In-growth of an electrically active defect in high-purity silicon after proton irradiation

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Defect-related energy levels in the lower half of the band gap of silicon have been studied with transient-capacitance techniques in high-purity, carbon and oxygen lean, plasma-enhanced chemical-vapor deposition grown, n-and p-type silicon layers after 2-MeV proton irradiations at temperatures at or just below room temperature. The in-growth of a distinct line in deep-level transient spectroscopy spectra, corresponding to a level in the band gap at $E_V + 0.357$ eV where E_V is the energy of the valence band edge, takes place for anneal temperatures at around room temperature with an activation energy of 0.95 ± 0.08 eV. The line disappears at an anneal temperature of around 450 K. The corresponding defect is demonstrated not to contain boron, carbon, oxygen, or phosphorus. Possible defect candidates are discussed. © *2013 AIP Publishing LLC*. [http://dx.doi.org/10.1063/1.4841175]

I. INTRODUCTION

A considerable effort has been put into the understanding of the correlation between dissolution of interstitial-type defects and transient-enhanced diffusion of dopants in silicon, in particular of boron.[1] Whereas the kinetics of large interstitial clusters, {113} defects and dislocation loops have been studied in detail[2,3] and comprehensive information has been accumulated on these defects, much less is known about the building blocks of these large defects such as the mono-, di-, tri-, and tetra-interstitial. The assignment of these small interstitial defects to experimentally observed defects has recently been reviewed by Pichler[4] and Carvalho *et al.*,[5] and the reader is referred to these reviews for detailed information.

The present investigation of defect levels in the lower half of the band gap of silicon was initiated in a search for electrically active, radiation-induced, interstitial type defects in silicon. As it is well established that carbon and oxygen impurities in silicon are effective sinks for self-interstitials,[4] a precondition for a successful outcome of such an investigation would be the use of carbon and oxygen lean silicon. Hence, the present investigation is based on carbon and oxygen lean plasma-enhanced chemical-vapor deposition (PECVD) grown silicon.

II. EXPERIMENTAL PROCEDURE

Defect levels in the lower half of the band gap of silicon produced by 2-MeV proton irradiations are studied using transient-capacitance techniques such as deep-level transient spectroscopy (DLTS) and minority-carrier transient spectroscopy (MCTS).[6] It is well established that for standard float-zone refined or Czochralski-grown silicon irradiated with MeV protons, carbon-, and oxygen-related defect complexes are produced, having energy levels in the lower half of the band gap;[7] spectral lines related to these complexes might dominate the DLTS spectra of p-type silicon depending on the O and C concentrations. Moreover, hydrogen-related defects are also known to give rise to strong signals in DLTS spectra of p-type silicon.[8] To avoid any confusion from these signals, the present experiments were performed in n- and p-type silicon layers grown epitaxially on a standard silicon substrate by ultra-high vacuum, plasma-enhanced chemical-vapor deposition (UHV PECVD). These epitaxial layers were grown at 700 °C to thicknesses between 5 and 10 μm, and both p- and n-type layers doped with B or P, respectively, to concentrations of about 5×10^{15} cm^{-3} (1-2 Ω cm), were produced. The oxygen and carbon concentrations in one of the p-type wafers with a 5 μm thick epitaxial layer, in which most of the experiments of the present investigation are based on, were determined by a secondary ion mass spectrometry (SIMS)-based bulk analysis technique called raster change (*Evans Analytical Group*), and were found to be below the detection limits of 2×10^{15} and 5×10^{15} cm^{-3} for C and O, respectively. Thus, the epitaxial layers are *carbon and oxygen lean*.

The PECVD equipment was controlled at the time of the depositions for metallic contaminations, and the test indicated a level below 1×10^{10} cm^{-3}. Moreover, the gases used were high-purity gases (99.9999) and the growth chamber was made from quartz. Hence, it can with confidence be assumed that metallic contamination of the epitaxial layers is not an issue in the present investigation.

