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Mechanical Properties and Dislocation Dynamics in III-V Compounds

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Abstract. — The dynamic activities of dislocations and the mechanical properties of various III-V compound semiconductors with a sphalerite structure are reviewed, including current results. Macroscopic stress-strain characteristics and yield strength are described quantitatively in terms of dislocation dynamics on the basis of knowledge of the dynamic behaviour of individual dislocations in these compounds. Various impurities in the compounds affect the mechanical strength through two kinds of effects on individual dislocations: one is the modification of dislocation mobility in glide motion and the other is the immobilization of dislocations. The velocities of rate controlling dislocations during deformation are deduced from analysis of the dynamic state of dislocations.

1. Introduction

The mechanical characteristics of a crystal such as yield behaviour are controlled by the nature of the collective motion of dislocations in the crystal. Extensive studies have been conducted to understand the macroscopically observed plastic behaviour of elemental semiconductors such as Ge and Si in terms of the dynamics of dislocations on a microscopic scale [1–6]. In such an approach the present author reported that the experimentally observed stress-strain characteristics of Si crystals can be described quantitatively using the model of dislocation self-multiplication and dislocation mutual interaction first proposed by Haasen's group [3, 4]. Moreover, the mean velocity and density of dislocations facilitating deformation have been determined for Ge and Si under a constant strain rate [3, 7, 8]. The characteristics of the collective motion of dislocations during deformation have been successfully described quantitatively using the model of Sumino [9].

It is interesting to know whether the mechanical properties of III-V compound semiconductors with a sphalerite structure can be understood in terms of the same dislocation models developed for elemental semiconductor crystals. It seems reasonable to suppose that the mechanical behaviour of III-V compounds will be similar to that of elemental semiconductors to some extent because of the similarities in atomic bonding and crystal structure, although it must be taken into account that basically different types of dislocations with different core

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structures, showing different dynamic characteristics and impurity effect, contribute to deformation simultaneously.

In this paper the characteristics of the mechanical properties of undoped and impurity-doped III-V compounds are reviewed, including updated results. Stress-strain behaviour and yield strength are measured as a function of the temperature and the strain rate, and are described on the basis of knowledge of the dynamic activities of individual dislocations. The yielding phenomenon is derived by calculation based on the model developed for Si [10] and the characteristics of the collective motion of dislocations during deformation are determined, from which the rate controlling dislocation velocities during the deformation are deduced.

2. Dynamic Activities of Dislocations

2.1. $\alpha$, $\beta$ and Screw Dislocations in a III-V Compound. — It is well known that a glissile dislocation in a III-V compound with sphalerite structure is extended into two Shockley partial dislocations bounding a strip of intrinsic stacking fault. In the terms of Hirth and Lothe [11], the dislocation is of the glide set and moves through the narrow-type $\{111\}$ interatomic planes. The width of the stacking fault is typically of the order of 10 nm and increases with increasing ionicity of the III-V compounds [12].

Dislocations are energetically stable when they are lying parallel to $(110)$ directions. Thus, the shape of a stable dislocation loop in a sphalerite structure crystal is hexagonal as schematically drawn in Figure 1a. A dislocation lying along some direction deviated from $(110)$ may be regarded as comprising a number of $(110)$ segments connected by kinks. Any dislocation with an edge component has geometrical dangling bonds along the dislocation core. Actually, such dangling bonds may be reconstructed for energetic reason. Dislocations in III-V compounds having geometrical dangling bonds emerging from group V atoms in the core are termed to be of the $\alpha$-type, while those with geometrical dangling bonds emerging from group III atoms are of the $\beta$-type. A $60^\circ$ dislocation marked $\alpha$ in Figure 1a consists of a $90^\circ$ Shockley partial and a $30^\circ$ Shockley partial both of $\alpha$-type as drawn in Figure 1b. A dislocation segment located on the opposite side of an $\alpha$-type segment in the same hexagonal loop is always of the $\beta$-type. Figure 1c shows that a $60^\circ$ dislocation of the $\beta$-type consists of a $90^\circ$ Shockley partial and a $30^\circ$ partial both of the $\beta$-type. A screw dislocation consists of a $30^\circ$ Shockley partial of $\alpha$-type and a $30^\circ$ Shockley partial of $\beta$-type, as shown in Figure 1d. Different types of dislocations in any compound have different mobilities and show different characteristics in the interaction with impurities.

