

# Enhanced oxygen diffusion in highly doped *p*-type Czochralski silicon

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The locking of dislocations by oxygen has been investigated experimentally in Czochralski silicon (Cz-Si) with different concentrations of shallow dopants. Specimens containing well-defined arrays of dislocation half-loops were subjected to isothermal anneals in the 350–550 °C temperature range, and the stress required to bring about dislocation motion at 550 °C was then measured. This dislocation unlocking stress was found to increase with annealing time due to oxygen diffusion to the dislocation core. The dislocation unlocking stress was measured in *n*-type Cz-Si with a high antimony doping level ( $\sim 3.4 \times 10^{18} \text{ cm}^{-3}$ ) and *p*-type Cz-Si with a low boron doping level ( $\sim 1.3 \times 10^{15} \text{ cm}^{-3}$ ). An analysis of the data taking the different oxygen concentrations into account showed that the rate of increase in dislocation unlocking stress was unaffected by the high level of antimony doping. This indicates that a high antimony doping level has no significant effect on oxygen transport for the conditions used in this experiment. However, in *p*-type Cz-Si with a high boron doping level ( $\sim 5.4 \times 10^{18} \text{ cm}^{-3}$ ), the dislocation unlocking stress was found to rise at a much faster rate than in Cz-Si with a low boron doping level or high antimony doping level. This enhancement in dislocation locking was by a factor of approximately 60 at 400 °C. By performing a numerical simulation to solve the diffusion equation for oxygen transport to a dislocation, the effective diffusivity of oxygen was deduced from the dislocation unlocking data to be  $2.7 \times 10^{-6} \exp(-1.4 \text{ eV}/kT) \text{ cm}^2 \text{ s}^{-1}$  in the highly boron doped Cz-Si. In the temperature range studied, the effective diffusion coefficient in the highly boron doped Cz-Si was found to be approximately 44 times higher than expected in low boron doped Cz-Si with an identical oxygen concentration.

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## INTRODUCTION

Czochralski silicon (Cz-Si) contains interstitial oxygen in a concentration of approximately  $10^{17}$ – $10^{18} \text{ cm}^{-3}$ , which is present as a consequence of the dissolution of the silica crucible that contains the melt. The presence of oxygen in silicon is both beneficial and detrimental from the point of view of the production of integrated circuits. Oxygen precipitates, which are formed by appropriate heat treatments, can act as gettering centers for unwanted metallic impurities.<sup>1</sup> Additionally, oxygen can improve the mechanical strength of wafers as it is known to immobilize dislocations that may be introduced during device processing.<sup>2–4</sup> However, excess precipitation can lead to dislocation generation, which adversely affects the wafer's mechanical stability (for a general discussion see Ref. 5).

For the benefits of oxygen in silicon to be fully exploited, a complete understanding of its transport is essential. The transport of oxygen in Cz-Si with a typical concentration of shallow dopants in the high temperature (700–1200 °C) range is well understood, and it is known that the diffusivity of oxygen  $D$  in this temperature range is described by  $D = 0.13 \exp(-2.53 \text{ eV}/kT) \text{ cm}^2 \text{ s}^{-1}$ .<sup>6</sup> In the low temperature range (350–600 °C) there is evidence that oxygen transport is strongly enhanced.<sup>7–9</sup> The difference in oxygen transport

between the temperature ranges is attributed to the different species that are thought to dominate transport in these ranges. At high temperatures the oxygen monomer is responsible for transport; at low temperatures the oxygen dimer is thought to be principally responsible,<sup>10</sup> and experimental evidence for this has been provided by a dislocation locking technique.<sup>9,11</sup> Because monomers and dimers are responsible for transport it is often helpful to describe oxygen transport in terms of an effective diffusivity, such that the flux of atoms  $J$  is given by  $J = -D_{\text{eff}}(dC/dx)$ , where

$$D_{\text{eff}} = \frac{C_O D_O + 2C_{O_2} D_{O_2}}{C_O + 2C_{O_2}} \quad (1)$$

and  $C$ ,  $C_O$ , and  $C_{O_2}$  are the total concentration of oxygen and the concentrations of the oxygen monomer and dimer, respectively, and  $D_O$  and  $D_{O_2}$  are the diffusivities of the oxygen monomer and dimer, respectively. Since  $C_O \gg C_{O_2}$  this becomes

$$D_{\text{eff}} \approx D_O + 2 \frac{C_{O_2}}{C_O} D_{O_2}. \quad (2)$$

The effect of the electrical doping level on oxygen transport in as-grown Cz-Si has been studied experimentally by secondary-ion mass spectrometry (SIMS).<sup>12–14</sup> The experimental results on the effect of high concentrations of antimony, boron, and arsenic on oxygen transport in Cz-Si are

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TABLE I. The effect of high concentrations of various shallow dopants on the transport of oxygen in Cz-Si.