Most of the diodes were either n^+p- or p^+n-mesa diodes, where the 0.3 μm thick n^+- or p^+-top layers were made by molecular-beam epitaxial (MBE) growth of silicon doped with Sb or B, respectively, to concentrations of around 5×10^{19} cm^{-3} at a temperature of 300 °C. Some of the mesa diodes were annealed at 700 °C for 30 min prior to proton

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irradiation in order to drive-out any hydrogen which might have been introduced during the various chemical treatments. No difference in the DLTS spectra was found between annealed and un-annealed diodes after irradiations demonstrating that hydrogen-related defects are not an issue when using these mesa diodes. One n^+p-mesa diode was made on float-zone refined, p-type silicon to investigate carbon- and oxygen-related defects. A few Schottky diodes were made on the PECVD-grown layers, by Al or Au evaporations through diode defining masks.

Irradiations were performed with 2-MeV protons to a fluence of typically $5 \times 10^{12}\,cm^{-2}$ with the diodes kept at room temperature or 280 K, and the DLTS and MCTS measurements were performed *ex-situ* and *in-situ*, on-line with the irradiation facility.

III. RESULTS AND DISCUSSIONS

A typical DLTS spectrum of a n^+p diode made from $\sim 2\,\Omega cm$ epitaxial p-type silicon, irradiated at room temperature with 2-MeV protons, and kept at this temperature for a couple of days prior to measurements, is shown in Fig. 1. The DLTS spectrum consists of two strong lines and some weak lines. The inset of the figure shows Arrhenius plots of the emission rate corrected for the T^2 dependence for the two pronounced lines in the spectrum. The DLTS-finger prints (level position in the band gap and apparent capture-cross section) extracted from these plots are $E_V + 0.19\,eV$, $\sigma_a = 4 \times 10^{-16}$ cm^2 and $E_V + 0.38\,eV$, $\sigma_a = 2 \times 10^{-14}$ cm^2 for the low- and high-temperature peaks, respectively, where E_V is the energy of the valence band edge, and σ_a is the apparent capture-cross section.[6] From its finger print, the low temperature line is assigned to the single-donor state of the di-vacancy.[9] The high-temperature line, which we will call the AU line in the following, will take the leading part in this paper. Its intensity is always found to be ~60% of the intensity of the di-vacancy line. The di-vacancy concentration is $\sim 5 \times 10^{13}\,cm^{-3}$, and that of the AU defect $\sim 3 \times 10^{13}\,cm^{-3}$. The small lines seen in the spectrum are probably related to the very small carbon and oxygen contents of the epitaxial p-type silicon layer; they will not be discussed any further in this paper.

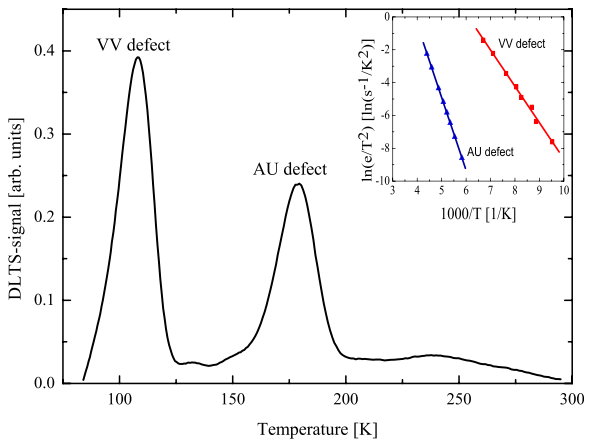

FIG. 1. DLTS spectrum of a n^+p-diode after 2-MeV proton irradiation to a dose of $9 \times 10^{12}\,cm^{-2}$ at room temperature. The diode was kept at room temperature for several days prior to DLTS measurement. The emission rate was $20\,s^{-1}$ and the pulse width 200 μs. The inset shows Arrhenius plots of the emission rate corrected for the T^2 dependence for the two pronounced peaks in the spectrum.

High-resolution Laplace DLTS (LDLTS)[10] investigations of the AU line were performed in order to check whether it involves more than one component. Only one line was found, and an accurate determination of the DLTS finger print was made resulting in $E_V + 0.357 \pm 0.005\,eV$ and $\sigma_a = 4 \times 10^{-15}$ cm^2. This finger print will be used in the following as being characteristic of the AU line.

