fig. 3.2. The saturation is due to the high growth rate, which contributes to the defect generation in the epitaxial layer. A XTEM image of such layers from the saturation region is shown in figure 3.3.

**Figure 3.2** The impact of the $\text{GeH}_4$ partial pressure on the carbon incorporation.

**Figure 3.3** XTEM figure illustrating the effect of increasing Ge amount in SiGeC layers.
After optimization of all growth parameters for silane-based epitaxy a process window for growing high quality epitaxial SiGeC layers is obtained as illustrated in figure 3.4. In chlorine-based epitaxy a low growth rate is obtained due to the etching of Si leading to a much smaller process window for selective epitaxy.

3.4 SiGeC growth on patterned substrates

Recent studies have demonstrated that the growth on patterned substrates differs from growth on blanket wafers [57-63]. This difference in the profile of the epitaxial layer originates from Si/SiO$_2$ filling ratio over the wafer (so-called global effect) and the opening size (so-called local effect). These effects in SiGeC layers appear as variation of growth rate, Ge and C amount and also severe changes in morphological appearance. This pattern dependency poses a serious problem in the applying of epitaxy process in device technology.

This local and global loading effect can be understood by the thermal and chemical behavior of the growth. The thermal effect depends essentially on the emissivity of the substrate, whereas the chemical effect is generated by the difference in diffusion length of the species as well as surface reaction mechanisms. In principle, the global effect can be measured by having a test chip in the mask with a large opening for SIMS or x-ray measurement and compare this experiment with a blanket reference sample.

Local loading effect is a more critical subject since the device openings are remarkably smaller than the test opening. There is the possibility to use AFM to measure the height of the epitaxial layer in the oxide opening. However, the Ge amount which is the main factor for the bandgap structure cannot be determined by these analyses. By designing a pattern with an array of small openings and applying HRXRD measurements on a large number of identical windows sufficient diffraction intensity can be collected to measure the strain and Ge content in the small openings.
Loading effect in non-selective epitaxy of SiGeC layers

In this process, the above-mentioned arguments about the local and global loading effect are also the origin of variation of both Ge amount and growth rate. As a consequence the carbon incorporation will be varied since it is dependent on the Ge amount. Introducing a polycrystalline Si seed layer on the oxide can change both the emissivity and the lateral diffusion of the species. Figure 3.5 a) and b) show AFM images of an opening with and without a seed layer. Note that the surface on the oxide is rough due to nucleated species. Figure 3.6 shows the impact of the Si seed layer thickness on Ge and C content in Si$_{0.8}$Ge$_{0.2}$ and Si$_{0.786}$Ge$_{0.21}$C$_{0.004}$ layers in the test chip (global effect) and the device openings (local effect) compared to blanket reference sample. Strain compensation occurs between Ge and substitutional C atoms in SiGeC matrix, thus the measured strain values correspond to apparent Ge amount in SiGeC layers. The figure exhibits that SiGeC layers grown on patterned substrates without any seed layer experience high local loading effect. In fact, the surface diffusion length of adsorbed species decreases when depositing a polycrystalline seed layer, since the species adsorbed on the polycrystalline Si layer are preferably incorporated on the polycrystalline layer. A too thin seed layer cannot efficiently reduce the loading effect, which also leads to a poor quality polycrystalline SiGeC layer. This is important for the device performance since this polycrystalline layer is the extrinsic base in a HBT structure. Nevertheless, AFM results show that a thick seed layer results in a rough single crystalline SiGeC layer. The reason for this surface roughness is injection of defects from the poly-epi transition line. The growth rate of SiGeC layers in small and big openings (not presented in the figure) are overlapped to the blanket wafers by employing an appropriate seed layer thickness. An optimum of 50-100 Å has to be employed for the SiGeC layers in order to obtain layers free of loading effect and a minimum defect density.

