End-of-range defects in germanium and their role in boron deactivation

F. Panciera, P F Fazzini, M. Collet, J. Boucher, Eléna Bedel-Pereira, Fuccio Cristiano

To cite this version:


HAL Id: hal-01922899
https://hal.science/hal-01922899
Submitted on 14 Nov 2018

HAL is a multi-disciplinary open access archive for the deposit and dissemination of scientific research documents, whether they are published or not. The documents may come from teaching and research institutions in France or abroad, or from public or private research centers.

L’archive ouverte pluridisciplinaire HAL, est destinée au dépôt et à la diffusion de documents scientifiques de niveau recherche, publiés ou non, émanant des établissements d’enseignement et de recherche français ou étrangers, des laboratoires publics ou privés.
End-of-range defects in germanium and their role in boron deactivation

F. Panciera, a) P. F. Fazzini, M. Collet, J. Boucher, E. Bedel, and F. Cristiano

LAAS, CNRS, 7 Avenue du Colonel Roche, 31077 Toulouse, France and UPS, INSA, INP, ISAE, LAAS, Université de Toulouse, 31077 Toulouse, France

(Received 25 May 2010; accepted 31 May 2010; published online 7 July 2010)

We investigated the thermal evolution of end-of-range (EOR) defects in germanium and their impact on junction thermal stability. After solid-phase epitaxial regrowth of a preamorphized germanium layer, EOR defects exhibiting dislocation loop-like contrast behavior are present. These defects disappear during thermal annealing at 400 °C, while boron electrical deactivation occurs. After the whole defect population vanishes, boron reactivation is observed. These results indicate that germanium self-interstitials, released by EOR defects, are the cause of B deactivation. Unlike in Si, the whole deactivation/reactivation cycle in Ge is found to take place while the maximum active B concentration exceeds its solubility limit. © 2010 American Institute of Physics. [doi:10.1063/1.3456537]

Recently, germanium is gaining a renewed interest as substitute of Si for microelectronic applications thanks to its attractive properties. Indeed, compared to silicon, germanium exhibits a four times higher hole mobility and requires lower process temperatures, reducing the problems related to the diffusion/activation of dopants in ultrashallow junctions and making germanium significantly better suited for integration with high-k dielectrics and metal gates.

In Si-based devices, preamorphization followed by solid phase epitaxial regrowth (SPER) is used to achieve high dopant activation. During such process, end-of-range (EOR) defects form just below the original amorphous/crystalline (a/c) interface. Upon subsequent annealing, these defects act as a source of free Si interstitial atoms (hereafter called interstitials) that are the cause of undesirable effects on dopant behavior, such as transient enhanced diffusion (TED) and electrical deactivation.

In germanium, the role of Ge interstitials during boron diffusion has been the object of a strong debate during the past years, and only very recently a direct experimental proof of boron TED was provided. On the contrary, a full experimental evidence of extended defects formation and evolution in Ge is still lacking, while no evidence at all exists concerning defects-related dopant electrical deactivation. First investigations on EOR defects in germanium, performed using transmission electron microscopy (TEM) did not reveal the presence of defects after SPER of amorphous Ge. In more recent paper, EOR defects were observed only in cross section TEM (XTEM) together with several other defects caused by the incomplete amorphization achieved at high ion implantation energy (1 MeV); a second paper showed a clear XTEM image of Ge EORs generated by Ge+ implantation, but surprisingly the reported plan-view (PV) TEM images were from P+ implanted specimens. In fact, following such an implant, P+-induced defects were not found by the same authors to be located just below the original a/c interface, as in the case of a Ge+ implant but were distributed over a much wider layer (down to about twice the a/c interface depth). Moreover, under similar implant and annealing conditions, P is known to precipitate into well visible clusters. Finally, a quantitative TEM study (in cross section specimens), conducted by Hickey et al. on a defect population produced by a nonamorphizing implant shows defects in germanium having an average size of about 10–15 nm that do not change in size during thermal annealing until complete defect dissolution at 750 °C.

