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Electronic effects on dislocation velocities in heavily doped silicon

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The electronic environment of the crystal markedly influences the velocity of dislocations in semiconductors. *n*-doping raises dislocation velocities in silicon over the undoped values. The behavior with *p*-doping is anomalous. For $T > 600°C$ *p*-doping does not change dislocation velocities. For $T < 600°C$ *p*-doping also increases the dislocation velocity, as found by Erofeev *et al.* The velocities of screw dislocations in *p*- and *n*-doped silicon have been investigated as a function of impurity concentration and temperature. The data can be explained using a model proposed by Frisch and Patel modified for Fermi-Dirac statistics, which assumes that the change in the dislocation velocity is proportional to the number of charged dislocation sites. In *n*-type silicon the dislocations behave as acceptors. A donor-like behavior is obtained for *p*-type silicon at $T < 600°C$, with levels in both types near the gap center. The observed change in the activation energy of dislocations between heavily doped and undoped silicon can be readily calculated from the change in the Fermi level between the doped and undoped crystal with dislocations.

I. INTRODUCTION

In previous work on heavily doped germanium it has been shown[1] that the velocity of dislocations increases with *n* doping and decreases with *p* doping with respect to the values for intrinsic crystals. Neutral impurities such as Sn and Si showed no effects on dislocation velocities and the doping effects could only be observed provided the concentration of impurities was higher than the intrinsic carrier concentration for the particular test temperature. Typically, for germanium, dopant concentrations exceeding 10^{18} cm^{-3} are necessary at a test temperature of 500 °C. For silicon[2] we have reported somewhat similar observations: *n*-type impurities raise the dislocation velocity over the undoped value. For temperatures ≥ 600 °C we have also reported no change in dislocation velocity[3] with *p*-type impurities. Subsequently, Erofeev, Nikitenko, and Osvenskii[4] showed that at a still lower temperature of 450 °C *p*-type impurities raise the dislocation velocity in silicon. In this study we wish to describe a more detailed examination of the role of *n* and *p* impurities on the velocity of screw dislocations in silicon over a range of temperatures 450–900 °C. The data are analyzed along the lines of a model developed by Frisch and Patel[5] modified for Fermi-Dirac statistics, which allows us to estimate more conveniently the nature and position of the dislocation energy level in silicon.

II. EXPERIMENTAL

A. Specimens and technique

Only dislocation-free crystals of silicon were used in our study even for impurity concentrations approaching 10^{20} cm^{-3}. Bars $25 \times 3 \times 2$ mm were cut with orientation shown in Fig. 1. A scratch was made on the wide face with a diamond point loaded to 30 g. The specimens were deformed in four point bending in a molybdenum jig heated by rf to the desired temperature. Dislocations were detected by x-ray topography and the distance dislocations had moved from the scratch during the time of loading gave the dislocation velocity. As expected, uniform motion was always observed between the loading points as shown in Fig. 2.

B. Comparison of present technique with other work

There have been some questions[4] about the comparison of velocity measurements based on dislocation motion from a scratch and the motion of individual loops. Previously we have shown that for carefully prepared specimens of germanium[6] the velocity results are identical for motion from a scratch or individual loops. To compare our technique with that of other workers[4] on mixed or 60° dislocations in silicon, we have measured 60° dislocation velocities on specimens close to the 123 direction by two methods (a) etch pitting and

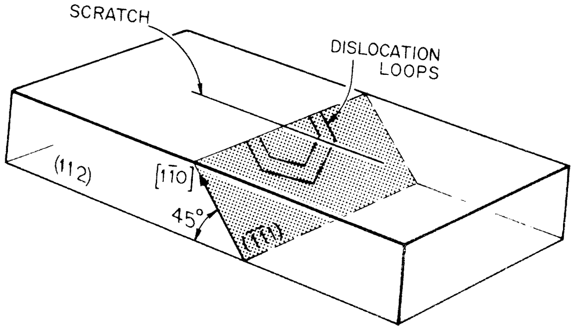

FIG. 1. Geometry of silicon crystals and glide system for four point bending experiments, showing loops generated from a scratch.

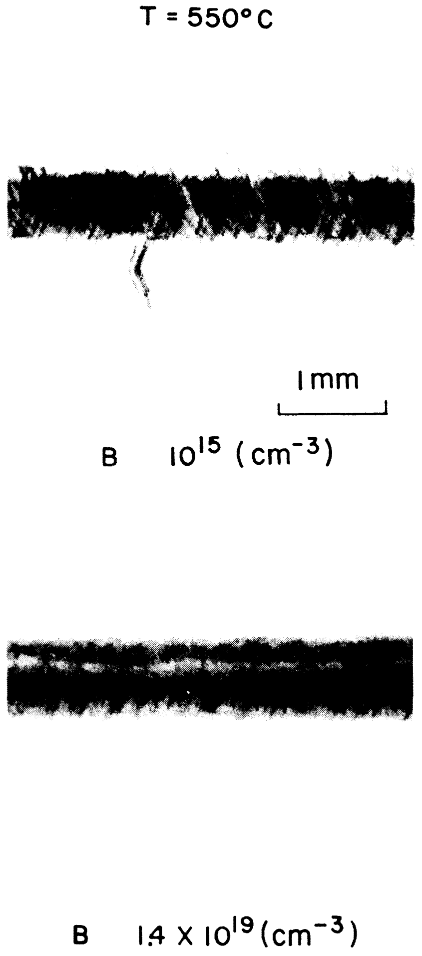

FIG. 2. (a) X-ray topographs of undoped p-type silicon deformed at $T = 550\,^\circ\mathrm{C}$; $\tau = 3$ kg/mm^2, $t = 93\,800$ sec. (b) p-type doped silicon 1.4×10^{19} cm^{-3} B deformed at $T = 550\,^\circ\mathrm{C}$; $\tau = 3$ kg/mm^2, $t = 57\,000$ sec.

(b) x-ray topography of dislocations moved directly from a scratch. The results are shown in Fig. 3. The etch pit and x-ray topography measurements gave identical values of dislocation velocity. The data in Fig. 3 follow a relation of the form $v \propto \tau^m$, where $m = 1.45$. The 60° dislocation velocity data of Erofeev *et al.*[4] are shown for comparison. These authors also report a slope $m = 1.45$ for low-doped silicon. Thus the value of m is reproducible among experiments, however, a theoretical understanding of this value is not available. The more recent work using individual loops and x-ray topography by George *et al.*[7] does not have data at the same temperature. We will compare later their results on screw dislocations with our work and show the close correspondence with the present work. The data in Fig. 3 give convincing evidence that the results between two different workers using different techniques and detection methods are quantitatively similar. We expect then that the results obtained below by our method of measuring dislocation motion directly from a scratch should give quantitatively similar results to measurements on the motion of individual dislocations.

