VIII International Conference on Computational Plasticity
COMPLAS VIII
E. Oñate and D. R. J. Owen (Eds)

ANISOTROPIC YIELD LOCUS OF ULTRAFINE-GRAINED ALUMINUM PRODUCED BY EQUAL CHANNEL ANGULAR EXTRUSION

F. El Houdaigui*; L. Duchêne; A.M. Habraken*; S. Poortmans***; B. Verlinden*****

*Department of Mechanics of materials and structures
University of Liège, Chemin des Chevreuils 1, 4000 Liège, Belgium
Email : Anne.Habraken@ulg.ac.be; felhoudaigui@ulg.ac.be

** COBO Department
Royal Military Academy, Avenue de la Renaissance 30, 1000 Brussels, Belgium

*** Department Metaalkunde en Toegepaste Materiaalkunde
Katholieke Universiteit Leuven, Dept. MTM, Kasteelpark Arenberg 44, 3001 Leuven, Belgium
Email: Bert.Verlinden@mtm.kuleuven.be; Stijn.Poortmans@mtm.kuleuven.be

Key words: Computational Plasticity, Nano-mechanics, Anisotropy, Aluminum, ECAE

Summary: *Macroscopic tensile, compression and torsion tests have been performed as well as texture measurements. Based on these data a phenomenological yield locus has been defined to describe the behaviour of an ultrafine-grained aluminum obtained by ECAE.*

1 INTRODUCTION

The investigated material was hot rolled commercial pure aluminum AA1050 with an initial grain size of about 100 µm. In order to obtain an ultrafine-grained (UFG) material, equal channel angular extrusion (ECAE) was performed at room temperature (RT), following route B_C for 8 passes. The final grain size was about 1µm. On this ECAE material state, macroscopic uniaxial tensile, axisymmetric compression and torsion tests were carried out at RT.

The crystallographic texture was measured and the corresponding Taylor-Bishop-Hill (TBH) yield locus was computed. The Lankford coefficients (r-values) calculated using a TBH model were used to determine the parameters of the Hill 1948 yield locus.

To account for tension-compression asymmetric behaviour caused by the ECAE process a simple approach is chosen: to use the Hill yield function shifted by a proper initial back stress defined by experimental test results. This rough model is validated by the yield strength obtained in a torsion test.

2 MECHANICAL TESTS

To characterize the ECAE material, uniaxial tensile, axisymmetric compression and torsion tests were carried out at RT. The tensile, compression and torsion true stress–true strain curves are shown in Fig. 1. The tensile and compression curves were computed from force displacement measurements assuming homogeneous stress and strain fields and taking into account a section defined by volume conservation. The torsion curve is based on the Nadaï formula[1]. The ECAE material exhibits a high tensile yield strength (σ_y^t = 186 MPa) and a small homogeneous deformation of only 2%. The latter is followed by strain softening coupled with necking phenomena, while fracture strain occurs after a technical strain of 13%. A low work hardening or even softening is often observed in UFG materials exhibiting a high defect density. For some severe plastic deformation processed metals such as ECAE, it is not surprising that the strain hardening capacity is close to being exhausted[2]. The lack of further strain hardening has sometimes been explained by the absence of dislocation activities or even a fundamental change of deformation mechanisms. It is well known that the ability to strain harden is important for stabilizing uniform tensile deformation, the relevance of this concept to nanostructured metals was not emphasized until recently. For compressive and torsion tests the yield stress is repectively σ_y^c = 144 and $\sigma_y^{torsion}$ = 107 MPa. A nearly constant flow stress approximately elastic to perfectly plastic response is exhibited after an initial stage of rapid strain hardening over a very small plastic strain (~1-3%). These results indicate that the strain hardening capacity is saturated very quickly for the tensile test. Dynamic equilibrium between hardening and softening is not obtained, while in compression and torsion tests this dynamic equilibrium is perfectly reached.

