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Positron annihilation study of silicon irradiated by different neutron doses

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Abstract. Single crystals of silicon grown in an argon atmosphere were irradiated with three different neutron doses: 6×10^{16}, 3.6×10^{17} and 1.2×10^{18} cm^{-2}. Positron annihilation spectroscopy indicates that there are two distinct annealing stages of V-type defects in the three neutron-irradiated silicon samples: one at about 200 °C and another at about 500 °C. The former is due to the annealing out of P–V and $(V_2\text{–O})^-$ complexes; the latter is due to the annealing out of $(V_2\text{–O})^0$ complexes. For samples irradiated with two different higher neutron doses, the intensity due to the radiation-induced monovacancy-type defects is above 58% and disappears at 550 °C; the intensity due to the radiation-induced divacancies is above 23%, increases above 200 °C and disappears at 600–650 °C. For the sample irradiated with a lower neutron dose the intensity due to the divacancies is only 7.3%, it disappears below 200 °C, and no secondary divacancies appear. The intensity due to the secondary divacancies is highly dependent on the neutron dose. There are two annealing stages of V_2-type defects at 150–200 °C and about 600 °C. The former is due to the approach of interstitial silicon atoms to divacancies localized in the cores of neutron-radiation-disordered regions, and the latter is due to annealing out of divacancies in the 'undisturbed' matrix of the silicon crystal and V_3–O complexes. For samples irradiated with two different higher neutron doses, monovacancy-type defects with higher intensities appear again at 600–650 °C, which indicates the nature of the '600–650 °C acceptor'. Above 700 °C, many dislocations with 231 ps $\leqslant \tau \leqslant$ 239 ps (sometimes both dislocations and monovacancy-type defects (240 ps $\leqslant \tau \leqslant$ 266 ps)), are formed, especially for samples irradiated with higher neutron doses.

1. Introduction

In neutron-irradiated silicon, many radiation defects and damage such as displaced silicon atoms and vacancies are produced since neutrons collide with silicon atoms. Although a considerable number of papers about this subject have been published, many questions remain unclear. Besides general interest in the fundamental aspects of neutron radiation defects and damage in semiconductors, there is also technological interest in a better understanding of defect production and annealing. Neutron-transmutation-doped silicon (NTD Si) has been used for fabricating power devices owing to its excellent dopant homogeneity and doping accuracy. However, radiation defects and damage may influence the electrical properties of silicon materials and devices greatly. NTD Si must be annealed. Rather empirical annealing procedures have been used so far. Therefore, a detailed study of the annealing behaviour of defects in neutron-irradiated silicon is still an important subject.

Positron annihilation spectroscopy (PAS) is sensitive to open-volume lattice defects in materials and therefore can be used to study neutron radiation defects such as vacancies, vacancy–impurity clusters and dislocations in silicon [1–3]. In particular, PAS is very useful for studying defects in NTD Si and is a non-destructive technique.

In spite of these researches the basic properties of defects in neutron-irradiated silicon have not been sufficiently clarified yet. For example, from the literature [2], divacancies are the dominant neutron-induced defect detected by PAS. This is very different from our results. Because radiation defects can combine with impurities in silicon and the Fermi level of irradiated silicon changes with the annealing temperature, the evolution and identification of defects become very complicated.

In this paper we present the results of Doppler-broadening and positron lifetime measurements of defects in floating-zone silicon grown in an argon atmosphere (FZ (Ar) Si) and NTD FZ (Ar) Si irradiated with three different neutron doses and annealed from room temperature (RT) to 1150 °C, discuss the annealing behaviour of defects and the influence of neutron doses on it and suggest an explanation for the annealing stages of V- and V_2-type defects.

2. Experimental procedure

The starting crystal silicon ingot used in this investigation was n-type FZ (Ar) Si with a high resistivity of 1500 Ω cm and an interstitial oxygen concentration of 1.5×10^{17} cm^{-3} determined by FTIR spectroscopy. The samples cut were 1.0 mm thick and the surfaces were polished for measurements. NTD Si samples 1, 2 and 3 were irradiated at RT in the light-water reactor of the Institute of Nuclear Energy Technology, Tsinghua University, with total neutron doses of 6×10^{16} cm^{-2}, 3.6×10^{17} cm^{-2} and 1.2×10^{18} cm^{-2}, respectively. The ratio of thermal to fast-neutron flux is 3.5. The 20 min isochronal anneal from RT to 1150 °C was carried out in an argon atmosphere.

The positron annihilation measurements were performed at RT at the Institute of Nuclear Physics, Technical University, Graz. The conventional lifetime spectrometer used has a time resolution of about 200 ps full width at half-maximum. The positron source itself was used in a sandwich configuration and was prepared by deposition of about 10 μCi of ^{22}NaCl on a 0.81 mg cm^{-2} aluminium foil. Each lifetime spectrum accumulated about 1×10^7 counts. After source correction the spectra were analysed and decomposed into two or three components.

The Doppler broadening of the annihilation line was measured using an intrinsic Ge detector with an energy resolution better than 1.18 keV at the 497 keV line (close to the 511 keV annihilation γ energy). About 2×10^6 counts were recorded for each spectrum and were repeated at least ten times. The usual lineshape parameter S, defined as the area of the central region of the peak divided by the total area of the spectrum, was used to characterize the spectra and calculated with a simple program.

3. Experimental results and discussion

3.1. Lifetime components

The positron lifetime can determine the effective electron density in silicon and depends on the concentration and distribution of the electrons in the defect in which the annihilation event occurs. It is possible to characterize multiple defect types simultaneously, since each positron lifetime is indicative of a specific defect type. The intensity of lifetime is related to the concentration and positron capture cross section of the defect.

Figure 1 is the two-component free-fitting curves for NTD Si sample 3. τ_1 is the short lifetime, which is the overall result of the lifetime of positron annihilation in a perfect crystal and the lifetime of positron capture states in vacancy-type defects; it is related to the concentration of defects. τ_2 is the long lifetime or 'defect' lifetime component, which is the lifetime of the positron captured by, and annihilated at vacancy-type defects in samples; it is related to the size and type of defects. I_1 and I_2 denote the relative intensities of the short lifetime and 'defect' lifetime, respectively.

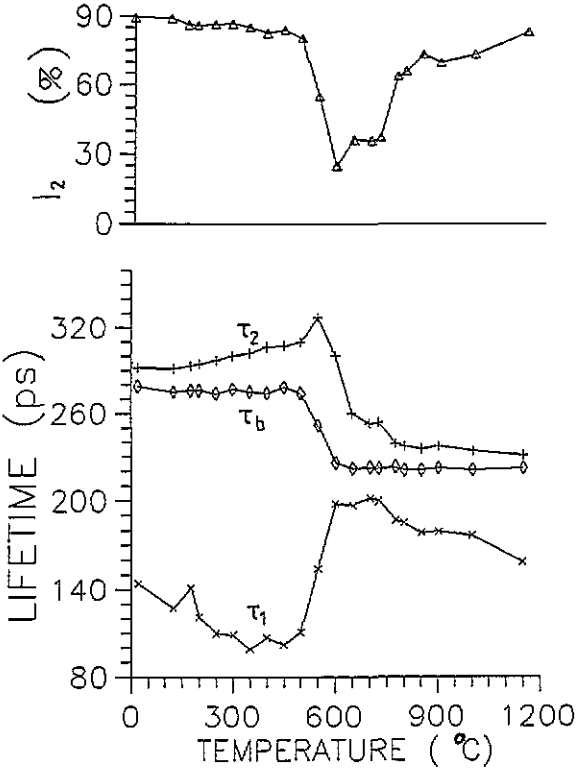

Figure 1. The bulk lifetime and decomposition of the lifetime spectra, obtained from unconstrained two-component analyses for NTD Si sample 3, as functions of the annealing temperature.

