

# Dynamic behavior of dislocations in InAs: In comparison with III–V compounds and other semiconductors

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The dynamic behavior of  $\alpha$  and  $\beta$  dislocations in both undoped and impurity-doped InAs crystals is investigated as a function of stress and temperature by means of the etch pit technique. Suppression of the generation of dislocations from a surface scratch is found in S doped InAs, which is interpreted in terms of dislocation locking due to impurity clusters and/or impurity-defect complexes as has been observed in other III–V compounds.  $\alpha$  dislocations move faster than  $\beta$  dislocations in the undoped and impurity-doped InAs. S donors reduce the velocities of both  $\alpha$  and  $\beta$  dislocations. On the other hand, Zn acceptors enhance the velocity of  $\alpha$  dislocations and reduce the velocity of  $\beta$  dislocations. Such features are related to an electronic state of the dislocations in the elementary process of the motion. The measured velocities are expressed using a simple empirical equation as a function of stress and temperature in the same manner as for other semiconductors. The linear dependence of the activation energies for dislocation motion on the band gap energy shows a clear distinction dependent on the group of semiconductors, i.e., the elemental and IV–IV compound, III–V compound, and II–VI compound. © 1998 American Institute of Physics. [S0021-8979(98)09320-7]

## I. INTRODUCTION

In III–V compound semiconductor crystals with the sphalerite structure, there are three basically different types of dislocation, namely  $\alpha$ ,  $\beta$ , and screw dislocations. It has been established that the dynamic motion of dislocations depends on the type of dislocation, on the material, and also on electrically active impurities and that certain kinds of impurities immobilize dislocations by locking as reported in detail by the present author for GaAs,<sup>1</sup> GaP,<sup>2</sup> and InP.<sup>3</sup> Such knowledge is now applied widely in growing crystals with low dislocation densities or even free from dislocations. Only one study<sup>4</sup> has been done to date on the velocities of dislocations within undoped InAs, and there has been no study on the effects of impurities. In order to clarify the fundamental process of dislocation motion and impurity interaction in semiconducting materials, it is necessary to establish knowledge of the dynamic properties of dislocations within InAs, in comparison with other III–V compound semiconductors.

InAs has a band gap of 0.36 eV at room temperature and is a suitable material for photodetectors in the 1–3  $\mu\text{m}$  range. The high electron mobility ( $2 \times 10^4 \text{ cm}^2/\text{V s}$  at room temperature) offers the potential of high frequency field effect transistors and sensitive Hall effect devices due to strong interactions with a magnetic field.<sup>5</sup> In such an application, dislocations should degrade the device efficiency, as is known for GaAs, GaP, and InP. Thus, a knowledge of dislocation velocities and the interaction of dislocations with various impurities certainly contributes to the development of the technology for reliable high performance devices.

This article reports on the dynamic behavior of dislocations within InAs, investigated by means of the etch-pit technique. The dislocation velocity in undoped InAs and the impurity effect on the generation and velocity of dislocations in Zn and S doped InAs are compared with other III–V compound semiconductors. The activation energy for dislocation motion in undoped crystals in a series of III–V compound semiconductors is discussed together with some elemental and IV–IV and II–VI compound semiconductors.

