LEGIBILITY NOTICE

A major purpose of the Technical Information Center is to provide the broadest dissemination possible of information contained in DOE's Research and Development Reports to business, industry, the academic community, and federal, state and local governments.

Although a small portion of this report is not reproducible, it is being made available to expedite the availability of information on the research discussed herein.

1

LA-UR -89-1241

Received OSTI
MAY 09 1989

LA-UR--89-1241

DE89 011181

TITLE POLYCRYSTAL PLASTICITY AS APPLIED TO THE PROBLEM OF IN-PLANE ANISOTROPY IN ROLLED CUBIC METALS

AUTHOR(S) Anthony D. Rollett, MST-6
Michael G. Stout, MST-5
U. Fred Kocks, CMS

SUBMITTED TO To be presented at "PLASTICITY-89", July 31-August 4, 1989, MIE University, Tsu, JAPAN, (Invited paper)

Los Alamos Los Alamos National Laboratory
Los Alamos, New Mexico 87545

DISTRIBUTION OF THIS DOCUMENT IS UNLIMITED

POLYCRYSTAL PLASTICITY AS APPLIED TO THE PROBLEM OF IN-PLANE ANISOTROPY IN ROLLED CUBIC METALS

A. D. ROLLETT, M. G. STOUT AND U. F. KOCKS

Mail Stop K765, Los Alamos National Laboratory, NM 87545, USA

ABSTRACT

A fundamental property of cubic metals is that slip occurs on close-packed planes in close-packed directions, which for the f.c.c. case results in 12 {111}<110> slip systems. This crystallographic restriction on the plastic behavior causes significant crystallographic preferred orientation (texture), hence anisotropy, to develop once a large strain has been imposed. Moreover, whereas annealing can generally "reset" the flow stress and ductility, it does not generally randomize the texture: therefore most metallic materials have some degree of texture and consequent anisotropy. The problem of earing in deep drawing can be simply related to the variation of r-value with angle from the rolling direction, i.e. the in-plane anisotropy of the sheet. The r-value can be calculated from a given texture with the use of a polycrystal plasticity model. The Los Alamos polycrystal plasticity (LApp) code is based on the Bishop-Hill single crystal yield surface (SCYS) but with a mildly strain-rate sensitive modification where the stress exponent is of order 30. This modification of the SCYS removes the ambiguity of slip system selection inherent in the Bishop-Hill formulation and permits other phenomena to be treated such as latent hardening and pencil glide. The use of LApp to simulate texture formation and consequent anisotropy is described. Experimental textures in the form of X-ray pole figures are analyzed with a Williams-Imhof-Matthies-Vinel (WIMV) code, as implemented by Kallena, to give full orientation distributions (OD's). The OD obtained this way contains approximately 5000 points on a 5° by 5° by 5° lattice; this is used to assign weights to approximately 1000 discrete orientations for calculations with LApp. The results of experimental measurements of textures and r-values in copper rolled to a reduction of 83% are compared with the simulation results of the polycrystal plasticity model for this case.

KEYWORDS

Plasticity, Texture, Anisotropy, r-values, Micro-mechanics, Copper, Polycrystal Plasticity, Orientation Distributions

INTRODUCTION

The theme of this paper is the quantitative prediction of material behavior and especially that of in-plane anisotropy of metal in sheet form. This is not a new topic but, whereas much previous work (e.g. da Costa Viana *et al.*, 1978) has concentrated on the behavior of annealed material, the results presented here focus on the as-rolled condition. This was motivated by the need to model anisotropy in the deformed state in problems such as metal forming. Experiments are described in which the f.c.c. metal copper was rolled to a strain of 2, the texture measured, and the in-plane anisotropy measured by means of sheet tensile tests. The results are expressed in the form of the variation of r-value with direction. Tensile tests on as-rolled f.c.c. metals are complicated by the small strain-to-failure that can be achieved (~0.2%) which necessitates the use of strain gauges. The texture of the rolled copper was determined by taking three X-ray pole figures, {111}, {200} and {220}, and calculating the three-dimensional orientation distribution (OD) with the WIMV method, as implemented by Kallend. The

OD is then used to weight a set of discrete orientations, distributed throughout orientation space, and this set is used with the Los Alamos polycrystal plasticity (LApp) code to simulate the anisotropy of the material. The operation of LApp is summarized and lastly comparisons of simulation with experiment are made.

