Deformation of a $\Sigma = 9$ (122) gb in silicon studied by HREM

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The role of grain boundaries (GB) during plastic deformation is well recognized. In spite of a large number of speculations and experimental studies (review 1, 2, 3), the local mechanisms of interaction between deformation induced dislocations and GB remain unclear. High resolution electron microscopy (HREM) has permitted to reveal what occurs at atomic scale in the particular case of the deformation of a twin GB E=9 in Si. The initial dislocation free bicrystals are obtained by Czochralsky method. The common tilt axis is [011], the misorientation angle 38°, the GB plane is (122)I (i for crystal I). The structure of the perfect \( \Sigma = 9 \) boundary has been previously determined either in Ge (4) and in Si (5). Deformation experiments are performed at the Ecole des Mines de Nancy (France) by A. Jacques and A. George. Compression tests are made along the \([26,7,20]_I\) axis equivalent to \([26,20,7]_{II}\) at 850°C. Traction tests has been made along the \([411]_I\) axis equivalent to \([411]_{II}\). The specimens are then thinned and observed in a JEOL 2000X microscope, in such a way that the GB is viewed end-on. The [011] axis common to both grains is parallel to the electron beam.

Previous observations using in-situ electron microscopy (6) and by RX topography (7) has been carried out on the same \( \Sigma = 9 \) bicrystals in Ge and Si. The main result is that the GB seems to act as a strong obstacle and a trap for lattice mobile dislocations. However, in cases of common slip for both grains dislocations can pass directly through the GB. The entrance of dislocations within the GB is viewed but the further behaviour of the grain boundary dislocations (GBD’s) is not determinable at this scale.

By HREM the observable dislocations are those lying along the [011] axis. The Burgers vector (BV) and the step height of the grainboundary dislocations are determined directly on the HREMicrographies using the method proposed by King and Smith (8). Due to the appropriate choice of the strain axis, the dislocations which fulfill the HREM observation criterion are those: (i) when the primary slip system \((\{111\}, \{100\}_I)\) or \((\{111\}, \{100\}_{II})\) and to the second slip system in primary planes \((\{111\}, \{101\}_I)\) or \((\{111\}, \{101\}_{II})\), 60° dislocations on the \((\{111\}_I)\) or \((\{111\}_{II})\) planes are also observable.

Pile-ups of 60° dislocations are often found but a lot of isolated dislocations are observed which reveal homogenisation and reduction of stresses by cross-slip. Screw dislocations are also found. 60° dislocations glide towards the GB remaining dissociate till the GB. In compression tests the leading partial is the \(90°\) partial dislocation whereas it is the \(30°\) partial in the traction test. Dislocations enter the GB and dissociate into GBD’s which BV belong to the DSC lattice. In compression the 90° partial disassociation is as soon as it touches the GB, whereas the 30° partial is still in the grain. The 90° partial disassociation residues are two primitive DSC vectors \(b_0\) and \(b_2\) one of which \(b_2\) is parallel to the GB and can easily glides within the GB and for this reason is not often found inside the GB. The \(30°\) leading partial cannot immediately dissociate because its BV is originally a DSC primitive vector. The 60°D must enter entirely within the GB before decomposition in the same three DSC vectors as in compression. It must be emphasized that, in these cases, the common assumption of energy reducing by dissociation in smaller DSC vectors does not hold because the incident partials belong to the DSC, and the \(b^2\) criterion gives the same balance for the \(90°\) partial decomposition.

It has been already pointed out (9) that the dissociation of the lattice dislocation and the incident partial BV belonging to the DSC must be taken into account to explain the behaviour of the GB under strain. Dissociated 60°D are observed on the \((\{111\}_I)\) and \((\{111\}_{II})\) planes. In these cases none of the partial BV belongs to the DSC lattice. However three primitive DSC vectors are found corresponding to their decomposition : two are equal to \(b_2\) parallel to the GB, the third one is equivalent to the incident 30° partial BV different from the previous 30° partial on the primary planes. The entrance of a 60°D leads to three DSC dislocations (three primitive DSC lattice vectors) within the GB, which are completely separated. This observation evidences the easily glide of DSC dislocation which BV is parallel to the GB but illustrates also the fact that DSC dislocations can also move by climb within the GB at 850°C. The lattice incoming dislocation completely looses its identity by decomposition within the GB. At this step its direct transfer in the other grain is highly improbable. Unfortunately, we are not able to detect screw dislocations in the GB and thus we cannot exactly determine their behaviour. However, because of their easy transfer through the GB, we can imagine that we never get an image of this step.

As GB’s move within the GB by glide and climb, they can interact. For instance the BV’s of the glissile residues due to the decomposition of 60° coming from grains I and II have opposite signs : they can annihilate together. A number of different resulting configurations has been observed. In the simplest cases it is possible to describe the history of the residues. As the deformation conditions are symetrical with respect to the grains the number of dislocations coming from both grain can be estimated to be the same and consequently the number of dislocations which BV are parallel to the GB. This observation evidences the easy glide of dislocations coming from both grain.

References 1: