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Rate controlling processes in creep of close packed metals at intermediate and high temperatures

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Résumé - De nouveaux modèles microstructuraux de fluage sont proposés qui s'accordent avec les paramètres d'activation expérimentaux et les observations métallographiques. Le glissement non compact des structures HC et CFC peut être décrit par un mécanisme de double décrochement. Il pourrait contrôler le fluage par glissement dans les sous-grains dans Mg aux températures intermédiaires et dans les métaux CFC aux hautes températures. Aux températures intermédiaires, des résultats récents sur le glissement dévié montrent que ce mécanisme pourrait être associé au fluage de Cu alors que dans Al la migration des sous-joints où le glissement non compact intervient, pourrait jouer un rôle important.

Abstract - New microstructural rate controlling models for high temperature deformation are suggested which are in satisfactory agreement not only with activation parameter values but also with various metallographic observations. Non-compact slip in the HCP and FCC structures can be described by a kink pair mechanism. This process could control dislocation glide through subgrains, and therefore the creep rate, for the cases of Mg at intermediate temperatures and FCC metals in some ranges of the high temperature domain. Recent data on cross slip show that it could be rate controlling in Cu at intermediate temperatures. In a similar temperature range, dislocation emission out of subboundaries plays an important role in Al.

1. Introduction

Numerous creep data show that the creep rate in stage II (constant strain rate) depends on various parameters including stress and temperature. In particular, its temperature dependance can be characterized by an activation energy, some examples of which are shown on fig. 1. This parameter seems to exhibit rather constant values within given temperature ranges. For sake of clarity we will define different temperature domains: the high temperature domain corresponds to an activation energy close to that for self diffusion while the intermediate temperature domain has a lower

activation energy. In addition, HCP metals exhibit a ultrahigh temperature range with an activation energy higher than that for self diffusion. The transition temperatures at the limits of the above domains have been discussed [3]. For cubic metals fig. 1a shows that this transition is not a constant fraction of the melting point. It seems to be higher for lower stacking fault energies (for instance, compare Cu and Al). A natural tendency has been to associate these experimentally measured activation energies to given mechanisms which would be rate controlling in the corresponding temperature intervals. These models have been reviewed recently [4], and

Fig. 1 - Activation enthalpies for stage II of creep as a function of temperature.

a) Idealized schematic representation for cubic crystals. Q_{SD} = self diffusion energy. From [1] ;

b) Zinc. Data from 4 different studies (quoted by [2]).

in the most popular ones the creep rate is claimed to be controlled by dislocation climb through self diffusion at high temperatures and by jog dragging screw dislocations for intermediate temperatures. Cross slip is also quoted by some authors although little is known about this dislocation mechanism. Nevertheless, a critical comparison of the creep activation energy with that for self diffusion [5], shows that the agreement can be questioned. The Arrhenius plots used to determine the creep activation energies are not always rectilinear and their slopes can yield energy values larger than those for self diffusion at very high temperatures (as observed for Silver for instance).

Although we believe that a complete description of the creep process is not available at the moment, we will try in this paper to emphasize some dislocation mechanisms which may be worth considering as creep rate controlling processes. Not only the temperature dependence of the creep rate will be considered but also its stress dependence. The latter can be expressed by a stress exponent as well as by an activation area, according to the experimental situations [6]. Metallographic observations are also necessary to derive realistic models. During creep, three elementary processes are necessary and a priori, anyone of these may be rate controlling. Dislocations have to be emitted, annihilated and in addition have to move between sources and sinks. The aim of this paper is to summarize the information and data which are available on these three processes. We will critically examine the role of glide on non compact planes and cross slip as elementary processes.

