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**Inhibited dynamic recovery and screw dislocation annihilation in multiple slip of FCC single crystals**

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3 **Inhibited dynamic recovery and screw dislocation annihilation**  
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6 **in multiple slip of FCC single crystals**  
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22

23 **Abstract**  
24

25 In face-centred cubic crystals, the stress-strain curves of  $\langle 111 \rangle$  crystals exhibit the  
26 peculiarity of being remarkably insensitive to dynamic recovery, as compared to other  
27 orientations. This effect is attested by several early experimental studies on copper,  
28 aluminium and silver crystals. The inhibition of dynamic recovery is attributed to a reduction  
29 of the rate of annihilation of screw dislocations. The underlying mechanisms are investigated  
30 in terms of the reversal of the direction of the resolved stress in the cross-slip plane, which  
31 occurs when going from  $\langle 001 \rangle$  to  $\langle 111 \rangle$  orientations. Two distinct processes governing  
32 dynamic recovery in  $\langle 111 \rangle$  crystals, acute cross-slip and backward annihilation, are  
33 examined and discussed. The second one arises because screw dislocation annihilation occurs  
34 along the direction opposite to that of the resolved applied stress in the cross-slip plane. It is  
35 concluded that, except possibly in aluminium, backward annihilation is always responsible for  
36 the inhibition of dynamic recovery in  $\langle 111 \rangle$  crystals.  
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51 **Keywords:**  
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53 Cross-slip, screw dislocations, dynamic recovery, crystals, stress-strain curves  
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## 1. Introduction

The present work discusses a question that arose during the modelling of strain hardening and its orientation dependence in face-centred cubic (FCC) crystals. Within a storage-recovery framework [1], the work hardening rate is obtained by combining the Taylor relation, which relates the stress to the density of stored dislocations, with an evolutionary law for the stored density. This evolutionary law allows estimating the net density resulting from dislocation storage and annihilation by dynamic recovery. The thermally activated character of this last process is usually thought to originate, for a large part, from the annihilation of screw dislocations by cross-slip. In this context, several current constitutive formulations derive from the one initially proposed by Teodosiu, Raphanel and Tabourot [2]. The parameters entering this type of model comprise the interaction strengths between slip systems that govern the flow stress, the mean free paths of dislocations, which govern the storage process and a recovery length, which is temperature- and strain rate-dependent [3]. The first two sets of parameters derive from elastic, mostly athermal, interactions of dislocations. Thus, they can be estimated from 3D dislocation dynamics simulations, as was done recently for the interaction strengths [4, 5]. It is not so for the recovery length, which is related to the annihilation properties of screw dislocations by cross-slip and, furthermore, depends on the unknown ratio of screw to edge dislocation densities in the microstructure.

Actually, it does not seem possible to rationalise all experimental results on the deformation of FCC single crystals by a single recovery length. Specifically, an examination of the existing experimental literature reveals that  $\langle 111 \rangle$  crystals exhibit an anomalously low rate of dynamic recovery, which is temperature and material dependent. Thus, the objectives

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3 of the present work consist in understanding the mechanisms of this inhibited recovery and  
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5 their relation with screw dislocation cross-slip and annihilation.  
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10 As shown in Section 2, the effect under investigation is clearly evidenced by comparing  
11 the stress-strain curves of  $\langle 001 \rangle$  and  $\langle 111 \rangle$  crystals and their temperature dependencies.  
12 The peculiarities of the  $\langle 111 \rangle$  orientation were mentioned several times in the early  
13 literature. Kocks [6] indicated that in polyslip of single aluminium crystals, there are two  
14 extremely different types of behaviour, typified by  $\langle 001 \rangle$  and  $\langle 111 \rangle$  tension. In silver  
15 crystals, Ramaswami, Kocks and Chalmers [7] noted that the work hardening rate in crystals  
16 of  $\langle 001 \rangle$  orientation decreases above the onset of dynamic recovery at a rate faster than in  
17 crystals of the  $\langle 111 \rangle$  orientation, which is attributed to extensive cross-slip in  $\langle 001 \rangle$   
18 crystals. In addition, several studies [7-9] mentioned that the specific behaviour of  $\langle 111 \rangle$   
19 crystals is somehow related to the fact that the resolved applied stress in the cross-slip plane  
20 has a direction opposite to the one that favours the cross-slip process. This explanation is  
21 investigated in Sections 3 and 4, where two distinct causes for the inhibition of dynamic  
22 recovery are identified and examined. Both derive from the annihilation of screw dislocations  
23 of opposite sign in the particular state of stress that characterises the  $\langle 111 \rangle$  orientation. The  
24 results are presented in terms of the relative ease of cross-slip and dislocation annihilation in  
25  $\langle 111 \rangle$  and  $\langle 001 \rangle$  crystals. A concluding discussion is presented in Section 5.  
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## 50 **2. Experimental aspects**

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55 This section focuses on a comparison between experimental shear stress - shear strain  
56 curves of  $\langle 001 \rangle$  and  $\langle 111 \rangle$  FCC crystals as a function of temperature. These curves present  
57 the characteristic features associated with multiple slip, especially an almost linear initial  
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3 slope followed by a pseudo-parabolic decrease of the strain hardening rate under the effect of  
4 dynamic recovery. The linear slope is akin to stage II of crystals initially oriented for single  
5 glide; it reflects the mostly athermal and material independent mechanism of dislocation  
6 storage. It is present in all deformation curves, but is practically reduced to an initial tangent  
7 at high relative temperatures, due to the onset of dynamic recovery. This last mechanism  
8 becomes significant beyond a more or less well-defined critical stress  $\tau_{III}$ . The subsequent  
9 pseudo-parabolic strain hardening behaviour is strongly temperature and material dependent.  
10 It invades the whole stress-strain curve above a temperature that is correlated with the  
11 stacking fault energy (typically slightly above 300 K in Al, around 500 K in Cu, and at a  
12 somewhat higher temperature in Ag).  
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29 Figure 1 reproduces results obtained by Hosford, Fleischer and Backofen [10] on  
30 aluminium crystals between 4.2 K and 273 K. At the two lowest temperatures, the work  
31 hardening rate is larger for [001] crystals than for  $[\bar{1}11]$  crystals. This is a common feature of  
32 the linear portion of the deformation curve in all FCC crystals. For a given stress, the work  
33 hardening rate is, as a rule, increasing with the number of active slip systems (see e.g. [11]).  
34 The stress-strain curves for the two types of crystals cross each other at 200 K. At 273 K, the  
35 [001] orientation exhibits a smaller strain hardening rate at all stresses, except perhaps at the  
36 onset of plastic flow. One finds similar deformation curves for aluminium at room  
37 temperature in [6] and in a review paper by Clarebrough and Hargreaves [12] (see fig. 27).  
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56 Fig. 1 about here  
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The same trend can be observed in copper single crystals, as illustrated by fig. 2, which  
reproduces experimental results by Kawasaki and Takeuchi [13] from 77 K to 873 K. In this  
case, there is always a cross over between the deformation curves of  $[\bar{1}11]$  and [001] crystals

