Effect of reversal of the dislocation glide direction on their mobility properties in germanium single crystals

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(Submitted 12 October 1981)
Zh. Eksp. Teor. Fiz. **82**, 1539–1550 (May 1982)

It is found that reversal of the dislocation gliding direction in germanium alters significantly the dependence of the dislocation velocity on the applied external stresses. It leads also to more than a twofold decrease of the effective activation energy of their motion compared with that of the initial crystal. The experimental results are compared with current theories of dislocation mobility in a deep Peierls potential relief. It is concluded that the asymmetry of the dislocation mobility may be due to a peculiar rearrangement of the point-defect structure during the course of the dislocation motion.

PACS numbers: 66.30.Lw

An almost unanimous opinion held until quite recently[1] is that semiconductor single crystals with covalent interatomic interaction forces are most highly promising for the experimental study of elementary shearing processes limited by the resistance of the crystal lattice proper and caused by gliding of perfect dislocations. Recent investigations[2] of the dislocation mobility in germanium and silicon single crystals have revealed a number of disparities between the experimental facts and the predictions of the theory concerning this fundamental problem, which stimulated in its time the establishment of modern concepts in the physics of material strength and plasticity.

Detailed investigations of the dependence of the dislocation velocity in a semiconductor on the stresses, on the temperature, and on the impurity density have shown[3-10] that if no account is taken of the action of point defects on the nucleation[11-14] of double kinks and their moving apart[14-16] it is impossible to explain the entire aggregate of the experimental data obtained even for the purest of the available real crystals. It was also convincingly demonstrated that the dislocations produced in these materials by plastic deformation are extended[17,18] and the crystals themselves are of record purity only with respect to the electrically active impurities of the third and fifth groups.

All this has raised a number of highly debatable points in the solution fundamental problems not only in the investigations of the dynamic properties of dislocations. These points are reflected also in investigations of the influence of dislocations on the energy spectrum of the carriers in semiconductors, in view of observations[19,20] of radical changes in the character of the electric activity of dislocations when the length and temperature of the heat treatment is varied in the course of plastic deformation of the crystal. These reduce to a separation of the influence of extension of the dislocations and of the state of the crystal with respect to point defects on the laws governing the dislocation motion and on the mechanisms whereby they influence the electronic properties of semiconductors.

We have therefore undertaken in the present study an investigation of the mobility of dislocations in germanium single crystals, upon inversion of their glide direction, as a function of the applied tangential stresses τ and of the temperature T. No such experiments have been performed to date on single crystals of semiconductors. It was hoped (for reasons considered in detail in the discussion of the results) that an analysis of the peculiarities of the dislocation motion over the already swept part of the glide plane will yield useful information for the study of the influence of the aforementioned factors under conditions when the action of one of them is decisive.

EXPERIMENT

The investigated samples were rectangular prisms with edge orientations [110], [112], and [111] and with dimensions 3.5×0.45×0.15 cm. They were cut from dislocation-free single crystals of n- and p-type germanium, grown by the Czochralski method, with antimony and gallium densities ~1.4×10^{13} and ~10^{14} cm^{-3}, respectively. The loading was by four-point flexure. The flexure axis of the sample coincided with the [112] direction. The initial procedure of introducing the dislocations was traditional.[21] Dislocation half-loops of semi-hexagonal shape were produced near a specially made by a scratch on the $(\bar{1}11)$ surface following the first loading of the sample. Sections of these half-loops, with 60° or screw orientation, crossed the $(\bar{1}11)$ surface at an angle 57°. They were revealed in the mass experiments by selective chemical etching. Some of the samples were investigated by x-ray diffraction.

The dislocations produced were displaced by the second loading in the "forward" direction, i.e., with the diameter of the half-loop increased to 600–1000 μm. After cooling, the crystal was again etched. An experiment with reversed motion of the dislocations was then carried out from this starting position.

The direction of the dislocation motion was reversed by two different methods. In the first, the sample was cooled after the forward displacement of the dislocations from the starting position, and the dislocations in the sample were revealed by selective etching. The sample was then turned over in the leading unit so that the faces subjected to compression and tension exchanged places, and was deformed again after heating to the required temperature. The reversal of the signs

 0038-5646/82/050891-07$04.00

of the stresses acting on the dislocations caused them to move in the "backward" direction, i.e., so that the diameters of the half-loops decreased. In this reversal method the dislocation velocities in the forward and backward motion were determined directly from the measurements of the distance between the etch pits that marked the positions of the dislocation before and after application of the load.

