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Following dislocation patterning during fatigue

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ABSTRACT

Precursors of failure are dislocation mechanisms at the nanoscale and dislocation organization at the mesoscale responsible for long-range internal stresses and lattice rotation. Detailed information on the link between both scales is missing, computationally and experimentally. Here we present a method based on x-ray Laue diffraction scanning providing time and sub-micron spatially resolved evolution of geometrical necessary dislocations in volumes that are similar to what advanced computational models can achieve. The approach is used to follow dislocation patterning during accumulation of fatigue cycles using a newly developed miniaturized shear device. Performed on Cu during cyclic shear, it reveals early dislocation patterning influenced by pre-existing dislocation structures. The quantitative information on non-homogeneous structure formation and its evolution corresponds to the need for synergies with continuum dislocation plasticity simulations of fatigue or any other type of plastic deformation.

1. Introduction

Plastic deformation in metals results from the motion of interacting dislocations developing heterogeneous structures with dislocation depleted and dislocation rich zones [1,2]. Advanced x-ray diffraction techniques have demonstrated a broad distribution of elastic strains within the dislocation depleted zones [3] and confirmed that these regions show intermittent dynamics and transient behaviour during loading [4]. While failure occurs at the macroscale, the pattern has a typical lengthscale at the micron scale and its formation involves multiple dislocation interactions and slip mechanisms at the nanometer scale. That is why the transition from homogeneous to heterogeneous dislocation microstructures is difficult to describe [5,6]. Computational models at all lengthscales have been used to understand particular aspects of dislocation patterning [7–11]. However, to capture fully the transition from uniform to non-uniform dislocation microstructures discrete dislocation models or density based dislocation dynamics models are required [9–15], and of equal importance, experimental methods must be available to validate the models.

In those computational models, dislocations are often separated into two different categories: statistically-stored dislocations (SSD), which evolve from random trapping processes during plastic deformation and geometrically-necessary dislocations (GND) [16]. The concept of GND was already introduced by Nye [17] to rationalize the compatibility of elastic-plastic deformation in materials experiencing strain gradients. GNDs introduce a characteristic length scale in the continuum formation of plastic deformation and give rise to deformation-induced long-range internal stresses [5,18]. The distribution of GND density can be calculated from experimental measurements of local crystallographic misorientation.

Electron channelling contrast imaging can spatially resolve dislocation patterns over several tens of micrometres close to the surface [19]. This technique however cannot distinguish between GND and SSD. GND density distributions in a 20 nm surface layer can be separately obtained by electron backscattering (EBSD) [20] with a nm spatial resolution. When combining EBSD with a serial sectioning, information on GNDs can be extended to the third dimension [21]. X-ray diffraction can resolve three-dimensionally (3D) and non-destructively local lattice rotation and elastic strain fields with a spatial resolution of the order of a micron depending

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on the beam diameter [22–26]. The technique is however slow [27]. Performed on thin sections of a deformed crystal it provided the probability density distribution of local misorientation [25].

The above-mentioned methods provide detailed information about the microstructure after a certain amount of strain, but they cannot teach us how patterning forms and evolves. To increase synergism with dislocation density based computational schemes, we designed an approach based on 2D Laue diffraction scans. As dislocation densities are integrated over the sample's thickness along beam direction, a template-matching technique is used for statistical information on the orientation spread in the beam direction. The method is applied during cyclic reversed shear at intermediate strain amplitudes on a Cu single crystal oriented for single slip. In this configuration, it is expected to form the so-called channel-vein structure before the gradual transformation into persistent slip bands [28]. Veins consist of an agglomeration of edge dipoles, surrounded by GNDs. Channels on the other hand are regions with very low SSD and no GNDs. A few phenomenological models have been put forward to explain this particular pattern [1,5,29], but there is no general understanding on the formation and evolution.

