



# Intrinsic Point Defects and Impurities in Silicon Crystal Growth

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The incorporation of intrinsic point defects into a growing crystal is affected by the presence of impurities that can react with vacancies and self-interstitials. The critical value of the ratio of the growth rate,  $V$ , to the axial temperature gradient,  $G$ , ( $V/G$  ratio) that separates the interstitial growth mode from the vacancy growth mode, is shifted by impurities, and this effect can be described by simple analytical expressions. Some impurities, such as oxygen, nitrogen, and hydrogen, trap vacancies and cause a downward shift in the critical  $V/G$  ratio (and also a fast increase in the fraction of trapped vacancies, on lowering  $T$ ). Other impurities, like carbon, trap self-interstitials, and cause an upward shift in the critical  $V/G$  ratio (and also an increase in the fraction of impurity interstitials, on lowering  $T$ ). The impurities affect both the incorporation and agglomeration stages of micro-defect production.

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The structural properties of as-grown silicon are controlled by the type and concentration of intrinsic point defects incorporated into the growing crystal.<sup>1-3</sup> An incorporation model<sup>1</sup> that is now commonly accepted, assumes a fast recombination of intrinsic point defects in the vicinity of the crystal/melt interface. The annihilation stage is effectively complete when the temperature is below the melting point by about 100°C. Since then, only one kind of defect, either vacancy or self-interstitial, is present in a supersaturated concentration while the competing defects rapidly disappear. In a simple approximation of a one-dimensional diffusion field,<sup>1</sup> there is some universal critical value of the ratio of the growth rate,  $V$ , to the axial temperature gradient,  $G$  ( $V/G$ ) that separates the vacancy growth mode (at higher  $V/G$ ) from the interstitial growth mode (at lower  $V/G$ ). Due to a radial and axial nonuniformity in  $G$  and  $V$ , the same crystal may contain both vacancy-type and interstitial-type regions separated by a well-defined boundary.<sup>2-4</sup>

The surviving intrinsic defects will agglomerate, upon the subsequent lowering of  $T$ , into various structural microdefects. For standard vacancy growth mode (at  $V/G$  well above the critical value), the basic microdefects are now recognized as voids<sup>5-7</sup> while for standard interstitial growth mode (at  $V/G$  well below the critical value), the basic microdefects were found to be interstitial-type dislocation loops.<sup>8-10</sup> The microdefect kind is drastically changed when  $V/G$  is close to the critical value, in other words, at a reduced concentration of incorporated point defects. At low vacancy concentration, oxide particles are formed instead of voids by joint agglomeration of vacancies and oxygen;<sup>2</sup> this effect is responsible for the formation of oxidation-induced stacking faults in a ring-like distribution.<sup>11-13</sup> At low self-interstitial concentration, so-called B-defects (globular interstitial aggregates) are formed instead of loops.<sup>4,9,14</sup>

The critical  $V/G$  ratio is of fundamental importance in the growth of dislocation-free silicon crystals of controlled microdefect properties. The reported numbers for the critical ratio scatter considerably, in the range from 0.12 to 0.2 mm<sup>2</sup>/min K.<sup>1,13,15</sup> One of the reasons for such a scatter is an uncertainty in the measured or calculated value of  $G$ , in the vicinity of the crystal/melt interface. However, there are also fundamental reasons for a scattered critical  $V/G$  ratio: (i) the effect of impurities on the incorporation of intrinsic point defects into the growing crystal, and (ii) the effect of the interface shape on the diffusion field of point defects in the vicinity of the interface.<sup>16</sup>

This paper is devoted to the impurity effect on the critical  $V/G$

ratio and to the related effect of trapping vacancies or self-interstitials by impurities, upon lowering the temperature.

## Critical $V/G$ Ratio for Pure Crystals and Flat Interface

The diffusion of point defects in the vicinity of the crystal/melt interface can be treated, approximately, as a one-dimensional problem, since the characteristic length of defect annihilation<sup>1</sup> is considerably smaller than the crystal diameter (this length is about 2 cm in modern Czochralski crystals, for  $G \approx 30$  K/cm). In this approximation, the crystal/melt interface is flat; the critical growth rate  $V_t$ , for the change-over from interstitial to vacancy incorporation, depends, strictly speaking, on the entire axial temperature profile  $T(z)$ . However, the most important feature of the  $T(z)$  profile is the near-interface value of  $G$  (defined as  $-dT/dz$ ). The  $V_t$  is dependent mostly on  $G$ . Particularly, for a linear axial profile of the inverse temperature and for temperature-independent diffusivities of the point defects, there is a precise analytical solution for the critical ratio (denoted by  $\xi_t$ ), through the point defect parameters<sup>17</sup>

$$\begin{aligned}\xi_t &= V_t/G \\ &= (1/kT_m^2)[(E - \varepsilon_i)D_iC_{im} \\ &\quad - (E - \varepsilon_v)D_vC_{vm}]/[C_{vm} - C_{im}]\end{aligned}\quad [1]$$

where  $E = (E_i + E_v)/2$  is the defect formation energy averaged over the two species,  $\varepsilon_i, \varepsilon_v$  are the drift energies that characterize the defect drift along the temperature gradient,  $D_i$  and  $D_v$  are the diffusivities, and  $C_{im}, C_{vm}$  are the equilibrium concentrations at the melting point ( $T_m$ ). The subscripts i and v refer to interstitial and vacancy, respectively. The critical growth rate,  $V_t$  corresponds to the zero total flux of the excess silicon, which is the difference of the interstitial and vacancy fluxes

$$\begin{aligned}J &= J_i - J_v \\ &= [-D_i\nabla C_i + D_v\nabla C_v] + [V(C_i - C_v)] \\ &\quad + [-D_iC_i\nabla(\varepsilon_i/kT) + D_vC_v\nabla(\varepsilon_v/kT)]\end{aligned}\quad [2]$$

where the three bracketed terms correspond to Fickian diffusion, convection, and thermal drift, respectively. The two concentration profiles,  $C_i(z)$  and  $C_v(z)$ , are related by the condition of fast recombination: the product of the two is equal to the product of the equilibrium concentrations,  $C_{ie}(T)C_{ve}(T)$ . Equation 2 can be integrated to find the flux,  $J$ , in dependence of the growth rate  $V$ .<sup>17</sup> The  $V_t$

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**Table I. Point defect parameters. The top line of numbers is after Ref. 17, the bottom row is for smaller formation energies.**

$C_{vm}$ ( $cm^{-3}$ )	$C_{im}$ ( $cm^{-3}$ )	$D_{im}$ ( $cm^2/s$ )	$D_{vm}$ ( $cm^2/s$ )	$E_{id}$ (eV)	$E_{vd}$ (eV)	$E$ (eV)	$\varepsilon_i$ (eV)	$\varepsilon_v$ (eV)
$1 \times 10^{15}$	$7.7 \times 10^{14}$	$4 \times 10^{-4}$	$5 \times 10^{-5}$	0.35	0.35	4.6	0	0
$1 \times 10^{15}$	$7.7 \times 10^{14}$	$4 \times 10^{-4}$	$5 \times 10^{-5}$	0.95	0.35	4	0	3

corresponds to zero flux. At  $V < V_t$ , the function  $J(V)$  is positive which corresponds to interstitial incorporation. Otherwise,  $J(V) < 0$  which corresponds to vacancy incorporation.

