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A REVIEW.

Microscopic Analysis of Sheet Metal Forming Operations

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ABSTRACT

A microstructural analysis of plastic deformation processes, based on transmission electron microscopy observations, is performed for polycrystalline copper and steel. In the particular case of copper, the stress-strain curves obtained in tension, present different levels depending on the angle between the tensile axis and the rolling direction. This behaviour is not the result of an anisotropic flow behaviour but is connected with the different values of the Taylor factor. There is a coupling effect between the grain size and the loading conditions in the microstructural behaviour of metals. In the tension of small-grained metals, internal stresses, due to the accommodation process between adjacent grains, act on the development of dislocation substructure; in large-grained metals the strain accommodation complexity is not important and the dislocation structure is similar to that of single crystals. In rolling, at low strains, the effect of grain size on the microstructural evolution is not relevant. Whatever the grain size, the imposed strain state leads to a high degree of constraint in each grain and consequently to the activation of a high number of slip systems. With increasing strain the grain size starts to influence the character of the plastic deformation and microbands with or without associated shear develop.

During complex strain paths, the amplitude of the strain path change is the most sensitive parameter controlling the plastic instability phenomenon. For each amplitude, the mechanical behaviour of metals is controlled by microstructural events such as similar sets of slip systems, latent hardening and the Bauschinger effect. The effect of grain size on the microstructural evolution of metals deformed under complex strain paths is shown in a way which is different to that observed during monotonic tests.

Depending on the grain size, severe changes in strain path can produce the development of microinstabilities (microbands) or contribute only towards the destruction of the previous dislocation cells. The stability of grains during deformation is an essential factor contributing to the persistence of the microinstabilities just at high strain levels. On the contrary, when grains rotate the slip rapidly saturates in the microbands. A thermal recovery performed after prestrain does not permit the development of microbands after reloading and accelerates the dynamic recovery processes during the initial stages of the subsequent deformation.

KEYWORDS

Plastic deformation, linear strain paths, complex strain paths, transmission electron microscopy, dislocation substructures, grain size, loading mode, amplitude of the strain path change, grain rotation, thermal recovery.

INTRODUCTION

In sheet metal forming operations, the forming limit is usually governed by strain localization due to plastic instability and fracture. Generally, the analysis of plastic instability phenomena is performed on the basis of the mathematical theory of plasticity. In this case, computational methods for predicting material processing failure have been used. Swift [1] first describes diffuse necking in thin sheets under plane stress states assuming that plastic instability occurs at a load maximum for proportional loading. The limit strains for localized necking condition have been derived by Hill [2] except for biaxial stretched sheets. Marciniak and Kuczinsky [3] have developed a theory (M-K theory) based on the assumption that necking develops from local regions of initial heterogeneity. Barata da Rocha, Barlat and Jalinier have proposed a theoretical approach to plastic instability phenomena in order to predict the forming limit diagrams in both linear and complex strain paths [4]. This model was tested by Gracio, Fernandes and Barata da Rocha on the basis of the comparison between experimental data obtained for copper and steel and the theoretically predicted values [5].

Unfortunately, all these models don't take into account the physical aspects of plastic deformation processes, namely, the response of the internal structures to various imposed conditions. Therefore, through these models one cannot interpret the experimental data correctly. The recent advances in transmission electron microscopy analysis offer new tools for the precise prediction of material behaviour. From the studies, concerning the microscopical aspects of plastic deformation, it appears that one of the main structural parameters influencing the mechanical behaviour of metals is the dislocation substructure. For copper single crystals, it was shown that for tensile axis orientation inside the standard triangle, layered cell structures are formed nearly parallel to the primary slip plane [6]. When tension is performed along a symmetry axis, closed cells appear, giving rise to either rectangular or equiaxed structures [6-8]. For polycrystals, internal stresses due to the accommodation between adjacent grains influence the development of the dislocation substructure. Two different types of models allow the calculation of the intergranular stress field. The first type assumes the stress-strain field to be homogeneous through the whole grain [9]. In this case, multiple slip is imposed inside the grains and closed cells are the predominant feature. In the second type of models every grain, is assumed to be divided in two zones: a grain interior and a grain boundary rim [10-12]. More recently, it was suggested that the deformation of polycrystals involving a reduced number of slip systems may cause the grains to subdivide into volume elements [13]. The principal objection to these simplified models is that they cannot predict the plastic behaviour of predeformed samples along another strain path, since changing a strain path promotes transient effects on the microstructural evolution [14-17].

The aim of this paper is to attempt to clarify the relationship between the macroscopic and microscopic aspects of plastic deformation on the basis of transmission electron microscopy observations. Several results obtained for copper and steel deformed under linear and complex strain paths are presented and discussed.

2. Linear Strain Paths

2.1. Anisotropic Flow Behaviour

One of the fundamental problems concerned with the mechanical behaviour of metals is the anisotropic flow behaviour. For example, in the case of polycrystalline copper with a mean grain size of 20 μm the true stress-true strain curves obtained for seven orientations Φ between the tensile axis and the transverse direction, can be clearly separated into two domains (**Fig.1**) [18, 19]: the curves corresponding to angles Φ of 0°, 15°, 30°, 45° and 60° which are, for all range of strains, at the same level and the curves corresponding to angles Φ of 75° and 90° which present higher level than the others. At first sight, these results are indicative of an anisotropic flow behaviour.

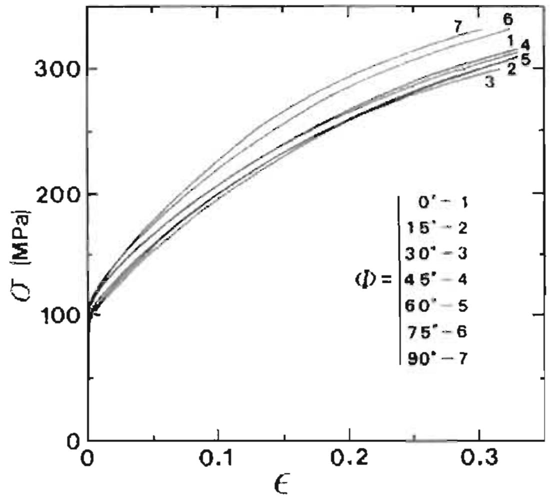


Fig. 1 True stress-true strain curves (obtained during tension of polycrystalline copper) for seven orientations Φ between the tensile axis and the rolling direction.

However, transmission electron microscopy observations have shown that whatever the angle Φ, two families of parallel dislocation walls or closed cells develop in the grains (**Figs. 2, 3** and **4**). A close correlation between the morphology of the cell structure and the number of the operative slip systems exists [6-8], and the presence of such structures shows that multiple slip mechanisms occur in the grains and,

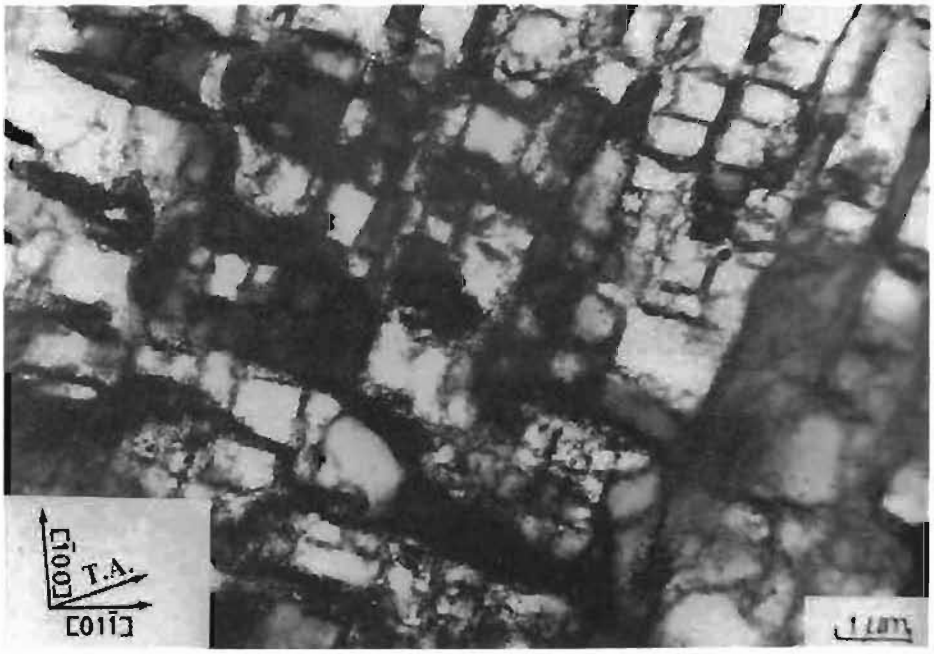


Fig. 2 TEM observation in copper showing closed cells developed in a 20 μm grain with a <110> axis normal to the sheet plane after a strain amount in tension equal to 0.15 (T.A., tension axis, parallel to the rollin direction).