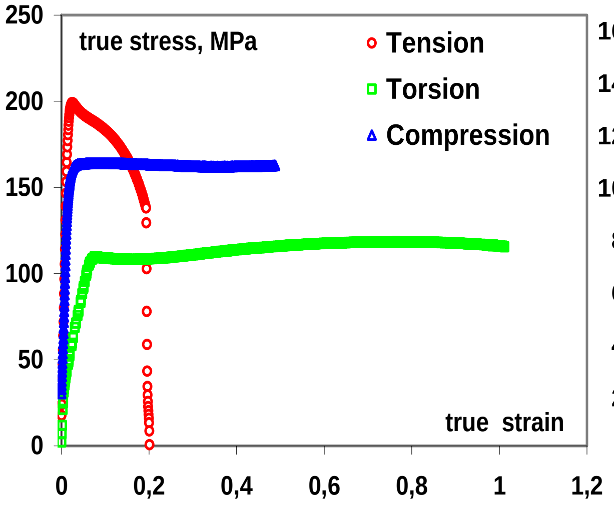

Fig. 1 : Tensile, compression and torsion true stress/true strain curves of ECAE-Al

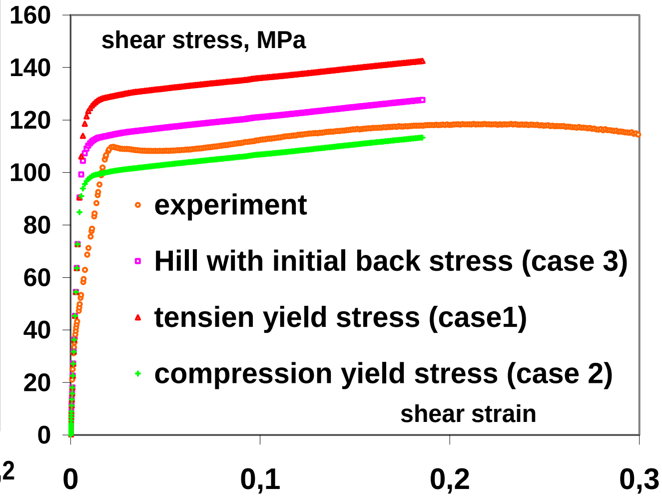

Fig. 2 : Numerical prediction of torsion for several cases.

3 IDENTIFICATION OF THE HILL YIELD LOCUS SHAPE

The Hill model implemented in Lagamine[3] finite element code was used. Hill (1948)'s yield locus[4] is defined by:

$$f(\underline{\sigma}) = F(\sigma_{22} - \sigma_{33})^2 + G(\sigma_{33} - \sigma_{11})^2 + H(\sigma_{11} - \sigma_{22})^2 \\ + 2L\sigma_{23}^2 + 2M\sigma_{13}^2 + 2N\sigma_{12}^2 - 2\sigma_F^2 = 0 \tag{1}$$

where $\underline{\sigma}$ is the stress represented by a 6 component vector, σ_F is the yield stress under uni-axial tension in a reference direction. The anisotropic shape of this yield surface is determined by the six coefficients *F, G, H, L, M* and *N*.

A full-constraints Taylor model was used to calculate the r-values (equation 2) of the studied ECAE material from its initial texture. In Fig. 3, the r-values are plotted as a function of the angle θ between the tensile axis and extrusion direction (ED), which corresponds to the former rolling direction (RD). This method takes into account only the crystallographic anisotropy. The link between r-values and the Hill parameters is defined by equation 2. The Hill parameters in Table 1 were identified from r-values (equation 3), obtained at 0°, 45° and 90° from the RD:

$$r_\theta = \frac{H + (2N - F - G - 4H)\sin^2\theta\cos^2\theta}{F\sin^2\theta + G\cos^2\theta} \tag{2}$$

$$r_o = \frac{H}{G} = 2.08\,,\quad r_{90} = \frac{H}{F} = 1.44\,,\quad r_{45} = \frac{2N - F - G}{2(F+G)} = 0.8 \tag{3}$$

After ECAE, Lankford coefficients significantly increase (Fig. 3). This improvement of plastic strain ratio is related to the texture and increases the formability of ECAE material.
In Fig. 4, the anisotropic yield locus computed by Hill model for ECAE material identified by Lankford coefficients is compared with TBH yield locus shape associated with initial crystallographic texture for ECAE material and hot rolled Aluminum.

