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Interaction of ion-implantation-induced interstitials in B-doped SiGe

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Abstract

B-doped $Si_{0.77}Ge_{0.23}$ of various surface-doping levels was used to investigate the evolution of implant damage and the corresponding transient enhanced diffusion of boron as a function of boron concentration. These layers were implanted with a non-amorphizing 60 keV, $1\times10^{14}\,cm^{-2}$ Si, and annealed at 750 °C. Plan-view transmission electron microscopy (PTEM) confirmed the formation and dissolution of dislocation loops. Transient enhanced diffusion (TED) is evident in the surface doped SiGe, but the low diffusivity of interstitials in $Si_{0.77}Ge_{0.23}$ and the presence of interstitial traps inhibited TED at the deeper B marker layer.

Keywords: SiGe; Ion-implantation; Annealing; Dislocation loops; Transient enhanced diffusion

1. Introduction

The insertion of SiGe alloys into integrated circuit technology is being materialized with modest documentation of the relationship between dopant transient enhanced diffusion (TED) and clustering kinetics in SiGe; however, there exist vast, detailed knowledge presented regarding transient enhanced diffusion in pure Si. Under interstitial injection, known boron interstitial clusters (BICs) nucleate in pure Si one order of magnitude below solid solubility resulting in poor activation and regulators of TED [1–6]. Substitutional boron atoms reduce the number of interstitials clustered in extended defects [4,7]. BICs in Si are metastable and will dissolve at high thermal cycles consistent with the dissolution of the immobile peak post-completion of TED. In the early stages of annealing, interstitial supersaturation is high and interstitial-rich clusters persist; in the latter stages, clusters are predominantly B-rich [8].

When annealing at a furnace temperatures of 750 °C, boron diffusion decreases with increasing Ge content [9] presenting preferential Ge–B bonding as reasoning for retardation of B diffusion in SiGe [9–11]. Point defect injection from annealing ambient experiments in Si-rich SiGe suggests that B diffusion remains interstitially mediated [12,13] with fractional interstitialcy constants comparable to that of pure Si [14]. Similarly to pure Si studies, SiGe is in need of identification of clustering kinetics and resulting B diffusion in an excess interstitial

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state while monitoring the effects of damage and point defect supersaturation [15]. The paper presents the extended defect evolution and ensuing B-TED arising from ion-implantation and annealing in $Si_{0.77}Ge_{0.23}$.

2. Experimental design

B-doped $Si_{0.77}Ge_{0.23}$ was grown atop of relaxed $Si_{0.8}Ge_{0.2}$ substrates via molecular beam epitaxy with varying surface B well concentrations: 8.0×10^{18}, 1.5×10^{19}, and $3.0 \times 10^{19}\,cm^{-3}$. These B surface wells extended 0.25 μm, and will be utilized to ascertain the evolution of extended defects in the presence of boron. At 0.50 μm below the surface, each sample contained a 5 nm B deep marker layer with a concentration of $3.0 \times 10^{18}\,cm^{-3}$, which was used to monitor the release of injected interstitials and the associated diffusion from the B-doped surface region. These structures were implanted with a non-amorphizing 60 keV, $1 \times 10^{14}\,cm^{-2}$ Si, and annealed at 750 °C. These implant and annealing conditions were chosen to mimic the processing steps, which yield visible subthreshold defects in Si [1–6]. The diffusion of B was measured by secondary ion mass spectrometry (SIMS), and the defect morphology was investigated by PTEM.