II. What do we know about glide on non compact planes during creep ?

Compact structures most easily deform by glide on compact planes, namely 111 planes in the FCC structure and the basal plane in

the HCP structure. Nevertheless, it has been known for some time that Magnesium can glide on the prismatic plane under certain circumstances [7] and that Aluminium, Copper and most FCC metals have been observed to glide on {110}, {100}, {112}, etc.. at sufficiently high temperatures [8], [9]. The importance of such glide mechanisms may have been underestimated. We will summarize first the main results concerning prismatic glide in Magnesium and its connection to the creep rate at intermediate temperatures. Then data on non compact slip in Aluminium in connection to high temperature creep will be presented. Both metals have some common features. In Magnesium, dislocations of Burgers vector $1/3 \langle 11\bar{2}0 \rangle$ are dissociated in the basal plane while in Aluminium the dissociation of $1/2 \langle 110 \rangle$ dislocations takes place on one out of the two possible {111} planes containing the Burgers vector. Moreover the width of splitting in the compact plane is weak in both cases (one or a few Burgers vector), the elastic constants are in the same order of magnitude, and the melting temperatures are alike. This should ensure a comparable mechanical behaviour.

However, the advantage of studying Magnesium is that glide on the basal plane which as a rule occurs easily, can be completely neutralized by choosing the proper deformation orientation. This allows one to observe pure prismatic glide and to investigate its properties in a larger stress temperature domain.

II-1. Prismatic glide in Magnesium

The macroscopic properties of prismatic glide have been published (see for instance [10-11]). At constant strain rate, the critical stress decreases as a function of temperature between 50 and 700 K which indicates a thermally activated mechanism [12]. The nature of this mechanism has been

discussed in terms of several contradictory models (see a review in [13]). In an attempt to bring some light in this controversy, microsamples of Magnesium single crystals have been deformed in situ in a Jeol 200 CX electron microscope, operating at 120 kV, below the irradiation threshold [13, 14]. In the microsamples, the basal plane was normal to the foil plane and contained the tensile axis. In such an orientation, only two prismatic glide systems are activated. In this type of experiment, dislocation velocities can be measured and the corresponding local shear stresses can be estimated through the radius of curvature of dislocations. Between 200 and 650 K, the microscopic features of prismatic glide can be described as follows :

i) Dislocation loops exhibit a very **anisotropic behaviour on prismatic planes** : screw portions are long and rectilinear, they move parallel to each other with a strong friction force. Conversely, the non screw portions glide much faster. As an

example, at room temperature the dislocation velocities are such that

$$V_{edge}/V_{screw} > 300$$

An example of gliding dislocation loops observed in situ is shown on fig. 2.
ii) For a given stress, it has been observed that the velocity of a screw dislocation varies linearly with its length.

These features are similar to those observed previously in BCC metals at low temperature, during in situ experiments [15]. They are typical of a Peierls mechanism which is schematically represented on fig. 3. Screw dislocations are split in the basal plane, so that their movement in the prismatic plane P is possible only by nucleating and propagating a pair of kinks of opposite sign. In this description, the only dislocation segments which lie on the non compact plane are the kinks. The rectilinear shape of the screws and the high velocity of the other segments

Fig. 2. Dislocation glide in a prismatic plane of Magnesium. Evidence of friction forces on screw dislocations α , β and γ . In situ experiment. 120 kV. 300 K.

Fig. 3. Schematic representation of a screw dislocation gliding on the prismatic plane P of an HCP crystal following the kink pair mechanism. L = distance between compact rows. $L = c/2 = \sqrt{2/3} a$. The hatched portions of the dislocation lie on the basal plane (perpendicular to P).

indicate that the time for propagating the kinks along the screw portions is much smaller than the time between two nucleations. In contrast to the jogged screw model, which has been proposed to account for a low mobility of screws, the kink pair model can account for the linearity of the screws. It also explains the proportionality between the length and the velocity since the probability of nucleating a kink pair (or the number of nucleation sites) is proportional to the dislocation length. Such a model has been proposed by Yoshinaga and Horiuchi [11] and fully developed by Escaig [16], [17].