## II. EXPERIMENT

Specimens were prepared from three kinds of crystal grown by the liquid-encapsulated Czochralski technique; a nominally undoped crystal (*n*-type) with a free carrier concentration of  $2.0 \times 10^{16} \text{ cm}^{-3}$ , a crystal doped with S to a concentration of  $1.3 \times 10^{18} \text{ cm}^{-3}$ , and a crystal doped with Zn to a concentration of  $3.7 \times 10^{17} \text{ cm}^{-3}$ . The density of grown-in dislocations in the crystals was of the order of  $10^3 \text{ cm}^{-2}$ . Specimens were sectioned into a rectangular shape, approximately  $2 \times 3 \times 15 \text{ mm}^3$ , with a long axis along the  $[\bar{1}10]$  direction and side surfaces parallel to the (111) and (11 $\bar{2}$ ) planes. The surfaces were chemically finished with a reagent of bromine–methanol following mechanical polishing. The specimen was stressed at elevated temperature by means of three-point bending in a vacuum. The bending axis was parallel to  $[\bar{1}1\bar{2}]$  for observation of the motion of  $\alpha$  and  $\beta$  dislocations. Dislocations were generated from a scratch drawn on the (111) surface along the  $[\bar{1}10]$  direction at room temperature with a diamond stylus. The displacements of dislocations generated from a scratch during stressing were measured from the etch pits developed by the RC-1 etchant<sup>6</sup> at 20 °C. The specimen surface was very smooth after stressing as determined by optical microscopy. Therefore, it was

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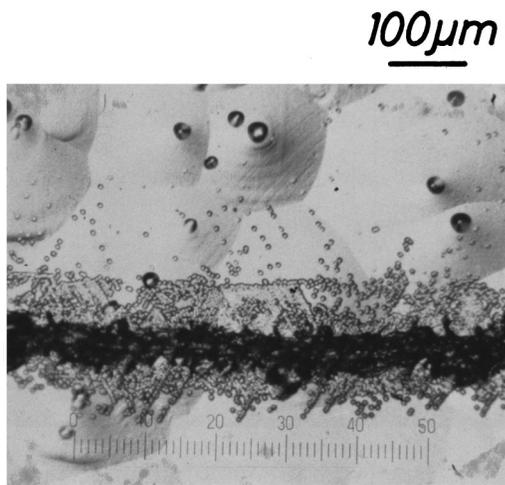


FIG. 1. Dislocation etch pits developed around a scratch on the (111) surface of undoped InAs subjected to a maximum resolved shear stress of 20 MPa by three-point bending at 350 °C for 30 s. The magnitude of the applied stress to the central region of the figure was estimated to be 10 MPa using the distance from the bending knife edges.

assumed that the effect of thermal decomposition of the InAs surface was not serious enough to affect the experimental results. The geometry of the specimen as well as the details of experimental procedure are described in previous articles.<sup>1-3</sup>

### III. RESULTS

Figure 1 shows the distribution of dislocation etch pits on the (111) surface of an undoped crystal that was subjected to a maximum resolved shear stress of 20 MPa by three-point bending at 350 °C for 30 s. The magnitude of the applied stress to the central region of the figure was 10 MPa estimated with the distance from the bending knife edges. Large etch pits showing a random distribution are related to the grown-in dislocations. Linear arrays of etch pits emerging from the scratch in both the upward and downward directions are those of dislocations emitted from the scratch. It is easily verified from a geometrical consideration<sup>1-3</sup> that dislocations that have moved in the upward direction are  $\alpha$  dislocations while dislocations that have moved in the downward direction are  $\beta$  dislocations. It is seen that dislocations are generated preferentially from the scratch under applied stress. A critical stress for dislocation generation from a scratch has been observed to exist in the S doped InAs crystal below which no dislocation generation takes place. This is similar to observations by the present author for III-V compounds doped with certain kinds of impurities.<sup>1-3</sup> The critical stress is found to depend on the temperature and the type of generated dislocations, i.e., whether they are of the  $\alpha$  or  $\beta$  type.

Figure 2 shows the critical stresses for generation of  $\alpha$  and  $\beta$  dislocations within InAs doped with S to  $1.3 \times 10^{18} \text{ cm}^{-3}$  as a function of temperature. The critical stress for generation of  $\beta$  dislocations is low at low temperatures and increases with temperatures above 350 °C, while the critical stress for generation of  $\alpha$  dislocations is high at low

