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The role of mixed-mode deformation at the crack tip on shear banding and crack propagation of ultrafine-grained copper

M. Goto [a,*], S.Z. Han [b], J.H. Ahn [b], T. Yakushiji [c], K. Euh [b], S.S. Kim [d], J. Lee [e]

[a] Department of Mechanical Engineering, Oita University, Oita 870-1192, Japan
[b] Structural Materials Division, Korea Institute of Materials Science, Changwon 642-831, Republic of Korea
[c] Department of Mechanical Engineering, Oita National College of Technology, Oita 870-1052, Japan
[d] Department of Materials Science & Engineering, Gyeongsang National University, Chinju 660-701, Republic of Korea
[e] Department of Metallurgy & Advanced Materials Engineering, Chagwon National University, Changwon 641-773, Republic of Korea

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ABSTRACT

Stress-controlled fatigue tests of ultrafine-grained copper round-bar specimens were conducted. Although the crack paths inclined 45° and 90° to the loading-axis were observed in the different locations, crack faces were extended along one set of maximum shear-stress planes, corresponding to the final ECAP shear plane. Profile of crack faces was examined, showing the aspect ratios of 0.38 and 1.10 for the cracks with 45° and 90° inclined path directions with respect to the loading-axis, respectively. The role of deformation mode at the crack tip on shear-banding and crack growth behavior were discussed in terms of the mixed-mode stress intensity factor.

1. Introduction

Over the past few decades, extensive studies have been made on the fatigue of ultrafine-grained (UFG) metals and alloys processed by equal channel angular pressing (ECAP). On the surface of cyclically deformed UFG metals, shear bands (SBs) extend over a much larger distance than the UFG grain size that is usually formed [1,2]. The SBs represent sites of severe cyclic strain localization in which fatigue cracks generate. Regarding SB formation in UFG metals processed by severe plastic deformation techniques, many investigators have observed morphological features and microstructure of SBs, discussing the formation mechanism of SBs. In cyclically deformed UFG Cu, Agnew and Weertman [3] observed extrusions and intrusions reminiscent of persistence slip bands, which are similar to those known from studies on Cu single crystals. Mughrabi and Höppel [4] observed two kinds of strain localization, in patches and in extended SBs in post-fatigued UFG copper, suggesting two possible mechanisms of SBs formation: (1) spreading of small regions/patches of local grain/subgrain coarsening, forming SBs; and (2) catastrophic shear localization, due to the strain path change from ECAP to fatigue, followed by the formation of new coarsened bands of localized shear deformation. These authors later suggested that the mechanism responsible for the SB formation was an interaction of cyclically induced cell/grain coarsening, which led to strain localization [5]. It was shown that the SBs orient at 45° with regard to the cyclic loading axis [1,6–11]. This means that the SBs are formed parallel to the plane of maximum resolved shear stress. Wu et al. [6,12] suggested that SBs in the cyclically deformed UFG copper developed inside the SBs that formed during the last pressing pass of ECAP, and that the SB formation was due to the orientated distribution of defects along the shear plane of the last pass. In Fe–36Ni alloy, the same conclusions regarding the role of shear plane of the last ECAP-pass on SB formation were shown [13]. Through electron backscatter diffraction (EBSD) analyses, several investigators observed no grain coarsening in cyclically deformed UFG copper (99.9% Cu [9], 99.8% Cu [14]), and indicated that the cooperative grain boundary (GB) sliding along the shear plane of the last ECAP pass could be the decisive mechanism in SB formation [9,15]. The importance of "precursor" SBs for the formation of fatigue-induced SBs was indicated for UFG Cu produced by accumulative roll bonding [16], and for UFG Al produced by cryorolling [17,18]; shear banding could be the result of the reactivation of pre-existent SBs in pre-worked UFG materials. However, clear evidence for a SB formation mechanism is still lacking.

* Corresponding author. Tel.: +81 97 554 7772; fax: +81 97 554 7764.
E-mail address: masagoto@oita-u.ac.jp (M. Goto).

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It has been shown that the low-cycle fatigue (LCF) life in a conventional grain-sized steel was almost completely dominated by the growth life of a single crack [19]. In UFG copper fatigued at high stress amplitudes corresponding to LCF regime, like the conventional grain-sized steel, the crack growth life accounted for over 90% of the fatigue life of plain specimens [20]. Therefore, the fatigue crack growth behavior should be clarified for the design of safe machine components and structures. Recently, in the high-cycle fatigue tests, the growth behaviors of long (millimeter-range) cracks in UFG metals were studied for compact-tension [21–24], single edge-notched [25–27], single edge bend [28] and center-cracked tensile specimens [29]. With regard to the growth characteristics of long fatigue cracks in UFG materials such as aluminum alloys, carbon steels, and titanium, the higher growth rates in low and medium values of the stress intensity factor (SIF) range and the lower growth thresholds have been reported. These phenomena appear to be attributed to the weakened roughness-induced crack closure caused by the much smoother fracture surface and lower deflection of the crack path. The smoother crack-path/fracture-surface is caused by decreased grain sizes and limited crack tip plasticity. Grain refinement and strain hardening limiting the dislocation mobility inevitably result in suppression of the crack tip blunting, therefore leading to faster crack propagation. Recent examples of the long-crack growth behavior from literature have been overviewed by Estrin and Vinogradov [30]. Meanwhile, in stain-controlled LCF tests and stress-controlled fatigue tests at high stress amplitudes corresponding to LCF regime, study on fatigue crack growth mechanism has been relatively rare, and only a few reports can be found [31–33]. In addition, since LCF cracks are generally initiated in and propagated along SBs, to study the role of SBs formed adjacent to the crack tip region on the growth behavior should be meaningful for investigation of the crack growth mechanism of UFG materials. However, little has been discussed about the effect of SBs on the crack growth behavior.