In this paper, we will first show that EOR defects in Ge can be observed both in XTEM and PVTEM and give information on their nature and thermal evolution. Then, we will provide an experimental evidence of the correlation between B deactivation/reactivation and defect thermal evolution and discuss the main features of the observed phenomenon with respect to what is known in silicon.

The germanium substrate materials used in this study were of (100) orientation, n type (Sb doped) with a background concentration of 2×1017 cm−3. Preamorphization was done using Ge+ ions 100 keV, 1×1015 cm−2 to obtain an amorphous layer of ~100 nm in agreement with the critical damage energy density model adapted to germanium by Koffel et al. Following the preamorphizing implant, boron was implanted to a dose of 1×1015 cm−2 with an energy of 10 keV. To prevent possible germanium oxidation and outdiffusion, a cap of 200 nm of SiO2 was deposited via plasma-enhanced chemical vapor deposition prior to thermal process. Rapid thermal annealing (RTA) was then performed in a N2 ambient with a ramp-up rate of 10 °C/s, at peak temperatures ranging from 400 to 900 °C for annealing times from 100 s to 3 h, to complete SPER and to activate dopants.

Chemical profiles were studied by secondary ion mass spectrometry (SIMS), using a CAMECA IMS-6f instrument, by collecting B+ secondary ions while sputtering with a 3 keV O2+ beam. Hall effect measurement by HL5500PC Hall–van der Pauw system was performed for sheet resistance (R_s), active dose (N), and drift mobility (μ_d) measurements. Samples were patterned with a “Greek cross” by photolithographic definition and plasma etching to ensure the confinement of the electric field lines. In this configuration, we estimate a 3% error for directly measured quantities (R_s and N) and 6% for drift mobility calculated as μ_d = 1/(εN × R_s),

a)Present address: Now at IM2NP, case 142 Faculté Saint Jérôme 13397 Marseille, cedex 20, France. Electronic mail: federico.panciera@im2np.fr.
EOR defects in Ge exhibit a very weak size evolution during isothermal anneals at 450 °C, they are already completely dissolved after 100 s of annealing times were finally performed on samples submitted to longer annealing processes at 400 °C, NH is found to decrease until it stabilizes at a value of 7.7 × 10^14 cm^−2 and N_H decreases almost to 0. The reason of this decrease is that the boron dopant atoms are deactivated after 100 s of annealing. As already evidenced by Hickey et al., in the case of nonamorphizing implants, our results indicate that EOR defects in Ge exhibit a very weak size evolution during thermal processes.

Finally, TEM was used to investigate defects in the EOR region. TEM specimens for (PV) and cross-sectional (X) observations were prepared by mechanical thinning and ion milling, performed employing low ion energy in order to avoid specimen heating. Samples were imaged using a JEOL 2010 microscope under weak beam dark field (WBDF) conditions using different diffraction vectors, g.

Figure 1(a) shows a XTEM image (g=422) of a sample submitted to annealing at 400 °C for 100 s. Following this annealing, the amorphous Ge layer is completely recrystallized and a band of defects is clearly visible at a ~100 nm depth from the surface. This depth is compatible with the expected position of the a/c interface. Under WBDF imaging conditions, defects appear as white dots on a dark background. The observed defect size is ~6 nm. Such a small size does not allow a direct determination of the defect nature (dislocation loops or {311}s), however this can be achieved thanks to quantitative analysis of PVTEM images, taken under different conditions. In fact, the contrast intensity, I, of a given defect is proportional to the scalar product of the diffraction vector g and the defect Burgers vector b, (1−g·b). Considering that each defect type exists in several differently oriented variants, the apparent defect density is expected to vary depending on the diffracting vector used for the TEM images. For instance, the ratio between the density of visible {311} defects under g=422 and g=400, is ~3, while for dislocation loops is very close to unity. Densities of visible defects in the two imaging conditions showed in Fig. 1(b) (g=422) and Fig. 1(c) (g=400) are substantially identical (around 1 × 10^13 cm^−2), therefore suggesting that the observed defects are not {311}s, but small dislocation loops. Taking an average radius of 3 nm, the total dose of Ge interstitials trapped in the defects results approximately equal to 5 × 10^13 cm^−2 in this sample.

