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Generation of dislocations during plastic deformation

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Abstract

The models of dislocation generation during plastic deformation are reviewed. The different types of dislocation generation, i.e. localized Frank–Read sources and multiplication by the double-cross slip mechanism, are observed in ceramic single crystals, semiconductors, metals, intermetallics and quasicrystals during in situ deformation in a high-voltage electron microscope. The results are discussed with respect to the ability of cross slip in different materials.

Keywords: TEM; In situ experiments; Frank–Read source; Dislocation multiplication

1. Introduction

In most crystalline materials, the dislocation density increases drastically during plastic deformation leading to work-hardening. This process may be described by an evolution law of the dislocation density containing a rate of dislocation generation minus an annihilation rate. The dependences of the dislocation density itself on the experimental parameters like strain, strain rate and temperature are well studied in many materials. However, the processes of dislocation generation and recovery are much less understood, at least on a quantitative level, although the basic models are about 50 years old, already.

The present paper gives a summary of the information on dislocation generation during plastic deformation obtained from in situ straining experiments by high-voltage electron microscopy (HVEM) performed by the authors on a number of different materials like ceramics, semiconductors, metals, intermetallics and quasicrystals. These experiments were carried out either in a quantitative double-tilting straining stage for room temperature [1] or a high-temperature straining stage for temperatures up to 1150 °C [2].

2. Models of dislocation generation

Since the homogeneous nucleation of dislocations requires stresses of about one tenth of the shear modulus, the generation of dislocations during plastic deformation occurs at much lower stresses as an elongation of the length of existing dislocations. The best-known mechanism of dislocation generation is the Frank–Read source [3], which may be characterized as a localized source. A dislocation segment lying on a slip plane is pinned at its ends, e.g. by nodes of the dislocation network or simply by changing onto another plane where it is not mobile, as sketched in Fig. 1. Under the applied stress, the mobile segment moves through different stages marked a to e. This localized source can operate repeatedly to emit a greater number of dislocations on the same slip plane.

The second mechanism is the double-cross slip mechanism suggested by Koehler [4] and Orowan [5] and experimentally first observed indirectly by Johnston and Gilman [6]. It is shown schematically in Fig. 2. A screw dislocation (a) moves on its glide plane, which is identical with the image plane. Cross slip of a segment of length L results in two superjogs J acting as pinning agents. The segment then multiplies similarly to the Frank–Read source. Simple line tension arguments show that both sources can act only if the segment length L is larger than a critical value

$$L_c = \frac{\mu b}{\tau} \quad (1)$$

where μ is the shear modulus, b the absolute value of the Burgers vector and τ is the local component of the acting stress. Using characteristic values, L_c is in the range of about 100–200 nm. In stage (c) of the double-cross slip mechanism, the branches adjoining the jogs have to pass each other on their parallel planes. This is only possible if the height of the jogs, i.e. the distance between the parallel glide planes, is larger than a critical value (dipole opening criterion)

$$h_c = \frac{\mu b}{[8\pi(1-\nu)\tau]} \quad (2)$$

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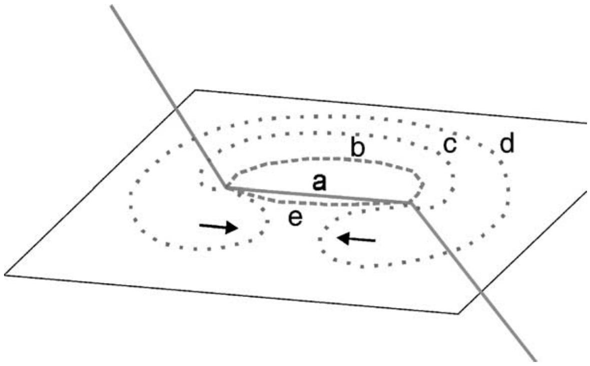

Fig. 1. Dislocation generation in a Frank–Read source.