Figure 3.5 AFM amplitude images showing a deposited SiGe in a 10×10 µm$^2$ opening and surroundings a) without and b) with poly seed layer.
Loading effect in selective epitaxy of SiGeC layers

In this growth, two competing processes occur i.e. growth and etching. This enables the opportunity to tune the growth and etching by adjusting the HCl partial pressure so that all formed nuclides on the oxide are etched and still some growth takes place in the opening. Meanwhile, the species on the oxide surface are still diffusitive towards the opening. This may stimulate the idea of having polycrystalline stripes around the oxide openings that will act as migration barriers for these species in a similar way as in the non-selective growth case. The area of these stripes is an almost negligible fraction of the Si/SiO₂ coverage ratio, thus no change in the global loading effect was observed. Figure 3.7 shows that the amount of local loading effect increases with increasing \( P_{Ge} \) and a reduction of ~25% was obtained in all SiGe samples by introducing polycrystalline stripes.

**Figure 3.6** Influence of seed layer thickness on the Ge concentration in SiGe and SiGeC NSEG layers.

**Figure 3.7** The pattern dependency on the loading effect in SEG epitaxial layers.
The integration of these stripes in the device process is also an important point. A chemical-mechanical polishing (CMP) was applied to remove the polycrystalline stripes after the epitaxial growth. Since the thickness of the epitaxial SiGeC layers were well below the oxide thickness, no surface damage was generated by CMP step. This shows that preparing the substrates with polycrystalline stripes can successfully be integrated in device fabrication. This method requires only one extra lithography step and it is more effective compared to the methods based only on tuning the growth parameters.

Figure 3.8 AFM amplitude images illustrating in a) opening with polycrystalline stripe, b) the opening after CMP treatment.
Chapter 4

Dopant incorporation and electrical characterisation

P-type doping of SiGeC layers

Highly p-doped layers are used as the base layer in advanced HBTs. Boron has evolved as the most suitable p-dopant due to the high solubility and process related reasons. Recently carbon has been integrated in these devices to overcome out-diffusion related problems [15, 19, 64-66]. The major mechanism in boron diffusion is the kick out process which is caused by the exchange of Si self-interstitials and substitutional boron atoms. The carbon atoms can consume these interstitials, which leads to a rapid decrease of boron diffusion. This is beneficial for processes which result in TED. The integration of both carbon and boron in SiGe layers adds a number of problems. Since both boron and carbon introduce tensile strain in the lattice (i.e. radius smaller than Si) they are believed to compete for the same lattice positions. For low contents the simultaneous incorporation of boron and carbon poses no major problems. However as the concentration increases the boron has a tendency to locate at the substitutional sites more preferentially compared to C, see figure 4.1.

![Figure 4.1 Effect of B on the substitutional C concentrations.](image-url)
It should be noted that the total carbon amount is constant and consequently more interstitial carbon must be formed. SIMS results in figure 4.2 illustrate that the presence of carbon in B-doped SiGe layers has no influence on Ge and B amount. The active dopant concentration and the mobility are two important electrical parameters for device application. An easy and fast measurement is the Hall measurement using the van der Pauw geometry (described in appendix A.3.) which can measure Hall carrier concentration and Hall mobility. In SiGe layers a number of factors are different compared to Si e.g. the valence band curvature has changed due to the strain induced by the Ge incorporation. This causes a change in scattering mechanisms in SiGe and we cannot apply the well-known Si Hall scattering factor ($r_H$) directly. By comparing Hall measurements with an independent technique that can measure the active dopant concentration, $r_H$ may be obtained.

In order to be active dopants the B atoms must locate at substitutional sites in the SiGe lattice, thus, compensating the strain. Then by measuring the strain compensation amount by applying a technique e.g. HRXRD (described in Appendix A.1) the active concentration can be calculated, see section 2.2. By comparing the HRXRD values with active boron concentration values obtained by Hall measurement $r_H$ for SiGe layers can be obtained, see figure 4.3. In this figure the theoretically calculated $r_H$ values are also illustrated. Our extracted values are consistent with the theoretical values [67].

**Figure 4.2** SIMS showing effect of C and B on the Ge concentration in a multilayer-sample.
Figure 4.3 Experimental and theoretical Hall scattering factor for different boron concentrations. Theoretical data from [67].
Chapter 5

Summary, conclusions and future outlook

This thesis deals with the epitaxy and characterization of SiGeC layers grown by reduced pressure chemical vapor deposition. The research has been focused on different aspects of the integration of carbon e.g. defect density and the electrical properties of the layers. It has been shown that a high carbon amount (several orders of magnitude higher than the equilibrium solubility) can be incorporated substitutionally in the SiGeC matrix (up to ~2% in selective epitaxy). However, the incorporation was observed to be extremely sensitive to the growth parameters. The process window for epitaxy of SiGeC layers was very small compared to e.g. SiGe epitaxy. No enhancement in electron mobility was obtained in Si$_{1-y}$C$_y$ layers even though an accurate optimization of growth parameters was performed (Paper A).