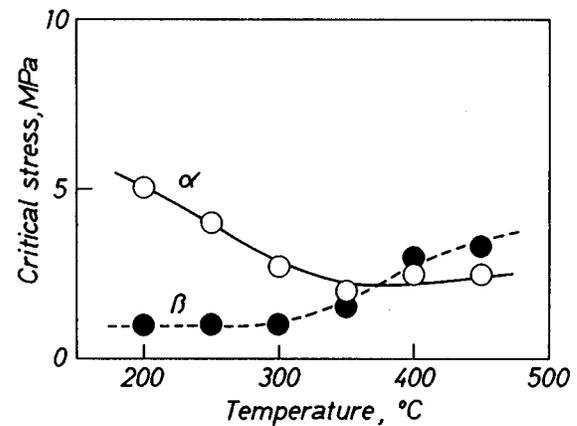


FIG. 2. Temperature dependencies of critical stress for the generation of  $\alpha$  and  $\beta$  dislocations in InAs doped with S at  $1.3 \times 10^{18} \text{ cm}^{-3}$ .

temperature and decreases with increasing temperature. Together with previous investigations for the critical stress of dislocation generation on impurity-doped III-V compounds,<sup>1-3</sup> it can be concluded that the suppression of dislocation generation is caused by the locking of dislocations with impurity atoms. Such critical stress was not observed for dislocation generation in the Zn doped InAs crystal, which might be related to the rather low concentration of impurities to detect a locking effect on dislocations.

The distance traveled by the leading dislocation in an array under a given stress divided by the stressing duration is taken to be the velocity of the dislocation under that stress. The velocity was measured as a function of temperature in the range between 200 and 450 °C, and of the resolved shear stress in the range between 2 and 20 MPa. Figure 3 shows the velocities of  $\alpha$  and  $\beta$  dislocations at 350 °C plotted

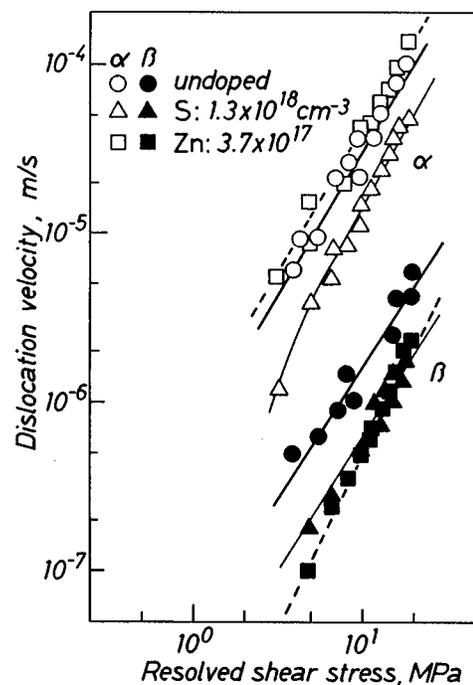


FIG. 3. Velocities of  $\alpha$  and  $\beta$  dislocations in various InAs crystals at 350 °C as a function of the resolved shear stress.

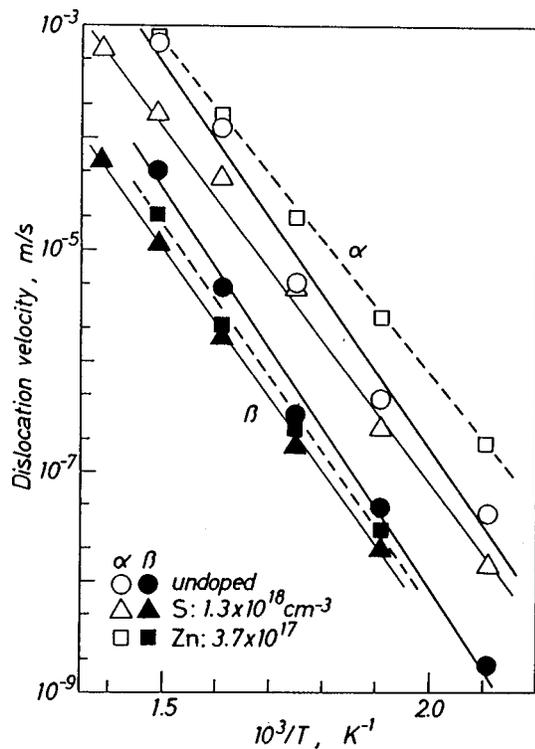


FIG. 4. Velocities of  $\alpha$  and  $\beta$  dislocations in various InAs crystals under a resolved shear stress of 20 MPa plotted against the reciprocal temperature.