III. IMPURITY CONCENTRATION DEPENDENCE OF DISLOCATION VELOCITY

A. n-type arsenic-doped silicon

The velocity of screw dislocations at constant stress and temperature as a function of impurity concentration in n-doped silicon is shown at 600 °C in Fig. 4. We note that at an impurity concentration of 10^{18} cm^{-3} the dislocation velocity begins to increase, and at 5.5×10^{19} cm^{-3} the velocity is over an order of magnitude larger than the intrinsic value. The intrinsic carrier concentration[8] for silicon at 600 °C is 3.4×10^{17} cm^{-3}. Essentially similar curves were observed at higher temperatures though the magnitude of the effect at a given impurity concentration decreased with increasing temperature.

B. p-type boron- and gallium-doped silicon at $T \geq 600\,^\circ\mathrm{C}$

The p-type results may readily be separated into two temperature regimes. Above 600 °C doping with gallium or boron impurity has no effect on dislocation velocity, to within an estimated experimental error of about 10%, even at concen-

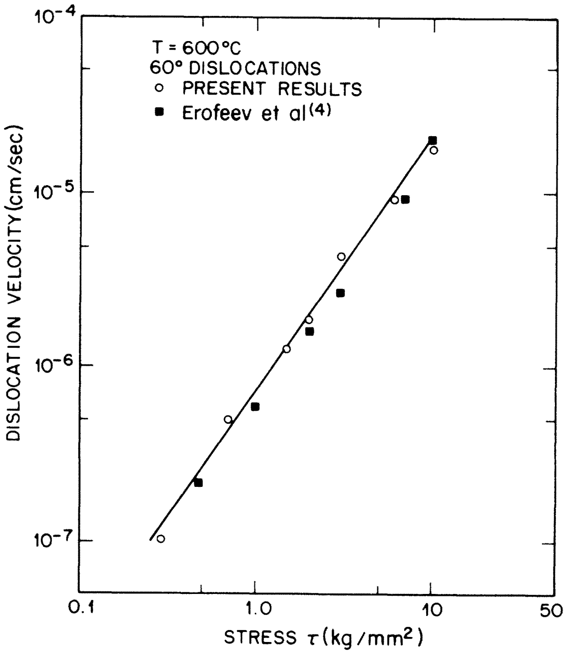

FIG. 3. Velocity vs stress for undoped n-type silicon at $T = 600\,^\circ\mathrm{C}$ for 60° dislocations. For comparison the results of Erofeev *et al.* (Ref. 4) are shown.

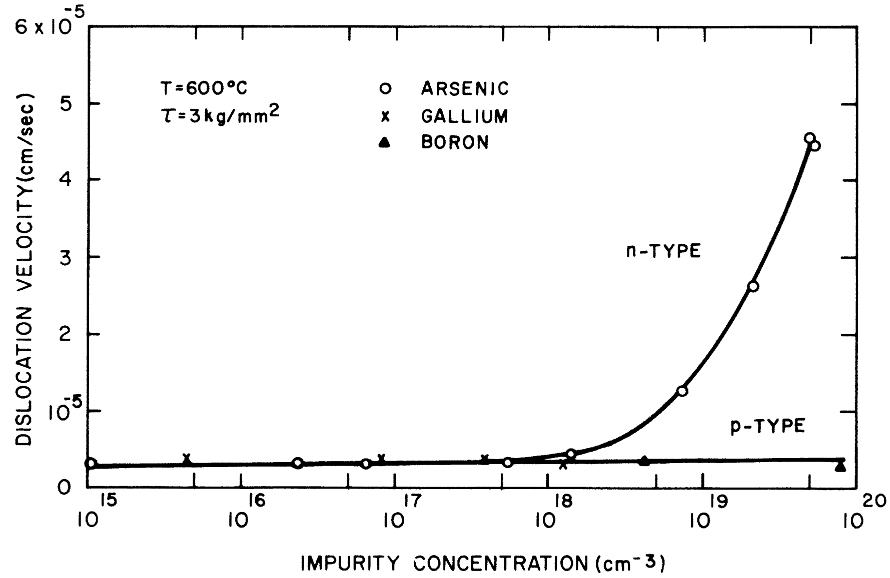

FIG. 4. Screw dislocation velocity vs impurity concentration for $T = 600$ °C; $\tau = 3$ kg/mm^2 for n- and p-doped silicon. Note that doping as high as 8×19 cm^{-3} does not change the dislocation velocity for p doping.

trations as high as 8×10^{19} cm^{-3} of boron. The results for p-type doping at $T = 600$ °C are shown in Fig. 4. It is evident that dislocation velocity is sensibly unchanged over the concentration range shown. At this same temperature of 600 °C Erofeev *et al.*[4] show what appears to be a small increase in velocity at about 9×10^{18} cm^{-3} of boron. The difference may be within the combined experimental errors. However, at a concentration level almost an order of magnitude higher than that of Erofeev *et al.*[4] we see no change of velocity. Furthermore measurements at $T = 900$ °C show v for 3×10^{19} cm^{-3} boron to be within 20% of the intrinsic value. Thus we conclude that for boron-doped crystals up to 3×10^{19} cm^{-3} and $600 < T < 900$ °C there is no detectable effect of the doping on dislocation velocity to within about 20%.

