Chapter 5

Substitution, diffusion of Sb and damage distribution after MeV carbon irradiation of Sb implanted Si

5.1 Introduction

In Chapter 4, we had discussed the 1.5 MeV Sb implantation in Si. In addition to investigating the Sb substitution and location in the Si lattice, we had also studied the depth dependent damage distribution produced during ion implantation. All the studies were carried out as a function of annealing temperature. Annealing of the ion-implanted Si lattice is essential for acquiring good Sb substitution in Si-lattice sites as well as for reduction in implantation induced damage. Studies demonstrate that radiation damage can also be effectively reduced by a subsequent ion irradiation [1, 2, 3, 4]. Earlier studies show that high energy (MeV) ion irradiation by a non-dopant element like O, C, F, Ne or As, at elevated temperatures (> 180°C), in silicon implanted with various dopants strongly suppress the formation of secondary defects produced in subsequent high temperature annealing [2, 3, 5, 6, 7, 8].

In the present study we have made a detailed investigation of the effect of high
energy irradiation on the damage distribution produced due to Sb ion implantation in Si. Moreover, we have also studied the modifications in Sb-substitution, distribution and diffusion due to HEII and subsequent annealings. In literature [9, 10], though most of the irradiation studies are performed at ambient temperatures (200 - 500°C), in the present study, the irradiation has been performed at room temperature (RT) so that substrate temperature does not influence the implantation induced damage or the dopant distribution. During irradiation, the energetic ions while cascading through the material can produce a distribution of point defects which may critically influence the growth of the damaged layer as well as the dopant behavior.

In this chapter we also discuss and extensively investigate the effect of low temperature annealing prior to high energy ion irradiation on the damage and dopant behavior of the MeV implanted Sb in Si. Sb exhibits a vacancy (V) mediated diffusion in surface oxidized silicon single crystals [11] as well as in MBE grown [12] and keV implanted [13] layers. However after MeV implantation, Sb diffusivity can depend upon implant damage, impurity-defect interactions, migration of defect complexes etc. These phenomena have not received any attention in literature. Moreover, with HEII we have been able to investigate and understand the influence of V-supersaturation created during HEII on the dopant behavior as well as its substitution in the Si lattice. This procedure produces a lattice structure with good Sb substitution and low overall damage density.

RBS/C technique has been used to monitor the re-growth of Si as well as the distribution and substitution, with annealing, of MeV implanted Sb in Si. Studies have been conducted for post-irradiated sample as well as for the sample which has undergone a low temperature annealing treatment prior to HE irradiation. Effect of isochronal annealings on these systems have also been investigated.

### 5.2 Experimental

A mirror polished (100)-oriented Si single crystal wafer (p-type, \(\rho=0.008-0.02 \ \Omega\cdot\text{cm}\) ) was used in the present study. Two samples were implanted at room temperature with a scanned beam of 1.50 MeV Sb\(^{2+}\) ions to a total dose of 5\(\times\)10\(^{15}\) cm\(^{-2}\). During
After implantation, irradiation was performed on one of the samples at RT (Sample A) with a C\(^{3+}\) beam of 8.00 MeV to a total dose of \(5\times10^{16}\) cm\(^{-2}\). This energy was selected to be high so that a significant point defect separation is established in Si [18] and the carbon implant (range \(R_C\approx7.18\ \mu m\)) is isolated from the Sb profile. During irradiation the sample was thermally isolated with the use of a quartz plate and the current on the target was kept lower than 0.3 μA so that the temperature rise of the substrate due to beam heating is minimal (< 50°C) [6]. This sample was then isochronally furnace annealed at 400, 600 and 800°C for 30 minutes.

The other sample implanted with 1.50 MeV Sb\(^{+2}\) ions to a total dose of \(5\times10^{15}\) cm\(^{-2}\) (Sample B), was first vacuum annealed at 600°C and then irradiated using a C\(^{3+}\) ion beam of 8 MeV to a total dose of \(5\times10^{16}\) cm\(^{-2}\) at RT. The double implanted sample (II) was further annealed at 800°C for 30 min.

RBS/C measurements were carried out along [100] and [110] direction using 2.05 MeV He\(^{+}\) ions. A surface barrier detector with 20 keV resolution was positioned at a scattering angle of 150° to detect the scattered particles. An incident charge of \(~15\ \mu C\) was collected for each spectra.