The second method made possible reversal of the dislocation motion without intermediate cooling and heating prior to the reversal of the sign of τ, by using a loading unit with six supports. The dislocations introduced in the starting position were moved first in the forward direction by the stresses produced by a load applied to the pair of internal supports. The transfer of the loading to the outermost supports, which was carried at the test temperature within several seconds, reversed the sign of the bending moment and of the direction of the dislocation displacement. The crystal was then cooled, and the final position of the dislocation was revealed by selective etching. The average range of the dislocations in the backward motion was calculated in this case from the measured distance between etch pits that marked the initial and final positions of the dislocation, with allowance for the average range of the dislocations in the forward direction, as determined from preceding experiments with cooling prior to reversal. Investigations[22] have shown that the dislocation range was linearly connected with the time of application of the load, both in the case of forward and backward motion, regardless of the sign of the acting stress, i.e., it was possible to determine from the measured range the rate of the stationary motion of the dislocation. The employed method of reversing the load made it possible to investigate dislocation mobility in the temperature interval 400–580℃.

It was noted earlier[23] that half-loops introduced in germanium can contract spontaneously, when annealed at high-temperature, under the influence of image forces and linear tension. Annealing at 400 and 500 °C of the crystal samples used by us, with starting dislocations that formed half-loops of large diameter, $\gtrsim 600\ \mu$m, has shown that the contracting stresses did reverse the motions of these dislocations within times comparable with the time of heating, cooling, and deformation.

EXPERIMENTAL RESULTS

When the signs of the stresses acting in the glide plane were reversed, it was observed that the rate of narrowing of the dislocation half-loops can exceed substantially (by two orders of magnitude) the rate of their expansion. A change of the dislocation mobility upon reversal was observed both for the 60° and screw parts of the half-loops, regardless of whether they were in initially compressed or initially stretched sections of the samples. The difference between the forward and backward dislocation velocities depended on the temperature and on the acting stresses, as well as on the conditions of the heat treatment of the crystal in the reversal process.

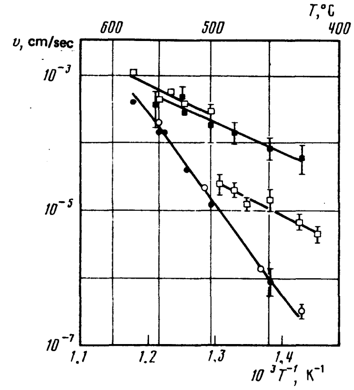

FIG. 1. Temperature dependence of the velocity of 60° dislocations gliding under the influence of a stress $\tau = 0.5\,\text{kgf/mm}^2$ in the forward direction (●), in the backward direction after intermediate cooling (□) and without intermediate cooling (■). Curve (○) corresponds to motion over the twice-swept part of the glide plane.

Figure 1 shows the temperature dependences of the velocities of 60° dislocations gliding under the influence of the same stress $\tau = 0.5$ kgf/mm^2 in the forward and backward directions. The rate of expansion of the dislocation half-loops (black circles), as in the preceding studies,[3,5,6,8] were characterized by a high dislocation-gliding activation energy of the $U \approx 2.6$ eV. In backward motion without intermediate cooling (black squares), the activation energy decreased to less than one-half, to ~1 eV. Intermediate heat treatment prior to reversal, consisting of cooling the sample to room temperature and heating it to the test temperature, led to a nonmonotonic dependence of $\log v$ on $1/T$ (light squares). The dislocation velocity v changed abruptly in a narrow temperature interval 490–500℃, above and below which the dependence of v on T can be approximated by an exponential function

$$v \sim v_0 \exp(-U/kT)$$

with different values of v_0 for $T > 500$℃ and $T < 500$℃, and with the same activation energy $U \approx 1$ eV for these temperature intervals. The velocities of the backward-moving dislocations after reversal without intermediate cooling did not differ (within the limits of experimental error) from the velocities measured after the intermediate heat treatment in the temperature interval 500–580℃, while at $T < 500$℃ they exceeded these values substantially.

Following a second reversal, i.e., after a second expansion of the dislocation half-loop, its motion over the twice-swept part of the glide plane is characterized by velocity and activation-energy values that practically coincide with the values obtained in the first motion in the forward direction (light circles). This result seems to reveal an asymmetry in the mobility of individual dislocations under condition of reversible motion over one glide plane.