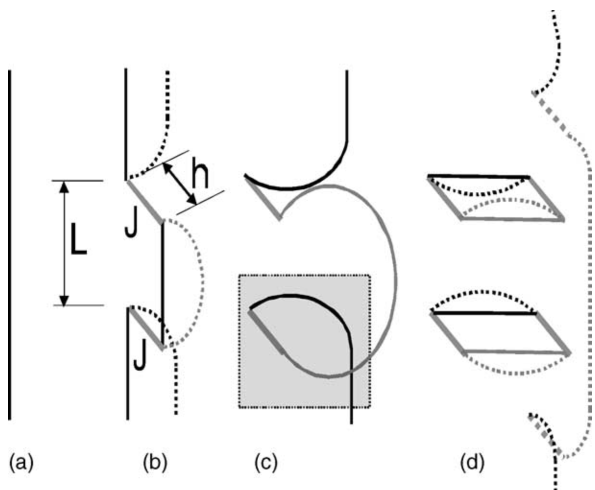

Fig. 2. Double-cross slip mechanism for dislocation generation.

with ν being Poisson's ratio. It is obvious that h_c is about 20 times smaller than L_c. Thus, both L_c and h_c are well below the foil thickness of about 500 nm in an HVEM in situ experiment, so that the mechanisms of dislocation generation can well be observed. If it is considered that the cross slip events show characteristic frequency distributions of L and h, it is reasonable to assume that the frequency of dislocation multiplication increases with increasing stress since sources with smaller values of L and h can be activated at higher stresses. In many cases, the intermediate configuration of stage (c) is metastable. It is characterized by the highlighted α-like configuration in Fig. 2, which is frequently observed in dislocation structures under stress. The double-cross slip mechanism usually emits only a single new dislocation loop. Since the generated dislocation moves on a plane parallel to the original one, slip may spread leading to a growth of the width of the slip bands, in contrast to the Frank–Read source which emits many dislocations on the same plane. Cross slip events happen during the motion of the dislocations. Accordingly, the increase in the dislocation density $d\rho$ should be proportional to the area dA swept by all dislocations. Considering the dependence of the creation rate of dislocations on τ, the creation rate may be written as

$$d\rho = w\tau\, dA = w\tau\rho\, ds = \left(\frac{w}{b}\right)\tau\, d\varepsilon \tag{3}$$

where w is a constant, ds the displacement of all dislocations and $d\varepsilon$ is the increment in shear strain. If the cross slip height is smaller than h_c in Eq. (2), the dislocation trails a dipole at each jog as outlined in stage (d) of Fig. 2. These dipoles may be terminated by glide of the jogs along the dislocation line. If the stress increases later on, yielding a smaller value of h_c, dipoles may open and emit additional new dislocations.

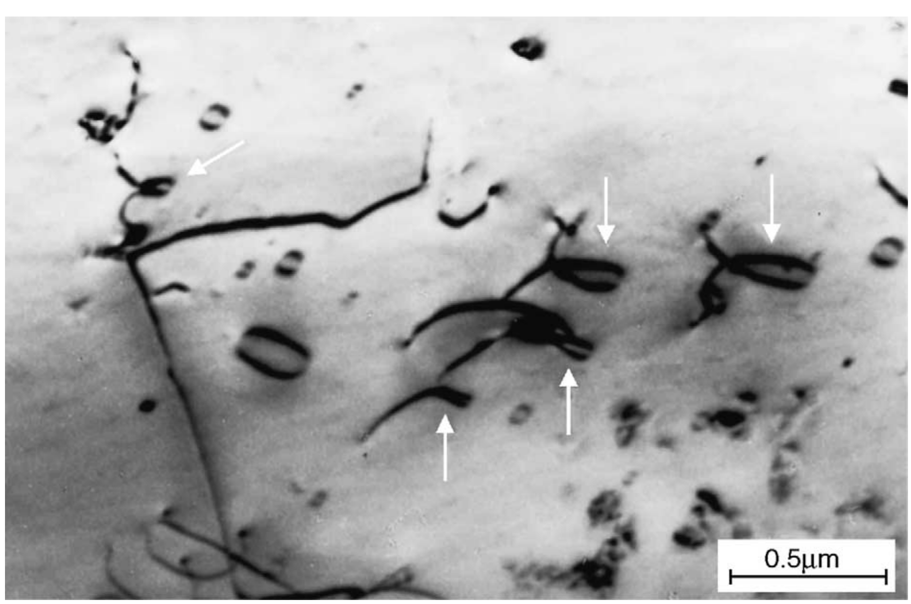

Fig. 3. Dislocation structure during in situ deformation of ZrO_2–10 mol% Y_2O_3 at 1150 °C.