According to a simple two-state trapping model [4], the bulk lifetime τ_b can be calculated from

$$1/\tau_b = I_1/\tau_1 + I_2/\tau_2. \tag{1}$$

The applicability of the trapping model can be verified if the calculated τ_b value is near 218–220 ps [5].

Table 1 shows the τ_2 values from free two-component fitting for the three NTD Si samples.

The main radiation defects in neutron-irradiated silicon are monovacancy (V)-type complexes such as oxygen–vacancy (O–V) and phosphorus–vacancy (P–V), and divacancies

Table 1. The 'defect' lifetime τ_2, obtained from unconstrained two-component analyses for the three NTD Si samples, as functions of annealing temperature.

Annealing temperature (°C)	τ_2 (ps)		
	NTD Si sample 1	NTD Si sample 2	NTD Si sample 3
20	285 ± 3	288 ± 1	292 ± 1
125	276 ± 5	290 ± 1	291 ± 1
150	278 ± 6	291 ± 1	293 ± 1
200	270 ± 6	294 ± 1	294 ± 1
250	261 ± 6	299 ± 1	297 ± 1
300	267 ± 7	300 ± 1	300 ± 1
350	263 ± 8	301 ± 1	302 ± 1
400	266 ± 8	305 ± 1	306 ± 1
450	263 ± 8	307 ± 2	307 ± 1
500	245 ± 5	309 ± 2	309 ± 1
550	268 ± 13	326 ± 5	327 ± 2
600	246 ± 9	274 ± 14	300 ± 9
650	255 ± 11	298 ± 16	260 ± 10
700	236 ± 4	238 ± 5	253 ± 15
725			254 ± 12
750	236 ± 4	235 ± 4	
775			239 ± 6
800	280 ± 21	245 ± 7	237 ± 6
850	254 ± 11	236 ± 6	235 ± 4
900	257 ± 7	234 ± 4	237 ± 5
1000	243 ± 9	234 ± 4	234 ± 4
1150	294 ± 12	282 ± 10	231 ± 2

(V_2) [6]. The positron lifetimes for V- and V_2-type complexes are 250–270 ps and 300–325 ps, respectively [1–3, 7, 8]. The phosphorus or oxygen atoms do not occupy any part of the vacancy. The observed 'defect' lifetime components τ_2 from free fitting for as-irradiated NTD Si samples 1, 2 and 3 are 285 ± 3 ps, 288 ± 1 ps and 292 ± 1 ps. This shows that τ_2 should be a mixture of the lifetimes related to the V- and V_2-type defects. On increase in the annealing temperature to 550 °C the lifetimes of NTD Si samples 2 and 3 increase to 326 ± 5 ps and 327 ± 2 ps because V-type defects anneal out and the relative concentration of V_2-type defects increases gradually.

As shown in figure 1, for NTD Si sample 3 which had been annealed below 500 °C, the intensity I_2 of the 'defect' lifetime due to the V- and V_2-type defects remains nearly unchanged (about 90%), which shows that most positrons are trapped in vacancy-type defects. Between 250 and 500 °C, τ_1 decreases to about 100 ps. Saturation trapping of positrons in defects due to the strong radiation damage could be expected; therefore, analyses give only an outline by the simple trapping model [4]. Above 550 °C, τ_2 and I_2 decrease; the calculated τ_b also drops to nearly the bulk lifetime value of unirradiated FZ Si; τ_1 increases to the maximum value, which indicates that most defects have annealed out. Dislocations could form in large-dose fast-neutron radiation silicon annealed at a high temperature [3], e.g. at 700 °C for 10 min [9]. Above 775 °C for NTD Si sample 3 and above 700 °C for NTD Si sample 2, most τ_2 values are in the range 231–239 ps (see table 1); the corresponding intensity I_2 is also rather large. We propose that 231–239 ps may be the positron lifetime due to dislocation. Snead and Lynn [3] considered that the lifetime for dislocations was 260 ps, a little smaller than that of V-type defects 270 ps. Between 700 and 725 °C for NTD Si sample 3 the τ_2 value is 253–254 ps, which may be due to a mixture of the lifetimes of dislocations and of V-type defects. The formation of dislocations induces a decrease in τ_1, but it seems that they have no obvious influence on the bulk lifetime.