C. p-type boron- and gallium-doped silicon at $T \leq 600$ °C

Below 600 °C p doping increases dislocation velocity as shown by Erofeev *et al.*[4] The velocity curve as a function of doping at a constant stress of 12 kg/mm^2 and $T = 450$ °C is shown in Fig. 5. The velocity increases with impurity doping. This is exactly the opposite of what occurs in germanium where p doping lowers dislocation velocity.[1] The velocity enhancement is not as strong for p doping as it is for n doping. For instance, at $T = 550$ °C the increase in velocity is a factor of 3.3 for a p-doped silicon (8×10^{19} cm^{-3} boron), while for n doping the increase is a factor of 30, an order of magnitude greater. The results of the measurements of Erofeev *et al.*[4] on 60° dislocations are shown for comparison. The general enhancement of ve-

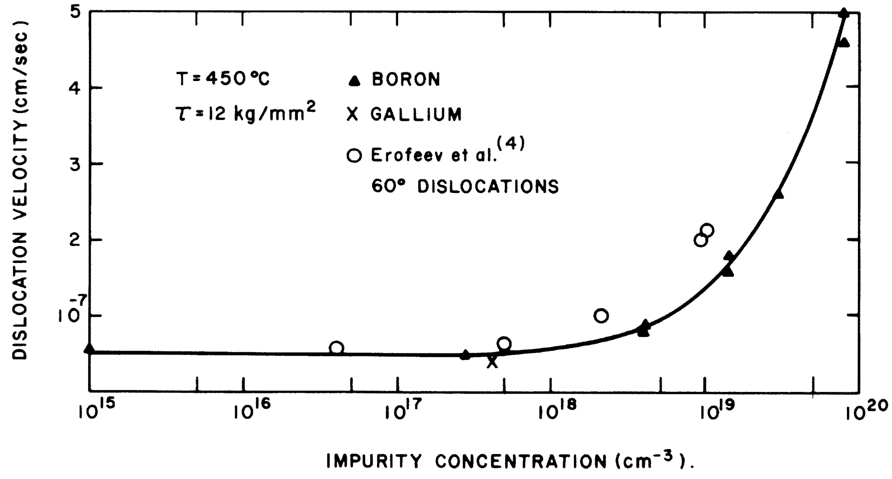

FIG. 5. Screw dislocation velocity vs impurity concentration for $T = 450$ °C; $\tau = 12$ kg/mm^2. Erofeev *et al.* (Ref. 4) points for 60° dislocations are shown for comparison.

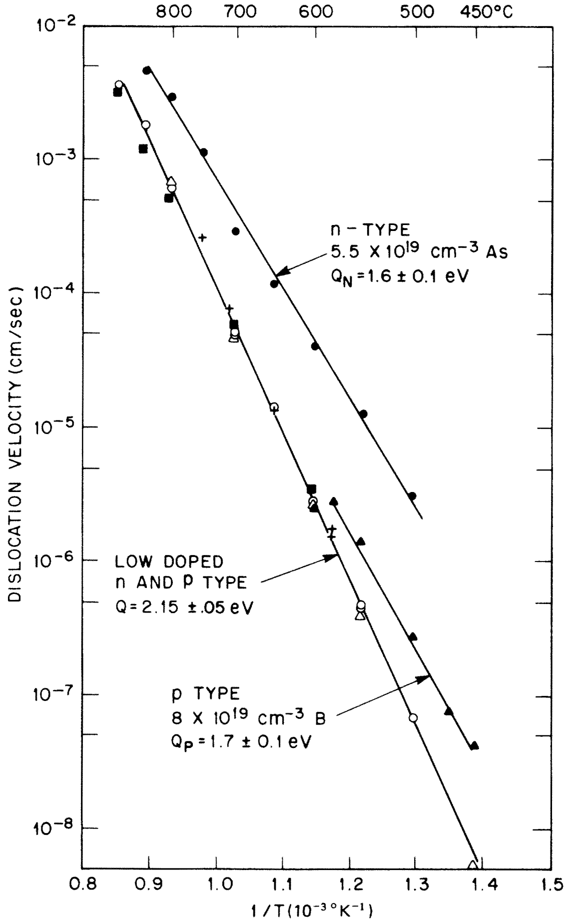

FIG. 6. Ln of screw dislocation velocity vs $1/T$ for undoped and n- and p-doped silicon; (○) 10^{14} cm^{-3} As; (△) 10^{15} cm^{-3} B; (▲) 8×10^{19} cm^{-3} B; (■) 3×10^{19} cm^{-3} B; (+) screw dislocation velocity data of George *et al.* (Ref. 7) at $\tau=3$ kg/mm^2.

locity with doping seems somewhat higher for 60° than for screw dislocations.

IV. TEMPERATURE DEPENDENCE OF DISLOCATION VELOCITY FOR UNDOPED AND HEAVILY DOPED CRYSTALS

The temperature dependence of dislocation velocity at constant stress follows the relation $v=v_0 e^{-Q/kT}$. In Fig. 6 we show ln$v$ plotted versus $1/T$ over the temperature range 450–900 °C at a constant stress $\tau=3$ kg/mm^2. Both boron- and arsenic-doped crystals in the low-doping range fall without much scatter on a straight line in Fig. 6 with an activation energy $Q_I=2.1\pm0.05$ eV. The points obtained by George *et al.*[7] for screw dislocations at the same stress as a function of temperature are also shown in Fig. 6. It is evident that their points are substantially on the same curve as our data. The lowest point at $T=450$ °C is the value normalized for $\tau=3$ kg/mm^2 from data taken at $\tau=12$ kg/mm^2.

For 5.5×10^{19} cm^{-3} arsenic doping the data can also be fitted with a straight line, however, the scatter observed at the high-doping levels is somewhat greater than that for the undoped crystals and the activation energy $Q_N=1.6\pm0.1$ eV. It is evident from Fig. 6 that the effect of doping is higher the lower the temperatures. At high temperatures, where the intrinsic carrier concentration approaches the doping concentration, the increase of dislocation velocity over the undoped crystals is not as pronounced. For instance, at 900 °C the increase in velocity for n doping is only a factor of 2.5 while at 500 °C it is almost a factor of 47 over the undoped values.

As indicated in the previous section the results for p doping for $T\geq600$ °C are similar to the results on undoped or low-doped silicon. This is indicated by the points (■) for T above 600 °C. The increase in velocity with boron doping becomes smaller than the experimental error in the temperature region $\gtrsim600$ °C. Below 600 °C p doping raises dislocation velocity. At 580 °C some slight increase is observed over the undoped value; by 500 °C the increase is almost a factor of 3.5 while at 450 °C the increase over low-doped silicon is about a factor of 8 for 8×10^{19} cm^{-3} of boron doping.

