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THE INFLUENCE OF THE TEXTURE ON YIELD STRENGTH AND STRAIN HARDENING OF HIGH PURITY ALUMINUM FOILS

WPŁYW TEKSTURY NA GRANICĘ PLASTYCZNOŚCI I UMOCNIENIE ODKSZTAŁCENIOWE FOLII ALUMINIOWYCH O WYSOKIEJ CZYSTOŚCI

It is well known that yield strength and strain hardening of metals strongly depend on the grain size and on the purity of samples. However, the crystallographic texture also plays an important role in terms of describing the mechanical properties. Usually it is difficult to distinguish between all these contributions to the material behavior. If one wants to measure the effect of texture alone without being disturbed by different kind of effects, it is necessary just to change the texture and keep everything else constant. We here report tensile tests of high purity aluminum foils of two different thicknesses. In order to separate the influence of the texture from grain size and impurity effects, we have cut 99.999% purity aluminum foils either parallel or diagonal from identical sheets. Further, the sample preparation included a heat treatment for two hours at 550°C. Thereby, a pronounced cubic texture was obtained. Using a tolerance angle of 10° for the Euler angles, we obtained 30.6% cubic texture for 270 μm thick foils and 36.2% cubic texture for 540 μm thick foils. Then tensile tests were performed, where the angle between the tensile direction and the rolling direction was either 0° or 45°. The latter samples showed a smaller flow stress at low strain but a higher ultimate tensile strength at the end. This result was nicely reproduced by the foils of either thickness. In the last experimental step, the texture of the deformed specimens was recorded. It was found that the texture evolution is clearly altered when the orientation of the samples is changed. Nevertheless, the deformed samples have in common that the orientations of the grains are spread over wide angles after the tensile tests. Finally, an attempt was made to interprete the measured flow stress with the Taylor model. According to this model the elongation of the sample is generated by five operating slip systems in any grain. Taylor used the minimum shear principle to select the five active slip systems among the 12 slip systems of the fcc geometry with respect to the orientation of the grains. The ratio of the overall shear of the active slip systems to the macroscopic strain of the sample is the so called Taylor factor. The average Taylor factor should be proportional to the measured flow stress at low strain. However, first calculations yield deviating results. A possible explanation for this discrepancy could be that the number of active slip systems is actually below five.

Keywords: plasticity, tensile tests, crystallographic texture, Taylor model, plastic anisotropy

Wiadomo, ze umowna granica plastyczności i umocnienie odkształceniowe metali silnie zależą od wielkości ziarna i czystości próbek. Jednakże, tekstura krystalograficzna także odgrywa ważną rolę przy określeniu własności mechanicznych. Zwykle trudno jest rozróżnić wpływ wszystkich tych czynników na zachowanie materiału. Jeśli chce się zmierzyć efekt samej tekstury bez zakłócania przez inne rodzaje zjawisk, konieczne jest zmieniać właśnie teksturę, a pozostawić wszystkie inne czynniki stałe. W pracy przedstawiamy próby rozciągania wysokiej czystości folii aluminiowych o dwóch grubościach. Aby oddzielić wpływ wielkości ziarna i zanieczyszczeń od efektu tekstury wycięto folie z aluminium o czystości 99.999% albo równolegle albo diagonalnie z identycznych arkuszy cienkich blach. Dalsze przygotowanie próbek obejmowało obróbkę cieplną w 550°C. W ten sposób uzyskano wyraźną teksturę sześcienną. Stosując kątowa tolerancję 10° dla kątów Eulera otrzymaliśmy 31% tekstury sześciennej dla folii o grubości 270 μm i 38% dla folii o grubości 540 μm. Próby rozciągania wykonano przy dla kątów pomiędzy kierunkiem rozciągania, a kierunkiem walcowania albo 0° albo 45°. Ostatnie próbki wykazały mniejszą wytrzymałość plastyczną przy niskim odkształceniu, ale wyższą końcową wytrzymałość na rozciąganie. Ten wynik ściśle powtórzył się dla każdej z grubości folii. W ostatnim etapie eksperymentu zbadano teksturę odkształconych próbek. Stwierdzono, ze rozwój tekstury wyraźnie zmienia się kiedy zmienia się orientacja próbki. Niemniej jednak, wspólne dla odkształconych próbek było, że próbie rozciągania orientacje ziaren były rozmyte w szerokim zakresie kątowym. Na zakończenie, podjęto próbę interpretacji zmierzonej wytrzymałości plastycznej przy pomocy modelu Taylora. Zgodnie z tym modelem wydłużenie próbki jest generowane przez pięć systemów poślizgu działających w każdym ziarnie. Taylor użył zasady minimalnego ścinania, aby wybrać pięć aktywnych systemów poślizgu – spomiędzy 12 systemów poślizgu dla geometrii RPC – w zależności od orientacji ziaren. Stosunek całkowitego ścinania w aktywnych systemach poślizgu do makroskopowego odkształcenia próbki nazywane jest współczynnikiem Taylora. Średni współczynnik Taylora powinien być proporcjonalny do zmierzonej wytrzymałości plastycznej