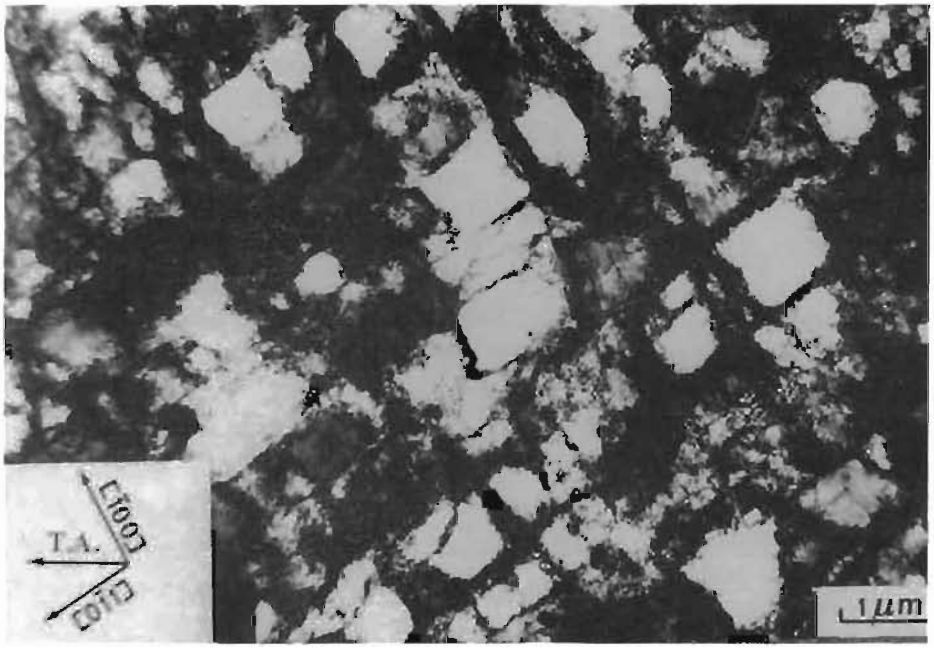


Fig. 3 TEM observation in copper showing closed cells developed in a 20 μm grain with a <110> axis normal to the sheet plane after a strain amount in tension equal to 0.15 (T.A., tension axis, making an angle of 45° with the rolling direction).

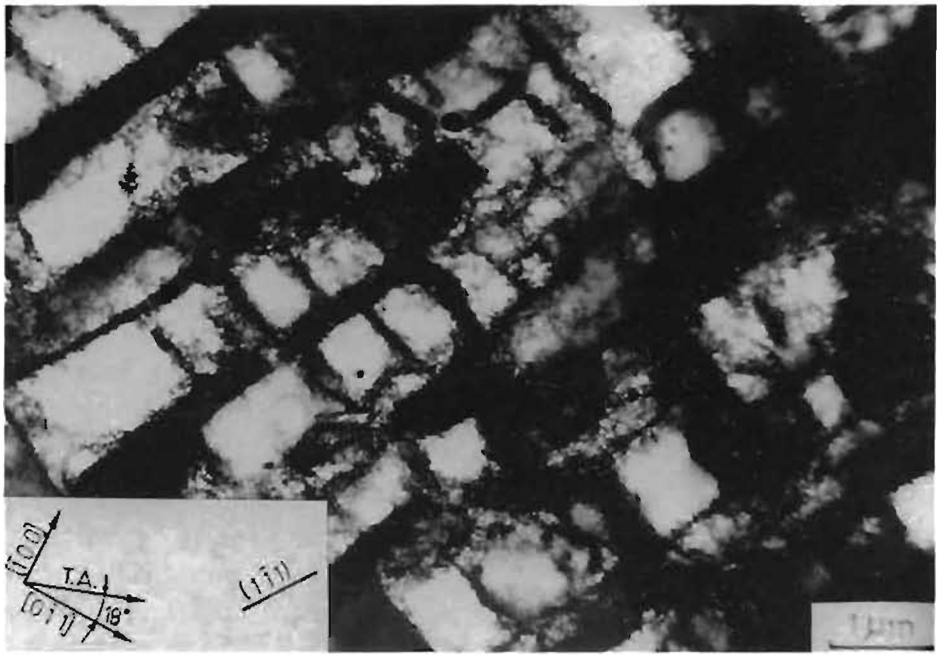


Fig. 4 TEM observation in copper showing closed cells developed in a 20 µm grain with a <110> axis normal to the sheet plane after a strain amount in tension equal to 0.15 (T.A., tension axis, perpendicular to the rolling direction).

consequently, a microstructural behaviour close to that proposed by Taylor takes place [9]. According to the theory of Franciosi and Zaoui [20-22], this type of cell morphology arises from an isotropic intragranular behaviour. On the other hand, it is well known that during monotonic tension the flow stress may be expressed as the product of the average Taylor factor M and the effective resolved shear stress τ which characterises the intragranular behaviour [9, 23]:

$$\sigma = M\tau$$

Because the intragranular behaviour is isotropic it appears from the equation that the different levels of the stress-strain curves are only connected with different values of the Taylor factor.

