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Analysis of damage and its influence on the plastic properties of copper

J. H. SCHMITT, J. M. JALINIER, B. BAUDELET
Laboratoire de Physique et de Technologie des Matériaux, Laboratoire Associé au CNRS No. 155, Université de Metz, France

The damage occurring during plastic deformation is studied in two coppers with similar matrices but containing two different distributions of particles. The damage is quantified by measurement of the relative density change along two different strain paths: uniaxial tension and equibiaxial expansion. This damage leads to different apparent rheological parameters (work-hardening index) for the two materials, the matrices of which are considered to have similar rheological parameters. This apparent mechanical behaviour is characterized by the forming-limit diagram of the materials studied.

1. Introduction

It has been pointed out in the literature that in a metal containing second-phase particles, ductile fracture occurs by initiation, growth and coalescence of voids between the particles and the matrix [1–6] or of cracks inside the particles [7–9]. The evolution of this plastic damage depends on such metallurgical parameters of the material as particle size, particle shape, inclusion distribution, chemical nature of the matrix, grain size, etc. These parameters are very often interdependent. The shape of the voids during plastic deformation, and therefore the void volume per particle, varies according to the strain path [10]. Experimentally, the damage is studied by direct observation using a microscope (optical, SEM or TEM) and measured by a macroscopic method such as the relative density change [10, 11]. A previous paper [10] has shown the influence of damage on the plastic instability of the material. This influence is characterized by the action of internal defects on the position of the forming-limit diagrams (FLD).

The aim of this paper is to evaluate both the influence of the particle distribution on the damage during deformation for different strain paths and the influence of the damage on the plastic behaviour of the material, everything else being constant. This is obtained for two materials with the same matrix but with different particle volume fractions.

2. Choice of materials

Two industrial coppers (99.9%) have been chosen. The first is a tough pitch copper (denoted Cu/a1) and the second an oxygen-free copper (denoted Cu/c1) (see Table I). The two metals are initially cold-rolled (half-hard) to a 2 mm thick sheet. The particle configuration is very different in the two

TABLE I Chemical compositon, rheological parameters (determined by tensile tests) and inclusion parameters of the two coppers. [fv: volume fraction of particles, N_S: particle density per unit surface area in the plane of the sheet, $\bar{D}$: mean inclusion diameter, n: work-hardening exponent (tensile test along rolling direction), $\bar{r}$ = mean anisotropy coefficient ($\bar{r} = \frac{1}{4}(r_0 + 2r_{45} + r_{90})$, with r_α the anisotropy coefficient in a tensile test direction making an angle α with the rolling direction)]

Sample	Chemical composition (wt%)			Inclusion parameters			Rheology	
	Cu	O	P	fv	N_S (mm^{-2})	$\bar{D}$ (μm)	n	$\bar{r}$
Cu/a1	99.9	0.1	–	3.7×10^{-3}	6×10^2	2.5	0.151	0.68
Cu/c1	99.92	–	Trace	2.5×10^{-3}	3×10^2	3.0	0.164	0.73

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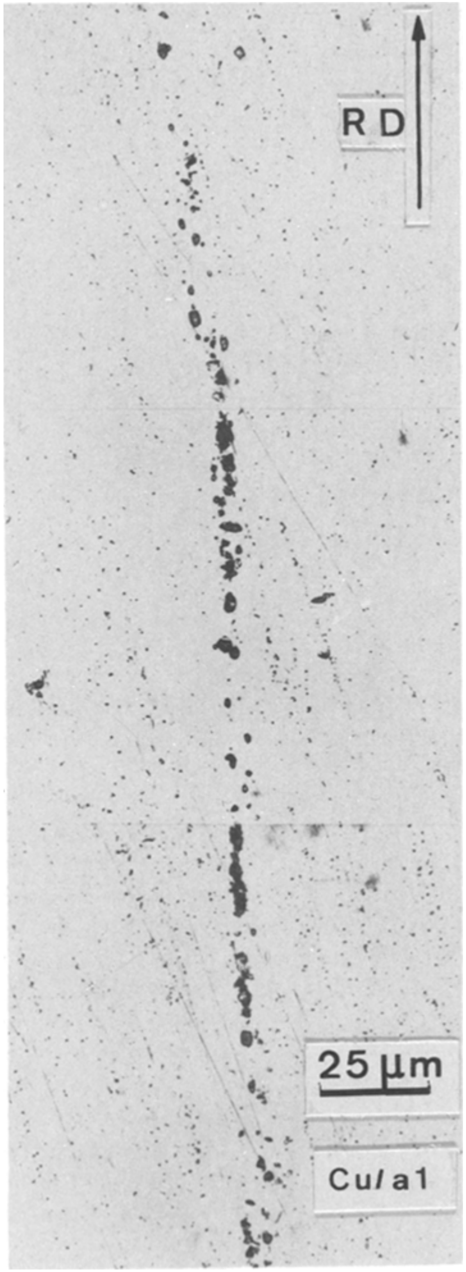