Under such conditions, the velocity of screw dislocations can be expressed as:

$$V = \sqrt{2/3} v_d (b^2/\lambda_c^2) L \exp -[\Delta G(\sigma)/kT]$$

with :

v_d = Debye frequency

b = Burgers vector

λ_c = critical length for nucleating a kink pair

L = dislocation length

$\Delta G(\sigma)$ = energy for nucleating a kink pair

The different parameters in this equation can be experimentally measured and compared to their theoretical values. The local stress is obtained from the curvature of non screw portions. The microscopic activation area A defined as :

$$A = (kT/b) \Delta \log V/\Delta \sigma$$

can be derived from the data of fig. 4 and is found to be close to $9b^2$. Such a small value is in agreement with the model predictions. Conventional macroscopic ex-

Fig. 4. Determination of the microscopic activation parameters for dislocation gliding on prismatic planes in Mg (in situ experiments). From [14].

periments provide higher values for the activation area, because of several complex processes which hinder the above one (e.g. stress dependant mobile dislocation density). Using dislocation velocity measurements at another temperature (373 K), the activation enthalpy of prismatic glide has been estimated to be about 0.8 eV at 340 K, a value which should increase up to 1.2 eV at 600 K (which is the athermal temperature for this process).

It is now interesting to compare these data to the results obtained by several authors in macroscopic creep experiments for polycrystals [18-22].

Fig. 5. Comparison of activation enthalpies for creep of polycrystals and for prismatic glide in Mg. Creep data from 4 different studies. Quoted by [14].

The corresponding activation energies are reported on fig. 5. By comparison with the value of the self diffusion energy (about 1.4 eV), these authors conclude that creep is either controlled by edge dislocation climb [18-21] or by the climb of jogs during dragging by screw dislocations [22]. However, a comparison of these creep activation energies with those for dislocations gliding on prismatic planes (fig. 5), shows that the agreement is reasonable. In addition, the creep experiments yield stress exponents ranging between 4 and 10. These values correspond to activation areas of between 50 and 100 b^2 which must be corrected for the stress dependant pre-exponential factor. The corrected values are close to those derived from in situ experiments [13-14].

Therefore screw dislocation motion within the subgrains, by thermally activated glide on the prismatic planes of Magnesium could very well control the creep rate between 400 and 600 K. Such a possibility should be more extensively studied.

II-2. Non compact glide in FCC Metals

Extensive studies of this deformation mechanism in Aluminium and other FCC metals are available. For Aluminium single crystals, the indices of the non compact planes have been determined as a function of the orientation [23]. In various FCC single crystals of [001] orientation (110) glide has been studied as a function of temperature (see e.g. [24]). For each metal, the temperatures at which (110) slip traces are visible have been determined.

As the stacking fault energy increases, the homologous temperature at which non compact glide is present decreases. There seems to be several common features of non compact slip, regardless of the type of activated plane. A complete study of (001) slip has been performed recently in [112] Aluminium single crystals [25] which were deformed both in creep and at constant strain rate (fig. 6).

In creep tests, as the temperature increases above 180 °C, (001) slip traces can be observed. Creep curves are not strongly influenced by the activation of non compact glide, except for the presence of a slight strain rate acceleration at the onset of the deformation. This can be explained by an extremely intense dislocation multiplication in the non compact planes as creep is initiated.

In contrast, stress strain curves appear quite different when the non compact glide systems are activated. At the onset of deformation, they exhibit a sharp yield drop followed by a plateau of low hardening rate. The stress corresponding to the yield drop can be directly related to the cri-

tical resolved shear stress (CRSS) for (001) slip [25]. A similar result has been found for (110) slip in [100] single crystals [24]. The decrease of this CRSS with temperature shows that this non compact slip is strongly thermally activated

Fig. 6. Stereographic projection showing the slip geometry of [112] single crystals. P₁, b₁ and P₂, b₂ are primary compact slip systems. P₃, b₃ has a higher Schmid factor (non compact system).