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3 and the  $[\bar{1}11]$  orientation is always harder at large strains. As temperature increases, the  
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5 position of the cross over decreases toward smaller stress and strain values. The same feature  
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7 is observed in another set of experiments performed in the same range of temperatures [11].  
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9 The reproducibility of the position of the cross over at room temperature can be checked from  
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11 Göttler's work [14] (see fig. 1).  
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17 Fig. 2 about here  
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21 The same pattern of behaviour seems to be obtained in the case of silver crystals, although  
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23 few stress-strain curves were published. From the results published by Mecking and Kocks  
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25 [15] (see fig. 6) and Kocks and Mecking [1] (see fig. 31), one can infer that the strain  
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27 hardening rate is larger for  $\langle 001 \rangle$  crystals than for  $\langle 111 \rangle$  crystals at room temperature and  
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29 that dynamic recovery becomes more active in  $\langle 001 \rangle$  silver crystals than in  $\langle 111 \rangle$  crystals  
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31 at higher temperatures.  
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37 Other experimental results confirm the peculiarities of the  $\langle 111 \rangle$  orientation. When an  
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39 FCC single crystal is prestrained at low temperature and further restrained at a sufficiently  
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41 high temperature, a drop in flow stress is recorded upon restraining. This work softening  
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43 phenomenon was first investigated by Cottrell and Stokes [16]. Miura and Hamashima [9]  
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45 further investigated work softening in aluminium between the prestrain temperature, 77 K,  
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47 and 293 K. The orientations investigated were  $\langle 111 \rangle$ ,  $\langle 001 \rangle$  and a single slip orientation.  
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49 Upon restraining, and after a transient, a significant drop in flow stress was recorded for the  
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51 last two orientations. In parallel, slip trace analysis showed the occurrence of coarse slip lines  
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53 and of extensive cross-slip activity. In contrast, the  $\langle 111 \rangle$  specimens did not exhibit work  
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55 softening and the sample surfaces only showed traces of fine crystallographic slip. The  
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57 authors concluded that work softening exhibits an orientation dependence that is controlled by  
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3 the aptitude to cross-slip. In these experiments, the drop in flow stress can be attributed to the  
4 sudden annihilation by cross-slip of the screw dislocations produced during the prestrain.  
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8 These dislocations could not mutually annihilate at low temperature because dynamic  
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10 recovery was not sufficiently active. Work-softening is particularly marked in aluminium  
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12 crystals since most of the activation energy for cross-slip is furnished by thermal fluctuations  
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14 at the restraining temperature. The particular behaviour of the  $\langle 111 \rangle$  crystals, again, points  
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16 out some inhibition of screw annihilations by cross-slip in this orientation.  
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22 Finally, in a publication by Miura and co-workers [8], it was shown that in silver crystals  
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24 at room temperature, several characteristic resolved stresses become significantly larger close  
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26 to the  $\langle 111 \rangle$  orientation than in other orientations. This concerns the stress  $\tau_{III}$  for the onset  
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28 of dynamic recovery on the stress-strain curves, but also the critical stress for twinning and  
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30 the fracture stress, which all behave in the same manner. The orientation dependence of  $\tau_{III}$   
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32 along the  $\langle 001 \rangle$  -  $\langle 111 \rangle$  boundary is reproduced in fig. 3. These  $\tau_{III}$  values are reasonably  
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34 well-defined in silver at 300 K, since they are much larger than the yield stresses (see fig. 31  
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36 of [1]).  
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41 Fig. 3 about here  
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45 From the examples given above, it appears that the inhibition of dynamic recovery and its  
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47 temperature dependence are more spectacular and manifests themselves in a larger  
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49 temperature range when one goes from aluminium to copper and silver, that is toward  
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51 increasing dissociation widths. This strongly suggests interpreting the inhibition of dynamic  
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53 recovery in  $\langle 111 \rangle$  FCC crystals in terms of an inhibition of the processes leading to the  
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55 annihilation of screw dislocations by cross-slip.  
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### 3. Inhibited screw annihilations

It is generally accepted that the process of dynamic recovery "moves along" [1] with the formation of a dislocation cell structure. Jackson reviewed in some detail this parallel evolution [17], emphasizing the role of cross-slip and screw dislocation annihilations. Specifically, dynamic recovery is usually schematized as follows. A mobile dislocation reaches a dislocation wall and is blocked at some point by a stress field barrier. If a stored screw dislocation of opposite sign exists in the immediate neighbourhood, annihilation by cross-slip may occur. As these events occur in the complex structure of dislocation walls, only simplified estimates can be performed in somehow idealised configurations. Thus, we assume that the incoming screw dislocation is immobilized in its slip plane against a local stress barrier, in the immediate vicinity of the stored screw dislocation. Thus, the two attractive dislocations constitute a dipole, which is embedded into an internal stress field. Their relative positions depend on the local environment and the plane of the dipole should be non-crystallographic in the more general case.

