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Résumé. — On fait un résumé critique des différents processus qui interviennent au cours de la déformation superplastique des polycristaux à grains fins dans des métaux ou des matériaux céramiques. Il apparaît que le glissement intergranulaire et la concentration des contraintes aux points triples jouent un rôle important dans la déformation superplastique. Le mécanisme de la rotation des grains qui s'effectue par cisaillement des couches granulaires est discuté en détail et correspond à la présence d'une structure équiaxe et à une sensibilité de la contrainte à la vitesse de déformation proche de 0.5. Dans le cas limite des contraintes petites, le mécanisme de la rotation des grains se transforme en fluage diffusion.

Abstract. — Various deformation processes taking part in superplastic behaviour of fine-grain metallic as well as ceramic materials are critically reviewed. An essential feature of stable large plastic deformations seems to be grain boundary sliding accompanied by stress concentrations at the triple junctions. Grain rolling induced by the shear deformation of grain layers is discussed in detail as the deformation mode which meets the experimental observations of equiaxed grain shape and of strain rate sensitivity close to 0.5. In the limit case of small stresses, the model based on grain rolling turns into diffusional creep.

1. Introduction.

Enormous strains can be reached by different mechanisms of plastic deformation and consequently there are different types of superplastic behaviour of materials [1]. A large deformation can develop in the regime of diffusional creep (type I superplasticity) or the superplastic flow can be enhanced by the occurrence of a fluid phase at the grain boundaries due to impurity segregation (type II). However, in this paper we will restrict our considerations to the case of fine-grain superplasticity which can be included into the classification by Raj [1], based on the influence of the grain boundary structure on deformation processes, as the type III superplasticity.

The microstructure of fine-grain superplastic materials is characterized by the preservation of equiaxed grains during the deformation. According to electron microscopy observations there is a very little dislocation activity within the grains. The fine-grain superplasticity was
found in metallic materials as well as in ceramic ones [2] in spite of the fact that the character of bonding forces and hence details of deformation processes are quite different. It is surprising how large strains over 600% in nickel-base intermetallic compounds [3] and over 300% in yttria-stabilized tetragonal zirconia [4, 5] were reached by tensile deformation of the materials that are typically brittle at ambient temperature.

Large strains of superplastic deformation seems to occur usually as a result of several competing mechanisms. We should mention here the diffusional transport of matter from compressive to tensile facets which would lead, however, to an elongation of the grains if the tensile facets remain perpendicular to the tensile deformation axis while the compressive facets are parallel to it. Similarly the deformation by intragranular dislocation flow would also cause the grain elongation if the grains do not change their orientation with respect to the deformation axis. However, such a behaviour would contradict the observation according to which the shape of the grains remains equiaxed during the superplastic deformation. It is generally accepted that grain boundary sliding is a dominant mechanism of superplastic deformation [6, 7]. Nevertheless, if the sliding velocity on all facets is assumed to be a constant, what would correspond to a viscous flow, the strain rate will be proportional to the applied stress contrary to the experimental observations according to which the stress exponent is often close to 2 [2, 4, 8]. Then the strain rate sensitivity for fine-grain superplasticity is close to 0.5, i.e. the strain rate is proportional to the second power of the stress. In order to describe such a non-linear behaviour of steady state plastic deformation, it is necessary to consider a non-homogeneous stress distribution which is related to the localization of deformation processes.

Recently, we have proposed a mechanism of superplastic deformation based on grain rolling [9] in which the requirement of the equiaxed grain shape and the changes of grain orientations to the deformation axis are naturally incorporated. It is assumed that the plastic deformation is fully carried by the grain boundaries. The grain sliding, which does not occur on all grain facets in the same way, leads to the stress concentrations at triple junctions that must be accommodated by another deformation process. Then the non-linearity of the strain rate, as a function of the applied stress, follows from the interaction of different deformation processes. Actually we try to put together various mechanisms of superplastic deformation and compose them into a complex process based on the grain rolling. Besides the grain sliding, the diffusional transport of matter [10-12] and the movement of dislocations along the grain interfaces [13-16] are taken into account. Microscopically the grain rolling is modelled by the movement of grain boundary dislocations (GBDs). The concept of grain boundary creep that was discussed essentially in connection with grain boundary sliding in [15], is extended in our model in such a way that GBD climb is also explicitly considered. However, let us point out that GBDs are used in our model only as a mathematical tool in the description of grain rolling, although there is no doubt that GBDs play an important role in high-temperature deformation of polycrystals [14].

