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2009 Semicond. Sci. Technol. 24 075002

(http://iopscience.iop.org/0268-1242/24/7/075002)

IOP Publishing Semiconductor Science and Technology
Semicond. Sci. Technol. **24** (2009) 075002 (7pp) doi:10.1088/0268-1242/24/7/075002

Radiation-induced defects in Czochralski-grown silicon containing carbon and germanium

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Received 22 December 2008, in final form 10 April 2009
Published 26 May 2009
Online at stacks.iop.org/SST/24/075002

Abstract
Formation processes of vacancy-oxygen (VO) and carbon interstitial-oxygen interstitial (C_iO_i) complexes in electron-irradiated Czochralski-grown Si crystals (Cz–Si), also doped with Ge, are investigated. IR spectroscopy measurements are employed to monitor the production of these defects. In Cz–Si with carbon concentrations [C_s] up to 1×10^{17} cm^{-3} and Ge concentrations [Ge] up to 1×10^{20} cm^{-3} the production rate of VO defects as well as the rate of oxygen loss show a slight growth of about 10% with the increasing Ge concentration. At high concentrations of carbon [C_s] around 2×10^{17} cm^{-3} the production rate of VO defects is getting larger by ~40% in Cz–Si:Ge at Ge concentrations around 1×10^{19} cm^{-3} and then at [Ge] $\approx 2 \times 10^{20}$ cm^{-3} this enlargement drops to ~13%, thus approaching the values characteristic of lesser concentrations of carbon. A similar behavior against Ge concentration displays the production rate of C_iO_i complexes. The same trend is also observed for the rate of carbon loss, whereas the trend for the rate of oxygen loss is opposite. The behavior of Ge atoms is different at low and high concentrations of this isoelectronic impurity in Cz–Si. At low concentrations most isolated Ge atoms serve as temporary traps for vacancies preventing them from indirect annihilation with self-interstitials. At high concentrations Ge atoms are prone to form clusters. The latter ones are traps for vacancies and self-interstitials due to the strain fields, increasing the importance of indirect annihilation of intrinsic point defects. Such a model allows one to give a plausible explanation for the obtained results. A new band at 994 cm^{-1} seen only in irradiated Ge-doped Cz–Si is also studied. Interestingly, its annealing behavior was found to be very similar to that of VO complexes.

1. Introduction

Oxygen and carbon are the most important impurities in Si. They are inadvertently added in the crystals during growth with the Czochralski method. Oxygen atoms occupy interstitial sites in the Si lattice with concentrations up to 2×10^{18} cm^{-3}. Carbon atoms occupy substitutional sites with concentrations up to 5×10^{17} cm^{-3}. Both impurities are electrically inactive. This is very fortunate for silicon-based devices, since the concentrations of donors and acceptors introduced for various applications are typically few orders of magnitude lower than those of oxygen and carbon. Thus, any problem that would arise from the electrical activity of both impurities is avoided. However, carbon and oxygen display a tendency to react with intrinsic point defects, i.e., vacancies and self-interstitials, during crystal growth or processing. They can also react between themselves and with other impurities and imperfections present in the crystal [1, 2]. Heat treatment, implantation and irradiation usually employed for the fabrication of silicon devices introduce vacancies and self-interstitials. Therefore, the interest to their reactions with carbon and oxygen impurities is very high.

Table 1. The germanium concentration, the oxygen and carbon concentrations before and after the electron irradiation, as well as the production rate of VO and C_iO_i defects and the rate of oxygen and carbon loss in the electron-irradiated (dose: 5×10^{17} cm^{-2}), Ge-doped Si samples used.

Sample name	[Ge] cm^{-3}	$[O_i]_o$ 10^{17} cm^{-3}	$[O_i]_{a.\,i.}$ 10^{17} cm^{-3}	$\eta_{\Delta Oi}$ cm^{-1}	$[C_s]_o$ 10^{16} cm^{-3}	$[C_s]_{a.\,i.}$ 10^{16} cm^{-3}	$\eta_{\Delta Cs}$ cm^{-1}	η_{VO} cm^{-1}	η_{CiOi} cm^{-1}
Cz–Si	0	9.56	9.30	0.052	<2.0	[a]	[a]	0.047	[a]
Cz–Si :Ge-1	1×10^{17}	9.60	9.05	0.110	2.0	<2.0	[a]	0.068	0.031
Cz–Si :Ge-2	7×10^{17}	6.50	6.20	0.060	<2.0	[a]	[a]	0.050	[a]
Cz–Si :Ge-3	1×10^{18}	10.00	9.42	0.116	3.0	<2.0	[a]	0.070	0.039
Cz–Si :Ge-4	4×10^{18}	5.55	4.88	0.134	10.0	4.9	0.102	0.072	0.051
Cz–Si :Ge-5	1×10^{19}	6.74	5.78	0.192	20.0	10.1	0.198	0.100	0.084
Cz–Si :Ge-6	5×10^{19}	7.60	7.30	0.060	<2.0	[a]	[a]	0.062	[a]
Cz–Si :Ge-7	1×10^{20}	8.77	8.03	0.148	3.7	<2.0	[a]	0.075	0.046
Cz–Si :Ge-8	2×10^{20}	7.70	6.53	0.234	18.0	10.0	0.160	0.085	0.077

[a]Below detection limit.