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przy niskim odkształceniu. Jednakowoż pierwsze obliczenia dały błędne, odchylone wartości. Możliwym wyjaśnieniem dla tej niezgodności mogłoby być działanie mniej niż pięciu aktywnych systemów poślizgu.

1. Introduction

The influence of the texture on the tensile strength is frequently described by the averaged Taylor factor. According to the Taylor model [1] the grains of a polycrystal are elongated parallel to each other so that every grain experiences the same strain as the macroscopic specimen. In general five slip systems per grain must be active to achieve this deformation. Taylor has selected those combinations of 5 out of 12 slip systems which lead to the least amount of internal friction. This method is known as the minimum shear principle. In his own article Taylor assumed that the cross contraction is exactly the same for all directions perpendicular to the tensile direction. This approach is justified for randomly textured samples. If there exists, however, a preferred orientation of the grains, the fraction of the cross contractions along width and thickness may be further optimized with respest to the minimum shear principle [2-4]. The fraction of the overall shear to the macroscopic strain along the tensile direction is the Taylor factor. One has to consider the volume fractions of the individual grains in order to obtain the averaged Taylor factor. Since Taylor assumed that the resistance to glide is the same for every glide system, the Taylor factor should be proportional to the tensile strength. We have therefore performed tensile tests with high purity aluminum foils to test the predictions of this model. In fact, there are some possible effects which might disturb the interpretation of experiments: Samples with different textures usually have different grain sizes, and hence the grain size hardening will dominate the tensile behavior. Moreover, at high strain the dislocation structure of different samples will in general not be the same, so that the effect of texture on the strength is outperformed by the influence of dislocation dependent work hardening. To avoid such effects, we have cut our foils into stripes either parallel or diagonal from identical sheets. Thereafter, the foils were heat treated under identical conditions. At the yield strength, the dislocation densities are very small for both sample types. In principle our foils had the same microstructure, but the tensile direction was changed. A similar test of the Taylor model was performed by Li et al. [4] with extruded aluminum alloys.

2. Sample preparation, microstructure and initial textures

99,999% purity aluminum foils with thicknesses of 270 μm and 540 μm were cut into stripes of 80×10 mm from a rolled sheet. The angle between cutting and rolling direction was either 0° or 45°. After cutting the foils were recrystallized for two hours at an annealing temperature of 550°C. Thereafter, the average grain size of the 270 μm thick foils was 312.5 μm and that of the 540 μm foils was 540 μm. The texture was recorded by X-ray with the Schulz reflection method. Using a tolerance angle of 10° for the Euler angles, we obtained a volume fraction of 30.62% cubic texture for the foils of 270 μm thickness, and 36.21% cubic texture for the foils with 540 μm thickness. Recalculated pole figures are depicted in figure 1.

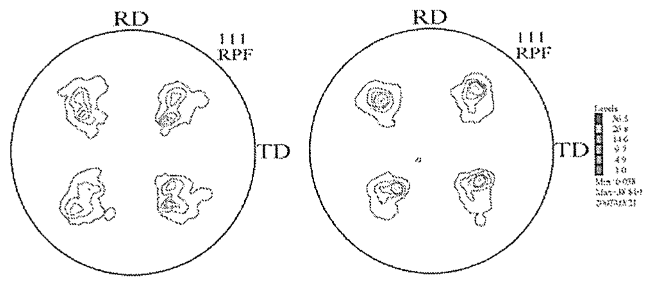

Fig. 1. Recalculated pole figures of the 1 1 1 reflex for the undeformed foils: Left: Foil with 270 μm thickness. Right: Foil with 540 μm thickness