EXPERIMENTS

The copper used in this work was an oxygen-free electrical grade copper (C10200 grade) which was purchased in the form of 25 mm rolled plate. This copper was rolled to 83% reduction which is equivalent to a von Mises strain of 2. Sheet tensile samples were machined out at 11° intervals to the rolling direction with dimensions 1.6 mm thick, 6 mm wide and a gauge length of 50 mm. Care was taken to avoid near-surface shear textures by machining out 1.6 mm thick samples from the 6 mm thick rolled plate. Additionally, the texture was measured from the outer surface of one of the tensile samples and it was verified that the texture was the same as at the center of the sheet. Two pairs of strain gauges were applied, one at the center and one near the end of the gauge section. The results showed that the two locations behaved very similarly. One gauge in each pair was parallel to the tensile axis to measure the tensile strain and one was perpendicular to it to measure the width strain. The loads and strain gauge displacements were recorded as a function of time in addition to the customary load-displacement history. Yield stresses were obtained as 0.2% offset stresses and r-values (ratio of width to thickness plastic strain increments) were obtained in the plastic region. The results are presented with the simulation results below.

TEXTURE MEASUREMENTS

Samples for texture measurement were cut from the rolled sheet so as to expose the central plane and were ground and polished following standard metallographic procedures. X-ray pole figures for the {111},{200} and {220} reflections were measured in a back-reflection goniometer on samples of material approximately 10 mm square. The pole figure data were corrected empirically for defocussing (drop-off in measured intensity beyond a tilt of about 60°), normalized and rotated to maximize the orthotropic sample symmetry expected in rolling. The data were then analyzed with a WIMV algorithm to obtain a discrete three-dimensional (3D) orientation distribution (OD). For simulation of anisotropy, a set of 1152 orientations was weighted according to the intensity of the experimentally determined OD at each discrete orientation. The orientations are arranged on a lattice in the 3-D orientation space at approximately 7.5° intervals. By taking advantage of cubic crystal symmetry and a diad axis on the sample Z-axis, the number of orientations is reduced to the 1152 used here. Lower crystal symmetries such as hexagonal, would require more orientations. The texture obtained is illustrated in Fig. 1 in the form of a discrete {111} pole figure in which each {111} pole of the discrete orientations used for subsequent simulation has been plotted with a symbol whose size corresponds to the square root of the weight assigned to that orientation. The pole figure shows that the texture is the standard f.c.c. rolling texture, dominated by the "copper" component at {112}<111>. Note that the pole figure is in effect a recalculated pole figure so the information at the edge of the pole figure is correct, unlike the experimental pole figure which has no information at the edge.

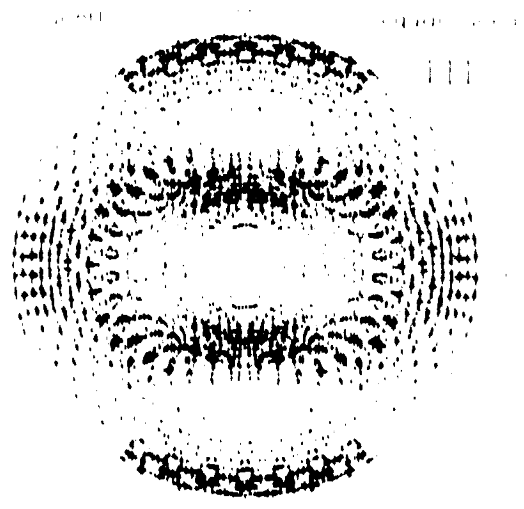

Fig. 1. {111} pole figure for copper rolled 83%, showing a normal f.c.c. rolling texture. This pole figure has been plotted from the discrete orientations used to represent the texture in computer simulations and the size of each symbol corresponds to the weight of that orientation.