Oxygen is the main trap for vacancies. Noticeably, most of the produced vacancies during irradiation are readily trapped by oxygen atoms to form VO pairs (the well-known A-centers). Although divacancies and V_2O are also formed, the concentration of VO centers is usually taken, to a first approximation, as a measure of the amount of vacancies produced by the irradiations. Upon thermal anneals VO centers react with oxygen and vacancies leading finally to the formation of a variety of multivacancy-multioxygen (V_nO_m) centers [3, 4]. Most of these V_nO_m centers introduce levels in the energy gap affecting the electrical activity of the material.

Carbon, being an isovalent element but having smaller covalent radii than that of Si, is the main trap for self-interstitials [2]. Noticeably, most of the produced self-interstitials during irradiation are selectively trapped by carbon atoms. Initially carbon interstitials formed through the kick-out mechanism (C_s+Si_I→C_i). Although self-interstitials clusters and also other defects as for example $O_i(Si_I)$ are formed as well, the concentration of C_i centers is usually considered as a measure of the amount of self-interstitials produced by the irradiation. C_is readily react with C_s and O_i atoms to form the C_iC_s and C_iO_i defects which are the main C-related defects [2] in C-rich Si. Both defects are electrically active introducing deep levels in the energy gap of Si. For higher doses these defects can trap self-interstitials leading to the formation of $C_iO_i(Si_I)$ and the CiCs(Si_I) complexes [2, 5].

The importance of the VO, C_iO_i and C_iC_s defects for the Si-based electronic industry is obvious and therefore the interest for their knowledge and control is very high. There are many ways used for improvement of Si materials in various applications. Enhancement of radiation and thermal resistance is possible by introduction of isoelectronic impurities in the lattice, among them germanium. Ge atoms being isoelectronic with Si are introduced at substitutional sites and are electrically inactive. Due to their larger covalent radii than those of Si they introduce internal strains which are expected to affect the production of defects, and the interactions between themselves. Generally, the characteristics of the various fundamental processes taking place at the various stages of material processing are affected. Carbon follows a manifold of reaction patterns in Cz–Si, as electrical and optical measurements [2, 6–11] have verified. These phenomena should be investigated particularly when another isovalent impurity as that of Ge is present. The purpose of this work is to study the effect of Ge doping on the production of VO and C_iO_i defects as well as the oxygen and carbon loss, in electron-irradiated Cz–Si.

2. Experimental details

The samples used in this work were cut from Czochralski prepolished Si wafers. Their dimensions were 20 × 10 × 2 mm^3. The corresponding initial germanium, carbon and oxygen concentrations in those samples are cited in table 1. The concentrations of Ge impurity were estimated from the mass ratio of Si and Ge in the melt provided by the supplier's certificate. For Ge-rich SiGe crystals Raman spectroscopy is known to be an effective tool for determining both the Ge content and residual strain [12]. Unfortunately, in our case for SiGe crystals with Ge concentrations of about 1 at.% or less this technique can be used only for a qualitative comparison of Raman spectra, since the relations between the frequencies of optical phonons and Ge content in this concentration range are unknown. Because of this the mass ratio of Si and Ge in the melt appears to be an only appropriate guide to the scale of the Ge content. The samples were irradiated with 2 MeV electrons, with a dose of 5×10^{17} cm^{-2}, using the Dynamitron accelerator at Takasaki-JAERI (Japan). After the irradiation all the samples were subjected to isochronal anneals up to 800 °C, in steps of $\Delta T \approx 10$ °C and $\Delta t = 20$ min. After each annealing step, the IR spectra were recorded at room temperature by means of a FT-IR spectrometer with a resolution of 1 cm^{-1}. The two-phonon intrinsic absorption was always subtracted by using a float-zone sample of equal thickness. As we see from figure 1, the two-phonon absorption was not removed completely. As a result the 605 cm^{-1} line due to substitutional carbon [2] was perturbed. There are some dips which overlap with this line and affect its shape and intensity. It is known from the literature [13] that a strong two-phonon absorption exists in the region of the 605 cm^{-1} line. Actually, the 605 cm^{-1} line is located very close to an intense lattice band at ∼ 610 cm^{-1} [14]. Thus, the C_s concentration in