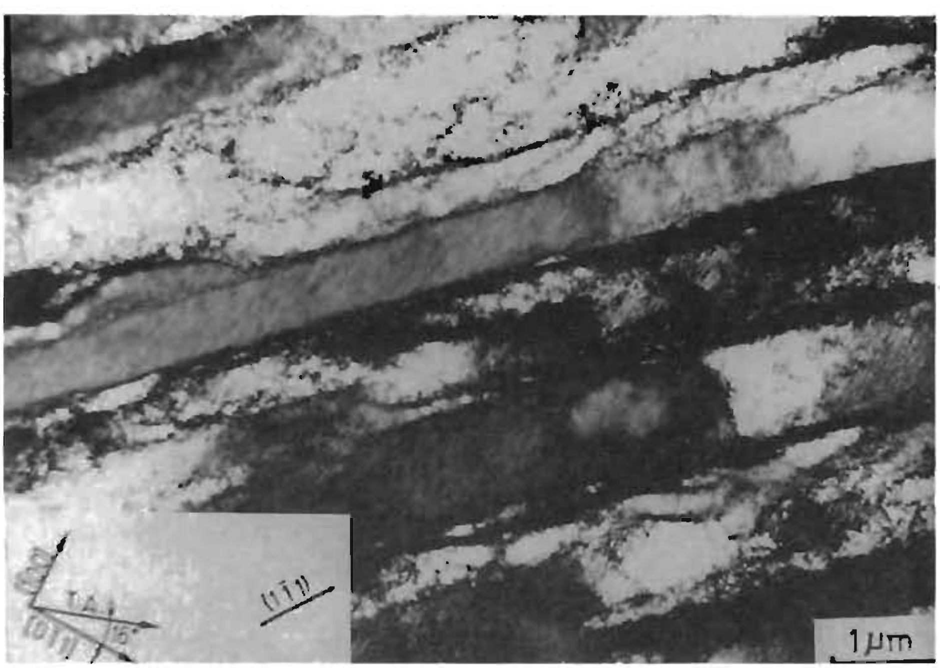


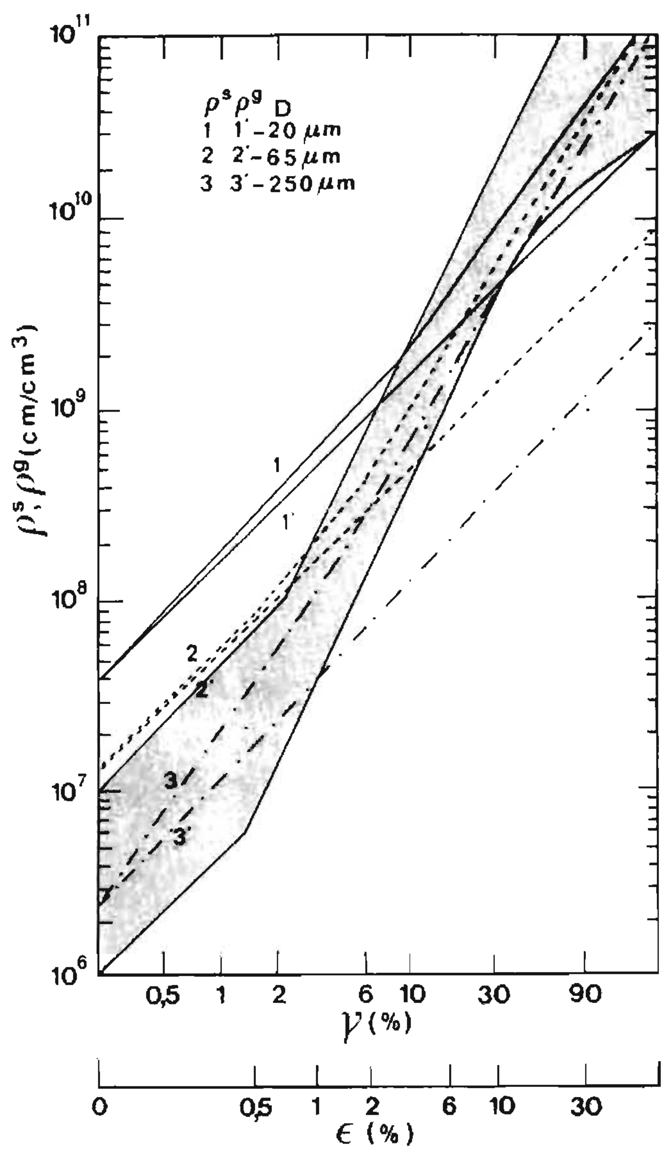
Fig. 5 TEM observation in copper showing one family of dislocation walls developed in a 250 µm grain with a <110> axis normal to the sheet plane after a strain amount of 0.15. (T.A., tension axis).

2.2. The Effect of Grain Size

2.2.1. Tension Deformation

During tension of polycrystalline copper, two different microstructural behaviours have been detected depending on the grain size [24]: for grain sizes between 20 and 65 mm two families of dislocation walls cross each other, leading to closed cells. Rather equiaxed cells are also detected. For grain sizes of 250 mm the microstructural behaviour is similar to that of single crystals: in most of the grains only one family of parallel dislocation walls develops (**Fig. 5**).

The different microstructural behaviours in tension have been explained on the basis of strain accommodation principles [10, 13, 24, 25]. In fact, the single crystal behaviour is influenced only by the presence of statistically stored dislocations. In polycrystals, in addition to the statistical dislocations the strain compatibility between adjacent grains generates a density of geometrically necessary dislocations. The complexity and the range of strains during which the accommodation process is important depends on the grain size [24, 25]. In large grains, after a low percentage of deformation, the statistical dislocations control the strengthening mechanism, as in the case of single crystals. On the contrary, in small grains the influence of the geometrical dislocations is noted just to the beginning of the formation of the cell walls (**Fig 6**). These geometrical dislocations are stored in the grain interior and increase the total density of dislocations, i.e. the stress in the core of the grain. So, in this case, whatever the grain orientation, the cell structure developed is composed at least of two families of parallel dislocation walls. It must be noted that after the development of the cell structure the accommodation is distributed over the cells [24, 26]. This implies that whatever the grain size, the same relation holds between the true stress and the inverse of the mean cell size.



2.2.2. Rolling Deformation

After rolling at low strains, the dislocation substructure is qualitatively similar for a wide range of grain sizes [27, 28]. They consist of a badly definedclosed cell structure (**Fig. 7**). Only from a quantitative point of view have some differences been detected: the cell size is smaller the smaller the grain size. The microstructural results in rolling are, at first sight, in contradiction with those obtained during tension test. However, it must be emphasized that in rolling, the grains

Fig. 6 Variation of the densities of the geometrically necessary dislocations (ρ^g) and of the statistically stored dislocations (ρ^s) with strain, during tension of polycrystalline copper with different grain sizes. In figure is also shown (shaded curve) the variation of the density of the statistically stored dislocations with strain, during tension of copper single crystals.

are more constrained than in tension. So, the existence of a coupling effect between the grain size and the loading conditions must be considered. Namely, in rolling the intragranular behaviour results from the balance between the increase of work hardening, due to an increased number of active slip systems required to accommodate the plastic deformation, and the accommodation work due to a difference between the grain deformation and the average strain.

After the initial stage of deformation, the dislocation structure evolves differently depending on the grain size [29, 30]. In large-grained copper badly defined microbands appear in some grains. At increasing strain, the microbands become well defined structures, and their density increased in relation to the density found at low strain. Moreover, a second family of microbands develop dividing the grains into domains (**Fig. 8**). The space between the microbands is occupied by the cell structure created during the first steps of deformation. In specimens with small grains (35 μm) the development of microbands only occurs at high strain (**Fig. 9**). In this case, large areas of the

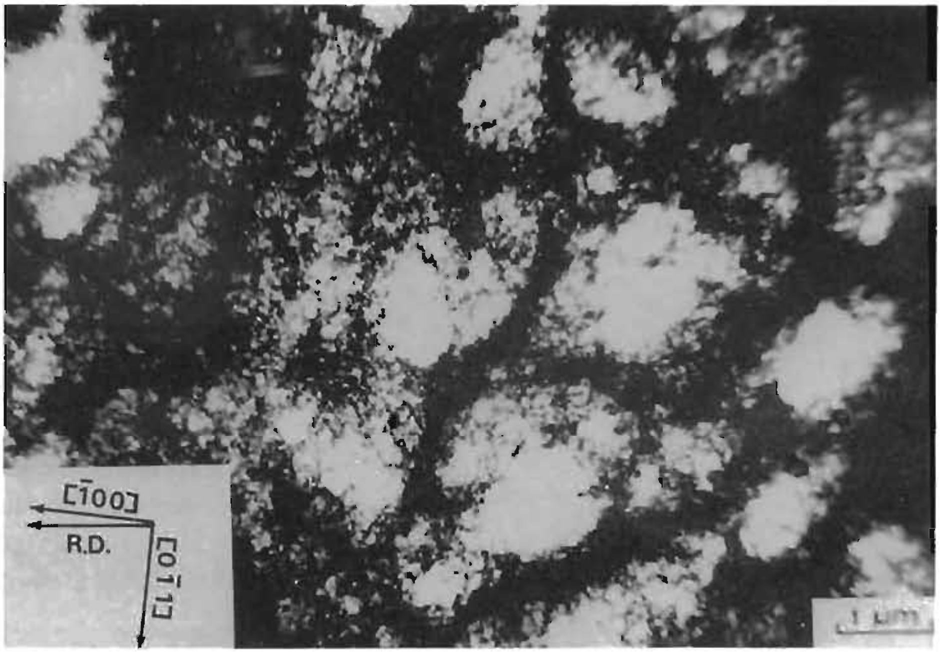


Fig. 7 TEM observation in copper showing equiaxed cells developed in a 250 μm grain after a rolling equivalent strain equal to 0.17. (R.D., rolling direction).

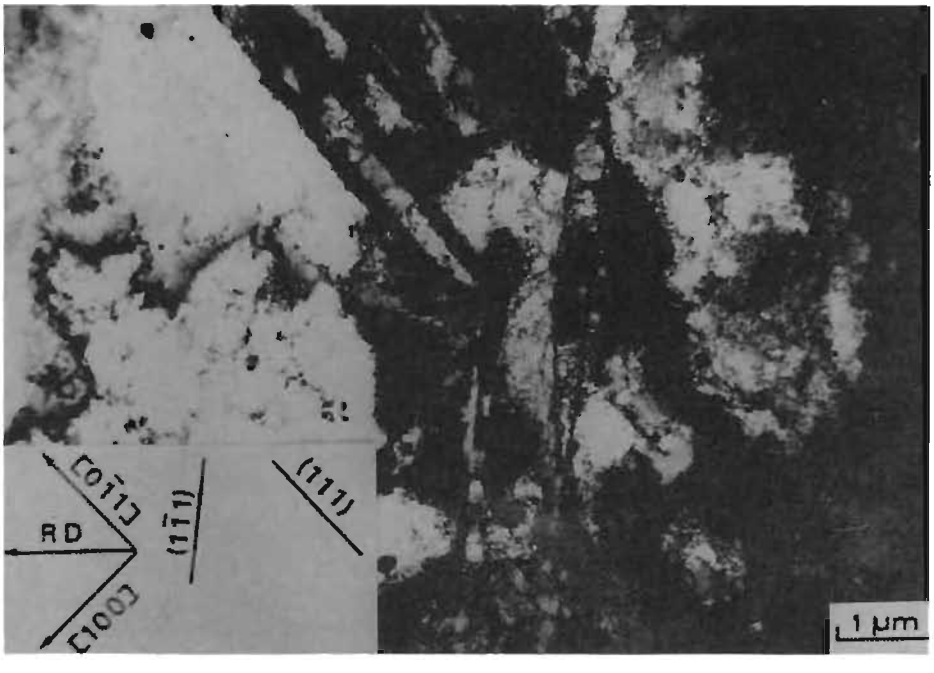


Fig. 8 TEM observation in copper showing two families of microbands developed in a 250 μm grain size specimen deformed in rolling at an equivalent strain equal to 0.45. In spite of the fact that the microbands are along {111} planes, no shear offsets have been detected in the initial structure. The microbands divide the grains into domains. (R.D., rolling direction).

grains are covered by the microbands which produce clear shear offsets in the previous structure. With increasing strain the density of microbands was enhanced, and the dislocation structure formed during the initial steps of deformation disappears completely.