Dopant	Concentration (cm <sup>-3</sup> )	Temperature range (°C)	Effect on oxygen transport	Reference(s)
Sb	$2.3 \times 10^{17}$ – $2.1 \times 10^{18}$	950–1100	None	12 and 13
Sb	$5 \times 10^{17}$	800	None	13
Sb	$1.3 \times 10^{18}$	500–800	Slows ( $E_A$ increased by 1.40 eV)	14
B	$8 \times 10^{18}$ – $1.2 \times 10^{19}$	1050	None	13
B	$8 \times 10^{18}$ – $1.2 \times 10^{19}$	800	Slows (factor of $\sim 5$ )	13
B	$4.4 \times 10^{18}$	500 and 800	None	14
B	$4.4 \times 10^{18}$	600–700	Slows (factor of $\sim 2$ –3)	14
As	$1.1 \times 10^{19}$	1050	None	13
As	$1.1 \times 10^{19}$	800	Slows (factor of $\sim 2.5$ )	13
As	$4 \times 10^{18}$ – $1.3 \times 10^{19}$	500–800	Slows ( $E_A$ increased by 0.64–0.68 eV)	14

summarized in Table I. High concentrations of all three elements have been found to provide a retardation effect on oxygen transport at 800 °C and below,<sup>13,14</sup> although the studies disagree on the temperatures at which the retardation takes place. In the study of Takeno *et al.* a high boron doping level was found to slow oxygen transport at 600 and 700 °C.<sup>14</sup> Since in the same work it was also found that oxygen transport is not affected by a high boron doping level at 500 and 800 °C, the authors attribute the retardation to the formation of immobile oxygen complexes. None of these elements has previously been shown directly to provide an enhancement in oxygen transport in as-grown material. However, there is also indirect evidence for the transport of oxygen being enhanced by a high boron concentration. One such piece of evidence is that oxygen precipitation, a process which is limited by oxygen transport, has been found to be enhanced in Cz-Si with a high concentration of boron.<sup>15</sup> Another is that oxygen transport, measured by SIMS, has been found to be enhanced in material into which boron had been diffused.<sup>16</sup> Density functional theory calculations indicate that the activation energy for diffusion of an oxygen dimer should be lower in highly doped *p*-type Cz-Si than in silicon with a low doping level.<sup>17</sup> The calculations predict an activation energy for oxygen diffusion in low doped Cz-Si to be 1.33 eV and in highly doped *p*-type Cz-Si to be 0.86 eV.

Dislocations in materials can be pinned by impurity atoms.<sup>2–4,9,11,18–21</sup> The critical resolved shear stress necessary to unpin a dislocation from the locking impurity is known as the *unlocking stress*. By studying the unlocking stress as a function of annealing time and temperature it is possible to deduce information on impurity transport and impurity-dislocation interactions.<sup>4,9,11,18</sup> This dislocation locking technique has previously been used to investigate the transport of oxygen in Cz-Si with a shallow dopant concentration of approximately  $10^{15}$  cm<sup>-3</sup> (Ref. 4, 9, and 11) and nitrogen in float-zone silicon.<sup>18</sup> In Cz-Si the technique reproduced existing values of the diffusivity of oxygen in silicon in the 700–850 °C temperature range<sup>4</sup> and provided evidence for the aforementioned enhanced oxygen transport in the 350–600 °C temperature range.<sup>9</sup> Furthermore, the oxygen concentration dependence of the enhanced oxygen diffusion provided experimental evidence that the enhanced transport is due to oxygen dimers.<sup>9,11</sup>

In this work, the locking of dislocations by oxygen in Cz-Si with different shallow dopant concentrations has been studied in the 350–550 °C temperature range. Since at these temperatures the diffusion of oxygen is likely to occur much faster than the diffusion of the shallow dopants in the Cz-Si investigated,<sup>22</sup> information on the electrical doping dependence of oxygen transport is deduced by analyzing the dislocation locking process in terms of oxygen transport to dislocations.

## EXPERIMENTAL METHOD

Specimens with approximate dimensions of  $0.65 \times 4 \times 30$  mm<sup>3</sup> were cleaved from (100) Cz-Si wafers and the cleaved edges were mechanically polished. The dopant and oxygen concentrations of the specimens used are summarized in Table II. Oxygen concentrations were determined by gas fusion analysis for the highly doped specimens and by Fourier transform infrared spectroscopy (FTIR) for the low doped specimens. Oxygen concentrations are stated to the DIN 50438/I standard.

