THE MECHANISM OF THE ENHANCEMENT OF DIVACANCY PRODUCTION BY OXYGEN DURING \( \text{(U)} \) STATE UNIV OF NEW YORK AT ALBANY DEPT OF PHYSICS G S OEHLEIN ET AL.
Numerical tests of possible models for the oxygen dependence of the divacancy introduction rate in silicon electron irradiated at room temperature were performed on a computer. Only the model in which oxygen traps Si self-interstitials can reproduce all the experimental data. Our modeling results (in conjunction with the experimental data) imply that during room-temperature electron irradiation of Si the indirect production of divacancies can be more important than the direct production of $V_2$ via the single-collision process.
The mechanism of the enhancement of divacancy production by oxygen during electron irradiation of silicon. II. Computer modeling

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I. INTRODUCTION

In a previous paper (Part I) we reported experimental results about the influence of the interstitial oxygen content of a Czochralski (Cz) silicon wafer on the divacancy production during room-temperature electron irradiation.\(^1\) We found that the divacancy \( [V_2] \) introduction rate varied linearly with interstitial oxygen concentration. Similarly the introduction rate of vacancy-oxygen centers (VO) displayed a linear dependence on interstitial oxygen concentration. The most remarkable result of our experiment was that the divacancy concentration correlated much better to the density of introduced vacancy-oxygen centers than to the initially present dissolved oxygen concentration and that the relationship between \( V_2 \) and VO was perfectly linear.

The VO density of a Si sample after electron irradiation is an indicator of how many single vacancies were available for capture by the oxygen atoms, i.e., the VO concentration is proportional to the steady-state concentration of vacancies and the length of the irradiation (ignoring saturation and other effects). The linear relationship between \( V_2 \) and VO indicates a possible dependence of the \( V_2 \) concentration on the steady-state vacancy concentration, i.e., the importance of the indirect production of divacancies.

Based on our results we proposed different possible models for the increase in divacancy introduction rate with increasing oxygen content. It is the purpose of the present paper to test these models numerically. Of the possible divacancy enhancement mechanisms mentioned in Part I only one can reproduce all our experimental results.

II. DEFECT PRODUCTION MODELING

A. Basic model

In our computer model of defect production\(^2\) we assume that single vacancies \( V \) and Si self-interstitials \( I \) are created at a constant rate \( G_0 \) during room-temperature electron bombardment of silicon. Simultaneously the irradiation introduces immobile divacancies \( V_2 \) and a corresponding number of interstitials at a constant rate \( G_v/o \). The mobile interstitials and vacancies diffuse and can annihilate each other or they can be captured by trapping centers. In addition two vacancies are allowed to combine and form a divacancy. The capture rate between any two defects is assumed to be of the form \( 4\pi (D_1 + D_2)|R_{12}| \), where \( D_1 \) and \( D_2 \) are the diffusion coefficients of the defects and \( R_{12} \) is an effective capture radius.\(^3\) For the boron-doped Cz silicon crystal \( (p \approx 7 \Omega cm) \) which we used in our experiment we restrict ourselves to these impurities: Varying interstitial oxygen concentrations, a substitutional carbon density of \( 5.0 \times 10^{15} \) at/cm\(^3\), and a substitutional boron content of \( 2.0 \times 10^{15} \) at/cm\(^3\).\(^4\)

It is known that the interstitial oxygen \( O_2 \) captures vacancies to form vacancy-oxygen centers VO,\(^5\) while the substitutional carbon \( C \), and boron \( B \), impurities trap Si self-interstitials.\(^6\) Interstitial carbon \( C_i \) is mobile at room temperature and can react with other defects. Interstitial boron \( B_i \), complexes with other Si self-interstitials to form large interstitial clusters \( B_i I_n \).\(^7\)

In the simplest version of our model the following defect reactions are assumed:

\[ V + I \rightarrow \phi \text{ (recombination)}, \]
\[ V + V \rightarrow V_2, \]
\[ V_2 + I \rightarrow V, \]
\[ V + O \rightarrow VO, \]
\[ VO + I \rightarrow O_i, \]
\[ I + C \rightarrow C_i, \]
\[ C_i + V \rightarrow C, \]
\[ VO + C_i \rightarrow C_i O_i, \]
\[ B_i + I \rightarrow B_i I, \]
\[ B_i + I \rightarrow B_i I_2, \]
\[ B_i I_n \rightarrow I \rightarrow B_i I_n. \]

\( V_2 \) and VO were perfectly linear.