We used 270 ps as the lifetime due to V-type defects for analyses of the lifetime spectra of FZ Si. The intensity due to V-type defects for as-grown FZ Si is 14.8%, it changes very little during annealing up to 1150 °C.

According to equation (1), the calculated bulk lifetime of FZ (Ar) Si annealed from RT to 1150 °C is between 218.9 and 222.2 ps; the average bulk lifetime is 220.5 ps. This agrees very well with the previous results [5]. Our result shows that V-type defects exist in FZ Si.

We used 270 ps and 320 ps as the lifetimes due to V- and V_2-type defects for analyses of the lifetime spectra of NTD Si. Their intensities I_2 and I_3 are shown in figure 2.

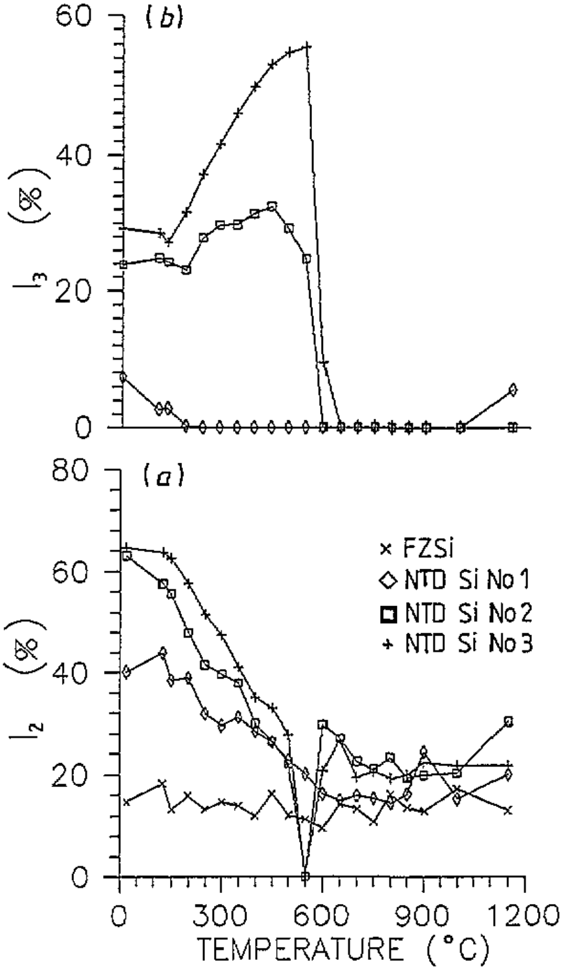

Figure 2. The intensities (*a*) I_2 and (*b*) I_3 of the 270 and 320 ps lifetime components for FZ Si (×) and NTD Si sample 1 (◇), sample 2 (□) and sample 3 (+) as functions of the annealing temperature.

The two-state trapping model has been expanded to accommodate the situation involving two kinds of defects [10]. The bulk lifetime τ_b, the concentrations C_2 and C_3, and the trapping rates κ_2 and κ_3 for V- and V_2-type defects, respectively, can be calculated:

$$\kappa_2 = [1/\tau_b - (1 - I_3)/\tau_2 - I_3/\tau_3](I_2/I_1) \tag{2}$$

$$\kappa_3 = [1/\tau_b - (1 - I_2)/\tau_3 - I_2/\tau_2](I_3/I_1) \tag{3}$$

$$\tau_b = (I_1/\tau_1 + I_2/\tau_2 + I_3/\tau_3)^{-1} \tag{4}$$

$$\kappa_i = \mu_i C_i \qquad (i = 2, 3) \tag{5}$$

where μ_2 and μ_3 are the specific trapping rates for V- and V_2-type defects, respectively.

