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# Modeling of Deformation Behavior and Dynamic Strain Ageing in ECAP

M. Vaseghi[a,*]

[a]*Assistant Professor, Faculty of Mechanical and Energy Engineering, Shahid Beheshti University, A.C., Tehran, Iran*

**Abstract**

Equal channel angular pressing (ECAP) is a substantial process for producing ultra-fine grains in bulk metallic materials by means of severe plastic deformation. In this study, microstructure evolution through dynamic ageing characteristics associated with the application of ECAP to Al6061 alloy at high temperature was investigated. The behavior of the material under ECAP, including the dislocation density and cell size evolution as well as precipitation development, was modelled using dislocation density-based model. Followed by ECAP, microstructural observations were undertaken using EBSD. The experimental investigations showed that after two pass of ECAP, a large amount of subgrains with low angle grain boundaries appears in the original coarse grains. The simulated cell size was in good agreement with the experiment, particularly with the observed rapid decrease of the cell size during the second pass slowing down from the forth pass onwards.



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*Keywords:* equal channel angular processing; microstructure evolution; dislocation; modelling.

## 1. Introduction

After three decades, ultra-fine-grained materials are attracting a great deal of consideration, Latypov et al. (2014). Of particular interest is grain refining by severe plastic deformation and other strengthening mechanisms that makes is possible to produce bulk materials with enhanced mechanical properties. A number of techniques were developed; by which metallic alloy can be deformed to extremely large plastic strains with negligible changes in their first and

* Corresponding author. Tel.: +98-21-7393-2694; fax: +98-21-7731-1446.
*E-mail address:* m_vaseghi@sbu.ac.ir


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last shapes, Vaseghi et al. (2010). Equal channel angular extrusion (ECAE) is considered to be the most promising for production of bulk, relatively uniform ultra-fine-structured materials, Baik et al. (2003).

Extremely large strains when add to precipitation hardening throughout dynamic strain ageing results incredible strength and hardness. The effect of die shape (cross section, channel angle, inner and outer curvature, the number of inlet and outlet channels and the length of them), the number of passes, the role of specimen rotation between consecutive passes, the ram speed as well as the working temperature were also investigated by researchers, previously. While SPD uses grain size strengthening and strain hardening of the processed materials for strengthening mechanisms, there is another strengthening mechanism involved in heat treatable Al alloys, i.e. precipitation hardening or age hardening. In order to estimate the magnitude of the strain introduced around the precipitates and to understand the deformation behavior in this complex process, a few attempt were performed.

However, none of the previous analyses have taken into account the strain hardening of the material by considering the microstructure evolution during ECAP. Indeed, from one hand, grain refinement by severe plastic deformation is related to the evolution of subgrain (dislocation cell) boundaries and on the other hand, by development the cell structure, the strain ageing phenomena takes more likely place. Hence, to understand the microstructure refinement and its effect on strain ageing behavior during ECAP, it is necessary to analyze the dislocation density evolution and the variation of the dislocation cell size during ECAP. Estrin et al. (1998) proposed the model that predicts the strain-hardening behavior of dislocation cell-forming crystalline materials at large strains in the two-dimensional cell structure and next the three-dimensional case, Toth et al. (2002). The aim of this study is to analyze the microstructure evolution of polycrystalline aluminum alloy during hot temperature ECAP in connection with the three-dimensional version of the dislocation density-based model. The outcomes include a description of the dislocation density evolution and the cell size variation, combined with dynamic strain ageing.

## 2. Dislocation density-based strain-hardening model

Here a brief review of three-dimensional version of the dislocation density-based strain-hardening model which was developed by Toth et al. (2002) is given and used in the present simulations. In this model a dislocation model previously developed to describe the deformation behavior of aluminum under equal channel angular pressing is applied to dynamical strain ageing during ECAE in Al6061 alloy. By evaluating certain of the parameters introduced in this model from experimentally determined information at various temperatures in the DSA range, a good quantitative physical interpretation of the dynamical strain ageing process in the alloy is obtained. This approach will combined with the dynamic strain ageing considerations, thus making it possible to simultaneously trace strain hardening and microstructure evolution. The dislocation population is partitioned into dislocations forming a cell structure and those contained within the cell interiors. This gives rise to the notion of a ‘three-phase material’: a cell interior with a relatively low dislocation density $\rho_c$, a precipitate-matrix interface with intermediate dislocation density $\rho_i$ and cell walls of width w with a higher dislocation density $\rho_w$. These three distinct dislocation densities are the internal variables of the model.