3. Experimental observations

The action of the different processes of the double-cross slip mechanism has been demonstrated in most detail by the in situ experiments on MgO single crystals [7]. Cross slip seems to be easy and appears frequently. Similar processes have been observed in ZrO_2–10 mol% Y_2O_3 at 1150 °C [8]. At these conditions, slip is very jerky suggesting that thermal activation does not control the dislocation motion. The multiplication events appear in a very instantaneous way, too. Nevertheless, the intermediate α-like configurations of stage (c) in Fig. 2 are observed frequently, as marked by the arrows in Fig. 3. In the in situ experiments with their relatively low foil thickness, a segment may cross slip which is terminated by the foil surface at one end so that only a single jog is created. In Fig. 3, many dislocation loops are visible, too, which may open at a later stage. As described above, the double-cross slip mechanism leads to sidewise spreading of slip. Since double-cross slip events of smaller height may lead to multiplication at higher stresses, the slip bands become narrower in strong materials and at low temperatures.

An example of the action of a localized source in an austenite grain in duplex steel [9] is presented in the sections of a video recording in Fig. 4. As in Fig. 3, only one fixed end of the mobile segment is contained in the foil and is marked by an arrow in Fig. 4a. The mobile branch emerges through the surface. It revolves a number of times and always emits a dislocation on the same plane which piles up against a phase boundary above the source. The piled-up dislocations cause a back stress which shields the applied stress and blocks the source after the pile up contains about three dislocations as in Fig. 4b. After some dislocations break through the boundary, the source operates again. Fig. 4c shows a configuration where a maximum of four dislocations pile up. This configuration was stable only for a very short time.

Figs. 3 and 4 are typical examples of the double-cross slip mechanism and the localized Frank–Read sources in crystalline materials. The information about dislocation generation in all other materials studied by the present authors is summarized in Table 1. It contains also the relevant slip and cross slip planes.

A special situation exists in quasicrystals. Although these materials do not exhibit translational symmetry, plastic deformation is carried by the movement of dislocations, as first observed directly during the in situ straining experiments on icosahedral Al–21 at.% Pd–8.5 at.% Mn single quasicrystals at high temperatures [19]. New dislocations are created by a mechanism similar to the double-cross slip mechanism [20]. However, frequently new dislocations emerge from a main slip band on other inclined planes. In connection with this, α-like configurations are observed as shown in Fig. 5. In contrast to the usual double-cross slip mechanism, the two branches adjoining the pinning center do not lie on parallel planes. Apparently, one branch extends on the original slip plane and the other one on a cross slip plane onto which the new dislocation spreads, as indicated by the straight lines in Fig. 5. Thus, multiplication in icosahedral quasicrystals seems to take place by a single-cross slip mechanism outlined in Fig. 6, where dislocations gliding on one (primary) plane PP emit dislocations on a secondary one SP.

(a)
(b)
(c)


Fig. 4. Sections of a video sequence of a localized dislocation source in an austenite grain in duplex steel: (a) 0 s; (b) after 58 s; and (c) after 126 s. From the cooperation in [11].

Table 1
Burgers vectors, slip and cross slip planes as well as type of dislocation generation in different materials