The dependence of the backward-moving dislocation velocity on the stress also differed from that obtained in forward motion for the same temperature inverval.

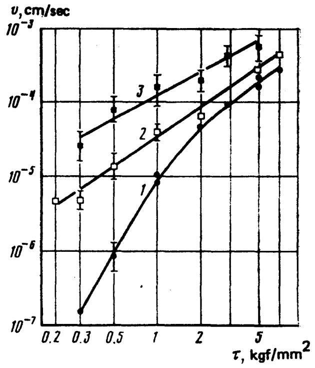

FIG. 2. Dependence of the velocity of 60° dislocations on the stress at $T = 450°$ C in forward motion (1), in backward motion after intermediate cooling (2) and without intermediate cooling (3).

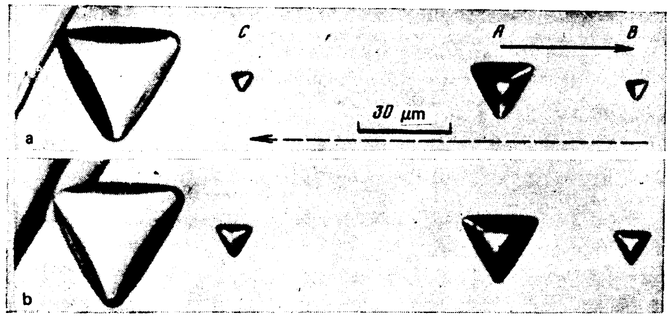

FIG. 3. a) Arrangement of etch pits in a Ge sample etched after reversal of the glide direction of a 60° dislocation at $T = 450$ °C and $\tau = 0.5$ kgf/mm^2. The large flat-bottom pit is located at the nucleation source, the time of motion is 90 min from the starting position A to the turning point B and 3 min to the final position C on the left. The arrows show the directions of motion of the dislocation. b) Picture observed after further etching.

Figure 2 shows the experimental data, plotted in $\log v$ and $\log \tau$ coordinates coordinates, for the forward and backward motions (and for both methods for reversing the load at $T = 450$ °C). The $v(\tau)$ dependence for a dislocation moving in the forward direction could be described in the investigated temperature interval, just as in Refs. 3, 5, 6, and 8, by the power law $v \sim \tau^m$ with $m \approx 3$ at $\tau < 2$ kgf/mm^2 and $m \approx 1.3$ at $\tau > 2$ kgf/mm^2. When the dislocation moved in the direction of the narrowing of the half-loop, no section with an abrupt dependence of v on τ was observed on the plot, which can then be approximated by a power-law function with $m \approx 1.3$ in the entire investigated range of stresses.

The difference between the dislocation velocities in the forward and backward directions depends on the conditions of the introduction of the dislocations in the starting position. The asymmetry of the mobility of the dislocations vanishes if they move in the backward direction over the same section of the glide plane that was swept in the course of starting at high temperature $T \approx 705$ °C and at velocity $v \approx 10^{-1}$ cm/sec. The asymmetry reappears for the same dislocations only if they are displaced into a crystal region that has not been subjected to shear before, at a lower temperature $T \approx 450$ °C with low velocity $v \approx 10^{-5}$ cm/sec in the forward direction, and then in the backward direction.

Etching the samples after reversal, without intermediate cooling, revealed an unusual change in the real structure of the crystal, due to the change in the direction of the dislocation glide. At low stresses and low temperatures, i.e., when the backward velocity exceeds the forward substantially, an acute-angle pyramidal pit was etched at the dislocation turning point [Fig. 3(a)]. With increasing etching time, this pit became flat-bottomed [Fig. 3(b)]. In these figures, the large flat-bottom pit marks the position of the dislocation at its initiation point. The smaller flat-bottom pit A corresponds to the starting point, and two acute-angle pyramidal pits C and B fix respectively the final position of the dislocation and its turning point upon inversion of the glide direction. The distance AB is equal to the range measured in preliminary experiments, when the dislocation was made to move in the forward direction under the same experimental conditions. A brief annealing at the test temperature (30 min) failed to reveal the pit at the point B after cooling. Nor was the pit observed at the place where the dislocation velocity decreased upon reversal of the glide direction (on going from contraction to expansion of the half-loop).