Based on the TEM observations it was possible to determine that the shear displacements produced by the microbands in small grains were relevant, while in large grains significant shearing did not occur [29, 30]. This observation brings about the question about the nature of microbands developed in small and large grains. It was verified that in large grains most of the microbands were not aligned with the {111} planes which indicates that they result from different combinations of slip systems. The function of these microbands is to assure the compatibility between differently deformed domains into which the grains are divided. On the contrary, in small grains the microbands were well aligned with the {111} slip planes, showing that their origin is strictly associated with localised shear. The TEM observations and the consequent crystallographic measurements allow to establish the correlation between the character of polycrystalline deformation, grain size and microstructural instabilities. The function of microbands is constant throughout a wide range of strains, and so the differences in polycrystalline deformation associated with grain size are determined by the presence of one or the other type of microinstability.

3. Complex Strain Paths

3.1. The Amplitude of the Strain Path Change

Industrial stampings of complex shape often involve multistage forming operations and linear strain paths can no longer be observed. A usual way to analyse the formability of sheet metal is by performing sequential strain paths [18, 19, 31, 32]. It was established that, whatever the sequence of loading modes imposed on the materials, they present the same mechanical properties if the cosine of the angle between the two vectors which represents the successive strain tensors is the same. For example, rolling-tension experiments with the tensile axis normal to the rolling direction produces the same mechanical effect of sequential tensile tests with the angle between the tensile axes equal to 55°[18]. This postulate allows one to conclude that the main factor influencing the mechanical behaviour of metals deformed under complex strain paths is the amplitude in the strain path change. The study of the effect of the amplitude in the strain path change on the mechanical behaviour of metals after reloading was performed on the basis of tension experiments [19]. It must be

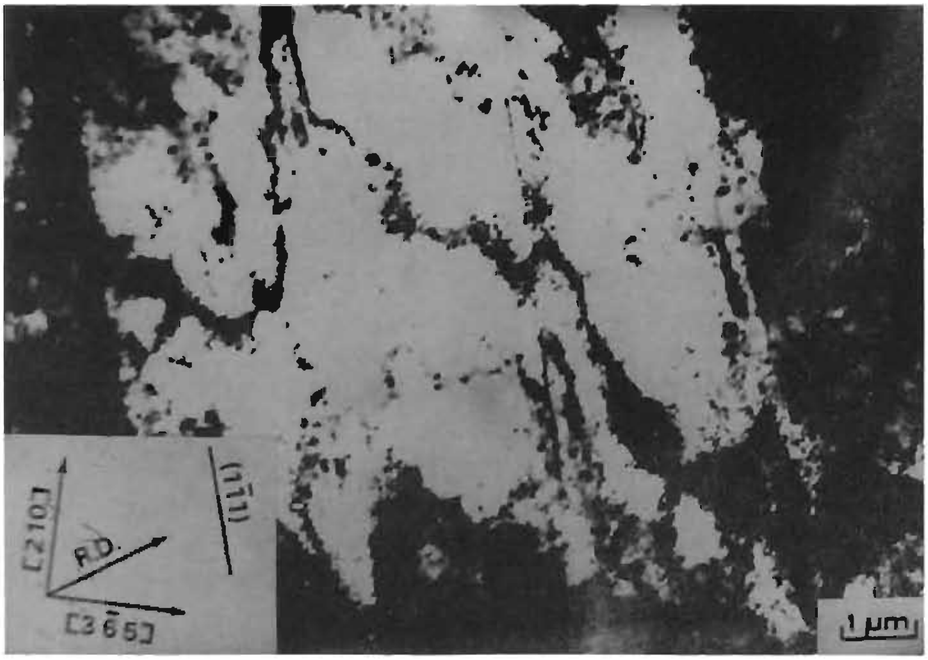


Fig. 9 TEM observation in copper showing microbands developed in a 35 μm grain size specimen deformed in rolling at an equivalent strain equal to 0.45. Note that the microbands are aligned with the {111} plane and produce clear shear offsets in the previous structure. (R.D., rolling direction).

initially noted that different parameters have been employed in order to describe the mechanical behaviour of metals deformed in tension-tension, namely, the reloading yield stress, the residual uniform strain etc. However, the parameter which better characterizes the effect of the change in the strain path on the mechanical behaviour of metals is the normalized reloading stress which corresponds to the ratio between the reloading yield stress (or back extrapolated stress) and the stress attained in a virgin sample deformed up to a strain amount equal to the value of the prestrain. The evolution of the normalized stress with seven angles Φ between the two successive tensile axes for polycrystalline copper is shown in **Figure 10**. The observed behaviour can be summarized as follows [19]:

(i) The higher the prestrain value the higher the level of the normalized reloading stress;

(ii) For Φ higher than 15o the normalized reloading stress increases and reaches its maximum for Φ between 45° and 60°;

(iii) A slight drop appears for Φ around 90o, though the value of the normalized stress is above unity.

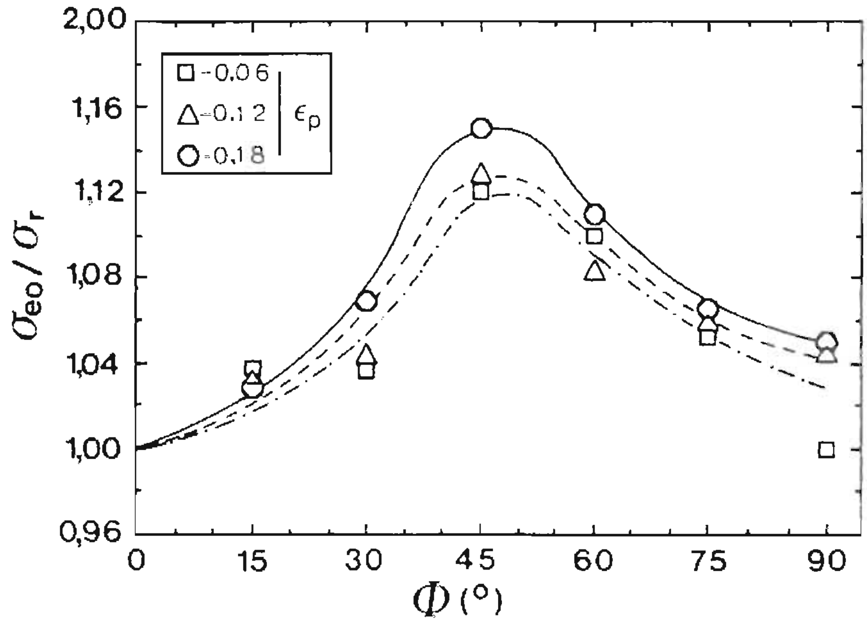


Fig. 10 Variation of the normalized reloading stress with the angle between the two tension axes (results obtained for polycrystalline copper).