Similar diodes have previously been irradiated with 2-MeV electrons to a dose which produces a di-vacancy line of an intensity similar to the one shown in Fig. 1. Under these conditions, the AU line was always smaller than 10% of the di-vacancy line, and was difficult to separate from other small lines in the DLTS spectra (see the DLTS spectra in Fig. 3 of Ref. 11 where the AU line should be found at ~200 K).

The position of the AU line in the LDLTS spectrum as a function of the electric field was investigated, as the effect of the electric field on the emission rate (Poole-Frenkel effect) contributes to the disclosure of the charge state of a defect center.[6] To vary the electric field on the diode, the reverse bias was varied from 5 to 20 V. At a temperature of 175 K, the AU line is found at an emission rate of $10\,s^{-1}$ in the LDLTS spectrum. In case of a Poole-Frenkel effect, and with the actual doping level of $5 \times 10^{15}\,cm^{-3}$, a change in reverse bias from 5 to 20 V would shift the line position in the LDLTS spectrum from 10 to $30\,s^{-1}$ which would be easily detectable. We do not, however, observe a shift when bias is changed from 5 to 20 V within the experimental uncertainty of about $\pm 5\,s^{-1}$. This strongly indicates that the energy level associated to the AU line is of the donor type.[6]

The projected range of 2-MeV protons in silicon is 48 μm (Ref. 12) and, with a reverse bias of 5 V which was typically used in the experiments, the depletion layer and thereby the investigated depth range extend less than 1 μm into the p-type layer. Thus, the depth region from which the spectrum of Fig. 1 stems is far away from the implanted hydrogen, and from end-of-range defects. Nevertheless, prior to irradiation, one diode was thinned down from the back to a thickness smaller than the range of the implanted hydrogen; a spectrum similar to the one shown in Fig. 1 was obtained demonstrating that end-of-range defects and implanted hydrogen have had no strong influence on the spectrum shown in Fig. 1.

It was soon realized that the intensity of the AU line grew with time when the diodes were kept at room temperature after irradiation. An example of this is shown in Fig. 2 which displays DLTS spectra measured after annealing at 320 K for two different times. A small reduction of the intensity of the di-vacancy line was observed to accompany the in-growth of the AU line. Diodes for a detailed investigation of the in-growth kinetics were irradiated with 2-MeV protons at a temperature of 280 K. This irradiation temperature was chosen as a compromise value for which the DLTS signals from the boron-vacancy pairs, which could otherwise obscure the analysis, have almost annealed out,[13] and the AU line has not yet grown significantly in intensity. The results from the isothermal-annealing experiment are shown

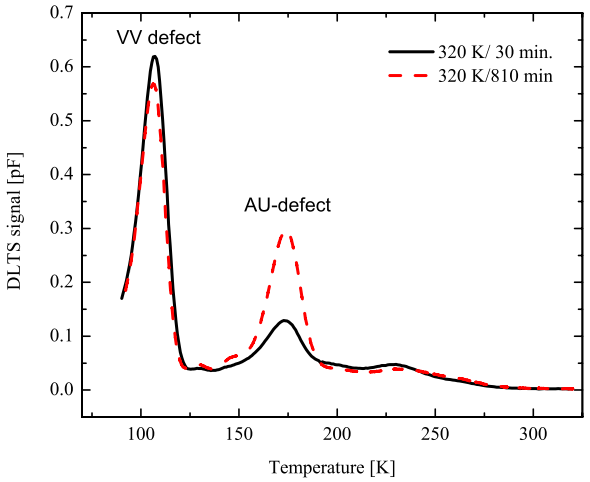

FIG. 2. DLTS spectra of a p^+n diode after 2-MeV proton irradiation at 280 K measured after annealing at 320 K for 30 and 810 min , respectively.

