IN-SITU PHOTOEXCITATION-INDUCED SUPPRESSION OF POINT DEFECT GENERATION IN ION IMPLANTED SILICON

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ABSTRACT

The formation of vacancy-related defects in n-type silicon has been studied immediately after implantation of He, Si, or Ge ions at 85 K using *in-situ* DLTS. A-center concentrations in He-implanted samples reach a maximum immediately after implantation, whereas, with Si or Ge ion implanted samples they continuously increase during subsequent anneals. It is proposed that defect clusters, which emit vacancies during anneals, are generated in the collision cascades of Si or Ge ions. An illumination-induced suppression of A-center formation is seen immediately after implantation of He ions at 85 K. This effect is also observed with Si or Ge ions, but only after annealing. The suppression of vacancy complex formation via photoexcitation is believed to occur due to an enhanced recombination of defects during ion implantation, and results in reduced number of vacancies remaining in the defect clusters. In p-type silicon, a reduction in K-center formation and an enhanced migration of defects are concurrently observed in the illuminated sample implanted with Si ions. These observations are consistent with a model where the injection of excess carriers modifies the defect charge state and impacts their diffusion.

INTRODUCTION

The accumulation of ion implantation-induced defects occurs via interactions of Frenkel pairs, dopants, and other impurities. In our previous work [1,2], we demonstrated that these interactions occurring from 85 K to ~700 K were strongly affected by both the ion mass and substrate impurities. It was found that mobile vacancies and interstitials emitted from the clusters formed in the collision cascades played an important role in the formation of defect complexes. It is also known that the migration and interactions of irradiation-induced point defects are dependent upon their charge state and temperature. For example, the migration activation energy of a double negatively charged vacancy in silicon is known to be 0.18 eV, while it is 0.45 eV in the neutral charge state [3]. The defect charge state depends on temperature, the dopant concentration, and the presence of deep levels, since it is determined by the defect state position relative to the Fermi level. Therefore, it is expected that external *in-situ* interventions will provide an electronic stimulation during defect generation/evolution due to additional ionization effects.

Control of defect formation by *in-situ* electrical interventions were reported by previous workers [4,5], where an *in-situ* reverse or forward bias on p/n junctions reduced defect formation in n-type samples irradiated with electrons or He ions. In addition, *in-situ* illumination provided suppressions of A-center, divacancy, and carbon-oxygen pair generation in silicon irradiated with MeV electrons [6]. Similarly, photoexcitation during ion implantation was shown to reduce defect production [7,8] and dopant transient enhanced diffusion (TED) during post implantation heat treatment [9]. In general, this photoexcitation effect can be understood by the generation of excess carriers which can change the charge state of beam-induced defects and thereby alter the kinetics of defect diffusion and interactions. In the present study, further investigations into the physical mechanism of *in-situ* photoexcitation impact have been carried out via comparisons of ion mass, implantation temperature, light sources, and substrate type.

EXPERIMENTAL

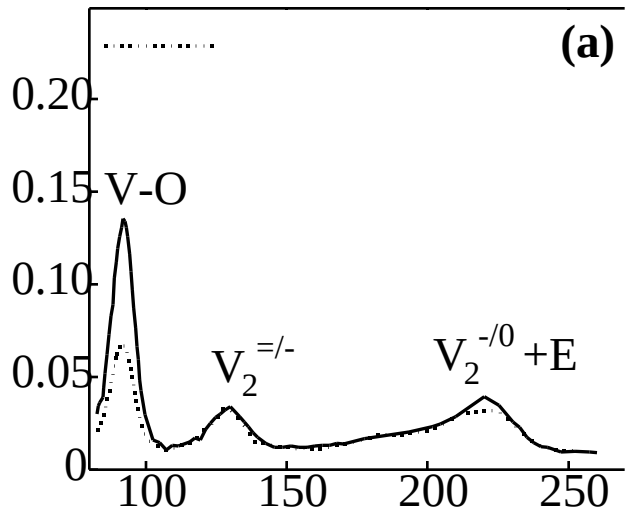
In this work, 1 _ 1mm^2 p$^+$/n diodes with a junction depth of ~0.3 µm were fabricated on n-type Czochralski (CZ) grown silicon substrates with carrier concentrations of ~ 2 _ 10^{15} cm^{-3}. These samples were mounted on the heating/cooling stage of an implantation target block with an electrical isolation. Unbiased diodes were illuminated/unilluminated during implantation with 0.6 MeV He, 2.6 MeV Si, or 4.2 MeV Ge ions to fluences of 0.3-2 _ 10^9 cm^{-2}. These ion energies have been chosen so that the implanted ion profiles were located within the space charge region of the biased diodes during subsequent DLTS measurements. In the *in-situ* illumination case, samples were exposed to a visible/IR halogen lamp or a mercury lamp predominantly emitting UV. The light intensity was controlled using the diode photocurrent, which was about 0.5 mA. The temperature increase due to illumination was measured by a thermocouple attached to the sample surface and shown to be below 5 K. In order to normalize this thermal effect, non-illuminated samples were heated 5 K higher than illuminated-samples. Immediately after implantation, the first *in-situ* DLTS scan was run from 85 K up to a desired annealing temperature, where the sample was maintained for one minute, after which the sample was quickly cooled to 85 K for the next annealing run. The DLTS measurements were performed using a lock-in technique with a 42 s^{-1} rate window. Also, unprocessed 5-6 Ω-cm p-type CZ reference samples were implanted with 2.6 MeV Si ions to a dose of 3 _ 10^8 at 295 K with and without *in-situ* illumination. These samples were examined by DLTS measurements after formation of Schottky barriers.