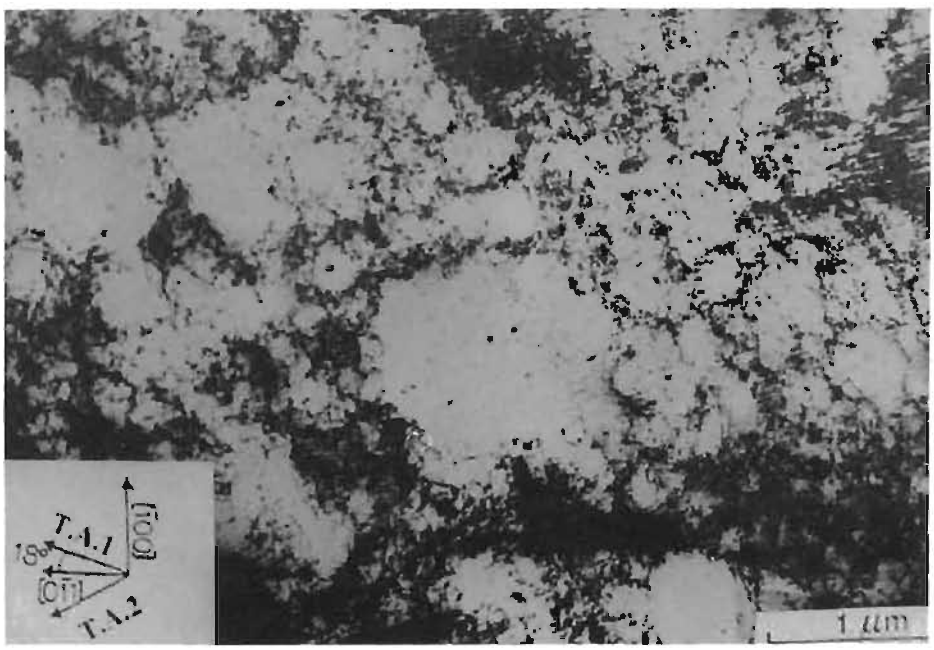


Fig. 11 TEM observation in copper showing a unorganized dislocation structure developed during the second deformation (strain approximately 0.025) in a grain with <110> axis normal to the sheet plane. The prestrain value is equal to 0.12 and the angle between the two tension axes is Φ = 45°. (T.A.1, axis of the first tension; T.A.2, tension axis of the subsequent deformation).

The explanation for such behaviour was given on the basis of TEM observations. For an angle Φ between 0 and 15° no noticeable change of the dislocation arrangement created during prestrain was noted after the change in strain path: the substructure which develops during the prestrain evolves continuously, the same as for monotonic tension. So, the weak evolution of the normalized stress is understood by the fact that in most of the grains the same set of slip systems is active for both strain paths. For larger amplitudes of strain path change (Φ = 45°and 90°) the dislocation walls created during the prestrain tend to disappear. After reloading, the dislocation structure is unorganized in most of the grains and dislocation tangles without any preferential orientation are the predominant feature (**Fig. 11**). These observations clearly show that after the change in the strain path, the activation of new slip systems occurs. Namely, such a strain path change requires the glide of dislocations with another Burgers vector. This fact explains the increase of the normalized stress for Φ values between 15 and 45o. In fact, the requirement of glide of dislocations with a new Burgers vector implies that a very low density of potentially mobile dislocations is available at the beginning of the reloading stage. A high reloading stress is then needed to initiate the multiplication process. For a Φ value equal to 90° an additional phenomenon was observed by TEM [19]. Straight dislocation segments seem to move from the wall vicinity towards the cell interior in the same way as was reported during a Bauschinger experiment (**Fig. 12**). This behaviour is related to the inverse activity on some slip systems during reloading. The internal stresses due to the dislocation walls tend to decrease the critical shear stress on the systems with inverted activity. Since this behaviour counteracts the latent hardening effect on the other slip systems, an overall decrease of the normalized stress occurs.

3.2 The Effect of Grain Size

Considering that the imposed prestrain values are not higher than 0.17 and that the type of preloading mode chosen is a rolling deformation, it appears that whatever the grain size the evolution of the dislocation structure after the change in strain path is independent of the previous structure. When the second loading mode is a tension test the strain accommodation principles followed in monotonic tension appear but in a different way. In the particular case of polycrystalline copper, in large-grained material, microbands develop just after the change in the strain path (**Fig. 13**). With increasing strain the density of microbands becomes higher. After a strain value approximately equal to the value of the prestrain the rolling substructure completely disappears and the grains are covered by one or

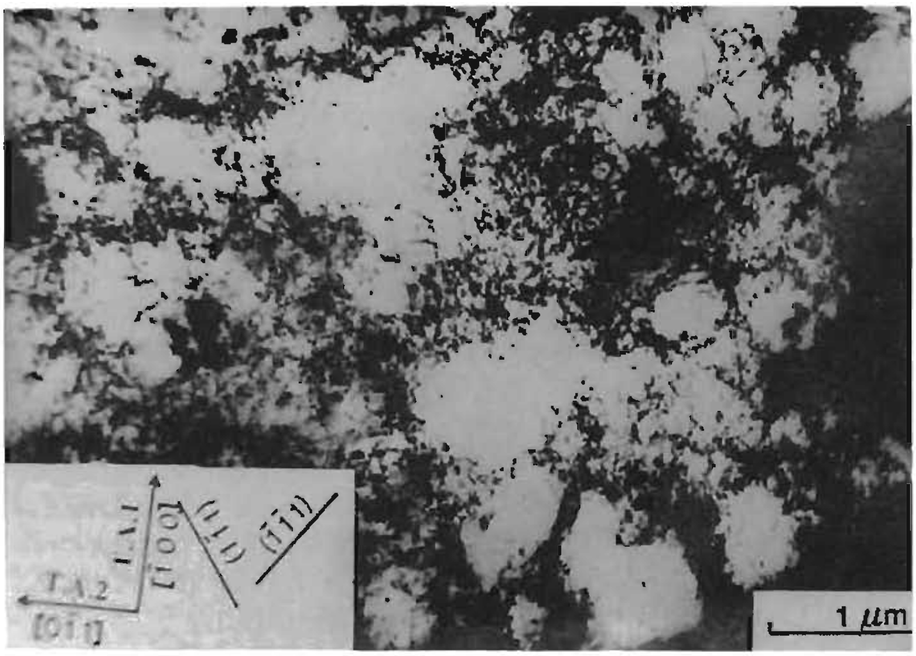


Fig. 12 TEM observation in copper showing the dislocation cell structure developed during the second deformation (strain approximately 0.0025) in a grain with a <110> axis normal to the sheet plane. The prestrain value is equal to 0.12 and the angle between the two tension axes is Φ = 90°. (T.A.1, axis of the first tension; T.A.2, tension axis of the subsequent deformation).

two families of parallel walls. On the contrary, in small grain size specimens, no trace of microscopic localized deformation is noted. The dislocation structure evolves, in a more or less continuous manner by dissolution of the prestrain structure, towards two intersecting families of parallel dislocation walls. On the basis of the calculation of slip activity during the reloading, it was concluded that microbands form only for peculiar grain orientations. Namely, the development of microbands is connected with an intense glide on one slip plane. Moreover, it was shown that there is no collective effect of softening induced by the microbands, i.e., microbands do not develop at the same time in all grains of a transverse section and do not propagate from one grain to another. In fact, the slip rapidly saturates in the microbands leading to an increase of microband density with increasing deformation. The dislocations that belong to the previous dislocation structure rearrange themselves and the frequency of interaction with the new mobile dislocations becomes stronger. Microbands give rise to lower energy configurations by adding dislocations with other Burgers vectors. For small grain sizes, uniform multiple glide finely distributed on two or more non-coplanar systems is necessary to accommodate the plastic deformation. The plastic flow inside the grains is stable and the previous microstructure evolves gradually towards that typical of the current path.

3.3. The Grain Rotation

A point of difference between the microbands which develop during rolling and during rolling-tension experiments is that in the last case the presence of microbands is not linked with a clear shear of the previous microstructure [27, 30]. Two main reasons have been pointed out to justify such differences. Localised shear occurs in the shear bands but the shear direction does not lie on the observation plane; the band development follows a two step mechanism. Recent experiments have proven the validity of the second hypothesis [34]. In fact, an important localised shearing occurs in microbands only for stable grain orientations compatible with a predominant planar slip. This analysis was based on the comparison between the microstructural behaviour of copper deformed in tension-tension (with the angle between the two tensile axes equal to 55°) and tension-shear experiments (with the shear direction normal to the tensile axis). Both complex strain paths lead to microband development after reloading (**Fig. 14**). However, during tension-tension experiments, the simple shearing associated with the activity of the microbands causes a rapid grain rotation. In this case, the behaviour of grains with microbands is close to that of a single crystal oriented for single slip. This means that after a certain value of strain, the requeriment of intragranular

