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Acoustic energy enabled dynamic recovery in aluminium and its effects on stress evolution and post-deformation microstructure

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ABSTRACT

It is now well established that simultaneous application of acoustic energy during deformation results in lowering of stresses required for plastic deformation. This phenomenon of acoustic softening has been used in several manufacturing processes, but there is no consensus on the exact physics governing the phenomenon. To further the understanding of the process physics, in this manuscript, after-deformation microstructure of aluminium samples deformed with simultaneous application of kilohertz range acoustic energy was studied using Electron-Backscatter Diffraction analysis. The microstructure shows evidence of acoustic energy enabled dynamic recovery. It is found that the subgrain sizes increase with an increase in acoustic energy density applied during deformation. A modified Kocks-Mecking (KM) model for crystal plasticity has been used to account for the observed acoustic energy enabled dynamic recovery. Using the modified KM model, predicted stress versus strain curves were plotted and compared with experimental results. Good agreements were found between predictions and experimental results. The manuscript identifies an analogy between microstructure evolution in hot deformation and that in acoustic energy assisted deformation.

1. Introduction

Acoustic softening has been used to improve several manufacturing processes by taking advantage of the associated reduction in stresses required to achieve and sustain plastic deformation. One of the first explanations proposed for this observed reduction in yield and flow stresses of a metal during deformation due to simultaneous application of acoustic energy was presented by Langenecker in 1966 [1]. In his proposed model, the reduction in stresses was attributed to the preferential absorption of acoustic energy at the lattice defects, more specifically dislocations, resulting in a reduction in the stress required to move the dislocations [1,2].

More recently, Yang et al. used longitudinal-torsional composite ultrasonic vibration for titanium wire drawing and observed a reduction in drawing force and better surface finish [3]. Abdullah et al. used longitudinal ultrasonic vibrations to reduce the axial forming forces and improve the surface quality during indentation formation of tubes [4]. Similarly, ultrasonic vibrations were shown to increase formability and decrease forming forces during forming of AA1050 sheets by Amini et al. [5]. Ultrasonic vibration assistance was shown to improve weld formation in friction stir welding by Zhong et al. [6]. In the hybrid additive-subtractive manufacturing process of Ultrasonic Consolidation, acoustic softening has been observed and shown to have significant effects on the process parameters [7–9]. Wire bonding is another process which uses acoustic softening to deform metal wires along with an athermal ultrasonic assisted diffusion phenomenon to bond the wires to a substrate [2,10,11]. Though the applications of acoustic softening to several manufacturing processes is wide-spread, understanding of the effects of acoustic softening on metals has not reached maturity. It is, therefore, important, particularly in the context of their microstructure which affects the eventual material properties to gain deeper understanding of the governing physics, which can result in further innovations in manufacturing.

Several attempts have been made to model the behavior of metals during deformation under the influence of acoustic energy irradiation. Rusinko modified the synthetic theory of irreversible deformation to develop an analytical model of acoustic softening and residual hardening by introducing a new term termed ultrasonic defect density [12]. Siddiq and Sayed modified the porous plasticity model to account for the acoustic softening effects during ultrasonic manufacturing processes [13]. Yao et al. modified the single crystal plasticity frame work to model acoustic softening and residual hardening during upsetting experiments [14]. However, these models do not provide a full explanation of the phenomenon of acoustic softening. Further, these models provide no information on microstructure evolution during the deformation process with simultaneous application of acoustic energy.

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The study of microstructure evolution is imperative in that it can provide insights into the acoustic softening phenomenon, and enables the development of a constitutive model that accurately captures the stress evolution during a static deformation process with simultaneous acoustic energy irradiation.

In the work described here, characterization of microstructure using Electron Backscatter Diffraction (EBSD) analysis of the aluminium samples (fine wires) after compression has been carried out. The microstructure characterization shows evidence of athermal dynamic recovery similar to that observed during hot deformation. A model based on the one-internal-variable crystal plasticity model called the Kocks-Mecking model has been used to predict the effect of simultaneous acoustic energy irradiation on stress evolution during compression. The model is similar to that used by Yao et al., but here it is used with several key modifications to account for athermal acoustic energy enabled dynamic recovery. Discrepancies in the results with those of Yao et al. have been highlighted and explanations provided [14]. The results observed here are significant for processes like wire bonding in which a similar setup is used to deform and bond fine wires and also for manufacturing processes like ultrasonic assisted forming, wire drawing and consolidation which use the phenomenon of acoustic softening.

2. Plasticity model

2.1. Kocks-Mecking model for crystal plasticity for thermal energy induced stress reduction

During hot deformation, as material is strained, new dislocations are generated in the material resulting in a rise in the stress required for further deforming the material. This is referred to as strain hardening. At the same time, dislocations annihilate, or entangle into low energy arrays forming subgrain boundaries, due to climb of edge dislocations made possible by the increased mobility of atoms at higher temperatures and cross glide of screw dislocation. When the rate of dislocation annihilation equals the rate of generation, a steady state is reached and there is no further increase in the stress [15–17].

