International Journal of Fatigue 63 (2014) 68–77

Contents lists available at ScienceDirect

International Journal of Fatigue

journal homepage: www.elsevier.com/locate/ijfatigue

Very high cycle fatigue of copper: Evolution, morphology and locations of surface slip markings

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ARTICLE INFO

Article history:
Received 21 August 2013
Received in revised form 31 December 2013
Accepted 5 January 2014
Available online 11 January 2014

Keywords:
Very high cycle fatigue
Persistent slip bands
Grain boundaries
Strain localization
Fatigue limit

ABSTRACT

The surfaces of commercially pure polycrystalline copper specimens subjected to interrupted 20 kHz fatigue tests in the very high cycle fatigue regime were investigated. The stress amplitude needed to form the early slip markings was found twice lower than the stress amplitude required to fracture which confirmed the results obtained by Stanzl-Tschegg et al. (2007). Three types of slip markings were classified according to their morphology and their location in the polycrystalline material. They are compared to slip markings observed during fatigue tests at frequencies lower than 100 Hz and numbers of cycles lower than 10^7. For 20 kHz fatigue tests, stress amplitudes ranging from 45 MPa to 65 MPa produce straight and long early persistent slip markings located along twin boundaries. Stress amplitudes lower than 45 MPa produce clusters of fine early persistent slip markings mainly located at triple junctions.

1. Introduction

There is currently a growing demand to investigate the very high cycle fatigue (VHCF) regime (higher than 10^7 cycles) to improve the reliability of design calculations and to extend the service lifetimes of some components such as engines. The shape of the S–N curve (plot of stress versus number of cycles to failure) varies with the material [1]. The hardest materials were found to exhibit a gradual decrease of the fatigue resistance with increasing number of cycles, whereas the softest alloys were found to present a plateau from 10^7 up to 10^9–10^{10} cycles [2]. Titanium and tantalum [3] and aluminium [4] exhibit a decrease of the fatigue resistance at a constant rate from 10^4 up to 10^8 cycles. The S–N curve for pure copper displays two slopes: the decrease of the fatigue resistance is stronger from 10^4 up to 10^7 cycles than from 10^7 up to 10^9 cycles [5–8]. All these results demonstrated that failure could occur beyond 10^7 cycles. However, the reason why such discrepancy as far as the VHCF response with regard to material is concerned is not clearly understood. Mughrabi [8] suggested distinguishing two classes of metallic materials when they are loaded in the VHCF regime: type I materials are pure ductile single-phase metals containing no extrinsic internal defects; type II materials contain internal defects such as precipitates or non-metallic inclusions. Most current studies have focused on the VHCF behavior of type II materials and revealed that, in the transition from high cycle fatigue (HCF) to VHCF, the origins of fatigue failure changed from surface to interior "fish-eye" fracture for instance at non-metallic inclusions [1,7]. For type I materials, only few investigations mainly on copper [9–12] are available regarding the shape of the fatigue life curve and the damage evolution [9–14]. Hessler et al. [9] and Weidner et al. [11] showed that cyclic strain localization in VHCF below the PSB threshold occurs in some form of PSBs. Despite the fact that there seems to be no possibility for the formation of a pronounced PSB structure at very low stress amplitudes, Mughrabi's model suggests that cracks initiate at the surface owing to the accumulation of very small but irreversible plastic deformation over very large number cycles [15]. In this work, the morphology and the location of slip markings observed after fatigue testing in the VHCF regime is clarified for pure copper. To perform experiments up to a very high number of cycles in a reasonable time, ultrasonic equipment at a testing frequency of 20 kHz was used. Three types of persistent slip markings are classified for the first time in terms of the stress amplitude ranges at which they appear predominantly. Slip markings morphology observed at 20 kHz and frequencies lower than 100 Hz are compared. The relationship between the appearance of persistent slip markings and failure is discussed.

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http://dx.doi.org/10.1016/j.ijfatigue.2014.01.007

2. Material and experimental procedure

2.1. Material and specimens

Oxygen-free high-conductivity commercially pure (99.95%) copper was tested. The polycrystalline material was hot rolled and supplied by Griset company. The specimens were extracted from the center of a 14 mm thick plate (Fig. 1). The mean grain size was about 30 μm. The microstructure contains about 0.3 volume fraction of annealed twins. The texture is quasi-isotropic. Fatigue tests were performed using a piezoelectric fatigue machine designed by Bathias and Paris [1]. Cylindrical hourglass shaped specimens were fatigued to get the S–N curve while flat hourglass shaped specimen were designed to carry out surface observations after interrupted fatigue tests. The specimens were designed to run in longitudinal vibration resonance with the piezoelectric machine at 20 kHz. The diameter in the center of the cylindrical specimens was 3 mm. The cross section in the center of the flat specimens was 2 mm thick and 3 mm wide. The dimensions of both specimen types are indicated on Fig. 2. After machining, the specimens were heat treated at 250 °C for 60 min to relieve the residual stress without change of microstructure. After thus thermal treatment, the tensile strength of the material is about 230 MPa. Then, all specimens were mechanically polished and soldered into a screw to attach them to the ultrasonic transducer. Finally, they were electrolytically polished to remove all hardened layers on the specimen surface. As a result, before testing, the specimens were mirror polished and without residual stresses. For all calculations, the Young's modulus was taken equal to 130 GPa [10].

