

Full length article

Controlling strength and ductility: Dislocation-based model of necking instability and its verification for ultrafine grain 316L steel

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ABSTRACT

A phenomenological dislocation-based approach is proposed to account for the necking phenomenon during tensile deformation of metals and alloys. The critical strain corresponding to the onset of tensile instability is predicted in a simple explicit form based on the Kocks-Mecking dislocation kinetics approach. The model strongly suggests that uniform elongation is controlled primarily by the rate of dislocation recovery. The role of the strain rate sensitivity in stabilizing uniform plastic flow is also elucidated. Model predictions are found to be in excellent agreement with experimental data obtained for ultrafine grained 316L steel produced by severe plastic deformation. The approach presented provides general cues for designing materials with enhanced ductility, including ultrafine grained and bulk nanostructured metals and alloys. The proposed recipe is based on microstructural control of the rate of dynamic recovery of dislocation.

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1. Introduction: motivation and goal

Most structural metals and alloys strained in tension fail by necking that sets in when a maximum on the engineering stress–strain curve is reached. This strain, which corresponds to the magnitude of uniform elongation, serves as a measure of ductility and is of great importance for many practical applications of structural materials. An excellent survey focussing on various aspects of the complexity of strain localization during plastic deformation of metals has been provided recently by Antolovich and Armstrong [1]. Controlling influences of temperature, strain rate, grain size and deformation mode on strain localization are evaluated and the critical role of dislocation mechanisms in strain localization is highlighted. The ductility issue is particularly acute for high strength materials, which usually possess a limited capability to deform uniformly. Particularly, bulk nanostructured or ultrafine grained (UFG) metals fall into this category of low ductility materials [2,3]. The analysis of a large body of experimental data

accumulated in numerous experimental studies [4] suggests that strength and ductility are mutually exclusive. Substantial efforts have therefore been invested in developing conceptual frameworks and methodologies for ductility enhancement [3,5–8]. The nature of premature strain localization leading to lower than desired uniform elongation is not fully understood from the microstructural viewpoint as yet. The most common explanation for the limited ductility of UFG materials is given in terms of the Considère instability criterion [9], which signifies the onset of necking at the point where the strain hardening coefficient h drops below the value of the flow stress σ for a given plastic strain rate $\dot{\epsilon}$. The critical condition reads:

$$h \equiv \left. \frac{\partial \sigma}{\partial \epsilon} \right|_{\dot{\epsilon}} = \sigma \quad (1)$$

This criterion is purely mechanistic [1,10], as it is based solely on the true stress vs. true strain relation. Therefore, understanding of the instability criterion in terms of the underlying microstructural processes is essential for control of microstructural characteristics aimed at delaying tensile necking thus improving ductility [11]. Despite the broad diversity of processing schedules employed

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for grain refinement by severe plastic deformation (SPD) and the ensuing microstructural states, the common trend is that a spectacular enhancement of strength upon SPD processing concurs with a considerable loss of ductility. As a general observation [3,4,12–14], it is fair to say that in UFG materials modified by SPD condition (1) is fulfilled already at quite low strains, Fig. 1. That is to say, the tremendous strain hardening achieved in the course of SPD is compromised by the attendant loss of uniform elongation ε_U . This is illustrated in Fig. 1, where the improvement of strength by equal-channel angular pressing (ECAP) is shown for the austenitic stainless 316L steel subjected to up to four passes through the die [15,16]. It is commonly believed that the main reason for the loss of ductility in UFG SPD materials is a combination of high flow stress and low strain hardening capability of SPD processed materials [14].

Although the ductility of a material is almost inevitably sacrificed after moderate pre-strain, e.g. after the very first ECAP pass, an appreciable improvement of ductility is often observed for giant pre-strains (e.g. with for the number of ECAP passes as large as 16–25) [12,17]. However, this improvement is admittedly modest. Significant improvement of ductility has, however, been documented in many cases after annealing of the UFG microstructure obtained by SPD. A bi-modal grain size distribution has been proven to be particularly conducive for enhancement of the post-ECAP ductility [6,18]. The improved ductility is commonly associated with relaxation of the internal stresses and reduction of excess dislocation density. However, this effect has not been rationalized in any quantitative modelling approach thus far.

