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ASYMMETRY OF ISOLATED DISLOCATIONS MOBILITY IN Ge AND Si SINGLE CRYSTALS

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Abstract - Change of dislocation velocities in Ge and Si caused by inversion of applied shear stress sign has been investigated as a function of temperature, stresses and thermal treatment conditions. The results obtained are discussed with account of the influence of dislocation splitting and dislocation-point defects interaction on the double kinks formation and extension.

In the last years the convincing evidence for the dislocation splitting in semiconductors has been obtained /1/. In this connection the influence of dislocation dissociation on dislocation mobility in Ge and Si has arisen a keen interest /2,3/. Still earlier it was established that the dislocation velocities in these crystals depend strongly on dislocation-point defects interaction /4/. In order to clarify the main factor affecting the dislocation mobility in a deep Peierls potential the experiments have been performed in the present work to investigate the effect of inversion of the dislocation motion direction on dislocation velocity.

The dislocation velocities were measured using large dislocation half-loops (600-1000 μm in diameter). The procedure of the dislocation introduction as well as loading by four-point bending around the [112] axis described in /5/. Part of the samples were cooled before each inversion of dislocation glide and etched to measure the dislocation pathways. On the other samples reversal of the stress sign was performed without intermediate cooling. Six-support bending jig was used to transfer the loading from the inner supports to the extreme ones /6/.

Fig. 1 shows the distances travelled by a 60° dislocation in P doped (up to 10^20 m^-3) Si grown by the floating zone method under the operation of shear stresses of opposite signs (τ=5 N/mm^2, T=650°C, t=1200 s). It is seen that the reversal of the motion direction is followed by a substantial increase of not only the average distance travelled by the dislocations but of the dispersion of the distance distribution too. The studies of the dependence of the dislocation average pathway on the loading time (Fig. 1b) have shown that the difference is due to a sharp increase (from 1.4 × 10^-5 to ≈10^-4 mm/s) of the dislocation velocities in the direction corresponding to the
narrowing of the half-loop. A change of the dislocation mobility was observed for $60^\circ$ and screw segments of half-loops on both initially extended and compressed sides of the sample. 

![Graphs showing dislocation mobility](image1)

Fig. 1 - Histograms (a) and the time dependence of the pathways (b) of the dislocations in Si. (1) - during expansion of the half-loop, (2) - during narrowing of the half-loop.

The difference between the dislocation velocities in opposite glide directions was found to be very sensitive to the temperature, stresses and to the conditions of the crystal heat treatment before the reversal of stress sign (Fig. 2).

![Graphs showing temperature dependence](image2)

Fig. 2 - Temperature dependences of the $60^\circ$ dislocation velocities in Si (a) ($\tau=10$ N/mm$^2$) and in Sb doped (up to $10^{15}$ m$^{-3}$) Ge (b) grown by Czochralski method ($\tau=5$ N/mm$^2$) during the half-loop extension (1), broadening of the half-loop along doubly swept slip plane part (2), during narrowing of the half-loop after the intermediate cooling (3) and without the intermediate cooling (4).

It is seen from Fig. 2 when the reverse cycle is repeated, i.e. upon a switch-over from the narrowing to the expansion of the half-loops, their motion along the doubly swept slip plane part is characterized by practically the same values of the velocity and activation energy as in the case of the first half-loop extension in asgrown crystal (curve 2). This result allows us to conclude that the asymmetry of the dislocation mobility is observed. The asymmetry magnitude decreases as the stresses increase.
Fig. 3 shows the dependence of the 60° dislocation velocities in Si (T=650°C, \(\tau=10\) N/mm²) in the direction corresponding to the narrowing of the half-loop on the annealing duration at \(T=680°C\) (1) and \(730°C\) (2) that preceded the inversion of the dislocation gliding direction. The dotted line indicates the velocity of the extending half-loop. It did not depend on the thermal treatment. It is seen, that the annealing causes a disappearance of the dislocation mobility asymmetry. However it can be restored in the same specimen if the dislocation was forced to move into the unswept part of the slip plane.

The results described evidence that the dislocation mobility asymmetry in Si and Ge is not caused by an obvious reason - the action of the line tension and image forces which result in a distinction of the effective forces magnitude acting on the broadening and narrowing half-loops. If this reason were essential one would expect the nonlinear dependence \(L(t)\) and the asymmetry increase with rising temperature \(/7/\), that contradicts experiment (Fig. 1,2). Moreover the annealing did not result in half-loop spontaneous shrinkage in our experiments, but led to the disappearance of the asymmetry effect. The latter fact gives evidence that possible asymmetry of the Peierls potential relief cannot stipulate the phenomenon observed too.