Fig. 7. Wavy slip lines corresponding to (001) glide after creep at 200 °C of [112] Al single crystals ($\tau = 3,6$ MPa). Scanning electron micrograph. On the left hand side, 111 slip traces are also visible near the sample head.

($\Delta H = 1.7$ eV for (001) glide and 1.4 eV for 110 glide). The slip lines at the sample surface after deformation allow us to define clearly in what temperature range non compact glide is activated. In [112] Aluminium single crystals at the lower temperatures, (001) slip lines appear wavy as shown on fig. 7. This phenomenon can be interpreted as multiple cross slip between compact and non-compact planes and is well explained by the kink pair mechanism extended to FCC metals (see below). At higher temperatures (400 °C) the slip lines become more rectilinear, indicating that non compact glide is more stable. By inspection of the sample shape after deformation, it is also possible to detect the activation of non compact glide. In [112] single crystals, the cross section which is initially circular evolves towards an ellipse. The orientation of the great axis of the ellipse depends on which type of glide is activated, with a difference in orientation of 90 ° between {111} and {001} glide. Another indication that dislocations glide on (001) is the existence of tilt walls which are observed after creep at 400 °C, as shown on fig. 8. The dislocations lie along the $[1\bar{1}0]$ direction, and high resolution electron microscope observations show that the Burgers vectors are $1/2 [110]$, i.e. they are of edge character and glide on (001) planes.

The dislocation mechanism associated with 001 slip has also been studied by different techniques. There is strong experimental evidence supporting a kink pair mechanism :

- Long straight screw dislocations have been observed in [112] single crystals of Al-11 wt% Zn after creep at 250 °C [25], [26]. This alloy which exhibits a creep behavior which is rather similar to Aluminium, offers the advantage that it is

Fig. 8. High resolution electron micrograph of a pure tilt subboundary made of $1/2 [110]$ (001) dislocations. Imaged using 150 kV electrons. SBP = subboundary plane. [110] foil cut from a [112] Al single crystal. Creep test at $T = 400$ °C, $\tau = 0.61$ MPa (Courtesy of M. Mills. To be published).

possible to freeze the dislocations under load [27]. Some examples of the substructure are shown on fig. 9. In addition, analysis of the other dislocation shapes showed that the loops were gliding simultaneously on different planes (pencil glide process), including compact and non-compact

Fig. 9. Straight screw dislocations corresponding to stage I of creep at 250 °C. Al-Zn solid solution. $\sigma = 4$ MPa. Pinned dislocation arrangements (Courtesy of F. Beltzung).

ones. The screw portions of the loops were straight, indicating friction forces on these dislocations [26].

- Wavy slip traces observed in bulk specimens (see fig. 7) and in thin foils during in situ deformation experiments in the electron microscope [25] are in agreement with instabilities of screw dislocations on (001).

This kink pair mechanism was first suggested for glide on non compact planes in FCC metals by Vanderschaeve and Escaig [28]. It has been further developed and confronted to deformation test data [25]. The screw dislocation in this case can split spontaneously into one of the two possible {111} planes. Kink pairs must then be nucleated and propagated on the (001) plane so that the dislocation can move on this plane. It has been verified that for the [112] crystal orientation used in this study, the stress acting on the 1/2 [110] screw dislocation and resolved on the ($\bar{1}11$) and (111) planes produces a force of the same magnitude on both planes (see the stereographic projection of fig. 6). This force is weak due to a low Schmid factor. In addition the applied stress alters the fault width by the same amount on both {111} planes. Consequently,

the screw dislocation has equal probabilities of cross slipping onto the ($\bar{1}11$) and (111), from the (001) plane and back to it. The dislocation motion is schematically represented on fig. 10. It cross slips on ($\bar{1}11$), (111) and (001) with an average motion on (001) which has the strongest Schmid factor. Glide on the two compact planes being friction free, glide on 001 will control the velocity of the screw dislocation. This explains the observation that at around 200 °C during creep of [112] single crystal, glide on (001) is unstable (wavy slip traces). However, at higher temperatures, the probability of nucleating kinks on (001) is larger (i.e. the friction decreases), and non compact slip becomes more planar.