In what follows, we only consider situations such that the incoming dislocation cross-slips and moves toward the stored dislocation. This motion occurs by a succession of slip and cross-slip events, during which interaction forces continuously increase. Nothing is known about the number and length of the glide steps, but this is not critical since what matters during the annihilation process is the interplay between the forces acting on the dislocations in the cross-slip planes. Thus, we define a generic configuration for annihilation by combining together all the cross-slip steps. The two screw dislocations are then set at the intersection of their slip planes with a common cross-slip plane, so that a single cross-slip event is needed to obtain annihilation (see fig. 5, below).

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6 The forces or stresses discussed below result from the externally applied stress and the  
7 attractive interaction between the two screw segments. Annihilation occurs under the  
8 combined effect of these forces provided that the incoming dislocation can first cross-slip and  
9 then move in the cross-slip plane in the direction of the stored screw dislocation. The latter  
10 may be either mobile or immobile in its cross-slip plane. In what follows, it is assumed to be  
11 immobile, but the conclusions drawn below are valid in both cases. This is due to the fact that  
12 the inhibition rates are estimated in relative values for the two orientations  $[001]$  and  $[\bar{1}11]$ ,  
13 so that possible correcting factors cancel out.  
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27 The stereographic plot of fig. 4 shows that the direction of the applied stress resolved in the  
28 cross-slip plane depends on specimen orientation.  
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33 Fig. 4 about here  
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37 We consider crystal orientations along the  $[001]-[\bar{1}11]$  boundary, as in fig. 3. For the  
38  $[001]$  orientation, the angle between the  $[1\bar{1}1]$  normal to the cross-slip plane and the  
39 specimen axis is smaller than  $\pi/2$ . It reaches the value  $\pi/2$  for the  $[\bar{1}12]$  orientation, which  
40 has a null Schmid factor in the cross-slip plane, and is larger than  $\pi/2$  for the  $[111]$   
41 orientation. The angle between the active Burgers vector,  $[\bar{1}01]$  and the slip plane normal is  
42 fixed. Thus, the direction of the applied stress resolved in the cross-slip plane changes sign  
43 when going from  $[001]$  to  $[\bar{1}11]$ . The consequences of this sign reversal on annihilations by  
44 cross-slip are shown in fig. 5. For each orientation of the stress axis, there are two possible  
45 configurations for mutual annihilation, depending upon whether the cross-slipping segment is  
46 above or below the stored one.  
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Fig. 5 about here

We examine first a single screw dislocation with a  $1/2[10\bar{1}]$  Burgers vector, which is not part of a dipole. For a tensile stress axis along  $[001]$  or  $[\bar{1}11]$ , the resolved applied stress in the  $(111)$  slip plane,  $\tau_{ag}$ , is acting along the direction parallel to the Peach-Koehler force,  $[\bar{1}2\bar{1}]$ . The direction of the resolved applied stress in the  $(1\bar{1}1)$  cross-slip plane,  $\tau_{acs}$ , is drawn from a straightforward calculation of the Peach-Koehler force. For the  $[001]$  orientation,  $\tau_{acs}$  is acting along  $[\bar{1}\bar{2}\bar{1}]$  (figs. 5 (a) and (c)), and the two consecutive directions of motion of a cross-slipping dislocation make an obtuse angle. This situation, which is commonly depicted in the literature, corresponds to *obtuse cross-slip*. For a  $[\bar{1}11]$  stress axis,  $\tau_{acs}$  is acting along  $[121]$  and the cross-slip direction makes an acute angle with the direction of motion in the slip plane (figs. 5 (b) and (d)). Thus, taking into account the sole effect of the applied stress, only *acute cross-slip* is permitted in the case of the  $[\bar{1}11]$  orientation. There are a few mentions of acute cross-slip in the early literature. For instance Ramaswami, Kocks and Chalmers [7] discuss it in the case of  $[\bar{1}11]$  silver crystals (see table 1 and fig. 10).

Consider now the additional effect of the interaction stress in the cross-slip plane ( $\tau_{int}$  in fig. 5) with the screw dislocation of opposite sign. The two stresses acting in the cross-slip plane,  $\tau_{acs}$  and  $\tau_{int}$ , are anti-parallel in configurations (b) and (c) of fig. 5. Annihilation only occurs when the interaction stress prevails over the resolved applied stress in the cross-slip plane. We call this process *backward annihilation*. The two stresses are parallel in configurations (a) and (d), which corresponds to the familiar process of *forward annihilation*.

Furthermore, in configurations (c) and (d) of fig. 5, the projection of the interaction force on the slip plane is along the direction opposite to that of the resolved applied stress. This

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3 reduces the glide stress that pushes the incoming dislocation against the obstacle and  
4 constricts it. As a consequence, the cross-slip probability is reduced (Section 4.2), with  
5 respect to configurations (a) and (b). This inhibiting factor is present in the two configurations  
6 such that annihilation can only occur in the acute direction of the cross-slip plane. In what  
7 follows, this process is called *acute annihilation*, by opposition to *obtuse annihilation*, in  
8 which the interaction stress favours the constriction process in the glide plane (figs. 5 (a) and  
9 (b)).

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22 As a result, two distinct effects may inhibit annihilations by cross-slip. The first one is  
23 acute annihilation and the second one is backward annihilation. There is only one easy  
24 configuration for annihilation, obtuse annihilation, which occurs for a [001] stress axis (fig. 5  
25 (a)). In what follows, it will be taken as reference. For each of the two configurations with a  
26  $[\bar{1}11]$  stress axis, annihilation may be inhibited for a different reason, backward annihilation  
27 or acute annihilation (figs. 5 (b) and (d)). For configuration (c), with a [001] stress axis, the  
28 two inhibiting processes are simultaneously present. Going from tension to compression  
29 changes the sign of the applied stresses in the slip and cross-slip planes but results in the same  
30 set of configurations for each orientation. Thus, no tension-compression asymmetry is  
31 predicted.