RESULTS AND DISCUSSION

The effects of photoexcitation on the formation defects during ion implantation of n-type silicon are shown by comparing the *in-situ* DLTS spectra in Figs. 1(a) and (b) for He and Ge ions with and without *in-situ* illumination. All spectra from these samples show the presence of the A -center (vacancy/oxygen pair), divacancy in the double negative charge state, and overlapping peaks at ~220 K corresponding to a mixture of single negatively charged divacancy (V_2) and E-center (vacancy/phosphorous pair). These peaks have previously been observed in n-type silicon after irradiation of electrons or ions [10,11]. The ratio of the A-center to V_2 + E amplitude in the He-implanted sample is ~5 times higher than in the Si-implanted sample and a further decrease of this ratio is observed in the Ge implanted sample. These observations indicate that single vacancies are more likely to be generated by light ion implantation and divacancies in the heavy ion implantation collision cascade. Note that the doubly negative charged divacancy peak at ~130 K does not reflect the total number of divacancies formed [12].

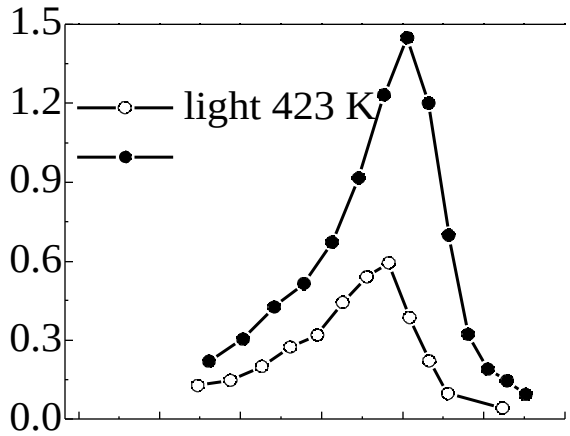
The modulation of electrically active defect formation is illustrated in Fig. 1 (a) by greater than two times reduction in the amplitude of A-center peak in the UV exposed sample implanted with He ions, while other peaks are negligibly affected. Also, in the Si-implanted sample the A-center and V_2 + E peaks are decreased by photoexcitation to a lesser degree. Regardless of ion species and implantation temperature, a reduction of vacancy related defects is observed in n-type silicon due to *in-situ* photoexcitation.

Similar to our previous work [1], A-centers are observed immediately after implantation of He, Si, or Ge ions into n-type silicon at 85 K. Next we examined A-center peak changes as a function of annealing temperature in samples implanted with He and Ge ions with and without *in-situ* photoexcitation. Figure 2(a) shows a reduction of A-center formation immediately after He implantation in the illuminated sample. This decrease is maintained at essentially the same level during subsequent post-implantation anneals; whereas in the illuminated Ge-implanted



samples suppression of A-center formation does not occur immediately, appears during subsequent anneals. This retardation behavior was also observed in the Si-implanted sample. Another interesting feature of these data is the kinetics of A-center development, i.e., slope of curves in Fig. 2, which strongly depends on the ion mass. For He, the A-center peak is already saturated during the first DLTS scan after implantation at 85 K, while in the Si or Ge-implanted sample, the formation of A-centers continues to increase with annealing temperature.

The mass dependence of the in-depth distribution of A-center and V_2 + E defects is also of interest. Figure 3 presents depth profiles for He and Ge ions implanted at 85 K with and without UV exposure. In the He-implanted case, the A-center concentrations were shown in Fig. 2 (a) to not significantly change due to subsequent anneals up to 150°C. Note that the concentration maxima of A-center and V_2 + E defects are ~0.2 μm shallower than the ion projected range (R_p). These locations were not affected by annealing temperatures or by photoexcitation, which does suppress the total concentration of A-centers more than two times, while the suppression of V_2 + E via illumination is negligible.