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\[ VO + I \rightarrow O_i, \]
\[ I + C \rightarrow C_i, \]
\[ C_i + V \rightarrow C, \]
\[ VO + C_i \rightarrow C_i O_i, \]
\[ B_i + I \rightarrow B_i I, \]
\[ B_i + I \rightarrow B_i I_2, \]
\[ B_i I_n \rightarrow I \rightarrow B_i I_n. \]
TABLE I. Simultaneous differential equations for the reactions (1-11).

\[
\begin{align*}
\frac{d[V]}{dt} &= G_0 - R(V, I)D_1 + D_1[V]/t - 2R[V, V]D_3[V]^2 \\
&\quad + R[V, I]D_3[V]/I \\
\frac{d[I]}{dt} &= G_0 + 2G_{0,o} - R[V, I]D_1 + D_1[V]/t \\
&\quad + R[I, I]D_3[I]/I \\
\frac{d[O]}{dt} &= -R[V, V]D_1[V]^2 + R[V, O]D_1[V + O] \\
&\quad + R[V, I]D_3[V]/I \\
\frac{d[O V]}{dt} &= R[V, I]D_3[V + O]/I - R[V, O]D_1[V + O] \\
&\quad + R[V, O]D_3[V]/I \\
\frac{d[C]}{dt} &= -R[C, V]D_1[V + C] + R[V, C]D_1[V + C] + D_2[V + C][V] \\
&\quad - R[V, O]D_1[V + O] \\
\frac{d[B_1]}{dt} &= -R[B_1, I]D_1[B_1]/I \\
\frac{d[B_2]}{dt} &= R[B_2, I]D_1[B_2]/I \\
\end{align*}
\]

always \( R^* = (X, Y) = 4\pi X(Y, Y) \).

These reactions require the system of simultaneous differential equations shown in Table I.\(^{11}\) We solved this system numerically by using a computer. Values for the various physical parameters which we used in the calculation are shown in Table II.

All capture radii are taken to be 5 Å which is appropriate for neutral defects and elastic interaction at 300 K.\(^{12}\) We neglect charge state effects, although they could be important.\(^{13}\) The values of the diffusion coefficients for the different species which are shown in Table II are based mainly on experimental values for the migration energy and the assumption that the jump frequency is equal to the Debye frequency.\(^{13}\) Our confidence in \( D_i \) is good.\(^{14,17}\) However, we are much less certain about the accuracy of \( D_{C_1}, D_{C_2}, \) and \( D_{B_1}.\)\(^{19}\) Ionization effects on the diffusion of the defects are neglected in this paper. The vacancy-interstitial generation rate shown in Table II is converted into fluence rate by assuming that the number of \( V, I \) pairs produced per 2-MeV electron is 0.5\(^{20}\) and that only 10% of all \( V, I \) pairs escape correlated recovery. Table II also shows the initial concentrations of the various defects.

B. Application to the question of the role of oxygen in divacancy production

In Part I we proposed different possible mechanisms for the oxygen dependence of the divacancy production rate in electron-irradiated silicon at room temperature. We tested those suggestions numerically with our computer model of radiation damage. Below we list the different mechanisms mentioned in Part I and the necessary modifications to the model described in Sec. II A.

1. Kinetics

Only the solution of the system shown in Table I is required.