against the resolved shear stress within various InAs crystals. In most cases the logarithm of the velocity is linear with respect to the logarithm of the stress for both types of dislocations with almost the same slope. The velocity of  $\alpha$  dislocations in the S doped InAs increases rapidly as the stress increases beyond the critical stress for dislocation generation and seems to show a break at a stress of about 5 MPa. From there, the velocity increases rather slowly with increasing stress at a rate almost the same as that of  $\alpha$  dislocations in the undoped and Zn doped InAs crystals.

Figure 4 shows the velocities of  $\alpha$  and  $\beta$  dislocations under a stress of 20 MPa plotted against reciprocal temperature. The measured velocities of the dislocations in the S doped InAs crystal under such a high stress are thought to be almost free from the influence of the generation stress of dislocations. It is seen that  $\alpha$  dislocations move faster than  $\beta$

dislocations by more than one order of magnitude in the temperature range investigated in all the InAs crystals, which is similar to the results on GaAs and GaP.<sup>1,2</sup>

It is well known in semiconductors that the dislocation velocity is influenced in a variety of ways by electrically active impurities.<sup>1-3</sup> It is seen in Fig. 4 that by doping with S donor, the velocities of  $\alpha$  and  $\beta$  dislocations are reduced compared to undoped InAs. Doping with Zn acceptor gives rise to an increase in the velocity of  $\alpha$  dislocations and, at the same time, a decrease in the velocity of  $\beta$  dislocations. This characteristic of dislocation velocity on the electrical dopants in InAs is dissimilar to that of GaAs,<sup>1</sup> InP,<sup>3</sup> and InSb,<sup>7</sup> being quite different from material to material, even in III-V compounds.

The enhanced velocity of dislocations detected in a semiconductor doped with a certain kind of impurity may be related to the electronic structure of the dislocations and the formation and/or migration of kinks on them as an elementary process of motion. Adopting the idea proposed by Hirsch<sup>8</sup> and Jones<sup>9</sup> for Si, the present author has proposed that a donor level and an acceptor level are associated with a kink on  $\beta$  dislocations within GaAs (Ref. 1) and a kink on  $\alpha$  dislocations within InP,<sup>3</sup> respectively. By analogy, a donor level seems to be related to a kink on  $\alpha$  dislocations within InAs. The detailed mechanism can be clarified by the direct measurement of the kink velocity in the future.

The velocity  $\nu$  of  $\alpha$  and  $\beta$  dislocations in the InAs crystals can well be described with the following empirical equation as a function of the stress  $\tau$  and the temperature  $T$  in the temperature range investigated:

$$\nu = \nu_0 (\tau/\tau_0)^m \exp(-Q/k_B T), \tag{1}$$

where  $\tau_0 = 1$  MPa and  $k_B$  is the Boltzmann constant. The magnitudes of  $\nu_0$ ,  $m$ , and  $Q$  determined for  $\alpha$  and  $\beta$  dislocations in the crystals are given in Table I.

#### IV. DISCUSSION

##### A. Generation of dislocations from a scratch

There exists a critical stress for the generation of dislocations from a scratch in the S doped InAs. The absence of dislocation generation from a scratch under low stresses is observed also for dislocations in GaAs and InP doped with some kinds of impurities as reported by the present author

TABLE I. Magnitudes of  $\nu_0$ ,  $m$ , and  $Q$  for  $\alpha$  and  $\beta$  dislocations in the InAs crystals.