Under such conditions, it is possible to derive the velocity of a screw dislocation along (001), following the same procedure as for prismatic glide [25]. Under the assumption of weak stresses, this yields :

$$v = a(L/\lambda_c^2)v_d b \exp -[\Delta G_{001}/kT]$$

with

a = distance between compact rows in (001)
 ΔG_{001} = activation energy for kink pair nucleation on (001)

Fig. 10. Schematic representation of screw dislocation glide on a (001) plane at temperatures at which non compact slip starts. The screw dislocation is seen edge on. The respective Schmid factors are indicated.

By measuring the critical stress for 001 slip at various temperatures and 3 different strain rates, an activation enthalpy of about $1.7 \text{ eV} \pm 0.2 \text{ eV}$ was found for this mechanism [25]. This value is rather inaccurate because non compact glide is observed in a restricted temperature range, close to the athermal plateau of the yield stress. This is due to the unavoidable competition between compact and non compact slip. As the temperature decreases, {111} slip is easier and dominates. From this enthalpy value, it is possible to deduce the free activation energy ΔG_0 which according to the model, is twice the kink energy U_k on 001. This yields

$$U_k = 0.72 \text{ eV}$$

This value is in reasonable agreement with the model predictions. It has also been checked by using a least square program that the rate equation of the model fits rather well the experimental $\sigma(T)$ dependance measured at 3 strain rates [25]. It is also possible to understand why the homologous temperature corresponding to the onset of non compact glide in different FCC metals increases as the dislocation dissociation increases, following the experimental observations of [29]. It is interesting to note that the need for a kink mechanism to account for the creep rate dependance on the stress, temperature and stacking fault energy has been postulated recently [30], although the physical origin of this mechanism was not proposed.

For the moment, it is difficult to explain why at a given temperature non compact glide takes over for normal slip and what are the factors which control its activation. Indeed, in the $[112]$ single crystals, the samples after deformation are divided into different regions in which either (001) or $\{111\}$ slip has been activated. Even at high temperatures $\{111\}$ slip may be present in constant strain rate tests, while in creep (001) slip takes over completely at 400°C . In this respect, the sample behaves differently in constant strain rate tests and under creep conditions. This may be due to the high deformation rates at the beginning of creep tests which may force non compact sources to operate.

It is also not possible at the moment to describe fully how the activation of non compact slip depends on the crystal orientation. As a first attempt, the Schmid law has been checked at temperatures for which the critical stress for non compact slip is no longer temperature or strain rate dependant. In the case of the two detailed studies now available ([25] for (001) glide and [24] [29] for (110) glide), the Schmid law is obeyed. Using this criterion, a standard triangle can be divided into 5 regions where the $\{111\}$, $\{001\}$, $\{110\}$, $\{112\}$ and $\{113\}$ slip systems exhibit the maximum Schmid factor (fig. 11). Other data [23] are presently being

Fig. 11. Slip planes with maximum Schmid factors for different crystal orientations.

analysed and seem to fit satisfactorily into the regions of fig. 11.

The role of non-compact slip in high temperature deformation of FCC crystals has been ignored or neglected. However, the experimental results reported above on $[112]$ Aluminium single crystals show the importance of 001 glide in creep above 180°C and at constant strain rate above 280°C . For the $[001]$ orientation, (110) is also an active glide system as the temperature increases. For other pure FCC metals, non compact slip has been observed in single crystals of Ni, Au, Cu, Ag [29].

Therefore, the results obtained for Aluminium may be extended to these metals, although this has not been checked systematically.