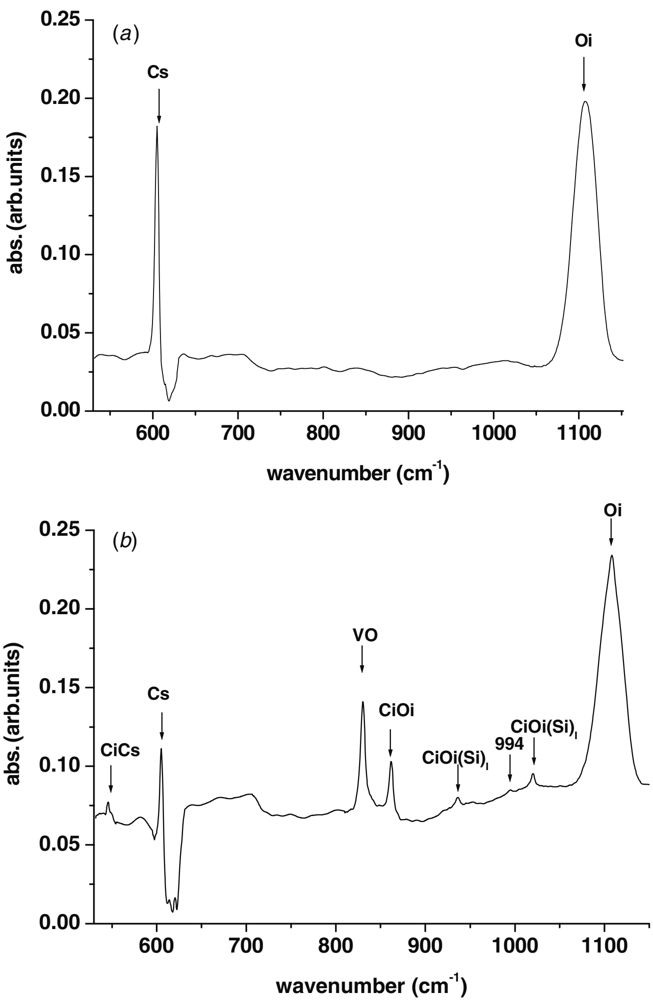

Figure 1. IR spectrum of the sample Cz–Si:Ge-5 prior to (*a*) and after irradiation (*b*).

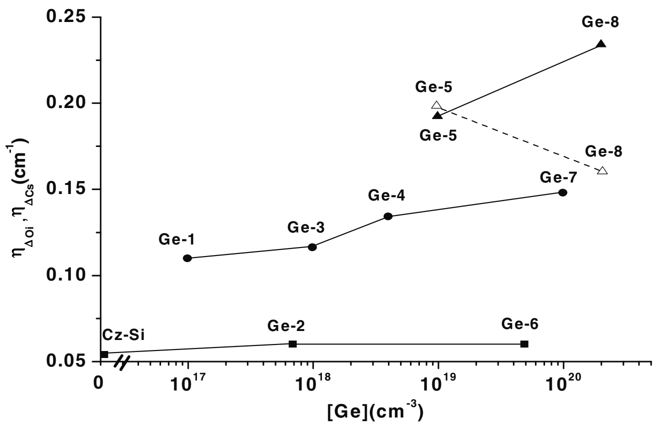

Figure 2. Rate of oxygen (solid lines) and carbon (dashed lines) loss versus [Ge].

our samples was determined from the strength of the 605 cm^{-1} band measured at room temperature in the following way. The ASTM procedure [15] with a conversion factor 1×10^{17} cm^{-2} [16] was used. A straight baseline from 595 to 610 cm^{-1} was drawn in a spectrum and the concentration of C_s was then calculated from the amplitude of the peak. The concentration of oxygen was determined by measurements of the 1107 cm^{-1} absorption band using a calibration coefficient of 3.14×10^{17} cm^{-2} [17]. This line of interstitial oxygen is very strong and its position in spectra is practically unperturbed by the lattice absorption. A straight baseline from 1050 to 1150 cm^{-1} was drawn in the spectra and from the amplitude of the peak we calculated the concentration of O_i. Note that in the experiments described below we are mainly dealing with the differences of oxygen and carbon concentrations prior to and after irradiation and, therefore, any errors in concentrations being of a few per cents are more or less compensated.

3. Experimental results and discussion

Figure 1 exhibits the IR spectra of Cz–Si:Ge-5 sample prior (a) and after (b) the electron irradiation. The spectra of other samples are similar, only with differences in the amplitudes of the bands of the corresponding defects. The well-known bands of the O_i (1107 cm^{-1}), C_s (605 cm^{-1}), VO (830 cm^{-1}), C_iO_i (862 cm^{-1}) are shown. Additionally, two bands at (936, 1020 cm^{-1}) of the $C_iO_i(Si_I)$ defect [2, 18], one band at 546 cm^{-1} of the C_iC_s defect [5, 19] and an unidentified band at 994 cm^{-1} are also shown.