### 5.3 Results and Discussion

#### 5.3.1 Effect of High Energy Ion Irradiation (HEII)

Figure 5.1 shows the random (a) and [100] aligned spectra from the as-implanted (b) and irradiated (c) sample as well as after annealing it to 400 (d), 600 (e) and 800°C (f). From the aligned RBS/C data of the as-implanted sample (b), it is clear that the top layer of the sample is completely amorphous. Using the GISA-3.95 ion scattering analysis code [14] the thickness of this amorphous layer was found to be 745 nm (this has been discussed in detail in Chapter 4). After irradiation at RT, the width of the amorphous zone in the Si lattice, created during implantation, remains
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**Fig. 5.1.** Effect of HE C⁺ irradiation: Random (a) and aligned backscattering spectra for as-implanted Si (b) with 1.5 MeV Sb and for as-irradiated sample (c) after HE C⁺ irradiation of this as-implanted sample are shown. Aligned spectra are also shown for the as-irradiated sample after it was sequentially annealed at 400°C (d), 600°C (e) and 800°C (f). A channeling spectrum from a virgin Si(100) (g) is also shown.

unaltered. However, a low thermal anneal at 400°C of this C⁺ irradiated sample results in drastic reduction of the Si yield indicating regrowth of the amorphous layer. Since solid-phase-epitaxial-growth (SPEG) of the amorphous layer in Si is expected to occur at temperatures $\geq 560^\circ$C [1, 15], our result is unexpected and very surprising. Furthermore, the crystalline quality of Si lattice deteriorates after 600°C anneal of the irradiated sample. The minimum yields ($\chi_{\text{min}}$) for the Si and Sb signals along with the fraction of Sb substitution at Si sites have been presented in Table 5.1. On further annealing at 800°C, the Si yield increases indicating that the crystallinity of the lattice has deteriorated.
The aligned yield of the Sb profile from the irradiated sample is almost similar to that of the as-implanted sample indicating that irradiation at RT results in almost no relocation of the dopant atoms (Fig. 5.1). However, on annealing at 400°C, the Sb yield decreases sharply giving rise to a high Sb substitution fraction \( (f_s) \) of 93%. Further annealing at 600°C, however results in an increase in the non-substitution fraction. This trend continues even after annealing at 800°C. The value of \( f_s \) at 600 and 800°C are 89% and 84% respectively (see Table 5.1). A marked increase in the aligned Sb yield beyond \( R_p \) at 800°C, may suggest some movement of the dopant in the Si lattice or possibly formation of Sb precipitates at the present Sb peak concentration of \( 1.7 \times 10^{20} \text{cm}^{-3} \). A hump in the Si profile [Fig. 5.1(f)] at a depth of \( \sim 400 \text{ nm} \) may thus be due to the distortions in the Si lattice caused by precipitates. A slightly higher Si yield and a reduced Sb substitution compared to that at 400°C are also observed at 600°C suggesting that some precipitate may form at this temperature also. The increased aligned-Si yield at 600 and 800°C can also be a consequence of the excess point defects which may be released prior to Sb-precipitation [16]. These point defects can play a very important role in the kinetics of the dopant migration [17]. This will be further discussed in the following sections.

The momentum transferred from the C\(^+\) projectile to the Si lattice, during irradiation, creates a slight but systematic displacement between the interstitial (I) and V distributions. After I-V recombination, excess I remain near \( R_C \) while the corresponding V surplus is found in the surface region [18]. Since \( R_C \) is an order of magnitude larger than \( R_p \), an excess V is present in the Sb region (\( \sim 50-800 \text{ nm} \)). This is also supported by TRIM analysis. The above results indicate that unlike the Sb implantation alone, the V supersaturation created during C\(^+\) irradiation not only induces the Si-regrowth with a low temperature annealing at 400°C, but also drives Sb into a (93%) substitution which is higher than that achieved in the un-irradiated case (Chapter 4). Influence of energetic ions in promoting the dopant substitution after MeV implantation, as observed in the present study [19], is an anomalous effect and has not been observed earlier. In this study, HEII was performed at RT unlike some other studies where Si regrowth was observed with ion irradiation at ambient temperatures of 200-500°C [1]. These earlier studies suggested that the beam induced
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Fig. 5.2A. Damage profiles for Si(100) implanted with 1.5 MeV Sb\textsuperscript{2+} after irradiation with 8 MeV C\textsuperscript{3+} ions, as well as after sequential annealing at 400, 600 and 800°C. \textit{R}_p denotes the range of Sb in Si(100).

