Impact of dopant profiles on the end of range defects for low energy germanium preamorphized silicon

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Abstract

As the industry continues to aggressively scale CMOS technology, the shift to lower energy ion implantation becomes essential. The consequent shallower amorphous layers result in dopant profiles that are in closer proximity to the end of range (EOR) damage and therefore a better understanding of the interaction between the dopant atoms and the EOR is required. A study is conducted on the influence of dopant profiles on the behavior of the EOR defects. Czochralski-grown silicon wafers are preamorphized with \(1 \times 10^{15} \text{ cm}^{-2}, 10 \text{ keV Ge}^+\) ions and subsequently implanted with \(1 \times 10^{15} \text{ cm}^{-2}, 1 \text{ kV B}^+\) ions. A sequence of rapid thermal and furnace anneals are performed at 750°C under a nitrogen ambient for periods of 1 s up to 6 h. Plan view transmission electron microscopy (PTEM) reveals a significant difference in the defect evolution for samples with and without boron, suggesting that the boron influences the evolution of the EOR defects. The extended defects observed for samples which contain boron appear as dot-like defects which are unstable and dissolve after very short anneal times. The defect evolution however, in samples without boron follows an Oswald ripening behavior and form \{311\}-type defects and dislocation loops. Hall effect measurements denote a high initial activation and subsequent deactivation of the dopant atoms which is characteristic of the formation of boron interstitial clusters. Diffusion analyses via secondary ion mass spectroscopy (SIMS) support this theory.

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1. Introduction

Standard CMOS technology employs the formation of amorphous layers via ion implantation, prior to the introduction of the dopant atoms into the silicon lattice, for ultra shallow junction (USJ) formation. An annealing process subsequent to implantation repairs the inherent lattice damage and activates implanted dopant atoms. However, the high interstitial supersaturation in the lattice during annealing results in unwanted complex and coupled processes such as clustering, diffusion and dopant deactivation. Transient enhanced diffusion (TED) [1] is a well-researched phenomenon [2,3,4,5] that also occurs. It is due to the formation of mobile dopant complexes via the coupling of silicon interstitials with the substitutional dopant atoms. The interstitial supersaturation associated with the EOR defects, which accompany amorphization provide the interstitials necessary for TED. Solid phase epitaxial re-growth (SPER) is capable of activating dopant levels that exceed solid solubility levels in silicon, at relatively low temperatures. The low thermal budget offers good junction depth control by reducing diffusion, which is highly beneficial for USJ formation. In addition, it provides excellent compatibility with the thermal requirements of high-K dielectric materials and metal gates. As the industry attempts to achieve highly doped, shallow layers, ultra low energy (ULE) dopant ion implantation is becoming increasingly important. These ULE implants do not require deep amorphous layers, therefore lower energy preamorphizing implants can be used. Low energy amorphizing implants reduce the number of recoil atoms into the EOR. However, lowering the implant energy also reduces the proximity of the EOR to the surface resulting in high interstitial supersaturations in the vicinity of the dopant profile. These interstitial concentrations have been demonstrated to cluster with boron atoms at concentrations far below its solid solubility [6].

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resulting in dopant deactivation. The proximity of the EOR to the boron profile therefore raises many concerns regarding the thermal stability of the activated junctions and the effect that the boron presence has on the EOR defect evolution.

2. Experiment

Two hundred mm, Czochralski grown, (100), n-type silicon wafers were preamorphized with 10 keV Ge ions at a dose of $1 \times 10^{15}$ cm$^{-2}$ and subsequently implanted with 1 keV, $1 \times 10^{17}$ cm$^{-2}$ B ions. Wafers which were not subjected to a boron implant served as the experimental control. All implants were conducted at a tilt and twist of 5° and 0°, respectively. Post-implantation processing included dicing the wafers into 1.4 cm x 1.4 cm samples and annealing at 750°C under nitrogen ambient for times ranging 1 s through 6 h. Anneals performed for less than 5 min were conducted in an AG Associates Heat Pulse 210T rapid thermal annealer (RTA). The RTA temperature profile consisted of a 1°C s$^{-1}$ ramping to the Ge preamorphization temperature, in those samples only subjected to the Ge preamorphization process. All implants were conducted at a tilt and twist of 5° to the normal to the surface with a raster of 250 μm. Sheet resistances were measured at 41°C on a MMR Technologies Inc., Hall and Van der Pauw measurement system. These measurements were conducted at 90 K to reduce experimental control in that tiny dot-like defects existed in the n-type substrate [8], thus allowing reliable sheet resistance measurements.