The concentration of radiation defects in materials is an important parameter. In the case of radiation-induced defects, this concentration depends on the kind of irradiation and irradiation temperature as well as the energy and dose of bombarding particles. Generally, the higher the dose, the higher the concentration. A quantitative description of the production of radiation defects is possible in terms of production rates which is defined as the concentration of radiation defects at a given dose divided by this dose. Correspondingly, the rate of impurity loss in materials due to irradiation is defined as the change in impurity concentration at a given dose divided by this dose. The production rates η_{VO} and η_{CiOi} of VO and C_iO_i defects, together with the rates of oxygen loss $\eta_{\Delta Oi}$ and carbon loss $\eta_{\Delta Cs}$, are cited in table 1. The calibration coefficients for the VO (830 cm^{-1}) band was taken 6.25×10^{16} cm^{-2} [20] and for the C_iO_i (861 cm^{-1}) band 1.1×10^{17} cm^{-2} [20].

Figure 2 shows the rates of oxygen and carbon loss versus Ge content; hereafter we use short labels of Ge-doped Cz–Si samples in all figures leaving the symbol of Ge and the relevant numbers of samples from table 1. Figure 3 shows the production rates of VO and C_iO_i defects as a function of Ge content, correspondingly. In the samples with low initial carbon concentrations, in the range from 4×10^{16} cm^{-3} down to 2×10^{16} cm^{-3} or below, the 605 cm^{-1} band of C_s cannot be seen in IR spectra. It is below the detection limit after the irradiation as a result of a loss of C_s from substitutional sites (see table 1). In the following in order to describe properly the production of defects as a function of the Ge content we shall divide the samples in three groups. Group I includes the samples Cz–Si, Cz–Si:Ge-2 and Cz–Si:Ge-6 with initial carbon concentrations below 2×10^{16} cm^{-3}, labeled carbon-lean materials. Group II contains the samples Cz–Si:Ge-1, Cz–Si:Ge-3, Cz–Si:Ge-4 and Cz–Si:Ge-7 with

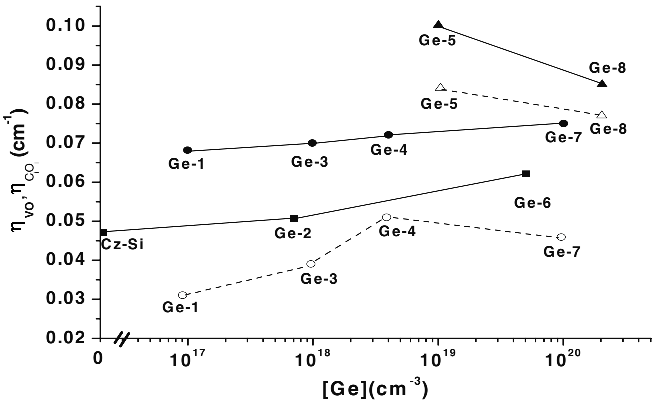

Figure 3. Production rate of VO (solid lines) and CiOi (dashed lines) defects versus [Ge].

carbon concentrations in the range from 2×10^{16} cm^{-3} to 1×10^{17} cm^{-3}, labeled carbon moderately doped samples. Group III includes the samples Cz–Si:Ge-5 and Cz–Si:Ge-8 with carbon concentrations around 2×10^{17} cm^{-3}, labeled carbon-rich materials.

Let us now consider the samples of group I. In the case of the irradiated reference sample Cz–Si without germanium, the rate of oxygen loss is about 0.052 cm^{-1} (see table 1). At the same time the production rate of VO defects was found to be about 0.047 cm^{-1}. It means that almost 90% of the total loss of oxygen is related to the formation of VO defects. The rest of the oxygen loss, about 10%, is probably due to the formation of other (vacancy-oxygen)-related complexes, e.g., V_2O defects. In two irradiated Ge-doped samples of group I, a small gradual increase in the VO production rate is observed against the increasing Ge concentration, at least up to concentrations of [Ge] = 5×10^{19} cm^{-3} (see figure 3). At such low concentrations of Ge, less than1 at.%, neither the threshold energy of elastic displacement of host atoms can be changed markedly nor the rate of Frenkel pairs separated into isolated vacancies and self-interstitials. Only the rate of indirect annihilation of vacancies and self-interstitials can be dependent on the concentration of Ge (see for instance [21]). These intrinsic defects are mobile at room temperature. In irradiated Cz–Si free vacancies are either trapped by interstitial oxygen atoms forming vacancy-oxygen pairs or they can annihilate when encounter self-interstitials. Ge atoms in the Cz–Si:Ge can successfully compete with oxygen for vacancy trapping in view of a favorable concentration ratio. These isoelectronic impurity atoms with a interatomic distance of several lattice spacings act as effective temporary trapping sites for vacancies, allowing them to survive from annihilation with self-interstitials. As a result, the production rate of oxygen-related radiation defects, mainly A-centers, is increased at the expense of the decreasing rate of indirect annihilation of intrinsic defects.