ion-beam interactions in the substrate control SPEG of the amorphised lattice. We observe that high energy irradiation at RT alone does not produce any regrowth in Si implying that the excess vacancies produced during irradiation, probably play a crucial role during annealing in Si-crystallization and Sb substitution. The increased channeling yield after 600 and 800°C anneal suggest some precipitation at these temperatures which can cause a reduction in \textit{f}_s. Antimony in Si is a system with a low solid solubility limit [20] and for metastable Sb concentration exceeding about 6–7 \times 10^{19} \text{cm}^{-3}, precipitates form during a heat treatment of \sim 800°C for 30 mins [17]. TEM and channeling studies show that on rapid thermal annealing of 80 keV Sb implanted Si at a dose of 5 \times 10^{15} \text{cm}^{-2}, at temperatures above 700°C, precipitates of Sb in the normal hexagonal phase were formed [21].

Depth dependent damage profiles were extracted for the RBS/C spectra shown in Fig. 5.1 using the multiple scattering formalism described in Chapter 2. The damage profiles are shown in Fig. 5.2A. The damage profile of the irradiated sample is similar
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Fig. 5.2B. Blow-up of Fig. 5.2A.

to that of the as-implanted sample (not shown). The top 745 nm from the sample surface is completely amorphous \((N_d/N \sim 1)\). On annealing at 400\(^\circ\)C, the damage in the lattice reduces drastically. The profile shows a hump at \(\sim 320\) nm (A) and a very broad damaged region (B) in Fig. 5.2A as well as in Fig. 5.2B, (which shows a blow-up of Fig. 5.2A). The damage profiles for the irradiated sample annealed at 600 and 800\(^\circ\)C indicate an overall rise in the lattice damage. The results indicate that HEII can stimulate recrystallization (or SPEG) in Si after 400\(^\circ\)C anneal, which is absent otherwise. The damage peak at A will mostly consist of un-recombined vacancies. The location of peak-A at a depth \(< R_p \) suggests this. However, since Sb is also present in this region, some Sb-vacancy complex formation may also take place. The broad damaged-region B in Fig. 5.2B indicates the presence of damage at the end-of-range (EOR). This may most probably be due to the presence of interstitial-point defects or their clusters, as formation of secondary defects at this temperature is unlikely. However, some secondary defect formation at this temperature, under the influence of HEII, though improbable cannot be completely ruled out.

Fig 5.3 shows the RBS/C profiles of the Sb regions after HEII and after sequential
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![Depth (nm)](image)

Fig. 5.3. RBS/C spectra of the Sb region showing redistribution and diffusion of the dopant after HEII and sequential annealing at 400, 600 and 800°C.

Annealing at 400, 600 and 800°C. Although the as-irradiated Sb profile exhibits a Gaussian distribution, a redistribution of the dopant is observed after annealing at 400°C. This indicates that some dopant-defect interaction may be taking place. An increased channeling intensity at 600 and 800°C compared to 400°C is also observed in Fig. 5.3. As discussed earlier this leads to a lower substitution fraction of 89% and 84% at 600 and 800°C respectively, compared to 93% at 400°C. The increased channeling intensities can be a result of some precipitate formation at 600 and 800°C. Studies [16, 17] also suggest that some V may get released prior to the formation of precipitates. Since Sb-diffusion is promoted by V, the in-diffusion of Sb towards the a/c interface, as noticed in Fig. 5.3 at 600 and 800°C, may be related to this excess V. The diffusion Sb in the Si lattice will be discussed in detail in the next section. Here we notice that the excess V may have contributed to the shift of damage-peak A (Fig. 5.2B) towards larger depths. We further observe that the shift is more after 800°C than 600°C anneal. Higher V-release at 800°C and subsequent Sb-diffusion
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Fig. 5.4. Redistribution and enhanced diffusion of Sb: Random spectra of the Sb region for the as-irradiated sample and after sequentially annealing it at various temperatures. After 600°C and 800°C anneals, profiles exhibit enhanced in-diffusion. (discussed later) can explain this. The damage in region B at 600 and 800°C will be mostly due to secondary defects formed at the a/c interface at these temperatures [25].