A controlled set of dislocation half-loops with a diameter of approximately 200  $\mu$ m was introduced into each specimen. This was done by indentation at uniform intervals along the length of the bar using a Vickers diamond tip (0.1 N load, 5 s dwell time) followed by a four-point bend at approximately 600 °C. Specimens made from the highly doped material were cooled to below 400 °C with the load still applied.

Specimens were then annealed in argon atmosphere at constant temperatures (from 350 to 550 °C) for different times (up to approximately 1600 h) to allow oxygen to diffuse to the dislocations. After the removal of a 30  $\mu$ m sur-

TABLE II. Dopant and oxygen concentrations of the Cz-Si used in this study.

Specimen	Dopant	Dopant concentration (cm <sup>-3</sup> )	Oxygen concentration (cm <sup>-3</sup> )
$n^+$	Sb	$(5.6\text{--}63) \times 10^{17}$	$1.06 \times 10^{18}$
$p^+$	B	$(2.7\text{--}8.0) \times 10^{18}$	$7.5 \times 10^{17}$
$p$	B	$1.3 \times 10^{15}$	$6.3 \times 10^{17}$

face layer by a chemical etch comprising HF (40%), HNO<sub>3</sub> (69%), and CH<sub>3</sub>COOH (glacial) in the ratio of 8:75:17, the specimens were subjected to a three-point bend at 550 °C. In a three-point bending configuration the stress varies approximately linearly from the outer to the central knife edges. This ensures that each set of dislocation half-loops is subjected to a different stress. After a defect-revealing etch, optical microscopy was used to find the set of dislocation half-loops that moved under the least resolved shear stress, i.e., the unlocking stress. It should be noted that a standard defect revealing etch, comprising CrO<sub>3</sub> (0.3M) and HF (40%) in the ratio of 5:4, was unsuitable as it stained the surface of the highly doped material. Therefore an etch comprising HF (40%), HNO<sub>3</sub> (69%), and CH<sub>3</sub>COOH (glacial) mixed in the ratio of 1:3:10 was instead used to reveal the dislocations.

In addition to the intentional annealing period used to decorate the dislocations with oxygen, the experimental technique also involves the dislocations being subjected to short periods at high temperatures while the experimental rig is cooling after the four-point bend stage and warming before the three-point bend process. During this time some diffusion of oxygen to the dislocation cores takes place and this is in addition to that which occurs during the intended anneal stage. The amount of oxygen accumulated during these steps is always small. However, for some specimens the effect is significant and a small correction is made to the measured unlocking stress.<sup>4</sup> The data are presented here after this correction.

## RESULTS

The dislocation unlocking stress as a function of annealing time at temperatures from 450 to 550 °C for low boron doped *p*-type Cz-Si with an oxygen concentration of  $6.3 \times 10^{17} \text{ cm}^{-3}$  and highly antimony doped *n*-type Cz-Si with an oxygen concentration of  $1.06 \times 10^{18} \text{ cm}^{-3}$  is shown in Fig. 1. For both sets of specimens, the increase in dislocation unlocking stress increases approximately linearly with annealing time. The gradient of this initial linear rise is on average 3.0 times higher in the case of the highly antimony doped material. As described later, this higher gradient is attributed to the higher oxygen concentration in the highly antimony doped *n*-type Cz-Si.

The dislocation unlocking stress as a function of annealing time at temperatures from 350 to 550 °C for highly boron doped *p*-type Cz-Si with an oxygen concentration of  $7.5 \times 10^{17} \text{ cm}^{-3}$  and low boron doped *p*-type Cz-Si with an oxygen concentration of  $6.3 \times 10^{17} \text{ cm}^{-3}$  is shown in Fig. 2. Again, the dislocation locking stress can be seen to increase approximately linearly with annealing time. The dislocation locking occurs at a much faster rate in the highly boron doped material. For instance, at 400 °C, the rate of increase in dislocation locking occurs approximately 60 times faster in the highly boron doped Cz-Si compared to low boron doped Cz-Si. The magnitude of the enhancement of the rate of dislocation locking cannot be explained by the slightly higher oxygen concentration, and it is believed that high boron doping enhances oxygen diffusion, as discussed later in this paper.