The characteristics of the morphology of a dislocation in motion in a III-V compound are seen in X-ray topographic images of dislocations in a deformed sample. Since the mobility of an $\alpha$ or a $\beta$ dislocation is remarkably different from that of a screw dislocation, a dislocation loop is usually not hexagonal in shape with well defined $(110)$ directions, but is smoothly curved [13].

2.2. Velocities of Dislocations in Undoped Crystals. — Figure 2 compares the velocities of $\alpha$, $\beta$ and screw dislocations in a series of undoped or low doped III-V compounds under an applied stress of 20 MPa plotted against reciprocal temperature [14-16]. The velocities of $60^\circ$ and screw dislocations in Si are also shown in the figure [17]. It is seen that dislocations in GaAs and InP are more mobile than in Si and GaP and less mobile than in InAs. The dislocation mobilities are approximately the same in GaAs and InP. However, there are two essential differences in the dislocation mobilities between GaAs and InP. Namely, the mobility of $\alpha$ dislocations is higher than those of $\beta$ and screw dislocations by about two orders of magnitude in GaAs while $\beta$ dislocations have a higher mobility than $\alpha$ and screw dislocations
Fig 1. — a) A dislocation loop on a slip plane, and atomic arrangements at the core of (b) α dislocations, (c) β dislocations and (d) screw dislocations in a III-V compound. The shaded regions denote an intrinsic stacking fault. Open circles correspond to group III atoms and solid circles correspond to group V atoms.
Temperature, °C

Fig 2. — Velocities of $\alpha$, $\beta$ and screw dislocations under a shear stress of 20 MPa in a series of undoped III-V compounds plotted against reciprocal temperature. Velocities of 60° and screw dislocations in Si are also shown for comparison.

Table I. — Magnitudes of $v_0$, $m$ and $Q$ for various types of dislocations in undoped or low doped compound semiconductors.

<table>
<thead>
<tr>
<th></th>
<th>$\alpha$ dislocations</th>
<th>$\beta$ dislocations</th>
<th>screw dislocations</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$v_0$ (m s$^{-1}$)</td>
<td>$m$</td>
<td>$Q$ (eV)</td>
</tr>
<tr>
<td></td>
<td>(±0.1)</td>
<td>(±0.05)</td>
<td></td>
</tr>
<tr>
<td>GaP</td>
<td>2.1 $\times$ 10$^2$</td>
<td>1.3</td>
<td>1.45</td>
</tr>
<tr>
<td>GaAs</td>
<td>1.9 $\times$ 10$^3$</td>
<td>1.7</td>
<td>1.30</td>
</tr>
<tr>
<td>InP</td>
<td>4.0 $\times$ 10$^4$</td>
<td>1.4</td>
<td>1.6</td>
</tr>
<tr>
<td>InAs</td>
<td>5.6 $\times$ 10$^4$</td>
<td>1.7</td>
<td>1.4</td>
</tr>
</tbody>
</table>

in InP. In both compounds screw dislocations are the slowest. The difference in the mobilities of $\beta$ and screw dislocations in InP is much smaller than the difference in the mobilities of $\alpha$ and screw dislocations in GaAs.

The velocities of all the types of dislocations in a variety of semiconductors are found to be expressed by the following equation as a function of the stress $\tau$ and the temperature $T$ in the wide ranges of stress and temperature where the crystals are ductile;

$$v = v_0 (\tau/\tau_0)^m \exp(-Q/k_BT),$$

where $v_0$, $m$ and $Q$ are constants, $k_B$ is the Boltzmann constant and $\tau_0 = 1$ MPa. Table I gives the magnitudes of $v_0$, $m$ and $Q$ for all the types of dislocations in undoped or low doped III-V compounds.
2.3. VELOCITIES AND IMMOBILIZATION OF DISLOCATIONS IN IMPURITY DOPED CRYSTALS

Impurity effects on the dynamic activities of dislocations can be divided into two categories: one is the effect on the velocity of dislocations in motion and the other is the immobilization of dislocations when they are moving slowly or at rest.

No generation of dislocations takes place under low stresses in crystals doped with certain kinds of impurities. In such cases the velocity of dislocations is found to be zero in the low stress range and to increase rapidly with the stress once the stress exceeds the critical stress for generation. The rapid increase in the dislocation velocity beyond the critical stress is related to the release process of dislocations immobilized by gettered impurities.