It is very likely that grain sliding occurs at least on some grain boundaries by the motion of physically real GBDs. Many configurations of GBDs observed by transmission electron microscopy have been reported in literature. Let us mention here as an example only the observations of GBDs taking part in the grain boundary creep [16]. Obviously the mobility of GBDs is strongly dependent on the type of GBDs which are closely connected with the atomic structure of the grain boundary. The experiments on oriented bicrystals clearly proved that the sliding rate is a function of grain misorientation [17-19]. The sliding rate of general grain boundaries is more than two orders of magnitudes higher than that of special or low-angle boundaries. The effect of grain boundary structure is also manifested by the extreme sensitivity of grain boundary diffusivity to the type of the boundary [20]. On the other hand, we should realize that a complete description of all grain boundaries in fine-grain materials is
practically impossible due to enormous number of parameters needed. As it is discussed in [21] the grain boundary character distribution has a significant influence on the properties of polycrystals but such effects are beyond the scope of this paper.

The triple junctions are supposed to be major obstacles to the motion of GBDs and as places where the pile-ups of GBDs are formed. If it is so the pile-ups of GBDs, considered in our model as a mathematical description of stress concentrations at the triple junctions, may attain a real meaning. The effect of a triple junction on grain sliding was demonstrated experimentally on tricrystals [22, 23]. The sliding displacements were not measured only on the boundaries in tricrystals but also in bicrystals containing the same boundary as one of three boundaries in the tricrystal. It was shown that the sliding rate was suppressed by the triple junction only up to a certain distance from the junction. Therefore, we can conclude that in the range of strain rates corresponding to a superplastic behaviour the plastic deformation seems to be governed by grain boundary sliding and by accommodation of the stress concentrations at triple junctions.


Various mechanisms of very large plastic deformations have been proposed in literature (for a review see for example [24, 25]). A great attention has been paid to the mechanism in which grain boundary sliding is combined with diffusional creep in the process of grain switching [10]. The configuration of four hexagonal grains passes through an intermediate state where one apex angle of all four grains in 90°. The grain switching corresponds to the tensile strain of 0.55. Let us note that the final state can be also obtained from the initial one by the rotation of 90°.

Grain boundary sliding is supposed to be accommodated by intragranular dislocation deformation in [26]. The lattice dislocations pile-up against the grain boundaries and their back stresses block the sliding. The deformation rate is then controlled by the climb of leading dislocations in pile-ups to appropriate annihilation sites along the grain boundaries.

In another model [27], the dislocation movement was considered to be concentrated only to a mantle attached to the grain boundaries. Both models [26] and [27] lead to the same dependence of the strain rate on the applied stress and the grain size, however, they differ in a numerical factor. According to the latter model the deformation rate is much faster what is in agreement with experiments indicating that the processes located at grain boundaries are more significant.

Dislocation processes situated at grain boundaries and their role in the superplastic deformation were considered both for single phase [13, 28] and two phase materials [29]. The interface dislocations can glide or climb according to the orientation of their Burgers vectors to the boundary plane and their motion may be affected by the interaction with the defects of the interface structure such as impurities, second-phase particles, precipitates, ledges etc.

The important feature of grain rearrangements during superplastic deformation of fine-grain materials is a grain rotation. A model in which grain rotations are accompanied by viscous interfacial sliding was analyzed in [30-33]. Since the stress concentrations were not considered in this model, the strain rate was found to be proportional to the external stress. The basic difference from our model consists in the fact that the sliding does not occur on all grain facets when the grains roll over their neighbours because some grains move connected together as explained in the following section. The main idea is that the macroscopic shear deformation of fine-grain polycrystals is not achieved by intragranular dislocation slip but by locally correlated displacements of grains in layers parallel to shear directions. The layers are composed of grains closely packed together into slices with the thickness of the grain size. This type of deformation mode can be described as grain rolling.

In our model the array of grains is divided into layers which are assumed to undergo a shear deformation during tensile straining. The grains in this mode of deformation roll locally together with their neighbours as shown in figure 1. For the sake of simplicity we consider a regular two-dimensional array of hexagonal grains. It is important that all the grains in certain group of grains rotate in the same direction. Notice that the grains do not roll over their neighbours as gear wheels. When the grains hold firmly together on the facets with the lowest inclination to the vertical tensile axis, the grains rotate in the same direction contrary to the revolving gears which move in the opposite directions.