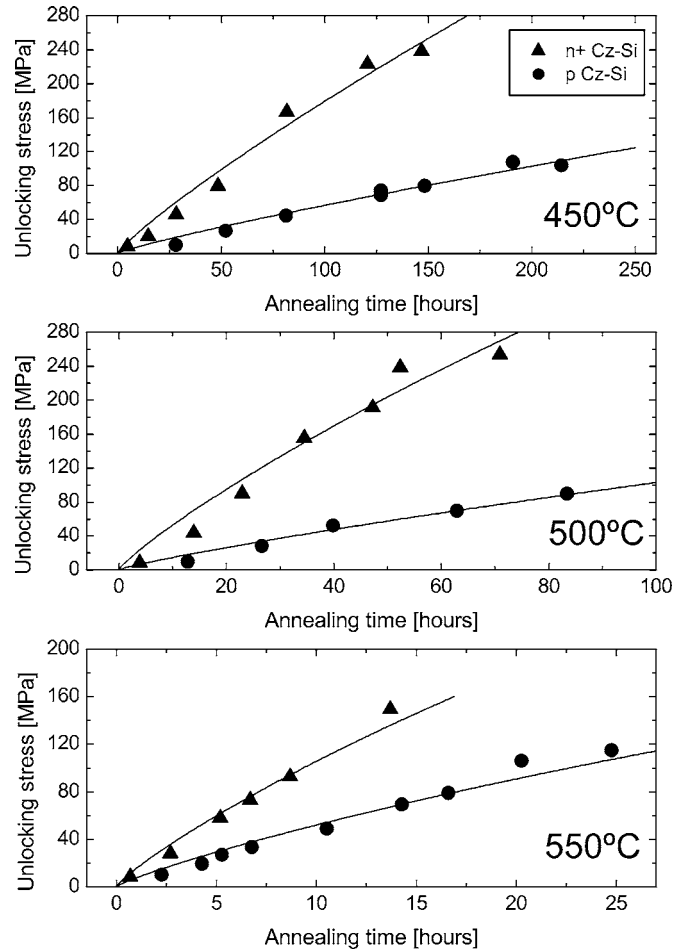


FIG. 1. Dislocation locking as a function of annealing time at different temperatures for *n*-type Cz-Si with a high antimony doping level with an oxygen concentration of  $1.06 \times 10^{18} \text{ cm}^{-3}$  (triangles) and for *p*-type Cz-Si with a low boron doping level with an oxygen concentration of  $6.3 \times 10^{17} \text{ cm}^{-3}$  (circles). The curves are from a numerical simulation of diffusion to a dislocation.

## MODELING

The dislocation locking process occurs due to oxygen transport to the dislocation core. Thus, from analyzing the dislocation unlocking stress as a function of annealing time and temperature, it is possible to deduce information on the transport of oxygen. The numerical simulation used to model oxygen diffusion to a dislocation in this present work is described in previous publications.<sup>4,9,11</sup> The value of the effective diffusion coefficient  $D_{\text{eff}}$  is chosen so that the numerical simulation fits the unlocking stress data presented in Figs. 1 and 2. For each set of experimental data presented, the unlocking stress as a function of annealing time given by the numerical simulation is shown as a curve on the figures.

In this paper, the following fitting parameters are used to simulate all the data obtained for all the different types of Cz-Si investigated: the entropy was taken to be  $S = 7.6 \times 10^{-5} \text{ eV K}^{-1}$ , the number of states at the dislocation core was taken to be  $C_a = 5 \times 10^{22} \text{ cm}^{-3}$ , the binding energy of the oxygen to the dislocation was taken to be  $\Delta G = 0.74 \text{ eV}$ ,<sup>4</sup> and the core radius was taken to be  $r_0 = 5 \text{ Å}$ .<sup>23</sup> Using the same values for all the conditions investigated allows the values of  $D_{\text{eff}}$  obtained in the different materials to be compared di-