2. Theoretical consideration: tensile necking criterion from the viewpoint of dislocation kinetics

Plastic deformation is obviously non-homogeneous at a dislocations scale. However, it appears to be reasonably homogeneous at a macroscopic scale until necking sets in, the cross-sectional area of the specimen shrinks, and the load-displacement curve begins to drop off. Virtually all instability criteria, including the Considère

one, Eq. (1), are based entirely on the constitutive relation between stress and strain, despite the fact that strain cannot be regarded as a state variable. Assuming that it is the dislocation density that can serve as such a “state variable” and describing a crystal under load as a dynamic system, Yasnikov et al. [19,20] have recently revisited this topic with an alternative approach. They considered the onset of tensile instability from the viewpoint of the evolution of dislocation ensembles during plastic deformation. A phenomenological approach developed by Kocks and Mecking [21] to account for the strain hardening behaviour of materials (cf. [22–25]) was extended further towards the point of macroscopic instability. This approach is based on two coupled equations. The first one is the Taylor relation between the shear stress τ for the operating slip system and the dislocation density ρ :

$$\tau = (\alpha_0 G b \sqrt{\rho}) \left(\frac{\dot{\gamma}}{\dot{\gamma}_0} \right)^{\frac{1}{m}} \quad (2)$$

where $\dot{\gamma} = d\gamma/dt$ is the plastic shear strain rate, G is the shear modulus, b the magnitude of the dislocation Burgers vector, α_0 and $\dot{\gamma}_0$ are material parameters. The temperature dependence is associated with thermally activated dislocation glide and resides in the parameters $\dot{\gamma}_0$ or m (where m is the inverse of the strain rate sensitivity of the flow stress). The scaling law prescribed by Eq. (2) holds quite generally regardless of the geometrical arrangement or the detail of the dislocation–dislocation interaction mechanism. A further equation, which is necessary to describe the evolution of the total dislocation density as a function of time in the process of straining, is given by

$$\frac{d\rho}{dt} = (k_1 \sqrt{\rho} - k_2 \rho) \dot{\gamma} \quad (3)$$

Here k_1 is a constant accounting for dislocation storage due to interaction of gliding dislocations with the forest ones or the dislocation cell walls; the coefficient k_2 is associated with the dynamic recovery leading to a loss of dislocation density due to dislocation annihilation. Accordingly, k_2 is strain-rate and

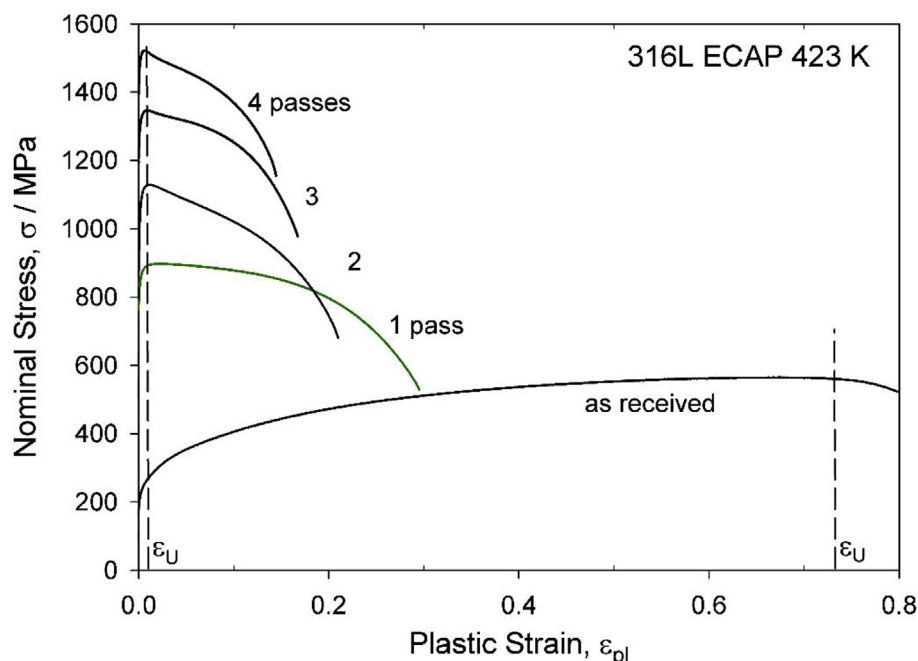


Fig. 1. Tensile stress–strain curves for 316L steel subjected to 0–4 ECAP passes performed at 423 K (adapted from Ref. [16]).

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