In as-irradiated NTD Si samples 1, 2 and 3 the intensities I_2 due to V-type defects are $40.1 \pm 3.4\%$, $63.1 \pm 0.9\%$ and $64.6 \pm 0.9\%$, and the intensities I_3 due to V_2-type defects are $7.3 \pm 1.5\%$, $23.8 \pm 0.9\%$ and $29.1 \pm 0.9\%$, respectively.

As shown in figure 2(*a*), the intensity I_2 due to V-type defects decreases during annealing from 125 to 300 °C for NTD Si samples 1 and 2, and from 125 to 400 °C for NTD Si sample 3. There is a 250–350 °C plateau on the annealing curve of I_2 for NTD Si sample 2. For NTD Si samples 2 and 3 the intensity I_2 continues to decrease above 400 °C; it decreases sharply above 500 °C and disappears at 550 °C. For NTD Si sample 1 it increases slightly at 350 °C and then decreases gradually. Two obvious annealing stages can be observed: one at about 200 °C and the other at about 500 °C.

The concentrations of phosphorus atoms due to the NTD process, i.e. the concentrations of P–V complexes due to the β-decay process of ^{31}Si atoms are about 1×10^{13} cm^{-3}, 6×10^{13} cm^{-3} and 2×10^{14} cm^{-3} in as-irradiated NTD Si samples 1, 2 and 3, respectively. The negatively charged P–V complexes are expected to be efficient positron traps. The annealing-out temperature of P–V complexes is 200 °C. Our measurements of deep-level transient spectroscopy show that the concentration of O–V complexes (A centres) is much higher than that of P–V complexes [11]. Girka *et al* [12] pointed out that a τ_2 of 250–270 ps is found in all proton-irradiated FZ Si and CZ Si with different oxygen concentrations but the intensity for irradiated CZ Si has the largest value; therefore, positrons are expected to be trapped intensively in O–V complexes and to be annihilated in them with a lifetime of 250–270 ps because of the comparatively high oxygen concentration in CZ Si. The concentration of V-type defects in NTD Si calculated from equation (5) is above 2×10^{16} cm^{-3}, much higher than that of P–V complexes mentioned above. The concentration of oxygen atoms is also much higher than that of phosphorus atoms. Therefore, O-related V-type complexes should be the dominant defects in NTD Si. However, Mascher *et al* [10] proposed that the A centre could be a shallow trap and it was not possible to measure A centres by PAS at RT.

In NTD Si, divacancy–oxygen complexes (V_2–O) can also be produced during irradiation with an approximately 50% concentration of A centres [13], and during annealing since some monovacancies, which are released from the annealing out of P–V and O–V complexes as well as neutron radiation-induced vacancy clusters, can move towards and combine with O–V complexes or since radiation-induced divacancies can combine with oxygen atoms. It begins to anneal from 120 °C, and disappears at 350 °C [13]. The positron lifetime due to the V_2–O complex may be about 270 ps [10] since it is essentially a vacancy plus a substitutional oxygen atom. Mäkinen *et al* [14] suggested that the annealing stage between 127 and 327 °C in proton-implanted silicon is due to the annealing out of negatively charged V_2–O complexes $(V_2\text{–O})^-$ [14]. Therefore, we propose that the first annealing stage of V-type defects at about 200 °C in this work is mainly due to the annealing out of P–V [15] and $(V_2\text{–O})^-$ complexes.

Our Hall coefficient measurements show that the free-hole concentration in NTD Si increases with increasing annealing temperature up to 550 °C, i.e. the Fermi-level position moves towards the valence band, and V_2–O complexes become neutral. Lee *et al* [16] have reported that neutral V_2–O complexes start to anneal out at about 350 °C. An abnormal annealing peak for NTD Si sample 1 at 300–400 °C and a shoulder for NTD Si sample 2 at

250–350 °C as well as the second annealing stage at about 500 °C in the annealing curves of V-type defects in NTD Si samples 2 and 3 may be attributed to the formation and annealing out of the $(V_2\text{–}O)^0$ complexes. However, its annealing-out temperature is higher than that for electron-irradiated silicon [13].