#### 4. Estimates for the annihilation rate

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53 Annihilation requires first that a cross-slip event occurs and next that the cross-slipped  
54 dislocation moves along the right direction. The objective is here to examine, in relative terms  
55 with respect to the [001] orientation, which of these mechanisms controls dynamic recovery  
56 in the  $[\bar{1}11]$  orientation.  
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#### 4.1. Stresses and distances between dislocations

The interaction stress,  $\tau_{int}$ , is of the form  $\tau_{int} = \mu b / 2\pi d$ , where  $d$  is the distance between the interacting screws. To investigate in dimensionless form the influence of this distance, we introduce a reference length scale  $\ell = \rho^{-1/2}$ , where  $\ell$  is the average distance between dislocation segments in a microstructure containing a uniform density  $\rho$  of forest dislocations. Each the two orientations considered here has identical Schmid factors in the glide and cross-slip planes. With the help of the Taylor relation, the resolved applied stresses  $\tau_{ag}$  and  $\tau_{acs}$  (Section 3) can be written in the form  $|\tau_{acs}| = |\tau_{ag}| = \alpha \mu b / \ell$ , with  $\alpha \approx 0.35$  [18]. Then,

$$\frac{|\tau_{acs}|}{|\tau_{int}|} = \frac{|\tau_{ag}|}{|\tau_{int}|} = 2\pi\alpha \frac{d}{\ell} \quad (1)$$

In cell walls, the average distance between dislocations is smaller than  $\ell$ . Therefore, a dislocation entering a cell wall interacts with nearest-neighbours at a distance  $d$  such that, in average,  $d/\ell < 1/2$ .

#### 4.2. Acute and obtuse annihilations

In Escaig's model [19], the activation energy for cross-slip is of the form  $\Delta G = \Delta G_o(1 - \beta\tau_g)$ , where  $\Delta G_o$  is the total activation energy and  $\tau_g$  is the net glide stress pushing the perfect screw dislocation against the barrier in its slip plane. This linear stress dependence holds for  $\tau_g < \Gamma/b$  ( $\approx 150$  MPa in copper), where  $\Gamma$  is the stacking fault energy. The coefficient  $\beta$  depends on the resolved stresses on the partial dislocations in the slip and cross-slip planes (see equations (1) of [20, 21]). It leads to an orientation dependence of the cross-slip probability, the so-called Escaig effect [21], which however, does not come into play here (cf. fig. 2b of [21]). In addition, there is also no effect due to the interaction with the immobile

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3 screw due the symmetry between acute and obtuse configurations. Therefore, the cross-slip  
4 probabilities for acute (*ac*) and obtuse (*ob*) annihilations are given by Arrhenius forms in  
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6 which all coefficients, including the prefactors, are the same. This justifies writing the ratio  
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10  $r_{csp}$  of these probabilities for in the form

$$11 \quad r_{csp} = \frac{P_{ac}}{P_{ob}} = \frac{\exp(V\tau_{gac}/kT)}{\exp(V\tau_{gob}/kT)}, \quad (2)$$

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14 where  $k$  is the Boltzmann's constant and  $T$  the temperature.  $V$  is the activation volume, which  
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16 is proportional to the square of the dissociation width under zero stress:  $V \propto (d_o/b)^2$ .  
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20 For obtuse annihilations (figs. 5 (a) and (b)), the glide stress  $\tau_{gob}$  in the slip plane is  
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22  $\tau_{gob} = \tau_{ag} + \tau_{int} \cos \theta$ , where  $\theta$  is the acute angle between the normals to the slip and cross-slip  
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24 planes ( $\cos \theta = 1/3$ ). In the case of acute annihilations (figs. 5 (c) and (d)), the effective stress  
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26 in the slip plane is  $\tau_{gac} = \tau_{ag} - \tau_{int} \cos \theta$ . It favours cross-slip, with a reduced probability,  
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28 provided that the glide stress is positive:  $\tau_{int} / \tau_{ag} < \cos \theta$ . From equation (1), this condition  
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30 defines a critical value of  $d/\ell$ ,  $(d/\ell)_c = \cos \theta / 2\pi\alpha \approx 0.15$ . Acute annihilations are only  
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32 possible when the distance between the two screws is above this threshold value.  
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47 With the help of equation (1), the ratio of the cross-slip probabilities given by equation (2),  
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49 is rewritten in the form:

$$50 \quad r_{csp} = \frac{\sinh\left(\frac{\tau_{ag}}{S} \left(1 - \frac{\cos \theta \ell}{2\pi\alpha d}\right)\right)}{\sinh\left(\frac{\tau_{ag}}{S} \left(1 + \frac{\cos \theta \ell}{2\pi\alpha d}\right)\right)}, \quad (3)$$

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3 where  $S = kT/V$  is the strain rate sensitivity. Use is made of hyperbolic sines in order to  
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5 account for inverse activation events occurring under small effective stresses near the  
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7 threshold value [22]. This ratio is plotted in fig. 6 as a function of the dimensionless applied  
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9 stress  $\tau_{ag}/S$  for several values of  $d/\ell$  larger than  $(d/\ell)_c$ .  
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12  
13 Fig. 6 about here  
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17 We consider first the case of copper. From experimental data on the activation volume, the  
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19 strain rate sensitivity is always slightly below or close to  $S \approx 0.7$  MPa between 250 and 410 K  
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21 and decreases outside this range [23, 24]. The ratio  $r_{csp}$  must be understood as the probability  
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23 for acute annihilations in conditions such that the probability for obtuse annihilations is one.  
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25 The resolved stress  $\tau_{ag}$  must then be at least of the order of  $\tau_{III}$  in [001] crystals, which is  
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27 never less than about 30 MPa at temperatures below 400 K (fig. 2 and [13]). Hence,  
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29  $\tau_{ag}/S \geq 40$ , and one can see from fig. 6 that the probability for acute annihilation achieve  
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31 non-vanishing values only when  $d/\ell > 2$ . As screw annihilations typically occur when  
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33  $d/\ell < 1/2$  (Section 4.1), acute annihilations require rather unlikely local fluctuations in  
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35 density and are practically forbidden. The same conclusion holds for materials with medium  
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37 and large dissociation widths, like Ni, Ag and Au, where the strain rate sensitivity is similar to  
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39 that of copper or larger. In aluminium, the dislocation core is narrow, so that the strain rate  
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41 sensitivity is large and the total activation energy for cross-slip is small, as compared to  
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43 copper. In addition, cross-slip is close to its athermal limit at this temperature, as manifested  
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45 in fig. 2 by a sharp bent of the stress-strain curves shortly after the yield stress. From [23, 24],  
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47 we have  $S \approx 3.1$  MPa at 300 K and from fig. 1 and [10], we draw  $\tau_{ag}/S = 2 - 3$  at 273 K and  
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49 slightly more at 200 K. Then, Fig. 6 shows that for  $(d/\ell)_c < d/\ell \leq 1/2$ , acute annihilations can  
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51 occur with a small but non-zero probability, typically between 0.05 and 0.15. Thus, in  
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contrast to other FCC crystals, acute annihilations are possible over a range of temperatures in aluminium, but with a small probability.