The effect described, in principle, would be explained with account of the dislocation splitting if it is essential in the formation of the dislocation mobility regularities in semiconductors. There are two main reasons for such an assumption. First of all the stacking fault width may change due to the sign stress inversion owing to the reversion of the forces ratio acting on the leading and trailing partial dislocations. In accordance with /3/ it may cause the transition from the uncorrelated to the correlated double kink formation on the partial dislocations as well as to the changes of the velocities and the activation energy of the complete dislocation motion. However the direct TEM measurements of the stacking fault width after the inversion of the dislocation gliding direction revealed no essential changes in splitting as well as in the constriction densities when the asymmetry effect was observed. Besides, by changing the bending axis orientation we have managed to realise the situation when the ratio of the forces acting on partial dislocations of the widening half-loop in one crystal was the same as for the narrowing half-loop in the another one. But the half-loop expansion velocities were equal to one another in the experimental accuracy limit. It evidences that the mentioned reason cannot cause the effect observed.

The second cause of the mobility asymmetry may be connected with the assumption, pointed out in /2/, that the velocity of the partial dislocation depends on their type and on the position in respect to the stacking fault. If this reason were essential the effect discussed would not observed for the screw dislocations, that
contradicts experiment. Moreover it is known that the velocities of the 60° segments of the expanding half-loop differ from each other only by 20–30% 

in such experimental conditions when we observed the increase of the narrowing half-loop velocity by two orders of magnitude. These give evidence that the dislocation dissociation did not play a significant role in the formation of the dislocation mobility peculiarities under discussion.

The effect observed can be explained unambiguously taking into account the effect of point defects on the double kinks formation and its expansion along the dislocation line. But one has to bear in mind that gliding through the crystal the dislocation changes the state of point defects in a crystal volume. It gathers impurity atoms. In our experiments the direct evidence for the process has been obtained. As a result of the sharp increase of dislocation velocity during the inversion of its gliding direction without crystal cooling the dislocation breaks away from the part of the impurity atmosphere. A one-dimensional set of the point defect complexes forms from this left atmosphere. They act as a line defect, which can be detected by etching (Fig. 4). We succeeded in revealing a local change of the surface electrical properties of Sb-doped (up to $10^{19}$ m$^{-3}$) n-Ge in the vicinity of the emergence point of this defect (Fig. 5).

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**Fig. 4** - Arrangement of etch pits in Si etched after the reversal of the 60° dislocation gliding direction (T=600°C, $\tau$=40 N/mm$^2$). The time required for the dislocation to go from the starting position (1) toward the right-hand turning point (2) was 1920 s, the time of the motion in opposite direction to the final position (3) was 1200 s. The arrows show the dislocation gliding direction.

**Fig. 5** - Current-voltage characteristics of the tungsten microprobe contact with n-Ge surface. The contact is positioned in etch pit of 60° dislocation (1), in etch pit of quasi-one-dimensional set of point defect complexes (2) and on the defect-free section of surface (3).

In the vicinity of the dislocation core the favorable conditions for the reaction of complexes formation appear due to the increase of the local impurity concentration. The impurity complexes formed cannot move with a dislocation because of their smaller diffusion coefficient than that of impurity atoms and are left in the slip plane behind the dislocation. It causes the formation of the specific etching traces behind the moved dislocation (Fig. 6) and an effect of the anomalous anisotropy of the mobility of the electrons moving along and across the parallel slip planes (Fig. 7).
Fig. 6 - Etch pit and specific trace along the dislocation slip plane in Si (a) and their image in an microinterferometer microscope (b).

Fig. 7 - Temperature dependence of the electron mobility in Si plastically deformed at 625°C. The current lines are directed perpendicular to the slip planes (3) and parallel to them (1,2-along and across the dislocations respectively).

It is known that the impurity atoms can facilitate the double kink formation /9,10/, that results in a decrease of the effective activation energy for the dislocation movement. On the other hand the necessity for the kink to overcome the barriers produced by point defects causes the increase of the activation energy for the dislocation gliding and results in a stronger dependence $v(T)$ /11,12/. Taking into account the described changes of the point defect states in the swept part of the dislocation slip plane one can explain a decrease of the activation energy for the dislocation motion during the narrowing of the half-loop and its dependence on the stresses and thermal treatment.

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