Increase of work hardening rate is due to an enhancement of the dislocation multiplication rate as a result of dislocation locking by precipitates and the subsequent generation of new mobile dislocations is the predominant processes dictating the high rate of strain hardening. It is necessary to introduce three main parameters in order to describe quantitatively the dislocation processes mentioned above. The first of these, U, is the rate of immobilization or annihilation of mobile dislocations in cell walls; the second, Q, measures the probability for generation of mobile dislocations; and the third parameter, A, is the rate at which mobile dislocations lock in the precipitate-matrix interface or annihilate in specimen surfaces, etc. Bergstrom and Roberts (1971). The definition of the parameter U implies that it is determined by the mean free path of mobile dislocations. Provided the assumptions stated are realized in practice, the variation of total dislocation density with strain is

$$\frac{d\rho}{d\varepsilon} = (U - A) + Q\rho \tag{1}$$

The total dislocation density, $\rho$, is given by a rule of mixtures, Toth et al. (2002):

$$\rho = f\rho_w + (1 - f)\rho_c \tag{2}$$

where $f$ denotes the volume fraction of the cell walls and $\rho$ is composed of a mobile dislocation density ($\rho_m$) and an immobile density ($\rho_i$). The variation of $\rho_i$ with $\varepsilon$ is determined by the dislocation generation, the immobilization of dislocations in cell walls and immobilization of dislocations in the precipitate-matrix interface; and any effect due to changes in dislocation structure is negligible.

By integrating of eq. (1) with the boundary condition $\rho = \rho_0$ at $\varepsilon = 0$ and the result is equal to eq. (2) gives

$$\rho = f\rho_w + (1-f)\rho_c = \frac{U-A}{Q}\left(1-e^{-Q\varepsilon}\right) + \rho_0 e^{-Q\varepsilon} \tag{3}$$

An important element of the model is the consideration of the evolution of the volume fraction of the cell walls, and it is found that f decreases with strain monotonically, Muller et al. (1996). The evolution of $f$ was approximated by Estrin et al. (1998) by the following empirical function:

$$f = f_\infty + (f_0 - f_\infty)\exp\left(\frac{-\gamma^r}{\tilde{\gamma}^r}\right) \tag{4}$$