Figure 1 Oxide particle alignment in the Cu/a1 copper in the plane of the sheet (RD = rolling direction).

coppers. The other metallurgical parameters are similar, but the particle distribution leads to a different evolution of these parameters in the two materials (such as the configuration of dislocations after cold rolling). It can be assumed, however, considering the similar constitution of the two metals, that the main parameter in the case studied is the particle configuration. The grain size and the

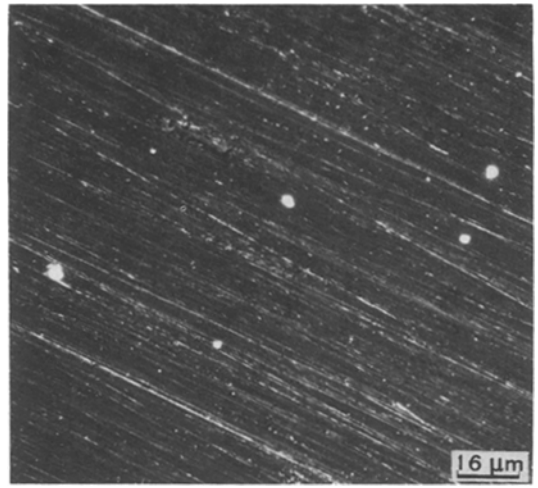

Figure 2 Particle distribution in the Cu/c1 copper in the plane of the sheet (in polarized light).

anisotropy coefficients are found to be similar, in agreement with the above assumption. It will be seen later that this hypothesis is reasonable.

Prior to plastic deformation, metallurgical observations are made on polished surfaces by optical microscopy and scanning electron microscopy (SEM). The two surfaces studied are the plane of rolling (the plane of the sheet) and the plane containing the rolling direction and the thickness direction. After mechanical polishing with a 3 μm diamond paste, the samples are electrolytically etched by a PRESI D-31 solution at 13 V for 1 sec. Low magnification shows a difference between the distribution of the particles in the two coppers. In the Cu/a1 copper, oxide inclusions are lined up along the rolling direction (Fig. 1). The length of the alignments is about 0.2 mm and the distance between two lines of particles is about 0.1 mm. The Cu/c1 copper contains a more random distribution of particles (mainly Cu_3P), as can be observed in Fig. 2. Moreover, the two metals (Table I) have different volume fractions and particle densities.

During cold-rolling deformation, the difference in plasticity between the particles and the matrix [12, 13] (hard particle in a soft matrix) leads to initial damage. This is characterized by either a fragmentation of the particles or cavitation at the interface between the inclusions and the matrix at the head of the inclusions along the rolling direction. Fig. 3 shows this initial damage in the two planes of the sheet under observation, for the Cu/a1 copper.