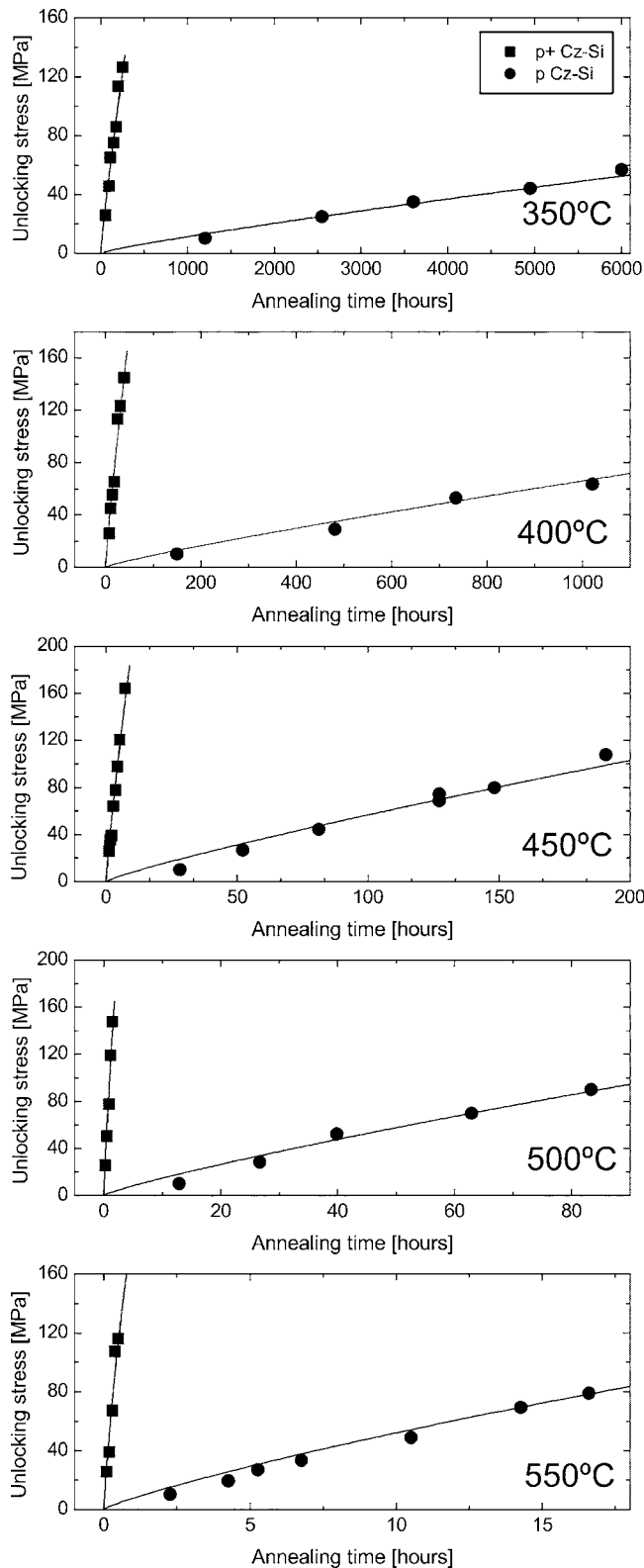


FIG. 2. Dislocation locking as a function of annealing time at different temperatures for *p*-type Cz-Si with a high boron doping level with an oxygen concentration of  $7.5 \times 10^{17} \text{ cm}^{-3}$  (squares) and a low boron doping level with an oxygen concentration of  $6.3 \times 10^{17} \text{ cm}^{-3}$  (circles). The curves are from a numerical simulation of diffusion to a dislocation.

rectly. This is a different approach to that taken in previous work using the same technique, where a small range of values were used for the entropy, number of states at the dislocation core, and the binding energy to provide the best over-

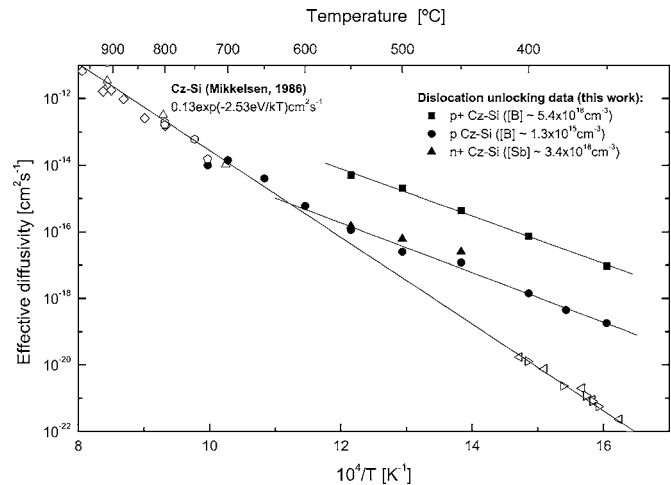


FIG. 3. Effective diffusivity of oxygen in silicon as a function of temperature. Data from the literature for Cz-Si with a low concentration of shallow dopants are denoted by open symbols (Ref. 6). Data from dislocation locking experiments are denoted by closed symbols. Closed circles are for low boron doped Cz-Si with an oxygen concentration of approximately  $7.5 \times 10^{17} \text{ cm}^{-3}$ , closed squares are for high boron doped Cz-Si with an oxygen concentration of approximately  $6.3 \times 10^{17} \text{ cm}^{-3}$ , and closed triangles are for high antimony doped Cz-Si with an oxygen concentration of  $1.06 \times 10^{18} \text{ cm}^{-3}$ .

all fit to a particular set of experimental data.<sup>4,9,11</sup> Therefore, the absolute values of the effective diffusivity presented in this work for the Cz-Si with the low boron concentration differ slightly from those presented in previous publications for which different fitting parameters were used. By fitting the simulation to the experimental data points, a value of  $D_{\text{eff}}$  is then determined at each temperature for each type of material. It is these values of  $D_{\text{eff}}$  which are taken as the measured value of effective oxygen diffusivity for the conditions studied.