3. Results and discussion

B$^+$ implant conditions were chosen such that the boron was positioned in the vicinity of the end of range. The 10 keV Ge$^+$ implant formed a continuous amorphous layer of thickness 23 nm, as illustrated in Fig. 1. The boron implant produced a peak concentration of $\sim 8 \times 10^{20}$ cm$^{-3}$ and a junction depth of 26 nm. It is evident that boron levels at the amorphous–crystalline interface were on the order of 3 $\times$ 10$^{14}$ cm$^{-2}$ with boron extending into the EOR region. WBDF PTEM images of the extended defects observed during the annealing process, in those samples only subjected to the Ge$^+$ preamorphization are depicted in Fig. 2. Through-

![As-implanted boron concentration-depth profile for the sample containing 1 keV, $1 \times 10^{17}$ cm$^{-2}$ B$^+$ and 10 keV, $1 \times 10^{15}$ cm$^{-2}$ Ge$^+$ implants, indicating the amorphous–crystalline interface.](image-url)
Annealed at 750°C.
in the profile is gone. This concurs with the PTEM analysis
that demonstrated dissolution of the EOR damage. Another
prominent feature in the SIMS plots is the enhanced diffu-
sion between 5 and 10 min which resulted in an approximate
15 nm increase in junction depth. Prior to and beyond this
time, the observed diffusion is considerably less with a mere
junction motion of approximately 5 nm during the 20 min in-
terval between 10 and 30 min, suggestive of a transient diffu-
station enhancement. This observation is therefore indicative
of a transitory supersaturation of silicon interstitials in the 5–10-
min annealing interval. Further examination of the SIMS
plots reveals that concentrations exceeding $5 \times 10^{18} \text{cm}^{-3}$
have moved in the direction of the Si/SiO$_2$ interface for all
profiles, similar to reports by Duffy et al. [12]. Progressive
motion toward the surface on the order of 1 and 1.9 nm were
achieved for the spike and 30 min anneals, respectively. Ad-
ditionally, there were increased boron concentrations at the
surface to values on the order of $1 \times 10^{19} \text{cm}^{-3}$ for the 10
and 30 min anneals. These characteristics are consistent with
previous reports of boron uphill diffusion [12,13] and may be
indicative of such a phenomenon. However, extraction of the
boron dose from the SIMS profiles indicated lower retained
boron doses than the implanted dose. The retained dose de-
creased with increasing anneal-times, suggesting boron dose
loss may have occurred during annealing, which could also
account for the motion of the profiles to the surface. Com-
parison of the increase in the boron surface levels between

![Fig. 3. Defect density as a function of annealing time for samples containing 10 keV, $1 \times 10^{15} \text{cm}^{-2}$ Ge$^+$ implants. Annealed at 750°C.](image)