Above 600 °C, some V-type defects appear again. The intensities I_2 for NTD Si samples 2 and 3 are obviously higher than that for FZ Si, especially at 600–650 °C and 1150 °C for NTD Si sample 2 and at 650 °C for NTD Si sample 3. The formation of the '600–650 °C acceptor' is an important characteristic during the annealing of neutron-irradiated silicon [17]. At 650 °C the higher concentration of V-type defects in NTD Si samples 2 and 3 may be due to the so-called '650 °C acceptor'. I_2 for NTD Si sample 1 is also large at 900 °C. Some high-temperature thermal defects such as dislocations and V-type defects may be produced above 700 °C.

As shown in figure 2(*b*), the V_2 defects in NTD Si sample 1 anneal out at 200 °C. No secondary V_2-type defects appear except for those with an I_3 of 5.6% at 1150 °C. This is because the concentration of radiation-induced V-type defects and monovacancies released from P–V and O–V complexes is too small to agglomerate into divacancies in NTD Si irradiated at lower neutron doses.

The intensity I_3 due to V_2-type defects undergoes a small decrease during annealing from RT to 200 °C for NTD Si sample 2 and from RT to 150 °C for NTD Si sample 3. Above 200 °C it increases obviously with increase in the annealing temperature, reaching the maximum value of 32.4% at 450 °C for NTD Si sample 2 and 55.6% at 550 °C for NTD Si sample 3. Then I_3 drops markedly, disappearing at 600 °C and at 650 °C, respectively. Two annealing stages can also be observed: one at 150–200 °C which is in a good agreement with that of NTD Si sample 1 and the other at about 600 °C.

Some interstitial silicon atoms can annihilate in the annealing temperature range between 100 and 150 °C [18]. The annealing stage of V_2-type defects at 100–200 °C in neutron-irradiated silicon is due to the approach of interstitial defects to divacancies localized in the cores of the neutron-radiation-disordered regions [19]. This could be the reason for the first annealing stage of V_2-type defects at 150–200 °C in the three different NTD Si samples.

Some secondary V_2 and/or V_3–O complexes may be formed owing to agglomeration of V-type defects. The V_3–O complex would have the V_2 character [20]. In addition, the divacancy is thought to start to migrate at about 170 °C [3]. Larger vacancy clusters, e.g. quadrivacancies (V_4), may also be formed owing to the migration of V_2 defects, but the maximum intensity is very small, less than 3.2%. Therefore, the formation of V_2 and/or V_3–O complexes should be responsible for the increase in I_3.

For NTD Si sample 2 annealed from 150 to 450 °C, the intensity I_2 decreases 29.1% and I_3 increases 8.2%; the corresponding changes in the positron-trapping rates of V- and V_2-type defects are 2.1×10^9 and $3.4 \times 10^8\ \mathrm{s}^{-1}$, calculated from equations (2) and (3). In electron-irradiated silicon the specific trapping rates of V_2 and $(P\text{–}V)^{-1}$ complexes are about $2 \times 10^{15}\ \mathrm{s}^{-1}$ [10, 21]. We can estimate roughly the defect reaction using these data. The ratio of the change in concentration of V-type complexes to that of V_2-type complexes is about 6. This shows that, during annealing, only some monovacancy-type defects (about 30%) take part in agglomeration to form V_2-type defects in NTD Si irradiated by 3.6×10^{17} neutrons cm^{-2}. The others may anneal out owing to direct annihilation with interstitial silicon atoms [18].

The annealing temperature of V_3–O is about 450 °C [18]. The annealing temperature of divacancies in the 'undisturbed' matrix of the silicon crystal is about 300–500 °C [19]. Here the second annealing stages of V_2-type defects at about 600 °C can be explained by the annealing out of V_2 and V_3–O complexes.