The effective diffusivity of oxygen as deduced from dislocation unlocking experiments is plotted in Fig. 3 for the Cz-Si with a high antimony doping level, Cz-Si with a low boron doping level, and Cz-Si with a high boron doping level. Also shown in the figure are data from the literature obtained by a variety of other techniques.<sup>6</sup> A best fit of the effective diffusion coefficients plotted in Fig. 3 gives the following expression for the effective diffusivity of oxygen in *p*-type Cz-Si with a low boron doping level:

$$D_p = 1.6 \times 10^{-7} \exp\left(-\frac{1.5 \text{ eV}}{kT}\right) \text{ cm}^2 \text{ s}^{-1} \quad (3)$$

in the 350–600 °C temperature range. The effective diffusivity of oxygen in *p*-type Cz-Si with the high boron doping level is given by the following expression:

$$D_{p+} = 2.7 \times 10^{-6} \exp\left(-\frac{1.4 \text{ eV}}{kT}\right) \text{ cm}^2 \text{ s}^{-1} \quad (4)$$

in the 350–550 °C temperature range.

## DISCUSSION

The data presented in Fig. 1 show that in the 450–550 °C temperature range, oxygen is transported to dislocations approximately three times faster in the case of the



highly antimony doped *n*-type Cz-Si with an oxygen concentration of  $1.06 \times 10^{18} \text{ cm}^{-3}$  than in the case of the low boron doped *p*-type Cz-Si with an oxygen concentration of  $6.3 \times 10^{17} \text{ cm}^{-3}$ . If it is assumed that the oxygen dimer is responsible for dislocation locking at the temperatures investigated, which is suggested by previous work,<sup>9,11</sup> then it can be shown that the effective diffusivity of oxygen depends linearly on the oxygen concentration and hence the gradient of the initial rise in dislocation unlocking stress depends on the square of the oxygen concentration in the specimens.<sup>11</sup> The average ratio of the gradient of the initial rises in dislocation locking stress between high antimony doped material and low boron doped material is approximately 3.0, and the square of the ratio of the oxygen concentrations is  $(10.6/6.3)^2 \approx 2.8$ . Within the experimental errors of the technique it is concluded that the difference in oxygen concentrations between the different materials accounts for the higher rate of dislocation locking in the highly antimony doped material and that a high concentration of antimony makes no detectable difference to the transport of oxygen. This effect can also be seen in the data plotted in Fig. 3, which show that the effective diffusivity of oxygen in the *n*-type Cz-Si with the high antimony doping level is slightly higher than that in the *p*-type Cz-Si with the low boron doping level. This small variation is due to the dependence of the effective diffusivity on oxygen concentration, as described above.

The data presented in Fig. 2 indicate that the oxygen transport to dislocations in the 350–550 °C temperature range occurs much faster in the high boron doped *p*-type Cz-Si with an oxygen concentration of  $7.5 \times 10^{17} \text{ cm}^{-3}$  than the low boron doped *p*-type Cz-Si with an oxygen concentration of  $6.3 \times 10^{17} \text{ cm}^{-3}$ . This cannot be explained by a difference in oxygen concentration between the specimens. Assuming that the oxygen dimer is responsible for transport gives an expected enhancement in the gradient of the initial rise in unlocking stress due to the oxygen concentration difference of  $(7.5/6.3)^2 \approx 1.4$ , whereas the transport of oxygen is found to be enhanced by a factor of approximately 60 at 400 °C. The effective oxygen diffusivity plotted in Fig. 3 is significantly higher in *p*-type Cz-Si with the high boron doping level compared to *p*-type Cz-Si with the low boron doping level at all temperatures studied. It is concluded that high boron doping strongly enhances oxygen transport in the 350–550 °C temperature range. In this temperature range, the effective diffusion coefficient in the highly boron doped Cz-Si was found to be approximately 44 times higher than expected in low boron doped Cz-Si with an identical oxygen concentration.

In this work it has been shown that oxygen transport in Cz-Si is not affected significantly by high antimony doping but is strongly enhanced by high boron doping. The measured activation energies for oxygen transport in low boron doped *p*-type Cz-Si and high boron doped *p*-type Cz-Si, given in Eqs. (3) and (4), respectively, are the same within experimental error. However, the experimentally measured effective diffusion coefficients vary significantly in their prefactors.