For understanding the dopant distribution and diffusivity following annealing of C⁺ irradiated sample, the random profiles of the antimony region for the as-irradiated as well as annealed samples have been analyzed (Figure 5.4). Although the as-irradiated profile remains similar to the Sb as-implanted profile (data not shown), the irradiated sample after a 400°C anneal exhibits an enormously redistributed dopant profile with several features. The Sb diffusion towards the a/c interface at higher temperatures is also very apparent. For quantitatively evaluating the dopant movement, these Sb RBS-profiles were fitted with five Gaussian peaks (results shown in inset). However, only three components (A, B, C) were necessary to achieve a good fit for the as-irradiated profile. The analysis was performed for all annealing temperatures.
and the results (peak positions and areas) are shown in Fig. 5.5. For the as-irradiated sample, Sb is primarily located in the central B feature, which represents the dopant located at $R_p$. Although feature B still contains the largest amount of Sb after 400°C anneal, it is flanked on either side by a new feature (P', P'') depicting a significant mobility of the dopant atoms. The diffusion at this stage is minimal with only feature C indicating a slight in-diffusion towards the bulk Si. However, after 600°C anneal, an enhanced diffusion of Sb towards the a/c interface is observed with Sb getting primarily transported to the region near P'' ($\sim 434\text{nm}$). This trend continues even after 800°C anneal when a high Sb concentration is seen around $\sim 511\text{nm}$. Phenomenon of enhanced in-diffusion after 600 and 800°C is also demonstrated in feature C, though the Sb concentration in this component is small ($\sim 10\%$) and does not vary much with temperature. Though the position of the central peak B at 345 nm ($R_p$) remains almost immobile, the amount of dopant under the peak varies as a function of annealing temperature. Noticeably, features P' and P'' at 400°C are observed at depths of
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\[ \sim R_p/2 \text{ and } \sim 3R_p/2 \] respectively. In addition to the excess V after irradiation, these two regions will also have a surplus of V and I respectively, due to Sb implantation [18]. These features in Sb profile, observed also at higher temperatures, may thus develop due to Sb-point defect interactions (forming e.g SbV, SbV$_2$, SbSi etc.). This Sb redistribution is not observed in absence of HE irradiation. Though SIMS studies show that dopants like As, P and B usually do not couple with point defects during SPEG [28], we observe that irradiation promotes this coupling for Sb. Since the diffusion is noticed only in the C$^+$ irradiated samples, V created during irradiation are promoting this diffusion. Though V mediated Sb diffusion has been observed earlier [11, 12, 13, 16], diffusion behavior in MeV implanted Sb-layers or in the presence of Sb related implant damage has never been investigated. Moreover, inward diffusivity towards the a/c interface, in the present study, is puzzling since TRIM analysis for 8 MeV C$^+$ ions exhibits a nearly uniform density of V in the dopant region. Though without direct evidence, a recent report speculates that the ripening of precipitates can release V in the lattice [16]. Additional excess V released below $R_p$, region where precipitates are predominantly evidenced at 800$^0$C [Fig. 5.1(f)], can explain the inward Sb diffusivity supporting the above hypothesis. A higher in-diffusion at 800$^0$C may be ascribed to the higher V flux at this temperature, compared to 600$^0$C.

The HEII thus promotes the 93% Sb substitution and redistribution in Si at 400$^0$C, which is almost absent otherwise. Dopant diffusion and redistribution is observed at 600$^0$C and 800$^0$C. The shift in damage peak at $R_p$ and the Sb in-diffusion towards a/c interface suggest that V release during precipitation may be controlling the Sb diffusion kinetics.