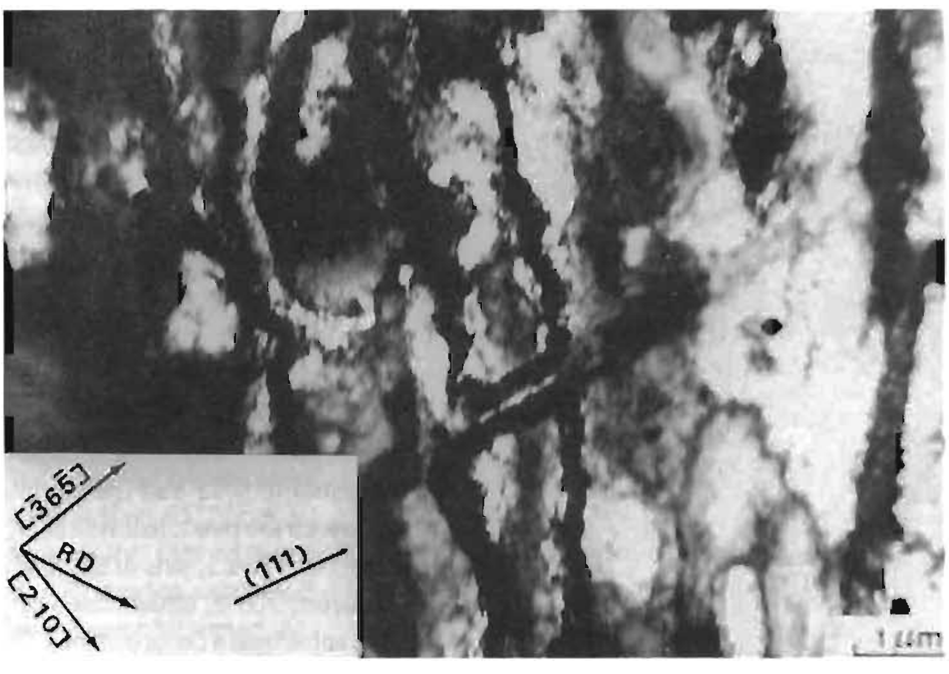


Fig. 13 TEM observation in copper showing a microband developed in a 250 μm grain size specimen deformed in tension (0.015) after prestrain in rolling (0.17). (T.A., tension axis).

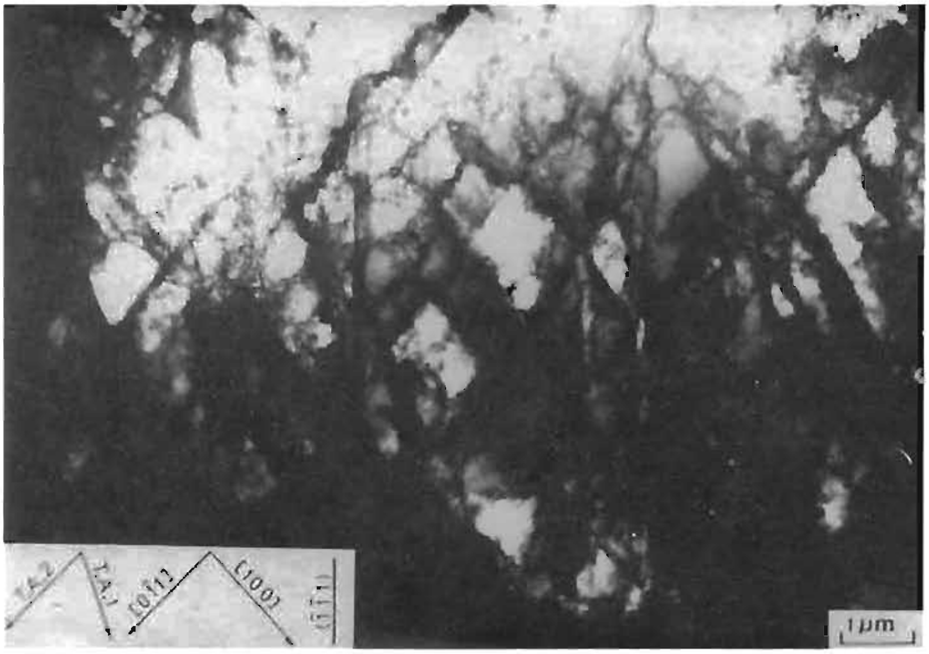


Fig. 14 TEM observation in copper showing microbands in a sample deformed in tension (0.015) after prestraining in tension (0.12). (T.A.1 and T.A.2, axes of the first and second tensions respectively).

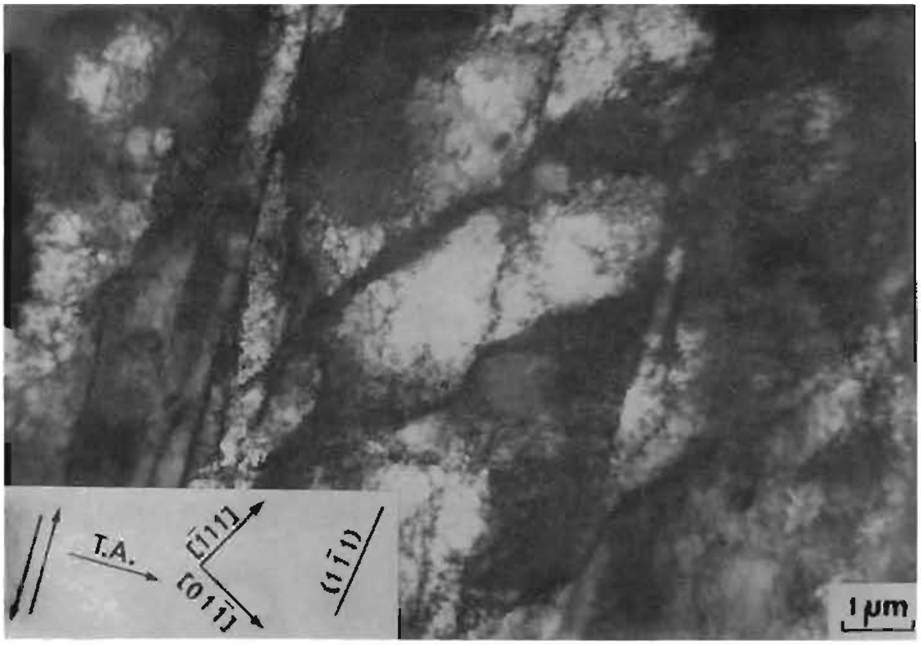


Fig. 15 TEM observation in copper showing microbands in a specimen deformed in simple shear ($\gamma = 0.26$) after prestraining in tension. Double arrows indicate direction of the applied shear stress. (T.A., axis of tension prestrain). Shear offsets produced by microbands at intersections with previous tension structure are clearly seen.

accomodation of the imposed **tension strain state** leads to the activation of other, non-coplanar slip systems. This implies that the microbands are not stable and quickly give rise to lower energy dislocation configurations. In the case of tension-shear experiments, during reloading the grains for which one slip system has the highest Schmid factor are oriented in such a way that the slip plane is normal to the sheet plane and the slip direction is close to the macroscopic shear direction or to the normal direction. When the slip direction is parallel to the shear direction, grain rotation is weak. The compatibility between **an intense activity on one slip plane and the** applied stress state makes the microbands persistent in the grains (**Fig. 15**).

3.4. Thermal Recovery Processes

As previously reported the mechanical behaviour of metals during complex strain paths depends essentially on the amplitude of the strain path change whatever the sequences of loading modes. Moreover, it was established that, for each prestrain amount, we find a limit value of the normalized reloading stress, after which the flow localization occurs immediately after reloading [18]. In agreement with this model, the occurrence of a plastic instability after a prestrain appears to be strongly related to the monotonic behaviour of the material (dependence on the n value of the monotonic tension). This implies, for example, that the instability presents higher sensitivity to the amount of prestraining in low carbon steel sheets than in copper sheets. Namely, in low carbon steel predeformed in rolling up to a strain value of 0.10, an orthogonal change in the strain path leads to the rapid appearance of flow localization, while in copper the value of prestrain which leads to plastic instability is of the order of 0.18 [18, 35]. Recent study performed in low carbon steel sheets have confirmed the validity of the above mentioned model [36]. However, it was also demonstrated that a thermal recovery performed after a prestrain value of 0.10 eliminates the occurrence of plastic instability after reloading. The TEM observations which have supported the microstructural analysis shown that after prestraining, badly defined, elongated cells were the main feature. Moreover, the cell interiors were not free of dislocations and, in some regions of the grains, dislocation tangles are also present. The recovery process performed after prestrain contributes to dislocation annihilation and rearrangement in regions of high dislocation density, transforming the cell walls into subboundaries (**Fig. 16**). The reason behind the non-existence of flow localization after reloading was found in the microstructural evolution during reloading. First it must be emphasized that microbands did not develop in the grains. Just after reloading the dislocation walls created during the prestraining tend to disappear (**Fig. 17**). The dislocation wall thickness increases and the dislocation density decreases in the walls. In some grains, dissolution of the pre-existing walls occurs. The process of the cell wall destruction occurs rapidly and the dislocation substructure becomes unorganised in all the grains. With increasing strain new dislocation walls, typical of the new loading conditions, appear in the grains (**Fig. 18**). This wall development is similar to that which occurs during a uniaxial tension test but with a smaller cell size, in agreement with the larger stress level during reloading. As observed during linear monotonic tension of steel, the new cell structure is mainly formed by one family of fairly straight dislocation walls. These microstructural results have demonstrated that when a thermal recovery is performed after prestraining strong obstacles are broken down in all the grains as the result of latent hardening and Bauschinger effects, and not only in local soft regions of different grains, as generally observed when microbands develop.