### Impurity Effect on $\xi_t$ through the Equilibrium Concentrations of Intrinsic Point Defects

The critical ratio,  $\xi_t$ , is sensitive to the equilibrium defect concentrations, mostly through the denominator of Eq. 1 since the two concentrations,  $C_{vm}$  and  $C_{im}$ , are nearly equal.<sup>17</sup> Therefore, even relatively small shifts in the concentrations (denoted by  $\delta C_{im}$  and  $\delta C_{vm}$ ) will cause an appreciable shift in  $\xi_t$ . This shift is approximated taking into account only a change in the denominator of Eq. 1

$$\xi_t = \xi_{to} / [1 + (\delta C_{vm} - \delta C_{im}) / (C_{vm} - C_{im})] \quad [3]$$

where  $\xi_{to}$  refers to the nondoped crystal. Generally, the equilibrium concentrations may be shifted due to

1. A shift in the Fermi level by a charged dopant that is present in the concentration not much lower than the intrinsic electron concentration. The equilibrium concentration of negatively charged defects (vacancies) will change in proportion to the electron concentration. This mechanism can explain well<sup>18</sup> the reported effect of boron impurity<sup>19</sup> which is to increase  $\xi_t$ . The boron acceptors reduce the electron concentration;  $\delta C_{vm}$  is negative and proportional to the boron concentration. The donor dopants are expected to reduce the critical ratio  $\xi_t$  by this mechanism.

2. A lattice strain induced by impurity atoms, if the impurity atom size is different from that of a host atom.<sup>20</sup> The impurity-induced lattice stress is a superposition of local stress fields around the impurity atoms and some constant pressure; only the latter stress component affects the equilibrium concentrations.<sup>21</sup> The strain effect on  $C_{vm}$  or  $C_{im}$  turns out to be negligible at the conventional doping level, below  $10^{19} cm^{-3}$ .

### Impurity Effect on $\xi_t$ Through Vacancy Trapping

There are some other impurities (oxygen, nitrogen, hydrogen, carbon) which also exhibit an effect on  $\xi_t$ .<sup>14,16,22,23</sup> These impurities are either neutral or of low concentration, and the mechanism of the impurity effect is through chemical reactions of impurity species with the intrinsic point defects. The two basic reactions are

1. Vacancy trapping and releasing by a major impurity state. Particularly, this applies to interstitial impurities with a small substitutional component; in this case the trapped (or bound) vacancies are identical to the substitutional impurity atoms.

2. Self-interstitial trapping and releasing by a major impurity state. Particularly, this applies to substitutional impurities with a small interstitial component; in this case, the trapped (or bound) self-interstitials are identical to interstitial impurity atoms, and trapping occurs by the kick-out reaction.

In this section we consider an impurity that traps vacancies. If the trapping and releasing reaction is fast, the free vacancies of the concentration  $C_v$  and the trapped vacancies of the concentration  $C_v^*$  coexist in the equilibrium ratio

$$C_v^*/C_v = (C_{vm}^*/C_{vm}) \exp(E_b/kT - E_b/kT_m) \quad [4]$$

where  $C_{vm}^*$  is the concentration of trapped vacancies at  $T_m$ , in equilibrium with the major impurity state, and  $E_b$  is the energy gain on

formation of one trapped vacancy. For simple atomic traps, like oxygen,  $E_b$  is identical to the binding energy of a vacancy to an impurity atom, and  $C_{vm}^*$  is just proportional to the total impurity concentration,  $C$ . The situation is a bit more complicated for nitrogen (see below) but Eq. 4 still holds. It should be noted that the total impurity concentration,  $C$ , is assumed to be much higher than the incorporated vacancy concentration (which is typically less than  $10^{14} cm^{-3}$ ), to neglect a small loss of the major impurity state for formation of trapped vacancies.

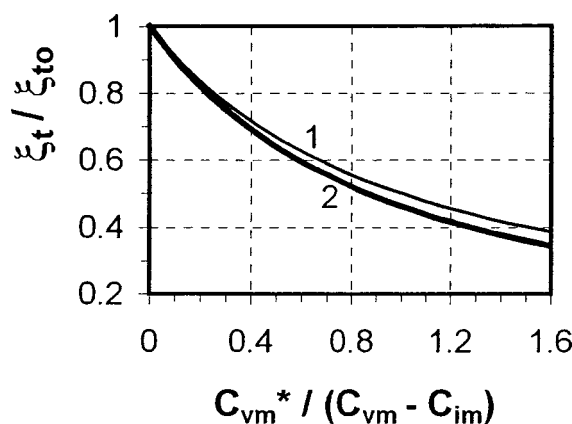
Within the incorporation model,<sup>1,17</sup> the concentrations of free vacancies and of self-interstitials at the crystal/melt interface are of equilibrium values ( $C_{vm}$  and  $C_{im}$ , respectively), due to fast consumption/generation reactions at the interface. In the presence of vacancy traps, the free vacancies, in the same concentration  $C_{vm}$ , coexist with the trapped vacancies; and the total concentration of the vacancy species is increased. Due to the fixed proportion (Eq. 4) between the free and trapped vacancies, both vacancy species are equally spent on the recombination with self-interstitials. Therefore, the vacancy convection flux in Eq. 2 should be modified by replacing  $C_v$  with the total concentration of the vacancy species,  $C_v + C_v^*$ . Since the denominator in Eq. 1 comes from the convection term, a shift in the critical  $V/G$  ratio due to the presence of trapped vacancies can be qualitatively described by replacing  $C_{vm}$  with the total concentration of the vacancy species at  $T_m$  which is  $C_{vm} + C_{vm}^*$ , in the denominator of the Eq. 1

$$\xi_t = \xi_{to} / [1 + C_{vm}^*/(C_{vm} - C_{im})] \quad [5]$$

This shift (due to the presence of trapped vacancies) is very similar to that caused by a shift in the free vacancy concentration (Eq. 3).

The formula (Eq. 5) would be precise at  $E_b = 0$  and with temperature-independent diffusivities, since then the flux (Eq. 2) is modified just by multiplying  $C_v$  by a constant factor,  $1 + C_v^*/C_v$ , and by simultaneously dividing  $D_v$  by the same factor. With nonzero  $E_b$  and the temperature-dependent diffusivities, one should find the critical ratio by numerical integration of the modified flux equation (Eq. 2), in order to find the function  $J(V)$  and the growth rate  $V_t$  of zero flux. The integration (quite similar to that described in Ref. 17) was performed using the point defect parameters specified by Table I. The parameter set includes the melting point values for the equilibrium concentrations ( $C_{vm}, C_{im}$ ) and for the diffusivities ( $D_{im}, D_{vm}$ ), and the energies of migration ( $E_{id}, E_{vd}$ ), formation ( $E$ ), and drift ( $\varepsilon_i, \varepsilon_v$ ). The measured temperature dependence of the concentration difference,  $C_{ve} - C_{ie}$ , indicates<sup>24</sup> that the formation energy of a self-interstitial,  $E_i$ , and that of a vacancy,  $E_v$ , are almost the same. It is thus adopted that  $E = E_i = E_v$ .