in Fig. 3 for the four mentioned temperatures. The full curves are fits to the data points with functions of the form[14]

$$\frac{I(t)}{I_{\max}} = 1 - e^{-a\cdot(t-t_0)},$$

where $I(t)$ is the time-dependent line intensity in the DLTS spectrum, I_{max} is the maximum intensity of the line, a is the temperature-dependent growth rate, and t_0 is the time at which the in-growth starts. From these fits, the growth rates at the four different temperatures were extracted, and they are plotted in an Arrhenius plot in Fig. 4. As it appears from Fig. 4, first-order kinetics prevails in the investigated temperature range; hence, the temperature dependence of the growth rate is given by[14]

$$a = k_0 e^{-\frac{E_a}{kT}},$$

where k_0 is the pre-exponential factor of the rate constant, E_a the activation energy, k the Boltzmann constant, and T the absolute temperature. This function has been fitted to the data points of Fig. 4, and the full line shows the result. The extracted activation energy for the in-growth is

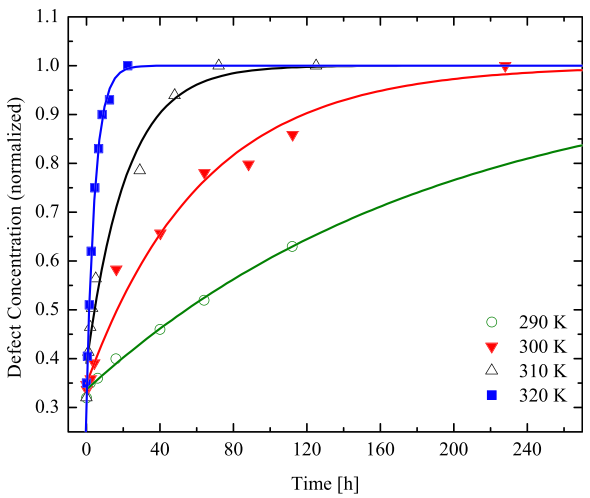

FIG. 3. Results from isothermal in-growth experiments with the AU-peak for the 4 mentioned temperatures after 2-MeV proton irradiations at 280 K. The full curves are fits to the data (see text).

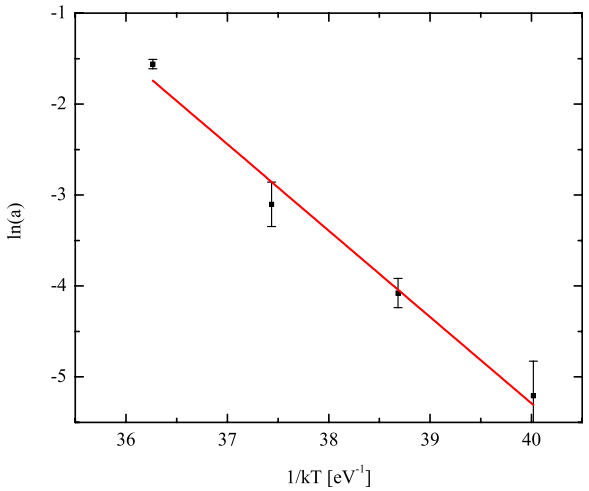

FIG. 4. Arrhenius plot of the in-growth rates, a, extracted from the data of Fig. 3. The full line is a fit to the data points (see text). The shown uncertainties are standard errors from the fits to the in-growth curves of Fig. 3.

$E_a = 0.95 \pm 0.08$ eV and the pre-exponential factor is $k_0 = 2 \times 10^{10}\,s^{-1}$. The small pre-exponential factor indicates that long range migration takes place, involving $\sim 10^3$ steps occurring at a rate of $\sim 10^{13}\,s^{-1}$.[15]

MCTS experiments were carried out on p^+n-diodes in order to examine if the AU defect was also formed in PECVD grown, n-type silicon after 2-MeV proton irradiation at room temperature. A typical MCTS spectrum is shown in Fig. 5. Two pronounced lines are observed: A line at a temperature of about 150 K which from its MCTS finger prints can be identified as the single-donor line of the phosphorus-vacancy pair (the E center).[11] The other prominent line at about 180 K has a MCTS finger print of $E_V + 0.357$ eV and $\sigma = 2 \times 10^{-15}$ cm^2 which is almost exactly the same finger print as found for the AU line. The presence of this line in a MCTS spectrum of n-type epitaxially grown silicon was already noted in Ref. 11; however, the extracted finger print was disturbed by the presence of a strong positive signal from the acceptor level of the E center. The agreement between the two sets of finger prints is remarkably good, strongly indicative of the same defect being responsible for the two lines.