### 4.3. Forward and backward annihilations

We now examine the relative ease of forward (figs. 5(a) and (d)) and backward annihilations (figs. 5 (b) and (c)). The effective stresses in the cross-slip plane are denoted  $\tau_f^*$  and  $\tau_b^*$ , respectively. From fig. 5, we have  $\tau_f^* = \tau_{int} + \tau_{acs}$  and  $\tau_b^* = \tau_{int} - \tau_{acs}$ . Making again use of equation (1), the ratios of the two effective stresses for forward and backward annihilations are rewritten:

$$\frac{\tau_f^*}{\tau_{acs}} = \frac{\tau_{int} + \tau_{acs}}{\tau_{acs}} = \frac{\ell}{2\pi\alpha d} + 1; \quad \frac{\tau_b^*}{\tau_{acs}} = \frac{\tau_{int} - \tau_{acs}}{\tau_{acs}} = \frac{\ell}{2\pi\alpha d} - 1 \quad (4)$$

Backward annihilations are forbidden for small distances between the two screws, as they require interaction stresses larger than the applied stress, that is  $d/\ell < 1/2\pi\alpha \approx 0.45$ . A plot of these two dimensionless stresses as a function of the scaled distance between dislocations is shown in fig. 7.

Fig. 7 about here

The relative weight of forward and backward annihilations,  $r_{bf}(d/\ell) = r_{ba}/r_{fa}$ , can be estimated from fig. 7. For  $0.2 < d/\ell < 0.45$ ,  $r_{bf}$  ranges between 0.4 and 0.06. Thus, backward annihilations are, indeed, impeded with respect to forward annihilations but are not vanishingly small. This conclusion derives from the consideration of a purely elastic process that is material-independent, not related to cross-slip probability and is not thermally activated.



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3 In summary, dynamic recovery in [001] crystals essentially occurs by obtuse cross-slip and  
4 forward annihilations (fig. 5(a)). This is because the annihilation process depicted in fig. 5(c)  
5 requires the operation of two inhibited processes, acute cross-slip and backward annihilations.  
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7 In  $[\bar{1}11]$  FCC crystals with medium and large stacking fault energies, acute annihilation (fig.  
8 5(d)) is practically forbidden. Obtuse annihilation with backward annihilation is possible (fig.  
9 5(b)) and this last process is responsible for the inhibition of dynamic recovery. Aluminum,  
10 with its narrow dislocation core, may constitute an exception to this rule, as acute  
11 annihilations are permitted to some extent.  
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## 25 **5. Discussion and concluding remarks**

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29 The present work relies on the assumption that the thermally activated nature of dynamic  
30 recovery and its material dependence can be treated in terms of the cross-slip mechanism.  
31 This view is, however, not unanimously accepted. One objection that is frequently raised is  
32 that dynamic recovery also involves the annihilation of non-screw dislocations (see [25] for  
33 discussion and references). According to recent studies [4, 26], the annihilation of non-screw  
34 segments can locally occur by the interaction of dislocations of same Burgers vectors lying in  
35 their slip and cross-slip planes. These so-called collinear annihilations concern a wide range  
36 of non-screw orientations. However, they occur in a purely mechanical manner and do not  
37 contribute to the global process of dynamic recovery. Thus, we feel it justified to treat the  
38 latter in terms of the annihilation of screw dislocations by cross-slip.  
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55 For the present purposes, which involved simple considerations, cross-slip was treated  
56 using the Escaig's model. A specific model for the annihilation of screw dipoles by cross-slip,  
57 which extends Hirsch's jog model for cross-slip [27], was developed by Brown [28]. In the  
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3 practice, Brown's model should not lead to conclusions different from those obtained here (M.  
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5 Brown, private communication).  
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10 Two main causes for the inhibition of screw dislocation annihilations were investigated,  
11 acute annihilations and backward annihilations. In FCC crystals, with the notable exception of  
12 aluminium, the inhibition of dynamic recovery is governed by backward annihilations. As this  
13 process is athermal, the dynamic recovery length is reduced by a constant factor when going  
14 from [001] to  $[\bar{1}11]$ , whatever the temperature. In aluminium, due to the narrow dislocation  
15 core, acute annihilations may also occur with a reduced rate and within a finite range of  
16 temperatures. This makes the modelling of dynamic recovery more challenging in this  
17 material. Consequences regarding strain hardening will be discussed in a separate study.  
18 Acute cross-slip occurs in 3D dislocation dynamics simulations and tests on dislocation  
19 annihilations at low applied stresses are in progress.  
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36 In materials with crystallographic structures other than FCC, dynamic recovery may be  
37 inhibited for one or several particular orientations. The distinction between obtuse and acute  
38 annihilation only vanishes when the slip and cross-slip planes are at right angles. However,  
39 backward annihilations are always present when the slip geometry is such that the direction of  
40 the resolved stress in the cross-slip plane changes sign within the standard stereographic  
41 triangle.  
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53 In conclusion, the sign reversal of the resolved applied stress in the cross-slip plane  
54 explains the significant differences that were recorded a few decades ago in the mechanical  
55 response of  $\langle 001 \rangle$  and  $\langle 111 \rangle$  crystals. This phenomenon, which is not mentioned in the  
56 current literature, is also important for the modelling of polycrystals. Indeed, in that case, the  
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3 mechanical behaviour of most of the individual grains is similar to that of single crystals with  
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5 <001> and <111> orientations [6, 29].  
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### 10 **Acknowledgements**

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15 The authors are indebted to Dr. W. Pantleon for drawing their attention to the inhibition of  
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17 cross-slip in <111> crystals. Helpful discussions with Prof. J. Bonneville and Dr. U.F. Kocks  
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19 are gratefully acknowledged.  
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### Figure captions

Figure 1. Shear stress - shear strain curves of  $[\bar{1}11]$  and  $[001]$  aluminium crystals between 4.2 K and 273 K. For clarity, the curves are shifted horizontally by a constant value (Redrawn from [10]).