The two parameter sets were actually used; the top line of numbers in Table I corresponds to the set specified in Ref. 18 under the assumption of a low migration energy of self-interstitials (here this energy is taken to be the same as the vacancy migration energy). Since the sum of the formation and migration energies for self-interstitial,  $E_i + E_{id}$ , is known to be 4.95 eV,<sup>25</sup> this assumption implies a rather high formation energy,  $E_i = 4.6$  eV. The quantum calculations<sup>26</sup> suggest a remarkably lower formation energies, about 4 eV for the vacancy. The interstitial migration energy (with  $E_i = E_v$ ) is then 0.95 eV. These values are used in the second param-



**Figure 1.** Relative shift in the critical  $V/G$  ratio caused by vacancy trapping impurity, in dependence of the normalized concentration of trapped vacancies at  $T_m$ . Curve 1: analytical expression (Eq. 5), curve 2: numerical computation at  $E_b = 2$  eV, for either of the two parameter sets.

eter set, shown in the bottom row of numbers in Table I. A high vacancy drift energy is now introduced. This is done for the following formal reasons: (i) to keep the same critical ratio  $\xi_{to}$  for the two sets, and (ii) to demonstrate that the drift, even as strong as that, is not essential for the impurity-induced shift in the critical  $V/G$  ratio. The second parameter set is quite close to that of Ref. 27. By the results of this paper, the second parameter set seems preferable.

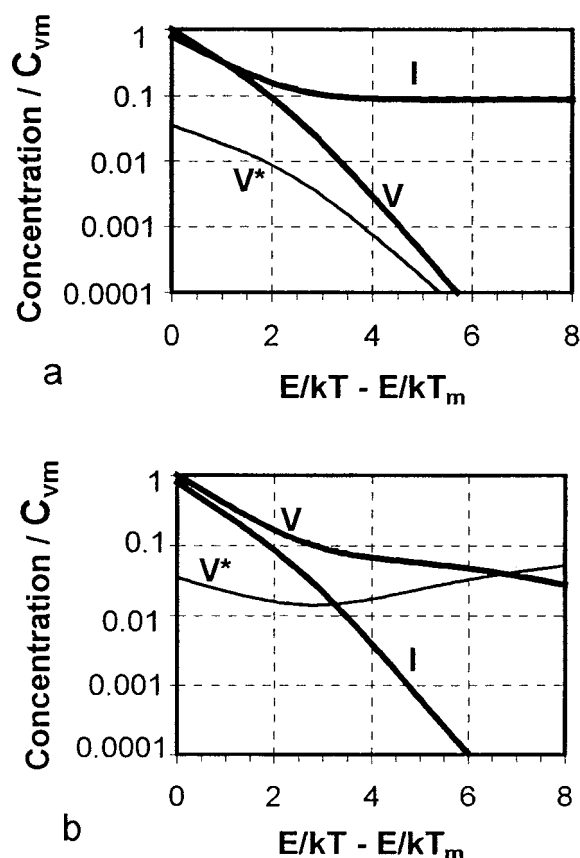
The dependence of the shifted ratio,  $\xi_t/\xi_{to}$ , on the impurity parameter  $C_{vm}^*/(C_{vm} - C_{im})$ , was computed for several values of binding energy  $E_b$ . The computed curves for the two parameter sets are almost identical, and they deviate only slightly from the simple formula (Eq. 5). An example is shown in Fig. 1, for  $E_b = 2$  eV (this value is representative for the particular impurities discussed below). The qualitative analytical expression (Eq. 5) contains just one intrinsic point defect parameter, the concentration difference  $C_{vm} - C_{im}$ , whereas the transport equation (Eq. 2) includes many parameters. And yet, Eq. 5 well describes the impurity-induced shift which is quite insensitive to all the parameters except  $C_{vm} - C_{im}$ .

An impurity that traps vacancies causes a reduction in  $\xi_t$ , and thus an expansion of the vacancy region at the expense of the interstitial region, when the two regions coexist in one crystal.

Another important consequence of vacancy trapping is to reduce the fraction of free vacancies in the incorporated vacancy community, on lowering  $T$ , according to Eq. 4. Below some characteristic "binding temperature"  $T_b$ , the bound (trapped) vacancies will prevail over the free vacancies. An example of computed axial profiles of self-interstitials (curve labeled I), free vacancies (curve labeled V) and trapped vacancies (curve labeled  $V^*$ ) is shown in Fig. 2, for particular values of  $V/V_t$  and  $C_{vm}^*/(C_{vm} - C_{im})$ . In the interstitial growth mode (Fig. 2a) both free and bound vacancies disappear by recombination with self-interstitials, and some self-interstitial concentration is incorporated. In the vacancy growth mode (Fig. 2b) the survived vacancy population consists of the two species,  $V$  and  $V^*$ , of some constant total concentration, but the ratio of the two species changes gradually in favor of trapped vacancies. The binding temperature corresponds to the intersection point of curves  $V$  and  $V^*$ .

When  $T_b$  is less than the void nucleation temperature  $T_n$ , vacancy trapping does not affect void formation and their properties very much. On the other hand, if  $T_b$  becomes greater than  $T_n$ , then void formation will be either strongly modified or completely suppressed, due to a lack of free vacancies. This effect is discussed further in the section below devoted to the nitrogen impurity.

**Vacancy trapping by oxygen.**—The oxygen impurity in silicon is well known to exist in the interstitial state. The substitutional (off-center) state of oxygen, or the vacancy/oxygen complex VO (known



**Figure 2.** Axial concentration profiles of free vacancies ( $V$ ), trapped vacancies ( $V^*$ ), and self-interstitials ( $I$ ) at  $C_{vm}^*/(C_{vm} - C_{im}) = 0.15$  and  $E_b = 2$  eV, for the second parameter set. (a, top) Interstitial mode,  $V/V_t = 0.7$ ; (b, bottom) vacancy growth mode,  $V/V_t = 1.5$ .