V. THEORETICAL ESTIMATES OF THE INCREASE OF DISLOCATION VELOCITY IN DOPED SILICON

For germanium Frisch and Patel[5] have proposed a model to account for the velocity behavior in crystals heavily doped with n and p impurities. The model postulates that kinks responsible for dislocation motion are associated mainly with charged dislocation sites. We make the further reasonable assumption that the enhancement of dislocation velocity, due to doping, $v_D\propto\theta$ where θ is proportional to the fraction of charged dislocation sites. It follows then that any mechanism such as chemical doping that increases the electron concentration will increase the charged dislocation acceptor sites in germanium,[1] and consequently raises the kink concentration and dislocation velocity. Similarly, doping with acceptors will reduce the kink density and hence the dislocation velocity. While a complete mathematical treatment presents severe problems, Frisch and Patel considered very simply the charge balance around an isolated dislocation and were able to estimate semiquantitatively the value of the ratio $R_D=v_D/v_I$, where v_D is the dislocation velocity in doped silicon and v_I the constant value reached at low doping. They found

$$R=\tfrac{1}{2}m(D^+-A^-)\pm\tfrac{1}{2}\{[m(D^+-A^-)]^2+4\}^{1/2}, \qquad (1)$$

where D^+ and A^- are the donor and acceptor concentration, respectively, and m is a single adjustable parameter. With a suitable choice of m, the theoretical calculated R fitted the experimental data in germanium and silicon reasonably well.

Explicit in Eq. (1) is the assumption that the concentration of the various species entering the charge-balance equation obeys Boltzmann statistics.

In order to obtain the energy level of the dislocation the adjustable parameter m must be obtained as a function of temperature. It has also been shown[3] that a plot of $\ln m^{-2}$ vs $1/T$ is a straight line whose slope gives the dislocation-energy level. For silicon it was found from the analysis on n-doped crystals that the dislocation-acceptor level[3] is almost at the conduction-band level. It was pointed out that the accuracy of this determination was poor since the inherent errors in determining m are magnified in taking the square of m. A more direct approach would be to treat the charge-balance relation by Fermi-Dirac statistics and, by fitting each R vs impurity-concentration curve, obtain the dislocation-energy level directly. The consistency of the dislocation-energy level at various temperatures is a measure of the reliability of this approach.

VI. APPLICATION OF FERMI-DIRAC STATISTICS TO ESTIMATE THE DISLOCATION VELOCITY RATIO R_D

To a very good approximation the velocities of single dislocations are observed to be the same as those measured experimentally for dislocation groups. We therefore write the charge-balance condition but consider only a small volume element enclosing the dislocation line. Within this volume we assume that the changes in the electrical properties from the host crystal owing to the dislocation are uniformly averaged. The radius of the cylinder is approximated by the Debye length. The model has been described by Frisch and Patel.[5] Within this volume we have

$$N_1 + N_2 + A = P_1 + P_2 + D , \qquad (2)$$

where the relations for N_1, P_1, N_2, and P_2 are the usual expressions for the thermal electron and holes and the ionized acceptor and donor impurities. The full expressions used are given in the Appendix. We confine our attention here to the terms A and D where

$$A/A_0 = 1/[(1/g)\exp(E_D - E_F/kT) + 1] \qquad (3)$$

is the fraction of charged dislocation sites if the dislocation is an acceptor, g is the appropriate degeneracy factor, E_D is the dislocation-acceptor level, and E_F is the Fermi level.

The ratio of the dislocation velocity in the doped to the undoped crystal

$$R = v_D/v_I = (\theta)_D/(\theta)_I , \qquad (4)$$

where $(\theta)_D$ is the fraction of charged dislocation sites, i.e., the Fermi factor of the doped crystal at the dislocation level E_D and $(\theta)_I$ for the undoped or low-doped crystal. The parameters A_0 and E_D are adjusted so as to fit the experimental data.

If the dislocation is a donor the fraction of charged donor sites is given by

$$D/D_0 = \{1 - [1/(1/g)\exp(E_D - E_F/kT) + 1]\} , \qquad (5)$$

where g is again the appropriate degeneracy factor.

VII. COMPARISON OF THEORY AND EXPERIMENT FOR THE DISLOCATION VELOCITY RATIO R

A. n-doped silicon

At a constant temperature the addition of n-type impurities shifts the Fermi level towards the conduction band and raises the dislocation velocity. We have also assumed that the velocity is proportional to the charged-dislocation sites. It follows then that we must consider the dislocation sites to be acceptorlike in nature. Donorlike behavior would result in a decrease in dislocation velocity. In Sec. XC we will discuss other evidence for the acceptorlike behavior of dislocations in n-doped silicon.

We use Eq. (2) to establish the Fermi level E_F as a function of the impurity concentration N with D_0 in Eq. (5) set to zero and A given by Eq. (3). The results that best fit the n-type data at $T = 600$ °C for a fixed τ are shown in Fig. 7 plotted as $\log R$ vs $\log N$, where N is the concentration of n-type impurities. The solid theoretical curves are for two values of $E_D = 0.5E_g$ and $0.6E_g$, where E_g (~ 1.0 eV at $T = 600$ °C)[9] is the gap energy measured from the valence band E_v. The open circles

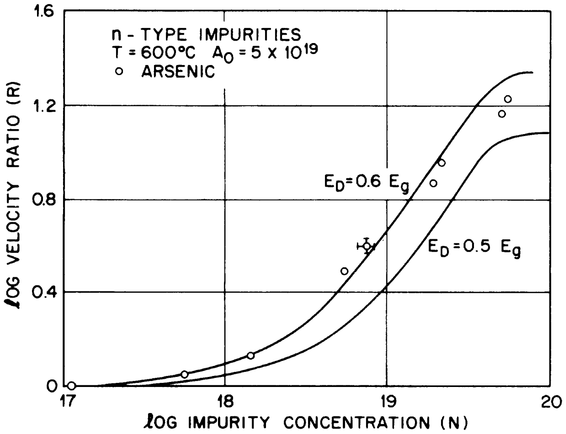

FIG. 7. Log of the velocity ratio (R) vs log of impurity concentration for n doping. Solid line theory, points experimental at $T = 600°$, $A_0 = 5\times10^{19}$. Curve for $E_D = 0.5E_g$ shows the dependence of R on choice of E_D. Estimated error bars are shown.