For Aluminium, the activation enthalpies for non compact slip, are in the range 1.2 to 1.9 eV depending on which system is activated. These values are in quite good agreement with the creep activation enthalpies above 200°C which range between 1.3 and 1.6 eV , according to the present authors [31]. Therefore the mechanism of non compact glide may well be rate controlling at the onset of the high temperature domain. Dislocation climb through self diffusion is frequently quoted for this temperature range, but this mechanism is difficult to detect from metallographic observations. Therefore slip on non compact planes could thus account for the high creep activation energies determined at high temperatures for Silver [5] and Copper [32-33]. Values higher than the self diffusion energy have been found without any convincing interpretations.

In the appropriate high temperature ranges, dislocation glide through the subgrains would proceed on non compact planes and could be rate controlling. In polycrystals, the same mechanisms could apply since creep activation enthalpies for single and polycrystals are comparable. For example, Al-11 wt% polycrystals were crept in stage II at 250°C and the dislocations were pinned under load at the end of the creep test [27]. Long straight screw dislocation segments were observed in thin foils, originating at some of the subboundaries and gliding through the subgrains on non compact planes. According to the grain orientations, compact or non compact slip can be activated. In a given set of grains, those favorably oriented for non compact slip would then deform more slowly at high temperatures, thus imposing the creep rate on the crystal. This effect should be studied more systematically.

III. How are dislocations emitted during creep ?

Recent observations which were performed in situ in Aluminium or after pinning the substructure under load in Al-11 wt% Zn, have shown that dislocations are emitted from subboundaries [27, 31, 34]. Subsequently they move across the subgrains according to the processes described above. Theoretical considerations which have been reviewed in [4], lead to the conclusion that very high local stresses - several times higher than the applied stress - are necessary to achieve this process since the dislocation mesh length in subboundaries is in fact very small. Evidence for such high stresses has been obtained directly at some points of the subboundaries by measuring the high curvature of dislocation segments, during in situ experiments (for instance see fig. 12). In

addition large curvatures are present at some subboundary nodes in creep substructures which were frozen under load in Al-11 wt% Zn during stage II at 250 °C. Local stresses 10 times higher than the applied stress have been measured. (See fig. 13 for an estimation of the local stresses in the crept specimen, as a function of the distance from the subboundary). Such high stresses are expected to be present in every substructure forming material. Indeed, these can be considered as composite materials consisting of hard zones (cell walls or subboundaries) and soft zones (cells or subgrains) which deform together [4]. This model has been

successfully applied to cell or fatigue walls which are dense, not planar, and consist of disordered dislocation configurations. However, the situation is different for subboundaries for several reasons :

i) Unlike dislocation tangles, subboundaries are fully relaxed structures. All dislocation segments have reached a position of minimum energy by climb, i.e. are submitted to a zero stress. In addition, if a local distortion is introduced in the subboundary geometry (e.g. through dislocation insertion), it disappears rapidly by climb.

Fig. 12. High stress build up at a creep subboundary. In situ creep test at 1 MeV, 300 °C on a precrept Al sample. The arrowed dislocation bows out as a function of time. In a) no stress is applied to the sample; b) just after loading R dislocation = 300 Å . c) the dislocation has escaped one node. From [34].

ii) Fig. 12 shows how local stresses are generated at some points of subboundaries during creep. The specimen has been cut out of a precrept sample. In fig. 12a, no stress is applied to the sample and the dislocation segments in the subboundary are more or less straight, as stated in i). As soon as an external stress is applied, the arrowed segment bows out while the others remain straight (Fig. 12b and c)). This is indication that it is submitted to a high local shear stress [35].

It can therefore be concluded that high stresses originate from a stress concentration process inside subboundaries, different from pile up formation which has never been observed. This process must be

faster than recovery by climb, which is very efficient in subboundaries, as mentioned in i).