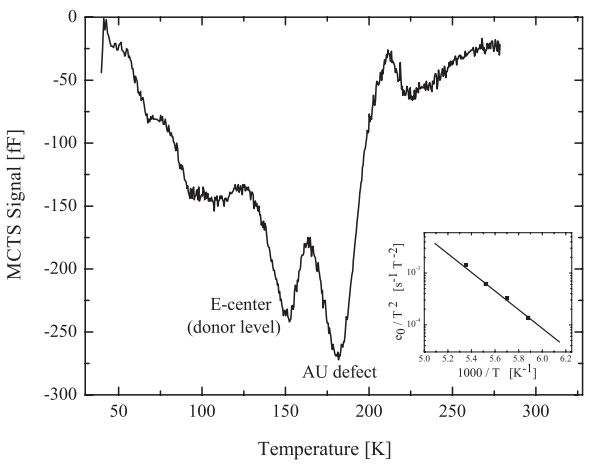

FIG. 5. MCTS spectrum of a p^+n-diode after 2-MeV proton irradiation at room temperature. The diode was kept at room temperature for several days prior to MCTS measurements. The inset show an Arrhenius plot of the emission rate corrected for the T^2 dependence of the AU-line.

Annealing experiments were performed on the AU line in epitaxial n- and p-type diodes after 2-MeV proton irradiations at room temperature, and the line intensities were followed by MCTS and DLTS measurements, respectively. The results are shown in Fig. 6. As can be seen, the AU line anneals at ~400 K when observed in p-type Si and at ~475 K when observed in n-type Si. As discussed above, there are indications as to the AU level being a donor level and, *qua* its position in the band gap the Fermi level is below the AU level for temperatures below ~450 K in the p-type Si, and in the n-type Si it is always above. This means, that for annealings below 450 K, the AU defect is in a definite charged state in p-type Si (positive if it is a single donor level), whereas in n-type Si, it is in a different charge state (neutral if it is a single donor level). Thus, the AU defects anneal in different charge states in the two cases, and it is quite common for defects in silicon that their mobility or stability depend slightly on their charge state. Note also, that for temperatures above ~450 K, the annealing of the AU line in p-type silicon slows down, and that this is the temperature at which the majority of the AU defects changes charge state. Hence, we may conclude that the AU defect is formed in both n- and p-type silicon, and consequently, neither B nor P can be involved in the defect. That boron is not part of the defect and has also been reached in a more direct way: Interstitial boron, B_i, is created in boron doped p-type silicon as a result of irradiation, and becomes mobile at 250–300 K with an activation energy of diffusion of 0.6 eV.[16] It then forms a number of secondary defects, e.g., B_iO_i with a level at $E_V + 0.26$ eV and B_iC_s with a level at $E_V + 0.30$ eV.[7] None of these levels corresponds to the AU level. More importantly is, however, that we have not found any correlation between the disappearance of the B_i related lines and the appearance of the AU line. The B_i defect has two levels in the upper half of the band gap, at $E_C = 0.13$ eV and $E_C = 0.37$ eV,[17] and they can be studied in p-type silicon by MCTS using pulsed light. We carried out these experiments after low-temperature irradiation, and followed both the B_i-related lines as a function of annealing and, parallel to that, followed the in-growth of the AU line using standard DLTS. As the B_i lines disappeared completely before the AU line had finished its in-growth, it can be concluded that boron is not involved in the AU defect.