Fig. 16 TEM observation in low carbon steel showing dislocation cells in a specimen predeformed in rolling (0.10) and after being recovered at 600 °C in vacuo for 1 h. The grain has a <112> axis normal to the sheet plane. (R.D., rolling direction).

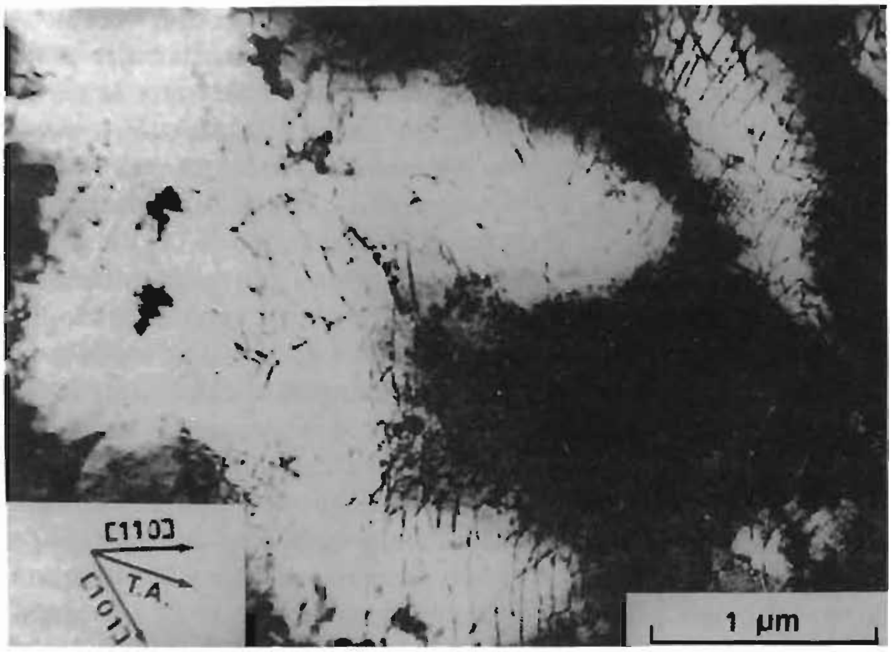


Fig. 17 TEM observation in low carbon steel showing the dislocation substructure developed at the initial stage of the reloading in tension, after prestrain in rolling (0.10) and recovery. (T.A., tension axis).

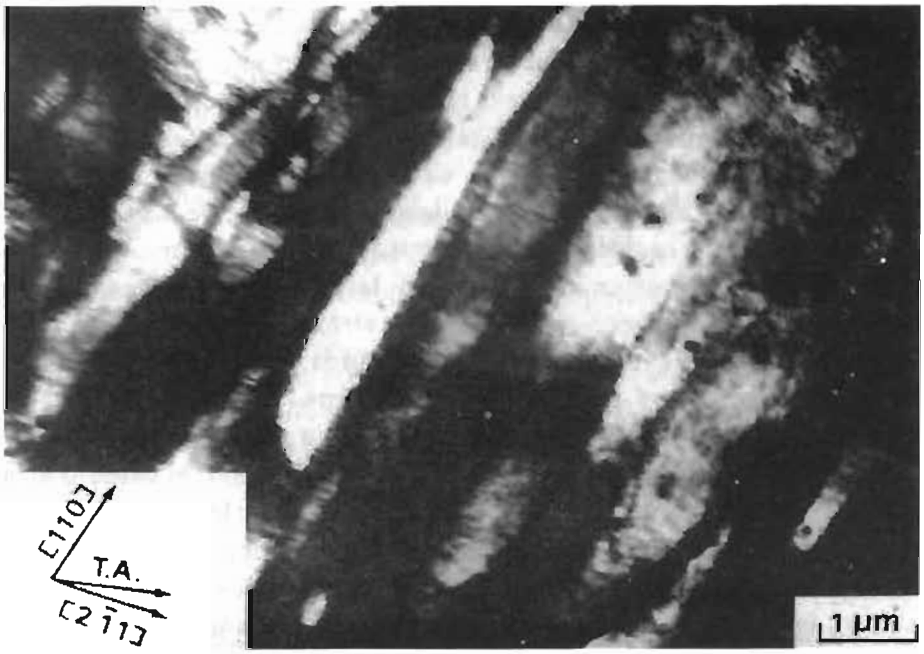


Fig. 18 TEM observation in low carbon steel showing parallel dislocation walls developed during the second deformation in tension, in a specimen previously predeformed in rolling and recovered. The grain has a <113> axis normal to the sheet plane. (T.A., tension axis).

4. CONCLUSIONS

The plastic deformation processes in copper and steel have been analysed on the basis of transmission electron microscopy observations. In the case of polycrystalline copper, the level of the stress-strain curves obtained in tension depends on the angle between the tensile axis and the rolling direction. This behaviour is not related to different evolutions of the dislocation microstructure but arises from different values of the Taylor factor. During tension, the microstructural behaviour of metals strongly depends on the grain size. In large-grained metals, just after the initial stage of deformation, the strengthening mechanism is controlled by the presence of statistically stored dislocations, as in the case of single crystals. In small-grained metals, the strain accommodation complexity is important just to the beginning of the formation of the cell walls; a density of geometrically necessary dislocations is then generated, leading to the increase of the total dislocation density. In rolling, at low strains, in addition to the intergranular accommodation, a high number of operative slip systems required to accommodate the plastic deformation must be considered. Whatever the grain size, the same cell morphology appears in the grains. At moderate strains, the grain size starts to influence the microstructural behaviour. In large grains, there occurs the development of microinstabilities which accommodate the deformation between the differently deformed domains into which the grains are divided. In small grains, microbands with associated shear develop creating shear offsets in the previous dislocation structure.

During complex strain paths, the mechanical behaviour of metals depends on the amplitude in the strain path change. For each amplitude, a typical microstructural event takes place. Namely, when the amplitude in strain path change is not significant, no noticeable change of the previous dislocation structure occurs; for severe changes in strain path the destruction of the dislocation cells created during prestrain takes place immediately after reloading. In large-grained materials, severe changes in the strain path promote the development of microbands. The stability of these microbands depends on grain rotation. During tension, the simple shearing associated with the activity of the microbands causes a rapid grain rotation and the microbands are not stable. During a shear test the slip direction can be parallel to the shear direction and the grain rotation is weak. In this case, the microbands persist in the grains just to higher strain levels. A thermal recovery process performed after the prestrain does not permit the development of microbands after reloading and accelerates the dynamic recovery processes during the initial stages of the subsequent deformation.

RESUMEN

Se realizó un análisis microestructural de los procesos de deformación plástica para el acero y cobre policristalino, basados en observaciones por microscopía electrónica de transmisión.

En el caso partícular del cobre, las curvas de esfuerzo y deformación, presentan diferentes niveles dependiendo del ángulo entre el eje de tensión y la dirección del laminado. Este comportamiento no es resultado de la reacción de flujo anisotrópico sino que está relacionado con los diferenrtes valores del Factor de Taylor. Existe un efecto en el comportamiento microestructural del material que se relaciona con el tamaño del grano y las condiciones de carga. En metales de granos pequeños la tensión, el esfuerzo interno, debido a los procesos de reacomodo entre granos adyacentes, induce al desarrollo de sub-estructuras de dislocaciones, en los metales con tamaño de grano grande, la complejidad del reareglo de la deformación no es importante y la estructura de dislocaciones es similar a la de un cristal simple. En el laminado, a baja deformación, el efecto del tamaño del grano en la evolución microestructural no es relevante. Para cualquier tamaño de grano, la imposición de la condición de deformación conlleva a un alto grado de compresión en cada grano y consecuentemente a la activación de un alto número de sistemas de desligamiento. Con el incremento de la deformación el tamaño del grano comienza a influir en el caracter de la deformación plástica y en microbandas con o sin desarrollo asociado.

