Interstitial injection in silicon after high-dose, low-energy arsenic implantation and annealing

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(Received 12 May 2005; accepted 14 September 2005; published online 7 November 2005)

In this work, we investigate the interstitial injection into the silicon lattice due to high-dose, low-energy arsenic implantation. The approach consists in monitoring the diffusion of the arsenic profile as well as of the boron profile in buried δ-doped layers, when amounts of the as-implanted arsenic profile are removed by low-temperature wet silicon etching. The experimental results indicate that the contribution of the implantation damage to the transient enhanced diffusion of boron, and thus the interstitial injection, is not the main one. On the contrary, interstitial generation due to arsenic clustering seems to be more important for the present conditions. © 2005 American Institute of Physics. [DOI: 10.1063/1.2130397]

High-dose, low-energy arsenic implantation is a process that is typically used for the formation of S/D regions in silicon MOS transistors. Compared to other dopants such as boron and phosphorus, fewer studies have been performed for the transient enhanced diffusion (TED) of arsenic.\textsuperscript{1–5} This is mainly because these effects are of moderate intensity at high temperatures and are more significant for very shallow junctions, less than 100 nm. In addition, research on arsenic is generally associated with high concentrations, where Fermi level dependence of dopant diffusion and point defects introduced by clustering or precipitation complicates the analysis of the implantation induced TED. In fact, there have been reported three sources for point defects that can contribute to the enhanced diffusion of arsenic and result in interstitial injection into the substrate. These sources are: (a) the end-of-range damage, (b) arsenic clustering, and (c) arsenic precipitation. Moreover, as the implantation energy is reduced and the implanted profiles are moving closer to the surface, the influence of the surface would be more severe.

In this work we investigate the interstitial injection caused by low-energy, high-dose arsenic implantation performed at room temperature. The approach consists in monitoring the diffusion both, of arsenic implanted with low energy and of boron in buried δ-doped layers, that exist below the implanted surface. Successive removal of Si implanted with arsenic is performed, using low-temperature oxidation ($<70^\circ$C), in order to control the total amount of damage that will influence the buried layer. The oxidation reaction under these conditions is not expected to influence the initial distribution of point defects, but will remove well-controlled depth slices from the damage profile.

Silicon wafers containing two boron δ-layers at a depth of 0.6 and 1.1 μm were implanted with arsenic at energies of 3 and 10 keV and dose of $2 \times 10^{15}$ cm$^{-2}$. The boron δ-layers were grown by chemical vapor deposition (CVD) and their concentration was of the order of $1 \times 10^{18}$ cm$^{-3}$. Cross-section TEM analysis revealed that the depth of the amorphous region was 6 nm and 13 nm, respectively, in good agreement with the predictions obtained by TRIM (Ref. 6) for the peak of the As profile in each case (6.8 nm and 13 nm, respectively).

The etching of the silicon substrate was performed in two steps: (a) an oxidization step into the oxidizing solution for 1 min, and (b) an immersion in 3% HF solution for 15 s, to dissolve the oxide. Details of the etching process are presented elsewhere.\textsuperscript{7} The combination of these two steps is referred as “one dip.” The thickness of the etched silicon was evaluated by AFM measurements of steps lithographically obtained on the silicon surface. In the case of 3 keV implant-

![Graph showing arsenic implantation profile](image-url)
The correspondence of implantation and annealing was observed in the 3 keV implantation case. The same diffusion takes place, both towards the surface and towards the bulk, after annealing at 900 °C for 30 s. We observe that the etched interstitials that remain in the samples after the successive etching steps decreases, in agreement to expected results. This is an indication that the etching process has been applied successfully. Although no significant diffusion of Arsenic was observed after annealing at 900 °C for 5 s, arsenic diffusion takes place, both towards the surface and towards the bulk, after annealing at 900 °C for 30 s. The same behavior was also observed in the 3 keV implantation case.

The corresponding SIMS profile of the boron δ-layers for the same samples is shown in Figs. 3 and 4. From these figures we can make the following remarks: (a) no significant diffusion of boron is observed after annealing at 900 °C for 5 s, while broadening of the boron profile is observable after annealing for 30 s; (b) the broadening of the boron δ-layers is, within experimental error, independent of the etching process for all implantation energies; and (c) the broadening of the boron profiles depends on the implantation energy and the distance much larger than the mean radius of the defect distribution.