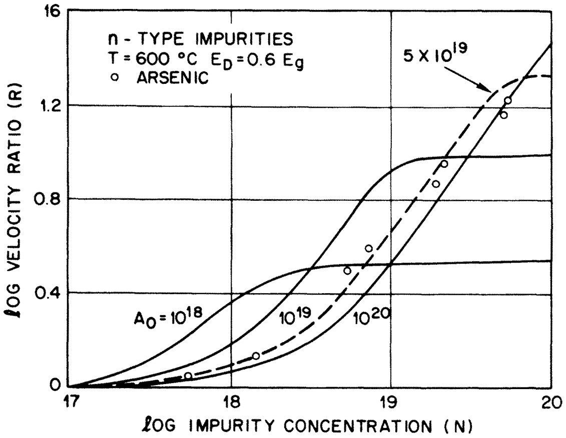

FIG. 8. Log$R$ vs log$N$ at $T = 600°$ and $E_D = 0.6\ E_g$ showing the dependence of the calculated curve on the choice of A_0. Note that the calculated curves are quite sensitive to the choice of A_0.

are the experimental data. The theoretical curve that best fits the data gives the acceptor level $E_D = 0.6E_g (\sim 0.6$ eV) above the valence band E_v. An estimate of the error bars is shown in Fig. 7.

The calculated curves in Fig. 7 also show that the best fit is obtained for $A_0 = 5 \times 10^{19}$ cm^{-3}, where A_0 or D_0 in the model is the average acceptor or donor state concentration owing to the dislocation and whose ionization state controls the dislocation velocity. The compliance of the fitted results to variations in A_0 was tested and it was found that for $A_0 = 10^{18}$ or 10^{19} cm^{-3} the curves saturate at low values of R (Fig. 8) and obviously cannot be fitted to the experimental data. For low A_0 saturation occurs at low R values because the Fermi level tends to saturate at lower values of impurity concentration. For $A_0 = 10^{20}$ cm^{-3} saturation is not apparent below an impurity concentration of 10^{20} cm^{-3} which is in poorer agreement with experiment. The best correspondence to the data is obtained at $A_0 = 5 \times 10^{19}$ cm^{-3}. Finally it should be noted that while an accurate microscopic calculation of A_0 as defined in the model cannot be carried out, its approximate magnitude should be amenable to physical argument. We will show in Sec. X B that a rough but reasonable estimate of A_0 does agree in order of magnitude with the fitted value.

At higher temperatures for $T = 800$ and 700 °C somewhat larger scatter is observed but the data lie between $0.5E_g < E_D < 0.6E_g$. We estimate from our method of fitting the data that the value of E_D is not determined to better than $E_D = 0.6 \pm 0.1$ eV above the valence band.[9]

B. *p*-doped silicon $T < 600$ °C

Since experimentally p doping also raises the dislocation velocity below 600 °C we must consider the dislocation level in p-type crystals as being donorlike. Acceptorlike behavior would result in a lowering of dislocation velocity with p doping.

Using the same calculation procedure as in Sec. VII A with $A_0 = 0$ in Eq. (3), we obtain the curves shown in Fig. 9. The solid curves are calculated for $E_D = 0.5E_g$ and $0.6E_g$ at $T = 450$ °C. The points are experimental. The curve for $E_d = 0.6E_g$ best fits the data at $T = 450$ °C. At 550 °C the points lie between $0.65E_g < E_d < 0.7E_g$. We conclude then that the donor level for the dislocation in p-doped silicon lies at $E_d = 0.6 \pm 0.1$ eV (above the valence band).

VIII. ACTIVATION ENERGY FOR DISLOCATION VELOCITY IN DOPED SILICON

In this section we attempt to calculate the observed change in activation energy over the value for the undoped crystals. The expression for the ratio of the dislocation velocities between the n-doped crystal with an acceptorlike dislocation level and undoped crystals is

$$R = \frac{v_D}{v_I} = \frac{f(\theta)_D}{f(\theta)_I} = \frac{1 + \exp(E_D - E_{FI}/kT)}{1 + \exp(E_D - E_{FD}/kT)}, \tag{6}$$

where E_{FI} is the Fermi level in the undoped crystal and E_{FD} is the level in the doped crystal. For p-doped crystals with donorlike dislocations a similar expression holds with a negative sign for the exponents. The physical source for the change in activation energy from Q_I to Q_D can easily be seen from Eq. (6). For relatively low doping (with $E_{FD} < E_D$ by several kT) in n-doped crystals with $E_D > E_{FI}$ we obtain

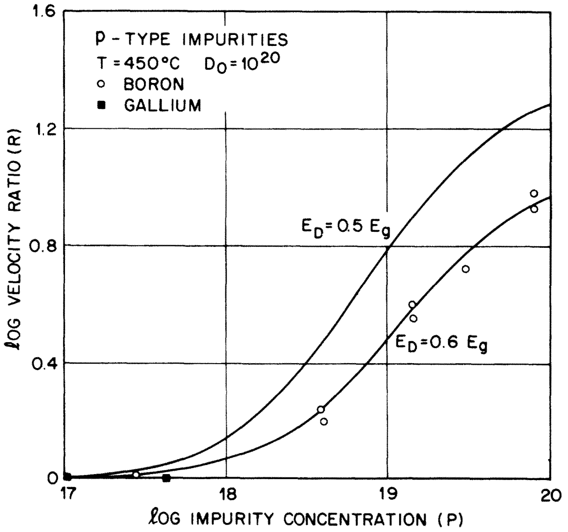

FIG. 9. Log$R$ vs log$P$ at $T = 450$ °C; $D_0 = 10^{20}$, solid curve: theory; points: experimental. Curve for $E_D = 0.5\ E_g$ shows that the calculated R is quite sensitive to the choice of E_D.

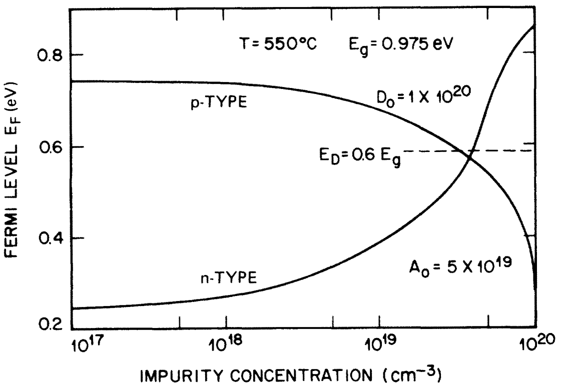

FIG. 10. Fermi level E_F measured from the valence band E_V vs impurity concentration at $T = 550\,°C$, and $E_D = 0.6\ E_g$.

$$v_D \approx v_0 \exp(-Q_I/kT) \exp(E_{FD} - E_{FI}/kT). \qquad (7)$$

In other words, the difference in the activation energy between the dislocation velocity in the doped (Q_D) and undoped (Q_I) crystal is simply the change in the Fermi level ($E_{FD} - E_{FI}$) between the doped and undoped crystal.