2. Breton's Suggestion\(^{21}\)

We have to include the capture of Si self-interstitials by interstitial oxygen and the possible interactions of the resulting complex with other defects. The following defect reactions were added to Eqs. (11)-(11):\(^{12}\)

\[
\begin{align*}
O_i + I &\rightarrow O_i I, \quad (12) \\
O_i I + V &\rightarrow O_i V, \quad (13)
\end{align*}
\]

3. \( V_2 O \) Center

We allow here for the formation of a \( V_2 O \) complex from VO. The respective concentrations of \( V_2 O \) and \( V_2 \) are added to get an "apparent" divacancy concentration. We added the following reactions to Eqs. (11)-(11):

\[
\begin{align*}
V + VO &\rightarrow V_2 O, \quad (14) \\
V_2 O + I &\rightarrow VO. \quad (15)
\end{align*}
\]

4. Impurity dependence of primary damage

a. Generation rate of \( V, I \) pairs. The introduction rate of \( V, I \) pairs is assumed to depend on the interstitial oxygen content. We therefore write

\[
G(O_i) = G_0 + C \{ O_i \}. \quad (16)
\]

b. Direct generation rate of divacancies. We allow the direct generation rate of \( I \) to depend on the interstitial oxygen concentration. The constant \( G_{v_2} \) is therefore replaced by

\[
G_{v_2}(O_i) = G_{v_2} + C \{ O_i \}. \quad (17)
\]

For both cases (a and b) we used \( C = 0.1 \) s\(^{-1}\).
increasing A-center density versus rising initial amounts of oxygen. In Brelot’s model there exists a nearly perfect linear relationship between \([VO]\) and \([O_2^-]\) over the whole range of \([O_2^-]\) values. This is the case for the other models only for larger \([O_2^-]\).

The relationship between \([V_2^-]\) and \([VO]\) after an electron dose of \(1.0 \times 10^{18}\) electrons/cm\(^2\) is shown in Fig. 3. The models \(V_2^-O + V_2^-G(O_2^-)\) and the simplest system (kinetics) all show declining divacancy concentrations with larger vacancy-oxygen concentrations. In Brelot’s model \([V_2^-]\) rises nearly linearly with \([VO]\), while for \(G_r^-\) (kinetics) the divacancy concentration is also increasing, but the relationship to \([VO]\) is not linear.

These qualitative trends of the different models are stable with respect to variation of the input parameters, i.e., diffusion coefficients, capture radii, and \(C\) (in Eqs. 16 and 17), within reasonable limits. We therefore believe that the results displayed in Figs. 1–3 reflect intrinsic characteristics of the models described in Part I and Secs. II B–II B 4, and they are not due to the fortuitous choice of a particular set of parameters.

We did not attempt to fit the experimental data quantitatively, since (as pointed out in Part I) it was not possible to convert the measured absorption coefficient values into defect concentrations.

**IV. DISCUSSION**

Our experimental findings reported in Part I give us three criteria which have to be met by the curves shown in Figs. 1–3. As mentioned in Sec. I we found that both \([V_2^-]\) and \([VO]\) increased linearly with increasing interstitial oxygen concentration and \([V_2^-]\) vs \([VO]\) was also linear. A look at Figs. 1–3 shows that only the numerical results of Brelot’s model compare favorably with these results. Only this model reproduces qualitatively the three linear relationships found by our experiments. The other models are in qualitative disagreement with our experimental data. On the basis of this we believe that the formation of \(O_2^-\) pairs is the cause for the increase in the divacancy introduction rate observed for increasing interstitial oxygen concentrations.
This conclusion is consistent with the annealing behavior of the O$_2$I pair, since the available experimental evidence indicates that the O$_2$I pair has the thermal stability required for a room-temperature irradiation.\textsuperscript{22} Although the O$_2$I-associated IR band at 935 cm$^{-1}$ is largest after low-temperature ($\sim$ 77 K) irradiation and decreases upon annealing to 310 K it is still present even after annealing at 350 K.\textsuperscript{22} We therefore conclude that O$_2$I interaction can take place during room temperature electron irradiation and that the resulting complex can be stable at room temperature.