![Fig. 4. PTEM WBDF images taken at g$_{312}$ imaging conditions of the ex-
tended defects for samples containing 1 keV, $1 \times 10^{15} \text{cm}^{-2}$ B$^+$ and 10 keV,
$1 \times 10^{15} \text{cm}^{-2}$ Ge$^+$ implants. Annealed at 750°C.](image)
the 5–10-min and 10–30-min anneal time intervals, clearly indicates significantly more boron diffusion to the surface occurred between 5 and 10 min. In accordance with the theories put forth by Cowern et al. [14], in which they suggest that the interstitial supersaturation gradient provides efficient boron transport to the surface, a much higher interstitial flux to the surface must exist in the 5–10-min interval. This high flux is indicative of a steeper interstitial gradient from the defect layer to the surface. The smaller increase in boron surface concentration between 10 and 30 min therefore represents a decline in the surface interstitial flux and thus a reduction in the driving force for interstitial diffusion. These observations reveal that the interstitial supersaturation levels in the lattice falls with progressive annealing, supported by the reduced diffusion in the tail of the profile between 10 and 30 min, mentioned above. This time frame also coincides with the rapid release of interstitials from the EOR damage which is also indicative of a reduction in interstitial supersaturation. Extended defects require high levels of interstitial supersaturation in their vicinity in order to exist as stable structures. Should these supersaturation concentrations fall, defect dissolution occurs releasing the trapped interstitials, in an attempt to sustain the necessary interstitial levels. Consequently, defect evolution into more stable structures cannot occur. The experimental results herein demonstrate that for samples containing boron, rapid defect dissolution occurs, which is symptomatic of a drastic decrease in interstitial supersaturation. On the contrary, in samples without boron, the EOR damage was seen to undergo Ostwald ripening and evolved into stable dislocation loops, suggestive of an adequate interstitial supersaturation. Based on these recognized differences in behavior, it is reasonable to infer that the boron presence influences the interstitial supersaturation and hence the EOR defect evolution. It is postulated that this occurs by boron clustering with the interstitial atoms. The boron implant conditions employed in this experiment confines the boron atoms to the vicinity of the surface resulting in high surface levels of $\sim 1 \times 10^{14} \text{cm}^{-2}$. Furthermore, high concentrations are also attained in the EOR region. It is well established that boron interstitial clustering occurs when a highly doped region is subjected to an interstitial supersaturation [6]. At 750 °C, the interstitial supersaturation is on the order of $1 \times 10^{4}$ according to the findings of Cowern et al. [15]. Therefore, the necessary conditions exist to enable clustering effects and boron interstitial cluster (BIC) formation. BICs grow by binding interstitials until a high supersaturation is sustained. [16,17] Recent experiments conducted by Mannino et al. [17] demonstrate that the interstitial supersaturation in samples containing boron is considerably lower when BICs are formed. It is therefore postulated that because the boron is in close enough proximity to the EOR, the BICs compete with the EOR defects for the interstitial supersaturation. [18,19] Hence, the supersaturation in the defect band falls to levels that are too low to support the extended defect existence and evolution, so they dissolve. BICs are known to have a detrimental effect on the electrical properties of the doped silicon host due to boron electrical deactivation [18,20] and carrier concentration mobility lowering. [21] Understanding the electrical behavior of the doped material is therefore essential to gaining further insight into the processes governing the defect evolution. Fig. 7 is the boron activation data represented as sheet resistance as a function of anneal-time. The most striking feature of the plot is the rapidity with which substitutional boron deactivates; demonstrated by the sharp increase in sheet resistance within the first 15 min of annealing. A spike anneal at 750 °C achieved a sheet resistance of $\sim 900 \Omega$ which increased to $\sim 1750 \Omega$ where it saturated. Measurements conducted on samples annealed for 4 h exhibited no improvements in boron activation, suggestive of the existence of boron in a very stable, inactive configuration. BICs have been shown to be very stable, existing up to 4 h after TED at 800 °C [6,21]. These deactivation characteristics

![Graph](Image 100x568 to 309x770)
are consistent with the formation BICs and strongly suggest their occurrence. The fact that the deactivation process saturates implies that the interstitial saturation levels may have fallen to levels that are too low to support further BIC formation. This coincides with the fall in supersaturation observed in the SIMS data as well as the rapid dissolution of the EOR associated with reduced interstitial levels. Typically, BICs are immobile and stationary peak concentrations observed in SIMS profiles are characteristic of the existence of these clusters. The SIMS results obtained in these experiments however, do not exhibit this characteristic. A possible explanation for this is that the immobile regions are hidden. An estimate of the clustered concentrations can be extracted from the electrical data assuming that the inactive concentrations are clustered. A more accurate calculation is obtained by utilizing the total retained dose determined from the SIMS analysis. On this basis, the clustered concentrations were estimated at $1 \times 10^{15} \text{cm}^{-2}$, which is below the maximum boron concentrations achieved by the implant. Consequently, the immobile portions of the profiles are not visible. The clustered boron concentrations are also two orders of magnitude higher than the boron levels in the EOR, suggesting that the boron deactivation is not confined to the EOR region. Deactivation of the boron atoms in the regrown crystal therefore necessitates the migration of interstitial atoms from the EOR to regions of high boron concentrations [9]. However, the driving force for interstitial motion is not clear. One possibility is interstitial annihilation at the surface. It is evident in the experimental control that interstitial annihilation at the surface does not deplete the supersaturation in the EOR sufficiently for defect dissolution [22]. Hence, it is reasonable to assume that surface does not play a role in the reduced interstitial levels in samples containing boron and thus, in the interstitial migration from the EOR region. The tensile strains associated with the initially high substitutional boron concentrations may induce interstitial diffusion from the EOR to compensate for the strain. This migration is plausible since the EOR region is situated such that it is not isolated from the effects of the strained layer. The migrating interstitials subsequently interact with the boron in the highly doped, strained area to form BICs which continue to bind interstitials competing with the EOR defects for the interstitial supersaturation. The EOR defects therefore never evolve into more stable structures.

4. Conclusion

Boron presence in the vicinity of the EOR plays a fundamental role in the EOR defect evolution. The rapid dissolution process of the EOR damage in samples containing boron signifies a reduction in interstitial supersaturation which is attributed to the formation of boron interstitial clusters. It is postulated that the tensile strain associated with the initially high boron activation may induce interstitial migration from the EOR region to the strained layer, resulting in the BIC formation which results in a rapid deactivation process. Consequently, the supersaturation in the EOR falls to levels that are too low to sustain the extended defect evolution and the defects ultimately dissolve.

References