also as A-center) is produced in irradiation experiments.<sup>28</sup> By annealing the irradiated samples, the A-centers disappear at  $T > 350^\circ\text{C}$ , presumably by attaching more oxygen atoms and not by dissociation. The binding energy of a vacancy to an oxygen atom may be large enough for the VO species to be present in appreciable amount even at high  $T$ , in equilibrium with free vacancies. Evidence of high temperature vacancy trapping by oxygen comes from the fact that, in the course of crystal growth, some vacancies escape consumption by voids; these "residual vacancies" control subsequent oxygen precipitation, giving rise to a complicated banded pattern.<sup>3,29</sup> This effect is accounted for if the free vacancies become bound (trapped) by oxygen, below some characteristic binding temperature  $T_b$  (estimated to be around  $1050^\circ\text{C}$ ); otherwise the vacancies would be completely consumed by the voids. A vacancy, due to its four dangling bonds, can absorb either one or two oxygen atoms, giving rise to VO and  $\text{VO}_2$  species. We assume that the reactions of transformation of  $V$  into VO (by vacancy diffusion) and of VO into  $\text{VO}_2$  (by oxygen diffusion) are fast enough to support the equilibrium ratio between the species concentrations  $C_v$ ,  $C_1$ , and  $C_2$  (of  $V$ , VO, and  $\text{VO}_2$ , respectively). It is also assumed that the free energy change, upon attachment of the first and the second oxygen atom to a vacancy, is the same. A qualitative argument in favor of such a simplification is that an oxygen atom in the A-center is close to the two of the silicon neighbors<sup>28</sup> leaving the two other silicon neighbors in the same state as that in the free vacancy. With this assumption, the mass-action law for the above reactions reads

$$C_1/C_v = C_2/C_1 = AC \exp(E_b/kT) \quad [6]$$



where  $C$  is the oxygen concentration; the prefactor  $A$  includes a change in the vibrational entropy due to the dangling bonds, on attaching an oxygen atom to a vacancy. The binding temperature may be defined by the condition  $C_1/C_v = C_2/C_1 = 1$ . Below  $T_b$ , this concentration ratio becomes high, and then  $C_2 \gg C_1 \gg C_v$ , which means that the prevailing vacancy state is  $VO_2$ . At temperatures greater than  $T_b$ , this ratio becomes small, and  $C_2 \ll C_1 \ll C_v$ . This means that free vacancies prevail with some small fraction of VO species and a much smaller fraction of  $VO_2$ .

The VO species (trapped vacancies) that are present at  $T_m$  will shift the critical  $V/G$  ratio according to Eq. 5; the concentration  $C_{vm}^*$  is equal to  $R_s C$  where  $R_s$  is the equilibrium fraction of substitutional oxygen (VO). The general expression (Eq. 5) is then presented in the form

$$\xi_t = \xi_{t0}/(1 + C/C_o) \quad [7]$$

where the constant  $C_o = (C_{vm} - C_{im})/R_s$ . The oxygen-induced shift in  $\xi_t$  was indeed found;<sup>16,22</sup> the dependence of  $\xi_t$  on the oxygen content,  $C$ , can be described by Eq. 7 with  $C_o = 4 \times 10^{18} \text{ cm}^{-3}$ ; the corresponding value of  $R_s$  is  $5 \times 10^{-5}$ . With specified value of  $R_s$ , both  $E_b$  and  $T_b$  can be estimated. In order to do so, we note that the equilibrium concentration,  $C_{2e}$ , of a  $VO_2$  species (that has no dangling bonds), can be written neglecting a vibrational entropy of this defect. In the equilibrium state ( $C_v = C_{ve}$ ), a  $VO_2$  species can be created without participation of a vacancy, by taking two oxygen atoms from the solid solution and removing a silicon atom to the sample surface. Then, the mass action law relates the atomic fractions of  $VO_2$  and of oxygen

$$(C_{2e}/\rho) = (C/\rho)^2 \exp(-E_2/kT) \quad [8]$$

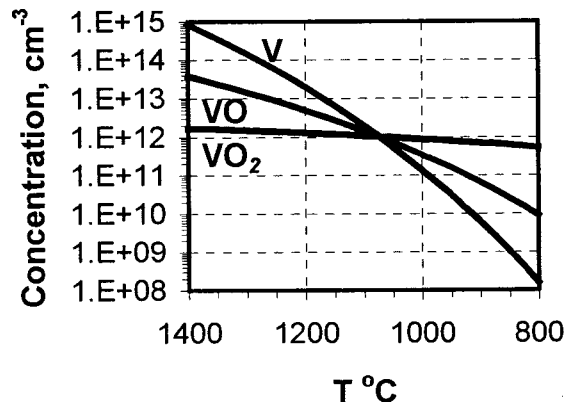
where  $\rho$  is on the order of the lattice site density ( $5 \times 10^{22} \text{ cm}^{-3}$ ) and  $E_2$  is the energy increment by the considered reaction. The relations (Eq. 6) specify the concentrations of VO and  $VO_2$  species in equilibrium with a vacancy solution

$$(C_1/C_v)^2 = C_2/C_v = (C_{2e}/C_{ve}) \quad [9]$$

The concentration  $C_2$  differs from the equilibrium value  $C_{2e}$  by including the vacancy supersaturation factor,  $C_v/C_{ve}$ . By comparing Eq. 9 and 4, the binding energy is expressed as  $E_b = (E_v - E_2)/2$ . At the melting point, the  $C_1/C_v$  ratio is identical to  $R_s C/C_{vm}$ , and Eq. 8 and 9 provide a relation between  $E_2$  and  $R_s$ ; the resulting value of the  $VO_2$  formation energy is  $E_2 = 0.3 \text{ eV}$ . Accordingly, the binding energy of a vacancy to an oxygen atom is  $E_b = 1.85 \text{ eV}$  (at  $E_v = 4 \text{ eV}$ ). Each equilibrium concentration,  $C_{2e}$  (of  $VO_2$ ) and  $C_{1e}$  (of VO) is now specified as a function of  $T$ , by Eq. 8 and 9. These concentrations are plotted in Fig. 3, at  $C = 8 \times 10^{17} \text{ cm}^{-3}$ . The binding temperature is close to  $1070^\circ\text{C}$ . The actual concentrations  $C_v$ ,  $C_1$ , and  $C_2$  are in the same ratio as the equilibrium concentrations  $C_{ve}$ ,  $C_{1e}$ , and  $C_{2e}$ .

The curves in Fig. 3 are based solely on the high temperature oxygen effect on  $\xi_t$ . They should be considered only as a rough description of the actual relation between the V, VO, and  $VO_2$  species, based on the assumptions made. And yet, the resulting value of the binding temperature is in a reasonable agreement with a rough estimate of  $T_b$  based on the residual vacancy consideration.<sup>29</sup> If  $E_v$  were assumed to be  $4.6 \text{ eV}$ , a too high value of  $T_b$  (about  $1100^\circ\text{C}$ ) would be obtained. This is an argument in favor of a lower formation energy ( $E_v = 4 \text{ eV}$ ) which is adopted from now on.

In conclusion, the equilibrium concentration of VO species (substitutional oxygen, or A-center) is relatively low, about  $4 \times 10^{13} \text{ cm}^{-3}$  at  $T_m$ . It is, however, enough for an appreciable shift (about 20%) in the critical  $V/G$  ratio. The critical ratio in Czochralski (CZ) silicon is always shifted downward, with respect to the "genuine" critical ratio for impurity-free silicon. On lowering  $T$ , the



**Figure 3.** Equilibrium concentrations of vacancy species V, VO, and  $VO_2$ , at the oxygen concentration  $C = 8 \times 10^{17} \text{ cm}^{-3}$ .

fraction of trapped vacancies increases, and approaches 50% at about  $1060^\circ\text{C}$ . Since the voids are formed at a higher temperature (at about  $1100^\circ\text{C}$ ), the vacancy trapping by oxygen is of a minor importance for the void formation.