![Diagram](image)

Fig. 1. — a) Shear deformation of grain layers causes the rotation of all grains in the same direction. b) The grain facets can be divided into compressive (double full line), tensile (dashed line) and sliding facets (single full line).

It is apparent that the facets perpendicular to the shear direction experience the largest shear stresses (for the vertical deformation axis) and serve, therefore, as sliding facets. The angle of inclination of grain layers to the deformation axis is always in the range 30°-60°. Since the total volume for shear deformation is preserved, the shortage of matter on the tensile facets almost perpendicular to the tensile axis is to be exactly compensated by the surplus of matter on the compressive facets with the low inclination to the tensile axis. The change of the mutual relations of three grains (initially in close contacts) after a shear deformation without any relaxation of their shapes is shown in figure 2a. It is assumed that the grains can slide relatively easily on the facets perpendicular to the shear direction. On the other hand, no sliding takes place on the facets with compressive and tensile normal stresses for the pairs of grains holding firmly together during the rolling.

It follows from the geometry of grain arrangements that the increase of the rotation angle, $\Delta \phi$, can be expressed by the increment of shear plastic deformation, $\Delta \gamma$, as

$$\Delta \phi = 3 \Delta \gamma / 4 = 3 \Delta \varepsilon / 2 .$$  \hspace{1cm} (1)

For the inclination of shearing layers close to 45°, the corresponding increment of the tensile strain, $\Delta \varepsilon$, is about $\Delta \gamma / 2$.

The grain sliding can be described by glide of grain boundary dislocations which will form pile-ups against the triple junctions (see Fig. 2b). For the grain size $d$, the length of the facet is $d/2$ and the number of dislocations in one branch of the pile-up determined by the theory of linear isotropic elasticity [34] is

$$N_s = (1 - \nu) (d/2 \, b_s) \, (\sigma/2 \, \mu) ,$$  \hspace{1cm} (2)

where $\nu$ is the Poisson's ratio, $b_s$ is the Burgers vector, $\sigma$ is the external stress, and $\mu$ is the shear modulus.
where $\sigma$ is the applied tensile stress, $\mu$ is the shear modulus and $\nu$ is the Poisson ratio. Each pair of dislocations on the sliding facet represents a dislocation loop that shifts the upper grain with respect to the lower grain by the Burgers vector $b_s$. The total shift which is equal to the Burgers vector of a macro-dislocation, $B$, leads to the grain rotation by the angle $\phi$

$$\phi d = N_s b_s = B.$$  

This rotation angle corresponds to a constant prestrain induced by easy dislocation glide on sliding facets but the main part of grain rotations is connected with the diffusion controlled climb of grain boundary dislocations on tensile and compressive facets.

The stress concentration caused at the triple junctions by grain sliding can be relaxed by the transport of matter from the compressive facets with a surplus of matter to the tensile facets with a matter shortage. In order to take into account the effect of the stress applied in this process, let us consider it as a steady state climb of grain boundary dislocations associated with the diffusional flow of vacancies. The dislocations climbing on compressive facets can move only under a steady supply of vacancies which are produced by the dislocations moving along the tensile facets.

We suppose that the acting forces pump the dislocations into the climbing facets through the triple junctions which are then the only sources of climbing dislocations. Actually, the repulsive force due to the sliding dislocation pile-up is the largest at the triple junctions where it strongly pushes all just nucleated dislocations into the climbing facets. Since the interaction forces of the climbing dislocations must balance the force at the tip of the sliding dislocation pile-up, the climbing dislocations form pile-ups similar to those on sliding facets (see Fig. 2b). The density of dislocations is given by the repulsive forces between the climbing dislocations with the Burgers vector of the same sign and increases towards the triple junctions. The dislocation density approaches zero at the middle of climbing facets where the dislocations of the opposite signs meet and annihilate. It can be shown [9] that the number of dislocations in the climbing pile-up is higher than the number for the sliding pile-up (cf. Eq. (2)) by the factor $\sqrt{(3/2)}$

$$N_c = \sqrt{(3/2)(1 - \nu)(d/2 b_s)(\sigma/2 \mu)}.$$  

Thus the average separation of climbing dislocations can be estimated as

$$\bar{L} = d/4 N_c = 2 \mu b_s / \sigma \sqrt{6(1 - \nu)}.$$  

Fig. 2. — a) Mutual positions of the grains induced by the shear deformation corresponding to the rotation angle $\Delta \phi$ (grain contours are depicted by full lines) with respect to the initial grain positions (dashed lines). b) Dislocation representation of stresses arising due to grain rolling.
The pile-ups of dislocations at triple junctions represented by macro-dislocations are also shown in figure 3.