Durante la compleja ruta de la deformación, el cambio de amplitud del desarrollo de la deformación es el parámetro controlador más sensitivo del fenómeno de la inestabilidad plástica. Por cada amplitud, el comportamiento mecánico de los metales es controlado por eventos similares tales como los desarrollados en los sistemas de deslizamiento, endurecimiento contínuo y el efecto Bauschinger. El efecto del tamaño del grano en la evolución microestructural de los metales bajo una compleja ruta de deformación es mostrado de una manera, la cual es diferente a las observadas durante las pruebas monotónicas. Dependiendo del tamaño del grano, los fuertes cambios en el camino de la deformación es un factor esencial para la persistencia de las microinestabilidades sólo a altos niveles de deformación. Por el contrario, cuando los granos son rapidamente rotados el deslizamiento

satura en microbandas. Una recuperación técnica efectuada después de predeformación no permite el desarrollo de microbandas después de ser recargado y acelera el proceso de recuperación dinámica durante el estado inicial y en subsecuente información.

REFERENCES

1. Swift H.W. (1952) Plastic Instability Under Plane Stress. J. Mech. Phys. Solids. **1**: 1-18.

2. Hill R. (1952) On Discontinuous Plastic States With Special Reference to Localized Necking in Thin Sheets. J. Mech. Phys. Solids. **1**: 19-30.

3. Marciniak Z. and Kuczynski K. (1967) Limit Strains in the Processes of Strench-Forming Sheet Metal. Int. J. Mech. Sci. **9**: 609-617.

4. Barata da Rocha A., Barlat F. and Jalinier J.M. (1984) Prediction of the Forming Limit Diagrams of Anisotropic Sheets in Linear and Non-Linear Loading. Mater. Sci. Eng., **68**: 151-164.

5. Gracio J.J., Fernandes J.V. and Barata da Rocha A. (1987) Theoretical Prediction of the Limit Curves for Simulation of Plastic Instability. In Computational Methods for Predicting Material Processing Defects (ed. M. Predeleanu) Elsevier Science Publishers. pp. 161-171.

6. Kawasaki Y. (1974) Cell Structures in Deformed Copper Single Crystals. J. Phys. Soc. Jpn. **36**: 142-148.

7. Kawasaki Y. (1979) Correspondence Between Layered Cell Structures and Slip Lines in Deformed Copper Single Crystals. Jpn. J. Appl. Phys. **18**: 1429-1438.

8. Kawasaki Y. and Takeuchi T. (1980) Cell Structures in Copper Single Crystals Deformed in [001] and [111] axes. Scripta. Metall. **14**: 183-188.

9. Kocks U.F. (1970) The Relation Between Polycrystal Deformation and Single-Crystal Deformation. Metall. Trans. **1**: 1121-1143.

10. Ashby M.F. (1970) The Deformation of Plastically Non-Homogeneous Materials. Philos. Mag., **21**: 399-424.

11. Thompson A.W., Baskes M.I. and Flanagan W.F. (1973) The Dependence of Polycrystal Work Hardening on Grain Size. Acta Metall. **21**: 1017-1028.

12. Hansen N. and Ralph B. (1982) The Strain and Grain Size Dependence of the Flow Stress of Copper. Acta Metall. **30**: 411-417.

13. Hansen N. (1990) Cold Deformation Microstructures. Mater. Sci. Techn. **6**: 1039-1047.

14. Wagoner R.H. and Laukonis J.V. (1983) Plastic Behavior of Aluminum-Killed Steel Following Plane-Strain Deformation. Metall. Trans. **14A**: 1487-1495.

15. Raphanel J.L., Rauch E.F., Shen E.L. and Schmitt J.H. (1987) Shear of Prestrained Steel Specimens. Scripta. Metall. **21**: 1087-1090.

16. Zandrahimi M., Platias S., Price D., Barret D., Bate, P.S., Roberts W.T. and Wilson D.V. (1989) Effects of Changes in Strain Path on Work Hardening in Cubic Metals. Metall. Trans. A **20**: 2471-2482.

17. Rauch E.F. and Schmitt J.H. (1989) Dislocation Substructures in Mild Steel Deformed in Simple Shear. Mater. Sci. Eng. **A113**: 441-448.

18. Vieira M.F., Schmitt J.H., Gracio J.J. and Fernandes J.V. (1990) The Effect of Strain path Change on the Mechanical Behaviour of Copper Sheets. J. Mater. Process. Technol. **24**: 313-322.

19. Schmitt J.H., Fernandes J.V., Gracio J.J. and Vieira M.F. (1991) Plastic Behaviour of Copper Sheets During Sequential Tension Tests. Mater. Sci. Eng. **A147**: 143-154.

20. Franciosi P. and Zaoui A. (1982) Multislip in

F.C.C. Crystals: A Theoretical Approach Compared With Experimental Data. Acta Metall. **30**: 1627-1641.

21. Franciosi P. and Zaoui A. (1982) Multislip Tests on Copper Crystals: A Junctions Hardening Effect. Acta Metall. **30**: 2141-2151.

22. Franciosi P. (1985) The Concepts of latent Hardening and Strain Hardening in Metallic Single Crystals. Acta Metall. **33**: 1601-1612.

23. Tome C., Canova G.R., Kocks U.F., Christodoulou N. and Jonas J.J. (1984) The Relation Between Macroscopic and Microscopic Strain Hardening in F.C.C. Polycrystals. Acta Metall. **32**: 1637-1653.

24. Gracio J.J., Fernandes J.V. and Schmitt J.H. (1989) Effect of Grain Size on Substructural Evolution and Plastic Behaviour of Copper. Mater. Sci. Eng. **A118**: 97-105.

25. Gracio J.J. (1994) The Double Effect of Grain Size on the Work Hardening Behaviour of Polycrystalline Copper. Scripta. Metall. et Materialia **31**: 487-489.

26. Tabata T., Tagaki K. and Fujita H. (1975) The Effect of Grain Size and Deformation Sub-Structure on Mechanical Properties of Polycrystalline Copper and Cu-Al Alloys. Trans. Jpn. Inst. Met. **16**: 569-579.

27. Fernandes J.V., Gracio J.J. and Schmitt J.H. (1993) Grain Size Effect on the Microstructural Evolution of Copper Deformed in Rolling-Tension. In Large Plastic Deformations: Fundamental Aspects and Applications to Metal Forming (eds. Teodosiu C., Raphanel J.L. and Sidoroff F.) Balkema. pp.219-228.

28. Gracio J.J., Fernandes J.V. and Vieira M.F. (1993) Effect of Grain Size on the Plastic Deformation of Copper Under Linear and Complex Strain Paths. Proc. 6th SPM Conference. pp. 53-63.

29. Gracio J.J., Ventura L.F. and Marques J.M. (1993) Microstructural Analysis of Copper Deformed in Rolling. Acta of the XXVIII Meeting of the Portuguese Society of Electron Microscopy, p. 44.

30. Gracio J.J. (1994) The Effect of Grain Size on the Microstructural Evolution of Copper Deformed in Rolling. Mater. Sci. Eng.: in press.

31. Schmitt J.H., Aernoudt E. and Baudelet B. (1985)
Yield Loci for Polycrystalline Metals Without texture. Mater. Sci. Eng. **75**: 13-20.

32. Raphanel J.L., Schmitt J.H. and Baudelet B. (1986) Effect of a Prestrain on the Subsequent Yielding of Low Carbon Steel Sheets: Experiments and Simulations. Int. J. Plast. **2**: 1-8.

33. Gracio J.J. (1992) Mechanical and Microstructural Behaviour of Copper Deformed Under Linear and Complex Strain Paths. Doctoral Thesis, University of Coimbra.

34. Fernandes J.V., Gracio J.J., Schmitt J.H. and Rauch E.F. (1993) Development and Persistence of Microbands in Copper Deformed Under Complex Strain Paths. Scripta. Metall. et Materialia **28**: 1335-1340.

35. Korbel A. and Martin P. (1988) Microstructural Events of Macroscopic Strain Localization in Prestrained Tensile Specimens. Acta Metall. **36**: 2575-2586.

36. Gracio J.J. (1994) Interaction Between Thermal Recovery and the Change in Strain Path in Low Carbon Steel. Mater. Sci. Eng. **A174**: 111-117.