	$\alpha$ dislocations			$\beta$ dislocations		
	$\nu_0$ (m/s)	$m$ ( $\pm 0.1$ )	$Q$ (eV) ( $\pm 0.1$ )	$\nu_0$ (m/s)	$m$ ( $\pm 0.1$ )	$Q$ (eV) ( $\pm 0.1$ )
Undoped InAs $n = 2.0 \times 10^{16} \text{ cm}^{-3}$	$5.6 \times 10^4$	1.7	1.4	$2.8 \times 10^4$	1.6	1.4
S doped InAs [S] = $1.3 \times 10^{18} \text{ cm}^{-3}$	$5.3 \times 10^3$	1.7	1.3	$1.4 \times 10^3$	1.6	1.4
Zn doped InAs [Zn] = $3.7 \times 10^{17} \text{ cm}^{-3}$	$7.8 \times 10^3$	1.7	1.2	$5.2 \times 10^2$	2.0	1.4

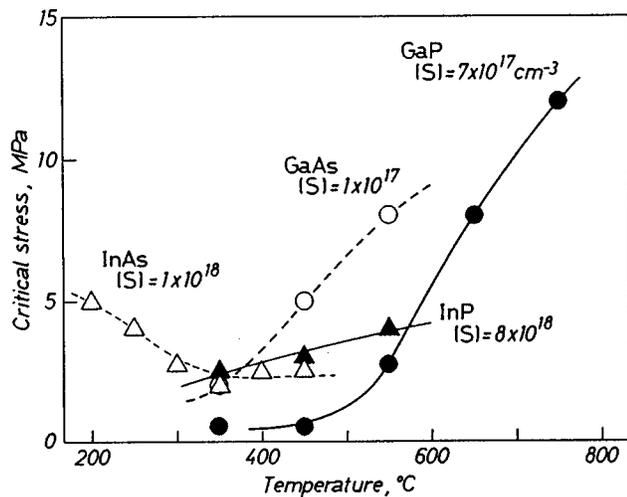


FIG. 5. Temperature dependencies of the critical stress for generation of  $\alpha$  dislocations in GaAs, GaP, InP, and InAs doped with S impurity. Numbers show the concentrations of S impurities in the crystals in units of  $\text{cm}^{-3}$ .

previously.<sup>1,3</sup> This has been interpreted in terms of the immobilization of dislocations due to their being locked by impurity atoms.<sup>1-3</sup> Similarly, we may attribute the existence of a critical stress for dislocation generation in the S doped InAs crystal to the immobilization of dislocations due to impurity segregation along the dislocations.

A comparison of the critical stresses for generation of  $\alpha$  dislocations in GaAs, GaP, InP, and InAs crystals doped with S impurity is shown in Fig. 5, in which the generation stresses are plotted against temperature. As seen in Fig. 5, S impurity suppresses the generation of dislocations strongly in GaAs and GaP, even for lower concentrations than those in InP and InAs. S impurity atoms are known to occupy the As sites in InAs and GaAs while the lengths of In-S and Ga-S bonds are shorter than those of In-As and Ga-As bonds, respectively, by about 5%, and also to occupy the P sites in InP and GaP while the lengths of In-S and Ga-S bonds are shorter than those of In-P and Ga-P bonds, respectively, by about 2%.<sup>10</sup> Thus, even if the same amount of S atoms segregate along dislocations through the elastic interaction due to the size misfit, we expect the suppression effects on dislocation generation to be almost the same in the case of InAs and GaAs, and also for InP and GaP. This is contrary to the experimental observations. Actually, the diffusivity of S atoms in InAs and InP is several orders higher than in GaAs and GaP in the relevant temperature range.<sup>11-13</sup> Thus, S impurity in GaAs and GaP is thought to form agents that are very effective in locking dislocations compared with that in InAs and InP. The above misfit related to a S atom in InAs or InP gives rise to an elastic interaction energy of about 0.5 eV at most between a dislocation and an individual S atom. Such a small magnitude of interaction energy is never effective in suppressing dislocation generation at elevated temperatures.<sup>14</sup>