The proposed mechanism is as follows [35 - 36]. Subboundaries are often observed to migrate, not only in Aluminium, but also in many different materials [4]. This migration is thought to occur by a translation of the dislocation network along a $\langle 110 \rangle$ direction, as sketched on fig. 14a, by a cooperative glide of all dislocation segments. This migration mechanism is in agreement with the creep subboundary geometry [37] and has also been observed in in-situ experiments [38]. However, in the case of real subboundaries containing imperfections, hard points are formed in the network which resist to migration as

Fig. 13. Effective shear stresses in units τ_a as a function from the distance of the subboundary. Al-11 wt% Zn. Creep test at 250 °C, 2,4 MPa. Stage II, 7% strain. From [27a].

sketched on fig. 14 b) to d), in the vicinity of extrinsic dislocations where cooperative movement is not possible. Then when the whole subboundary pulls on these points, dislocations are forced closer together (fig. 14 c) and d)) and high local internal stresses are built up. This process occurs as soon as the external stress is applied and in spite of fast recovery by short distance climb [35, 36].

This model has been developed for creep of Aluminium in the intermediate temperature domain [36]. It has been established

in this case that among the three types of dislocation segments present in the boundary, two are gliding on {111} planes, and one in screw orientation on a {001} type plane. Therefore, the migration velocity is dependent on the rate at which the latter segment can glide along 001. This velocity is proportional to the rate at which high internal stresses are built up, which is turn is related to the emission rate of dislocations out of subboundaries, and therefore the creep rate. Fig. 15 is a schematic representation of the cooperative glide process in the subboundary during migration. On fig. 15a), three dislocation segments in the subboundary network are represented, namely x_1, b_1 , x_2, b_2 and x_3, b_3 , with corresponding slip planes P_1, P_2, P_3 . For migration to occur by glide, a kink has to be nucleated on (100) on the screw dislocation x_1, b_1 so that it may glide on that plane. The shear stress acting on dislocation x_2 will help in nucleating the kink at the lower node, as shown on fig. 15b).

Under such conditions, the subboundary velocity is :

$$V_{migr} = v_d(b^2/h) \exp -(\Delta G_m/kT)$$

with :

h = subboundary mesh size

$\Delta G_m = U_k + \tau b \Delta l - \sigma b A_2$

U_k = kink energy in {100}

τ = energy per unit length of dislocation

Δl = increase in length of x_1 and x_2 from Fig. 15a to 15b, $\Delta l \ll A_2/\tau$

A_2 = activation area $\sim bh/2$ [36]

The deformation rate due to subboundary migration only is :

Fig. 14. Schematic representation of the stress concentration mechanism at creep subboundaries. a) Migration of the subboundary network by glide of the dislocation segments. b) An extrinsic dislocation in the network (dotted line) glides in its own slip plane. c)d) Strong repulsion forces are generated locally during migration (arrows). From [35].

Fig. 15 - Schematic representation of the subboundary migration process (see text).

$$\dot{\epsilon}_{\text{migr}} = (b/Lh) V_{\text{migr}}$$

with L = mean subgrain size.

The total creep rate includes $\dot{\epsilon}_{\text{migr}}$ as well as dislocation emission out of the subboundaries which yields

$$\dot{\epsilon} = (k+1) \dot{\epsilon}_{\text{migr}}$$

with $k = 3$ to 30 between 20 and 200°C [36].

This model provides an explanation for the stress concentration at subboundaries. It accounts for an activation area which is proportional to the subboundary mesh size, in agreement with experimental observations at intermediate temperatures, when the power law of creep breaks down. It predicts a free activation energy ΔG_0 which should be half that for high temperature creep due to dislocation glide on non compact planes through the subgrains (nucleation of one kink in the former case, of a kink pair in the latter). This is in reasonable agreement with the activation energies measured in creep at intermediate and high temperatures respectively [36].