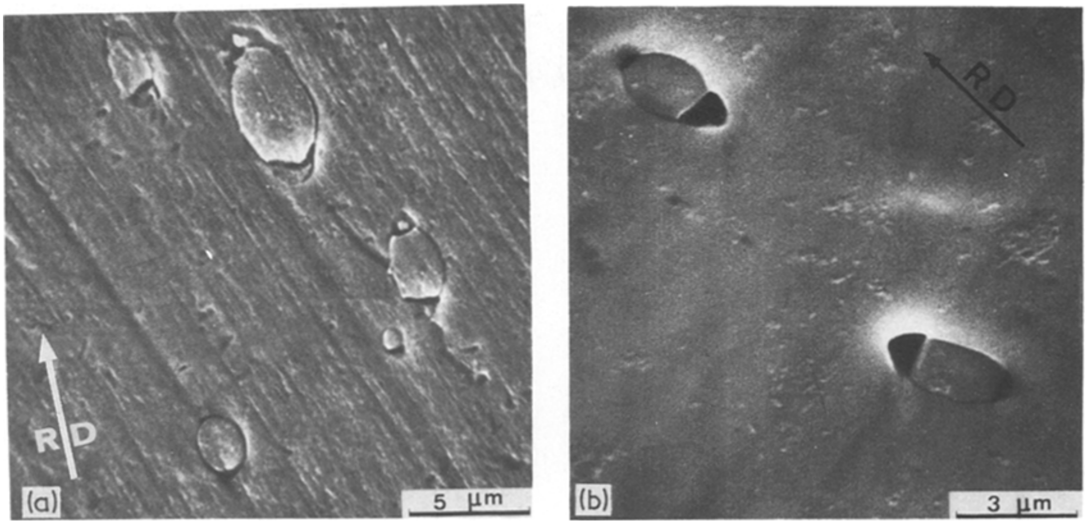

Figure 3 Oxide particles and initial damage in the Cu/al copper before deformation (after cold working). RD = rolling direction. (a) In the plane of the sheet. (b) In the thickness plane of the sheet.

3. Damage

3.1. Metallurgical observations

3.1.1. Experiments

Two principal tests are performed: a tensile test on an ISO 50 sample* at room temperature, using an Instron-type tensile testing machine, and a bulge test using Jovignot's method. These tests allow comparison of the damage along two different strain paths ($\rho = -0.5$ and $+1$ with $\rho = \epsilon_2/\epsilon_1$, where ϵ_1 and ϵ_2 are the two principal strains in the plane of the sheet)†.

After deformation, the principal strains in the plane of the sheet are measured on the samples from a grid of 1 mm diameter circles initially photoprinted on the undeformed sheets. The samples are prepared by the method described in Section 1, and are observed by SEM.

3.1.2. Results on tough pitch copper

Three points arise from the metallurgical observations after the tensile or bulge tests:

(a) At first, the relative density changes during deformation can be estimated by the measurement of void surface per particle for a given strain and along a given strain path (Fig. 4). For instance, after uniaxial tension up to a major principal strain ϵ_1 of 0.48, the void surface area is about 3×10^{-6} mm^2 per particle in the plane of the sheet. Considering the thickness of the particles (about 2×10^{-3} mm) and their density (see Table I), the relative density changes for the initial state can be estimated at -1.5×10^{-3} for a tensile test along the rolling direction up to $\epsilon_1 = 0.48$. This value will be compared later with that obtained by another method described in Section 3.2.

(b) For a given copper the influence of the strain path on the shape of the void around the particles is illustrated in Fig. 4. After uniaxial tension along the rolling direction, the cavities are very elongated and resemble cracks in the plane of the sheet (for example, at a local strain $\epsilon_1 = 0.80$, the ratio L/W is equal to 3.6, with L and W the length and the width of the void respectively). Near the necking strain, there is coalescence between neighbouring voids along an alignment of particles, as shown in Fig. 4a. On the other hand, after equibiaxial tension the voids are more circular in the plane of the sheet ($L/W = 1.5$ at $\epsilon_1 = \epsilon_2 = 0.42$). Moreover, this difference in the shape of the voids arising from two strain paths is amplified by a difference in the volume of the voids. In fact, the volume of the cavities per particle is found to be greater as the deformation becomes more equibiaxial. These results are qualitatively in agreement with the Rice and Tracey [14] model of cavity growth.

(c) This difference in void shape has a great importance when the particle distribution is considered. If the particles are randomly distributed and rather far from one another, the evolution of

*Gauge length of 50 mm and width of 12.5 mm.
†Neglecting the anisotropy.