POLYCRYSTAL PLASTICITY MODEL, LApp

The basis of LApp is the Taylor model of polycrystal deformation in which it is assumed that the deformation of each grain is identical to that of the polycrystal. This preference of compatibility over stress equilibrium has been verified by comparison of Taylor-based simulations with experimental textures (Kallend & Davies, 1972). When large strains are applied and the grain shape has a large aspect ratio, however, it is sometimes appropriate to deviate from the Taylor model: this deviation is known as Relaxed Constraints (RC) which was originally introduced (Honneff & Mecking, 1978) to account for certain features of f.c.c. rolling textures. In LApp, the RC model means that a stress boundary condition is applied to the two shear components in uniaxial or plane-strain compression on the rolling plane; the strain in those two components is then determined by the crystal orientation. The basis of the RC model is that when the grain shape reaches a large aspect ratio at large plastic strains, stress equilibrium becomes a more reasonable model than strain compatibility (Kocks & Canova, 1981). For the limiting case of a rate-insensitive material, as discussed by Bishop & Hill (1951), there is a finite set of discrete stress states that will satisfy an arbitrary strain increment applied to a single crystal (grain) in an arbitrary orientation. These discrete stress states are vertices in the single crystal yield surface. Although this leads to a computationally efficient method, there lacks a unique solution for the shear rates on the 12 {111}<110> slip systems of an f.c.c. crystal because the vertices activate either 6 or 8 slip systems simultaneously. Therefore we make the physically realistic assumption (Kocks *et al.*, 1975) that the relationship between resolved shear stress, τ, and shear rate, γ, on an individual slip system is rate-sensitive:

$$\frac{\dot{\gamma}}{\dot{\gamma}_0} = \left(\frac{\tau}{\tau_0}\right)^n, \quad n = 30 \tag{1}$$

where γ_0 and τ_0 are a pair of reference values. The exponent is chosen to be high enough that the texture evolution corresponds to that of actual, nearly rate-insensitive materials but is as low as possible for computational convenience. This constitutive relation leads to a set of five independent non-linear equations,

$$D_{ij} = \dot{\gamma}_0 \, m^s_{ij} \left(\frac{m^s_{mn} \ \sigma_{mn}}{\tau^s_0}\right)^n, \tag{2}$$

where D is the applied (polycrystal) strain rate, σ is the applied stress on the single crystal, m^s is the Schmid factor matrix for the slip system s and $\dot{\gamma}_0$ is a reference strain rate. The orientation of the grain is accounted for by rotating each slip system into the frame used to describe the grain shape. This set of equations is reduced from a nominal nine to five by applying Lequeu's normalization to the stress and strain and the equations are solved iteratively by Newton's method. When RC model applies, only three components of the strain must be satisfied so the number of equations is reduced to three. The method of determining the volume fraction of grains (orientations) that are deforming in RC is based on that described by Tiem et al. (1986); for these calculations, each grain is either deforming under the Taylor assumption (five strain components to be satisfied) or two components are relaxed (three strain components to be satisfied). Once a solution for the stress and distribution of shear rates has been obtained for each grain, averages are taken to obtain the properties of the polycrystal.

When a tensile test is modeled, it is necessary to apply stress boundary conditions to the polycrystal since some components of the strain are determined by the anisotropic properties - the texture - of the material. This is not a straightforward procedure with the Taylor model because of the assumption that the microscopic strain is identical to the macroscopic strain. Therefore an initial guess of an axially symmetric tension is used for the strain but those strain components that correspond to components of the average stress that do not meet the boundary conditions, which for a tensile test are zero stress in all directions except the tensile axis, are altered iteratively until the stress boundary conditions on the polycrystal are satisfied (Rollett *et al.*, 1986). The criterion is that the difference between the expected value of a stress component and the average value over the polycrystal must be less than small fraction, say 0.2, of the mean deviation in that component. The r-value can then be calculated from the final value of the strain.

The small plastic strains obtained in the experiments were of the same order as the total elastic strains. In this case the Taylor model is not expected to apply (Budianski and Wu, 1962). To investigate the other extreme, a modification of LApp was made to simulate the so-called "Sachs" model in which a uniform stress is applied to each grain, the maximum resolved shear stress is used to select a single active slip system and the strain response of the polycrystal is then taken to be the average over the polcrystal. "Uniform stress" in this model means that the direction of the stress is held constant (as a tensile stress) but the magnitude of the stress varies from grain to grain. The strain response of each crystal is normalised such that the imposed tensile strain is satisfied, that is, a single component of the strain is satisfied. Figure 2 shows the results obtained by simulating tensile tests on the texture shown in Fig. 1, using both the standard RC model and the "Sachs" model described above. At a rolling strain of 2, the volume fraction of grains deforming in RC is 0.64.

