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Strain localization and fatigue crack initiation in ultrafine-grained copper in high- and giga-cycle region

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ABSTRACT

Initiation of fatigue cracks and early crack propagation in high-cycle and giga-cycle region in ultrafine-grained copper prepared by equal channel angular pressing was experimentally investigated. The cyclic slip localization takes place in cyclic slip bands whose length substantially exceeds the grain size. The mechanism of the crack initiation under load controlled cycling does not require the dynamic grain coarsening often indicated in literature. Fatigue cracks initiate in slip bands, which develop predominantly in zones of near-by oriented grains. The damage mechanism consists in localized slip resulting in development of cavities and voids arranged along the planes of highest cyclic shear stress and in formation of surface relief. A process of growth and linking of cavities and voids produced by the irreversible cyclic slip by dislocation movement generating point defects governs the early stage of the development of fatigue cracks. Sufficiently large cracks created by this mechanism and lying in suitably oriented long slip bands finally transform into fatigue cracks propagating by common opening mode with a plastic zone generated at their tip.

1. Introduction

Mechanisms of fatigue crack initiation and subsequent crack propagation has been a matter of intensive research in the past. For conventionally grained (CG) materials the comprehensive survey of basic knowledge on crack initiation, early crack propagation and propagation of long cracks can be found in many textbooks, e.g. [1–3]. Historically, the great deal of the pool of basic knowledge on fatigue damage mechanisms, development of dislocation structures, cyclic plastic strain localization, crack initiation and its relation to microstructure have been conducted on f.c.c. materials, particularly on Cu. The same holds for the pioneering investigation of fatigue performance of UFG material processed by severe plastic deformation [4,5]. Cu belongs from this point of view to the most thoroughly investigated materials. This is also the reason why it was chosen for investigation in this study.

The simple general rule indicates that the value of fatigue or endurance limit for most CG steels and copper alloys is 35–50% of the ultimate tensile strength σ_{UTS} [2]. This signalizes promising fatigue resistance of UFG materials and makes them attractive for engineering applications. Indeed, the fatigue lives of UFG specimens were generally found to be longer than those of CG specimens when fatigue tests were conducted under stress control and expressed in terms of S–N plots [6]. UFG Cu prepared by equal channel angular pressing (ECAP) with σ_{UTS} = 387 MPa exhibits under stress-controlled fatigue loading endurance limit of 150 MPa (on the basis of 10^8 cycles) [7]. However, the rule, relating the fatigue limit to the σ_{UTS} is sometimes violated in UFG materials. For instance, for the UFG Cu having σ_{UTS} 390 or 420 MPa (in dependence on the details of the ECAP procedure and related UFG structure), the experimentally determined endurance limit was only 80 MPa [8], which is very close to the value characteristic for an annealed CG copper. These substantially unmatched results were recently discussed, e.g. in [6,7]. The explanation is currently sought in the stability of the UFG structure under fatigue loading and, consequently, in the details of the fatigue crack initiation and early crack propagation. Nonetheless, the up-to now available knowledge is not sufficient to explain the observed effects consistently.

Fatigue limit of CG easy cross-slip materials, like Cu and Al under constant stress amplitude loading depends only very weakly on the grain size [9]. A weak decrease of the endurance limit (based on 10^9 cycles) expressed in terms of the total strain amplitude with increasing grain size was reported in [10]. Generally, it can be summarized that the fatigue life curves expressed both as S–N curves or dependences of number of cycles to failure on the total strain amplitude depend on the grain size insignificantly. This holds especially for endurance limits defined for 10^7 cycles [9]. Based

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on results published e.g. in [9–11] it can be concluded, that the fatigue strength of Cu in high-cycle region is nearly insensitive to the grain size ranging from 3 to 1200 μm. The explanation of this insensitivity to the grain size utilizes the idea, that the dislocation cell structure, which develops in CG materials due to fatigue, masks the effect of the grain size [9]. UFG materials prepared by ECAP also exhibit a cell structure. However, it differs in details from that developed in Cu under fatigue loading. The UFG material reveals substructural features – subgrains, dislocation cells and X-ray coherent diffraction regions. The structure bears traces of the ECAP process. The cells are often elongated in the shear direction along the last ECAP plane. The details of the structure are dependent on the ECAP route; the Bc route yields the structure of highest equiaxiality. Further, the scatter of the cell size is often substantial. Quantitative determination of grain or cell size of materials processed by ECAP is complicated by the fact that the size is varying broadly between hundreds of nanometers and some microns and by not well-defined boundaries in transmission electron microscopy (TEM) or electron back scattering diffraction (EBSD) images. The mutual orientation of structural units cannot be satisfactorily described as homogeneous high-angle random orientation. That is why the term grain size is often replaced by the term cell size to point out that there are areas containing crystallites with very similar orientation. There are regions where the low angle boundaries are frequent. These regions can be described as "zones of near-by oriented grains" and they seem to play an important role in the localization of cyclic plasticity, crack initiation and early crack propagation [12].