An x-ray diffraction investigation has shown that the etch pits C and B are not due to a stacking fault, and no new dislocations were introduced upon reversal of the load. Therefore the appearance of the pyramidal etch pit at the turning point can be due only to the fact that the dislocation has left behind part of its point-defect atmosphere. Out of this atmosphere there was formed a one-dimensional aggregate of impurity complexes, which caused local changes in the chemical potential and selective etching of the sample at their location. An investigation of the current-voltage characteristics of clamped contacts of germanium with thin tungsten needles has shown that this defect, just as a 60° dislocation, exhibits acceptor action in n-Ge. This is evidence that the point defects left after the turning of the dislocations are not the principal doping impurities in the given crystal.

DISCUSSION OF RESULTS

The possibility of distinguishing between the glide processes in opposite directions was first discussed as applied to metals with bcc lattices quite some time ago.[24] The idea of the polarity of the glide in these materials stemmed from an analysis of the dependence of the yield point on the sample orientation, on the temperature, and on the type of stressed state. It was assumed that it is due to singularities in the structure of the core of the screw dislocation in the bcc crystals, the motion of which calls for a unique re-extension of the core. It is well known that the asymmetry of the mobility of dislocations is by far not the only possible cause of the onset of anisotropy of the macroscopic characteristics of plastic deformation in a crystal. This deformation can also be singularities in the multiplication of the dislocations. In the case of bcc metals, attempts at experimentally observing the asymmetry of the mobility of individual dislocations upon inversion of the direction of their glide were unsuccessful.[25] On the other hand, the difference between the properties of dislocation multiplication in these samples, when tested under compression and tension, was directly confirmed in experiment.[26,27]

The dependence of the dislocation ranges in NaCl crystals on the time of action of an external force, and also on the direction of this action, was investigated in Ref. 28 by selective chemical etching. The Peierls stress in this material is negligibly small, and the dislocation mobility at low τ is completely limited by the barriers connected with the point defects. In contrast to the results obtained with Ge in Ref. 22 and in the present study, it was established in Ref. 28 that, regardless of the sign of acting stresses, the path length l of the dislocation varied nonlinearly with time t of its motion. Reversal of the direction of the dislocation glide changed the character of the nonlinear $l(t)$ dependence. The sign of the effect was the same whether the half-loop expanded or contracted when the dislocation was brought to the starting position. These facts offered evidence of a change in the regularities that govern the detachment of the dislocations from the blockades upon inversion of the glide direction, and the authors of Ref. 28 attributed them to an "increase in the perfection of the structure of the glide plane" after passage of the dislocation over it, i.e., to a decrease in the density of the blocking centers.

The asymmetry of the dislocation mobility observed experimentally in Ge single crystals upon reversal of the direction of their glide can be due in principle to several causes. Let us analyze in succession their possible manifestations in the investigated experimental situation.

An increase in the mobility of dislocations that move in a direction corresponding to a decrease of the dislocation-half-loop diameter can occur if the dislocation is acted upon by additional "contracting" stresses due to the linear tension and image forces, which cause the dislocation to tend to decrease its length and occupy a position perpendicular to the surface,[29] so as to minimize these forces. However, measurement of the mobility of dislocations perpendicular to the observation surface (110), which were introduced in samples with edge orientations [001], [110], and [1$\bar{1}$0], has shown that their forward and backward velocities do not differ from those measured in samples with dislocations that are inclined to the surface. Thus, the image forces do not determine the observed increase of the mobility of dislocations moving in the backward direction. The negligible contribution made to the observed effect by the linear tension forces is evidenced by the absence of backward motion of the dislocations when the samples are annealed in the investigated temperature interval. In addition, if the contraction stresses were to have a decisive influence on the magnitude of the effect, one should expect a nonlinear dependence of the dislocation path on the time, as well as an increase in the difference between the forward and backward velocities with increasing test temperature, but this is not confirmed by experiment (see Fig. 1).

An asymmetry of the mobility of the dislocations in the Peierls relief might be possible if this relief were asymmetrical with respect to the dislocation glide direction. However, measurement of the velocities of 60° dislocations with equal sign of the Burgers vector, located on the compressed and stretched sides of the sample and gliding in opposite directions under the influence of stresses of equal magnitude, has shown that the values of v differed for them by not more than 20–30%, as was noted also earlier.[21] Thus, the possible asymmetry of the potential relief of the crystal lattice proper cannot be the cause of the difference observed in the present study between the dislocation velocities in the forward and backward directions.