**Vacancy trapping by nitrogen.**—The nitrogen impurity is thought to be present in the di-interstitial (molecular) state,  $N_2$ , at room temperature.<sup>30</sup> Though it dissociates on raising  $T$ , producing atomic interstitials,  $N_1$ , the first-principle calculations<sup>31,32</sup> suggest that the di-interstitial state prevails even at  $T_m$ . The authors of Ref. 32 calculated the energies of various nitrogen/vacancy species: substitutional nitrogen (VN, a complex of a vacancy and one interstitial nitrogen atom), substitutional molecule ( $VN_2$ ), and also the  $V_2N_2$  complex, all produced from the basic molecular state,  $N_2$ . The species concentrations, in equilibrium with the solid solution of nitrogen molecules, can be expressed by a mass-action law using the reported energies. A change in the vibrational entropy upon formation of these species will be neglected, and the resulting numbers should be considered as rough estimates. The square brackets will be used to denote the atomic fraction of the species (the ratio of the concentration to the site density  $\rho$ ).

For atomic nitrogen interstitials

$$[N_1] = [N_2]^{1/2} \exp(-2.15 \text{ eV}/kT) \quad [10]$$

Formation of substitutional nitrogen, VN, for the equilibrium case ( $C_v = C_{ve}$ ), can be performed without assisting vacancies (by removal of a silicon atom to the crystal surface), and the mass-action law reads quite similarly to Eq. 10 with the formation energy of  $1.74 \text{ eV}$ .<sup>32</sup> When, however, this species is produced by trapping a vacancy, the concentration of VN is proportional to  $C_v$ , and therefore, includes the supersaturation factor  $C_v/C_{ve}$

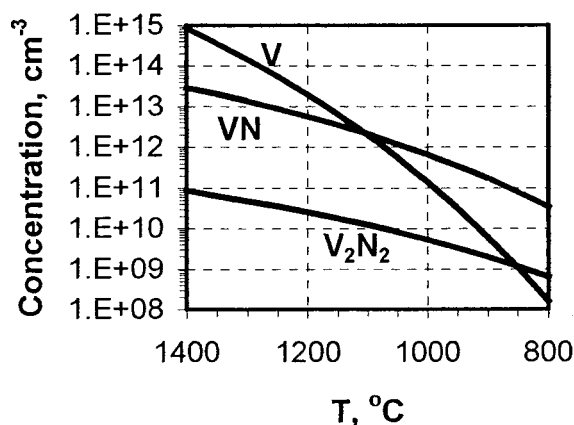
$$[VN] = (C_v/C_{ve})[N_2]^{1/2} \exp(-1.74 \text{ eV}/kT) \quad [11]$$

The concentrations of the other vacancy/nitrogen species are expressed similarly

$$[VN_2] = (C_v/C_{ve})[N_2] \exp(-2.33 \text{ eV}/kT) \quad [12]$$

$$[V_2N_2] = (C_v/C_{ve})^2[N_2] \exp(-1.26 \text{ eV}/kT) \quad [13]$$

Assuming for the total nitrogen concentration,  $C$ , a typical value of  $10^{15} \text{ cm}^{-3}$ , the calculated concentrations at  $T_m$  (when  $C_v/C_{ve} = 1$ ) are all much lower than the concentration of the molecular form. The latter is thus close to  $C/2 = 5 \times 10^{14} \text{ cm}^{-3}$ . The concentrations of the other species, as defined by Eq. 10-13, are as follows (in units of  $\text{cm}^{-3}$ ): atomic interstitials  $N_1$ :  $1.9 \times 10^{12}$ ; sub-



**Figure 4.** Equilibrium concentrations of vacancy species V, VN, and  $V_2N_2$  in nitrogen doped silicon, at the nitrogen concentration  $C = 10^{15} \text{ cm}^{-3}$ . The  $VN_2$  species is of too low concentration to show in the figure.

stitutional species VN:  $3.1 \times 10^{13}$ ;  $VN_2$ :  $5.5 \times 10^7$ ; and  $V_2N_2$ :  $8.6 \times 10^{10}$ . The formation energy is the lowest for the  $V_2N_2$ . It induced the authors of Ref. 31, 32 to suggest that  $V_2N_2$  is the prevailing vacancy/nitrogen species. However, due to the difference in the prefactors in Eq. 11 and 13, the actually prevailing species (at  $T_m$ ) is VN, the substitutional nitrogen atom. The concentration ratio  $C_v^*/C_v$  is specified by Eq. 11 and reduced to the general form of Eq. 4: the energy gain  $E_b = E_v - 1.74 \text{ eV} = 2.36 \text{ eV}$ . Equation 5, with Eq. 11 for the concentration of the trapped vacancies, is presented in the form

$$\xi_t = \xi_{t0} / [1 + (C/C_0)^{1/2}] \quad [14]$$

where the constant  $C_0 = 5.6 \times 10^{16} \text{ cm}^{-3}$ . The difference from the previous case of oxygen is only in the dependence of the concentration of trapped vacancies,  $C_{vm}^*$ , on the total impurity concentration:  $C_{vm}^*$  is now proportional to  $C^{1/2}$ , due to the dominance of molecular interstitial state. The expected downward shift in  $\xi_t$  is about 12%, which is consistent with the observed effect of nitrogen to expand the vacancy region in Czochralski crystals grown at a ramped rate.<sup>22</sup>

The equilibrium concentrations of the vacancy species, at  $C = 10^{15} \text{ cm}^{-3}$ , are plotted in Fig. 4. The binding temperature is close to  $1115^\circ\text{C}$ , if the contribution of the  $V_2N_2$  species is neglected. For the actual case of supersaturated vacancy solution, the relative contribution of the  $V_2N_2$  species is increased due to the square dependence on the supersaturation, according to Eq. 13. To include this effect, one should find the nonequilibrium concentrations of all the vacancy species using Eq. 11-13 and specifying the total incorporated concentration of vacancies, which is typically about  $10^{14} \text{ cm}^{-3}$ .<sup>17</sup> The relative contribution of  $V_2N_2$  species, though enhanced, turns out to be still small at  $T_b$ : about 8%. Therefore, the dominant type of trapped vacancies at  $T_b$  is still the substitutional nitrogen, VN.

