Effect of carbon codoping on boron diffusion in amorphous silicon

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The effect of carbon codoping on boron diffusion in amorphous silicon is investigated during low temperature annealing. The diffusivity of boron is unaffected by carbon codoping, but the fraction of mobile boron is observed to increase with increasing carbon concentration. A concomitant reduction in boron clustering is also observed at higher carbon coimplant concentrations, consistent with a change in the local trap concentration. This is consistent with carbon possibly acting as a trap site for boron and thereby changing the size and dynamics of the boron cluster formation. © 2008 American Institute of Physics. [DOI: 10.1063/1.2975833]

Achieving ultrashallow junctions with high levels of dopant activation and abrupt dopant profiles is crucial for advanced complementary metal oxide semiconductor devices. Preamorphization of Si prior to B implantation has several advantages for the fabrication of ultrashallow junctions including an increase in the activation of B due to dopant incorporation during low temperature solid phase epitaxial regrowth (SPER) of the amorphous layer.[1] However, it has also recently been reported that SPER results in significant B redistribution due to the fact that B diffusivities in amorphous (a−) Si are approximately four orders of magnitude greater than those in crystalline (c−) Si at the same temperature.[2,3]

a-Si is a continuous random network with dangling and floating bonds that act similar to point defects in c-Si.[4] While several diffusion mechanisms for B in a-Si have been proposed, the most plausible of these are based on concentration-dependent diffusion and trap-limited diffusion. In the case of trap-limited diffusion the intrinsic structural defects (dangling and floating bonds) trap dopant atoms and diffusion proceeds via dopant motion between trap sites. The degree of dopant diffusion is then proportional to the concentration of trap sites. Previous work in c-Si showed that substitutional C acts as an effective trap site for excess interstitials, thereby suppressing transient enhanced diffusion (TED) of B.[5,6] When TED is effectively suppressed, C has no reported effect on the active B concentration, but recent work has shown that during high temperature annealing, boron interstitial clusters dissolve when a C coimplant is present even though the B solubility limit remains at approximately 2×10^{20} cm^{-3} and TED is observed.[7,8]

C is known to slow SPER in a-Si, and previous work on B diffusion in a-Si coimplanted with F showed enhanced B diffusion due to the fact that F slowed the SPER.[9,10] This was a direct consequence of the fact that B spent a longer period in the a-Si phase. C coimplantation may similarly increase the net B diffusion by retarding SPER.

We measured the effect of C coimplantation on B diffusion in a-Si while avoiding the effect of C on SPER. The aim is to determine whether C affects B motion in a-Si as it does in c-Si. If C acts as a trapping species for dangling and floating bonds in a-Si or for B, then it will also help reveal the dominant diffusion mechanism for B in a-Si.

Si wafers of the (001) orientation were preamorphized from the surface to a depth of approximately 850 nm by cooling them to approximately 77 K and implanting them with Si ions (5×10^{15} Si$^+$/cm^2 at 500 keV, plus 1×10^{15} Si$^-$/cm^2 at 95 keV). This was followed by a C$^-$ implant of either 3×10^{15} or 1×10^{15} cm^{-2} at 15 keV. The control for this study received no C$^-$ implant after Si$^+$ preamorphization. The samples then received a B$^+$ implant of 1×10^{15} cm^{-2} at 3.5 keV. Due to C$^-$ implantation constraints the B implantation energy and projected range were increased from previously reported work.[3,11] Postimplant anneals were conducted in a quartz tube furnace under N$_2$ ambient. The anneal times and temperatures were chosen so as to limit the diffusion measurements to only the amorphous phase (i.e., prior to interaction with the advancing amorphous/crystalline interface).[12] Cross-sectional transmission electron microscopy (XTEM) was used to measure a-layer thicknesses before and after annealing to confirm that the advancing amorphous/crystalline interface had not reached the B profile. Secondary ion mass spectroscopy (SIMS) profiles for B were collected using an O$_2^+$ beam with a net impact energy of 2 kV and a beam current of 50 nA, while profiles for C were collected using a Cs$^+$ beam with a net impact energy of 3 kV and a beam current of 20 nA. In order to estimate B diffusivity the diffusion length was measured at a B concentration of 3×10^{19} cm^{-3} and equated to $2\sqrt{(D_B t)}$.