3. Results and analysis

In Fig. 1, both implanted and unimplanted SIMS profiles of these structures annealed at 750 °C for 3 h are shown in comparison to the as-grown profile. The B diffusion at the marker layer appears to be intrinsic in nature exemplifying minor diffusion with no assistance from point defect injection. This amount of diffusion is consistent with previous work indicating that B would travel 6.6 nm after a 3 h, 750 °C anneal [16]. With point defect injection from similar implant conditions and thermal budgets, B migrated over 50 nm in B-doped SiGe with no B doping at the surface [16]. The diffusion of the implanted B wells increases with decreasing B concentration. The diffusion lengths of the 8.0×10^{18}, 1.5×10^{19} (not shown) and $3.0 \times 10^{19}\,cm^{-3}$ are 28, 22, and 18 nm, respectively. The post-anneal profiles of the unimplanted and implanted $3.0 \times 10^{19}\,cm^{-3}$ well samples lie right on top of each other with no enhancement. The amount of enhancement at the B surface well is more pronounced than at the marker layer indicating that the presence of B is obstructing point defect assisted diffusion at the marker for both B concentrations.

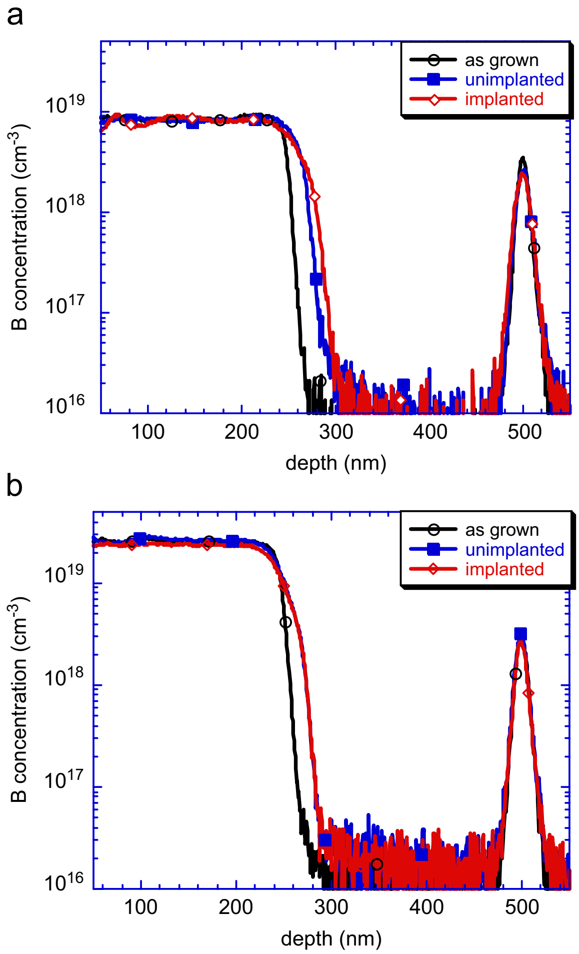

Fig. 1. SIMS profiles of B in $Si_{0.77}Ge_{0.23}$ as grown, unimplanted plus anneal, and implanted plus anneal for the (a) $8.0 \times 10^{18}\,cm^{-3}$ and (b) $3.0 \times 10^{19}\,cm^{-3}$ B well samples. The implant condition was 60 keV, $1 \times 10^{14}\,cm^{-2}$ Si, and the ensuing anneal was a 3 h, 750 °C cycle.

After 3 hours of annealing long after the dissolution of extended defects, the effective diffusivity enhancement is less than 4 for all of the B-doped SiGe samples well within the window of intrinsic diffusion; however, without the presence of a B well at the surface, the effective diffusivity enhancement was ~15 [16]. It appears that B may serve as an additional, effective interstitial trap in $Si_{0.77}Ge_{0.23}$, and TED was being modulated by the presence of B at the surface. In order to attain a full assessment of B-TED in SiGe, an investigation of extended defect evolution in the presence of B is necessary.