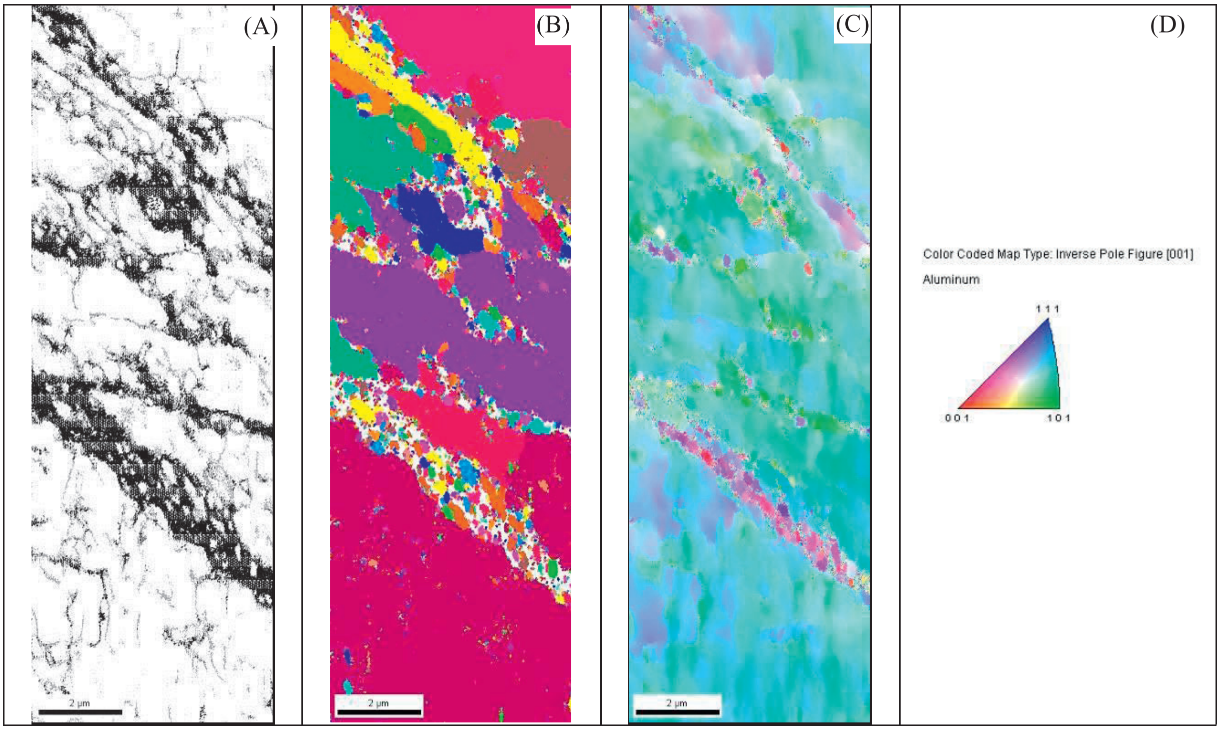


Fig. 1. (A) the boundaries map; (B) orientation imaging microscopy; (C) inverse pole figure obtained from the transverse section of the sample after two pass ECAP at 200 ◦C; (D) standard stereographic projection.

where $f_0$ is the initial value of $f$, $f_\infty$ its saturation value at large strains and the quantity $\tilde{\gamma}^r$ describes the rate of variation of $f$ with resolved shear strain $\gamma^r$. The fact that $f_\infty$ is significantly smaller than $f_0$ implies that the subgrain walls become sharper in the course of straining. The average cell size $d$ is directly related to the total dislocation density through:

$$d = \frac{K}{\sqrt{\rho}} \tag{5}$$

where $K$ is a proportionality constant and also decreases with the accumulation of the total dislocation density in the course of ECAP. It is well established that the flow stress is related to the total dislocation density $\rho$ as:

$$\sigma = \sigma_0 + \alpha G b \left( \frac{U - A}{Q} \left(1 - e^{-Q\varepsilon}\right) + \rho e^{-Q\varepsilon} \right)^{\frac{1}{2}} \tag{6}$$

At hot temperatures, the solute diffusivity is too low for any interaction with mobile dislocations and the parameters U and A will be independent of strain in the range $\varepsilon_{luders} \le \varepsilon \le \varepsilon_{necking}$ necking. The grain size strengthening mechanism is generally interpreted by pile-up of dislocations and the effectiveness of grain boundaries in strengthening will depend on the misorientation angle between them. The stress concentration by the pile-up dislocation at the subgrain/grain boundaries leads to the dislocation activated in the neighboring grain. A smaller misorientation angle between the two grains makes it easier to transfer the dislocation movement to the neighbor grain at the same slip system. So the subgrain boundary strengthening effect is smaller than the grain boundary.

