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# NON-EQUILIBRIUM FEATURES OF CONTINUOUS RECRYSTALLIZATION PROCESS AT SEVERE PLASTIC DEFORMATION OF COPPER

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**Abstract.** It is shown that separation of scalar dislocation density into two parts, giving densities of mobile and immobile dislocations, is providing a possibility to predict continuous dynamic recrystallization (DRX) process caused by severe plastic deformation of metallic materials. The proposed approach is making it possible to predict a rather extent set of material microstructure properties that can be measured experimentally, based on a minimum set of "tuning" parameters. Received predictions are verified by comparison to experimental results and predictions received using another known dislocation plasticity and dynamic recrystallization models**.**

## 1. Introduction

Nowadays none of the known models can offer a correct prediction of deformation that should be applied to a sample in order to receive desired defect structure with a-priori given physical properties. At the same time the available dislocation based plasticity models [1–5] are able to make a reliable prediction of resultant material grain size after one or several ECAP passes [5]. Unfortunately, these models contain several "trimming" parameters that should be evaluated experimentally. Further development of ultrafine-grained (UFG) structure material engineering requires an analysis of the existing models of SPD, a study of influence of "trimming" parameters on the received solutions, a study of the possibility to provide clear physical meaning and measurability to these parameters, a study of validity limit of these models, etc. The final result should consist in appearance of a rather universal material behaviour model having no "trimming" parameters and applicable for prediction of a wide range of processes and materials.

## 2. Modification of the classical model for a separate account of mobile and immobile dislocation densities

A model proposed by Mayer and co-authors [2, 3] extends the applicability limits and improves the classical model [1]. For quasistatic case it can be reduced to a classical model with the principal distinction from this model consisting in separation of dislocation density $\rho_D$ into two parts giving densities for mobile $\rho_D^{mob}$ and immobile $\rho_D^{im}$ dislocations. Evolution of

dislocation densities (taking into account the Orowan equation $\dot{\varepsilon}_{plast} = \rho_D^{mob} b |V_D|$ for plastic strain rate) is given by the following kinetic equations [2]:

$$\frac{d\rho_D^{mob}}{d\varepsilon} = \left( k_g \, b \, \sigma_y^0 + \alpha G b^2 k_g \sqrt{\rho_D^{im} + \rho_D^{mob}} \right) - \frac{V_C (\rho_D^{mob} - \rho_D^0) \sqrt{\rho_D^{im}}}{\dot{\varepsilon}_{pl}} - k_\alpha^m (2\rho_D^{mob} + \rho_D^{im}),$$

$$\frac{d\rho_D^{im}}{d\varepsilon} = \frac{V_C (\rho_D^{mob} - \rho_D^0) \sqrt{\rho_D^{im}}}{\dot{\varepsilon}_{pl}} - k_\alpha^i \rho_D^{im}, \qquad (1)$$

Here $k_g = \eta / \varepsilon_L b$ is the coefficient responsible for dislocation generation, $\alpha \sim 0.5$ is the Taylor constant [3], $\varepsilon_L = 8\text{eV}/b$ is the elastic energy of dislocation length unit [2, 6], $\eta \sim 0.1$ is the fraction of work of plastic deformation stored in the form of defects. Thus, the model [2, 3] adds two new parameters to the ones used within the classical model [1]: the rate of dislocation immobilization $V_C$ and the annihilation constant for immobile dislocations $k_\alpha^i$ [7]. Figure 1 presents the scalar density for mobile (a) and immobile (b) dislocations as a function of strain. Local maximum for mobile dislocation density is clearly observed in compliance with experimental observations [8].