For high doping levels with ($E_{FD} > E_D$ by several kT) the lower exponential in Eq. (6) can be ignored and we obtain

$$v_D = v_0 e^{-Q_I/kT} e^{E_D - E_{FI}/kT} . \qquad (8)$$

Thus in this case the additional activation energy is given by ($E_D - E_{FI}$) and is no longer dependent on doping. The full Eq. (6) will show a smooth transition from the activation energy in Eq. (7) to that in Eq. (8). As an example we see from Figs. 6 and 10 that 5.5×10^{19} cm^{-3} As corresponds to the high doping level and we expect

$$\begin{aligned} Q_D &= Q_I + (E_D - E_{FI}) = 2.15 - 0.35 \text{ eV} \\ &= 1.8 \pm 0.1 \text{ eV (calculated)} \\ &= 1.6 \pm 0.1 \text{ eV (measured)} . \end{aligned}$$

For p doping we expect the activation energy to decrease with doping but only up to an amount $E_D - E_{FI} \sim 0.2 \pm 0.1$ eV, while the observed decrease is $\sim 0.45 \pm 0.1$ eV. Although the model correctly predicts that the p-doping effects on dislocation velocity will be smaller than for n doping, the disagreement in the detailed comparison is somewhat outside the estimated errors.

IX. DISLOCATION VELOCITY BEHAVIOR IN HEAVILY DOPED SILICON AND GERMANIUM

To summarize the behavior of dislocations in heavily doped silicon; (a) n-type doping raises dislocation velocity over the undoped value and (b) the behavior in p-doped silicon is anomalous, for $T > 600\,°C$ heavy doping does not change dislocation velocity from its undoped value. For $T < 600\,°C$ p doping also raises dislocation velocity although the effect is not as high as with n doping.

The behavior observed in silicon is quite different from the effect observed in germanium. The results in germanium can best be summarized in Fig. 11, which reproduces earlier work showing the velocity behavior as a function of temperature. Clearly n-type arsenic doping raises dislocation velocity while p-type gallium doping lowers it compared to the intrinsic value. It has also been shown in the germanium work that neutral impurities do not influence the dislocation velocity, which remains unchanged at the intrinsic value. We have argued earlier[5] that the germanium results can be explained if the dislocation is acceptorlike in both p- and n-doped germanium. The behavior in silicon is somewhat more complex as shown by the evidence in the previous sections.

X. SUMMARY AND DISCUSSION

A. Experimental

The velocities of dislocations in both silicon and germanium are influenced by the type and amount of electrically active impurities in the crystal. In germanium n doping raises, while p doping decreases the dislocation velocity compared to the intrinsic value. In silicon n doping also

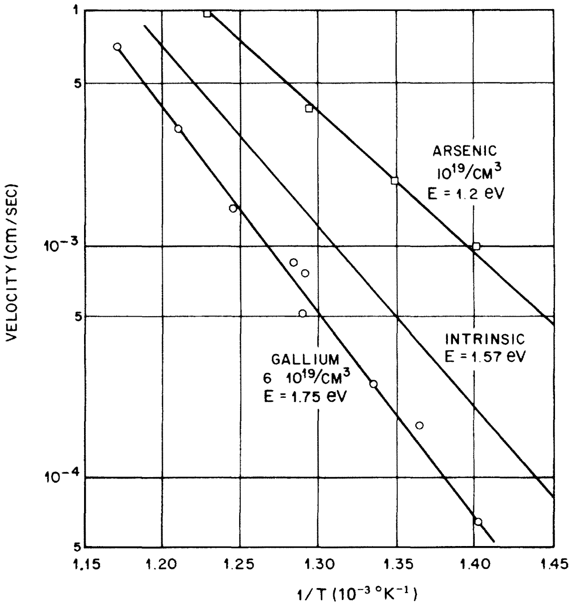

FIG. 11. Ln velocity vs $1/T$ for arsenic and gallium-doped crystals at a constant shear stress $\tau = 6$ kg/mm^2. Activation energy is higher for gallium doping and lower for arsenic doping compared to intrinsic crystals. After Patel and Chaudhuri (Ref. 1).

raises dislocation velocity but the p-doping behavior is anomalous. For $T \gtrsim 600$ °C the dislocation velocity for p doping is indistinguishable from the undoped value (Fig. 6) within the experimental spread of our measurements. Erofeev *et al.*[4] report that the curve of dislocation velocity vs $1/T$ falls below the undoped value at $T \gtrsim 600$ °C. Our experimental measurements up to 900 °C and at a doping level a factor of 3–4 greater than theirs does not bear out their conclusion since we would have expected a wider difference at 900 °C if the experimental curves were indeed diverging for our carrier concentration of 3×10^{19} cm^{-3}. From the relatively small spread of the undoped and p-doped velocities for $T \gtrsim 600$ °C in their data we do not feel that Erofeev *et al.* are justified in concluding that a real difference exists. While the scatter in our data does not allow us to state with certainty that there is a definite break in $\ln v$ vs $1/T$ for p doping around $T \simeq 600$ °C, we can with good justification assert that the differences between p doped and undoped velocities above $T \simeq 600$ °C are within experimental error. For $T \lesssim 600$ °C p doping raises dislocation velocity. In this range of temperatures we confirm the data of Erofeev *et al.*

B. Theory

The theory we have used to fit the experiments is a two parameter theory, since both A_0 and E_D must be adjusted in Eq. (3) [and D_0 and E_D in Eq. (5)] to fit the n- and p-doping results, respectively, at a given temperature. A single choice of A_0 and E_D is found to reproduce the velocity vs doping behavior over 2–3 orders of magnitude in dopant concentration. Once A_0 or D_0 has been determined for a given temperature we find that a good fit of the experimental data can be obtained with the same A_0 or D_0 values at other temperatures with very nearly the same E_D value. The spread of the dislocation-energy level E_D is not more than about 0.1 eV over the temperature range for a given doping type.[9]

Both A_0 and D_0 are analogous to N_A and N_D in the expressions for the ionized acceptor and donor concentrations in Eqs. (A2) and (A4) in the Appendix. One may imagine that A_0 and D_0 represent the concentration of available charge sites along a single dislocation. The problem arises in trying to describe a cutoff radius for a region around a dislocation. If for simplicity we assume that the Debye radius defines this cutoff value, then for a carrier concentration of 10^{19} cm^{-3} the value of A_0 or D_0 is of the order of 10^{20} cm^{-3}, a value that is not unreasonable compared to the values of 5×10^{19} and 10^{20} assumed for A_0 and D_0, respectively, in fitting the experimental data.