Material load axis	Burgers vector	Slip planes	Cross slip planes	Frank–Read sources	Double-cross slip mechanism
NaCl⟨1 0 0⟩	1/2⟨1 1 0⟩	{1 1 0}	{1 0 0}		[10][a]
MgO⟨1 0 0⟩	1/2⟨1 1 0⟩	{1 1 0}	{1 2 2} {2 1 1}		[7]
ZrO_2⟨1 1 2⟩	1/2⟨1 1 0⟩	{1 0 0}	{1 1 0} {1 1 1}		[8]
Si, Ge	1/2⟨1 1 0⟩	{1 1 1}	{1 1 1}	[11]	
Al–Zn–Mg	1/2⟨1 1 0⟩	{1 1 1}	{1 1 1}		[12]
Al–1Ag	1/2⟨1 1 0⟩	{1 1 1}	{1 1 1}		[13]
Al–8Li	1/2⟨1 1 0⟩	{1 1 1}	{1 1 1}		[14]
Duplex steel austenite	1/2⟨1 1 0⟩	{1 1 1}		[9]	
Duplex steel ferrite	1/2⟨1 1 1⟩	{1 1 0} {1 1 2} {1 2 3}	{1 1 0} {1 1 2} {1 2 3}		[9]
Ti–6Al	$a/3\langle 1 1 \bar{2} 0\rangle$	$\{0 0 0 1\}$ $\{1 \bar{1} 0 0\}$	$\{0 0 0 1\}$ $\{1 \bar{1} 0 0\}$	[15]	[15]
γ-Ti–52Al	1/2⟨1 1 0⟩	{1 1 1}	{1 1 1}		[16]
NiAl⟨1 1 0⟩	⟨1 0 0⟩	{1 0 0} {1 1 0} {2 1 0}	{1 0 0} {1 1 0} {2 1 0}		[17]
$MoSi_2$⟨2 0 1⟩	1/2⟨1 1 1⟩	{1 1 0}		[18]	

[a] Cross slip studied by metal surface decoration.

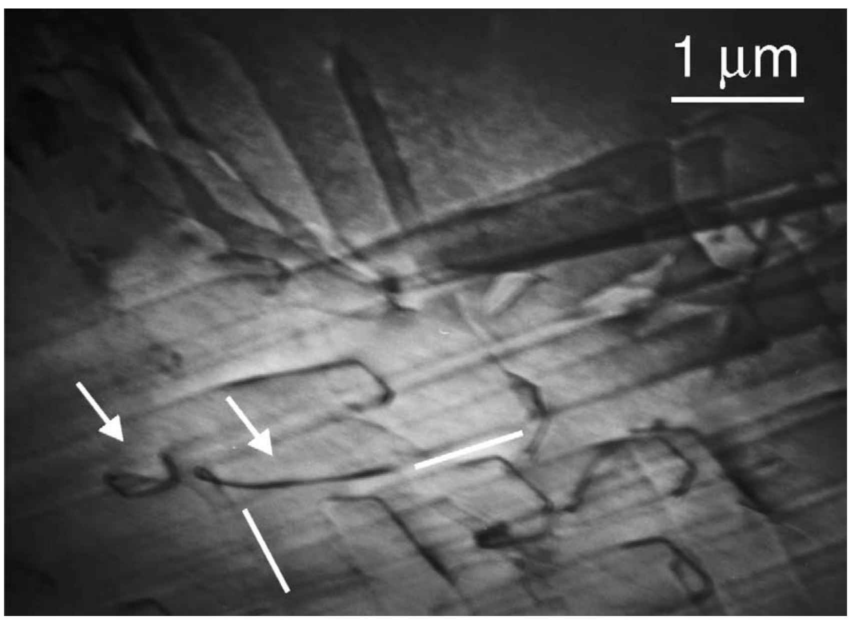

Fig. 5. The α-like dislocation configurations during in situ deformation of icosahedral Al−21Pd−8.5Mn at 750 °C.

4. Discussion

As shown above, HVEM in situ straining experiments reveal the different mechanisms of dislocation generation during plastic deformation. However, the understanding on a quantitative level is still poor. In contrast to the localized Frank–Read sources, dislocation multiplication requires cross slip. While in the face-centered cubic metals the cross slip planes are of the same type as the primary slip planes resulting in an equal dislocation mobility, in most other materials the mobility on the cross slip planes is lower than on the primary ones. In ZrO_2–10 mol% Y_2O_3, e.g. the flow stress on the cross slip planes is about 30% higher than on the primary ones [21]. The flow stress ratio between both planes certainly influences the width of the cross slip events. Statistical data on the frequencies of cross slip are very rare. An exception are the data on NaCl single crystals obtained from surface heavy metal decoration of slip steps of individual dislocations [10]. Here, the frequency of cross slip decreases strongly with increasing cross slip height h (approximately exponentially for short heights). Using the dipole opening criterion of Eq. (2) for multiplication with appropriate parameters, only a small fraction of cross slip events at the long end of the distribution may lead to multiplication, as was modeled first in [22]. Both, the average cross slip height and the total frequency depend on doping the crystals, which influences the flow stress, but not on temperature. Cross slip is initiated by long-range internal stress fields. Only in materials where extended dislocations have to constrict before cross slip, the initiation