Figures 3a and 3b show how the velocities of $\alpha$, $\beta$ and screw dislocations depend on the electrical properties of impurities in a semi-quantitative manner for GaAs and InP, respectively. The velocities of dislocations shown are those at 450 °C under a stress of 20 MPa, where dislocations are free from the influence of impurity immobilization.

In, P and Al isovalent impurities in GaAs or Ga and As in InP do not give appreciable influence on the velocities of all types of dislocations in motion even when their concentrations are in the order of $10^{19}$–$10^{20}$ cm$^{-3}$. However, the donor impurities Si, S and Te drastically reduce the mobilities of all types of dislocations in GaAs. Conversely, S and Sn enhance the mobility of $\alpha$ dislocations while reducing the mobilities of $\beta$ and screw dislocations in InP. As a consequence, the $\alpha$ dislocation is the fastest among all types of dislocations in InP heavily doped with donor impurities. Contrarily, the acceptor impurity Zn enhances the mobilities of $\beta$ and screw dislocations while reducing the mobility of $\alpha$ dislocations in GaAs. Zn reduces
drastically the mobilities of all types of dislocations in InP as donor impurities in GaAs do. The velocities of dislocations in various impurity doped III-V compounds can also be expressed by equation (1).

It is shown theoretically that the interaction between a dislocation and an individually dispersed impurity atom never gives rise to an appreciable effect on the velocity of the dislocation moving at an elevated temperature [18]. In and Al isovalent impurities show no effect on the velocities of all types of dislocations, and seem to be dispersed individually. The impurities which give rise to an appreciable retarding effect on the dislocation motion in GaAs or InP are in the state of cluster or complex for which the energy of interaction with a dislocation is higher than about 3 eV. S, Si and Te impurities in GaAs and Zn impurities in InP are thought to be in such a state.

The enhancement of dislocation velocities may be related to the electronic structure of the dislocations. Adopting the idea proposed by Hirsch [19] and Jones [20] for Si, if a kink in the stable configuration on a 30°β Shockley partial dislocation in GaAs accompanies some donor level, the thermal equilibrium concentration of kinks on the dislocation may increase in a p-type crystal since the free energy of the system decreases with doping of acceptor impurities. If some donor level is associated with the saddle point configuration of a kink on a 30°β partial in GaAs, the kink motion along the dislocation is enhanced since the height of the potential barrier for the kink motion is reduced. In both cases the velocity of β dislocations in GaAs is increased by doping of acceptors as observed experimentally in Figure 3. Enhancement of the mobility of α dislocations in InP due to doping with donor impurities is accounted for if we assume that a kink in the stable state or in the saddle point configuration on a 90°α partial dislocation accompanies some acceptor level.

Though immobilization of dislocations has been found mainly in In doped GaAs by some groups [18, 21], such strong immobilization can be observed in many cases as for β and screw dislocations in Zn doped GaAs, for α dislocations in Te doped GaAs, and for dislocations in Zn and S doped InP, etc. [14, 16, 18]. The strength of dislocation immobilization due to impurity gettering depends on the diffusivity of impurities, the concentration of impurities, the reactivity of impurities at the dislocation core and the type of dislocation [14–16]. In GaAs, In impurities immobilize selectively α and screw dislocations, and Al impurities selectively immobilize β dislocations. Since both impurities belong to group III elements and the strength of immobilization is not related to the covalent radii of the impurities, the immobilization effect on dislocations cannot be interpreted in terms of Cottrell cloud locking by impurities. Probably, some kind of reaction incorporating impurity atoms takes place at the dislocation core. The dislocations are immobilized by pinning at an impurity cluster and/or complexes.

3. Mechanical Characteristics of III-V Compounds

3.1. Undoped Crystals

3.1.1. Stress-Strain Characteristics. — Figures 4a and 4b show stress-strain curves of GaAs and InP crystals, respectively, at various temperatures measured in compressive deformation along the [123] direction under a shear strain rate of 2 × 10^{-4} s^{-1} [16, 22]. The dislocation densities of the crystals prior to deformation tests are in the range 10^3–10^5 cm^{-2}. InP crystal shows stress-strain characteristics similar to those of GaAs approximately in the same temperature range.