The fatigue strength of cyclically loaded Cu is determined mainly by cyclic slip localization, initiation of fatigue cracks and early crack propagation. The complete solution of the damage by fatigue needs together with the knowledge on cyclic slip localization mechanism primarily the knowledge on the crack initiation process, including the crack path in the early fatigue crack stage. The mechanism of crack initiation known from CG materials cannot be directly applied to UFG structures. In the CG Cu the crack initiation and early crack propagation is related to the persistent slip bands (PSBs). Important role plays the surface roughness (extrusions and intrusions), which develops during cycling on the free surface. The characteristic dimension of specific dislocation structure below the surface relief, i.e. the width and length of the ladder-like PSB structure in the case of low-amplitude loading or the dimension of layers of dislocation cells in the case of high-amplitude loading, substantially exceeds the characteristic structural dimension of UFC Cu. Here, contrary to the case of CG Cu or Cu single crystals, where the characteristic dislocation structures develop, no specific dislocation structures were detected [6].

Summarizing the up-to now state of art, there is no satisfactory explanation and understanding of the cyclic slip localization, crack initiation and early crack propagation in UFG Cu at present. Simultaneously, it is obvious that the knowledge obtained on CG copper cannot be directly utilized for UFG structures. The aim of this study is to investigate the initiation and early crack propagation process in UFG Cu and to contribute to the completion of the knowledge on this phenomenon.

2. Material

Copper of 99.9% purity was processed by ECAP. Cylindrical billets of 20 mm in diameter and 120 mm in length were produced by eight passes through the die using the route Bc (90° rotation after each extrusion). The ECAP procedure was carried out at room temperature. After the last ECAP path cylindrical semi-products of 16 mm in diameter and 100 mm length were turned from the billets.

Microstructure as observed by means of light microscopy (LM) on polished and etched ($K_2Cr_2O_7 + H_2SO_4 + H_2O$) two perpendicular axial and on transversal sections of the cylindrical billet is shown in Fig. 1. Some degree of directionality of structure can be seen on both axial sections, Fig. 1a and b. The structure on transversal section does not exhibit any clear traces of directionality. The

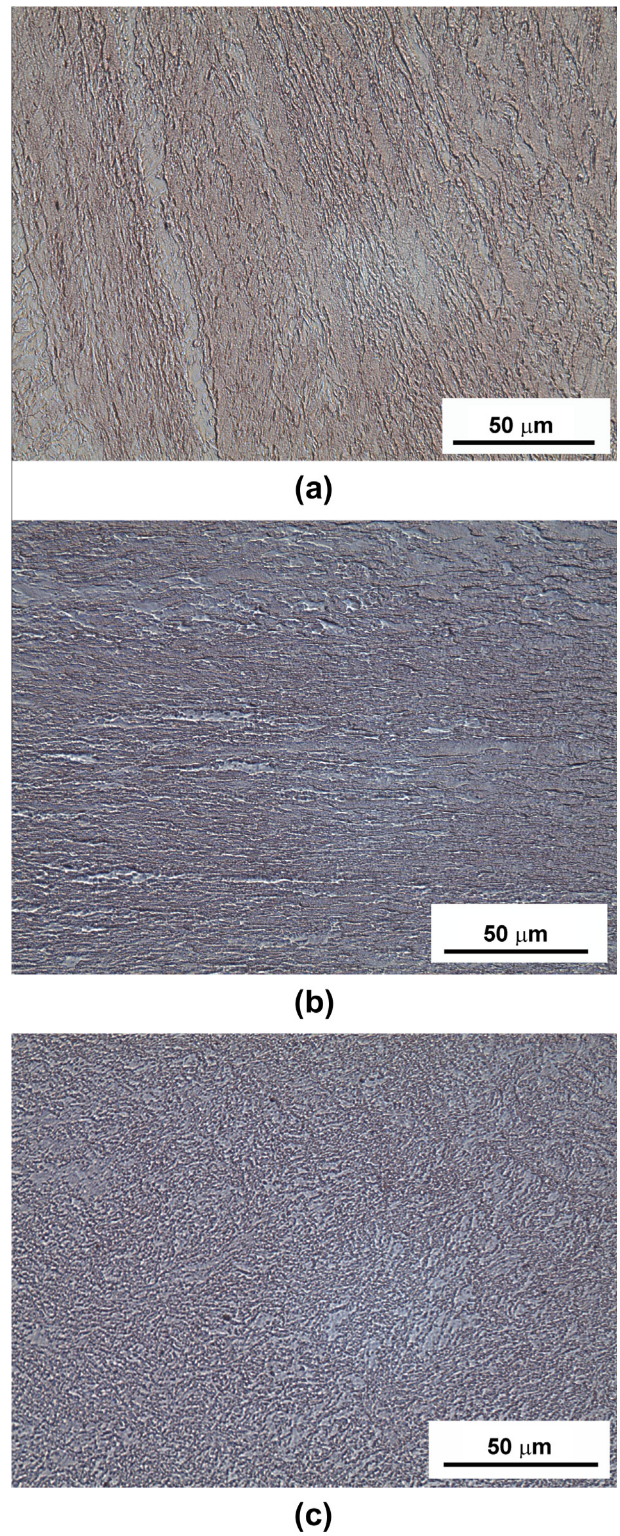

Fig. 1. Microstructure of UFG Cu as observed by light microscopy on two perpendicular axial sections (a and b) and on transversal section (c) of an ECAPed billet.