The pattern dependency (loading effect) of SiGeC epitaxy has also been studied and methods to decrease this problem have been proposed (Paper B). These methods are based on decreasing the flow of formed species on the oxide towards the oxide openings. As migration barriers polycrystalline stripes around the oxide openings were applied in selective epitaxy. These stripes reduced the local loading effect significantly and after epitaxy they were removed by using chemical mechanical polishing. In non-selective epitaxy a polycrystalline seed layer on the oxide with thickness of 50-100 Å was necessary to eliminate the pattern dependency of the epitaxial process.

The incorporation of B in SiGe layers was studied by using high-resolution reciprocal lattice mapping (HRRLM) and Hall measurements (Paper C). The results showed that the growth rate increased by increasing the boron concentration and this had no influence on Ge amount. The strain compensation amount induced by boron in SiGe matrix was measured by HRRLMs and thereafter, these values were converted to substitutional B concentration. By comparing the dopant concentration obtained from these analyses with Hall measurement the Hall factor for SiGe layers for different boron concentrations was estimated in good agreement with theoretical values.

The incorporation of B in SiGeC layers has also been investigated. The presence of B had no influence on the carbon atomic concentration in SiGeC layers. Our investigations showed that the boron atoms in SiGeC layers locate preferably at substitutional sites in contrary to carbon atoms at both substitutional and interstitial sites. The substitutional carbon concentration decreased with increasing boron amount in SiGe matrix. Since the atomic concentration of carbon was constant then the number of carbon interstitials is believed to be increased in presence of boron.
**Future outlook**

The future of high performance SiGe-based devices looks very promising. The formation of strained or relaxed high quality SiGe or Si layers on SOI or other types of substrates are extremely important in order to push the device performance further. More interpretation of C in device performance and characteristics has to be studied. Finally, new applications of C in SiGe-based materials have to be explored.
Appendix

Characterization techniques

A.1 X-Ray Diffraction (XRD)

W. L. Bragg made the first explanation of x-ray diffraction from crystals with relation A1, the well-known Bragg law:

\[ 2d \sin(\theta) = n\lambda \]  

where \(d\) is the distance between parallel planes, \(\lambda\) is the wavelength of the x-rays and \(\theta\) is the angle of the incident beam to the surface. Suppose there is an incoming wave to a crystal material, then the first plane will only reflect about \(10^{-3}\) to \(10^{-5}\) of the incident radiation. This means that up to \(10^5\) planes can contribute to the total diffracted radiation in a perfect crystal. In this case, the diffracted beams are obtained when reflections from parallel atom planes interfere constructively. However, this constructive interference will only occur at certain values of \(\theta\). Moreover, the Bragg law is limited by the condition \(\lambda \leq 2d\), this is the reason why x-rays with small values for \(\lambda\) (Cu \(K_\alpha = 1.5406 \text{ Å}\)) are used.

**High Resolution X-Ray Diffraction (HRXRD)**

The XRD equipment that has been used in this study is the Philips X’pert system see figure A1. A copper target X-ray tube with tungsten filament at 45 kV bias yields beside the continuous spectrum the characteristic Cu \(K_\alpha\) and \(K_\beta\) radiation where the latter is easily filtered with slits. A monochromator with four Ge (220) crystals is needed to erase the Cu \(K_{\alphaII}\) x-rays and this set-up also decreases the divergency of the incident beam to 12 arcsec. By applying a mirror and an aperture in the primary optics an intense beam with a spot size varied between \(100 \times 100 \ \mu m^2\) to \(10 \times 10 \ mm^2\) can be obtained. By inserting two Ge (220) crystals in front of the detector the divergency of the diffracted beam can be decreased.
Rocking curve measurement