The curves of both crystals, especially at lower temperatures, are characterized by a remarkable stress drop after yielding followed by a gradual increase in the stress with the strain due to work hardening. The magnitude of the upper yield stress and the amount of stress drop after
yielding both depend very sensitively on the temperature. The stress drop becomes small or is absent as the temperature is raised. The overall stress-strain characteristics of both crystals depend sensitively on the temperature and the strain rate. A decrease in the strain rate brings about the same effect as an increase in the temperature. Such characteristics of stress-strain curves are commonly observed in the elemental semiconductors Ge and Si [3, 7, 8]. Here, it should be noted that the InP crystal shows a much more remarkable stress drop after yielding than the GaAs crystal.

The stress-strain behaviour of all the III-V compounds does not depend systematically on the density of grown-in dislocations contrary to the elemental semiconductors [3]. This implies that grown-in dislocations in III-V compounds are essentially immobile during deformation of the crystals. It seems that the surfaces of crystals of III-V compounds involve some irregularities that act as very effective generation centres for dislocations.

Figures 5a and 5b show the magnitudes of the upper yield stress and lower yield stress, respectively, in a series of undoped or low doped III-V compounds plotted against the reciprocal temperature of deformation at a shear strain rate of $2 \times 10^{-4}$ s$^{-1}$. The temperature ranges where different compounds show similar mechanical strengths correspond well with the temperature ranges where they have similar dislocation mobilities as seen in Figure 2. This reflects the fact that the mechanical behaviour of any compound is basically controlled by the dislocation mobility.
Fig. 5. — Magnitudes of (a) the upper yield stress $\tau_{uy}$ and (b) the lower yield stress $\tau_{ly}$ in various III-V compounds for deformation under a shear strain rate of $2 \times 10^{-4}$ s$^{-1}$ plotted against reciprocal temperature.

The upper and lower yield stresses, denoted here collectively by $\tau$, of any crystal are well expressed as a function of the temperature $T$ and the strain rate $\dot{\varepsilon}$ by the following empirical equation:

$$\tau = A\dot{\varepsilon}^{1/n} \exp(U/k_B T),$$  \hspace{1cm} (2)

where $A$ is a constant that depends on the crystal. The magnitudes of $n$ and $U$ for the upper and lower yield stresses in various crystals are given in Table II.
Table II. — Magnitudes of \( n \) and \( U \) for the upper yield stress \( \tau_{uy} \) and the lower yield stress \( \tau_y \) and the steady state values of the effective stress \( \tau_{eff} \) in various III-V compound semiconductors.

<table>
<thead>
<tr>
<th></th>
<th>( \tau_{uy} )</th>
<th>( \tau_y )</th>
<th>( \tau_{eff} )</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>( n )</td>
<td>( U ) (eV)</td>
<td>( n )</td>
</tr>
<tr>
<td>GaP</td>
<td>2.8</td>
<td>0.68</td>
<td>2.9</td>
</tr>
<tr>
<td>GaAs</td>
<td>3.6</td>
<td>0.46</td>
<td>4.7</td>
</tr>
<tr>
<td>GaSb</td>
<td>4.8</td>
<td>0.50</td>
<td>4.1</td>
</tr>
<tr>
<td>InP</td>
<td>2.7</td>
<td>0.63</td>
<td>3.2</td>
</tr>
<tr>
<td>InAs</td>
<td>3.6</td>
<td>0.50</td>
<td>3.3</td>
</tr>
<tr>
<td>InSb</td>
<td>2.1</td>
<td>0.45</td>
<td>2.2</td>
</tr>
</tbody>
</table>

3.1.2. Effective Stress. — The effective stress \( \tau_{eff} \) which is necessary to keep dislocations moving at a certain velocity was measured by the strain-rate cycling technique. The variations in the effective stress with the strain are shown in Figure 4 for undoped GaAs [22] and InP [23]. The effective stress in any crystal is constant with respect to the strain in the deformation stage after the lower yield point, which means that the mean velocity and the density of moving dislocations are both constant with respect to the strain in the deformation stage. Such dynamic state of moving dislocations has been termed the steady state of deformation, and has already been found to be realized in Ge and Si [3, 7, 8]. A steady state is realized also in other III-V compounds [23, 24].

The steady state value of the effective stress has been found to depend on the temperature and the strain rate. Figure 6 shows the dependence of the steady state values of the effective stress \( \tau_{eff} \), in a series of undoped or low doped III-V compounds, on the temperature for deformation under a strain rate of \( 2 \times 10^{-4} \text{ s}^{-1} \). The temperature and strain rate dependencies of the steady state value of the effective stress of any crystal are again well expressed by a relation of type equation (2). The magnitudes of parameters \( n \) and \( U \) for various crystals are given also in Table II.