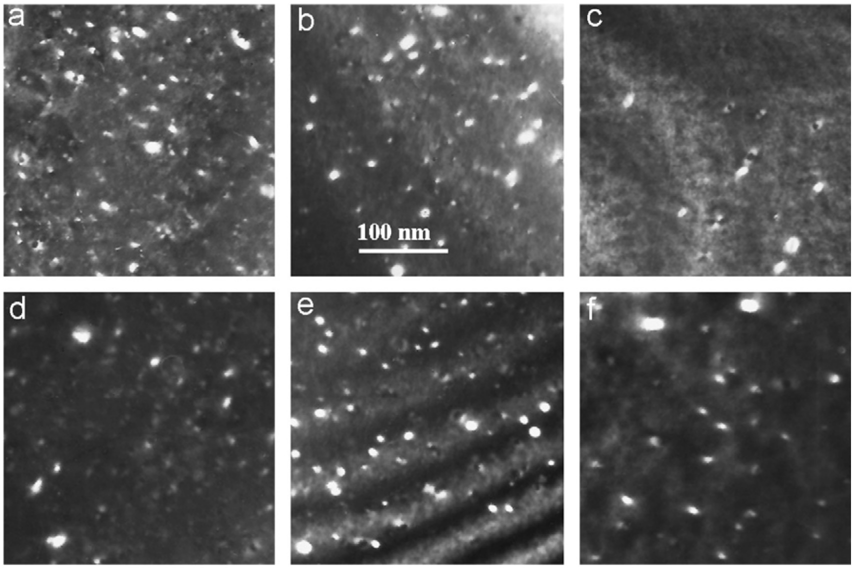

Fig. 2. PTEM images of B-doped $Si_{0.77}Ge_{0.23}$ with 60 keV, $1\times10^{14}\,cm^{-2}$ Si annealed 750 °C for the $8.0\times10^{18}\,cm^{-3}$ B well sample (a) 5 min, (b) 10 min, and (c) 120 min and $3.0\times10^{19}\,cm^{-3}$ B well sample (d) 5 min, (e) 10 min, and (f) 120 min.

Fig. 2 presents weak beam dark field g_{220} (for defects are visible when $\boldsymbol{g}\cdot\boldsymbol{b}\times\boldsymbol{u}\neq0$, where $\boldsymbol{g}$ is the reciprocal lattice vector corresponding to the diffraction plane, $\boldsymbol{b}$ is the Burger's vector of the dislocation, and $\boldsymbol{u}$ is the line direction of the dislocation) PTEM images of implanted and annealed B-doped $Si_{0.77}Ge_{0.23}$. The PTEM micrographs show that interstitials precipitate into dislocation loops in the presence of B. The evolution of dislocation loops in B-doped $Si_{0.77}Ge_{0.23}$ is similar to undoped SiGe revealing a high density at 5 and 10 min [16,17] in the ripening stages. At the longer anneals, a decrease in density is observed marking the duration and onset of dissolution. The defect densities of 8.0×10^{18} and $3.0\times10^{19}\,cm^{-3}$ B-doped $Si_{0.77}Ge_{0.23}$ with respect to time are comparable.

Quantitative TEM [18] was utilized to monitor variations in defect morphology with increasing B concentrations, and the correlation between trapped interstitials and doping level is shown in Fig. 3. To further examine the effects of B on defect morphology in $Si_{0.77}Ge_{0.23}$, a comparison between the stability of dislocation loops in $Si_{0.77}Ge_{0.23}$ and {3 1 1} defects in Si has been established. Similar experiments in Si reveal that increasing B concentration led to the extinction of {3 1 1} defects due to B competing and trapping interstitials [4]. Even though the interstitial supersaturation has decreased considerably after 15 min of annealing, trapped interstitials behave differently in the presence of Ge. In Si, B has a stronger influence on the observed defects showing a drastic decrease in trapped interstitials with increasing B concentration with near disappearance of {3 1 1} defects at B concentrations of $1\times10^{19}\,cm^{-3}$ resulting from equivalent self-implantation and a 15 min, 740 °C anneal. Trapped interstitial areal density is definitely lower in B-doped $Si_{0.77}Ge_{0.23}$ than in the no B surface well sample [16]. Nevertheless, the seizure of interstitials by dislocation loops is persistent at higher thermal budgets in both undoped and doped $Si_{0.77}Ge_{0.23}$ than {3 1 1} defects in pure Si. Considering the