IV. How does dislocation annihilation take place ?

Among the three elementary processes much less is known about dislocation annihilation. In situ experiments have shown that dislocations can be inserted into subboundaries where they annihilate by short distance climb [31]. On the other hand, annihilation inside subgrains is expected to occur by cross slip of adjacent screw segments [39]. Although not directly observed, this mechanism has been proposed to explain the high temperature domain of semi conductors [40], and HCP metals [21]. In all these studies the critical stress for cross slip as well as its temperature and rate dependance was estimated from the τ_{III} properties. This stress corresponds to the onset of stage III on the stress strain curves. Nevertheless a recent study of the cross slip process in Copper has

shown that in fact τ_{III} corresponds to more complex processes. In addition, it was possible to determine the true activation parameters for the former mechanism.

A new method has been developed to create an avalanche of cross slip events at the elastic limit of Copper single crystals, after an appropriate pre deformation [41-42]. Experiments were performed between 150 K and 470 K. Cross slip could be observed between 250 and 407 K, the critical stress decreasing from 27 MPa to 21 MPa over this temperature range. The corresponding activation volume was measured at the elastic limit through load relaxation experiments. True values were obtained through successive load relaxation experiments to account for the machine stiffness. Values close to $350 b^3$, independant of stress and temperature, were obtained between 250 and 407 K. This value is in excellent agreement with the predictions of Escaig's model for cross slip [16, 43], using the dislocation widths which were formerly measured by weak beam experiments in Copper [44]. The variations of the elastic limit and the activation volume with temperature are presented on fig. 16. Corresponding activation energy values could be obtained and ΔG_0 was found to be equal to 1.15 ± 0.37 eV, in reasonable agreement with the model. The validity of Escaig's model of cross slip is also supported by the measured asymmetry of the yield stress in tension and compression [41].

If one compares the above value of ΔG_0 to creep activation energies for Copper, it is not too different from that measured at intermediate temperatures (about 1.45 eV, see fig. 1a)). However, similar data on other metals are required in order to confirm that cross slip is the process which controls at these temperatures.

V. Conclusions

Rather than presenting an extensive review of proposed creep mechanisms (such reviews can be found in [2,6,45-47]), we

Fig. 16 - Critical stress for cross slip and corresponding activation volume V_c as a function of temperature. v = atomic volume. Cross slip takes place at the elastic limit between 250 and 407 K. From [42] .

have tried to propose some outlines for new models which agree with microstructural observations and yield reasonable activation parameter values for creep of compact structures.

Dislocation transport through the subgrains by glide on the prismatic plane could be rate controlling in creep of Magnesium at intermediate temperatures. In situ observations of this non compact glide process have provided useful quantitative information on the activation parameters of this mechanism. Similarly, non compact glide of dislocations through subgrains, on 001, 110... planes may be rate controlling in creep of Aluminium at the onset of the high temperature domain. It could be an alternative mechanism in other FCC metals in the high temperature range when the creep activation energy is higher than that for self diffusion.

However at intermediate temperatures in Aluminium, dislocation emission out of the subboundaries is a significant process with respect to creep. This mechanism is associated with subboundary migration by glide which continuously regenerate high local internal stresses necessary to extract dislocations from subboundaries (in spite of a continuous recovery in the subboundaries by short distance climb). The migration rate is controlled by glide of screw segments of the boundary on 001 planes. This occurs by the nucleation of a single kink at a node, the activation area

of the process being proportional to the mesh size. As the temperature increases, this mechanism becomes athermal and glide through the subgrains on non compact planes becomes rate controlling, through the nucleation of a pair of kinks. Dislocation annihilation has also to occur in creep but information is lacking about this process. The activation parameters for cross slip in Copper are known and different from those defined from τ_{III} values). However, these parameters are in good agreement with those of creep at intermediate temperatures for this metal.

We conclude by identifying these three areas as requiring further investigation:

- i) More data are needed, concerning non compact slip in compact structures,
- ii) the cross slip experiments should be extended to other metals and,
- iii) high temperature activation parameters creep should be remeasured more carefully with emphasis on metallographic information.

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