5.3.2 Effect of Pre-annealing and HEII

Figure 5.6 shows RBS/C results measured along the [100] direction from Sample B. The 1.5 MeV Sb implanted Si was annealed to 600$^0$C prior to HEII with a carbon (C) beam of 8 MeV energy for Sample B. Results of Sb-implanted sample annealed to 600$^0$C are discussed in Chapter 4 and in this section we discuss the post-irradiation results. After irradiation, the II sample was further annealed at 800$^0$C.
Annealing of the as-implanted sample at 600°C (Fig. 5.6b) results in recrystallization of the amorphised lattice leading to a $f_s$ of 69% (discussed in Chapter 4). Irradiation with a carbon beam at 8 MeV (II), results in very little modification of the Sb profile (Fig. 5.6c) when compared to the sample before irradiation. However, the Si region exhibits a higher channeling yield indicating some production of defects due to C irradiation. The $\chi_{min}$ values obtained from the Si and Sb yields are 16.5% and 37% respectively and $f_s$ being 75%. The increase in Si yield can be due to production of Si and Sb recoils in the surface and sub-surface regions. Some strain in the Si lattice, due to excess V introduced in this region during HEII, can also cause this increase. Annealing of the II sample at 800°C (Fig. 5.6d) results in a reduction in both the Si and Sb channeling yield as well as $\chi_{min}$ (also see Table 5.1). At this temperature, the $f_s$ is $\sim$ 94% along the [100] direction. Measurements were also performed along [101] channeling direction. Channeling along the [101] direction leads to a $f_s$ of 95%. Fig. 5.7 shows the angular scans for [101] axis obtained from the 800°C annealed sample. The nearly equal critical angles for Si and Sb ($\psi^{Si}_{1/2} = 0.41^\circ$ and $\psi^{Sb}_{1/2} = 0.41^\circ$).
and $\psi_{1/2}^{Si}=0.43^\circ$) and the lower minimum yields indicate that most of the Sb atoms are present at regular Si sites. The very similar axial angular scans of Sb and Si along the [100] channel (data not shown) also confirm this. With a 94% substitution along [100] axis and 95% along [101], we assume a net Sb-substitution of 94% at 800°C. The Sb profile exhibits no redistribution or diffusion in the Si lattice at this stage.

Fig. 5.8 shows the damage profiles after 600°C annealing, post-irradiation (II) and after 800°C annealing of Sample B. On annealing the as-implanted sample at 600°C, recrystallization of the thick amorphous layer formed due to implantation takes place by SPEG. As discussed earlier in Chapter 4, the damage profile of the 600°C annealed sample reveals the existence of three peaks A, B and C at depths of 310 nm, 480 nm and 780 nm respectively. The peaks A and B correspond to the regions having clusters of vacancies and interstitials respectively, and peak C is due to the formation of secondary defects at the a/c interface. After HEII, an increase in damage is observed which is predominantly in surface - 2$R_p$ region. This increase in damage will be primarily due to Sb and Si recoils that are produced due to irradiation with high energy ions. The damage beyond the a/c interface has also

**Fig. 5.7.** Angular yield profiles along [101] axis for 2.05 MeV He$^+$ ions incident on the C-irradiated and 800°C annealed Sb-implanted Si(100) sample. The normalized yields from Si and Sb within the implanted region are shown.
increased slightly. On annealing the II sample at 800°C, recombination of the point defects take place. The damage in the surface region has reduced with feature A (~310nm) barely visible. Around region B some damage is still present. Vacancies are expected to be predominant at the surface and region-A, whereas interstitials and secondary defects will be primarily present in regions B and C respectively. By comparing the Fig. 5.8 with Fig. 5.2B (and Table 5.2) we observe that, after 800°C annealing, EOR damage is much less for the sample that underwent pre-annealing treatment and HEII (Sample B) than for the sample with no HEII (Chapter 4) or the sample with HEII and no pre-annealing (Sample A). Thus, the annihilation of point defects during 600°C pre-annealing treatments seems to affect the EOR damage formation after subsequent irradiation and annealing. This result can be of immense consequence in IC fabrication technology. Furthermore, the pre-annealed and HEII sample (Sample B) exhibits the best substitution of 94% after 800°C anneal and demonstrates no Sb diffusion as observed for HEII sample (with no pre-annealing.
Substitution, diffusion of Sb and damage distribution after MeV carbon irradiation, treatment) at 600 and 800°C.