With an increase in radiation dose, the production and annealing behaviour of radiation defects have some obvious characteristics; the concentration of defects produced in as-irradiated NTD Si obviously increases and for example, the intensities I_2 (63.1% and 64.6%) and I_3 (23.8% and 29.1%), the average lifetime (261 ps and 276 ps) and S (0.5195 and 0.5233) for NTD Si samples 2 and 3 irradiated by higher neutron doses are much higher than those for NTD Si sample 1 (40.1%, 7.3%, 227 ps and 0.5096, respectively).

Compared with NTD Si sample 1 the intensities I_2 and I_3 for NTD Si samples 2 and 3 are higher during annealing below 600 °C. The V-type defects disappear at 550 °C, but the concentration and annealing-out temperatures of V_2-type defects are much higher. It seems that radiation doses have a larger influence on the formation and annealing behaviour of secondary divacancy-type defects.

3.2. The parameter S and average lifetime

The parameter S is a measure of the annihilation probability of positrons and valence electrons.

Figure 3 shows the change in S and average lifetimes of the FZ Si samples and NTD Si samples 1, 2 and 3 with annealing temperature. Here, the experimental error for S is within ±0.0003.

According to a simple two-state trapping model [4], the average positron lifetime $\bar{\tau}$, or the statistical mean of the positron-lifetime distribution, is defined as

$$\bar{\tau} = (\tau_1 I_1 + \tau_2 I_2)/(I_1 + I_2). \tag{6}$$

The average lifetime and S indicate the average effect of the formation and recovery of defects. Therefore, their changes with annealing temperature are very similar (see figure 3).

For as-grown FZ (Ar) Si, S is 0.5072; the average lifetime is 219 ps.

Since trapping in open-volume defects decreases the width of the Doppler-broadening energy spectrum, S for defects is generally greater than that for bulk silicon. An increase in S represents an increase in the concentration of defects.

For as-irradiated NTD Si samples 1, 2 and 3, S increases to 0.5096, 0.5195 and 0.5233, respectively; the average lifetimes are also very high, reaching 229 ps, 261 ps and 276 ps, respectively; the calculated bulk lifetimes are 216 ps, 227 ps and 260 ps, respectively, owing to neutron radiation defects. The estimated concentrations of displaced silicon atoms are about 1.5×10^{17} cm^{-3}, 9×10^{17} cm^{-3} and 3×10^{18} cm^{-3}, respectively [22].

For NTD Si samples 2 and 3, S and the average lifetime are still very high during annealing at temperatures lower than 550 °C. The $\bar{\tau}$ values are 236 ps and 271 ps, respectively at 500 °C. The calculated τ_b is between 260 and 227 ps for NTD Si sample 3 annealed below 500 °C, much longer than τ_b for FZ Si. Therefore, the simple trapping model is not applicable to this situation, and the analyses give only an outline of neutron-induced defects in NTD Si sample 3 annealed below 500 °C. However, for NTD Si sample 2 its τ_b value equals the bulk lifetime of FZ Si (except for the unannealed sample). We can say that for samples irradiated with two different higher neutron doses, the intensities due to the radiation-induced V- and V_2-type defects are above 58% and 23%, respectively. Above 600 °C, S and $\bar{\tau}$ decrease greatly, down to the stable value (0.5066–0.5072 and 219–221 ps, respectively), nearly the same as the corresponding values for FZ Si. This shows that most radiation defects and secondary defects can be removed at about 600–650 °C.

For NTD Si sample 1 irradiated with a lower neutron dose, $\bar{\tau}$ and S are nearly equal to those for unirradiated FZ Si except that $\bar{\tau}$ is always a little larger, but the S values are a little larger at some annealing temperatures and a little smaller at other annealing