In the theory we have assumed that the kinks which enhance dislocation motion are associated with the charged dislocation sites. Based on the experimental data we must then conclude that, for n-type doping which enhanced dislocation velocity, the dislocation is acceptorlike. For p doping at $T \lesssim 600$ °C the observed enhancement of velocity allows us to postulate that the dislocation is donorlike. If the dislocation were acceptorlike, then p doping, which moves the Fermi level E_F towards the valence band E_v, would decrease the charge on the dislocation and hence reduce the kink density and the dislocation velocity. Based on these arguments we find that the dislocation level is acceptorlike in n-doped silicon with level $E_D \simeq 0.6 \pm 0.1$ eV above the valence band. For p-doped silicon the level is donorlike with a level $E_D \simeq 0.6 \pm 0.1$ eV above the valence band.

The anomalous behavior for p-type silicon is in qualitative accord with the model predictions but fails in detailed comparison. Thus we can fit the data at $T = 450$ °C assuming $E_D \simeq 0.6 \pm 0.1$ eV but this parameter leads to velocity enhancements at $T > 600$ °C of ~50–200% while no enhancement is observed within ±20–30%. However, if E_D is not temperature independent but increases from 0.6 to 0.7 eV for $450 < T \lesssim 700$ °C the model would correctly predict the null effect within experimental error observed at the higher temperatures. Such movement of E_D may not de unreasonable, but in the absence of confirming evidence is only an *ad hoc* conclusion to rectify the discrepancy.

C. Comparison with electrical measurements

The discussion would not be complete without a mention of the very extensive and often contradictory experimental findings on the electrical effects of dislocations in semiconductors measured below room temperatures in crystals where the doping levels are typically five or six orders of magnitude below ours. From the extensive literature we confine ourselves to the results of two groups of workers which point up the dillemma in the experimental results on the electrical effects of dislocations.

In two papers Glaenzer and Jordan[10] conclude from carrier lifetime and conductivity measurements that (i) dislocations in n-type silicon (N_d ~ 10^{13}–10^{15} cm^{-3}) act as acceptors with a level 0.6 eV above the valence band E_v, at room temperature; (ii) in p-type silicon, dislocations are donors with a level that decreases with an increase in doping level from 0.5 to 0.38 eV above the valence band E_v.

In a series of papers principally on Hall effect and conductivity measurements on silicon and germanium, the group of Göttingen (Haasen and Schröter,[11] Labusch and Schettler[12]) find (i) above 300 °K dislocations behave as acceptors in both p-type germanium and silicon at low doping (10^{13}–

TABLE I. Comparison of dislocation-energy levels in silicon determined from dislocation-velocity and electrical measurements.

	Dislocation energy level		
		Electrical measurements[a]	
Silicon type	From present velocity measurements[b]	Glaenzer and Jordan	Schröter and Haasen
n	0.6 ± 0.1 eV (acceptor)	0.6 eV (acceptor)	0.3 eV (neutral)
p	0.6 ± 0.1 eV (donor)	0.5–0.38 eV (donor)	

[a] Energy levels from the electrical measurements are for room temperature and below.

[b] The dislocation levels from the present experiments for *n*-type Si are determined at 500 °C < T < 850 °C and *p* type at 450 < T < 600 °C.

10^{14} cm^{-3}); (ii) below about 300 °K for silicon and 80 °K for germanium dislocations behave as donors at low doping (the temperatures quoted are nominal and depend on the dislocation density and doping levels) and (iii) neutral levels are at $E_D = 0.09$ eV (Ge) and $E_D = 0.3$ eV (Si) above the valence band E_v.

We note that the energy levels are nowhere near agreement between these two groups for silicon. Similar discrepancies exist among other workers and Haasen and Schröter[11] have in fact acknowledged and discussed this situation in their work. The differences in these findings may lie partly in how the analysis is made. The reader is referred to the various papers referenced here.[11,12]

We must, however, point out that our results appear more consistent with those of Glaenzer and Jordan both as to the donor and acceptor behavior in *p*- and *n*-doped silicon and also with their values of E_D which are close to ours, as shown in Table I. In fairness we must also state that our measurements are in a very different regime of temperature and impurity concentration compared to the electrical measurements, and perhaps a strict correspondence between the two types of results, velocity and electrical measurements, may not be expected. Nevertheless the qualitative agreement described above could not be evaded. The largest discrepancy between our results and that of Glaenzer and Jordan, which is in the magnitude (but not the electrical nature) of the dislocation level of *p*-Si, may be due to the fact (see Glaenzer and Jordan[10]) that this level is very sensitive to the doping level and perhaps the temperature. Our failure to see any doping dependence of the dislocation velocity above 600 °C in *p*-Si would indicate, according to the Patel-Frisch model, that the Fermi level has not passed through, or come close to, the dislocation level. This may imply that the dislocation level has moved close to the valence or conduction-band edges. As to why this is so cannot be explained on the basis of our simple statistical model.

We have found that the activation energy Q_D for the dislocation velocity in the high *n*-doping limit ($E_{FD} \gg E_D$ by at least several kT) is given by $Q_D \simeq Q_I - (E_D - E_{FI})$, where Q_I is the activation energy for the undoped crystal and $(E_D - E_{FI})$ is the difference between the dislocation level and the Fermi level in the intrinsic crystal (provided $E_D > E_{FI}$ by at least several kT). For the low doping limit ($E_{FD} < E_D$ and $E_D > E_{FI}$ by several kT) we obtain $Q_D = Q_I - (E_{FD} - E_{FI})$ where $(E_{FD} - E_{FI})$ is just the change in Fermi level with doping.