Figure 2. Shear stress - shear strain curves of  $[\bar{1}11]$  and  $[001]$  copper crystals between 77 K and 873 K (Redrawn from [13]).

Figure 3. Orientation dependence of the stress  $\tau_{III}$  for the onset of stage III in silver crystals at 300 K.  $\phi$  is the angle between the  $[001]$  orientation and other orientations along the  $[001]-[\bar{1}11]$  boundary. After Miura, Takamura and Narita [8].

Figure 4. The geometry of cross-slip in the standard stereographic plot. The primary slip system is  $[\bar{1}01](111)$  and the cross-slip system is  $[\bar{1}01](1\bar{1}1)$ . The Schmid factor in the cross-slip plane,  $F_{CS}$ , vanishes along the  $[011]-[\bar{1}12]$  boundary, as the normal to the  $(1\bar{1}1)$  plane is perpendicular to the loading axis.

Figure 5. An incoming screw dislocation (+) interacting with an immobile dislocation of opposite sign (-). The configurations for crystals deformed along  $[001]$  and  $[\bar{1}11]$  stress axes, are shown respectively in the left and right columns. The two consecutive arrows represent the directions of the applied stresses on the (+) dislocation, resolved in the slip and cross-slip planes.  $\tau_{int}$  is the interaction stress acting along the direction of the attractive interaction force in the cross-slip plane. Configurations (a) and (c) correspond to obtuse cross-slip and

