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FRACTURE IN TEXTURED THIN BRASS SPECIMENS

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INTRODUCTION

Antisotropic plastic behaviour of thin brass strip is still of interest to the radiator industry. Usually, information about anisotropic behaviour is based on results of tensile tests. Directionality of mechanical properties is determined directly by testing strips cut at different angles to the rolling direction. Plastic anisotropy is induced by the preferred orientation of crystals.

The r-value [1 - 3], defined by the ratio of the width to through - thickness strain is determined from tensile test data. It results from existing texture. The importance of the r-value has already been sufficiently elucidated [2, 4 - 7]. Recently [8, 9], a corrected value of r, r_k, was introduced (see later). It was found constant during tensile deformation of specimens cut from aluminum, copper and α-brass sheets.

Extensive studies of the variation of preferred orientation during tension testing of sheet specimens have shown great changes of crystallite orientation and the formation of characteristic tension textures [10 - 13]. This texture evolution was examined with respect to the dependence of r-value on ε. It is of interest to consider the influence of texture on fracture of thin brass specimens in tension, and this is the subject of the present study.

SHEET SPECIMEN IN TENSION

When sufficiently wide (width at least five times the thickness) strips, cut from sheet at an angle α to the rolling direction (RD) are pulled in tension, generally they form a localized neck. The angle θ of neck inclination to the tensile axis (TA) can be obtained from the equation [14]:

$$a \tan^2\theta + 2b \tan\theta - c = 0 \tag{1}$$

where

$$a = H + (2N - F - G - 4H) \sin^2\alpha\cos^2\alpha$$

$$b = (N - F - 2H) \sin^2\alpha - (N - G - 2H) \cos^2\alpha\sin\alpha\cos\alpha$$

$$c = a + F\sin^2\alpha + G\cos^2\alpha$$

The anisotropy parameters H, F, G and N refer to the state of anisotropy immediately preceding necking.

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In order to solve the quadratic equation (1) we express the parameters H, F, G, N in terms of r_0, r_{45}, r_{90} (r-value in RD, at π/4 and π/2 to RD, respectively) and X (yield stress in RD direction). Assuming, in the case of an anisotropic body [14] that the strain - increment equations are analogous to those of Lévy - Mises, we can obtain:

$$H = \frac{r_0}{X^2(1 + r_0)}$$

$$G = \frac{H}{r_0} \tag{2}$$

$$F = \frac{H}{r_{90}}$$

$$N = \frac{0.5 + r_{45}}{X^2}$$

It is seen that in this approach (equation (2)) the yield stress, X, has no effect on the value of θ derived from equation (1). This angle depends only on the strain rates r_0, r_{45} and r_{90}.

MATERIAL AND METHODS

Experiments were performed on specimens cut from brass strip (Cu-32%Zn) of commercial purity, 0.17 mm thick. Its grain diameter was about 15 μm. Initial texture is represented by a quarter of a {111} pole figure (Figure 1). In Figure 1 ideal orientations are marked which correspond to the main component of texture. This (493)[3$\bar{2}$2]5° (5° = π/36 rad) orientation was obtained from three - dimensional texture analysis [13], while the (252)[$\bar{7}\bar{4}$3] is usually accepted for α-brass recrystallized at low temperature and after heavy cold work.

Special grips, mounted on an Instron universal testing machine were constructed to insure uniaxial loading. Tensile tests on strips of 10 mm width and 20 mm gauge length were performed at a crosshead speed of 0.0167 mm s^{-1}. The measurements of r-value determination were carried out on unloaded specimens with an optical microscope. Thin samples for X-ray diffraction were prepared. Transmission and reflection methods were used for pole figure determination. Mo K_α radiation was used in the X-ray diffraction measurements.

TEXTURE EVOLUTION DURING THE TENSION TEST

X-ray investigations were performed for deformed specimens cut at angles 0, π/12, π/4, π/3, π/2 rad to RD. Particular care was taken to follow the changes in RD and CD (cross direction). These were analyzed in detail [9, 13] by means of a three - dimensional [15, 16] orientation - distribution function (ODF). Figure 2 presents the quarters of {111} pole figures, recalculated from the ODF for specimens stretched to the limit of uniform elongation ε_u (in all cases ε_u was equal to about 0.30). For the sample cut in RD (Figure 2a) the main component of texture is {112}<$\bar{1}\bar{1}$1> and for that cut in CD (Figure 2b) the main component is {110}<1$\bar{1}$1>.

Preferred orientation in the remaining specimens was investigated only by pole figures obtained experimentally. One of these {111} pole figures, namely that of a sample cut at angle π/12 and stretched to ε_u is given in Figure 3. The final main component is described by a [132]<111> ideal orientation. The arrows in Figure 3 correspond to rotation axes 01 and 01a. They show the final movement {111} poles of the main components 1 and 1a (both of the type [493]<322>5° and considered with respect to the reference system at the angle π/12 to RD) from the initial to the end positions. They are derived from slip rotations induced by tension.

