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On the stress exponent of dislocation climb velocity

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Abstract

The few available examples of pure climb plasticity indicate that the stress dependence of the corresponding dislocation velocity is larger than the simple proportionality predicted by classical models. A possible explanation is proposed on the basis of a comparison with dislocation glide in semiconductors. It is subsequently shown that high-temperature creep of several materials, in particular superalloys, may be controlled by the same process.

Keywords: Dislocations; Climb; Quasicrystal; Superalloy

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1. Introduction

The velocity of climbing dislocations is usually considered to be proportional to stress, σ, with activation energy equal to that of self diffusion. If the density of mobile dislocations is assumed to vary as the stress squared, this yields the so-called natural creep law, for which the strain rate is proportional to σ^3 (Nabarro [1]). Since most creep experiments yield stress exponents larger than 3, it can be inferred that – unless ad hoc assumptions are made – the stress dependence of the dislocation velocity is higher than expected. There are however very few studies of high-temperature plasticity proceeding by pure climb, for which the stress dependence of climb velocity can be measured unambiguously. These experiments are reviewed, and the corresponding stress–velocity dependences are discussed and compared to that of dislocation glide in semiconductors, in order to propose possible common interpretations.

2. Examples of pure climb plasticity

2.1. Hexagonal close packed (hcp) Be and Mg

The most comprehensive studies of climb under an externally applied stress have been made by Le Hazif et al. [2], in Be, and by Edelin and Poirier [3], in Mg. When compressed along their *c*-axis, these two metals deform by pure climb of *c*-dislocations. Vacancies absorbed by these dislocations are regenerated by another set of climbing dislocations, with *a*-type Burgers vectors, not activated by the applied stress. This corresponds to Nabarro creep mechanism.

In Mg, the density of *c*-dislocations has been measured as a function of strain, for various stresses and temperatures. The corresponding dislocation velocity, deduced from the Orowan equation, is independent of strain, and proportional to $\sigma^{2.8}$. The corresponding activation energy is 1.80 eV, larger than the self-diffusion energy (1.43 eV). These results which do not fit with the classical model have not been fully explained yet.

2.2. Quasicrystalline Al–Pd–Mn

More recently, dislocation motion in icosahedral Al–Pd–Mn at intermediate and high temperatures has been investigated by Mompiou et al. [4–6]. When compressed along a five-fold axis, samples deform by climb motion of dislocations with Burgers vector parallel to the compression axis, and trailing vacancy-type planar faults (called phason walls). The vacancies absorbed during this process are provided by another set of climbing dislocations, not activated by the applied stress, and trailing interstitial-type phason walls (Fig. 1). This situation is thus identical to that observed in hcp metals. Since Schall et al. [7] showed that the density of dislocations is approximately proportional to σ^2, the strain rate given by the Orowan law, with a dislocation velocity proportional to

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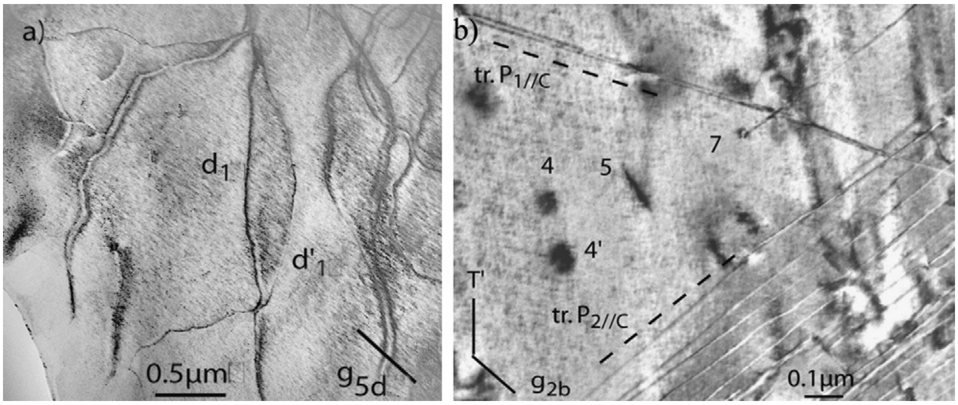

Fig. 1. Dislocations moving by pure climb in Al–Pd–Mn deformed at 300 °C, under a high confining pressure. Dislocations accounting for plastic deformation (those climbing in the plane perpendicular to the compression axis) are shown in (a) whereas dislocations exchanging vacancies with the former ones (those climbing in planes parallel to the compression axis) are shown in (b). From Mompiou et al. [4].

stress, should be proportional to σ^3. In reality, experiments yield stress exponents comprised between 4 and 10 (Brunner et al. [8], Messerschmidt et al. [9]), indicating that the velocity of climb is not proportional to stress. Here again, this discrepancy is difficult to explain in the frame of the available models.