According to the above estimates, the nitrogen impurity is a powerful trap for vacancies, in comparison to oxygen. The void formation in nitrogen-doped CZ silicon occurs under condition of well-pronounced trapping of vacancies by nitrogen. The trapping kinetics is either by dissociation of  $N_2$  and subsequent reaction of  $N_1$  with vacancies, or by reaction of V directly with  $N_2$ , to produce VN and  $N_1$ . The interstitial nitrogen atoms produced by the latter reaction will trap vacancies, contributing to the overall rate of VN formation. We assume that, by one or both of these paths, the equilibrium between the free and bound vacancies is maintained (a quantitative proof of this assumption requires a separate study). In this case, the free vacancy loss to voids is accompanied by a release of

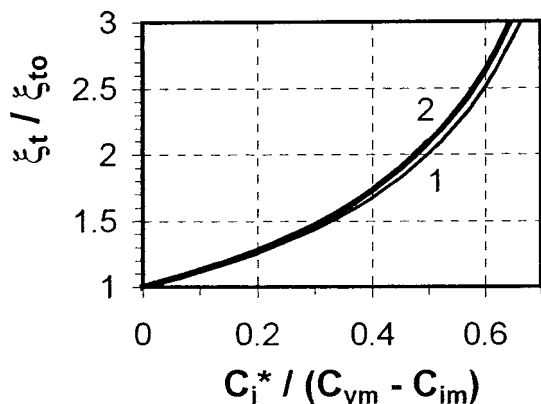
trapped vacancies, to support the equilibrium  $C_v/C_v^*$  ratio, and finally, almost all vacancy species are consumed in the formation of voids. A reduction in the free vacancy fraction is equivalent to a reduced effective mobility of the vacancy population (of which only a fraction is mobile). The density of produced voids is inversely proportional to this effective mobility, to the power of 1.5.<sup>2</sup> At reduced vacancy mobility (and thus at reduced vacancy flux to the voids), a larger density of voids will be produced before the vacancy loss to the voids becomes fast enough to suppress further void nucleation. Accordingly, the void density in nitrogen-doped silicon should increase, at the same growth parameters. A reduced concentration of free vacancies results in a lower nucleation temperature of voids and, therefore, in a further reduction in the effective vacancy mobility. This effect contributes into the increased void density. Both an increase in the void density and a downward shift in the void nucleation temperature, due to nitrogen doping, were indeed reported.<sup>33-35</sup> Whether these effects can be described quantitatively by the vacancy trapping alone, needs further investigation.

On the other hand, nitrogen is well known to completely suppress void formation in float-zoned (FZ) silicon.<sup>36,37</sup> It can be explained by the difference in the void formation temperature between CZ and FZ crystals. In CZ crystals the void formation is thought to be assisted by oxygen, which is either adsorbed at the void surface<sup>38</sup> or involved in nucleation of the primary oxide particles which are then transformed into voids, by cavitation.<sup>39</sup> As pointed out above, the nucleation temperature, in the absence of nitrogen, is close to the vacancy binding temperature by nitrogen. The free vacancy concentration, though somewhat reduced by formation of VN, is still sufficient for the production of voids: in increased density, due to a reduced effective vacancy mobility. In nitrogen-free FZ silicon,  $T_n$  is expected, due to the lack of oxygen, to be substantially lower than that in CZ silicon (although this is not yet proven experimentally). This value of  $T_n$  is well below the nitrogen-controlled binding temperature (which is around  $1115^\circ\text{C}$ ). Under such conditions, provided the concentration of nitrogen is sufficient to trap all the vacancies, the free vacancy concentration is so reduced upon lowering  $T$  as to completely prevent void formation. The vacancy species, VN, will be completely preserved in the crystal.

This conclusion is supported by the results obtained by the Pt-diffusion technique used to count grown-in vacancy species in both CZ and FZ samples.<sup>40,41</sup> The counted vacancy species are not necessarily only free vacancies but also other species like  $VO_2$  or VN (these are counted by replacing oxygen or nitrogen by a Pt atom). It was found that the Pt concentration is very low both in CZ samples and in FZ samples nondoped with nitrogen,<sup>40,41</sup> which shows that the initially incorporated vacancies, in a concentration of about  $10^{14} \text{ cm}^{-3}$ , are almost completely consumed by the voids. The FZ samples studied in Ref. 40 were nitrogen-doped, unlike those reported in Ref. 41 (private communication by Pichler and Quast). The Pt concentration in nitrogen-doped FZ samples was up to  $10^{14} \text{ cm}^{-3}$ , which proves that the incorporated vacancies are preserved from agglomeration, most likely as VN species. The presence of these grown-in point defects may account for a strong increase in the resistivity, due to formation of deep centers, observed in nitrogen-doped FZ crystals after a very short annealing at 900 or  $1000^\circ\text{C}$ .<sup>42</sup> A plausible mechanism of this effect is a reconstruction of initially inactive VN species into deep centers. The deep nitrogen-related centers (presumably substitutional nitrogen atoms) were also found in implanted samples.<sup>43</sup>

In conclusion, the nitrogen effect in FZ crystals is to expand the vacancy region but to shrink the void-containing portion of this region.

**Vacancy trapping by hydrogen.**—Hydrogen was the first impurity reported to affect formation of grown-in microdefects.<sup>14</sup> It was found that the critical growth rate, for disappearance of swirl defects (loops), was strongly reduced, by a factor of 2, in the presence of hydrogen, roughly, in concentration of  $10^{15} \text{ cm}^{-3}$ . The concentration of vacancy-hydrogen complexes can be estimated from Eq. 5 at



**Figure 5.** Relative shift in the critical  $V/G$  ratio caused by interstitial trapping impurity, in dependence of the normalized concentration of impurity interstitials at  $T_m$ . Curve 1: analytical expression (Eq. 16), curve 2: numerical computation at  $E_k = 2$  eV, for either of the two parameter sets.

$\xi_t/\xi_{to} = 0.5$ ; it is as high as  $2 \times 10^{14} \text{ cm}^{-3}$  (at  $T_m$ ). No further details of vacancy trapping by hydrogen can be given, due to the lack of information on the vacancy-hydrogen complexes.

#### Impurity Effect on $\xi_t$ Through Self-Interstitial Trapping

The effect of predominantly substitutional impurity with a minor interstitial component is quite symmetric to the above case of vacancy trapping. Now the self-interstitials are trapped by the major impurity species, to create, by the kick-out reaction, a minor interstitial impurity state. Again, we assume that the self-interstitials (of concentration  $C_i$ ) and the impurity interstitials (of concentration  $C_i^*$ ) coexist in the equilibrium ratio

$$C_i^*/C_i = (C_{im}^*/C_{im}) \exp(E_k/kT - E_k/kT) \quad [15]$$

where  $E_k$  is the energy gain by the kick-out reaction, and  $C_{im}^*$  is the concentration of impurity interstitials, in equilibrium with the major substitutional component, at  $T_m$ . In other words,  $C_{im}^* = R_i C$  where  $R_i$  is a small fraction of interstitial impurity state at  $T_m$ , and  $C$  is the total impurity concentration.