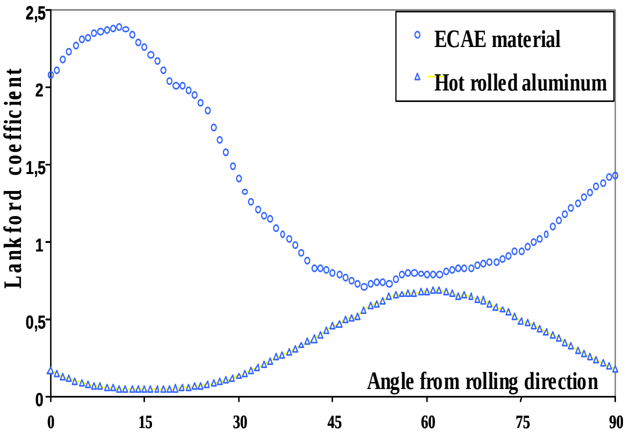

Fig. 3 : r-values calculated from the crystallographic texture for ECAE material and hot rolled Al

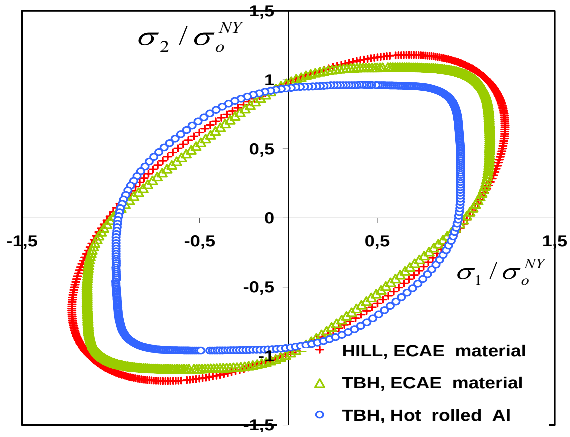

Fig. 4: Yield locus shape computed by Hill model and TBH law

Table 1: parameters of Hill's model obtained from Lankford coefficients of ECAE-Al.

F	G	H	N=L=M
0.94	0.65	1.35	2.07

For this ECAE material, the yield strength under uniaxial tension σ_y^t is about 20% higher than in compression σ_y^c. Notice that tension-compression asymmetry disappears for larger grain sizes[5]. The difference in yield strength for tensile and compression test is probably due to different dislocation mobility along grain boundaries under tensile or compression state, due to pre-deformation by ECAE. Based on the above experimental results, a back stress is introduced to shift the center of the yield surface to account for this tension-compression asymmetry. The position of the center of the initial yield locus is given by the initial Back Stress *(BS)* after ECAE: $BS = (\sigma_y^t - \sigma_y^c)/2 = 21$ MPa. The size of the initial yield locus (σ_o^{NY} : plastic radius of the yield locus) is given by the average between the tension and compression strength $\sigma_o^{NY} = (\sigma_y^c + \sigma_y^t)/2 = 165$ MPa.
The yield strength obtained in a torsion test is compared with the one calculated using different sizes of yield locus (Fig. 2). The test simulations were performed either with the size of the yield locus determined respectively by yield limit in tension (case 1) and compression (case 2) with no back stress or in (case 3) with the new size of yield function σ_o^{NY} and with the initial *BS*, no hardening is assumed. Note that the shape of the Hill yield locus is the identical for all simulations. The yield surface with new size σ_o^{NY} and with the initial *BS* gives the best fit of torsion test ($\sigma_y^{torsion}$ = 107 MPa) compared to other simulaions. Note that the apparent hardening in torsion is due to the Nadai formula and its inaccurate description of the progressive yielding of the section. The pure shear stress-strain curve is effectively

perfectly "elastic-plastic" as checked in the FE computation. However Fig. 2 applies the Nadaï formula on both numerical and experimental results to allow comparison.