In a recent communication Haasen,[13] by estimating "ΔE" the energy gained by forming a double kink, has also calculated the activation energy for dislocation velocity in doped crystals. In his approach he estimates the electrostatic energy associated with a kink "ΔE" as a function of the fraction of occupied dislocation sites. The relevant fraction f is then found from a knowledge of the Fermi level, the electrostatic interaction energy of charges on a dislocation line, and a knowledge of the neutral dislocation level obtained from low-temperature Hall-effect measurements on *p*-doped silicon. The calculated ΔE is now taken to be the change in activation energy with doping. A comparison between our theoretical estimate and Haasen's and the experimental activation energies is shown in Table II.

In conclusion, we feel that our derivation of the electrical parameters of dislocations from dislocation-velocity measurements has the following advantages:

(i) The measurements in effect are on a single dislocation. They do not rely on a special arrangement of dislocations as is necessary for the electrical measurements.

(ii) Trace impurities in the 10^{14}–10^{15} cm^{-3} range that are difficult to prevent from diffusing into silicon during heating are negligible compared to the concentration range 10^{18}–10^{20} cm^{-3}, where the

TABLE II. Comparison of experimental activation energies for dislocation motion in doped silicon with theoretical estimates.

		Activation energy for dislocation velocity			
Silicon type		Present experiment (screw)	Present theory (screw)	Expt.[4] (60°)	Theory Haasen[13] (60°)
n	As (5.5 × 10^{19} cm^{-3})[a]	1.6 ± 0.1 eV	1.8 ± 0.1 eV		
	(10^{19} cm^{-3})	1.8 ± 0.1 eV	2.0 ± 0.1 eV	1.5	1.04 eV
p	B (8 × 10^{19} cm^{-3})[a]	1.7 ± 0.1 eV	1.95 ± 0.1 eV		
	(10^{19} cm^{-3})	1.9 ± 0.1 eV	2.1 ± 0.1 eV	1.75	1.8 eV

[a] The activation energy values for 5.5 × 10^{19} cm^{-3} and 8 × 10^{19} cm^{-3} are from experiment, whereas at 10^{19} cm^{-3} of As or B they represent interpolated values. Owing to inherent experimental uncertainties differences of 0.1–0.2 eV are reasonable in the activation energies.

main effects are observed on dislocation velocity. They are, of course, crucial for the electrical measurements.

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APPENDIX

The expressions for the thermal- and the ionized-electron and hole concentrations are outlined below. The thermal-electron concentration is

$$N_1 = 5.449\times 10^{15}\left(T\,\frac{m_e^*}{m_e}\right)^{3/2}\mathfrak{F}_{1/2}\left(\frac{E_F-E_g}{kT}\right). \tag{A1}$$

The ionized-acceptor concentration is

$$N_2 = \frac{N_a}{1+0.25\exp[(E_a-E_F)/kT]}. \tag{A2}$$

The thermal-hole concentration is

$$P_1 = 5.449\times 10^{15}\left(T\,\frac{m_h^*}{m_e}\right)^{3/2}\mathfrak{F}_{1/2}\left(\frac{E_v-E_F}{kT}\right). \tag{A3}$$

The ionized-donor concentration is

$$P_2 = N_d\left(1-\frac{1}{1+0.5\exp[(E_d-E_F)/kT]}\right), \tag{A4}$$

where electron density of states effective-mass ratio is $m_e^*/m_e = 1.082$, and the hole density of states effective-mass ratio is $m_h^*/m_e = 0.594$. $\mathfrak{F}_{1/2}$ is the Fermi integral given in Ref. 14, E_F the Fermi level, E_c and E_v the conduction- and valence-band levels, respectively, E_g the energy gap, and N_a and N_d are the total concentrations of chemical acceptor and donor impurities. E_a and E_d are the acceptor and donor impurity levels and the constants 0.25 and 0.5 in (A1) and (A4) are the ground-state degeneracy factors for the acceptor and donor levels, respectively.

The use of this model at high doping levels ($\sim 10^{20}$ cm^{-3}) will lead to some errors resulting from the impurity band broadening at the high impurity concentrations. Generally this will introduce uncertainties of ~ 0.1 eV which arise from the impurity bandwidths at these doping levels.[14] The model also treats the valence and conduction bands of silicon as parabolic which is approximately correct at low doping levels, but may lead to some errors in high carrier concentration p-type material due to the warped valence band.

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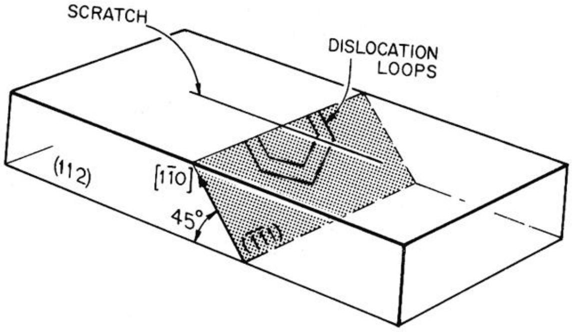
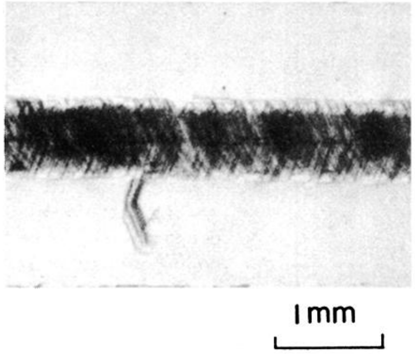
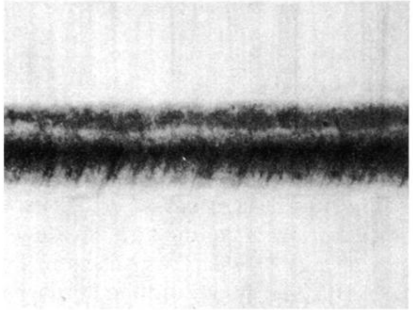


FIG. 1. Geometry of silicon crystals and glide system for four point bending experiments, showing loops generated from a scratch.

T = 550°C



B 10^{15} (cm^{-3})



B 1.4×10^{19} (cm^{-3})

FIG. 2. (a) X-ray topographs of undoped *p*-type silicon deformed at $T = 550\,°C$; $\tau = 3$ kg/mm^2, $t = 93\,800$ sec. (b) *p*-type doped silicon 1.4×10^{19} cm^{-3} B deformed at $T = 550\,°C$; $\tau = 3$ kg/mm^2, $t = 57\,000$ sec.