3. Discussion

These two examples show that the stress dependence (stress exponent) of the climb velocity can be definitely larger than unity. This difference with respect to classical climb theories can be tentatively explained by the following considerations.

3.1. Jog-pair nucleation at high stress

In situ experiments in quasicrystals [5] showed that climbing dislocations have polygonal shapes with edges parallel to dense directions (Fig. 2). This indicates that climb involves a difficult jog-pair nucleation (Mompiou et al. [5], Takeuchi [10]). Under such conditions, the climb velocity must be described by the theory of Hirth and Lothe [11], originally developed for dislocation glide in semiconductors. In semiconductors, the velocity of glide in the “kink-collision regime” is:

$$v \propto \frac{\sigma}{kT} \exp\left(-\frac{(1/2)U_{\mathrm{kp}}^{(\mathrm{c})}(\sigma) + U_{\mathrm{m}}}{kT}\right), \tag{1}$$

where $U_{\mathrm{kp}}^{(\mathrm{c})}(\sigma)$ is the energy of a kink-pair at its threshold position and U_{m} is the kink migration energy. By analogy, the velocity of climb in the “jog-collision regime” can be expressed as:

$$v \propto \frac{\sigma}{kT} \exp\left(-\frac{(1/2)U_{\mathrm{jp}}^{(\mathrm{c})}(\sigma) + U_{\mathrm{sd}}}{kT}\right), \tag{2}$$

where $U_{\mathrm{jp}}^{(\mathrm{c})}(\sigma)$ is the energy of a jog-pair at its threshold position and U_{sd} is the self-diffusion energy. The apparent stress exponent is accordingly [11]:

$$n = \frac{\Delta \ln v}{\Delta \ln \sigma} = 1 + \frac{\sigma}{2kT}\frac{\partial U_{\mathrm{jp}}^{(\mathrm{c})}(\sigma)}{\partial \sigma}. \tag{3a}$$

At very high stresses, where usual approximations are not valid, this parameter must however be expressed as (Caillard and Martin [12]):

$$n = \frac{\sigma\Omega}{kT}\frac{\exp(\sigma\Omega/kT)}{\exp(\sigma\Omega/kT) - 1} + \frac{\sigma}{2kT}\frac{\partial U_{\mathrm{jp}}^{(\mathrm{c})}(\sigma)}{\partial \sigma}, \tag{3b}$$

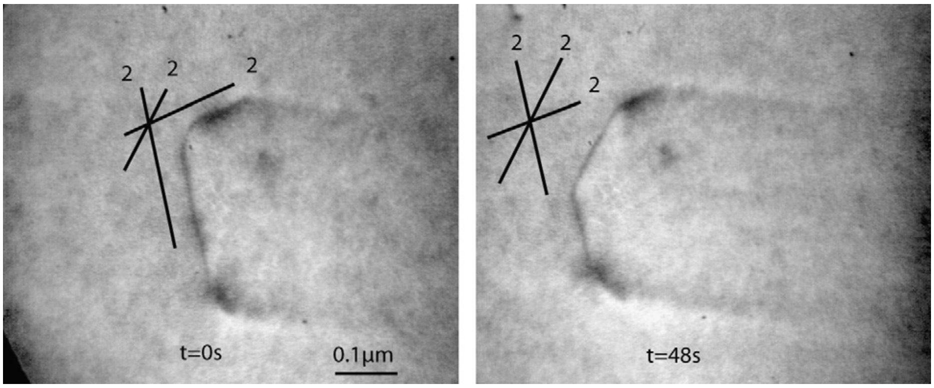

Fig. 2. Dislocation with a five-fold Burgers vector, moving by pure climb in Al–Pd–Mn. In situ experiment at 740 °C, from Ref. [5]. Note the polygonal shape with edges parallel to two-fold directions, showing difficult jog-pair nucleation.