The observed suppression of dislocation generation in III-V compound crystals is attributed not to the interaction between a dislocation and individual impurity atoms, but to that between a dislocation and some clusters and/or defect complexes including impurity atoms, as concluded in the

previous articles on GaAs, GaP, and InP.<sup>1-3</sup> It is thought that such clusters or complexes are more effectively developed at the core of a dislocation in GaP and GaAs than in InP and InAs. The single bond energy of Ga-S is estimated to be 3.5 eV which is larger than that of In-S (3.0 eV).<sup>15</sup> However, the above difference seems to be not enough to explain the facts. Possibly, some stable structure may be constructed in cooperation with a few impurities, or intrinsic point defects such as P impurity in Si as calculated by Heggie, Jones, and Umerski.<sup>16</sup> A model leading to a satisfactory explanation for the observed facts for various semiconductors needs to be clarified in the future.

## B. Dislocation velocities in III-V compounds and other types of semiconductors

In comparison with the results previously reported by the present author,<sup>1-3</sup> the present results on InAs confirm that the velocities of dislocations in III-V compounds depend strongly on the type of dislocation. The velocity of  $\alpha$  dislocations is higher than those of  $\beta$  and screw dislocations in GaAs,<sup>1</sup> GaP,<sup>2</sup> and InAs, while the velocity of  $\beta$  dislocations is higher than those of  $\alpha$  and screw dislocations only in InP.<sup>3</sup> Commonly, in the III-V compounds, screw dislocations are the slowest. In addition, dislocations in GaAs (Ref. 1) and InP (Ref. 3) are known to be more mobile than in GaP (Ref. 2) and less mobile than in InAs, which has verified the conclusion<sup>2</sup> that in III-V compound semiconductors, dislocations move faster in In-V crystals than in Ga-V crystals, while dislocations move faster in III-As crystals than in III-P crystals.

Most of the experimental data on dislocation velocities in various semiconductors can be expressed by an empirical equation of the type of Eq. (1). Such results are generally interpreted in terms of the double kink mechanism described below. The motion of dislocations in InAs is thought to be controlled also by the same mechanism as those in other semiconductors.

The elementary process of dislocation glide in a semiconductor consists of the thermally activated nucleation of a double kink on a straight dislocation line lying along the Peierls valley and the subsequent expansion of the generated kink pair along the dislocation line. In such a kink mechanism, a dislocation needs the process of bond breaking for motion. Thus, the resistance against the motion of dislocations in covalently bonded materials should be controlled by the need for bonding electron to become excited from the valence band to the conduction band.<sup>17-19</sup>

Figure 6 shows the relation between the measured activation energy  $Q$  for the dislocation motion in Eq. (1) and the band gap energy in a series of III-V compounds. The relation in the elemental semiconductors Si (Ref. 20) and Ge (Ref. 21) and in the IV-IV compound SiC (Ref. 22) and in some II-VI compounds with the sphalerite structure<sup>23-26</sup> is also shown for comparison. The band gap energy is taken at a typical temperature for dislocation motion, i.e., at  $0.6T_m$ , where  $T_m$  is the melting point. Since the band gap energy of HgTe and HgSe are -0.14 and -0.06 eV, respectively, they are taken to be 0 eV in Fig. 6. Though there exists little scatter except in the group of the elemental and IV-IV com-