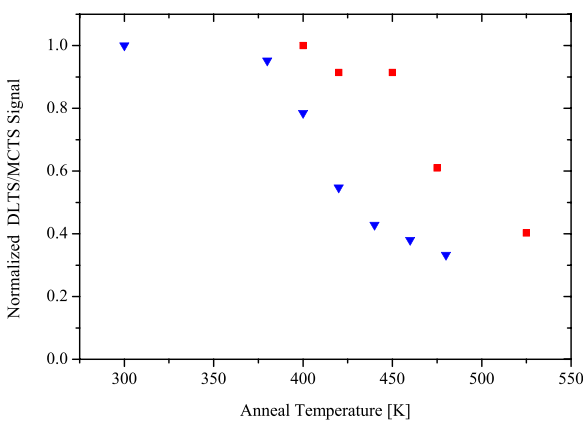

FIG. 6. Results from 15-min isochronal annealing experiments with the AU-line observed in (■) MCTS spectra of 2-MeV proton irradiated p^+n diodes, and (▼) DLTS spectra of 2-MeV proton irradiated p-type Schottky diodes. To avoid interference from the E-center line, the MCTS results are normalized to the 400 K data point as the E-center line has annealed at this temperature.

Several defects in Si grow in with activation energies in this range after particle irradiation. The complex of interstitial carbon and interstitial oxygen, C_iO_i, has associated to it a level in the band gap at $E_V + 0.35$ eV which would be very difficult to distinguish from the AU line.[17] In irradiated diodes made from p-type, float zone refined silicon, which contain oxygen and carbon of the order of 10^{16} cm^{-3}, this line is always significant in the DLTS spectra. The C_iO_i defect forms when C_i, created as a result of the irradiation, becomes mobile at temperatures of about 320 K with an activation energy of diffusion of 0.80–0.87 eV.[16] The C_i defect has a level in the lower half of the band gap at $E_V + 0.28$ eV which is clearly visible in diodes of p-type, float-zone refined silicon.[7] We have irradiated a diode made from p-type float-zone refined silicon with 2-MeV protons at 280 K, and have observed a strong line related to C_i which converts in an almost one-to-one ratio to C_iO_i after an annealing at about room temperature. The C_i line was not observed in diodes made from the PECVD-grown silicon in agreement with the SIMS measurements demonstrating that in the PECVD samples the carbon concentration is very low. Moreover, as the oxygen concentration is also very low, it can be excluded that the AU line is the C_iO_i line.

Markevich *et al.*[18] has reported the observation of a radiation induced defect in p-type Cz Si with high oxygen concentration ($\sim 1 \times 10^{18}$ cm^{-3}), and low carbon ($\leq 2 \times 10^{15}$ cm^{-3}) and boron ($\leq 2 \times 10^{14}$ cm^{-3}) concentrations which in some respects resemble the AU defect. They used 4-MeV electron irradiations with the Si diodes at room temperature, and found two lines with DLTS and LDLTS at $E_V + 0.123$ eV and $E_V + 0.358$ eV which disappeared in the temperature range of 90–120 °C upon 30 min isochronal annealing. The apparent capture-cross of the $E_V + 0.358$ eV defect was found to be $\sigma_a = 4.3 \times 10^{-15}$ cm^{-2}. The two levels were shown to belong to two different configurations of the same bistable negative-U defect, and, supported by *ab-initio* calculations, they concluded that the defect was an oxygen dimer with a trapped Si-self interstitial, IO_{2i}. This interpretation was later on confirmed by Yarykin and Weber[19] using 5-MeV electron irradiations of Cz-Si samples at room temperature, and they furthermore demonstrated a linear relationship between the oxygen dimer concentration and the concentration of the $E_V + 0.358$ eV defect observed using DLTS. The DLTS fingerprints of the $E_V + 0.358$ eV line are remarkably similar to that of the AU line, and the annealing temperatures are similar. However, we do not find that the AU defect is bistable with a component at $E_V + 0.123$ eV, and, more importantly, the oxygen concentration in the PECVD grown layers is so low that the concentration of oxygen dimers will be vanishingly small ($\lesssim 10^{10}$ cm^{-3}).[20] Thus, the AU defect cannot be the IO_{2i} defect.

Another somewhat similar defect to the AU defect has been observed by Cho *et al.*[21] and Yarykin *et al.*[22] after low-temperature He, Si, and Ge ion implantations at MeV energies into Schottky diodes of p-type Czochralski silicon and p-type epitaxial silicon using on-line DLTS measurements.