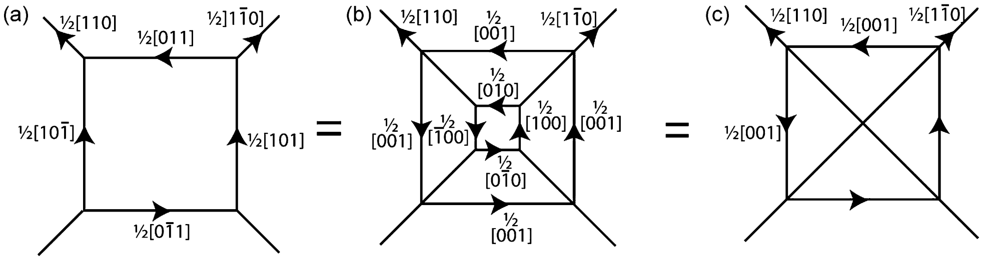

Fig. 3. Diagram showing the equivalence between (a) a γ/γ' interface dislocation networks in a superalloy crept at high temperature, and (b and c) a square network of 1/2[1 1 0] and 1/2[1 $\bar{1}$ 0] dislocations, relaxing the misfit, superimposed to square 1/2[0 0 1] dislocation loops, involved in pure climb deformation. Arrows indicate the orientations of the dislocation lines.

where Ω is the atomic volume. Eq. (3b) yields $n = 1.6$ in Mg, which is definitely larger than unity, but still lower than the experimental value $n = 2.8$. It also yields a too low stress exponent value in Al–Pd–Mn.

3.2. *Possible role of impurities*

It is striking that Eq. (3a), originally derived for semiconductors (with U_{kp} instead of U_{jp}), is generally not verified in these materials either. For instance, dislocation velocity measurements by double etch pits in GaAs and InSb yield n comprised between 1 and 2.6, whereas $n = 1$ would be expected on account of a low applied stress (Erofeeva and Osspyan [13], Choi et al. [14], Yonenaga and Sumino [15]). The origin of this discrepancy can be found in the experiments by Imai and Sumino [16], in Si. These authors measured dislocation velocities as a function of stress and temperature, by X-ray topography. The theoretical value $n = 1$ is obtained only in ultra pure Si, where dislocations remain perfectly straight during their motion. In Si containing various amounts of impurities, the same behaviour is observed only at moderate stresses. Indeed, n increases to 2, and dislocations become wavy, when the stress decreases to low values. These experiments show that Eq. (3) is no more valid when kinks interact with impurity atoms, which occurs preferentially at low stresses. This can be interpreted by a decrease of the mean free path of kinks, due to their interaction with solute atoms (Dour and Estrin [17], Sumino and Yonenaga [18]).

We suggest that the above conclusions can be transposed to climb, although the parameters involved correspond to different physical processes. In other words, the elastic interaction between climbing jogs and atomic-size obstacles should be taken into account.

4. Consequences for other materials

The above considerations show that deformation controlled by pure dislocation climb motion should not be rejected when $\Delta \ln \dot{\varepsilon}/\Delta \ln \sigma > 3$.[1] This concerns in particular materials exhibiting strong indications of extensive dislocation climb, described below.

Fragmentary observations show that prismatic dislocation loops, with [0 0 1] Burgers vectors parallel to the strain axis, develop during the high-temperature deformation of NiAl (Fraser et al. [19], Srinivasan et al. [20]). Mixed climb (namely motion in a plane intermediate between a glide plane and the climb plane perpendicular to the Burgers vector) is also controlling the high-temperature creep of TiAl alloys, according to Malaplate et al. [21]. Lastly, creep of superalloys in the temperature range 700–1000 °C takes place in the same way as above, by dislocation climb and exchange of vacancies between γ/γ' interfaces parallel and perpendicular to the load axis, respectively (Carry and Strudel [22], Svoboda and Lukas [23], Pollock and Argon [24], and Pollock and Field [25]). Results showing that high-temperature creep of superalloys is also controlled by the same kind of mechanism are described below.

Creep of superalloys in tension at very high temperature (1000 °C) generates a raft structure perpendicular to the straining axis (case of a negative misfit between γ and γ'). For a pure climb mechanism, dislocation motion must take place in planes perpendicular to the [0 0 1] straining axis, i.e. in the interface plane. Such a motion has actually been suggested by Pollock and Field [25], and Link et al. [26]. This should increase the vacancy concentration, and activate other systems acting as vacancy sinks. The latter have been observed by Louchet and Ignat [27], Eggeler and Dlouhy [28], Epishin and Link [29], and Link et al. [26]. They consist of dislocations with [1 0 0] and [0 1 0] Burgers vectors perpendicular to the straining axis, climbing in (1 0 0) and (0 1 0) planes, respectively. Note that the existence of such dislocations, moving under osmotic forces only (no climb component due to the applied stress), is a direct proof of a climb mechanism. The nucleation, motion and annihilation of dislocations climbing in the interfaces are now discussed in details. Fig. 3a is the schematic description of one node of the interfacial network, located in a (0 0 1) plane perpendicular to the straining axis. Fig. 3a–c shows that this network can be decomposed into[2]: (i) two sets of intersecting edge dislocations with 1/2[1 1 0] and 1/2[1 $\bar{1}$ 0] Burgers vectors, respectively, which accommodate the misfit between γ and γ', and (ii) square loops with 1/2[0 0 1] Burgers vectors, which are those required for pure climb plasticity. The latter are thought to appear at each intersection point, grow and even-

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[1] and also when the activation energy is different from that of self diffusion, although this aspect is not discussed here.