Up to now, SB formation and LCF crack growth behavior of UFG materials have been mainly discussed from the viewpoints of microstructure and morphological features of surface damage. On the other hand, the discussion from the mechanical viewpoints should be done for a better understanding of the fatigue damage of UFG materials. However, such studies are few and certain questions remain unanswered. The objective of this study is to investigate the physical background of the formation mechanism of crack growth paths and the effect of shear banding on the crack growth behavior in terms of the mixed-mode deformation at the crack tip.

2. Experimental procedures

The material used was pure oxygen-free copper (99.99 wt% Cu). Before ECAP, the materials were annealed at 500 °C for 1 h (grain size: about 100 μm). Fig. 1a shows a schematic of the ECAP die used in this investigation and the orientation of fatigue specimens relative to the final pressing direction. The die had a 90° angle between intersecting channels. The angles at the inner and outer corners of the channel intersection were 90° and 45°, respectively. Repetitive ECAP was accomplished according to the Bc route (after each pressing, the billet bar was rotated 90° around its longitudinal axis). Each rod was subjected to eight sequential passes of pressing at room temperature. The pre-ECAP mechanical properties were 232 MPa tensile strength, 65% elongation, and Vickers hardness of 63 (load: 2.9 N). After eight ECAP passages, the properties changed to 438 MPa, 28%, and 141, respectively.

Analysis [34] has shown that the equivalent strain, ε, after one pass is given by the following relationship:

$$\varepsilon = \frac{1}{\sqrt{3}}\left\{2\cot\left(\frac{\Phi}{2}+\frac{\Psi}{2}\right) + \Psi\operatorname{cosec}\left(\frac{\Phi}{2}+\frac{\Psi}{2}\right)\right\} \quad (1)$$

Eight extrusion passes resulted in an equivalent strain of about 7.8. In the earlier analysis of Segal [35], the calculation was conducted assuming a well-lubricated billet with a square-cut die (Φ = 90°, Ψ = 0°). In practice, problems with friction at the die walls may be avoided by the use of appropriate lubricants. Under this condition, the billet moves as a ridged body and, apart from small end-zones, undergoes simple shear in a thin layer as it moves from the first channel into the second. It was shown through experimental measurements and simple finite element analysis for a die with a sharp outer corner, that the die makes it difficult to completely fill the die corner [36]. Kim et al. [37] suggested that the outer corner angle used for the prediction of strain should be an arc curvature of the workpiece, not the die corner angle Ψ because of gap formation between the workpiece and the die. This problem has been avoided in some experiments by using a die with a round outer corner [38]. Two-dimensional finite element analysis on the plastic deformation behavior of the materials during ECAP process with a round die corner and a frictionless condition was conducted [39], showing that the lesser shear zone in the outer part of the workpiece occurs due to the faster flow of the outer part compared with the inner part within the main deformation zone (area ABC in Fig. 1). The calculated average effective strain from finite element method (FEM) was 0.82, which was lower than the theoretical value, ε = 0.907, calculated using Eq. (1). This lower value is attributed to the formation of the lesser deformation zone in the outer part. If the lesser deformation zone is ignored, the average value from FEM is in good agreement with the theoretical value. The effective strain distribution by FEM [37] along the width direction of the workpiece (15 mm diameter , Φ = Ψ = 90°) normal to the pressing direction showed that the lesser deformation occurred within the zone 5 mm away from the bottom (outer part). Therefore, it is likely that the effective strain of the present specimen (5 mm diameter) cut from the inner parts of ECAPed workpiece (10 mm diameter) is close to the theoretical value, nevertheless the inhomogeneous microstructures were produced by 8 passes through the die. This point will be referred to later in Section 4. Accordingly, authors believe that the outer corner curvature in the die has negligible effect on the mechanistic discussion of the present study. Here, "shear plane" is defined approximately by a red dotted line illustrated in Fig. 1a.

Fatigue specimens (Fig. 1b) 5 mm in diameter were machined from their respective processed bars. Although the specimens had shallow circumferential notches (20-mm notch radius and 0.25-mm notch depth), the fatigue strength reduction factor for this geometry was close to 1, meaning that they could be considered plain. The fatigue specimens were electrolytically polished (approximately ≈ 25 μm from the surface layer) prior to mechanical testing to remove any preparation-affected surface layer. Polishing was carried out at 25 °C using an electrolyte consisting of 600 mL of phosphoric acid, 300 mL of distilled water, and 100 mL of sulfuric acid. Prior to testing, a small blind hole (both diameter and depth of 0.1 mm; Fig. 1c) was drilled as a crack starter on the middle surfaces of the plain specimens.

All fatigue tests were performed at room temperature using a rotating–bending (R–B) fatigue machine (constant bending-moment type) operating at 50 Hz. The fatigue damage on the specimen surface and the fracture surface was observed using an optical microscope (OM) and a scanning electron microscope (SEM). The crack length, l, was measured along the circumferential direction of the surface using a plastic replication technique. The stress value referred to is that of the nominal stress amplitude, σ_a, at the minimum cross-section (5-mm diameter).

The cross-section perpendicular to the press direction was observed in EBSD analyses. The specimens were ground using silicon carbide paper and polished with polycrystalline 3- and 1-μm diamond suspension. Final polishing was performed using