Both crystal strain effects and electronic effects should be considered with regards to the mechanism of enhanced oxygen transport in highly boron doped Cz-Si. A high boron concentration generally causes the contraction of the silicon lattice<sup>24</sup> together with a short range distortion of the lattice around individual boron atoms, and this could affect the diffusion of oxygen. However, a high level of antimony doping would also introduce a lattice strain (in this case tensile) and the results presented in this work show that doping Cz-Si with a high concentration of antimony does not affect oxygen transport. Thus, it seems unlikely that lattice strain by itself is responsible for the enhancement in oxygen diffusion observed in highly boron doped material. While there is no direct evidence that electronic effects are responsible for the enhanced oxygen transport observed in highly boron doped Cz-Si, this remains a possible explanation since for the doping concentrations used the material remains extrinsic over the entire temperature range studied.

The effect of Fermi level position on oxygen diffusion has been considered by Adey *et al.*<sup>17</sup> They developed a theoretical model which suggests that the activation energy for oxygen dimer diffusion will be lower in highly doped *p*-type material than in *p*-type Cz-Si with a low concentration of shallow dopants. They suggested, that the oxygen dimer in highly doped *p*-type Cz-Si exists in a double positive charge state, which diffuses by alternating between square and staggered forms with an activation energy of 0.86 eV. However, in material with a low concentration of shallow dopants the model suggests that the uncharged oxygen dimer is most stable and that it diffuses with an activation energy of 1.33 eV. The values of activation energy found in this work given in Eqs. (3) and (4) represent those for oxygen transport and not diffusion. If the binding energy of the oxygen dimer is known then the activation energy for oxygen dimer diffusion is the sum of the values found for transport and the binding energy.<sup>9</sup> The binding energy of the oxygen dimer in low doped Cz-Si has been measured to be approximately 0.3 eV by Murin *et al.*,<sup>25</sup> but has not been measured in highly doped Cz-Si. Thus, using the value of Murin *et al.* for low doped Cz-Si and the activation energy in Eq. (3) puts the oxygen dimer diffusion activation energy at approximately 1.8 eV. Thus, even if the dimer binding energy in highly doped Cz-Si is significantly smaller than 0.3 eV, the results presented in this paper are not consistent with a 0.47 eV reduction in the activation energy for diffusion between the highly doped and low doped *p*-type Cz-Si proposed by Adey *et al.*

Sueoka *et al.* found that a high boron concentration led to enhanced oxygen precipitation in the 700–1000 °C temperature range.<sup>15</sup> They suggested, that this is due to enhanced oxygen transport, which has been found in this present work. However, other workers<sup>13,14</sup> have found that the effective diffusivity of oxygen is unchanged or actually reduced by high boron doping (see Table I). In this work it has been found that a high antimony concentration in Cz-Si has little effect on oxygen transport in the 450–550 °C temperature range. This contradicts the results of Takeno *et al.* who found

that the activation energy for oxygen transport in such material is increased by 1.40 eV in the 500–800 °C temperature range.<sup>14</sup>

## CONCLUSIONS

The locking of dislocations by oxygen in Cz-Si with different concentrations of shallow dopants has been investigated experimentally. In the 450–550 °C temperature range, the gradient of the rise in dislocation unlocking stress was found to be approximately a factor of 3 higher in highly antimony doped *n*-type Cz-Si with an oxygen concentration of  $1.06 \times 10^{18} \text{ cm}^{-3}$  than in low boron doped *p*-type Cz-Si with an oxygen concentration of  $6.3 \times 10^{17} \text{ cm}^{-3}$ . This increase in the rate of dislocation locking is due to the higher oxygen concentration in the highly antimony doped Cz-Si. It is therefore concluded that doping Cz-Si with antimony in levels of approximately  $3.4 \times 10^{18} \text{ cm}^{-3}$  makes no difference to oxygen transport in the temperature range investigated.

In the 350–550 °C temperature range, the gradient of the rise in the dislocation unlocking stress was found to be much higher in highly boron doped *p*-type Cz-Si with an oxygen concentration of  $7.5 \times 10^{17} \text{ cm}^{-3}$  compared to that in low boron doped *p*-type Cz-Si with an oxygen concentration of  $6.3 \times 10^{17} \text{ cm}^{-3}$ . The rate of increase in dislocation unlocking stress at 400 °C was enhanced by a factor of approximately 60, which cannot be explained by the higher oxygen concentration in the highly doped Cz-Si. It is concluded that boron doping in levels of approximately  $5.4 \times 10^{18} \text{ cm}^{-3}$  strongly enhances the transport of oxygen in Cz-Si. From numerical solving the diffusion equation for oxygen transport to a dislocation, the effective diffusivity of oxygen in highly boron doped *p*-type Cz-Si is deduced to be  $2.7 \times 10^{-6} \exp(-1.4 \text{ eV}/kT) \text{ cm}^2 \text{ s}^{-1}$ . The effective diffusion coefficient in the highly boron doped Cz-Si was found, on average, to be approximately 44 times higher than expected in low boron doped Cz-Si with an identical oxygen concentration.