In order to have a better idea of the obtained results it is useful to discuss about the various mechanisms that influence TED of dopants in the case of low-energy amorphizing implants in silicon. It is accepted that the origin of TED phenomenon is the interstitial supersaturation induced by the implantation damage. The recoiled interstitials that survive the recombination with vacancies at the crystalline part of the substrate form small clusters during the first stages (550–700 °C) of the subsequent annealing. As the annealing temperature increases these clusters are transformed in more stable extended defects (311) defects (700–800 °C) and dislocation loops (800–1000 °C). An interstitial supersaturation is maintained during the Ostwald ripening process that is taking place during the formation and evolution of the extended defects, as these evolve from (311) defects to dislocation loops and until the final dissolution of the latest. When the band of the defects is located away from the surface (at a distance much larger than the mean radius of the defect distribution) the ripening process is conservative and the number of bounded interstitials within the defects remains almost constant. This number represents the recoiled interstitials survived the recombination with vacancies at the crystalline part. This is the part of the implantation damage responsible for anomalous dopant diffusion. However, when the surface is very close to the defect band (as in our case) the ripening process is nonconservative since the surface is a strong sink for interstitials.

According to that, we expect that decreasing the distance between the defect band and the surface (by removing silicon) would lead to a reduction of the interstitial supersaturation within the defect band (due to more effective interstitial loss) and consequently to reduction of the enhanced diffusion of boron δ-layers. This is expected to be more evident in the case where the etched depth was more that the thickness of the crystalline part.
the amorphous layer. However, as we can notice from the comparison between the reference sample (nonetched) and the samples that were etched at the surface, no significant differences in the diffusion of the boron profiles were observed. This indicates that in our case boron TED is not only due to the supersaturation maintained during extended defect evolution. In fact, there is a strong possibility that in our case almost all the defects might have dissolved in all the samples even before the start of the 5 s annealing at 900 °C. This is supported by recent literature results indicating that no extended defects survived when a 5 keV, $10^{15}$ cm$^{-2}$ arsenic implantation was annealed at 850 °C for 10 s. In addition, TEM analysis of the samples implanted at 3 and 10 keV and annealed at 900 °C for 30 s, showed that no extended defects (dislocation loops) survived the annealing process, for all the samples (etched and nonetched), probably due to the proximity of the surface, which acts as an interstitial sink absorbing the interstitials from the defects band and leading to their dissolution. The combination of the above arguments leads to the evidence that the existence of a secondary interstitial source, which needs a specific thermal budget for its activation, is very possible. However, as we have stated earlier, there are two additional sources for point defects, especially in the case on high arsenic concentration. When the arsenic concentration exceeds the solid solubility of arsenic in silicon, arsenic clustering may occur. Electrical activation studies have attributed arsenic clusters to be the cause of dopant deactivation in silicon. These studies have also determined a critical concentration for arsenic cluster formation, which is a function of the annealing temperature. Theoretical studies and experimental evidence indicate that As clusters form around a vacancy with the consequent injection of silicon interstitials. The proposed mechanism can be described by the reaction

$$n \text{As} \leftrightarrow \text{As}_n^V + I,$$

where $n$ assumes values between 2 and 4.

In a recent work Solmi et al. estimated that about one-third of the TED of arsenic, implanted at 35 keV with a dose of $5 \times 10^{15}$ cm$^{-2}$, observed in the first 20 min of annealing at 800 °C is due to the defects produced by clustering. It is very probable that this is the main mechanism that is responsible for boron TED in our experimental conditions. Within the frame of the above assumption the independence of the boron profile broadening on the etching process for fixed implantation energy could be explained as follows. If the density of the formed clusters (and consequently of the injected interstitials) depends on the maximum arsenic concentration this result is expectable since, as we can see from Fig. 2, it is almost of the same order of magnitude in all the cases due to the dopant pile-up at the surface. On the other hand the energy dependence of the boron profiles broadening could be explained assuming that the high concentration part of the arsenic profile is more extended in the higher implantation energy case leading to a higher density of clusters.

Another possible source for point defect injection is due to arsenic precipitation. It has been reported that extended defects at the projected range are absent when the arsenic concentration approaches the critical value for precipitation. This is probably due to vacancies released during SiAs precipitation leading to the annihilation of extended defects. Vacancy injection could explain the absence of broadening in the boron profiles after 5 s of annealing via a mechanism that reduces the interstitial supersaturation level within the extended defects band. However, the formation of SiAs precipitates is not very probable in our experimental conditions due to the very high values of the solid solubility of As in Si and the difficulty of nucleation of this conjugate phase.

In this work we have investigated the interstitial generation by low-energy, high-dose arsenic implantation performed at room temperature and their subsequent injection into the substrate. Information about the mechanism that contributes mainly to silicon interstitial injection, under these conditions, was presented.

The financial support of EU IST Project No. 2000-30129 FRENDETECH (Front-End Models for Future Technology) is acknowledged. We also acknowledge Philips for providing the wafers with the boron δ-layers.