secondary phase in the material investigated in this work was found to vary from 10 to 50 nm (Fig. 1).

Moreover, transmission electron microscope (TEM) observations revealed that each grain was surrounded by a dark band (Fig. 1), The dark area persisted when the objective aperture was removed. This indicated that the band is caused by absorption, which can be explained by the presence of the heavy element Bi in this region (here called the diffusion layer).

EDAX analysis confirms the presence of Bi (Fig. 2) in this region. After correction of the absorption coefficients, as discussed by Duncumb and Reed [5], the amount of Bi in the diffusion layer is estimated to be less than 2 at%. The width of the diffusion layer is found to vary greatly from boundary to boundary and probably depends on the crystallographic orientation of the grains. Whether, apart from this diffusion layer, another oxidation layer exists, or whether both layers are identical is not clear from the present data. However, it is obvious that the boundary layer model needs some modification, at least in the case of the boundary layer material investigated here.

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On the existence of initial damage in sheet metal

It is well established that ductile fracture of a material containing second-phase particles occurs by nucleation, growth and coalescence of voids generally associated with inclusions. Some theoretical models try to account for growth of cavities [1, 2], but the parameters controlling the nucleation of cavities during plastic deformation at low temperature ($T \leqslant 0.2\,T_\mathrm{f}$, T_f corresponding to the melting point) are still being discussed [3].

It is possible that a certain amount of plastic damage due to hot and cold rolling or ingot porosity exists before any plastic deformation of a sheet metal occurs. It could modify the evolution of the damage, suppressing the initial gap due to nucleation, but the existence of such initial damage is still being debated.

In this paper, some metallographic evidence of initial damage in tough pitch copper, 3003 aluminium alloy, aluminium-killed steel and interstitial free steel are noted (these materials are described by Schmitt *et al.* [4], Jalinier *et al.* [5, 6]). The influence of initial damage on the formability of the materials is studied taking into account the kinetics of the growth of the cavities in the metal.

Different experimental methods can be used to determine the amount of damage in a material. Some of these methods were discussed in a previous paper [5]. Most methods (e.g. relative density change measurements, hydrogen diffusion, positron annihilation) only provide information on the relative damage between a plastically deformed sample and an as-received one. An absolute method of determining initial damage is metallographic observation by optical or electron microscopy. The damage existing in the reference sample can be estimated by the measurement of the surface

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voids associated with particles in the plane of observation. Nevertheless, many areas must be analysed to maintain accuracy because of the heterogeneity of the sheet and of the smallness of the observed area (e.g. $20\,\mu m \times 20\,\mu m$ at a magnification of 4.000 using a scanning electron microscope).

The literature gives some proof of plastic damage due to both hot rolling [8–10] and cold rolling [7, 9, 11]. According to the shape of the particle and to the ratio of ductility of the particle to the matrix; damage occurs either by decohesion at the interface between the particle and the matrix or by fracture of the particle. When the material contains brittle particles and has a large surface energy, the damage due to cold rolling occurs by cracking inside the particles. The cracks are very often orthogonal to the rolling direction. This is observed in 3003 aluminium alloy with hard $Al_6(Mn, Fe)$ particles (Fig. 1) and in interstitial-free steel for some titanium particles (Fig. 2). The cracks created are too narrow (about $0.1\,\mu m$ in the $Al_6(Mn, Fe)$ particles) to allow the flow of the matrix into them. In Fig. 3, it is clear that this phenomenon can exist but, even in large cracks the flow is difficult and damage can still remain [9].

In numerous cases the initial damage occurs by decohesion at the particle–matrix interface.

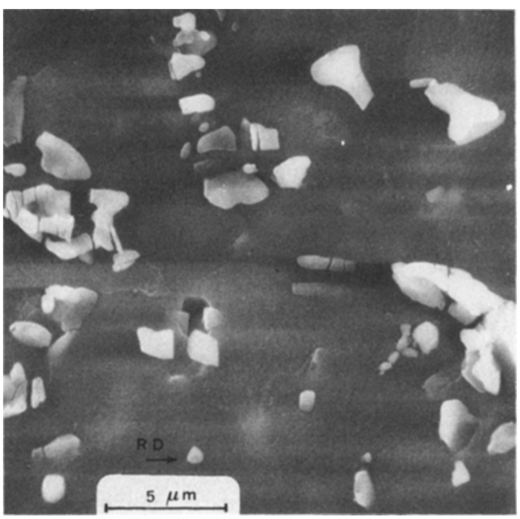

Figure 1 Scanning electron microscopy (SEM) observation of cracked and uncracked particles on the surface of 3003 aluminium alloy. RD represents the rolling direction.