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3 configurations (*b*) and (*d*) to acute cross-slip. Forward annihilations occur in configurations  
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5 (*a*) and (*d*) and backward annihilations in configurations (*b*) and (*c*).  
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10 Figure 6. The ratio of cross-slip probabilities for acute and obtuse annihilations as a  
11 function of the resolved applied stress in the slip plane  $\tau_{ag}$ , scaled by the strain rate sensitivity  
12 *S*. This ratio is drawn for several values of the distance between the two attractive  
13 dislocations above the threshold value  $(d/\ell)_{\min} = 0.15$ .  
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22 Figure 7. The effective stresses for forward (*fa*) and backward (*ba*) annihilations, scaled  
23 by the resolved applied stress in the cross-slip plane,  $\tau_{acs}$ , as a function of the scaled distance  
24 *d/ℓ* between the interacting screw dislocations. Backward annihilation can only occur below  
25 the threshold value  $1/2\pi\alpha$ .  
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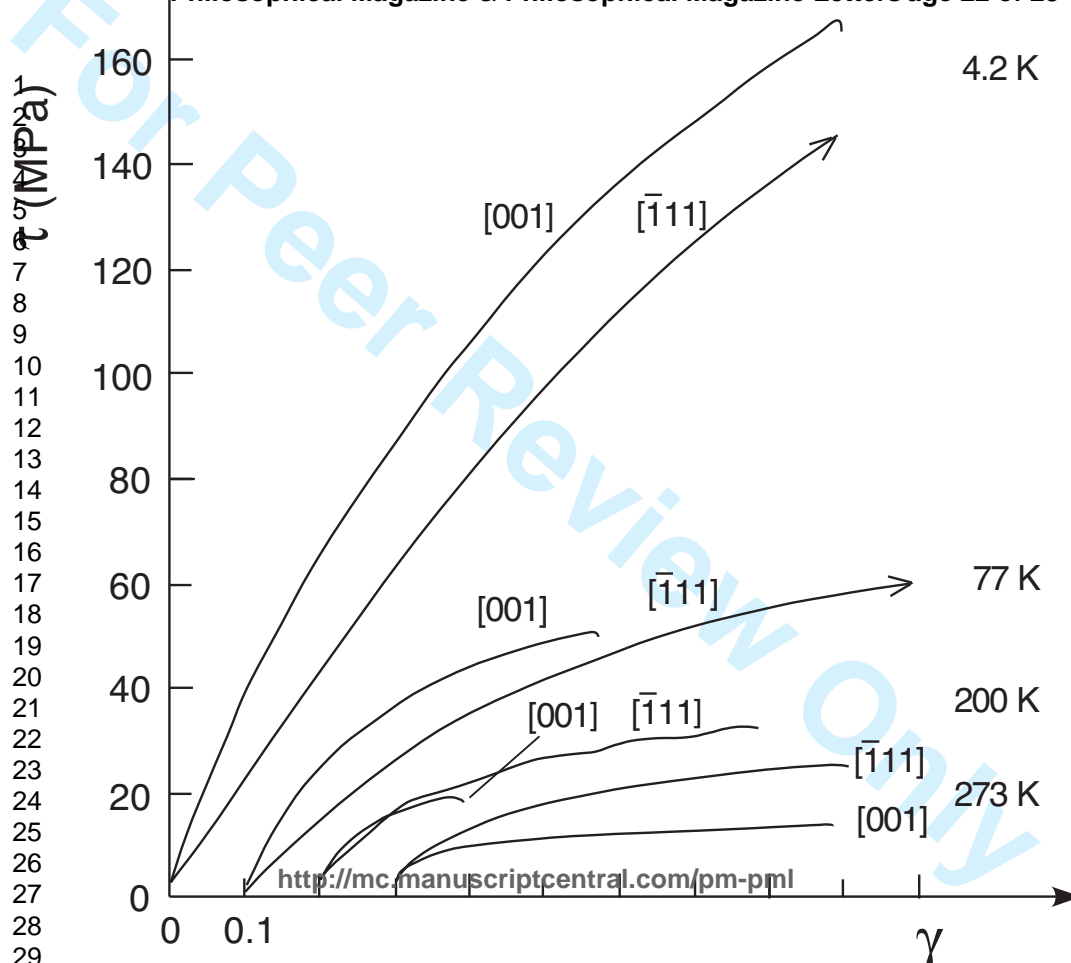
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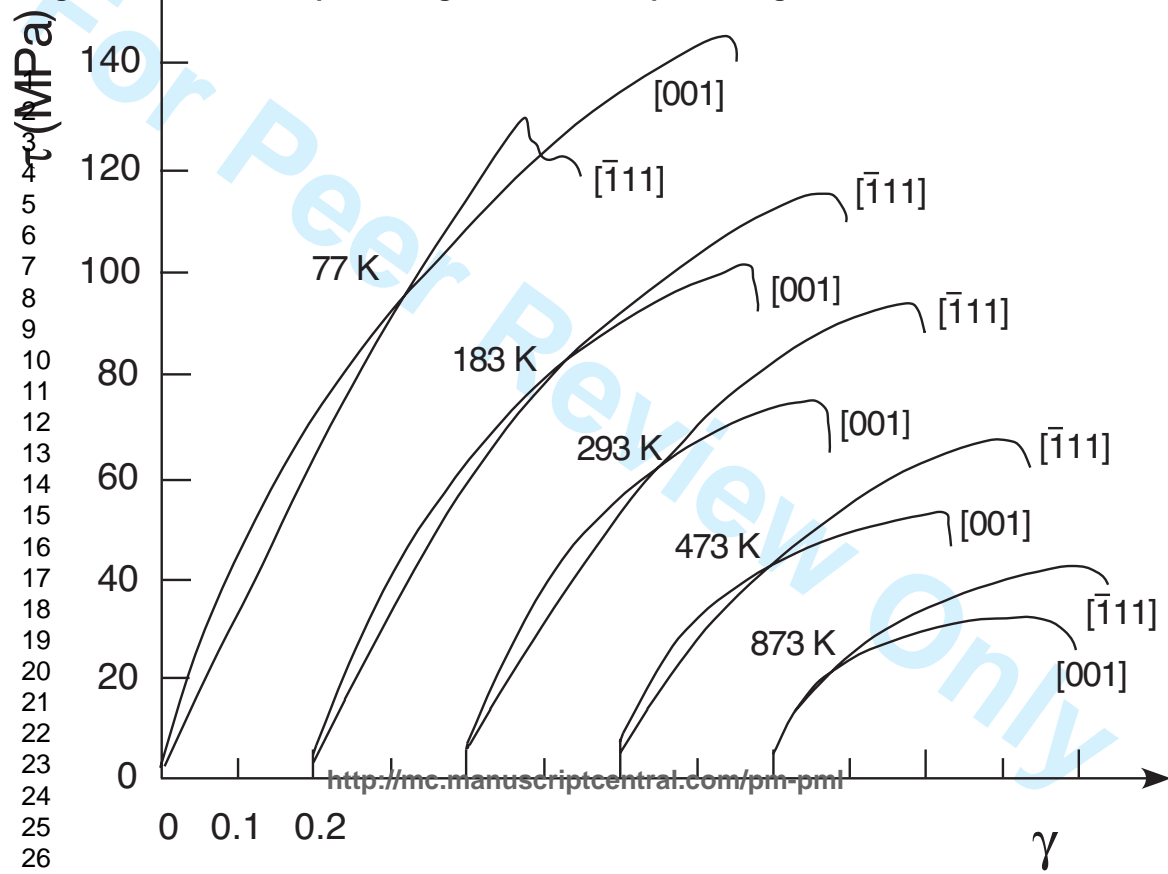
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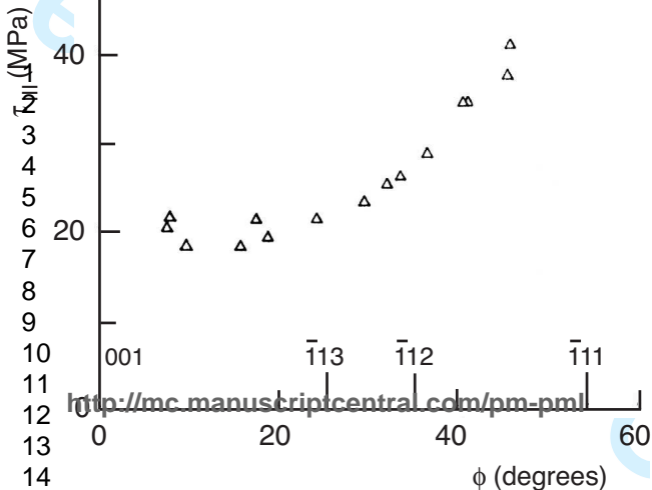
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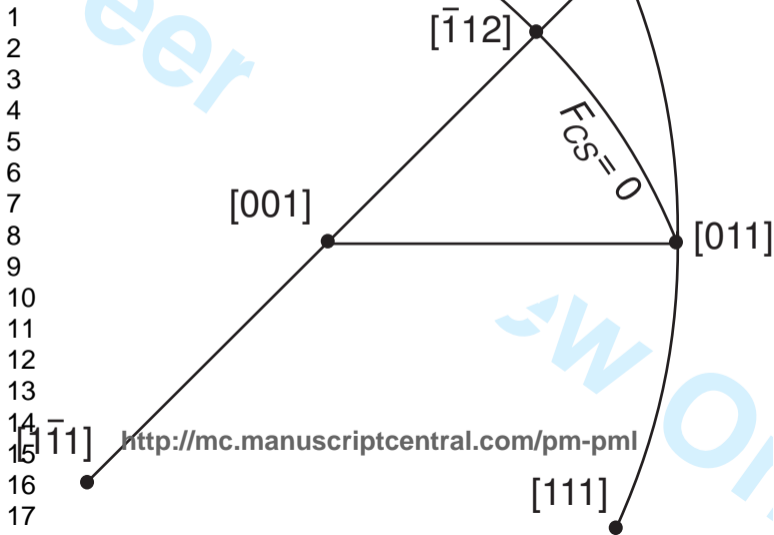
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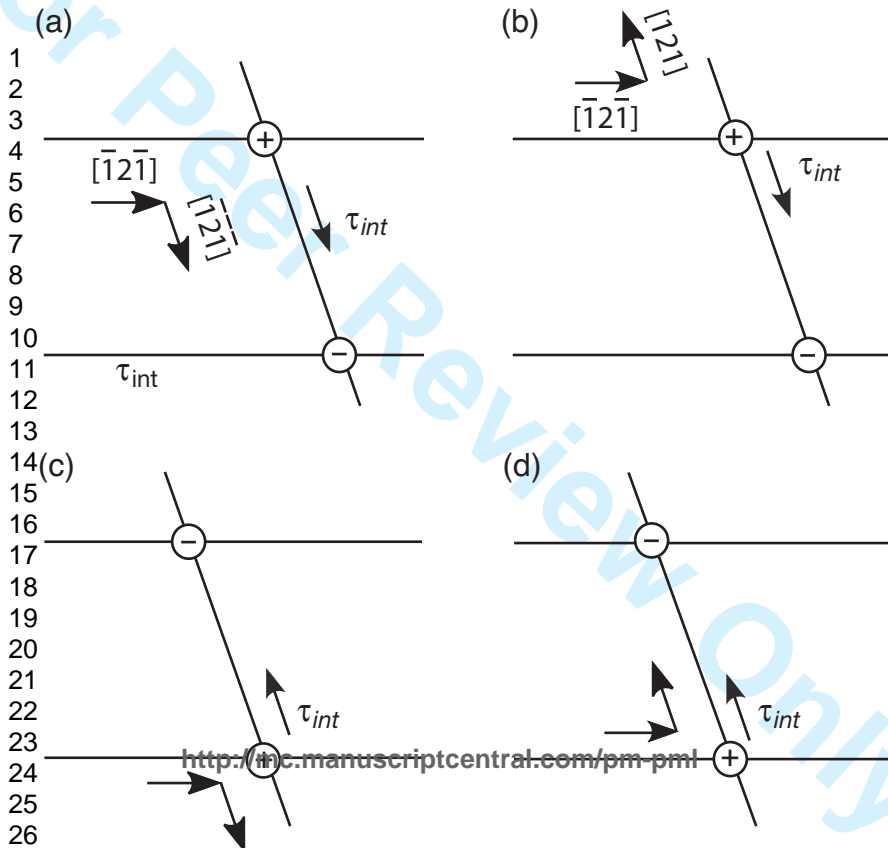