In order to investigate defect thermal evolution, PVTEM were finally performed on samples submitted to longer annealing times (900 and 2700 s) or higher temperature (450 °C for 100 s); at 400 °C, defects do not change in size up to 900 s anneal and rapidly dissolve after longer times; at 450 °C, they are already completely dissolved after 100 s (not shown). As already evidenced by Hickey et al., in the case of nonamorphizing implants, our results indicate that EOR defects in Ge exhibit a very weak size evolution during thermal processes.

Figure 2 shows the depth profiles of B implanted in preamorphized Ge after different annealing conditions. There is no evidence of boron diffusion during the annealing at 500 °C for 15 min (red circles in Fig. 2), even if, under similar conditions, Simoen et al. have observed boron TED in Ge. However, in that study, boron profiles were more abrupt than in our case (4 nm/dec and 50 nm/dec, respectively) and therefore more sensitive to weak diffusion effects. On the other hand, at an annealing temperature of 900 °C, boron diffusion becomes observable, as already demonstrated in literature; almost the totality of the boron profile is immobile and diffusion broadening occurs only for concentrations below 2 × 10^18 cm^−3. This concentration corresponds to the maximum B solubility in Ge at high temperature, as reported in literature. Hall effect measurements in VDP configuration provide a measure of the sheet resistance Rs, Hall mobility (μ_H), and Hall dose (N_H). The Hall scattering factor r_H relates μ_H and N_H to the drift mobility μ_d and the active dose N through the formulas μ_d=μ_H/r_H and N=r_H×N_H. Different values for the Hall scattering factor r_H have been proposed. Unfortunately, there is not yet a universally recognized value, however, r_H is usually found not to deviate significantly from unity. In the following, we will therefore present the raw Hall effect data without r_H correction. Figure 3 reports values of Rs (a) and N_H (b) for samples submitted to process at 400, 450, and 500 °C for various times. As the measured mobility remains approximately constant around 110 cm^2/V s for all samples (not shown), Rs is inversely proportional to N_H. The lowest thermal budget (400 °C for 100 s) provides a Hall dose of 8.7 × 10^14 cm^−2, which corresponds to the total activation of the implanted boron profile. This high level of active concentration, close to 2 × 10^20 cm^−3, has also been achieved in other works, when SPER is performed. After longer processes at 400 °C, N_H is found to decrease until it stabilizes at a value of 7.7 × 10^14 cm^−2, indicating that B deactivation occurs. On the other hand, during isothermal anneals at 450 and 500 °C, N_H steadily increases, suggesting that boron atoms are reactivated in such conditions until N_H reaches the initial "total" activation value (8.7 × 10^14 cm^−2 after 2700 s at 500 °C).

The observed deactivation/reactivation phenomenon is well known for B junctions in Si. Currently accepted...
explanations\textsuperscript{4,5} assume that dopant deactivation is due to the formation of electrically inactive boron-interstitial-clusters (BICs), caused by the migration of interstitials released from the dissolving EOR defects toward the doped region. Once the whole defect population vanishes, dissolution of BICs occurs, resulting in boron reactivation. A similar mechanism is the only satisfactory explanation of the deactivation/ reactivation phenomenon we observe in Ge. Moreover, it is strongly supported by our observations on defects evolution. Boron deactivation occurs within the same time interval necessary to achieve complete defects dissolution while the total deactivated boron dose is of the same order of magnitude \((\sim 1 \times 10^{14} \text{ cm}^{-2})\) as the density of interstitials initially trapped in the EOR defects. Finally, reactivation takes place at higher temperatures, i.e., after the total defect dissolution.

The mechanism that drives deactivation/reactivation in Si and in Ge appears to be the same, however, in Ge, the whole deactivation/reactivation cycle surprisingly takes place while the maximum active B concentration largely exceeds its solubility limit, explaining why it is possible to achieve, in Germanium, highly activated defect-free \(p^+\text{-}n\) junctions.

The authors wish to thank the TEAM service of LAAS-CNRS for the technical support.