Further support for the correctness of our conclusion comes from the following consideration. In Brelot’s model the divacancy increase in oxygen containing silicon is not related to the chemical nature of the oxygen impurity, but only to the fact that oxygen functions as a Si self-interstitial trap. Therefore replacing oxygen by any other interstitial trap should affect a similar divacancy increase. This actually has been observed for carbon.\textsuperscript{23,24} Alternatively we could directly vary the steady-state vacancy concentration by changing the concentration of vacancy sinks (impurities, surfaces, etc.), and we should observe a divacancy change. Such an experiment has been done by Wang et al.\textsuperscript{25} They studied the surface-defect distributions of electron-irradiated silicon by transient capacitance techniques. They observed a depletion of defects near the surface and attributed it to vacancy trapping at the surface. Further they found that both VO and V$_2$ had the same spatial dependence. We reanalyzed their data for the divacancy and the VO center. Plotting $[V_2]$ vs [VO] gives the graph shown in Fig. 4. We note that $[V_2]$ depends linearly on [VO]. The behavior is in qualitative agreement with the curve of Brelot’s model in Fig. 3. This further supports the notion that oxygen can influence the steady-state vacancy concentration by interstitial trapping.

The significance of our experimental result that $[V_2]$ can be proportional to [VO] (i.e., that $[V_2]$ reflects the steady-state single-vacancy concentration) and the subsequent explanation of this relationship by Brelot’s model is in its relation to the question of divacancy production per se. Since there is conclusive experimental evidence that divacancies can form by the direct collision process during electron irradiation of silicon\textsuperscript{26,27} for a long time the formation of divacancies via combination of two vacancies was thought to be negligible. We can now answer the question of what fraction of a certain divacancy concentration introduced by room-temperature electron irradiation was produced by a direct collision process and what fraction is due to the agglomeration of single vacancies. In our experiment we found:\textsuperscript{1}

$$\alpha_{V_2} = 0.57\alpha_{VO} + 0.25 \text{ cm}^{-1}. \tag{18}$$

Since the fraction of $\alpha_{V_2}$ which is proportional to $\alpha_{VO}$ is due to the combination of two vacancies while the constant gives an upper limit of directly introduced divacancies we can rewrite Eq. (18) as

$$\alpha_{V_2} = \alpha_{\text{indirect}} + \alpha_{\text{direct}}, \tag{19}$$

with $\alpha_{\text{indirect}} = 0.57\alpha_{VO}$ and $\alpha_{\text{direct}} = 0.25 \text{ cm}^{-1}$. For the largest $\alpha_{VO}$ we had $\alpha_{V_2} = 1.2 \text{ cm}^{-1}\text{.}^1$ We then get $\alpha_{\text{indirect}} = 0.95 \times 0.25$ and we therefore have that the indirect production rate of divacancies can be four times as large as the direct production rate during room-temperature irradiation with 2-MeV electrons. It is interesting that Barnes found a similar ratio in a low-temperature (76 K) neutron-damage experiment. Barnes observed immediately after irradiation at 76 K (i.e., with the single vacancies immobile) only 25% of the maximum $[V_2]$ value which was obtained after a 330-K anneal.\textsuperscript{28} Therefore in his case the indirect production rate of $V_2$ was three times as large as the direct production rate.

V. SUMMARY

Numerical tests of possible models for the oxygen dependence of the divacancy introduction rate in at room-temperature electron-irradiated silicon were performed on a computer. Only one of the models can reproduce all the experimental results.\textsuperscript{1} The other models give results which are in qualitative disagreement with the experimental data. We envision the mechanism responsible for the enhancement of divacancy production as follows: Interstitial oxygen prevents by the capture of Si self-interstitials interstitial-vacancy recombination. This in turn increases the steady-state vacancy concentration and ultimately enhances the divacancy concentration via agglomeration of two single vacancies.

Our experimental and modeling results show that the indirect formation of divacancies during room-temperature irradiation of silicon with 2-MeV electrons can be more important than the production of $V_2$ by the single collision process.

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We assume that $R(B,J,J)$ and $R(B,J,J)$ are the same. We also do not solve explicitly for $[O,C]$ and for the $[B,J]$.}


J. L. Lindström, G. Oehrlein, A. E. Jaworowski, and J. W. Corbett, J. Appl. Phys. 53, 8486 (1982) [Part I], and Fig. 7 therein.