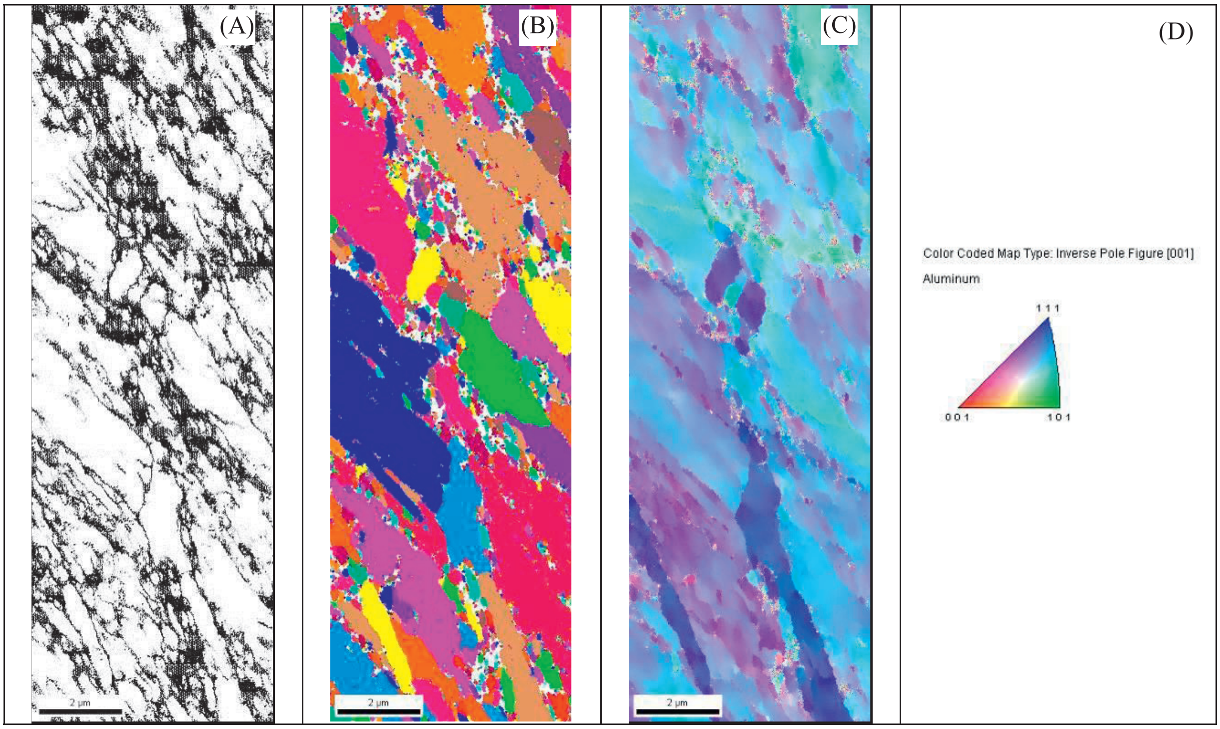


Fig. 2. (A) the boundaries map; (B) orientation imaging microscopy; (C) inverse pole figure obtained from the transverse section of the sample after four pass ECAP at 200 ◦C; (D) standard stereographic projection.

To verify the suggestion about misorientation angle between grains and/or subgrains one might attempt measurements through electron backscattered diffraction (EBSD). A better understanding of superposition laws would still be welcome in situations where a large number of different obstacles to dislocations are present (forest dislocations, several types of precipitate phases, small grains). In the some zone where the amount of deformation reaches to critical level, one can expect that near-full suctioning is achieved during ECAP processing. Also, substantial strength recovery is possible in these zones due to reprecipitation of low-temperature phases such as co-clusters/GP zones. As deformation proceeds, LAGBs gradually transform into HAGBs and are rendered indistinguishable from original grain boundaries, Vaseghi et al. (2014). Typical microstructures are illustrated in the form of EBSD maps in Figs. 1 and 2; from (A) the boundaries map, (B) orientation imaging microscopy, (C) inverse pole figure of the ECAPed sample at 200 ◦C and (D) standard stereographic projection. Black lines indicate high angle boundaries (⩾15◦), gray lines indicate low angle boundaries (15◦ >θ> 2◦). After two pass, an approximate lamellar structure is observed, where the grain aspect ratio is typically greater than 2. The microstructure consists of a mixture of high angle boundaries, as well as dislocation boundaries of low misorientation. After treating more passes the microstructure has refined and become more equiaxed. The result could be concluded from Fig.3 witch shows the misorientation angles calculated by the model.

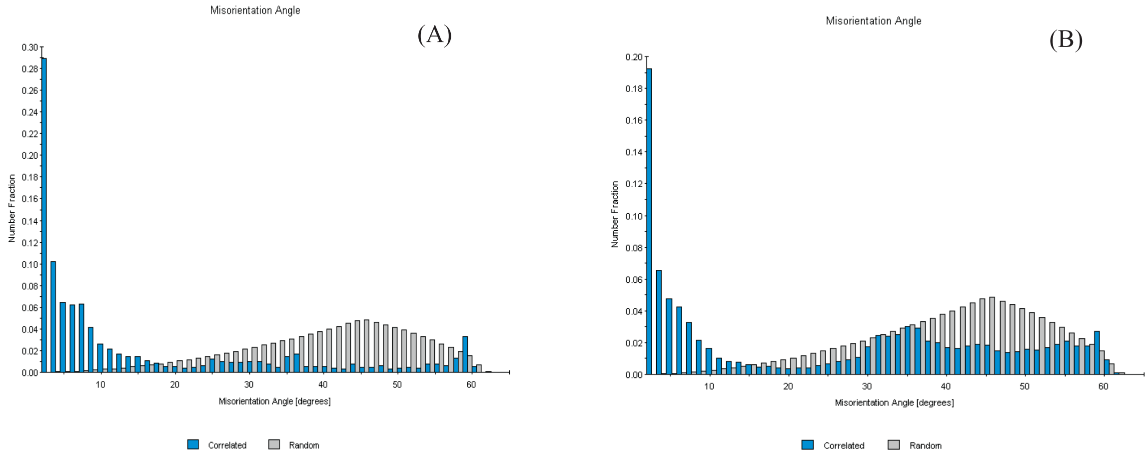


Fig. 3. Change in the misorientation distribution after (A) two; (B) four passes of ECAPed sample.

## 3. Conclusions

The microstructure evolution through dynamic ageing characteristics associated with the application of ECAP to Al6061 alloy at high temperature was investigated. The model indicated that the strength contributed by dislocation density is much higher than the strength contributed by grain size in the ECAPed billet. The simulated cell size was in good agreement with the experiment, particularly with the observed rapid decrease of the cell size during the second pass slowing down from the forth pass onwards.

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