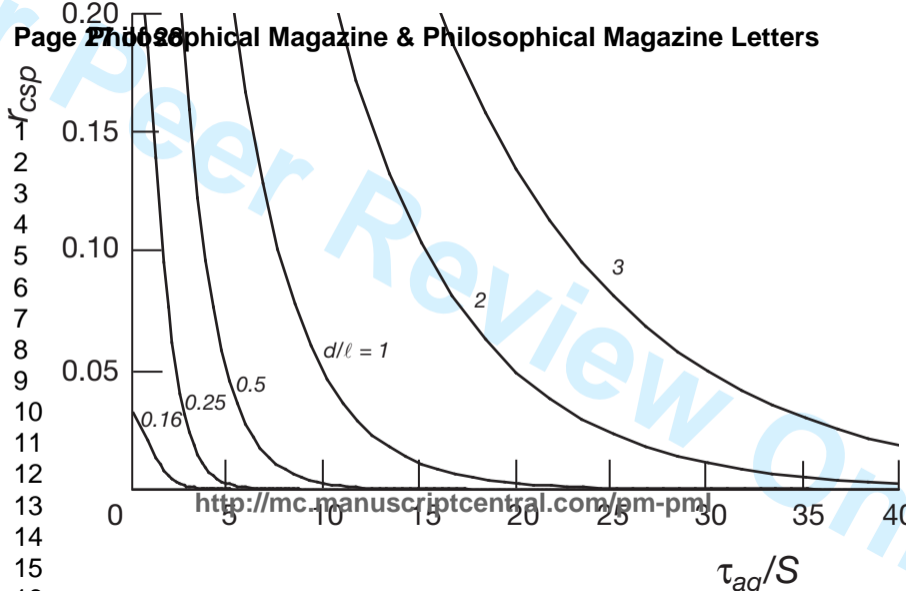


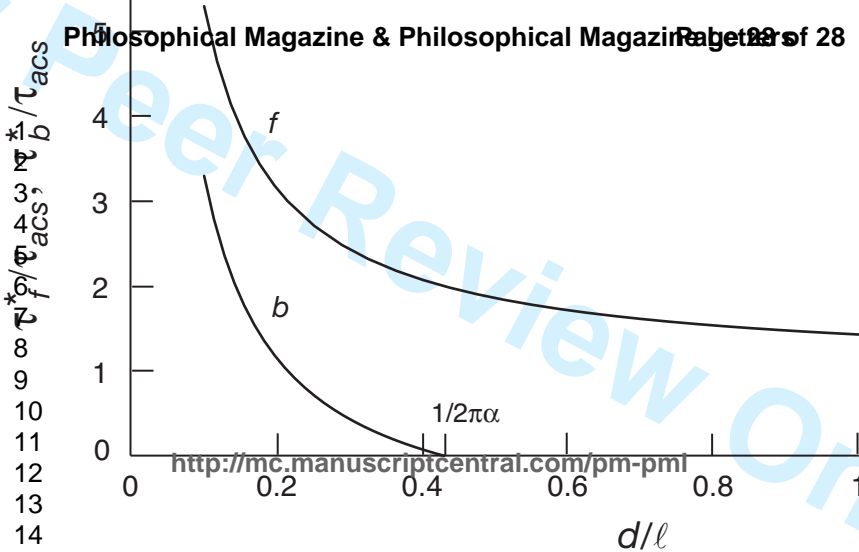


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