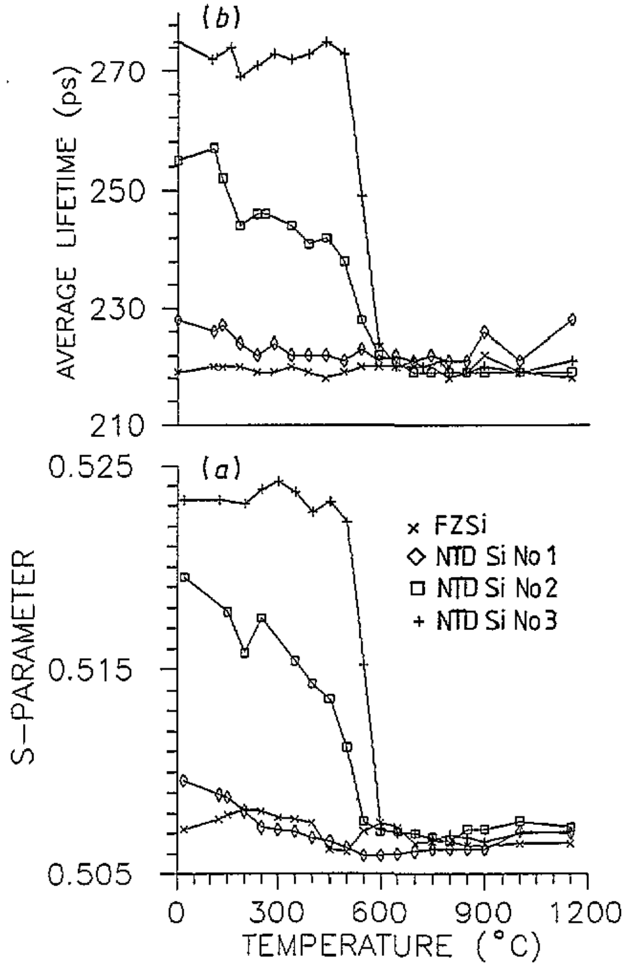

Figure 3. (*a*) The *S*-parameter and (*b*) the average lifetime of FZ Si (×) and NTD sample 1 (◇), sample 2 (□) and sample 3 (+) as functions of the annealing temperature.

temperatures although the concentration of V-type defects is also high below 500 °C. *S* decreases obviously from RT to 250 °C and then reaches the lower stable value. $\bar{\tau}$ decreases gradually with increase in annealing temperature but is always a little higher than that for FZ Si; that is especially obvious at 900 and 1150 °C and indicates that only a sufficiently high concentration of V-type defects can influence these parameters, and that divacancies have an even greater influence.

The annealing behaviours of V- and V_2-type defects agree well with those of *S* and the average lifetime. For example, *S* and $\bar{\tau}$ for NTD Si samples 2 and 3 are very high below 550 °C owing to the high concentration of defects. During annealing above 600 °C, *S* remains at the lower stable value but the higher intensities due to V-type defects in NTD Si sample 2 induce values of *S* a little larger than those of unirradiated silicon. One very interesting feature is that there are obvious increases in average lifetime at 900 and 1150 °C for NTD Si sample 1. As shown in figure 2, for NTD Si sample 1, I_2 is rather large and its increases are 8% and 5% at these two temperatures, respectively; V_2-type defects with an intensity of 5.6% appear at 1150 °C; a long lifetime of 1.9 ns with an intensity of about 0.3% also appears, perhaps owing to the production of microvoids at these two temperatures; another reason may be the influence of the source substrate on S, but the average lifetime is obtained through source correction and therefore is more sensitive to defects.

4. Conclusion

There are always V-type defects with an intensity of 10–18% in FZ Si during annealing from RT to 1150 °C. In NTD Si irradiated with 6×10^{16} neutrons cm^{-2} the main defect is the V type; V_2 defects have an intensity of only 7.3% and anneal out at a temperature lower than 200 °C. In NTD Si irradiated with 3.6×10^{17} neutrons cm^{-2} and 1.2×10^{18} neutrons cm^{-2}, the intensities due to V- and V_2-type defects are above 58% and 23%, respectively. The former anneals out at 550 °C; the latter increases above 200 °C, decreases greatly above 550 °C and anneals out at 600 and 650 °C. It seems that radiation doses have a marked influence on the formation and annealing behaviour of secondary V_2-type defects.

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