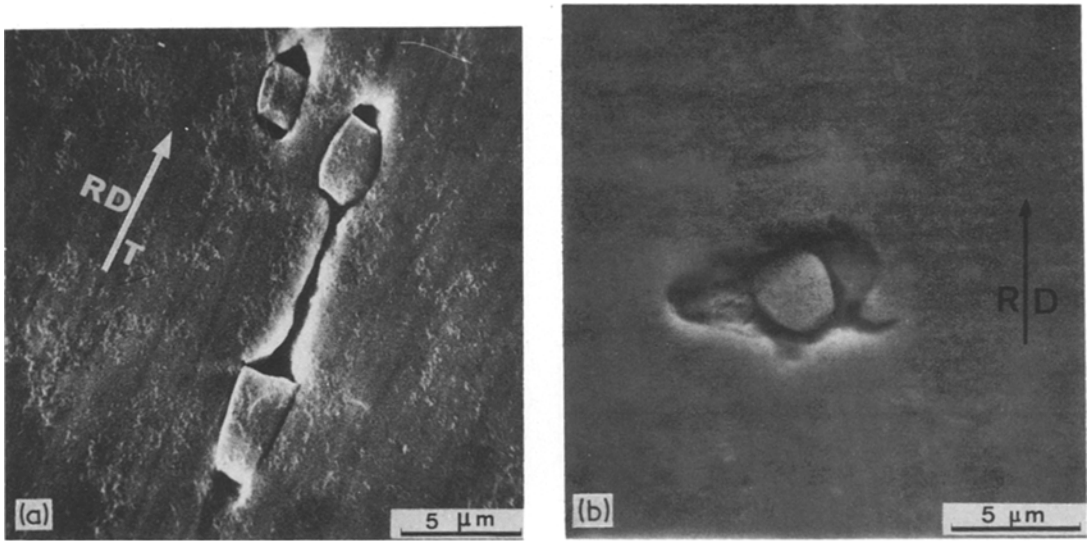

Figure 4 SEM observations of void formed around oxide particles in the Cu/al copper in the plane of the sheet. (a) After a tensile test in the rolling direction (RD) up to a local major principal strain (ϵ_1) of 0.80. Note the coalescence between two voids along the tensile direction (denoted by T). (b) After an equibiaxial stretching up to $\epsilon_1 = \epsilon_2 = 0.42$.

the cavities will have little influence on each other. On the other hand, if the particles are lined up in a given direction, as it is the case in the Cu/a1 copper, the evolution of the shape and the volume of the voids has a great influence on the strains at which coalescence and fracture occur. For a tensile test carried out along the rolling direction (i.e. along the alignment of the inclusions), coalescence will be unidirectional (see Fig. 4a). This coalescence appears close to the onset of plastic instability (i.e. for $\epsilon_1 = 0.37$ in the uniaxial tension test). As the cavities do not extend in the cross directions, even after coalescence, the material can undergo important post-necking deformation before fracture occurs. This explains why, for a necking strain of 0.37, fracture of the sample occurs for a local strain ϵ_1 of 1.45. For the case of biaxial tension deformation, cavities grow in two directions. Coalescence takes place along the alignments of particles and large voids can extend from one alignment to another. This leads to a small difference between the necking strain and the fracture strain in equibiaxial tension (necking strain 0.52 and local fracture strain 0.99); in all cases, the strain path between necking and fracture is closed to plane strain ($\rho = 0$).

3.2. Measurements of relative density changes

Ratcliffe's method [15] is used to estimate the plastic damage due to the formation of voids around the particle, by measuring the relative density change $\Delta d/d$ before and after deformation [10]. This method is accurate (accuracy less than $\pm 5 \times 10^{-5}$) and allows the measurement of many samples. However, since the measured quantity is the sum of various effects, some parameters are masked in such a method; for instance, the shape of the voids, which depends on the strain path. Some differences between the relative density changes along different strain paths and for different metals can, however, be understood through the metallurgical observations described above.

Fig. 5 shows the relative density changes of the two coppers for the two strain paths studied ($\rho = -0.5$ in the rolling direction and $\rho = 1$). For a given metal, the damage is greater for the equibiaxial deformation than for the uniaxial one at a given strain ϵ_1. Moreover, the initial behaviour of the two curves is different: a delay in the increase of the damage is observed for the uniaxial test. Along the same strain path the damage created in the Cu/a1 copper is greater than in the Cu/c1. These differences can be explained by the metallurgical observations carried out earlier.