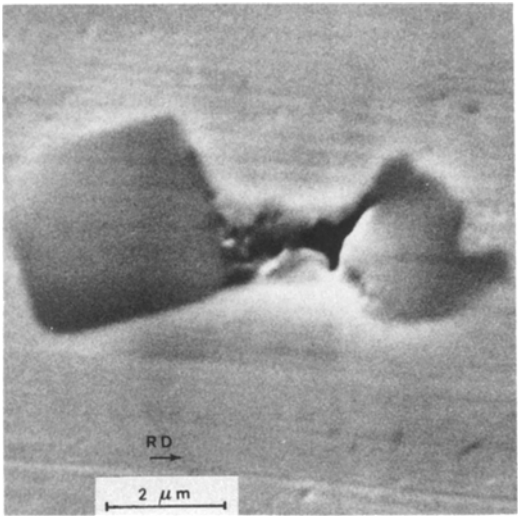

Figure 2 SEM observation of a void formed by the crack of a titanium particle on the surface of an interstitial free steel.

Cavities are created at some points of stress concentration, generally along the rolling direction as shown in Fig. 4 for tough pitch copper and Fig. 5 for aluminium-killed steel. The shape of the cavities obtained in the tough pitch copper sheets are similar to those obtained after a compression test [12].

During cold rolling, the pressure due to the rollings is not homogeneous along the sheet. Relative density change measurements along the rolling and transverse directions of a Sollac steel shows this fact (Fig. 6). The edges of the

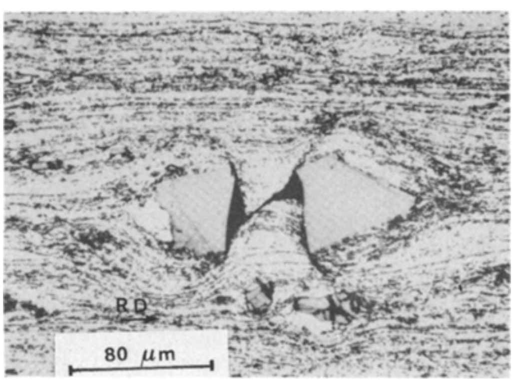

Figure 3 Optical microscopy observation of a large primary crystal broken in rolling leaving a gap within fragments (Mondolfo [9]).

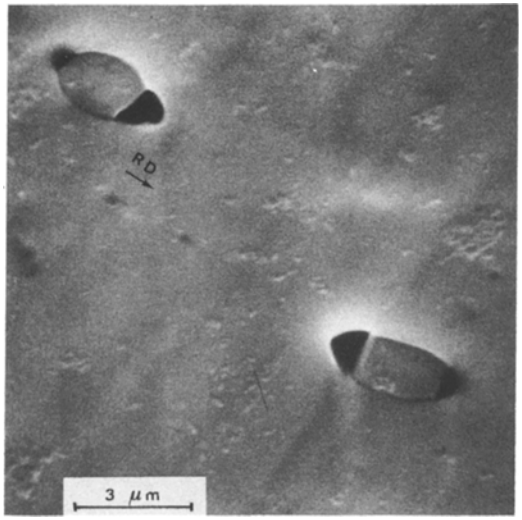

Figure 4 SEM observation of cracks created at the end of particles on the surface of a tough pitch copper.

sheet are more damaged compared with the middle areas. Along the rolling direction, periodic fluctuations appear which could probably correspond to an eccentricity of the rolls. The density change values are significant, the accuracy of the method being of $\pm 5 \times 10^{-5}$ [5]. This result is in agreement with those obtained on the same steel by hydrogen absorption [13].