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- <sup>1</sup>D. Gilles, E. R. Weber, and S. Hahn, *Phys. Rev. Lett.* **64**, 196 (1990).
- <sup>2</sup>S. M. Hu, *Appl. Phys. Lett.* **31**, 53 (1977).
- <sup>3</sup>K. Sumino and I. Yonenaga, in *Semiconductors and Semimetals*, edited by F. Shimura (Academic, New York, 1994), Vol. 42, p. 449.
- <sup>4</sup>S. Senkader, K. Jurkschat, D. Gambaro, R. J. Falster, and P. R. Wilshaw, *Philos. Mag. A* **81**, 759 (2001).
- <sup>5</sup>A. Borghesi, B. Pivac, A. Sassella, and A. Stella, *J. Appl. Phys.* **77**, 4169 (1995).
- <sup>6</sup>J. C. Mikkelsen, Jr., *Mater. Res. Soc. Symp. Proc.* **59**, 19 (1986).
- <sup>7</sup>S.-T. Lee, P. Fellingner, and S. Chen, *J. Appl. Phys.* **63**, 1924 (1988).
- <sup>8</sup>S. A. McQuaid, B. K. Johnson, D. Gambaro, R. Falster, M. J. Ashwin, and J. H. Tucker, *J. Appl. Phys.* **86**, 1878 (1999).
- <sup>9</sup>S. Senkader, P. R. Wilshaw, and R. J. Falster, *J. Appl. Phys.* **89**, 4803 (2001).
- <sup>10</sup>U. Gösele and T. Y. Tan, *Appl. Phys. A: Solids Surf.* **28**, 79 (1982).
- <sup>11</sup>A. Giannattasio, J. D. Murphy, S. Senkader, R. J. Falster, and P. R. Wilshaw, *J. Electrochem. Soc.* **152**, G460 (2005).
- <sup>12</sup>M. Pagani, *J. Appl. Phys.* **68**, 3726 (1990).
- <sup>13</sup>T. Ono, G. A. Rozgonyi, E. Asayama, H. Horie, H. Tsuya, and K. Sueoka, *Appl. Phys. Lett.* **74**, 3648 (1999).
- <sup>14</sup>H. Takeno, K. Sunakawa, and M. Suezawa, *Appl. Phys. Lett.* **77**, 376 (2000).
- <sup>15</sup>K. Sueoka, M. Akatsuka, M. Yonemura, T. Ono, E. Asayama, and H. Katahama, *J. Electrochem. Soc.* **147**, 756 (2000).
- <sup>16</sup>D. Heck, R. E. Tressler, and J. Monkowski, *J. Appl. Phys.* **54**, 5739 (1983).
- <sup>17</sup>J. Adey, R. Jones, D. W. Palmer, P. R. Briddon, and S. Öberg, *Phys. Rev. Lett.* **93**, 055504 (2004).
- <sup>18</sup>J. D. Murphy, C. R. Alpass, A. Giannattasio, S. Senkader, R. J. Falster, and P. R. Wilshaw, *Nucl. Instrum. Methods Phys. Res. B* **253**, 113 (2006).
- <sup>19</sup>K. Sumino and M. Imai, *Philos. Mag. A* **47**, 753 (1983).
- <sup>20</sup>I. Yonenaga, K. Sumino, and K. Hoshi, *J. Appl. Phys.* **56**, 2346 (1984).
- <sup>21</sup>V. I. Orlov, H. Richter, A. Fischer, J. Reif, T. Müller, and R. Wahlich, *Mater. Sci. Semicond. Process.* **5**, 403 (2002).
- <sup>22</sup>P. M. Fahey, P. B. Griffin, and J. D. Plummer, *Rev. Mod. Phys.* **61**, 289 (1989).
- <sup>23</sup>A. S. Nandedkar and J. Narayan, *Philos. Mag. A* **56**, 625 (1987).
- <sup>24</sup>J. Kucytowski and K. Wokulska, *Cryst. Res. Technol.* **40**, 424 (2005).
- <sup>25</sup>L. I. Murin, T. Hallberg, V. P. Markevich, and J. L. Lindström, *Phys. Rev. Lett.* **80**, 93 (1998).