<table>
<thead>
<tr>
<th>Sample</th>
<th>Condition</th>
<th>$\chi_{\text{Si}}^m$</th>
<th>$\chi_{\text{Sb}}^m$</th>
<th>$f_s$</th>
<th>Total Si displ (cm$^{-2}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Un-irradiated</td>
<td>as-implanted</td>
<td>1</td>
<td>1</td>
<td>-</td>
<td>$29.6 \times 10^{17}$</td>
</tr>
<tr>
<td>(Chapter 4)</td>
<td>400°C</td>
<td>1</td>
<td>1</td>
<td>-</td>
<td>$29.6 \times 10^{17}$</td>
</tr>
<tr>
<td></td>
<td>600°C</td>
<td>0.12</td>
<td>0.39</td>
<td>69%</td>
<td>$3.5 \times 10^{17}$</td>
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<tr>
<td></td>
<td>800°C</td>
<td>0.09</td>
<td>0.19</td>
<td>89%</td>
<td>$3.6 \times 10^{17}$</td>
</tr>
<tr>
<td>HEII Sample A</td>
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<td>1</td>
<td>-</td>
<td>$29.5 \times 10^{17}$</td>
</tr>
<tr>
<td>(Chapter 5)</td>
<td>400°C</td>
<td>0.13</td>
<td>0.19</td>
<td>93%</td>
<td>$4.2 \times 10^{17}$</td>
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<tr>
<td></td>
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<td>0.24</td>
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<td></td>
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<td>0.295</td>
<td>84%</td>
<td>$6.3 \times 10^{17}$</td>
</tr>
<tr>
<td>Pre-annealed &amp; irradiated Sample B (Chapter 5)</td>
<td>600°C</td>
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<td>0.39</td>
<td>69%</td>
<td>$3.5 \times 10^{17}$</td>
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<td></td>
<td>II</td>
<td>0.165</td>
<td>0.37</td>
<td>75%</td>
<td>$5.7 \times 10^{17}$</td>
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<tr>
<td></td>
<td>800°C</td>
<td>0.11</td>
<td>0.165</td>
<td>94%</td>
<td>$3.3 \times 10^{17}$</td>
</tr>
</tbody>
</table>

Table 5.1: The crystallinity, substitution fraction and total Si displacement in the 1.5 MeV Sb implanted Si(100), without irradiation, after HEII and sequential annealing at 400, 600 and 800°C. Also shown is the $\chi_{\text{min}}$, $f_s$ and total Si displacement in the pre-annealed sample, followed by irradiation and 800°C anneal (see text for more details).

Table 5.1 gives a comparison of the crystallinity of the Si matrix, the substitution fraction ($f_s$) and the total number of displacements in the Si matrix for the three samples studied in Chapters 4 & 5 (i.e., un-irradiated, HEII and 'pre-annealed & HEII' samples). A comparative figure of the resultant damage profiles is shown in Fig 5.9. The studies indicate that SPEG of MeV Sb implanted Si(100) occurs after an anneal at 600°C (Chapter 4). The damage profile indicates 3 regions where the damage in the system is primarily accumulated. The total number of Si displacements
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present in the system at this stage is \(3.5 \times 10^{17} \text{cm}^{-2}\). Annealing at 800°C results in a better crystalline quality of the Si lattice at the surface region. However, the damage profile and Table 5.2 indicate the presence of a substantial amount of secondary defects at the a/c interface and a total damage of \(3.6 \times 10^{17} \text{cm}^{-2}\). This is nearly equivalent to the number of Si displacements present at 600°C. HEII, using carbon ions at RT, of the Sb implanted Si(100) (Sample A, Chapter 5), has no effect on the damage distribution of the amorphised lattice. However, after an isochronal anneal at 400°C, the amount of damage at the surface region is almost equivalent to that of the 600°C annealed - unirradiated sample as discussed in Chapter 4 (see Fig. 5.9). Beyond this surface region there is an overall increase in damage which could be due to the Si and Sb recoils created during HEII. Some damage is also observed at the a/c interface which may be mostly due to interstitials and recoils, as secondary defect formation is unlikely at this stage. A high substitution of 93% is observed at this stage though the total damage in the lattice is much higher than that in the 800°C-annealed unirradiated sample. The Sb profile exhibits a redistribution after an anneal at 400°C. Annealing at higher temperatures results in some precipitate formation which causes dopant diffusion towards deeper depths. Damage profiles indicate the presence of large density of secondary defects at the a/c interface at these temperatures. High energy ion irradiation (HEII) of the Sb implanted and 600°C pre-annealed sample (SampleB, Chapter 5), results in large increase in damage at the surface and subsurface region. However, on annealing this II sample at 800°C, it is observed that there is a high percentage of substitution \(f_s = 94\%\) of Sb at Si sites. A reduction in the overall damage in the lattice is also observed, with very little damage at the a/c interface. The total number of Si displacements is \(3.3 \times 10^{17} \text{cm}^{-2}\), which is less than the total damage in any other sample studied here. No diffusion or redistribution of the dopant (Sb) is observed after annealing at 800°C.