The initial damage influences the evolution of

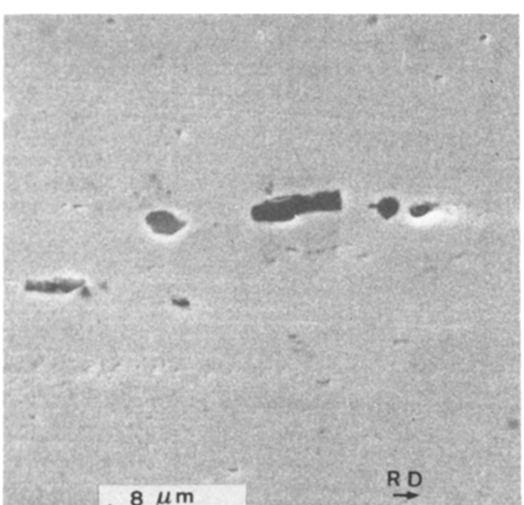

Figure 5 SEM observation of decohesion around a hard particle of alumina in the thickness of an aluminium-killed steel.

the damage during plastic deformation in different ways. If the particle is hard compared to the matrix, then stress concentration occurred. This stress concentration may or may not be sufficient to break the particle during a test. Therefore, in some particular cases, only a small amount of nucleation can be present after cold rolling. So, the damage increases especially at the particles which were broken before the plastic test. Such a phenomenon is observed in the 3003 aluminium alloy studied in which the percentage of broken particles increases from 11% ($\pm$ 3%) in the initial state to 17% ($\pm$ 3%) at fracture strain both in unaxial and biaxial tension tests [5]. A similar point is also described by Souza Nobrega *et al.* [7] on an aluminium-killed steel. The initial number of defects is $1.0 \pm 0.2 \times 10^3\,\mathrm{mm}^{-2}$, the number of defects after tensile deformation to an equivalent stress of 0.23 is $1.1 \pm 0.2 \times 10^3\,\mathrm{mm}^{-2}$ and $1.0 \pm 0.1 \times 10^3\,\mathrm{mm}^{-2}$ after an equibiaxial deformation to an equivalent stress of 0.36.

It has been shown that initial damage can create some elongated voids at the end of the particles. In some cases, this initial shape introduces an anisotropy of the damage. For instance, in tough pitch copper coalescence of the voids during the tensile test along the rolling direction occurs along the direction of the particles and creates long oriented defects [4]. In all cases the initial shape influences the evolution of the shape of the voids. Rice and Tracey [2] assume that the voids created after nucleation are spherical and then predict a very different evolution of aspect ratio according to the biaxiality of the deformation: the cavities are rather circular after equibiaxial tension and quite elongated after uniaxial tension. When the initial voids are not spherical, the differences between the two strain paths is less clear. In fact, the result on a low carbon steel gives an initial aspect ratio (major to minus axis of the void) of 1.4. After a deformation of 0.4 in equibiaxial stretching, this ratio becomes about 2.0, which means that it is slightly more elongated but the mean surface of the void is increased by a factor of about ten.

In order to determine plastic instability in the plane of the sheet a calculation is performed taking a damage function into account. This calculation is described by Jalinier and Baudelet [14]. The damage is represented by the volume

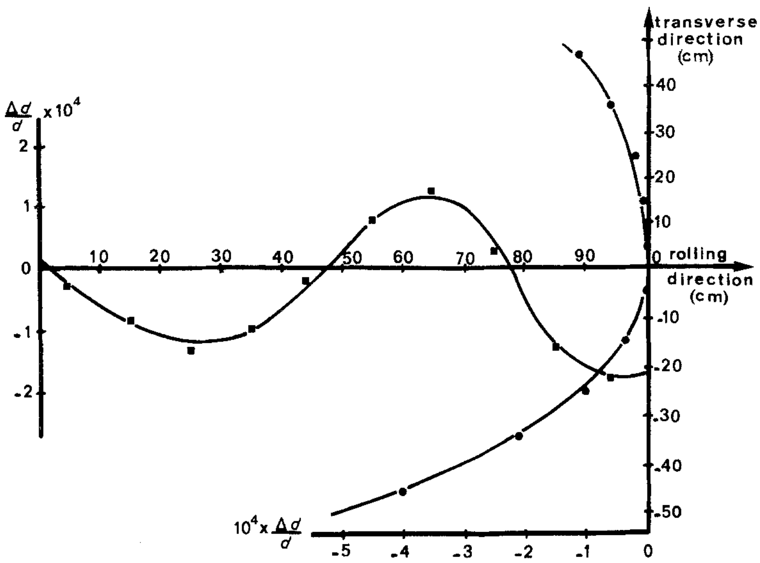

Figure 6 Representation of the relative density changes along the rolling and transverse directions of an aluminium-killed steel. The rolling gives a sheet of 1 m^2.