3. Tensile tests and deformed textures

A micro tensile machine in combination with a laser speckle extensometer as non contacting optical strain sensor was employed to perform the tensile tests [5]. The sample is illuminated by two collimated laser diodes with a distance of 41,74 mm between the laser spots. A magnified picture of the laser speckle pattern is recorded repeatedly by two CCD cameras, which are connected to a computer using a frame grabber card for image processing. The displacement of the speckle pattern is thus calculated with use of the cross correlation function. The strain resolution of this laser speckle system is 10^{-5}. In order to ensure the reproducibility of the results the tensile tests of any sample type were performed 3 to 5 times. Full stress strain curves can be seen in the figure 2. A comparison of the yield strength and the ultimate tensile strength of all foils is given in table 1. Obviously, the foils with the rolling direction parallel to the tensile direction had the higher yield strength but the lower ultimate tensile strength. The recalculated pole figures of the deformed samples can bee seen in figure 3.

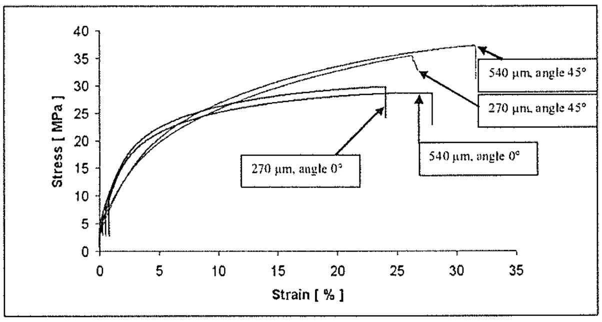

Fig. 2. Technical stress strain curves of foils which were cut in an angle of either 0° or 45° to the rolling direction. For the comparison we have chosen the curves with medium ductility out of at least 3 curves for each sample type

TABLE 1

Comparison of the yield strength and the ultimate tensile strength of the foils. The yield strength is an average value from 3-5 samples. For the ultimate tensile strength we have taken the value from the foil with medium ductility of any sample type

Foil thickness	Angle between rolling and tensile direction	Yield strength at 0.2% plastic strain	Ultimate tensile strength
270 μm	0°	6.0 +/- 0.3 MPa	29.8 MPa
270 μm	45°	5.0 +/- 0.3 MPa	35.5 MPa
540 μm	0°	5.0 +/- 0.3 MPa	28.65 MPa
540 μm	45°	3.8 +/- 0.4 MPa	37.3 MPa

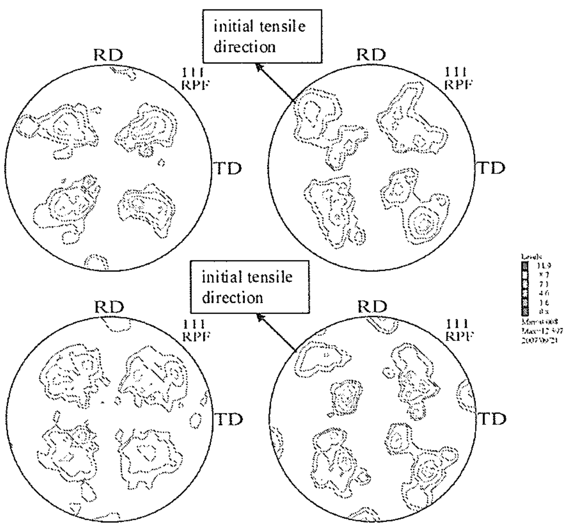

Fig. 3. Recalculated pole figures of the deformed samples. Top left: 270 μm thick, orientation angle 0°. Top right: 270 μm thick, orientation angle 45°. Bottom left: 540 μm thick, orientation angle 0°. Bottom right: 540 μm thick, orientation angle 45°

4. Averaged Taylor factors and r – values

In order to evaluate the average value of the Taylor factor, we have to consider the volume fractions of the grains. They are obtained from the orientation distribution function (ODF) of the textures. The ODF is defined as the volume fraction of grains oriented along some direction g

$$ODF(g) := \frac{\Delta V(g)}{V dg} \tag{1}$$

where the direction g is usually identified using the 3 Euler angles φ_1, ϕ and φ_2. The ODF may be normalized in the sense that for a perfect random texture, where all orientations occur with the same frequency, the ODF takes the value 1 everywhere. We have received the ODF of our recorded textures from a commercial program. This, of course, reduces the amount of computation work that we have to do ourselves. Nevertheless, there are two details of the calculation one must be careful about: First, intervals of equal size in the Euler space lead to a higher density of interpolation points for small angles of ϕ. This distortion of the Euler space must be compensated by a weighting coefficient when the averaged Taylor factor is evaluated. Second, the ODF sometimes contains informations about the volume fractions in redundant style. This means that some distinct physical orientations occur repeatedly in the table of the ODF. Any such physical orientation should be counted only once when the average value of a physical quantity is calculated.

Since our undeformed foils showed a pronounced cubic texture, Taylor's assumption that the cross contractions were the same along all directions perpendicular to the tensile axis is not useful here. Instead, the plastic anisotropy of the material must be considered as demonstrated by Bunge [2]. The ratio of the cross contractions along width and thickness are optimized with respect to the minimum shear principle [2-4]. However, any single grain still experiences the same deformation as the macroscopic sample. Thus, the strain rate tensor D of the sample in the laboratory coordinate system takes the form

$$D = \begin{pmatrix} 1 & 0 & 0 \\ 0 & -q & 0 \\ 0 & 0 & q-1 \end{pmatrix} \tag{2}$$

whereby the tensile direction points along the x – axis, the width is measured along the y – axis and the thickness along the z – axis. The r – value is defined as the ratio of the momentary cross contractions along width and thickness. It reads as

$$r = \frac{q}{1-q} \tag{3}$$

In order to obtain the optimized r – vale, the averaged Taylor factor is calculated consecutively for different values of q, and finally that value is adopted as the result, which gave the least value of the Taylor factor. The results for the r – values and the correlated Taylor factors for the various foils are summarized in table 2.

TABLE 2
Comparison of calculated r – values and averaged Taylor factors of the foils. The deformed textures used in this table were recorded after rupture

Foil thickness	Angle between rolling direction and tensile axis	Deformation status	r – value	Averaged Taylor factor
270 μm	0°	undeformed	1.17	2.516
270 μm	45°	undeformed	0.00	2.727
270 μm	0°	deformed	1.42	2.565
270 μm	45°	deformed	0.01	2.743
540 μm	0°	undeformed	0.97	2.535
540 μm	45°	undeformed	0.01	2.644
540 μm	0°	deformed	0.97	2.737
540 μm	45°	deformed	0.04	3.026

5. Discussion

The comparison of our experimental results for the yield strength measured at 0.2% plastic strain with the evaluated Taylor factors of the undeformed samples deviates from the predictions of the Taylor model. According to the calculated Taylor factors, one would expect that the foils with the rolling direction parallel to the tensile axis should exhibit the lower yield strength than the foils where the angle between rolling direction and tensile axis was 45°. However, the tensile tests showed the opposite behavior. On the other hand, the differences between the predictions of the averaged Taylor factors for the sample orientations considered here are rather small. There are a number of possible reasons which might be responsible for the observed deviations from the Taylor model:

First, a surface roughening of the foils was observed. This clearly indicates an inhomogeneous deformation of the grains which cannot be described by the Taylor model. But it is unclear how the surface roughening could invert the predictions of the Taylor model. Second, the Taylor model neglects slip on planes other than (1 1 1) planes. Indeed, dislocations on (1 0 0) planes were found with TEM observations in fcc crystals [6]. The Schmid factors of the slip systems on the (1 0 0) planes are much smaller for the orientations where the rolling direction is parallel to the tensile axis compared to orientations where this angle is 45°. So this effect might contribute to the deviation from the Taylor model as observed in the experiments. Third, there arises the question whether really 5 slip systems are active per grain. The Taylor model certainly plays the role of an upper bound model. Five slip systems per grain are necessary to achieve a full constraint model. If one somewhat relaxes the geometrical constraints and permits small misfits of the plastic strain between neighboring grains the number of active slip systems may be reduced to 3 or 4 [7-11]. Relaxed constraint Taylor models were first developed to explain the rolling textures of fcc metals. However, from observations of slip bands one gets the impression that the number of active slip systems is also below five during tensile tests. Further, the RC models may improve the satisfaction of the stress condition between the grains. In contrast, the stress equilibrium at the grain boundaries was violated in the FC Taylor model. It has been reported that the coincidence between tensile tests and theoretical predictions may be enhanced with a realxed constraint Taylor model [4].

The comparison of the measured ultimate tensile strength clearly shows that the foils which were cut in an angle of 45° to the rolling direction were stronger at the end of the tensile test. However, the Taylor factors calculated on the basis of the deformed textures showed only a small difference for the foils of either orientation. So it does not seem that the final distribution of orientations alone can be responsible for the differences of the ultimate tensile strength. It may be speculated that the dislocation structure which evolved during the deformation process makes the difference.

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