For deformation at higher temperatures and lower strain rates, the increased mobility of dislocations enables the annihilation to be higher. Therefore, a steady state is reached at lower overall dislocation densities, which results in larger subgrain sizes. On the other hand, for deformation at lower temperatures and higher strain rates, the generation of dislocations is faster due to which the steady state is reached at higher overall dislocation densities, and therefore smaller subgrain sizes. The Kocks-Mecking model embodies this phenomenon of dislocation density evolution by using a single internal variable dependent on dislocation density (Eq. (3)).

It relates the plastic strain rate $\dot{\gamma}^p$ to the shear stress τ through the kinetic equation given by [18,19]

$$\dot{\gamma}^p = \dot{\gamma}_o \exp\left(\frac{-\Delta G}{kT}\right) \tag{1}$$

$\dot{\gamma}_o$ is called the pre-exponential factor, and ΔG is the Gibbs free energy. Gibbs free energy is a function of the obstacle distribution. It is related to total free energy ΔF via the equation [18,19]

$$\Delta G = \Delta F\left(1 - \left(\frac{\tau}{\hat{\tau}}\right)^p\right)^q \tag{2}$$

p and q have values 3/4 and 4/3 respectively and $\Delta F = 0.5\mu b^3$ [19]. μ is the shear modulus and b is the burgers vector.

The KM model is based on a single internal variable $\hat{\tau}$ called mechanical threshold which depends on dislocation density ρ. The mechanical threshold is a demarcation between thermally activated flow and viscous glide; below the mechanical threshold, plastic flow is only due to thermal activation and above it due to rate sensitive viscous glide [20]. The relation between mechanical threshold $\hat{\tau}$ and ρ is given by [18,21,22]

$$\hat{\tau} = \alpha\mu b\sqrt{\rho} \tag{3}$$

where, α is a numeric constant.

The evolution of dislocation density with plastic strain is controlled by two terms. First is the dislocation storage term that causes athermal hardening. It is inversely proportional to the average spacing between dislocations, and therefore, directly proportional to $\sqrt{\rho}$. The second term is the dislocation annihilation term which accounts for the dynamic recovery due to the cross-slip of screw dislocations, and climb of edge dislocations. It is proportional to ρ [18,21,23]. For a detailed derivation for each term, readers are referred to [21].

$$\frac{d\rho}{d\gamma^p} = \frac{d\rho^+}{d\gamma^p} + \frac{d\rho^-}{d\gamma^p} = k_1\sqrt{\rho} - k_2\rho \tag{4}$$

where γ^p is the resolved shear strain in the slip plane. The coefficient for the dynamic recovery term k_2 is given by the equation [18,23]-

$$k_2 = k_{20}\left(\frac{\dot{\gamma}^p}{\dot{\gamma}_o^*}\right)^{-1/n} \tag{5}$$

where k_{20} is a numeric constant.k_2 is strain rate and temperature dependent. For low temperatures, n is inversely proportional to temperature T and $\dot{\gamma}_o^*$ is constant. However, at high temperatures, $\dot{\gamma}_o^*$ is given by Arrhenius equation [18,24]

$$\dot{\gamma}_o^* = \dot{\gamma}_{oo}^*\exp\left(\frac{-Q_d}{kT}\right) \tag{6}$$

where Q_d is activation energy for self-diffusion or dislocation climb, k is the Boltzmann constant and T is temperature. For high temperatures, n is constant between 3 and 5 [18,20,21]. The demarcation between low and high temperatures is usually defined as 2/3 of the melting temperature [24].

Finally, to relate shear stresses and strains in a single crystal material to macroscopic axial stresses and strains in polycrystalline materials, Taylor's factor is used.

$$M = \frac{\sigma}{\tau} = \frac{\gamma^p}{\varepsilon^p} \tag{7}$$

M is microstructure dependent. The evolution of M is assumed to be much slower than the evolution of dislocation density. M is, therefore, assumed to be constant [18].

2.2. Modification to KM model to account for acoustic softening

The after-deformation microstructure of the aluminium samples deformed under the influence of acoustic energy shows evidence of strain rate and athermal acoustic energy dependent dynamic recovery analogous to those observed in hot deformation. The details about the microstructure characterization results are provided in Results and Discussion section. To account for this acoustic energy induced dynamic recovery, Eq. (4) for $\dot{\gamma}_o^*$ is modified as follows-

$$\dot{\gamma}_o^* = \dot{\gamma}_{oo}^*\exp\left(\frac{-Q_d\chi}{kTE}\right) \tag{8}$$

E is the energy density in J/m^3, and is given by $E = a^2\omega^2\rho_{Al}$, where a is the amplitude of vibration, ω is the frequency in rad/s and ρ_{Al} is the density of aluminium. χ is a constant with units J/m^3.

The authors propose that the effect of acoustic energy here can be thought of as the minimum amount of acoustic energy density required to achieve acoustic softening. Increase in the amount of acoustic energy density used during deformation results in an increase in the value of $\dot{\gamma}_o^*$ in Eq. (6), which in turn causes increased in k_2 in Eq. (5). As k_2 changes, the dislocation density (therefore microstructure) evolution changes accordingly, as modeled in Eq. (4).