5.4 Summary and conclusions

In this chapter, the effect of high energy ion irradiation (HEII) using an 8 MeV C\(^{+3}\) beam, on the dopant (Sb) distribution and substitution in Si(100) lattice has
been studied. The damage distribution in the system has been extracted from the RBS/C data using the multiple scattering formalism. Implantation of 1.5 MeV Sb in Si(100) at a dose of $5 \times 10^{15} \text{ions/cm}^2$ leads to the formation of a completely amorphised top layer (745 nm) in the Si lattice. HEII using carbon ions at RT, results in no change in the amorphised layer or the damage in the system. However, after 400°C anneal, the vacancy supersaturation created during irradiation stimulates a growth in silicon and induces a high Sb substitution of 93%. Annealing at 400°C also promote Sb-defect interactions resulting in redistribution of Sb in the Si lattice. This is accompanied by some surface, sub-surface damage as well as higher damage at a/c interface. Annealing at 600 and 800°C, surprisingly causes a worsening of substitution fraction and a large accumulation in overall damage. The Sb-channeling results indicate some precipitate formation at this stage. Precipitate formation in turn can lead to release of V which seem to promote an inward diffusion of Sb towards the a/c interface. The development of the precipitates and secondary defects, may be responsible for the increased damage observed at this stage. RBS/C studies on
the 600°C pre-annealed HEII sample exhibits an enormous increase in surface and sub-surface damage. However, an 800°C anneal of this sample produces a high Sb substitution of 94%. Furthermore, the damage in this sample is much lower than the 400°C annealed sample (Sample-A) which had similar Sb substitution fraction. Table 5.1 and 5.2 clearly establish that HEII of the 600°C pre-annealed sample (Sample-B) exhibits a combination of the lowest overall damage and the best substitution after an 800°C anneal. Moreover at this stage no precipitation or Sb-redistribution was observed. These results suggest that though the V-supersaturation created by HEII may be necessary for a good Sb-substitution, pre-annealing treatment is essential for the reduction of lattice damage.

<table>
<thead>
<tr>
<th>Sample Condition</th>
<th>Total Si displ (cm⁻²) (0 - 20 nm)</th>
<th>Total Si displ (cm⁻²) (20 - 400 nm)</th>
<th>Total Si displ (cm⁻²) (400 - 600 nm)</th>
<th>Total Si displ (cm⁻²) (600 - 1100 nm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Un-irradiated (Chapter 4)</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>600°C</td>
<td>0.06 × 10¹⁷</td>
<td>1.30 × 10¹⁷</td>
<td>0.68 × 10¹⁷</td>
<td>1.50 × 10¹⁷</td>
</tr>
<tr>
<td>800°C</td>
<td>0.03 × 10¹⁷</td>
<td>0.84 × 10¹⁷</td>
<td>0.74 × 10¹⁷</td>
<td>2.00 × 10¹⁷</td>
</tr>
<tr>
<td>HEII Sample A (Chapter 5)</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>400°C</td>
<td>0.07 × 10¹⁷</td>
<td>1.50 × 10¹⁷</td>
<td>0.96 × 10¹⁷</td>
<td>1.63 × 10¹⁷</td>
</tr>
<tr>
<td>600°C</td>
<td>0.09 × 10¹⁷</td>
<td>1.90 × 10¹⁷</td>
<td>1.29 × 10¹⁷</td>
<td>2.60 × 10¹⁷</td>
</tr>
<tr>
<td>800°C</td>
<td>0.10 × 10¹⁷</td>
<td>2.10 × 10¹⁷</td>
<td>1.42 × 10¹⁷</td>
<td>2.66 × 10¹⁷</td>
</tr>
<tr>
<td>Pre-annealed &amp; irradiated Sample B (Chapter 5)</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>600°C</td>
<td>0.06 × 10¹⁷</td>
<td>1.30 × 10¹⁷</td>
<td>0.68 × 10¹⁷</td>
<td>1.50 × 10¹⁷</td>
</tr>
<tr>
<td>II</td>
<td>0.11 × 10¹⁷</td>
<td>2.30 × 10¹⁷</td>
<td>1.20 × 10¹⁷</td>
<td>2.10 × 10¹⁷</td>
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<tr>
<td>800°C</td>
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<td>1.20 × 10¹⁷</td>
<td>0.71 × 10¹⁷</td>
<td>1.32 × 10¹⁷</td>
</tr>
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Table 5.2: The total Si displacement in the 1.5 MeV Sb implanted Si(100) at various depths after annealing at 600 and 800°C. The total number of Si displacements after HEII and sequential annealing at 400, 600 and 800°C is shown. Also shown is the total Si displacement at various regions in the 600°C pre-annealed sample, followed by irradiation and 800°C anneal. (see text for more details).
Bibliography


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