The applied stress, acting on dislocations easily moving on the sliding facet, gives rise to a force at the triple junction opposing all the forces of the dislocations in the sliding pile-up. On the other hand, the dislocations on climbing facet are stuck in their positions waiting for vacancies that would enable them to move. The sum of their back-forces compensates the force due to the sliding pile-up at the triple junction by which new climbing dislocations are supplied.

![Diagram of dislocations and vacancy pile-ups](image)

Fig. 3. — Macro-dislocations at the triple junctions representing the pile-ups of grain boundary dislocations that arise due to grain sliding. The stresses are relaxed by diffusional flow of matter from the compressive to the tensile facets (marked by arrows).

4. Relaxation of stress concentrations.

The grain boundary sliding which builds up stress concentrations at triple junctions should be accompanied by diffusional transport of matter to accommodate such enhanced stresses. Let us consider the transport of matter by means of vacancy flow. The vacancy sources and sinks are merely the climbing dislocations that move in a nonconservative way on climbing facets. In a steady state, the dislocation emission rate is a constant and, therefore, the number of dislocations passing through any point of grain boundaries per unit time is also constant. The vacancies diffuse under the condition of constant rate of vacancy production and/or consumption everywhere along the climbing grain facets. From this condition we can easily determine the vacancy flux which is maximum at the triple junctions and zero at the middle of climbing facets. The vacancy concentration is then given by the diffusion equation. Only the diffusion along grain boundaries, which is much faster than the bulk diffusion, was considered. In principle, it would also be possible to take into account the bulk diffusion of vacancies. However, for the sake of the problem simplicity only the one-dimensional diffusion along grain facets in our two-dimensional model was calculated. The variation of the vacancy concentration, \( c \), along three pairs of climbing facets is shown in figure 4. In average the vacancies move in the direction parallel to the shearing grain layers (see Fig. 3).

The strain rate of tensile deformation, \( \dot{\varepsilon} \), is related to the emission rate of grain boundary dislocations, \( n \), according to equations (1) and (3)

\[
\dot{\varepsilon} = 2 \, \dot{\phi} / 3 = 2 \, nb_z / 3 \, d .
\]

(6)
Fig. 4. — Vacancy distribution along climbing facets. The upper part of the figure shows the undeformed grains which overlap on the compressive facets and make spaces on the tensile facets.

Notice that it is also proportional to the rate of grain rotation, $\phi$. Actually, the dislocation emission rate is a number of dislocations on the length equal to the magnitude of climb velocity, $v_c$. Therefore, the emission rate is inversely proportional to the dislocation spacing

$$n = v_c/L.$$  (7)

It is obvious that the process of superplastic deformation is controlled by the speed at which the matter is transported from the compressive facets to the tensile ones. This speed is determined in our model by the velocity of climbing dislocations.

Following [34] the expression for the climb velocity in the case of grain boundary diffusion is

$$v_c = 3 \delta D_g \Omega \sigma / 2 b_k k TR,$$  (8)

where $D_g$ is the grain boundary diffusion coefficient, $\delta$ is a boundary thickness, $\Omega$ is the atomic volume, $k$ is the Boltzmann constant, $T$ absolute temperature and $R$ is a distance to places with the equilibrium concentration of vacancies.

Finally, considering the average separation of dislocations $\bar{L}$ in equation (7), the constitutive equation for the strain rate of superplastic deformation can be expressed using equations (6)-(8) as

$$\dot{\varepsilon} = \delta D_g \Omega \sigma / k T R \bar{L} d.$$  (9)

It is apparent from figure 4 that the vacancy concentration attains the equilibrium mean value, in average, in the distance of one half of the facet length, i.e. $R = d/4$. Inserting for $\bar{L}$ the expression (5), the strain rate will become

$$\dot{\varepsilon} = 2 \sqrt{6(1 - \nu)} \delta D_g \Omega \sigma^2 / \mu b_k k T d^2.$$  (10)

Since the climb velocity is proportional to the force acting on climbing dislocations, which is proportional to the applied stress, and the spacing of climbing dislocations $\bar{L}$ is inversely proportional to the applied stress, the deformation rate is proportional to the second power of the applied stress as observed experimentally [2, 4, 8]. Moreover, the strain rate is inversely proportional to the second power of the grain size.
5. Discussion.