Since it can be assumed[18] that the dislocations move in germanium in an extended state, it is necessary to analyze the possible influence of the extension of the dislocation on the nature of the discussed phenomenon. In the general case partial dislocations in the field of the same normal stresses can be acted upon by different forces because of the inequality of the tangential stresses acting in the direction of the Burgers vector of each of them. This should lead to a narrowing of a stacking fault moving in one direction and to an expansion in the opposite glide direction of the perfect dislocation. Since the width d of the stacking fault determines[30] the critical stress

$$\tau_c = k_0/d^2 \quad (k_0=\mathrm{const}),$$

at which the activation energy of the dislocation motion changes as a result of a transition from uncorrelated to correlated formation of double kinks on the partial dislocations that make up the perfect dislocation, one should expect a change in the velocity of the dislocation upon reversal of the glide direction.

The influence of large external stresses ($\tau \approx 40$ kgf/mm^2) acting in the direction of the Burgers vector of a perfect dislocation in Si, on the width of a stacking fault in its core was investigated in Ref. 31. The sample orientation was chosen such that the external stresses should have caused a decrease in the width of the stacking fault compared with its equilibrium value in the unloaded state. Experiment has shown, however, that both broadening and narrowing of the stacking fault is observed. To explain the results, the authors of Ref. 31 took into account, in the balance of the forces acting on the partial dislocations, the quasiviscous friction force applied by the lattice. They assumed that this force is different for different types of partial dislocations (30° and edge) and depends also on their location relative to the stacking fault (leading or terminal). Thus, allowance for the degree of extension of the dislocation gives some grounds for suggesting the appearance of a difference between the velocities of the perfect dislocation when the direction of its glide is reversed, as a result of two causes: the change in the width of the stacking fault, and as a result of the different mobilities of the partial dislocations, due to the change of the sequence at which they follow relative to the stacking fault.

To check on these assumptions, we have performed experiments in which it was possible to change the width of the stacking fault without changing the sequence of the partial dislocations. We used samples with the aforementioned two different orientations, where the normal stresses acted along the directions [110] and [001], respectively. In these cases the ratios of the forces acting on the partial dislocations that lead and terminate the stacking fault are the inverse of each other, so that the width of the stacking fault at the same value of τ should differ for them in the same was as when the dislocations move in opposite directions at the edge orientations [110], [1$\bar{1}$2], and [$\bar{1}$11] used in the study. Measurements have shown, however, that the velocities of the 60° sections of the dislocation half-loops, which expand in samples of these two orientations, practically coincided in the investigated temperature and stress interval. This offers direct evidence that a change in the width of the stacking fault because of the difference of the forces acting on the partial dislocations does not exert a substantial influence on the total velocity of the entire complex under the given experimental conditions.

As for the influence of the sequence of the partial dislocations on the velocity of the perfect dislocation, it should be noted that this factor stems from the assumption[31] that the dislocation of velocity can be described in terms of quasiviscous deceleration, which is not physically justified for Ge and Si crystals. However, even if we make the more rigorous assumption that the velocity of a partial dislocation is determined by the dependence of the Peierls relief on the position of the dislocation relative to the stacking fault, this still does not explain the observed asymmetry of the dislocation mobility. In fact, in this case it should be observed only for 60° dislocations, for which the leading and lagging partial dislocations are different (30° and edge), and would be absent for screw dislocations consisting of two 30° dislocations. However, the velocities of the forward and backward gliding screw dislocations differ by more than one order of magnitude, i.e., just as in the case of 60° dislocations. Moreover, the velocities of the 60° sections of half-loops with different sequences of the partial dislocations, emerging to the observation surface on opposite sides of the scratch, differ from each other by not more than 20–30% in the case of motion in the forward direction, as noted in Refs. 32–34 under experimental conditions when an appreciable increase (by 50–100 times) of the backward velocity was noted. Thus, these experimental facts indicate that the observed phenomena cannot be explained by taking into account only the extension of the dislocations.