4 ANISOTROPIC HARDENING FOR ECAE MATERIAL

When the plasticity is controlled by the dislocation mode, the relation between flow stress and dislocation density ρ is given by $\sigma_F = M_T \alpha G b \sqrt{\rho}$, where M_T (=3.065) is the tensile Taylor factor accounting for a polycrystal, α is a numerical factor ($\alpha \approx 0.25$) depending on the character of the dislocation considered, G is the shear modulus (26 GPa for Al) and b is the Burgers vector (b=0.286 *nm*). σ_F defines the flow stress at a given dislocation density measured in a 'reference condition'. The evolution equation of the total dislocation density is formulated by:

$$\frac{d\rho}{d\varepsilon} = M_T\left(k_1\sqrt{\rho} - k_2\rho\right) \tag{4}$$

where k_1 and k_2 are respectively the dislocation generation and annihilation constants.
In the vast majority of UFG materials, tensile tests presents hardening followed by softening. Therefore a softening term is added in the integration of equation (4) only for tensile test:

$$\sigma_F - \sigma_{0.2\%} = Q_1\left(1-e^{-b_1(\varepsilon-\varepsilon_{0.2\%})}\right) - Q_2\left(1-e^{-b_2(\varepsilon-\varepsilon_u)}\right) \tag{5}$$

where $Q_1 = \chi.(k_1/k_2)$, $b_1 = (M_T/2).k_2$ and $\chi = M_T\alpha G b$.

From the macroscopic experimental results (Fig. 1), it is possible to estimate the constants in the hardening part (b_1=340, Q_1=14 MPa) for tensile test, (b_1 =170, Q_1=20 MPa) for compression test and in the softening part for tensile state (Q_2=14 MPa, b_2=30). The highest annihilation constant was observed during tensile test, i.e. more than 2 times the annihilation constant of the compression test. Any hardening function proposed for a general ECAE behaviour model must depend on the plastic state, either in a phenomenological way or in an approach based on the microscopic phenomena. The monotonic torsion data will allow first validations but tests with strain paths change are strongly required to check the predictive capacity of such a model.

5 CONCLUSIONS

Three types of tests were performed to fully characterize the initial yield locus and tobegin the investigation of the strain hardening behaviour. Tests with strain path change are required to achieve a better understanding in the strain hardening behaviour. The simulations showed that the anisotropic Hill function with a proper initial back stress is able to recover monotonic stress-strain curves. Further validation will be the FE prediction of the shape of the equatorial section of a compressed sample.

Acknowledgements: The authors would like to acknowledge the Belgian Science Policy for financial support through Contract P5/08. A. M. Habraken is mandated by the National Fund for Scientific Research (Belgium).

REFERENCES

[1] F. Székely, I. Groma, J. Lendvai. Nonlocal effects in torsional deformation. Materials Science and Engineering A277 (2000) 148-153
[2] Y. M. Wang, E. Ma. Three strategies to achieve uniform tensile deformation in a nanostructured metal. Acta Materialia 52 (2004) 1699-1709.
[3] Lagamine users guide, FE code developed by M&S dept in ULG.
[4] R. Hill. A theory of the yielding and plastic flow of anisotropic metals, Proc. R. Soc. London 193A (1948).
[5] A.C. Lund, C.A. Schuh. Strength asymmetry in nanocrystalline metals. Acta materialia (2005) in press.