Previous work found that retarding SPER with a coimplantation species enhances the resulting B diffusion in a-Si due to the longer time spent in the a-Si phase.[10] Since C also slows SPER, the XTEM of both C coimplant conditions during annealing at 600 °C was used to determine the extent of SPER retardation. Figure 1 shows XTEM images of amorphous layers after annealing at 600 °C for 14 min. No significant reduction in SPER velocity was observed for the lower C fluence of 1×10^{15} cm^{-2}, but the velocity is observed to decrease by approximately 5% when the C fluence is increased to 3×10^{15} cm^{-2}. In order to avoid SPER related
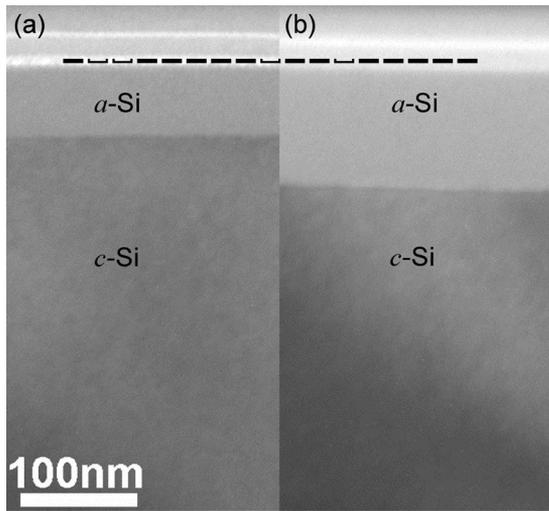
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FIG. 1. XTEM after implantation and annealing at 600 °C for 14 min with C$^-$ implantation at (a) 1×10^{15} cm^{-2} and (b) 3×10^{15} cm^{-2} C. Hatched line indicates sample surface.

B diffusion, all control and C coimplanted samples were annealed such that the B profile remained within the amorphous phase. The amount of regrown *a*-Si was not used as an experimental parameter.

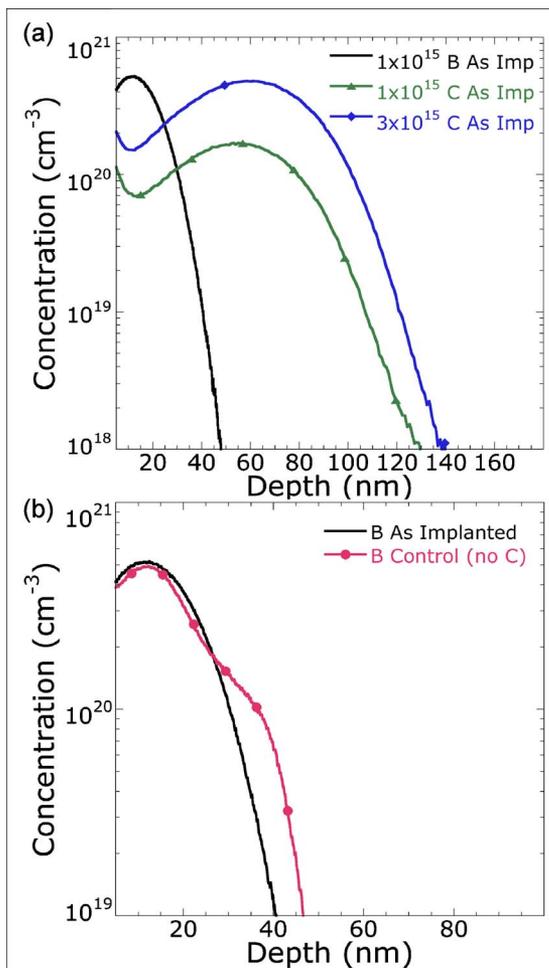Figure 2(a) shows the as-implanted dopant profiles for B

FIG. 2. (Color online) SIMS profiles of (a) as-implanted B and C concentrations and (b) the B control (no C) after annealing at 600 °C for 14 min.

and C under both C coimplant conditions. The C$^-$ implant energy corresponds to a projected range of approximately 60 nm, which overlaps with the tail of the B profile. The C fluence corresponds to a ratio between C$^-$ and B$^+$ ions of 1:1 or 3:1. When the C fluence is equal to that of B, the peak C concentration is equivalent to the immobile fraction breakpoint for B in *a*-Si, or approximately 2×10^{20} cm^{-3}. When the C fluence is increased to 3×10^{15} cm^{-2}, the peak C concentration matches the peak B concentration of approximately 5×10^{20} cm^{-3}. The first 5 nm of the dopant profiles from the surface were discarded as SIMS artifacts, and the B results were subsequently normalized to the fluence. Figure 2(b) confirms the characteristic diffusion profile of B in *a*-Si at higher implantation energies: a "knee" that is consistent with the presence of an immobile fraction above concentrations of 2×10^{20} cm^{-3} and a second mobile fraction below this value. This is indicative of B clustering and concentration-dependent B diffusion as discussed previously.[3]

Figure 3(a) shows B motion for all three implant conditions after annealing at 500 °C for 28 h. No significant C redistribution is observed under these anneal conditions and annealed C profiles are omitted for clarity. These data also show that the diffusion length at a B concentration of 3×10^{19} cm^{-3} is approximately 7 nm for all C coimplant conditions. It can therefore be concluded that the presence of the C coimplant has no effect on the B diffusivity, which is calculated at a concentration well below the immobile fraction breakpoint of B in *a*-Si. In contrast, a change is observed in the characteristic knee region of the B profile for the high C fluence, i.e., at B concentrations close to 2×10^{20} cm^{-3} where the immobile and mobile fractions differentiate. The fraction of immobile B appears to decrease, while the mobile B fraction increases proportionally. This is despite the fact that the immobile fraction breakpoint and the slope are consistent with the B profile in the absence of C. These phenomena are repeated under annealing at 550 °C for 2.5 h [Fig. 3(b)] and 600 °C for 14 min [Fig. 3(c)]. The effect of C on the B profile is equivalent under all current experimental conditions and is therefore considered to be temperature independent. Significantly, the concentration of C present in the region where B is immobile is equivalent to the increase in the mobile B concentration (approximately 1×10^{20} cm^{-3}). The B and C profiles also intersect at the clustering breakpoint of B in *a*-Si.