They found a line in their spectra (which they called the K2 line) with the DLTS finger print $E_V + 0.36\,eV$ and $\sigma = 6.6 \times 10^{-15}\ cm^2$ which grew in at temperatures between 200 and 250 K after 2.4-MeV Ge implantations. In the diodes made from Czochralski silicon, this line merged with the C_iO_i line, and only in diodes made of epitaxial silicon, they could observe the undisturbed K2 line which disappeared completely at 300 °C. They did not observe the K2 line when using forward bias injection in p-n junctions prepared on n-type Czochralski silicon, and this led them to conclude that the defect could involve boron as it was only observed in p-type Si. The fact that the line grew-in at about the temperature at which the vacancy in p-type Si (the neutral charge state) becomes mobile led Cho *et al.*[21] to conclude that the defect could include a vacancy. Thus, their conclusion was that the K2 defect is a vacancy-related complex including one or more boron atoms. Although the K2 and AU defects have much in common, there are marked differences: The AU defect grows in at a higher temperature than does the K2 defect, and it is also observed in n-type silicon. A major difference between the experimental conditions in which the K2 line and the AU defects are observed is that the K2 defect is formed in a very dense collision cascade produced as a result of a 2.4-MeV Ge implantation in silicon, whereas the AU defect forms in a much less dense collision cascade from the 2-MeV protons. As already noted by Cho *et al.*,[21] the K2 defects might be located in partly electrically compensated regions which might thus not be accessible by DLTS. Moreover, the compensated regions might also anneal during the annealing experiments which could confuse the picture. Thus, the K2 and AU defect could well be the same defect; however, the AU defect is situated in less disturbed surroundings.

It is well established that by using transient capacitance techniques, as we do in the present investigation, an unambiguous identification of a defect is not possible; identifications based on these techniques always rely to some degree on educated speculations. Above, we have demonstrated that neither carbon, oxygen, boron nor phosphorus is part of the AU defect, and that none of the impurity-related defects known to us with DLTS signatures similar to that of the AU defect can explain the experimental findings for the AU defect. This seems to point to a defect entirely composed by native defects, the vacancy and the self-interstitial. In the following, we will briefly discuss which type of defect the AU defect could possibly be.

Let us first discuss defect candidates in silicon made from vacancies. In the collision cascades of the 2-MeV protons, mono-vacancies, di-vacancies, and, to a lesser extent, and probably closer to the end-of range of the protons, also tri-vacancies are formed. The appearance of the AU line at temperatures just above room temperature cannot be correlated with the migration of neutral vacancies which takes place at temperatures around 200 K with an activation energy of 0.45 eV (Ref. 16) or with the migration of vacancies released from the BV pairs at their anneal temperatures ≤ 280 K.[13] The di-vacancies are stable to temperatures of around 550 K (Ref. 16) which excludes that it is the precursor of the AU defect. Besides, the change of the di-vacancy concentration when the AU defect grows in is small compared to the change of the AU defect which suggests that the di-vacancy is not involved in the AU defect. Recent reports on the tri-vacancy in silicon[23,24] have shown that this defect is a bistable defect which has two donor levels in the lower half of the band gap at $E_V + 0.106\,eV$ and $E_V + 0.193\,eV$, and that it becomes mobile at temperatures higher than 200 °C. These observations exclude that the AU defect is related to the tri-vacancy. Thus, this discussion of vacancy defects points towards an AU defect which does not include vacancies.

There is no firm experimental evidence as to levels in the band gap related to the mono-interstitial, but there are some recent theoretical estimates based on first-principles calculations.[25,26] Lopez and Fiorentini[25] found a double-donor and a single-donor level at 0.4 eV and 0.7 eV from the valence band, respectively, and a single-acceptor level very close to the conduction band; Jones *et al.*,[26] however, found that the double- and single-donor levels overlap at an energy of 0.7 eV from the valence band, and they did not report the finding of any acceptor level. Although this does not exclude that the AU defect could be the mono-interstitial, in n-type silicon, where the AU defect is also observed, the mono-interstitial is known to start migrating at a temperature of about 150 K which is seen by the appearance of interstitial carbon as a result of the interaction between migrating mono-interstitials and substitutional carbon impurities.[17] This excludes that the AU defect could be the mono-interstitial.