A noncontradictory qualitative treatment of the obtained experimental data can be offered only if account is taken of the spatial inhomogeneities of the Peierls relief, which are produced by point defects. It is known[2] that, depending on the density and state of these defects, which determine the appearance and motion of double kinks, the dislocation velocity can be different for the same values of τ and T. For large stresses, when the kinks jump over the barriers due to the point defects without activation, the mobility of the dislocations is determined mainly by the production of double kinks in the field of the point defects. For the simplest case when they are produced on each point defect,[8] the dependence of the dislocation velocity on τ and T, according to Ref. 11, is described by the expression

$$v(\tau,T)=a\bar{l}\frac{E}{E_{\rm dp}}\left(1+\frac{al_c}{\bar{l}b}\exp\frac{E_{\rm dp}}{kT}\right)J_0(\tau,T), \tag{1}$$

where $\bar{l}$ is the average distance between the defects; l_c is the critical dimension of the double kink, determined from the equality of the forces of the mutual attraction of the kinks to the external force that repels the kinks from each other, $E_{\rm dp}$ is the energy of the interaction of the dislocation with the defect; a is the distance between the valleys of the potential relief; b is the Burgers vector of the dislocation; J_0 is the probability of production of a double kink in a crystal without point defects;

$$E=U_{\rm dk}-kT\ln\left(1+\frac{al_c}{\bar{l}b}\exp\frac{E_{\rm dp}}{kT}\right),$$

$U_{\rm dk}$ is the energy of production of a double kink of critical size in the absence of defects, and depends little on τ.

The dependence of v on τ is determined in this case mainly by the factor J_0 in accord with the expression

$$J_0=(8\pi)^{1/4}\frac{D_k}{b^3}\left(\frac{\tau ab^2}{kT}\right)^{1/2}\left(\frac{\tau}{G}\right)^{1/4}\exp\left(-\frac{U_{\rm dk}}{kT}\right), \tag{2}$$

where G is the shear modulus, D_k is the kink diffusion coefficient, $D_k\sim\nu_D b^2$ (ν_D is the Debye frequency).

It follows from (1) and (2) that the effective activation energy of motion of the dislocations under the influence of point defects that initiate the production of double kinks decrease compared with $U_{\rm dk}$, in first-order approximation, by an amount $E_{\rm dp}$:

$$U=U_{\rm dk}-E_{\rm dp}.$$

The dependence of v on τ turns out to be weak in this case:

$$v\propto\tau,\quad \tau<0.1\tau_P$$

(τ_P is the Peierls stress).

The stopping of the kinks by defects, which influences the mobility of the dislocations at small stresses, gives rise to a stronger dependence of v on τ (Refs. 15, 16). When many of the double kinks are simultaneously produced on a dislocation, this dependence is given by

$$v=(2a\bar{l}\nu_0^*\nu_f^0)^{1/2}\left(1+\frac{E_{\rm kp}}{\tau ab\bar{l}}+\frac{l_c}{\bar{l}}\right)^{1/2}\exp\left(-\frac{U_{\rm dk}+E_{\rm kp}}{2kT}-\frac{E_{\rm kp}}{2\tau ab\bar{l}}-\frac{l_c}{2\bar{l}}\right). \tag{3}$$

Here $E_{\rm kp}$ is the energy of interaction of the kink with the point defect, and ν_0^* and ν_f^0 are frequency multipliers defined in Ref. 15. In this case the dependence of v on τ can be approximated by a power-law function with an exponent $m=3$ (Ref. 3), while the effective activation energy is

$$U\approx U_{\rm dk}-E_{\rm dp}+E_{\rm kp}.$$

In the case of forward motion of the dislocations in the described experiments the dependence of v on τ agrees qualitatively, just as in Refs. 6 and 8, with expressions (3) at $\tau<\tau_0<2$ kgf/mm^2 and with (1) at $\tau>\tau_0$.

In the case of backward motion it can be described approximately in this entire interval of τ by Eq. (1) alone.

The change of the character of the $v(\tau)$ dependence and the decrease of U upon reversal of the direction of the dislocation glide can be explained within the framework of the theories of their motion in an inhomogeneous Peierls relief, if it is assumed that the real structure of the Ge single crystal, with respect to point defects located near the glide plane, is substantially altered. When the dislocation half-loop is expanded, the point defects produce in the initial state an obstacle to the propagation of the kinks, and the latter overcome the obstacle at $\tau < \tau_0$ with the aid of thermal fluctuations. In this range of τ, the $v(\tau)$ dependence (curve 1 of Fig. 2) is described by expression (3). Higher stresses ensure above-barrier motion of the kinks, under the conditions of which the increase of v with increasing τ is determined by Eq. (1).