For all anneal conditions, the addition of C at a fluence of 3×10^{15} cm^{-2} increases the mobile B fraction by around 10% while having no effect on the diffusion length. There are several possible explanations for these observations. It is plausible that C could be inducing a steric effect. Unlike other coimplant species (Ge or F), C is smaller than B and Si and may not serve as an atomic-level block for B motion. Instead the addition of C is expected to alter the amorphous matrix, potentially opening the matrix for B to move into mobile positions. However with C only accounting for 1 at. % of the amorphous matrix, there is not enough C under the current implantation parameters to account for the magnitude of changes seen. C could also be acting as a local trap site in *a*-Si, effectively "tying up" dangling Si bonds. The binding energy for C–Si bonds is relatively strong at 4.7 eV.[13] Mirabella *et al.*[14] proposed a dangling bond mediated B diffusion in *a*-Si in which the extent of B diffusion is limited by the density of dangling bonds. This would mean that if C
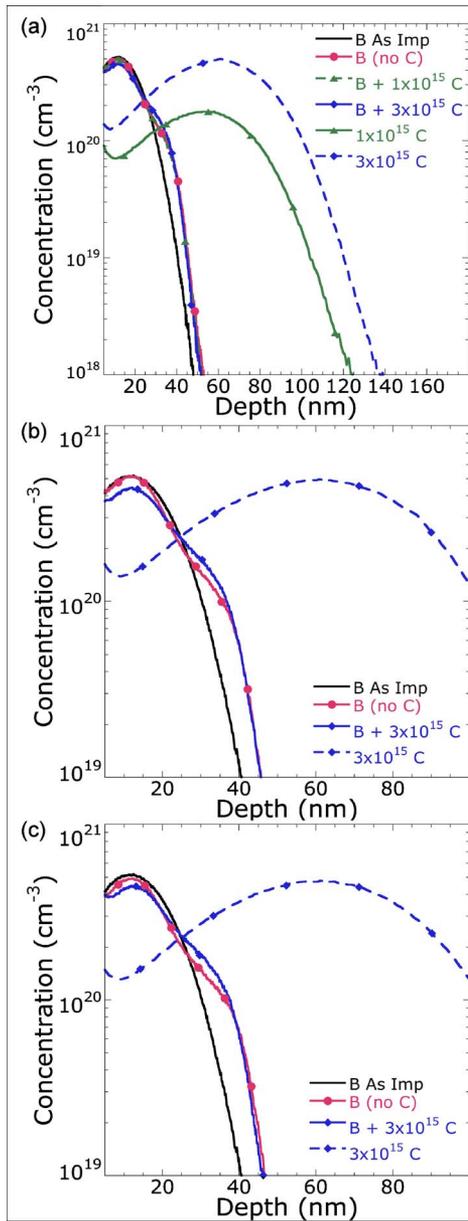
FIG. 3. (Color online) SIMS profiles for as-implanted and annealed samples at (a) 500 °C for 28 h, (b) 550 °C for 2.5 h, and (c) 600 °C for 14 min. SIMS profiles for B+1×10^{15} C are omitted for clarity from (b) and (c).

acted as a trap site for dangling bonds the resulting B diffusion would slow down with increasing C fluence. This is inconsistent with the data presented in Fig. 3. With the binding energy for B–C bonds (4.6 eV) being higher than that for B–Si (3.6 eV), it is also possible that B is preferentially binding with C.[13] In *c*-Si, the CB split interstitial has a lower

binding energy than either C_2I or B_2I, making it the energetically favored cluster.[15] Since C reduces the concentration of B clusters, C may be acting as a trap site for B, in effect changing the size and dynamics of the B cluster by replacing B and preventing other B atoms from joining the clusters. This could lead to an increase in mobile B if less B atoms are present in the clusters. If the excluded B atoms are also unable to form new clusters, this could lead to an additional increase in mobile B. This would simultaneously have no impact on the overall B motion. Further experimentation is needed to determine the precise effect of C on the B clusters.

In summary, the effect of C coimplantation on B diffusion and clustering in *a*-Si was investigated. C was found to have no effect on the diffusivity of the mobile B fraction but did influence B clustering matching, effectively reducing the fraction of immobile B and increasing the fraction of mobile B. Based on these observations it was argued that C has a direct effect on B trapping, thereby impacting the ability of B to cluster.

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