Let us now turn to the samples of group II which contain moderate concentrations of carbon. As is seen in figures 2 and 3, the rate of oxygen loss as well as the production rate of VO defects are considerably larger than those of the samples in group I. This is also true for the production of the C_iO_i defects (see figure 3). In these samples most of free self-interstitials are captured by carbon substitutional atoms converting them to carbon interstitials. Carbon interstitials being mobile at room temperature are trapped by carbon substitutional atoms and oxygen interstitials. Thus the more stable C_iC_s and C_iO_i complexes are formed. The concentration ratio between C_iO_i and C_iC_s defects produced is reported to be approximately proportional to the concentration ratio between O_i and C_s impurity atoms [22]. Thus, in these samples the C_iO_i complexes should be dominant carbon-related defects. As can be seen in table 1, for the carbon moderately doped samples the overall production rate of VO and C_iO_i defects is close to the rate of oxygen loss. The rate of oxygen loss measured for the samples of group II turned out to be larger than that in group I due to the trapping of self-interstitials by substitutional carbon atoms. In this way the rate of indirect annihilation of vacancies and self-interstitials becomes less important in the samples of group II, leading to a higher production rate of VO defects. Together with this, the production of C_iO_i defects is also getting much more pronounced. In other words, the annihilation of primary radiation defects is noticeably suppressed due to the carbon presence resulting in larger production rates of VO and C_iO_i defects, approximately by 20% for VO defects in relation to the samples of group I (see figure 3). Additionally, the increase in the production rate of VO and C_iO_i defects versus Ge content in carbon moderately doped materials could be explained in the model suggested above for the samples of group I. In this model Ge atoms act as temporary traps for vacancies preventing them from annihilation with self-interstitials. As a result, the production of VO complexes increases. At the same time the production of C_i defects and, therefore, C_iO_i complexes increases, too.

Let us now consider the group III samples which contain high carbon concentrations. At Ge concentrations of 10^{19} cm^{-3} to 2×10^{20} cm^{-3} the production rates of VO and the C_iO_i defects as well as the rate of oxygen and carbon loss are considerably increased (figures 2 and 3) in comparison to the samples of group II. The observed increases can easily be explained by the increasing concentration of carbon, just on the basis of the model discussed above for the samples of group II. As to the dependence of the production rates of VO and the C_iO_i defects on Ge content we observed that at [Ge] = 1×10^{19} cm^{-3} the production rates η_{VO} and η_{CiOi} are equal to 0.100 cm^{-1} and 0.084 cm^{-1}, respectively. However at higher concentrations of Ge, [Ge] = 2×10^{20} cm^{-3}, the η_{VO} and η_{CiOi} are smaller, 0.085 cm^{-1} and 0.077 cm^{-1}, correspondingly (see table 1 and figure 3). These results provide strong support to the conclusion that most of oxygen atoms pair with vacancies and most of carbon atoms react with self-interstitials. It should be noted that the rate of carbon loss against Ge concentration displays a similar trend, whereas the trend for the rate of oxygen loss turned out to be opposite (see figure 2). Interestingly, photoluminescence studies on the behavior of radiation-induced defects in Ge-doped Cz–Si with [Ge] = 2×10^{20} cm^{-3} and [Ge]$\leqslant 1 \times 10^{19}$ cm^{-3} revealed some significant differences in the spectra [23]. These differences were attributed in [23] to inhomogeneous distributions of Ge atoms

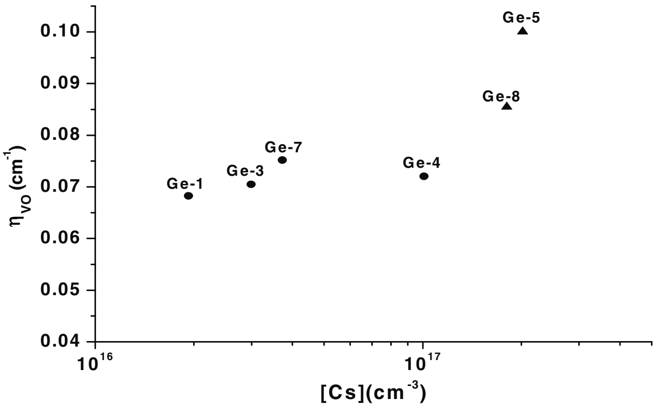

Figure 4. Production rate of VO defect versus [C$_s$].

in Cz–Si:Ge leading to the appearance of Ge clusters at high concentrations of Ge. Such clusters introduce local strain fields. The associated strain gradients assist in trapping of intrinsic defects by these clusters, thus promoting their enhanced annihilation at these sites. As a result, the average concentrations of vacancies and self-interstitials in a crystal are decreased. Therefore, one can expect that the production rates η_{VO} and η_{CiOi} should decrease, too. Obviously, this effect is reliably established in the present experiments (see figure 3). It might be well to point out that the idea of an increasing rate of indirect annihilation of vacancies and self-interstitials in Cz–Si with high Ge content was also suggested earlier for explanation of electrical data [24, 25].