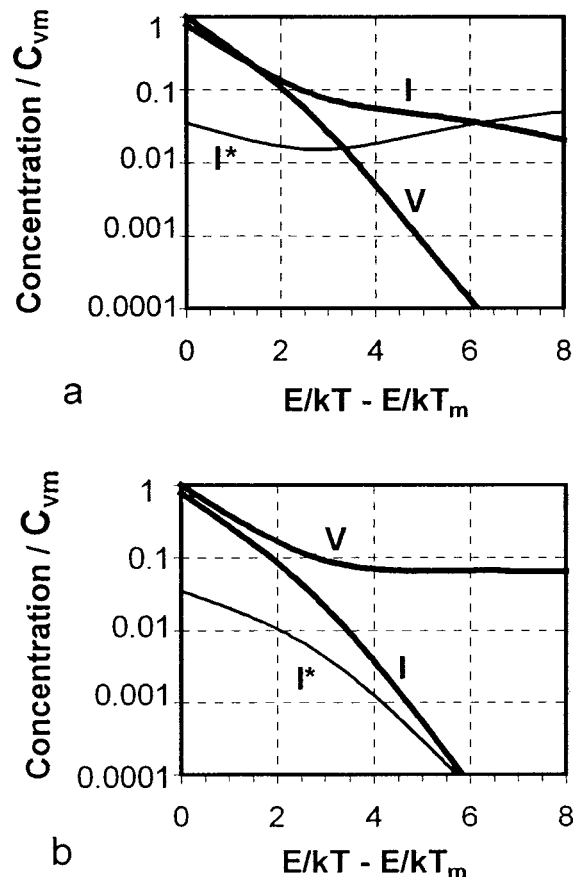
At the melting point, the self-interstitials are present in the equilibrium concentration  $C_{vm}$  along with the trapped (impurity) interstitials, in concentration  $C_{vm}^*$ . Both interstitial species are consumed upon recombination with vacancies. The interstitial convection flux is now modified by replacing  $C_i$  with the total interstitial concentration  $C_i + C_i^*$ . A shift in the critical ratio is qualitatively described by replacing  $C_{im}$  with  $C_{im} + C_{im}^*$  in the denominator of the Eq. 1

$$\xi_t = \xi_{to} / [1 - C_{im}^*/(C_{vm} - C_{im})] \quad [16]$$

In the precise formulation, the modified flux equation 2 is integrated, to get the  $J(V)$  function and define the critical growth rate  $V_t$  of zero flux. The computed dependence of the shifted critical ratio on the impurity parameter  $C_{im}^*/(C_{vm} - C_{im})$  is compared (Fig. 5) to that defined by the qualitative formula, Eq. 16. Like in the case of vacancy trapping, a shift in the critical  $V/G$  ratio is insensitive to a particular parameter set; it is well described by the analytical expression, Eq. 16.

The impurity that traps self-interstitials induces an increase in  $\xi_t$ , and thus an expansion of the interstitial region at the expense of the vacancy region, when the two regions coexist in same crystal.

On lowering  $T$ , interstitial trapping results in a gradual replacement of self-interstitials by impurity interstitials, according to Eq. 15. Below some characteristic kick-out temperature,  $T_k$ , the impurity interstitials will prevail over the self-interstitials. An example of computed axial profiles is shown in Fig. 6, for particular values of



**Figure 6.** Axial concentration profiles of self-interstitials (I), impurity interstitials ( $I^*$ ), and vacancies (V) at  $C_{im}^*/(C_{vm} - C_{im}) = 0.15$  and  $E_k = 2$  eV, for the second parameter set. (a, top) Interstitial mode,  $V/V_t = 0.7$ ; (b, bottom) vacancy growth mode,  $V/V_t = 1.5$ .

$V/V_t$  and  $C_{im}^*/(C_{vm} - C_{im})$ . The profile of self-interstitials is labeled I, that of impurity interstitials  $I^*$ , and that of vacancies V. In the interstitial growth mode (Fig. 6a), the interstitial species are incorporated in some constant concentration, but the ratio of the two species is gradually changing, in favor of impurity interstitials. The  $T_k$  corresponds to the intersection point of the curves I and  $I^*$ . For the vacancy growth mode (Fig. 6b), both kinds of interstitials decay, due to recombination with vacancies, and some vacancy concentration is incorporated.

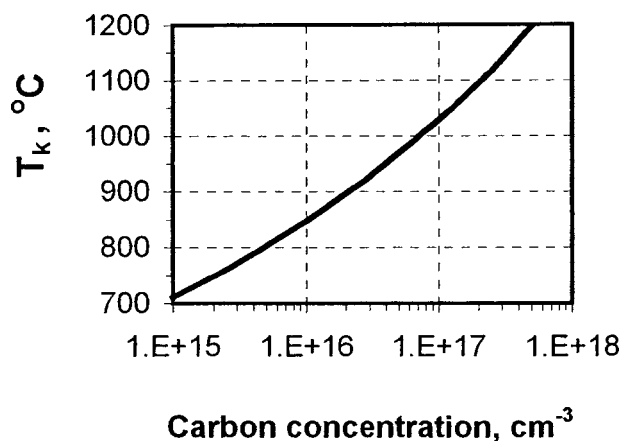
**Interstitial trapping by carbon.**—Carbon impurity was reported<sup>23</sup> to cause a change from vacancy growth mode to interstitial growth mode, at a doping level of  $1.7 \times 10^{17} \text{ cm}^{-3}$ . The  $V_t$  was intermediate between 3 and 4 mm/min in nondoped crystal, and increased to over 4 mm/min in carbon-doped crystal. This effect allows only for a rough estimate of the carbon-induced shift in  $\xi_t$  by about 20%. Equation 16 for a shift is reduced to the form similar to Eq. 7, only now the addition sign in the denominator is replaced by the subtraction sign

$$\xi_t = \xi_{to} / (1 - C/C_o) \quad [17]$$

where the characteristic constant  $C_o = (C_{vm} - C_{im})/R_i$ . The estimated shift corresponds to  $C_o \approx 10^{18} \text{ cm}^{-3}$ , and the interstitial carbon fraction at  $T_m$  is estimated as  $R_i = 3 \times 10^{-4}$ .

This important carbon parameter, in conjunction with some other reported data on carbon,<sup>44</sup> makes it possible for a comprehensive description of carbon behavior in a wide temperature range. First, we make use of the effective diffusivity of carbon,  $D(T)$ , and the





**Figure 7.** The kick-out temperature,  $T_k$ , for carbon impurity in silicon (below  $T_k$ , the carbon interstitials prevail over the self-interstitials).

solubility,  $C_{se}(T)$ , measured in a range of high  $T$  (the references are cited in the review of Ref. 44). The product  $DC_{se}$  turns out to be very small in comparison to the self-diffusivity product,  $D_i C_{ie}$ , which means that the carbon in-diffusion is limited by the supply of interstitial carbon, and not by out-diffusion of kicked-out self-interstitials. In such a case, the effective diffusivity is reduced to the diffusivity of the carbon interstitials,  $D_i^*$ , multiplied by the equilibrium fraction of these mobile species:  $D = D_i^* (C_i^*/C) = R_i D_i^*$ . The  $D(T)$  dependence is characterized by an activation energy of 3.1 eV; the diffusivity extrapolated to the melting point is  $1.1 \times 10^{-9}$  cm<sup>2</sup>/s. Since the value of  $R_i$  at  $T_m$  was just estimated, we get an estimate for the carbon interstitial diffusivity at  $T_m$ :  $D_i^* = 3.5 \times 10^{-6}$  cm<sup>2</sup>/s. The interstitial carbon diffusivity is known<sup>44</sup> at very low  $T$ , and thus  $D_i^*(T)$  can be approximated by an Arrhenius plot in the whole range from low  $T$  up to  $T_m$

$$D_i^* = 4.4 \times 10^{-4} \exp(-0.7 \text{ eV}/kT) \text{ cm}^2/\text{s} \quad [18]$$

The resulting migration energy, 0.7 eV, is slightly less than the value of 0.88 eV deduced from only low temperature data.<sup>44</sup> With known  $D(T)$ , we can now write down the equilibrium fraction of interstitial carbon,  $R_i(T)$

$$R_i(T) = 4.8 \times 10^3 \exp(-2.4 \text{ eV}/kT) \quad [19]$$

We are interested in the ratio of self- and carbon-interstitials,  $C_i^*/C_i$ . This ratio is expressed through  $R_i$ , from the mass-action law for the kick-out reaction

$$C_i C_i^* = K(T) = C_{ie}/R_i$$

and hence

$$C_i^*/C_i = R_i(T)C/C_{ie}(T) \quad [20]$$

The last of these expressions is just a form of Eq. 15, and the kick-out energy,  $E_k$  for carbon is now specified:  $E_k = E_i - 2.4 \text{ eV} = 1.6 \text{ eV}$ . The kick-out temperature  $T_k$  is defined by the Eq. 20 at  $C_i^*/C_i = 1$ . The dependence of  $T_k$  on the carbon concentration,  $C$ , is shown in Fig. 7.