2.2. Experimental procedure

The piezoelectric transducer of the ultrasonic fatigue machine converts an electronic signal into a mechanical displacement. This displacement is amplified via a horn and transmitted to the specimen. It was calibrated at the beginning of each series of tests with a laser extensometer. The displacement was found to be perfectly sinusoidal in the course of time. Assuming pure linear elastic behavior, the strain and stress distributions along the specimen were calculated using a one-dimensional approach (the mechanical variables depend only on x coordinate) in forced vibration regime. The strain was then measured by means of a small strain gage stuck on the specimen surface center to check the calibration. The strain R ratio defined as $\frac{\varepsilon_{min}}{\varepsilon_{max}}$ was -1. In the following, the stress amplitude rather than the strain amplitude is used to define the fatigue test. In hour-glass shaped specimens, the strain and stress amplitudes strongly decrease towards the specimen ends. They are 97% of maximum at places 1.25 mm and 1.5 mm apart from the center for the cylindrical and flat specimens, respectively (Fig. 2). Additional 3D elastic finite element simulations in forced vibration regime, carried out on the hour-glass shaped flat specimen, showed that the stress amplitude at the border of the smallest cross section is slightly greater (~5%) than the stress amplitude at the center of the cross section.

To get the S–N curve for a number of cycles higher than 10^6, hour-glass shaped cylindrical specimens were continuously fatigued up to failure at various stress amplitudes. Failure occurred when a crack had grown large enough to decrease the natural frequency of the system below the standard operating range (19.5–20.5 kHz) leading to a machine stop. The specimens were cooled with compressed air to avoid self-heating due to intrinsic dissipation. After the fatigue test, they were broken in liquid nitrogen for fracture surface examination.

In addition, fatigue loadings performed on the hour-glass shaped flat specimens were interrupted repeatedly in order to perform surface studies by optical microscopy, scanning electron microscopy (SEM), electron back scattering diffraction (EBSD) technique and atomic force microscopy (AFM) after different numbers of cycles. The stress amplitude range [30–60 MPa] was chosen much smaller than for the S–N curve to observe the early traces of plasticity on the surface specimen. In this stress amplitude range, the specimen heating was found lower than 20 °C so the specimens were not cooled. The morphology of the surface patterns was observed by means of a Hitachi S3600 FEG-SEM. EBSD analyses clarify the places of the plastic markings with regard to the polycrystalline microstructure. AFM was operating directly to the surface specimen (no replica) to keep the details of the roughness. The AFM Vecco nanoscope V was used in tapping mode in ambient air and room temperature. The AFM has a V-shaped Si_3N_4 tip with a spring constant of about 20 N/m of the cantilever beam. The side angle of the tip is 17 nm and the radius of its curvature 8 nm.

3. Results

3.1. S–N curve

The S–N curve of the studied copper obtained by ultrasonic fatigue machine at 20 kHz is shown in Fig. 3. It displays two regimes. In the first regime, failure occurs between 3.6×10^6 and

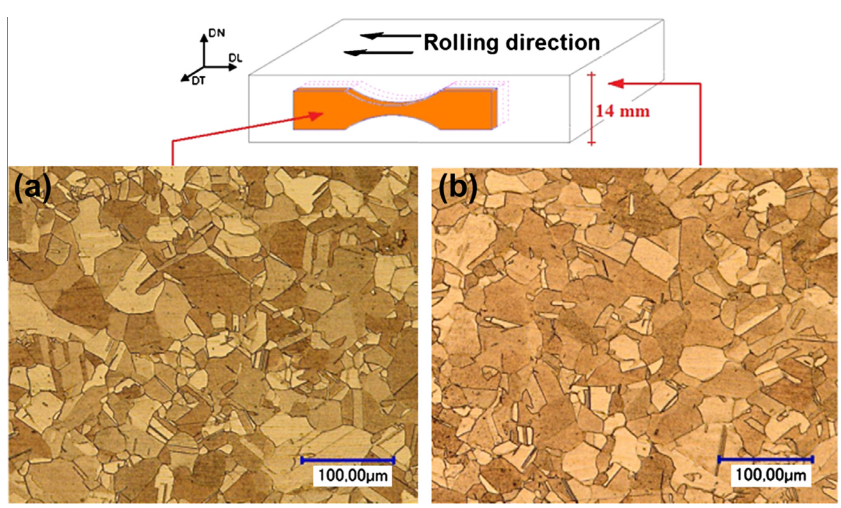

Fig. 1. Polycrystalline pure copper microstructure: cross section (a) in rolling direction and (b) perpendicular to the rolling direction.