At first sight, the rate of oxygen loss in Cz–Si with high Ge content is felt out of the explanation given above. However, such behavior may also be the characteristic of Ge clustering if one takes into account that the strain associated with such clusters can lead to the formation of other kinds of oxygen-related complexes, besides the VO defects, not seen in this study. Future experiments can help to gain an insight in this issue.

Figure 4 shows the production rate of VO defects versus [Cs] for the Cz–Si:Ge samples. It is clearly seen that the concentration of VO defects increases with the increasing carbon concentration. Interestingly, there are some conflicting reports on the role of carbon on the production rates of VO complexes telling about the lack of any effects, on the one hand [26], or about an enhanced production of VO defects, on the other [27]. The observed increase of VO production in our studies on Cz–Si:Ge could be attributed to the presence of both carbon and Ge impurities. Concerning the effect of Ge on the production of VO defects, we argue that the presence of Ge atoms in the Si lattice reduces the annihilation rate of vacancies and self-interstitials in the following way: a fraction of mobile vacancies is temporarily trapped by Ge atoms during irradiation, thus partially suppressing their uncorrelated annihilation with mobile self-interstitials ($V+Si_I \rightarrow Si_s$); the excess self-interstitials are absorbed by other unsaturated sinks in the lattice. As a result, the quasi-stationary concentration of vacancies during irradiation should increase. This, in turn, gives rise to the increasing production rate of VO centers. Concerning the effect of carbon, its ability to capture

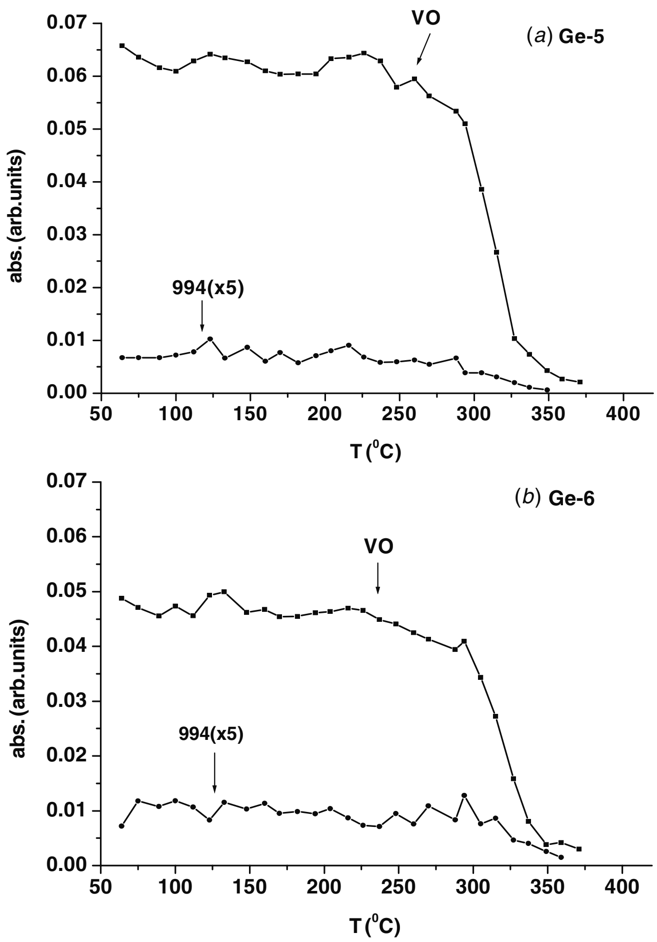

Figure 5. Thermal evolution of the VO and 994 cm^{-1} bands in the samples (*a*) Cz–Si:Ge-5 and (*b*) Cz–Si:Ge-6.

self-interstitials decreases their number, thus suppressing mutual annihilation with vacancies. Carbon also reduces the possibility of VO destruction ($VO+Si_I \rightarrow Oi$) in the course of irradiation. Both effects, that is the reduction of available self-interstitials due to the presence of C and the increase of available vacancies due to the presence of Ge, are expected to result in an increase of the production rate of VO defects.

Finally, we are presenting some preliminary results concerning the 994 cm^{-1} band detected only in IR spectra of electron-irradiated Ge-doped Cz–Si but not in Cz–Si. Figure 5 shows the evolution of this band in the course of isochronal annealing, together with the evolution of the 830 cm^{-1} band associated with VO defects for the samples Cz–Si:Ge-5 and Cz–Si:Ge-6. They clearly show a similar behavior. However, their behavior in dependence on the Ge content turned out to be different (see figures 3 and 6). Actually, for the samples of groups I and III, the behavior is similar inside each group, whereas in group II the trends are different. In general, the intensity of the 994 cm^{-1} band is decreased with increasing content of carbon (figure 7). In contrast the corresponding behavior of VO defects is just opposite (figure 4). It should be mentioned that in neutron irradiated Cz–Si, two weak bands at 914 cm^{-1} and 1000 cm^{-1} have been reported [28, 29]. They have similar annealing behavior with the VO defect and have been attributed to a