2. Experimental method

2.1. Shear device

A dedicated miniaturized deformation rig was built to apply reversible shear. A schematic view of the machine is presented in Fig. 1. The rig has dimensions of $200 \times 900 \times 150$ mm^3. The frame is made of an AlMgSi alloy. On the back side an opening of 100° is created for a large angular acceptance. Reversible shear is induced by two pins that sequentially push on either side of the sample. The pins are mounted on two Smaract linear actuators (SLC-1760) with 7N blocking force, sub-nanometer scan resolution and 41 mm travel range. Two 5N load cells (Transducer Techniques, USA) measure the applied force on the pins. In order to compensate gravity a counterweight is installed for the lower linear actuator. On the top side of the specimen a cantilever is installed in between pin and sample. This system measures the actual displacement of the upper pin with a Renishaw Tonic optical encoder with 20 nm resolution.

The shear device is controlled with in-house written LabVIEW routines. During the complete test both pins remain into contact with the sample. When one pin pushes on the sample the second pin retracts. A PID-control loop ensures that the force on this second pin remains zero.

2.2. Sample preparation and orientation

The samples are machined from $19 \times 19 \times 2$ mm^3 copper single crystals with 99.999% purity, purchased from Mateck GmbH, Germany. First bores and grooves are machined. Then the samples are annealed at 800° for 2 h with long cooling time, mounted on the sample holder and finally machined by electric discharge machining to a Miyauchi's geometry [30] as shown in Fig. 2a. It consists of two external immobile parts fixed with screws and a central moving area displaced with the pins from the shear rig. These parts are separated by channels and by two symmetric milled 150 μm thick grooves, corresponding to the shear zones. These shear zones are furthermore locally thinned by picosecond pulsed

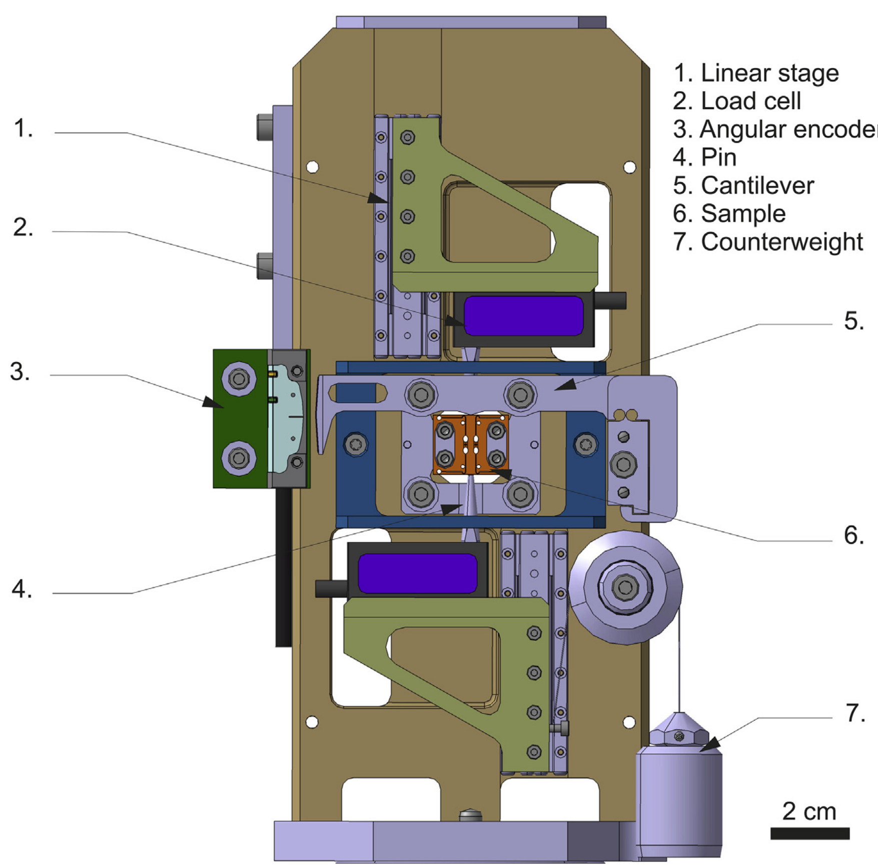

Fig. 1. Schematic view of the deformation rig that allows applying reversible shear.