If the AU defect should be the di-interstitial, then it must have been formed either from migration of mono-interstitials or directly in the collision cascade. The former case would be characterised by an in-growth where the activation energy should be that of the migration energy of the mono-interstitial. According to the theoretical estimates of the band-gap levels of the mono-interstitial discussed above, the mono-interstitial is expected to be positively charged in p-type and neutral in n-type silicon. The observed migration temperature of the mono-interstitial in n-type silicon of about 150 K is in agreement with the theoretical estimate of the activation energy of 0.26 eV by Lopez and Fiorentini[25] for the neutral mono-interstitial, but for the doubly positively charged interstitial, they find a value of 1.2 eV; thus the migration temperature in p-type silicon is expected to be significantly higher than 150 K. The small pre-exponential factor to the rate constant observed for the in-growth of the AU defect supports its formation via long-range migration. Against it speaks, however, that the mono-interstitials are positively charged according to theory and will, consequently, repel each other. If the di-interstitial is formed directly in the collision cascade, which could be supported by the observation of a larger introduction rate with protons than with electrons, then a reconfiguration of the di-interstitial at room temperature as discussed by Kim *et al.*[27] and Lee[28] could explain the observed in-growth. A single-donor level at $E_V + 0.3\,eV$ was suggested for the di-interstitial by Eberlein *et al.*[29] based on the first principles density-functional method employing Gaussian orbitals (AIMPRO); this level fits well to the donor level observed for the AU defect.

The activation energy for the migration of the di-interstitial has been theoretically estimated by several groups

to values between ~0.2 and ~1 eV.[30] From this, it cannot be excluded that the di-interstitial could be the precursor of the AU defect which could then be the tetra-interstitial that is expected to be rather stable.[4] Coomer *et al.*[31] have shown with *ab-initio* calculations that the tetra-interstitial has only one level in the bandgap, a donor level between $E_V + 0.16$ and $E_V + 0.27$ eV. In contrast to this are *ab-initio* calculations by Kohyama and Takeda[32] demonstrating the lack of levels related to the tetra-interstitials in the band gap.

There is a difference between the formation of the AU defect in p- and n-type silicon: Whereas the AU-line in p-type silicon is only significant for 2-MeV proton irradiations or heavier particle irradiations, and barely detectable for a 2-MeV electron irradiation to a fluence resulting in the same concentration of di-vacancies as in the case of the proton irradiation, in n-type silicon, it is significant for electron as well as for heavier particle irradiations. This behaviour is understandable if in n-type silicon, the AU defect is formed from two or more electrically neutral or oppositely charged migrating defects formed as a result of the electron irradiation, whereas in p-type, it is only formed as a primary defect preferentially produced by proton irradiations. This scenario would occur if the migrating defects, which form the AU defect in n-type silicon, change sign from neutral to positively charged when going from n-type to p-type silicon. This scenario suggests that the AU defect could be a defect complex possibly involving self-interstitials.

IV. SUMMARY

In summary, irradiation induced defects in high-purity, carbon and oxygen lean UHV-PECVD grown n- and p-type silicon layers have been studied after 2-MeV proton irradiations with transient capacitance techniques. The ingrowth of a distinct line in DLTS spectra, called the AU line, corresponding to a level in the band gap at $E_V + 0.357$ eV for anneal temperatures of around room temperature has been studied. The activation energy for the in-growth is $E_a = 0.95 \pm 0.08$ eV and the pre-exponential factor is $k_0 = 2 \times 10^{10}\,s^{-1}$. Its anneal temperature is found to be ~400 K when positively charged and ~475 K when neutral, assuming it to be a singly charged donor level. It is demonstrated that neither boron, carbon, oxygen nor phosphorus is part of the defect, and that the constituents of the defect must be search for among the native defects. From a discussion of small vacancy and self-interstitial defects, it is concluded that the AU defect is most probably a small interstitial-type defect.

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