Scanning the incident angle, \( \omega \) with fixed 20 (\( \omega \) scan) probes different orientations of planes with a given plane separation, while a simultaneous scan of \( \omega \) and 20 with a rate ratio of 1:2 (\( \omega \)-20 scan) probes different orientations of planes with a given direction. A \( \omega \) or \( \omega \)-20 scan is called a rocking curve. A \( \omega \)-20 scan of an epitaxial Si/SiGe(C) structure generates two peaks: the Bragg condition is satisfied for both Si and SiGe(C). The composition can be obtained from the split angle of the layer peak from the substrate. X-rays may interfere inside the layer due to internal reflection between interfaces. This interference causes optical fringes around the layer peak. The number of fringes depends mostly on the quality of the interface. Extended defects such as dislocations and stacking faults affect the interfacial quality of the epitaxial layer by roughness. This means that the first sign of defects in the layer is a decreasing number of fringes in the XRD scan. By measuring the distance (difference in angle) between two neighboring fringes (\( 2\Delta\theta \)) the thickness \( t \) can be calculated by the following relationship:

\[
2\Delta\theta \cos(\theta) \cdot t = \lambda
\]

For multilayer samples the interpretation is somewhat more difficult and simulations are required.

Reciprocal lattice mapping

A high-resolution reciprocal lattice mapping (HRRLM) can be obtained by performing \( \omega \)-20 scans for a range of incident angles \( \omega \pm \Delta\omega \). HRRLM at the asymmetrical (113) reflection is very sensitive to reveal the defect generation. The incident beam in these HRRLMs is 2.81°, which provides the possibility to measure a large volume of the layer revealing any minor defects (or relaxation). It makes this type of measurement very attractive.
for performing an accurate optimization of the growth parameters in a state-of-the-art epitaxial process compared to the ordinary rocking curve measurements.

In these maps, the position of the layer peak relative to the substrate peak provides the mismatch parameters and the feature of the peaks illustrates the defect density. The out-of-plane and in-plane lattice mismatches can be calculated from data obtained from HRRLMs from an asymmetrical reflection:

\[ f_z = \frac{\Delta a_z}{a_{sub}} = \frac{a_{lay} - a_{sub}}{a_{sub}} = \frac{\sin \theta_{sub} \cos(\omega_{sub} - \theta_{sub})}{\sin \theta_{lay} \cos(\omega_{lay} - \theta_{lay})} - 1 \]  

(A3)

\[ f_{x,y} = \frac{\Delta a_{x,y}}{a_{sub}} = \frac{a_{lay}^{x,y} - a_{sub}}{a_{sub}} = \frac{\sin \theta_{sub} \sin(\omega_{sub} - \theta_{sub})}{\sin \theta_{lay} \sin(\omega_{lay} - \theta_{lay})} - 1 \]  

(A4)

where “a” is a lattice parameter and the indices “sub” and “lay” stand for the substrate (Si) and the layer (SiGeC), respectively. The symbols z and x,y refer to out-of-plane and in-plane components, respectively. For more information about x-ray characterisation see e.g. [68-69]

A.2 Atomic Force Microscopy (AFM)

AFM is one method among the family of scanning probe microscopy (SPM) techniques using a sharp tip that scans the surface at a very short distance. AFM often uses an etched Si tip that probes the forces between tip and surface. The motion of the tip is monitored by a laser beam that is reflected on the backside of the cantilever to a position sensitive photo detector. In this arrangement a small change in cantilever position will alter the position of the beam in the detector which can be converted to the vertical movement of the tip. This enables 3-dimensional images to be obtained when scanning a surface. Several modes of AFM are possible e.g. contact and tapping mode. In the contact mode the tip is in physical contact with surface. However, the tapping mode is often used due to the fact that a number of friction related problems are avoided, e.g. adhesion and electrostatic forces, etc. In this case the cantilever is excited to vertical oscillations close to the resonance frequency. As the tip approaches the surface and begins to “tap”, the resonant frequency is reduced due to the energy loss when contacting the surface. During the operation as the tip encounters a bump or trench the room for oscillating is decreased or increased leading a change in amplitude. This change is monitored by the above-mentioned optical lever and a digital feedback loop then adjusts the tip-sample separation to keep a constant amplitude and force. Piezoelectric crystals are used to control the cantilever oscillations as well as movement on the microscopic level in x, y and z direction. The resolution capabilities are extremely good in the vertical direction on the order of Ångströms. The lateral resolution is somewhat poorer within >5nm [70].
A.3 Hall measurement