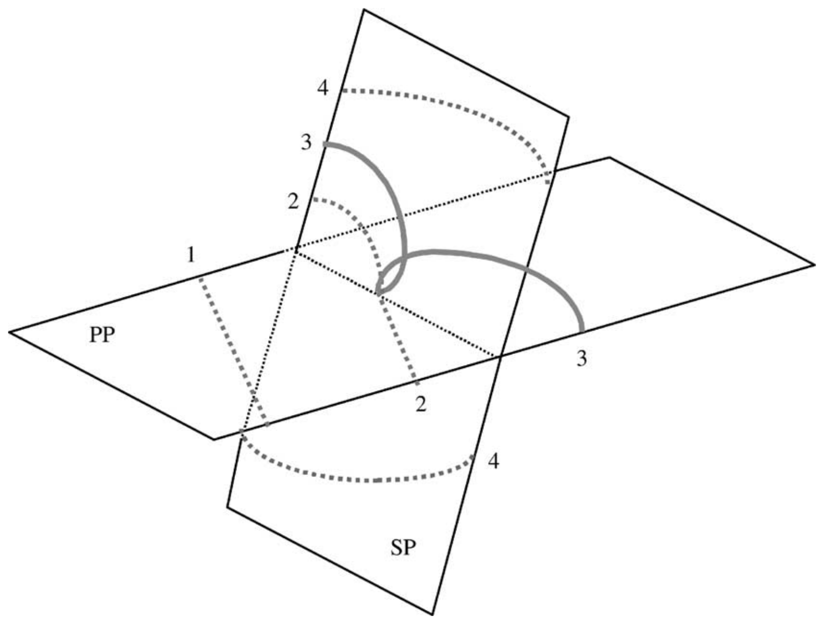

Fig. 6. Schematic representation of dislocation multiplication in a quasicrystal by a single-cross slip mechanism.

process is thermally activated. The multiplication event, i.e. the bowing after Eq. (1) and the bypassing after Eq. (2) are also athermal processes. This is confirmed by the very instantaneous character of multiplication in ZrO_2, where the flow stress is also of athermal nature at the respective temperature. As a consequence, the multiplication rate of Eq. (3) contains the applied stress but not explicitly the temperature. However, the latter enters the rate of dislocation annihilation since climb is involved in this process so that the evolution of the dislocation density depends strongly on temperature, as discussed, e.g. for Al–Pd–Mn quasicrystals in [23].

While cross slip is a prerequisite of dislocation multiplication, localized sources should operate if cross slip is limited. This is obvious for dislocations with $1/2\langle 1\,1\,1\rangle$ Burgers vectors in $MoSi_2$, where only $\{1\,1\,0\}$ planes exist as easy slip planes. In the other cases, the situation is not as clear, e.g. in Ti–6 at.% Al, where the slip and cross slip planes are of different type and both mechanisms of dislocation generation occur. The type of dislocation generation certainly influences the spreading of slip. Planar slip, as it is observed, e.g. in $MoSi_2$ and Ti–6Al requires the operation of localized sources. Slip localization is frequently accompanied with plastic instabilities (e.g. [24]). Both phenomena are usually discussed on the basis of processes which soften the material, e.g. by destroying long- or short-range order (as in Ti–6Al) thus facilitating dislocation motion in a narrow region. At the same time, the mechanism of dislocation creation must generate the new dislocations within this narrow region. This may occur either by the localized Frank–Read sources or by multiplication at high stresses where short cross slip events are sufficient for the emission of a new dislocation.

Note

Recently, it turned out that plastic deformation of icosahedral Al–Pd–Mn quasicrystals is mainly carried by climb of dislocations or a combination of climb and glide (F. Mompiou, M. Feuerbacher, D. Caillard, International Symposium on in situ Electron Microscopy, Nagoya, 20–22 January 2003; U. Messerschmidt, M. Bartsch, submitted to Scripta Mater.). Nevertheless, the dislocations move on well defined planes, most probably owing to special core configurations. During multiplication, the dislocations change from one plane to another, as described in the text, but the process is not glide and cross glide but combined climb and glide on different planes.

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