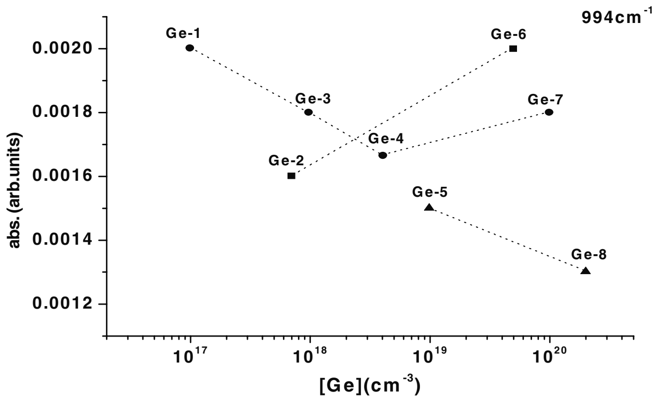

Figure 6. Amplitude of the 994 cm^{-1} band versus [Ge].

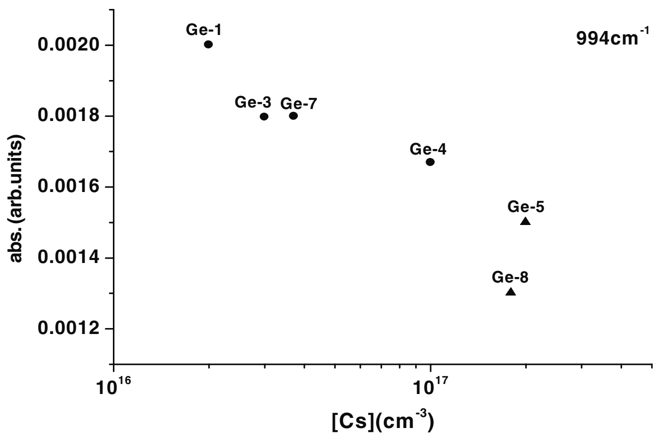

Figure 7. Amplitude of the 994 cm^{-1} band versus [C_s].

[VO+O_i] structure considered as precursor to the VO_2 defect. However these bands appear in the spectra upon isochronal anneals at 200 °C in contrast with the 994 cm^{-1} band which appears immediately after irradiation. At the present stage a definite assignment cannot be made. The defect may originate from a vacancy-oxygen complex perturbed by other nearby impurity atoms, e.g. C and Ge atoms in the vicinity of a VO complex. Because of this the annealing temperature of the VO defect and the VO perturbed by additional impurity atoms may practically be the same.

4. Conclusions

We studied the effects of Ge doping on the production rates of irradiation-induced VO and C_iO_i defects in Cz–Si:Ge crystals with various carbon concentrations. Our results showed that for samples with [C_s]$\leqslant 1 \times 10^{17}$ cm^{-3} the production rates of VO and C_iO_i defects increase slightly with the increasing Ge content, at least up to [Ge] $= 1 \times 10^{20}$ cm^{-3}. It is thought that the isolated Ge atoms act as temporary traps for vacancies, thus preventing them from indirect annihilation with self-interstitials. Because of this, the concentration of vacancies available for pairing with oxygen atoms is growing. The rate of oxygen loss displays a similar behavior. For samples with higher carbon concentrations about 2×10^{17} cm^{-3} the production rates of VO and C_iO_i defects turned out to be even larger. At [Ge] $= 2 \times 10^{20}$ cm^{-3} these rates become smaller as compared to the rate at [Ge] $= 1 \times 10^{19}$ cm^{-3}. A plausible explanation is based on the notion that at [Ge]$>1 \times 10^{20}$ cm^{-3} Ge atoms are distributed inhomogeneously, forming various clusters, along with isolated Ge atoms. Such clusters can serve as additional indirect annihilation sites of vacancies with self-interstitials due to their strain fields. Because of this, the production rates of VO and C_iO_i defects as well as the rate of carbon loss is getting smaller as compared to those observed in samples with [Ge] $= 1 \times 10^{19}$ cm^{-3}. However, the rate of oxygen loss displays an opposite behavior. A weak band at 994 cm^{-1} observed only in IR spectra of electron-irradiated Ge-doped Cz–Si shows the same annealing stage as that of VO defects. However, its production rate against Ge and C_s concentrations is different in behavior. The origin of this band is still uncertain.