Hall effect measurements are an excellent way to determine carrier mobility, density of carriers and carrier type in a material [70]. In case of a geometry like figure A3, an electric current density, \( J_x \) flows in the x direction. Moreover, a magnetic field \( B_z \) is applied in the z direction, which causes the traveling charges to bend in either positive or negative y-direction. In order to obtain steady state flow of carriers in the x-direction the current flow need to be zero in the y-direction. In that case the bent carriers will form an electric field in the opposite direction. This electric field constitutes a voltage so-called Hall voltage, \( V_H \). The Lorentz force, \( \mathbf{F} \) on the carriers can be written as following:

\[
\mathbf{F} = -q(\mathbf{E} + \mathbf{v} \times \mathbf{B})
\]

where \( q \) is the charge of the carrier, \( \mathbf{E} \) the electric field, \( \mathbf{v} \) velocity of carrier and finally \( \mathbf{B} \) the magnetic field. Furthermore, it is convenient to define Hall coefficient, \( R_H \):

\[
R_H = \frac{E_y}{J_x \cdot B_z}
\]  

(A6)

The sign of this \( R_H \) determines the carrier type. Furthermore, since \( E_y = V_H \cdot d \) equation A6 can be rewritten as:

\[
R_H = \frac{V_H \cdot d}{J_x \cdot B_z}
\]

(A7)

where the right side represents quantities that can be obtained directly from measurements.
In the simplest case i.e. one carrier type in one sub-band the Hall scattering factor, $r_H$ is determined by the following relation:

$$R_H = \frac{r_H}{q \cdot n_H} \quad (A8)$$

where $n_H$ is the Hall active carrier concentration. This $r_H$ is a material, temperature and magnetic field dependent factor between 0.2 and 1.7 causing a discrepancy between the measured Hall concentration/mobility ($n_H/\mu_H$) and the real values. The Hall mobility, $\mu_H$ can be calculated through the following relation:

$$\mu_H = \frac{\sigma}{q \cdot n_H} \quad (A9)$$

where the conductivity $\sigma$ may be obtained from resistivity measurements.

A.4 Secondary ion mass spectrometry (SIMS)

A focused beam of high-energy ions (\sim 15 keV) called primary ions, e.g. $\text{Ar}^+$, $\text{Cs}^+$, $\text{O}^-$ is sent on the surface of the sample to be analyzed. The choice of ion is determined by the element that should be detected. Those ions pull out ions, called secondary ions, from the sample and a small fraction of the secondary ions are analyzed by a mass spectrometer. Concentration can be determined with accuracy \pm 10\%. It is essential to apprehend that this method gives the total atomic concentration. In other words, this method does not permit to distinguish substitutional from interstitial atoms [71].

A.5 Transmission electron microscopy (TEM)

In order to study the crystal quality and defects in the epitaxial layers, cross-sectional TEM has been used. With this technique samples are ion-milled to a thickness which is transparent to the electron beam. Accelerated electrons pass through the sample to visualize an image on a fluorescent screen below the sample. The maximum resolution is of the order of some Ångströms which enable an extremely detailed study of the specimen. However, the
sample preparation is destructive and time consuming making this technique quite expensive [72].
Bibliography


Epitaxy and characterization of SiGeC layers grown by reduced pressure chemical vapor deposition


Epitaxy and characterization of SiGeC layers grown by reduced pressure chemical vapor deposition


Epitaxy and characterization of SiGeC layers grown by reduced pressure chemical vapor deposition

Appended papers

**Paper A**
*Growth of high quality epitaxial SiGeC layers by using chemical vapour deposition*

**J. Hållstedt**, E. Suvar, P. O. Å. Persson, L. Hultman, Y. -B. Wang och H. H. Radamson


**Paper B**
*Methods to reduce the loading effect in selective and non-selective epitaxial growth of SiGeC layers*

**J. Hållstedt**, E. Suvar, C. Menon, M. Östling och H. H. Radamson


**Paper C**
*Strain and electrical characterisation of boron-doped SiGeC layers grown by chemical vapor deposition*