[2] with no change in the long-range strain and stress fields.

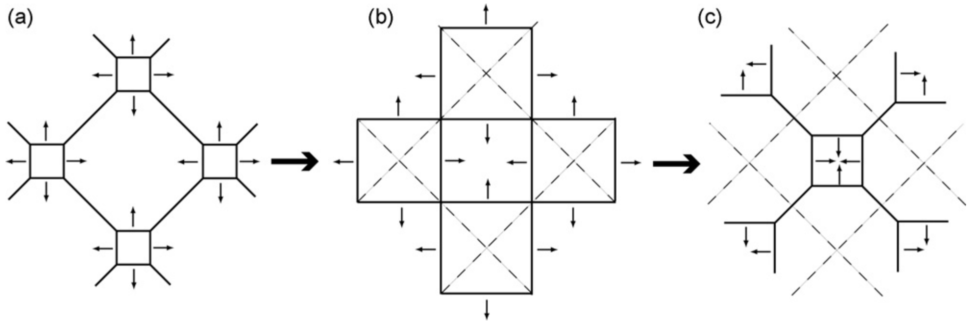

Fig. 4. Schematic description of the nucleation, growth, and annihilation of climbing 1/2[0 0 1] dislocation loops, in the γ/γ′ interface of a superalloy.

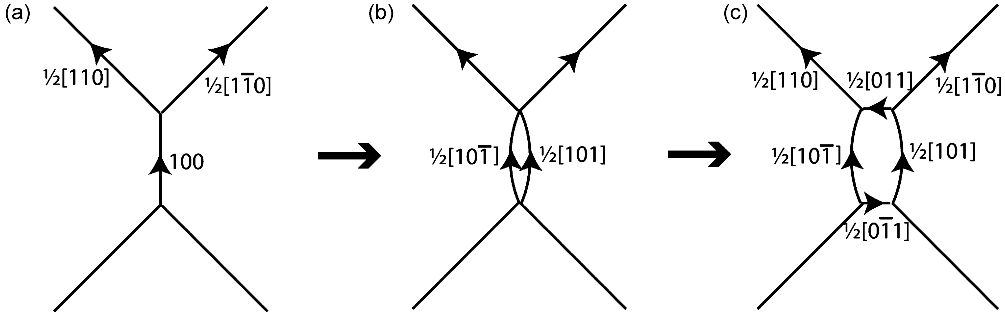

Fig. 5. Nucleation of a dislocation loop at a junction.

tually annihilate, as described schematically in Fig. 4a–c. The nucleation of new square loops can take place at [1 0 0] junction segments (always present in real networks) by the decomposition $[1\,0\,0] \Rightarrow 1/2[1\,0\,\bar{1}] + 1/2[1\,0\,1]$ (Fig. 5a and b). This reaction does not require any extra elastic energy, as the two decomposition products are at right angle. Note that configurations such as in Fig. 4b, with segments curved in opposite directions (on account of opposite climb forces), have been abundantly observed in crept samples (Benyoucef [30]). According to this model, the density of climbing dislocations should be proportional to the density of nodes in the interfacial network, i.e. independent of stress. The stress dependence of the creep rate should accordingly be that of the dislocation climb velocity. However, high internal stresses make the determination of the effective stress exponent difficult [22].

5. Conclusions

In spite of its implication in the high-temperature mechanical properties of materials, dislocation climb remains a poorly known mechanism. It has however been shown that:

- The stress dependence of the climb velocity is more complex than predicted by simple models. In particular, a comparison between the Hirth and Lothe model applied to dislocation glide in semiconductors, and to dislocation climb in metals and alloys, has shown that the nucleation of jog-pairs, and their subsequent interaction with solute atoms, must be taken into account.
- The high-temperature deformation of hcp metals (along the *c*-direction), quasicrystals, and superalloys can be described by the same Nabarro creep mechanism, for which two sets of dislocations move by pure climb and exchange vacancies.

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