It has been shown for the Cu/a1 copper that the difference between the evolution of the voids along two strain paths is such that the void volume at a given equivalent strain is greater in a biaxial than in a uniaxial test. This difference explains the density change measurements and is in agree-

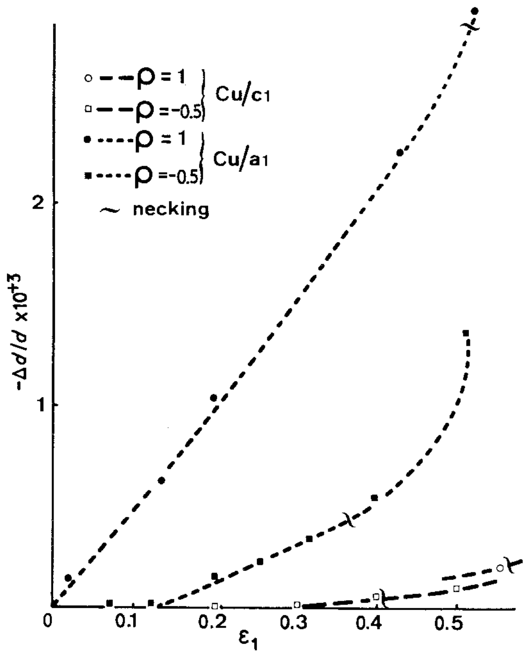

Figure 5 Relative density change along the strain paths ρ equal to 1 and $-\frac{1}{2}$ for the two coppers.

ment with experimental results obtained previously on steels [10]. The measured values for the threshold of the two curves for the Cu/a1 copper can be explained by the initial damage. Biaxial stresses tend to open the initial voids in the first stages of the deformation, thus involving a density change as early as the beginning of the deformation. On the other hand, uniaxial stresses in the rolling direction tend to close the voids in the first stage of the deformation.

During tensile tests along the rolling direction, differences between the behaviour of the two coppers can be found. In the Cu/a1 copper, where stress concentration around the particles is important due to the proximity of the aligned particles [9], the growth of the cavities is faster than in the Cu/c1 copper which has a larger mean particle spacing. This explains the threshold strains on the two curves of relative density change in the tensile tests. Furthermore, the difference between the particle densities in the two coppers leads to the existence of two types of damage evolution which depend on the deformation. Thus the ratio of the density of particles in the two coppers is similar to the ratio of the two slopes of the curves of density change during the tensile test before the coalescence of the voids (N_v in Cu/c1 is approximately half that in Cu/a1). The combination of these two metallurgical parameters (the particle density and the particle distribution) leads to a great divergence in the values of $\Delta d/d$. For instance, in Fig. 5 one finds that, at an equivalent strain of 0.4, $\Delta d/d = -5.5 \times 10^{-4}$ for the Cu/a1 copper and $\Delta d/d = -0.5 \times 10^{-4}$ for the Cu/c1 copper. These differences between tough pitch copper and OFHC copper are in agreement with the experimental results of Rogers [11].

Thus there is a good correlation between the voids observed near the particles and the damage measured by density changes. The value of the volume fraction of voids calculated above for the Cu/a1 copper $\Delta d/d = -1.5 \times 10^{-3}$ at $\epsilon_1 = 0.48$ during tensile testing) is in good agreement with the value of the curve of Fig. 5. Consequently, it seems justified to consider the inclusion parameters as the main ones controlling the plastic damage, more especially as it has been verified that there is no significant influence of the dislocation density on the density changes (a deformed specimen whether subsequently annealed or not gives the same value for the relative density change).

4. Influence of the damage on the plastic behaviour of the material

4.1. Influence on the rheological parameters

When a test is performed to assess the rheology of a material, only the apparent values of the strains and stresses are measured. This leads to an apparent rheology which is different from that of the matrix (the real material without any damage) because of the damage [16, 17]. For example, for the two coppers studied the work-hardening index has been found to be equal to 0.164 for the Cu/c1 copper and 0.151 for the Cu/a1 copper during a tensile test in the rolling direction. The mean size of dislocation cells is small compared to the mean distance between particles, except along a particle alignment in the Cu/a1 copper. This would lead to a similar behaviour of the dislocations (and hence a similar microstructure) in the two materials, except along an alignment of particles in the Cu/a1 copper. In this case, the mean distance between particles is of the order of the cell size and, so, contributes to more hardening. It then appears that the work-hardening index is similar for the two matrices, but slightly greater for the Cu/a1 copper. In reality, the work-hardening index has been found to be greater for the Cu/c1 copper. So the difference in the apparent behaviour is qualitatively explained