As a result of the interaction between a dislocation moving in the forward direction and point defects, the state of the latter is transformed in such a way that they no longer act as barriers to the propagation of the kinks in the investigated range of τ in the case of backward motion of the dislocation. It is this which determines the absence of a section with a stronger dependence of v on τ for shrinking dislocation half-loops (curves 2 and 3 of Fig. 2), as well as the decrease of the effective activation energy of their motion (by an amount E_{kp}). In the case of non-activation jumping of the kinks over the barriers connected with the transformed point defects, the $v(\tau)$ dependence should already be described by Eq. (1) at all values of τ. In the new state, they cause also an additional decrease of U because of the higher values of U_{dp}, and it is this which ensures the considerable difference between the effective activation energies of the dislocation motion in the opposite directions. Within the framework of the discussed mechanism it is also possible to present a noncontradictory explanation of the influence, described above, of the conditions that precede the reversal of the sign of the external stresses on the characteristics of the mobility of dislocations with decreasing diameter of the half-loops. They are due to the different changes in the real structure of the crystal in the course of its heat treatment.

A stable transformation of the state of point defects adjacent to glide planes of the dislocation in Si and Ge single crystals was already confirmed directly by experiment.[35-39] In these experiments on Ge, the direct symptom of one of these processes, namely adsorption of impurities on a moving dislocation, is the onset of an etch pit at the turning point of the dislocation under the influence of the part of impurity atmosphere left by it. This effect, however, does not make a decisive contribution to the mechanism of the asymmetry of the dislocation mobility. On the contrary, the very departure of the impurity is due to the abrupt increase of the dislocation velocity. It was observed thus that impurity complexes can appear also when the dislocation moves in the forward direction without reversal of the load, under the action of a stepwise increasing stress, and mark the position of the dislocation at the instant of time when its velocity is increased by approximately two orders of magnitude by application of a larger stress (Fig. 4). The velocity of the dislocation after it leaves an impurity complex does not differ in this case from that usually measured at the same stresses and temperature.

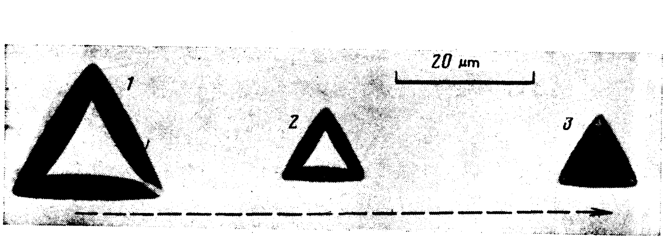

FIG. 4. Arrangement of etch pits in Ge sample etched after the action of a stress that increased stepwise from 0.5 to 3 kgf/mm^2 at T = 450 °C. 1) Starting position, 2) section corresponding to position of dislocation at the instant of abrupt increase of the velocity, 3) final position of 60° dislocation. The arrow shows the direction of its motion.

The asymmetry of the mobility of dislocations in semiconductors can be due to processes previously not accounted for these materials, namely the transformation of the state of the point defects near the dislocation and the part of the glide plane swept by the dislocation. Its mechanism can consist, e.g., of a reorientation of the asymmetric point defects in the field of dislocation microstresses. The possibility of such processes was invoked in discussions of the anisotropy of macroscopic characteristics of plastic deformation of alkali-halide crystals[40,41] irradiated under load or in electric field, of data on internal friction in semiconductors,[42] etc. This anisotropy is a natural consequence of the existence of several equivalent orientations of a point defect in a crystal, corresponding to a minimum of its energy. Nor is it excluded that the transformation of complicated associative defects is stimulated by a redistribution of the chemical bonds as a result of dislocation motion, occurring at the instant when a moving dislocation cuts through the complex. The features of the fine structure of the dislocation core, including stacking faults, sections of transition of dislocation from a gliding into reshuffled set, jogs, etc. can facilitate the structural transformations of the point defects.

We note in conclusion that an investigation of the asymmetry of the dislocation mobility in semiconductor single crystals is of interest not only from the point of view of analyzing the fundamental aspects of the problem, but also for the solution of the important practical problem of finding ways of increasing the plasticity of material so as to lower the brittle-damage threshold.

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Translated by J. G. Adashko