We should point out that similarly to the Ashby and Verrall grain switching mechanism [10], the plastic deformation of real polycrystals has a stochastic nature since the grains are not as regular as assumed in simplified models. The deformation may start at places where the grain layers are inclined 45° to the loading axis since the shear stress acting on such layers is maximum. In the course of tensile deformation the inclination of grain layers decreases down to 30° because of grain rotation. The resulting grain configuration is then practically stable similarly as in the Ashby-Verrall model and the deformation will continue at other parts of the polycrystal where the grain configurations are more conveniently oriented to the loading axis. Due to the overall motion of the grains, the locally stable grain configurations may become active again when their orientation is altered by the rotations of grain complexes in their neighborhood. The most important difference between our model and the Ashby-Verrall model consists in the fact that there is no intermediate state in our model since the grain rolling is a continuous deformation process during which the grains still preserve their equiaxed shape.

In this paper we have not discussed the influence of the orientation of a grain configuration with respect to the tensile axis on the local strain rate. Strictly speaking all the quantities such as \( N_s, N_{\text{cr}}, \dot{\varepsilon} \) and the forces acting on the dislocations depend on the inclination of shearing layers [9]. However, the expressions can be well approximated by the values for the inclination of 45° lying in the middle of the allowed angle range 30°-60° as it has been done in the previous sections.

The grain rotations as a part of general grain rearrangement during superplastic deformation can be directly observed in scanning electron microscope [35]. It was found that the grains never rotate more than 45° and often change the sense of rotation when the accumulated rotation exceeds the angle of about 20° or 30°. The local tensile strain of 0.55 corresponds to the rearrangement of grain configuration during the plastic deformation from the inclination of shearing layers of 60° with respect to the tensile axis to the inclination of 30°. It means that the whole range of possible orientations of the grain configuration is fully covered by this amount of plastic deformation. The total grain rotation of about 50°, that would arise, is hence the maximum rotation which can be expected to be observed. However, when the deformation starts for the most convenient inclination of 45°, the grains will rotate only up to the angle of about 25°. Then the deformation will proceed in another part of the sample and the observed grains may continue to rotate in a different direction when another system of grain layers becomes conveniently oriented for the shear deformation. This way of behaviour which follows from our model is strongly supported by the experimentally observed changes of the sense of grain rotation.

It is obvious that our model is limited by several simplifications. Because of its two-dimensional character it is not possible to describe in its frame new grains emerging during plastic deformation at free surfaces as it is observed in the scanning electron microscope. Moreover, the assumption that all the grains are regular hexagons of the same size is not also very realistic. Nevertheless, the characteristic feature of superplastic deformation, i.e. the preservation of equiaxed grains, which is intimately related to the grain rearrangement by rotations, is properly considered.

The stress concentrations are introduced to the model by the triple junctions and are described by dislocations pile-ups. Resulting nonlinear dependence of the strain rate on the applied stress is directly associated with the appearance of enhanced stresses at triple junctions. The nonlinearity of the deformation process is a consequence of the competition between grain boundary sliding that occurs only on certain facets and diffusional transport of
matter along the other facets. It means that different modes of deformation are active simultaneously affecting each other. Notice that the uncorrelated grain boundary sliding or diffusional transport of matter (Nabarro-Herring or Coble creep) are both linear processes. If they are treated independently, the strain rate as a linear function of the applied stress would be obtained in both cases.

The superplastic deformation can occur only in a certain range of strain rates. For the slower strain rates corresponding to the low applied stresses, the plastic deformation may undertake a regime of diffusional creep. On the other hand, at the faster strain rates, i.e. at the high stresses, wholly dislocation mechanisms can operate. In our model when the limit of small stresses is considered, the spacing of climbing dislocations can be larger than the facet length, \( \bar{L} > d/4 \). However, since one half of the grain facet is an upper limit for the spacing of neighbouring dislocations, \( d/4 \) should be inserted into equation (9) instead of \( \bar{L} \) from equation (5). Then the strain rate will be proportional to the applied stress and inversely proportional to the third power of the grain size

\[
\dot{\varepsilon} = 16 \delta D_g \Omega \sigma / kT d^3 .
\]  

This is a well known expression for the diffusional creep of Coble type.

6. Conclusions.

The analysis of various processes that may contribute to the development of large superplastic strains indicates that the processes situated at the grain boundaries are the most significant. It has been shown that the essential features of the superplastic behaviour of fine-grain materials can be explained by a special deformation mode based on grain rolling that describes the competition between the sliding on conveniently oriented grain facets and the diffusional transport of matter on the other facets.

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