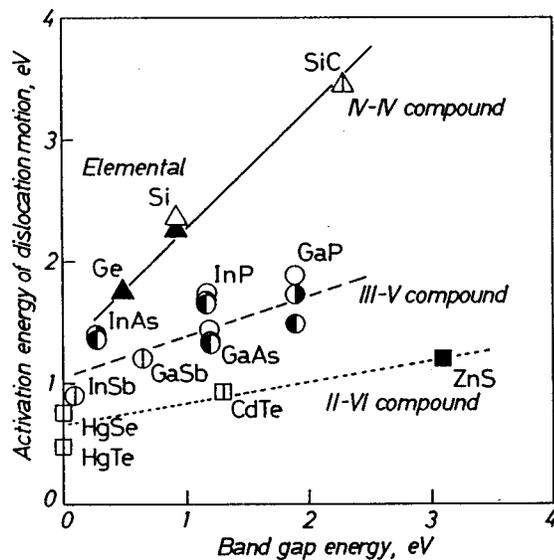


FIG. 6. Activation energy of dislocation motion as a function of the band gap energy at  $0.6T_m$  ( $T_m$ ; melting point) for various types of semiconductor. Triangles, circles, and squares indicate elemental and IV-IV compound [SiC (Ref. 22)], III-V compound and II-VI compound semiconductors [CdTe (Ref. 23), HgTe (Ref. 24), HgSe (Ref. 25), and ZnS (Ref. 26)], respectively. Open, full, left-hand filled and right-hand filled marks are for screw,  $60^\circ$ ,  $\alpha$ , and  $\beta$  dislocations, respectively. The marks with a vertical line inside mean that the activation energy is estimated from the mechanical tests.

pond semiconductors, there is a clear distinction among the groups of semiconductors in the relation between the activation energy  $Q$  and the band gap energy. It is found that the activation energy increases linearly with the band gap energy in each group of semiconductor. The slope of the line is different among the groups of semiconductors and increases in the order of II-VI, III-V compounds, and elemental and IV-IV compound semiconductors. This may be related to the fact that II-VI and III-V compound semiconductors have a comparatively stronger ionicity. The real physical mechanism, however, may be a more complicated, detailed evaluation, which is a task for the future.

From the above phenomenological relation, however, we can deduce some features about the dislocation mobilities within new materials in which it is difficult to measure the dislocation velocities directly. GaN and ZnSe are being developed as a blue laser material and dislocations may affect on the degradation of such devices, but no data on dislocation motion is available. In the relation in Fig. 6, we can estimate the activation energy of dislocation glide to be 2 and 1 eV in GaN (band gap energy: 3 eV) and ZnSe (2.3 eV), respectively. This means that dislocations are more mobile in ZnSe than in GaN, i.e., GaN is less likely to suffer from the motion and multiplication of dislocations leading to degradation than ZnSe.

## V. CONCLUSION

The dynamic behavior of  $\alpha$  and  $\beta$  dislocations in undoped InAs and impurity-doped InAs with S and Zn to a concentration of  $1.3 \times 10^{18}$  and  $3.7 \times 10^{17} \text{ cm}^{-3}$ , respectively, were investigated as a function of stress and tempera-

ture by means of the etch-pit technique. The generation of dislocations from a surface scratch is found to be suppressed in S doped InAs. Suppression is interpreted in terms of dislocation locking due to impurity clusters and/or impurity-defect complexes in comparison with other III-V compound semiconductors.

$\alpha$  dislocations move faster than  $\beta$  dislocations within the undoped and impurity-doped InAs. In undoped materials dislocations move more easily in InAs than in GaP, GaAs, and InP. S donors reduce the dislocation velocities of both  $\alpha$  and  $\beta$  dislocations, while Zn acceptors enhance the velocity of  $\alpha$  dislocations and reduce the velocity of  $\beta$  dislocations. Such features differ from material to material, even in III-V compounds, and seem to be attributed to an electronic state of a dislocation in the elementary process of the motion. The dislocation velocities can be expressed using an empirical equation as a function of stress and temperature similar to that which is ascertained for other semiconductors.

The activation energies for the motion of dislocations within various groups of semiconductors show a linear dependence on the band gap energy. The dependence is different among the elemental and IV-IV compound, III-V compound, and II-VI compound semiconductors. The dislocations in ZnSe are estimated to move more easily than in GaN.

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