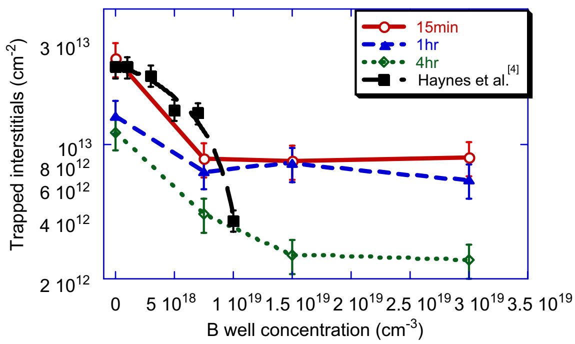

Fig. 3. Areal density of interstitials in extended defect observed in $Si_{0.77}Ge_{0.23}$ after a 60 keV, $1\times10^{14}\,cm^{-2}$ Si annealed at 750 °C (reference data from [4] was annealed at 740 °C for 15 min) as a function of B dopant concentration.

melting temperature of $Si_{0.77}Ge_{0.23}$ versus Si, the nucleation, ripening, and dissolution of defects in $Si_{0.77}Ge_{0.23}$ is rather impressive illustrated by the existence of dislocation loops after 4 h of annealing in the presence of B exceeding concentrations of $8.0\times10^{18}\,cm^{-3}$. The trapped interstitials decrease with increasing anneal time, but are fairly independent of B concentration in the B-doped samples annealed for 1 h or less. In the B-doped samples after 15 min and 1 h of annealing, trapped interstitials are somewhat constant with increasing B concentration. Only at longer anneals does the trapped interstitial density start to decrease with increasing B concentration.

It is clear that the density of trapped interstitials and defects are hindered by the presence of B. However, once dislocation loops form in B-doped $Si_{0.77}Ge_{0.23}$, the loops are stable. These two findings suggest that B competes for interstitials and hinders the nucleation of dislocation loops, but ripening and dissolution of loops is independent of B concentration.

4. Discussion

Boron interstitial clusters in Si are metastable in nature existing past thermal budgets of 740 °C for 4 h while concurrently dissolving and contributing to TED. After the implant and 750 °C, 3 h anneal in B-doped $Si_{0.77}Ge_{0.23}$, the TED is insignificant compared to B-doped Si [6]. Although the dislocation loops are more stable in $Si_{0.77}Ge_{0.23}$ than {3 1 1} defects in Si, these loops do exemplify clear dissolution within minutes of annealing at 750 °C releasing interstitials. Thus, the lack of TED suggests that the dislocation loop-free interstitials exhibit reduced mobility and/or are being trapped by some uncanny interstitial complex associated with the presence of B.

The defect morphology and the nature of diffusion at the B marker suggests that B persists as a successful trap extending 250 nm. It should be noted that the slower dissolution of defects and lack of TED also suggests that interstitial mobility is significantly reduced in the presence of Ge as compared to pure Si alluding to additional trapping mechanisms. These interstitial complexes are strong, stable competitors during the nucleation stages of dislocation loops indicated by the decreasing initial concentration of interstitials in loops with increasing B concentration. Once the dislocation loops are nucleated, the loops undergo ripening and dissolution with little dependence on increasing B concentration.

5. Conclusions

When 60 keV, $1\times10^{14}\,cm^{-2}$ Si interstitials are injected into 8.0×10^{18}, 1.5×10^{19}, and $3.0\times10^{19}\,cm^{-3}$ B-doped $Si_{0.77}Ge_{0.23}$, 750 °C post-annealing leads to the growth and dissolution of stable dislocation loops. TED arising from 750 °C anneal is observed after implantation at the 8.0×10^{18}, 1.5×10^{19}, and $3.0\times10^{19}\,cm^{-3}$ B-doped surface wells and decreases with increasing B concentration. In B-doped $Si_{0.77}Ge_{0.23}$, the existence of B hinders the nucleation of dislocation loops but the loop coarsening and release of interstitials is uninterrupted.

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