3 2. IMPURITY DOPED CRYSTALS. — The general features of the stress–strain curves of GaAs crystals doped with various kinds of impurities are similar to those of undoped crystals in the temperature range lower than about 600 °C [25]. The magnitude of the upper yield stress and the stress drop after yielding become large with an increase in the concentration of Si impurity doped in GaAs. The characteristics of the stress-strain curve of GaAs doped with In or Zn at a high concentration are similar to those of the undoped GaAs. Contrarily, for deformation at high temperatures under a low strain rate the stress-strain curves of GaAs crystals doped with In or Zn impurities show no stress drop after yielding and many fine serrations, known as the Portevin-Le Chatelier phenomenon, have been observed in Si and GaAs doped with a high concentration of a certain kind of impurity [26, 27]. Figure 7 shows the upper yield stress dependency on temperature in the deformation of impurity doped GaAs under a shear strain rate of \( 2 \times 10^{-4} \text{ s}^{-1} \). The temperature dependence of the upper yield stress of impurity doped GaAs is strong at low temperatures but weak at high temperatures, except for GaAs doped with Si at rather low concentrations.

In the low temperature region, the upper yield stress is the highest in GaAs doped with Si at a concentration of \( 4 \times 10^{18} \text{ cm}^{-3} \), while those of GaAs doped with In or Zn at a concentration of \( 2 \times 10^{20} \) or \( 2 \times 10^{19} \text{ cm}^{-3} \) are close to that of undoped GaAs. The yield stress of Si doped GaAs
increases with increasing concentration of Si impurity. The yield stresses of GaAs doped with Zn or In at concentrations of \(2 \times 10^{19}\) or \(2 \times 10^{20} \text{ cm}^{-3}\) are lower than those of the GaAs doped with Si at a concentration as low as \(1 \times 10^{17} \text{ cm}^{-3}\). The yield stresses depend strongly on the strain rate. As seen in the previous section, Si donor impurities strongly reduce the velocities of all the types of dislocations, while isovalent impurities have no appreciable effect on them. The reduction of the velocities of dislocations results in an increase of the yield stresses as seen in Figure 7. Though Zn impurities enhance the velocities of \(\beta\) and screw dislocations, the effect is compensated by the reduction of the velocities of \(\alpha\) dislocations, the result of which is that the yield stress of Zn doped GaAs is nearly equal to that of undoped GaAs. Thus, we can conclude that in the low temperature range the effect of the impurities on the dislocation velocity in glide motion plays an essential role in determining the mechanical strength of the crystal, which Si impurities are most effective at enhancing yield and flow stress.

Contrarily, in the high temperature region, the yield stress of In doped GaAs is higher than that of the Si doped GaAs, and the yield stress of Zn doped GaAs and Si doped GaAs are comparable. The yield stress of In or Zn doped GaAs depends very weakly on the strain rate in the high temperature region as observed in Si doped with a high concentration of P [26]. Such anomalous temperature dependence of the yield stress in In or Zn doped GaAs may be related to some kind of interaction of moving dislocations with In and Zn impurities at high temperatures. Indeed, In and Zn impurities have a strong pinning effect on \(\alpha\) or \(\beta\) dislocations as well as on screw dislocations, as seen in the previous section. Thus, we can conclude that the mechanical strength of GaAs doped with certain kinds of impurities at high temperatures is controlled by dislocation locking by impurities, and that In impurities are most effective in this respect.

A similar feature of stress-strain characteristics to that in impurity doped GaAs is also seen in impurity doped InP [16]. Zn doped InP shows yield stresses higher than those of undoped or S doped InP over the whole temperature range investigated, since Zn impurities retard the

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Fig. 6 — Magnitudes of the steady state value of the effective stress \(\tau_{\text{eff}}\) in various III-V compounds for deformation under a shear strain rate of \(2 \times 10^{-4} \text{ s}^{-1}\) plotted against reciprocal temperature.
velocities of all the types of dislocations while S impurities enhance the velocity of $\alpha$ dislocations and reduce the velocities of $\beta$ and screw dislocations. The enhancement of the yield strength by dislocation immobilization due to impurity locking is observed at high temperatures in Zn and S doped InP.