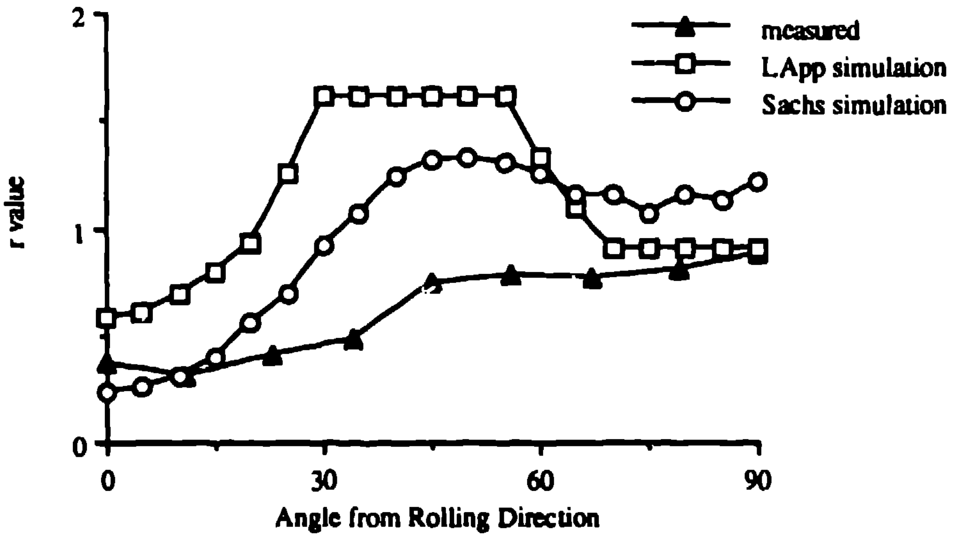

Fig. 2. Plot of the variation of r-value with angle from the rolling direction for rolled copper, for both experimental (solid symbol) and simulated (open symbol) results.

DISCUSSION

The LApp results show that the anisotropy calculated from the texture of rolled copper is such that there is a pronounced peak in the r-value at 45° to the rolling direction. This agrees with previous results for simulated rolling textures (Rollett *et al.*, 1986). The experimental r-values, however, are all less than unity and show only a gradual rise in r-value as the test direction deviates from the rolling direction. This experimental result is in disagreement with the results of Hirsch (1978) who measured r-vlaues at 0°, 45° and 90° and found a peak at 45°. There is better agreement with the results of Stephens (1968) who measured r-values in rolled pure copper every 10° and obtained values all less than unity. The difference between our experimental and LApp results are ascribed to the fact that the measured plastic strains are of the order of the elastic strains for the reasons discussed above. The results of the Sachs calculation show an improvement in the fit in that the r-values fit well at small angles and plateau at high angles. Previous work on this topic has found good agreement between anisotropies calculated with the Taylor model (Rollett *et al.*, 1986) and earing behavior in deep drawing. Deep drawing tests involve large plastic strains, however, and here the Taylor model does apply.

ACKNOWLEDGEMENTS

Fruitful discusisons with Dr. S. Panchanadeeswaran of the ALCOA Center are acknowledged. This work was supported by the U.S. Department of Energy, Office of Basic Energy Sciences, Division of Materials Sciences.

REFERENCES

Bishop, J. F. W. and Hill, R. (1951). *Phil. Mag.*, 42, 1298.

Budiansky, B. and Wu, T. T. (1962). *Proc. 4th Congr. Appl. Mech.*, 1175.

da Costa Viana, C. S., Davies, G. J., and Kallend, J. S. (1978). *ICOTOM-5*, Aachen, W. Germany, 265.

Hirsch, J., Musick, R. and Lücke, K. (1978). *ICOTOM-5*, Aachen, W. Germany, 2, 437.

Honneff, H. and Mecking, H. (1978). *ICOTOM-5*, Aachen, W. Germany, 1, 265.

Kallend, J. S. and Davies, G. J. (1972). *Phil. Mag.* 25, 471.

Kocks, U. F., Argon, A. S. and Ashby, M. F. (1975). *Prog. Mat. Sci.*, 19.

Kocks, U. F. and Canova, G. R. (1981). *Proc. 2nd Riso Symposium*, Denmark, 35.

Rollett, A. D., Canova, G. R. and Kocks, U. F. (1986). In: *Formability and Metallurgical Structure* (Sachdev, A. K. and Embury, J. D., Eds.), TMS-AIME, 147.

Stephens, A. W. (1968). Ph.D. thesis, Univ. of Arizona.

Tiem et al. C., Berveiller, M., and Canova, G. R. (1986). *Acta metall.*, 34, 2139.