Beside the normal tendency of rotation of the TA towards the <111> direction it was observed that the ND direction tended to occupy a {112}, {110} or intermediate, for example [123], position. The orientations with the TA parallel to <100> direction were relatively stable.

EFFECT OF PLASTIC ANISOTROPY

The anisotropy coefficient r was calculated according to the formula [2]:

$$r = \frac{\varepsilon_B}{\varepsilon_H} = \frac{\ln B/B_o}{\ln H/H_o} \tag{3}$$

where ε_B - strain in width direction B

ε_H - strain in through - thickness direction H.

From the linear dependence [8, 9] of the width strain ε_B on ε (= ln l/l , where l and l are the initial and final gauge lengths respectively) and r_k value was derived:

$$r_k = \frac{\beta}{1 - \beta} \tag{4}$$

where β is the slope of the line $\varepsilon_B = b + \beta\varepsilon$, where b is a constant. It is assumed constant during deformation in tension up to the limit of uniform elongation (this was found in the case of specimens cut of aluminum, copper and brass sheets). The r_k values at angles 0, π/12, π/6, π/4, π/3, 5π/12 and π/2 rad to RD are given in Table 1.

From the initial state ODF and using the Bishop - Hill [17] or the Taylor theory of deformation [18] theoretical, r_{th}, values were calculated [19]. These are listed in Table 1. There are some differences which may be attributed to the theory utilized.

Values of r at angles 0, π/4, π/2 rad (Table 1) were used to resolve the quadratic equation (1) in Table 2, these calculated and measured inclination angles θ of local necks which are formed before the onset of the shearing instability are shown.

DISCUSSION AND CONCLUSIONS

Preferred orientation changes considerably during tensile deformation. However, an assumed constant value of r_k (with regard to conditions imposed on anisotropy parameters) enables a comparison of θ values (Table 2).

There are still differences between θ values measured directly and those deduced from theoretical and experimental r-values. This could be explained by the influence of texture evolution and the theory utilized. An effect due to width to thickness ratio B/H may be expected also; e.g. in steel [20] the measured value of the angle θ was 59.8 x π/180 rad for B/H = 36 whereas for B/H = 6.5, it was equal 56.25 x π/180 rad. In our case, B/H is greater than 60 and thus it may possibly have an effect.

Though the texture is well pronounced there is still a large dispersion around the ideal orientation. This would cause an accommodation to the external conditions induced by stress state and in consequence reduce the role of texture. Surface observations give supplementary information. As would be expected, specimens appear more rough as elongation increases. At the limit of uniform elongation or just after diffuse neck creation, parallel striations suddenly occur, inclined at an angle θ to TA (Table 2 - the measured angles). Fracture mostly appeared along one of these lines, although sometimes at an angle different from those listed in Table 2.

The features observed indicate the formation of shear deformation bands. One of these becomes active at the end stage of deformation. The position of the shear band can be derived from macroscopic parameters of plastic anisotropy. This phenomenon was examined metallographically. On the upper edge of the longitudinal cross-section of the specimen cut at 5π/12 rad to RD and deformed up to ε_u small cracks starting at the surface (Figure 4a) may be seen. In the neck and across an active macroscopic deformation band in this specimen (Figure 4b) we can observe initiation of fracture. The characteristic waviness of the edges appears to result from deformation achieved by shear fracture beginning at one of these steps. It seems rather intergranular in origin.

From these observations it can be concluded that texture plays a significant role in the plastic behaviour of thin, fine grained brass strip, as approximately described by the macroscopic anisotropy parameters. Measured θ values are intermediate between those calculated from r_k and r_{th}, since theoretical values depend upon the assumption made, while experimental values were obtained from texture evolution. The process is clearly complex and requires further investigation.

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Table 1 Anisotropy Coefficient r for Cu-32% Zn Brass Strip (α - Angle Between TA and RD)

Angle to RD rad	r_k mean value	r_{th}
0	0.865	1.060
π/12	0.741	1.427
π/6	1.031	1.608
π/4	1.145	1.570
π/3	1.111	1.317
5π/12	0.921	0.917
π/2	0.669	0.478

Table 2 Calculated and Measured Local Neck Inclination θ

Angle to RD α rad	Calculated from r_k		Calculated from r_{th}		Measured
	θ_1^*	θ_2^*	θ_1^*	θ_2^*	θ^*
0	55.74	-55.74	54.35	-54.35	56
π/12	57.27	-54.07	58.90	-49.77	58
π/6	57.67	-53.62	60.38	-49.16	58
π/4	57.33	-54.59	60.03	-51.90	60
π/3	57.02	-56.20	59.65	-55.84	55
5π/12	57.42	-57.24	60.03	-59.08	56
π/2	57.66	-57.66	60.37	-60.37	60