For the Ge-implanted sample, see Fig. 3(b), the A-centers continuously increase during subsequent anneals after implantation; however, no A-center peak occurs, even under zero filling pulse voltage, which analyze a depth significantly shallower than R_p. A large increase in the A-center concentration was observed in the near R_p region during annealing at 150°C (not shown in Fig. 3), which was much higher than that of the E-center contained in the V_2 + E peak. Simultaneously, the impact of photoexcitation produces an A-center concentration reduction above R_p. Similar behavior has also been observed for Si-implants [13]. It is commonly accepted that an A-center is generated via the trapping of a mobile vacancy by interstitial oxygen. Therefore, it is expected that a vacancy source whose activity has a wide temperature window is created in the collision cascade formed by low temperature Si or Ge ion implantation. It is revealed that the E-center concentration is reduced by photoexcitation during implantation of Ge ions after annealing the samples at 150°C where E-centers are dissociated; whereas the V_2 + E concentration reduction is only ~10%. Note that the maximum divacancy concentration in Fig. 3 (b) is remains near R_p.

Since the oxygen interstitial concentration is essentially the same in all samples, we believe that photoexcitation influences on the formation of vacancy sources, which depend on the ion mass. The suppression of defect formation via photoexcitation during ion implantation is attributed to a more effective Frenkel pair recombination process. In the case of He ion

implantation, defects are only sparsely distributed in the collision cascade allowing the enhanced recombination of defects due to photoexcitation to directly occur during implantation. Thus, vacancy sources for additional A-center formation during annealing are negligible. In case of heavy ion implantation, defect clusters produced in the dense collision cascades provide a

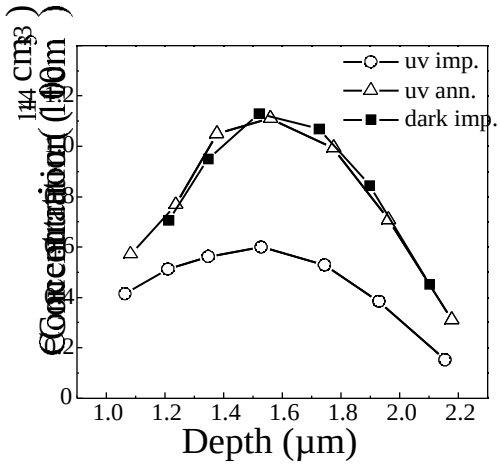


Fig. 4. Depth profiles of A-centers in samples implanted with 3_10^{8} cm^{-2} Si ions at 2.6 MeV with (open circle symbols) and without (solid symbols) photoexcitation at 85 K. Tn89 ’m ¯ ‡25 660.m 225

vacancy source which remains after the low temperature implantation, even though photoexcitation has reduced the number of vacancies stored in the damaged regions by an enhanced recombination process.

To investigate effects of illumination during post implantation annealing, samples were heated to 300 K with and without UV irradiation immediately after implantation. UV exposure annealing does not impact the formation of A-centers as shown in the depth profiles of Fig. 4. This observation indicates that the effects of photoexcitation occur via additional electronic excitations due to electronic energy loss of implanted ions. However, since the implantation-induced ionization current measured in the diode was about two orders lower than the photocurrent of illumination, it is believed that the effects of photoexcitation are determined by Frenkel pair recombination during ion implantation, where the concentrations of mobile vacancy and interstitial are high. We also found a similar reduction in defect formation via *in-situ* illumination with a halogen lamp whose spectrum does not have UV component. Thus, the effects of photoexcitation are solely caused by the injection of excess carriers.

To investigate the effects of *in-situ* illumination in p-type silicon, unprocessed CZ samples were implanted with Si ions at room temperature with and without *in-situ* illumination. DLTS spectra measured after Schottky diode fabrication show divacancies, K-centers (containing carbon/oxygen pair), and the H2 peak [2]. Figure 5 shows the depth profiles of divacancies and K-centers. In the *in-situ* illuminated-case, the K-center concentrations are significantly decreased above R_p, whereas slightly increased concentrations are seen in region deeper than R_p. Simultaneously, the maximum concentrations of divacancy and K-center have been decreased and shifted ~0.3 μm deeper in the *in-situ* illuminated-sample. Similar effects of illumination have

been observed in our low temperature He or Ge ion implanted samples. It is known that electronic excitations enhance the migration of point defects in several charge states during ion implantation [14]. The shifted defect distributions are probably attributed to the enhanced athermal migration [15], which occurred during ion implantation by photoexcitation. Evidence of the absence of enhanced defect migration in low dose, n-type samples is not conclusive at this time. However, it may be a factor in the fact that defect complexes form earlier in n-type than in p-type silicon at low temperature.

CONCLUSIONS

The suppression of A-centers via photoexcitation is observed immediately after implantation of light ions at 85 K. Additional A-center formation is barely seen in the sample implanted with light ions during subsequent anneals. The reduction of defect formation via photoexcitation in the sample implanted with heavy ions is not immediately present after implantation, but appears during heat treatment. These observations indicate that photoexcitation during low temperature ion implantation reduces the number of vacancies in the defect clusters via an enhanced recombination of Frenkel pairs. The remaining defect clusters formed near the R_p of heavy ions act as continuous vacancy sources for additional A-center formations during subsequent annealing.

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