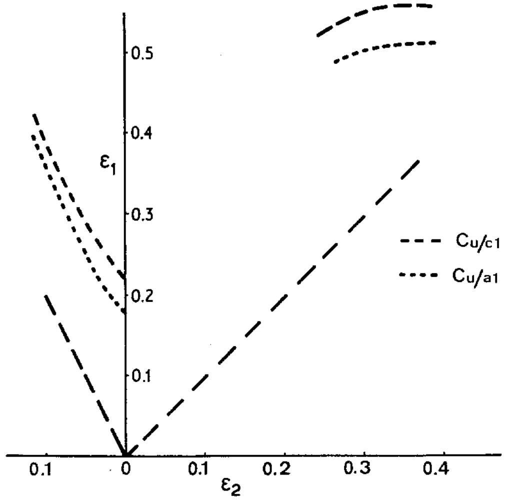

Figure 6 Forming-limit diagrams at necking of the Cu/a1 and Cu/c1 coppers.

by the difference between the amount of damage created. Indeed, it has been pointed out that the relative density changes are greater for Cu/a1 copper than for Cu/c1 during a uniaxial test.

The apparent work-hardening index is not an accurate parameter for use in plastic instability calculations because the instability is controlled by the mechanical behaviour of the matrix, i.e. by the matrix work-hardening index and also by the evolution of the geometrical defect due to the damage. This is not equivalent to a calculation using the apparent parameter which masks the damage phenomena. For example, according to the flow theory of plasticity, no localization can be predicted in the expansion range by taking into account only the apparent rheology and not the defects.

4.2. Influence on the forming-limit diagram

As pointed out previously, the two coppers have similar matrix work-hardening behaviour. Hence, the principal cause for the differences in the forming-limit diagrams (FLD) at necking is the evolution of the damage created in each material. According to measurements of the relative density change, the FLD of the Cu/cl copper must be at higher strains than the FLD of the Cu/al copper (Fig. 6) [18].

Furthermore, the difference in the values of relative density change between the two coppers is greater for a biaxial tension test than for a uniaxial test. In agreement with this result, the gap in major principal strain ϵ_1 between the two FLDs is about 0.05 in equibiaxial stretching and 0.025 in pure tension.

5. Damage model

In this analysis the damage is characterized by a scalar parameter; the relative density change. This parameter does not describe the shape and the distribution of the voids at each step of the straining. For example, the initial distribution of the particles after rolling shows alignment along the rolling direction (specially in Cu/al). This anisotropy of the damage leads to different behaviour in the rolling and transverse directions; for example, the fracture strains are 1.45 and 0.89 respectively (for Cu/al). However, the relative density changes are very similar in the two cases: $\Delta d/d$ is equal to -6.8×10^{-4} in the rolling-direction tensile test and -8.7×10^{-4} in the transverse-direction tensile test, both measured at the fracture strain. The only way to characterize the anisotropy of the distribution of the damage and the anisotropy of the shape of the voids is to represent the damage by a tensor. The main difficulty in such an approach is to find the proof tests to assess every term of this tensor. Furthermore, a tensor is generally very unfavourable to plastic instability calculations. Measurements of relative density change are useful variables for quantifying, the damage. They must be associated with metallurgical observations, however, in order to obtain a physical model of the damage suitable for calculations.

6. Conclusions

In this analysis it has been demonstrated that the rate of damage has an influence on the plastic behaviour of the materials. Two kinds of copper with the same matrix but with different volume fractions of particles have shown the following properties:

(a) the rate of damage measured by the relative density changes is found to be correlated to the particle content;

(b) the rate of damage is greatly affected by the strain path, in agreement with metallurgical observations of the damage mechanisms; the amount of damage is greater for an equibiaxial than for a tensile deformation;

(c) the apparent rheological parameter (work-hardening index) is affected by the damage and it is lower the greater the damage;

(d) the difference between the two FLDs of these two materials can be a measure of the influence of the geometrical defect created by the damage.

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