*θ x 0.017453 rad

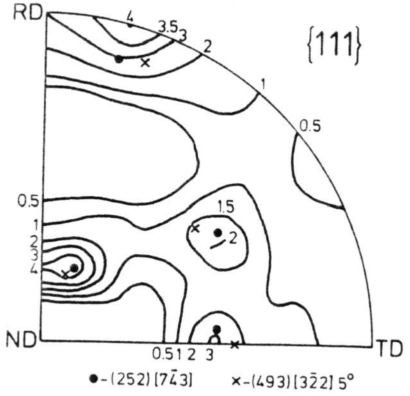
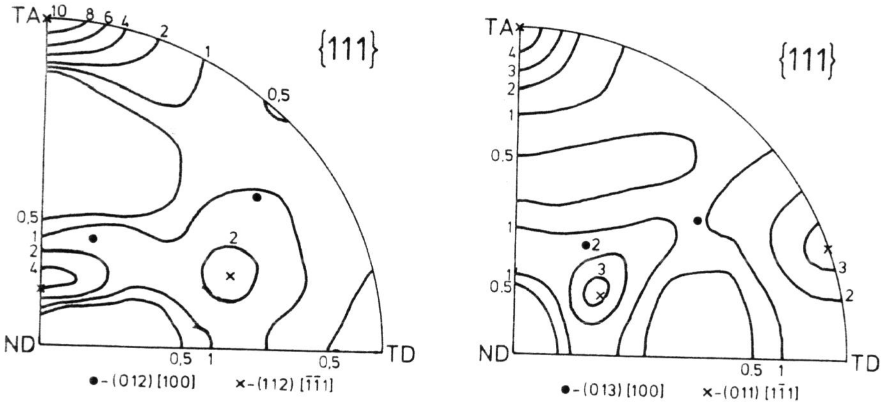


Figure 1 {111} Pole Figure of Cu-32% Zn Brass Strip in Initial State, Recalculated from ODF [13] (RD - Rolling Direction, CD - Cross Direction, ND - Normal Direction)



(a)

(b)

Figure 2 {111} Pole Figures for Strips Pulled in TA (TA - Tensile Axis) up to the Limit of Uniform Elongations

(a) TA Along RD
(b) TA Along CD

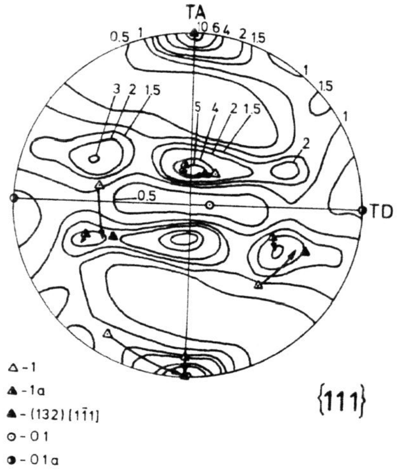
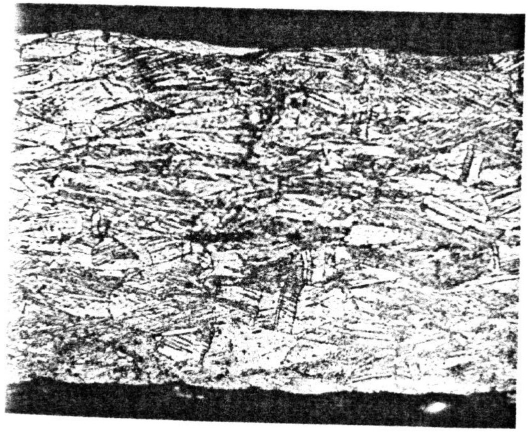
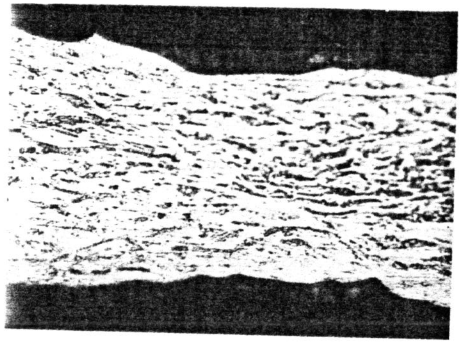


Figure 3 {111} Pole Figure of Specimen at Π/12 Rad to RD and Deformed up to the Limit of Uniform Elongation



(a)



(b)

Figure 4 Microstructures in the Longitudinal Cross-Section of a Specimen Cut at 5Π/12 Rad to RD

(a) Uniform Elongation, 500 X
(b) Neck, 400 X