References

[1] Bender H and Vanhellemont J 1994 *Oxygen in Silicon in Handbook of Semiconductors, Materials, Properties and Preparations* ed T S Moss and S Mahajan (Amsterdam: North Holland) p 1637
[2] Davies G and Newman R C 1994 *Carbon in monocrystalline Si in Handbook of Semiconductors, Materials, Properties and Preparations* ed T S Moss and S Mahajan (Amsterdam: North Holland) p 1557
[3] Londos C A, Fytros L G and Georgiou G J 1999 IR studies of oxygen–vacancy related defects in irradiated Si *Defects Diffus. Forum* **171–172** 1–31
[4] Corbett J., Watkins G D and McDoland R 1964 New oxygen infrared bands in annealed irradiated Si *Phys. Rev.* A **135** 381–5
[5] Londos C A, Potsidi M S, Antonaras G D and Andrianakis A 2006 Isochronal annealing studies of carbon related defects in irradiated Si *Physica* B **376–377** 165–8
[6] Londos C A 1988 Aspects of the defect reactions related to carbon impurity in Si *Japan J. Appl. Phys.* **27** 2089–93
[7] Londos C A 1987 Annealing studies of defects pertinent to radiation damage in Si *Phys. Status Solidi b* **102** 639–44
[8] Londos C A 1987 Deep-level transient spectroscopy studies of the interstitial carbon defect in silicon *Phys. Rev.* B **35** 6295–97
[9] Londos C A 1990 Carbon-related radiation damage centres and processes in p-type Si *Semicond. Sci. Technol.* **5** 645–48
[10] Davies G 1989 The optical properties of luminescence centres in silicon *Phys. Rep.* **176** 83–188
[11] Chappel S P and Newman R C 1987 The selective trapping of self-interstitials by interstitial carbon impurities in electron irradiated silicon *Semicond. Sci. Technol.* **2** 691–94
[12] Tsang J C, Mooney P M, Dacol F and Chu J O 1994 Measurements of alloy composition and strain in thin Ge_xSi_{1-x} layers *J. Appl. Phys.* **75** 8098–108
[13] Johnson F A 1959 Lattice absorption bands in silicon *Proc. Phys. Soc. London* **73** 265–72
[14] Vook F L and Stein H J 1968 Infrared absorption bands in carbon- and oxygen-doped silicon *Appl. Phys. Lett.* **13** 343–46
[15] 1986 *ASTM Book of Standards* F123–86 p 252
[16] Regolini J L, Stroquert J P, Ganter C and Siffert P 1986 Determination of the conversion factor for infrared measurements of carbon in silicon *J. Electrochem. Soc.* **133** 2165–68
[17] Baghdadi A, Bullis W M, Croarkin M C, Yue–zhen L, Scace R I, Series R W, Stallhoffer P and Watanabe M J 1989 Interlaboratory determination of the calibration factor for

the measurement of the interstitial oxygen content of silicon by infrared absorption *J. Electrochem. Soc.* **136** 2015–24
[18] Londos C A, Potsidi M S and Stakakis E 2003 Carbon-related complexes in neutron-irradiated silicon *Physica* B **340–342** 551–55
[19] Lavrov E V, Hoffmann L and Nielsen B Bech 1999 Local vibrational modes of the metastable dicarbon center (C_s–C_i) in silicon *Phys. Rev.* B **60** 8081–86
[20] Davies G, Lightowlers E C, Newman R C and Oates A S 1987 A model for radiation damage effects in carbon-doped crystalline silicon *Semicond. Sci. Technol.* **2** 524–32
[21] Emtsev V V, Mashovets T V and Mikhnovich V V 1992 Frenkel pairs in germanium and silicon (review) *Sov. Phys. Semicond.* **26** 12–25
[22] Lindstrom G *et al* 2001 Radiation hard silicon detectors—developments by the RD48 (ROSE) collaboration *Nucl. Instrum. Methods Phys. Res.* A **466** 308–26
[23] Sobolev N A and Nazave M H 1999 Defects incorporating Ge atoms in irradiated Si:Ge *Physica* B **273–274** 271–74
[24] Khirunenko L I, Shekhovtsov V I, Shinkarenko V K, Shpinar L I and Yaskovets I I 1987 Characteristics of radiation defect formation processes in Si:Ge crystals *Sov. Phys.—Semicond.* **21** 345–47
[25] Golubev V G, Emtsev V V, Klinger P M, Kropotov G I and Shmartsev Yu V 1992 Processes of formation of radiation defects in Si:Ge at 4.2, 78 and 300 K *Sov. Phys.—Semicond.* **26** 328–29
[26] Bean A R, Newman R C and Smith R S 1970 Electron irradiation damage in silicon containing carbon and oxygen *J. Phys. Chem. Solids* **31** 739–51
[27] Akhmetov V D and Bolotov V V 1980 The effect of carbon and boron on the accumulation of vacancy-oxygen complexes in silicon *Radiat. Eff.* **52** 149–52
[28] Londos C A, Georgiou G T, Fytros L G and Papastergiou K 1994 Interpretation of infrared data in neutron-irradiated silicon *Phys. Rev.* B **50** 11531–34
[29] Londos C A, Sarlis N, Fytros L G and Papastergiou K 1996 Precursor defect to the vacancy-dioxygen center in Si *Phys. Rev.* B **53** 6900–03