4. Dislocation Dynamics in Deformation of III-V Compounds

4.1. Derivation of Stress-Strain Characteristics. — Both the deformation characteristics of III-V compounds and the dynamic state of dislocations during macroscopic deformation are similar to those known in elemental semiconductors such as Si and Ge. Thus, the plastic deformation of all III-V compounds is seen to be controlled by dislocation processes essentially the same as those operating in elemental semiconductors. Hence, we can deduce the stress-strain characteristics, especially yielding phenomenon by a model based on the dislocation dynamics. Some results have been given in reference [10].

Plastic deformation of a crystal proceeds by means of the motion of a fairly high density of dislocations. The flow stress $\tau_a$ of a crystal is the stress needed to move dislocations within the crystal at a certain velocity which is determined by the strain rate and the density of dislocations in motion. The stress consists of two components: one is the effective stress $\tau_{\text{eff}}$ and the other is the internal stress $\tau_i$ which is the component to overcome the resistance originating from the interaction between dislocations,

$$\tau_a = \tau_{\text{eff}} + \tau_i,$$

$$\tau_i = \frac{G b N^{1/2}}{\beta},$$

where $G$ is the shear modulus, $b$ is the magnitude of the Burgers vector, $N$ is the density of dislocations in the crystal and $\beta$ is an interaction parameter of the order of 3 to 4.
In a sphalerite crystal three different types of dislocations which have remarkably different mobilities from each other contribute to plastic deformation. As shown in the previous section, in most III-V compounds $\alpha$ dislocations move at velocities orders of magnitude higher than those of other types of dislocations. Conversely, $\beta$ dislocations move faster than other types of dislocations in InP. In such cases the deformation rate of a crystal is controlled by the motion of screw dislocations and the plastic strain rate $\dot{\epsilon}_{\text{pl}}$ of a crystal is given by the following equation, from the geometrical consideration of a dislocation loop if only dislocations on the primary slip system are active:

$$\dot{\epsilon}_{\text{pl}} = 2N_s \bar{v}_s \delta,$$  

(5)

where $N_s$ is the density of screw dislocations, being almost equal to the density of all the moving dislocations, and $\bar{v}_s$ is the mean velocity of screw dislocations [22].

The secondary slip systems also operate during crystal deformation. Of importance in these compounds is that the faster type of dislocations, such as $\alpha$ dislocations in GaAs, of the secondary slip system move much faster than screw dislocations of the primary slip system under any applied stress as deduced from their velocities in Figure 2. Thus, the role of dislocations of the secondary slip system is much more significant in comparison with elemental semiconductors.

The plastic strain rate of a crystal with respect to the primary system may be modified as:

$$\dot{\epsilon}_{\text{pl}} = 2(N_s^{\alpha} \bar{v}_s^{\alpha} + \gamma N_s^{\beta} \bar{v}_s^{\beta}),$$  

(6)

where $N_s^{\alpha}$ and $\bar{v}_s^{\alpha}$ are the density and mean velocity of screw dislocations in the primary slip system, respectively, and $N_s^{\beta}$ and $\bar{v}_s^{\beta}$ are those of the secondary system. The parameter $\gamma$ is a geometrical factor related to the Schmid factor of the secondary slip system and is smaller than unity.

The mean velocity $\bar{v}_s^{\alpha}$ and $\bar{v}_s^{\beta}$ are obtained by substituting the following effective stresses on the primary and secondary slip systems into equation (1). The effective stresses on the primary and secondary slip system are given

$$\tau_{\text{eff}}^{\alpha} = \tau_{\alpha}^{\alpha} - GbN_s^{\alpha} \bar{v}_s^{\alpha} / \beta - GbN_s^{\beta} \bar{v}_s^{\beta} / \beta^*,$$  

(7)

$$\tau_{\text{eff}}^{\beta} = \tau_{\alpha}^{\beta} - GbN_s^{\alpha} \bar{v}_s^{\alpha} / \beta - GbN_s^{\beta} \bar{v}_s^{\beta} / \beta^*,$$  

(8)

where $\tau_{\alpha}^{\alpha}$ and $\tau_{\alpha}^{\beta}$ are the resolved shear stresses of the applied stress on the primary and secondary slip system, respectively, and $\beta > \beta^*$.