It was found<sup>45</sup> that the nucleation temperature  $T_n$  of interstitial microdefects in FZ crystals strongly increased with increasing the carbon content. To understand this effect, it is important to know the ratio of self- and carbon-interstitials at the microdefect formation stage. Since the reported nucleation temperature<sup>45</sup> turns out to be well above  $T_k$ , the carbon interstitials are minor species, and their concentration  $C_i^*$  is defined by Eq. 20 to be proportional to the carbon content,  $C$  (and to the total concentration of interstitial species that is close to  $C_i$ ). The strong carbon effect on  $T_n$  can be

understood if the critical interstitial clusters are made not of the prevailing species (self-interstitials) but of minor species, carbon interstitials.

In modern CZ crystals the carbon content is low, about  $3 \times 10^{15} \text{ cm}^{-3}$ . The kick-out temperature, 770°C, is well below the nucleation temperature reported to be over 900°C.<sup>15</sup> Yet the fraction of carbon interstitials at  $T_n$  is still appreciable, 10% at 930°C. It is likely that the interstitial agglomeration in CZ crystals is also controlled by the carbon impurity.

The carbon-containing globular clusters may be identified with the so-called B-defects which are present along with the interstitial-type loops (A-defects) and also form a separate B-band at the edge of an interstitial region.<sup>3,4</sup> A possible interpretation is that the B-defects are the precursors of the A-defects, and that the  $B \rightarrow A$  conversion occurs if the B-defects are large enough for that.

### Conclusions

Vacancy trapping by impurities can affect both the stage of incorporation of intrinsic point defects into the growing crystal and the lower-temperature stage of vacancy agglomeration into voids.

1. At the incorporation stage, the critical  $V/G$  ratio (that separates interstitial and vacancy growth modes) is shifted downward in favor of vacancy incorporation. This effect is due to an increased total equilibrium concentration of the vacancy species (free and trapped vacancies) at the melting point. A shift in the critical  $V/G$  ratio is well described by a simple analytical expression. The shift is appreciable (more than 10%) if the equilibrium concentration of trapped vacancies is higher than about  $2 \times 10^{13} \text{ cm}^{-3}$ . Since oxygen (at a concentration of about  $8 \times 10^{17} \text{ cm}^{-3}$ ), nitrogen and hydrogen (at a concentration of about  $10^{15} \text{ cm}^{-3}$ ) induce an appreciable downward shift in the critical  $V/G$ , it is concluded that these impurities trap vacancies even at high  $T$  (near the melting point), and the equilibrium concentration of trapped vacancies is larger than the value indicated above. (i) For oxygen, the trapped vacancies are thought to be substitutional oxygen atoms (known as A-centers from the low temperature irradiation experiments). (ii) For nitrogen, there are several species that can be considered as trapped vacancies. However, rough estimates indicate that the prevailing trapped vacancy is just substitutional nitrogen atom: a complex VN of a vacancy and an interstitial nitrogen atom.

2. At the melting point, the free vacancies greatly exceed the trapped species in concentration, for the impurities considered here. However, on lowering the temperature, the fraction of trapped vacancies increases rapidly, and these species become dominant below some binding temperature,  $T_b$ . An enhanced trapping at low  $T$  will affect the void formation in the vacancy growth mode. If the void nucleation temperature,  $T_n$ , is greater than  $T_b$ , then vacancy trapping is not important for the void production. In the opposite case, when  $T_b$  is well over the temperature at which the voids would be formed without trapping, the free vacancies will already be strongly trapped before the void production can start. Here, void formation is completely suppressed, and the grown-in crystal will contain the trapped vacancies (point defects) instead of voids. For the intermediate case (a moderate trapping at  $T_n$ ) the void properties will be modified in an essential way: the nucleation temperature  $T_n$  decreases, due to a decreased concentration of free vacancies, and the effective mobility of the vacancy population is reduced which results in a higher void density. All three of these possibilities seem to occur commonly.

1. For oxygen,  $T_b$  is roughly estimated to be about 1060°C. Voids in Czochralski (CZ) silicon are known to form at a higher  $T$  (at  $T_n$  typically close to 1100°C). Vacancy trapping by oxygen is thus not very important in void production. The trapping of vacancies by oxygen is nonetheless a very important effect in CZ silicon growth; due to this effect, some vacancies escape consumption by voids. These residual vacancies control subsequent oxygen precipitation, giving rise to a complicated banded pattern and the so-called anomalous oxygen precipitation.<sup>29</sup>

2. For nitrogen, the estimated binding temperature is relatively high, around 1115°C. In CZ silicon, it corresponds to the intermediate trapping case and therefore to a decreased void formation temperature and to an increased void density.

3. For nitrogen in FZ silicon the void production temperature (in material not doped with nitrogen) is expected to be considerably lower than that in CZ crystal, due to the effect of oxygen on void production. This case corresponds to a strong vacancy trapping and suppression of the void formation (provided the amount of nitrogen is enough to trap all the vacancies). In this case, the as-grown material is expected to contain the VN species instead of voids.

The available data on the above impurities are qualitatively consistent with this general concept of vacancy trapping. The list of impurities that trap vacancies is likely to include more examples. For instance, the donor dopants may trap the vacancies due to Coulomb attraction. There are some indications for this effect<sup>23,46</sup> but quantitative data are still lacking.

Some impurities can trap self-interstitials, which results in similar effects. Particularly, for a substitutional impurity with a minor interstitial component, the trapped interstitial is identical to the interstitial impurity atom. The presence of trapped (impurity) interstitials, along with self-interstitials, can affect both the incorporation and agglomeration stages of microdefect formation. At the incorporation stage, the critical  $V/G$  ratio is shifted upward, in favor of interstitial incorporation. On lowering  $T$ , the self-interstitials are gradually replaced by impurity interstitials; the latter species are dominant below some characteristic  $T_k$ . An upward shift in the critical  $V/G$  was reported for the carbon impurity, and  $T_k$  was estimated in dependence of the carbon content. The interstitial agglomeration occurs at a temperature higher than  $T_k$ , and yet it is affected by the presence of carbon, presumably due to the involvement of mobile carbon interstitials in formation of critical interstitial clusters.

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