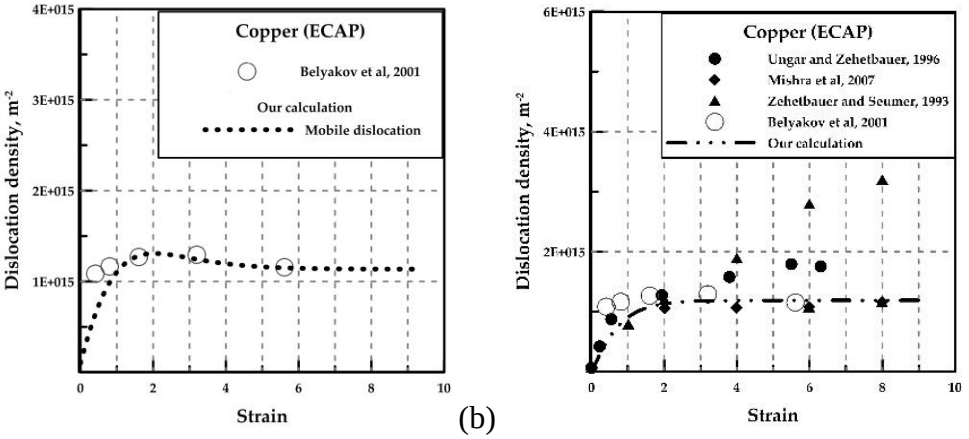



**Fig. 1.** Dislocation density of mobile (a) and immobile dislocations as a function of strain. The dots stand for the number of experimental data from [8, 9] for copper.

Numerical simulation of ECAP process using dislocation density based models will require a set of model parameters. The following values of the model parameters were used to compute data: $k_\alpha = 5.7$, $k_\alpha^i = 7$, $k_\alpha^m = 4$, $\delta_0$=0.02, $\eta = 0.1$, $\varepsilon_L / b = 8\text{eV}$, $V_C = 2.5 \cdot 10^{-6}$ m/s, $\rho_D^0 = 10^{13}\text{m}^{-2}$, $\rho_D^{mob} = 10^{13}\text{m}^{-2}$, $\rho_D^{im} = 10^{14}\text{m}^{-2}$. New plasticity models were implemented as USERMAT FORTRAN subroutine for ANSYS commercial finite element software. The subroutine is executed during element results calculation and provides a possibility to update local dislocation densities in every point of the deformed sample. Simulation results provided for analysis include sample shape for all simulation substeps along with the history of all state variables for all the sample points. Experimental geometry was chosen to comply with the experiments [6] (see Fig. 2a).

### 3. Continuous dynamic recrystallization model

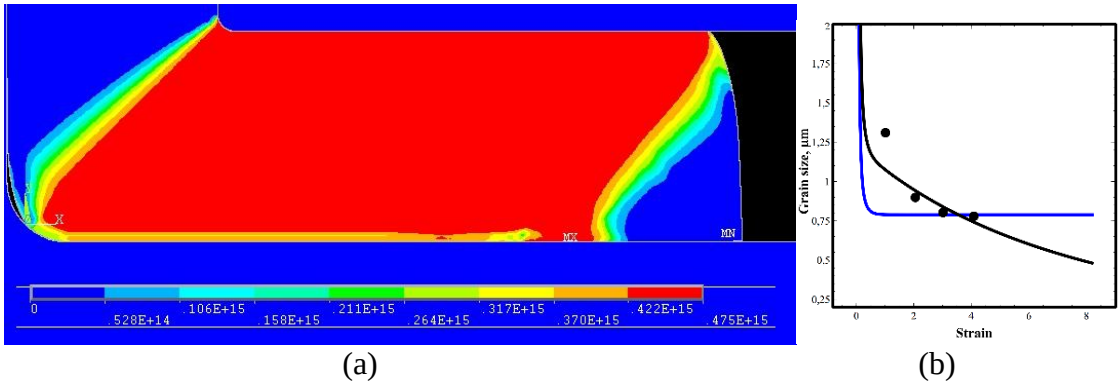
Dislocation density and grain size can be coupled once an assumption about dynamic recrystallization (DRX). Dynamic recrystallization is the result of dislocation substructure evolution and dislocation cell diameter as a function of dislocation density can be received as [10]:

$$d = \beta \rho_D^{-1/2}, \tag{2}$$

Here $\beta$ is a coefficient dependent on dislocation cell geometry. It can be expected that (2) should give a mean value for a size of grain appearing in material after intensive plastic deformation. The problem is that the model (2) of recrystallization is correct for equilibrium processes, which is not true for SDP processes. This approach does not give a possibility to predict evolution of grain structure from large grains to ultrafine grains (UFG) under ECAP. Due to this reason, in [10] authors introduce a dependency for $\beta$: $\beta = \beta_\infty + (\beta_0 - \beta_\infty)\exp(-\gamma\varepsilon)$, which results in appearance of three additional fitting parameters. An alternative approach can be proposed should one suppose the degree of defect structure nonequilibrium to be an important factor controlling the process of dynamic recrystallization. In [7] volume fraction of consolidated triple junctions $J$ was introduced, then the mean distance between the elements of stable structure will be given by $d = \beta \rho_D^{-1/2} / J$. Supposing, for simplicity, that the rate of nonequilibrium grain boundaries fraction change is proportional to the ratio between mobile to immobile dislocations, the following form of recrystallization law can be proposed [7]:

$$\begin{cases} J = p^3, \\ \dot{p} = \alpha[V_C(\rho_D^{mob} - \rho_D^0)/\rho_D^{im}], \end{cases} \tag{3}$$

where parameter $p$ is the fraction of nonequilibrium boundaries of dislocation cells in dislocation structure. The approach is shown to be applicable to predict experimentally observed correlation between the accumulated strain and the observed resultant grain size (see Fig. 2b). The best coincidence between the studied experimental data and predictions given by (3) is achieved for $\alpha$ = 0.083.




(a) (b)

**Fig. 2.** (a) Distribution of total dislocation density and (b) DRX curve: the black line corresponds to calculation by (5) with $\beta = 2.6$. The blue line corresponds to calculation using (2) with $\beta = 21$. The circles correspond to the experimental data from [10].

## 4. Conclusions

It is shown that in many cases the simplest classical model can be used in order to predict a defect hardening of the deformed material. Additional features for dislocation density based models given by separation of dislocations into mobile and immobile [2, 3] make the prediction of defect structure evolution more realistic, providing a possibility to account the changes in material fine structure. New approach for DRX has been proposed. Among the advantages of this approach is the introduction of a single additional parameter (to be found from correspondence to experimental data) controlling DRX process instead of three fitting parameters introduced in another [10] known approach to the problem while the prediction accuracy is not reduced.

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**References**

[1] G.A. Malygin // *Physics-Uspekhi* **169** (1999) 979.
[2] A.E. Mayer, K.V. Khishchenko, P.R. Levashov, P.N. Mayer // *Journal of Applied Physics* **113** (2013) 193508.
[3] V.S. Krasnikov, A.E. Mayer, A.P. Yalovets // *International Journal of Plasticity* **27** (2011) 1294.
[4] Y. Estrin, L.S. Toth, A. Molinari, Y. Brechet // *Acta Materialia* **46** (1998) 5509.
[5] S.C. Baik // *Materials Science & Engineering A* **351** (2003) 86.
[6] A.E. Dudorov, A.E. Mayer // *Bulletin of Chelyabinsk State University. Physics* **39(19)** (2011) 48. (In Russian).
[7] V. Bratov, E.N. Borodin // *Materials Science & Engineering A* **631** (2015) 10.
[8] A. Belyakov, T. Sakai, H. Miura, R. Kaibyshev // *Philosophical Magazine Letters* **80** (2000) 711.
[9] A. Mishra, B.K. Kad, F. Gregori, M.A. Meyers // *Acta Materialia* **55** (2007) 13.
[10] R. Lapovok, F.H. Dalla Torre, J. Sandlin, C.H.J. Davies, E.V. Pereloma, P.F. Thomson, Y. Estrin // *Journal of the Mechanics and Physics of Solids* **53** (2005) 729.