Dislocations multiply generally from the centres as jogs located on the screw part of dislocations. Such jogs originate from the cutting with dislocations of other slip systems. Thus, the density of such multiplication centres is proportional to the density of screw dislocations of the relevant slip system. Further, screw dislocations are generated through the motion of the fastest type of dislocations. Such dislocations are $\alpha$ dislocations in GaAs and $\beta$ dislocations in InP. Thus, the multiplication rates of screw dislocations of the primary and secondary slip systems may be given by the following equation:

$$\frac{dN_s^{\alpha}}{dt} = K N_s^{\alpha} \bar{v}_s^{\alpha} \tau_{\text{eff}}^{\alpha} + K^* N_s^{\alpha} N_s^{\beta} \bar{v}_s^{\beta} \tau_{\text{eff}}^{\beta},$$  

(9)

$$\frac{dN_s^{\beta}}{dt} = K N_s^{\beta} \bar{v}_s^{\beta} \tau_{\text{eff}}^{\beta} + K^* N_s^{\beta} N_s^{\alpha} \bar{v}_s^{\alpha} \tau_{\text{eff}}^{\alpha},$$  

(10)

where the subscript “f” stands for the fastest type of dislocations. The parameter $K$ in equations (9) and (10) is a multiplication factor of dislocations in the each slip system, while $K^*$ is a multiplication factor through the intersection with the other slip system.
STRENGTH AND DISLOCATION DYNAMICS IN III-V

Fig. 8. — Calculated stress-strain curves of (a) GaAs and (b) InP which compare well with Figures 5a and 5b, respectively.

The magnitudes of $v_0$, $m$ and $Q$ for screw dislocations can be deduced from the experimentally obtained magnitudes of $n$ and $U$ for the effective stress $\tau_{\text{eff}}$ in the steady state on the basis of
Table III. — Magnitudes of $m^*$ and $Q^*$ determined from the effective stress in the steady state of deformation.

<table>
<thead>
<tr>
<th></th>
<th>$m^*$</th>
<th>$Q^*$ (eV)</th>
</tr>
</thead>
<tbody>
<tr>
<td>GaP</td>
<td>1.0</td>
<td>2.2</td>
</tr>
<tr>
<td>GaAs</td>
<td>1.6</td>
<td>1.4</td>
</tr>
<tr>
<td>GaSb</td>
<td>1.1</td>
<td>1.1</td>
</tr>
<tr>
<td>InP</td>
<td>1.2</td>
<td>1.2</td>
</tr>
<tr>
<td>InAs</td>
<td>1.4</td>
<td>1.4</td>
</tr>
<tr>
<td>InSb</td>
<td>1.1</td>
<td>0.8</td>
</tr>
</tbody>
</table>

Fig. 9. — Velocities of screw dislocations in III-V compounds under a shear stress of 10 MPa plotted against reciprocal temperature.

The magnitudes of $m^*$ and $Q^*$ determined for III-V compound semiconductors are given in Table III. It is seen that the values of $m^*$ are in the range of 1.0–1.6 and those of $Q^*$ in the range of 0.8–2.2 eV.

Figure 9 shows the velocities of screw dislocations under a shear stress of 10 MPa plotted against reciprocal temperature as deduced from equation (11) in III-V compounds. The deduced velocities of dislocations in GaAs, InP, GaP and InAs agree satisfactorily with those determined directly by the etch pit technique.
Now we are able to determine the velocity equation of the dislocations which are rate controlling in the deformation of III-V compounds, except where dislocation immobilization due to impurity gettering does not occur. The procedure may be applied to estimate the dislocation velocity in crystals with a high density of grown-in dislocations or at high temperatures, where it is usually difficult to make direct measurements of the dislocation velocity.

5. Conclusions

Mechanical properties and dislocation dynamics have been investigated in undoped and impurity doped III-V compound semiconductors on the basis of knowledge of the dynamic activities of individual dislocations.

The stress-strain characteristics of undoped compounds are controlled by dislocation processes which are essentially the same as those in elemental semiconductors of Si and Ge. The yield strength of undoped III-V compounds increases with a decrease of the mobilities of individual dislocations in them. The mechanical strength is influenced by impurities in the compounds through their effects on individual dislocation. The effect of impurities on the dislocation velocities in glide motion plays an essential role in determining the mechanical strength of the crystals at low temperatures. The immobilization of dislocations by dislocation locking by certain kinds of impurities controls the mechanical strength of the crystal at high temperatures.

Stress-strain curves are derived successfully by calculation on a theoretical model with the experimentally obtained mobilities of dislocations and a consideration of the activation of dislocations of the primary and secondary slip systems. The velocities of rate controlling dislocations during deformation are deduced as a function of stress and temperature.

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