fraction of the voids. The cavities are assumed to be randomly distributed and to have a similar size. The relative thinning of the sheet represents the defect which increases from the initial value. A statistical calculation gives the evolution of the critical defect with strain. Then a plastic instability calculation on a two slice material (one with damage and one without damage) gives the strain at which localization of the flow occurs. This calculation takes into account the aggravation of the initial defect (according to the model of Marciniak and Kuczynski [15]) and the amount of damage due to plastic growth of the voids. Through this model the relative influence of the initial damage on the limiting strain depends on the kinetics of evolution of the defects. In a low carbon steel the evolution of the defect is a hyperbolic line function of the equivalent strain, i.e. a rapid increase of the defect occurs. In such materials, the influence of the initial damage is small compared to the created damage which is about ten times greater than the initial damage. It explains why in such cases the theory of Marciniak and Kuczynski [15] required an important ($\sim 1\%$) initial geometric defect to fit the experimental data. When the evolution of the defect is smoother, the initial defect takes on some importance. For instance in a 3003 aluminium alloy the defect increases more linearly with the equivalent strain [16]. In order to compare the contribution of the initial damage to the limiting strain, the calculation gives a forming limit of 0.36 in tension taking account only of initial damage and 0.77 taking account of only the growth of the damage (initial defect $f = 0$) for an experimental limiting strain of 0.33 in tension.

Finally, it has been pointed out that initial damage due to hot and cold rolling can exist in many sheet metals. The evolution of the damage during straining strongly depends on the size, shape and number of initially damaged particles. The calculation to predict plastic instability has to take into account this fact, especially if there is little nucleation during the plastic tests or if the growth of cavities is small during plastic deformation. Consequently the relative methods to quantify plastic damage have to be completed by metallographic observations in order to determine the amount of initial damage.

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Transverse cracks in glass/epoxy cross-ply laminates impacted by projectiles

Impacted composite laminates may fail with several different failure mechanisms either occurring separately or in some combination. In $[(0°)_5/(90°)_5/(0°)_5]$ glass/epoxy cross-ply laminates impacted by cylindrical projectiles with different impactor nose shapes and lengths, a sequential delamination mechanism is dominant, initiated by a generator strip, of width approximately equal to the impactor diameter, cut from the first lamina by two through-the-thickness cracks parallel to the fibres of the first lamina [1]. Blunt-nosed projectiles produced clearer generator strips than hemispherical-nosed projectiles [2].

In addition, an observable, almost even distribution of fine transverse cracks in the 0° direction was noted on the front and back of the impacted plates in the 0° fibre direction, as shown in Fig. 1 for blunt-nosed projectiles. Some transverse cracks extend the full length of the specimen in the fibre direction, while others are combinations of several cracks. Similar transverse cracks in the middle, 90° orientated lamina can also be found along the fibre direction when the plates are illuminated by a strong back-light.

Similar types of transverse cracks with evenly distributed crack spacings in cross-ply laminates have been observed in different static loading conditions [3–12].

Recently, several authors have investigated in detail these characteristic transverse cracks in uniaxial tensile tests of cross-ply laminates and defined the factors controlling this evenly distributed crack spacing [13, 14]. Garrett and Bailey [13] have found the spacing of transverse cracks to be dependent both on the thickness of the transverse ply and on the applied stress. Generally, the higher the applied stress and the smaller the transverse ply thickness are, the smaller is the average crack spacing. The above discussion on transverse cracking relates mainly to static tensile tests but can be applied directly to transverse cracking in the impacted laminates in a qualitative manner as follows.

In order to do this, the crack spacing on the front and back faces of each lamina of each specimen was measured and the arithmetic mean was then calculated. The mean transverse crack distance (MTCD) was found as a function of impactor velocity and is shown in Fig. 2 for five impactor types. The two marks for each specimen correspond to the MTCD measured on the front and back laminas, respectively. For blunt-nosed impactors (Fig. 2a), the upper points correspond to the

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