Orientation-dependent fatigue damage in planar slip metals

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Abstract. Shear banding as a ubiquitous phenomenon is mostly observed in planar slip metals under complex plastic deformation conditions, especially in cyclic deformation. The deformation substructures and local damages during fatigue vary from grain to grain, which depends greatly on the crystallographic orientation. Despite previous extensive studies, the orientation-dependent deformation substructure is still elusive due to the complexity of damage evolution and the limitation of spatial resolution in both reciprocal and real sample spaces by conventional characterization methods. In this study, we conducted synchrotron-based X-ray micro-diffraction (μXRD) experiments to explore the dependence of localized elastic strains on grain orientation. It is demonstrated that the typical intersection of slip shear bands (SSBs) with Lomer-Cottrell locks (LCLs) are the potential damage locations, which are most likely to form in the near [001]//LD grains. The advanced μXRD with ability of spatially quantify the localized strains/fine orientations opens new opportunities for understanding the microscopic mechanisms of damage and failure in bulk materials.

1. Introduction

Shear bands are the narrow layers of materials undergoing intense shearing strains. Under a fast monotonic or low-strain-amplitude cyclic load, cumulative damage gradually develops within the microscale shear bands such as dislocation slip bands, leading to the eventual material failure. SSBs have been extensively observed in the planar slip alloys subjected to cyclic deformation [1-3], in which damage accumulation and resulting instabilities represent one of the most common modes of material failure. It is still a significant challenge to rationally evaluate the cyclic failure life of structural materials due to a lack of knowledge on the microscale stress heterogeneity across SSBs and the criterion of localized instability.

Deformation substructures and microscale stresses are developed during cyclic deformation, which can be used as an important microstructural indicator to scale the progress of fatigue damage. It is in a close correlation with primary slip systems activated during fatigue of planar slip materials. Detailed investigations on the evolution of deformation substructure and its dependence of crystal orientation are of great value in the understanding of fatigue damage. The traditional characterization techniques such as EBSD can be used to obtain the nanoscale misorientation distribution on the surface of deformed sample. However, the stress relaxing on the sample surface may destroy the configuration of defects and alter the microscale stress field. The synchrotron-based X-ray micro-diffraction (μXRD) technique provides a powerful tool to study the microscale distribution of elastic strain and grain orientation within three-dimensional deformed sample. From the microscale stress field, it is also possible to determine
the configuration of various defect structures with submicrometer resolution, being the fundamental element of substructures.

AL6XN SS is a typical austenite stainless steel (SS) with low stacking fault energy. As a result, planar slip prevails, yielding rich planar slip bands during fatigue cycling, which makes it an ideal target material for the study of orientation-dependent fatigue damage.

2. Experimental procedure
The AL6XN SS has a single-phase face-centered-cubic (FCC) structure. Samples for mechanical testing were cut from the sheet after 1160 °C solution treatment for 2 hours followed by water quenching and hot rolling with a final thickness of 15 mm. The final grain sizes range from 80 to 120 µm. The standard rod-shaped samples were used in fatigue tests with gauge section of 8 mm in diameter and 20 mm in length. The strain-controlled fatigue test was conducted under a total strain control with the maximum strain amplitude of 0.3%, the cycle asymmetry coefficient R = -1, and the load frequency of 0.5 Hz at room temperature in air.

The µXRD [4-8] was used to explore the strain heterogeneity of grains in a fatigued bulk AL6XN SS. Spatially-resolved 3D µXRD measurements were conducted by using differential-aperture X-ray microscopy (DAXM) at beamline 34-ID-E of the Advanced Photon Source in Argonne National Laboratory. An X-ray microbeam (polychromatic or monochromatic) is focused to about 0.5 µm × 0.5 µm by a pair of elliptical Kirkpatrick-Baez focusing mirrors. Figure 1a is a schematic illustration of the diffraction geometry with the loading direction (LD), transverse direction (TD) and normal direction (ND) labeled in light blue. Diffracted X-ray beams will be recorded on an X-ray area detector.

![Figure 1](image_url)

**Figure 1.** Schematic of µXRD experiments. (a) Diffraction geometry. (b) The polychromatic Laue pattern. (c) The monochromatic Laue pattern.

The polychromatic and monochromatic Laue spots are illustrated in Fig. 1b and 1c, respectively. The point-to-point micro-orientation can be calculated from the polychromatic Laue spots, and the high accuracy lattice strain along ND can be calculated from the energy scanning for monochromatic Laue
spots. The depth resolution is provided by scanning a 100-μm-diameter platinum wire as diffracted-beam profiler in a knife-edge fashion. By triangulation, the origins of the X-rays can be determined on a pixel-by-pixel basis. A typical scanning step of 1.4 μm gives the depth resolution of ~1.0 μm. The detailed information on μXRD is well described previously [6].

3. Results and discussion

3.1. Orientation-dependent deformation substructure

The micro-orientation with an order of magnitude higher on precision than Electron Back-Scattered Diffraction (EBSD) [9] can be obtained from the analysis on monochromatic Laue spots in the μXRD experiments. Better yet, we can get the point-to-point local strain tensor map from the calculations, which provides valuable data for studying orientation-dependent features.

![Figure 2](image)

**Figure 2.** Polychromatic μXRD scanning results. Inverse-pole figure (IPF) map (a) and εzz (b) showing grain orientation and corresponding strain distribution of a point-to-point scan area. (c) and (d) are the local amplifications of (a) and (b). (e) and (f) are the IPF (parallel to LD/Y) and coordinate system.

It has been well revealed that the orientation-dependent residual stress is widely existed in fatigued materials. The stress is related to the deformation substructure, which influences the damage of fatigue greatly [10-12]. In this study, the distribution of residual strain is revealed in the monochromatic μXRD experiments. An inverse pole figure (IPF) map and the corresponding residual strain map are shown in Fig. 2a and 2b, respectively. Fig. 2c and 2d are their local amplification focussed on several grains. A distribution of orientation dependent residual strain with fine spatial resolution on a bulk fatigued steel is directly visualized. The red/reddish grains with their [001] axis nearly parallel to LD ([001]/LD) in Fig. 2a are consistent with the blue grains having a larger compressive strain along ND in Fig. 2b. More detailed, the deformation substructures of a large number of grains are displayed in the residual strain
maps (Fig. 2b and 2d), from which it can be concluded that the near [001]//LD grains undergo severe heterogeneous deformation, with the formation of obvious shear bands through more than one slip systems. This microstructural future is further confirmed by the monochromatic $\mu$XRD scan within a [001]//LD grain and a [110]//LD grain.

The diffraction spots of (480) and (006) crystal planes are used to characterize the local features of the [001]//LD grain and the [110]//LD grain in the monochromatic $\mu$XRD. The deformation substructure is clearly revealed from the full width at half maximum (FWHM) maps (Fig. 3a and 3b). Shear bands are merged into background (matrix information) in the [110]//LD grain (Fig. 3b), but they are very obvious in the [001]//LD grain (Fig. 3a). The complex evolution of deformation substructure is important to scale the progress of fatigue damage.

Figure 3. Deformation substructures of AL6XN SS after 29,000 cycles. (a) A map of the (480) diffraction peak FWHM of a [001]//LD grain. (b) A map of the (006) diffraction peak FWHM of a [110]//LD grain. They were measured by monochromatic X-ray.

It should be emphasized that the orientation-dependent deformation substructure is particularly remarkable in cyclic deformation of planar slip materials, in which no mesoscopic grain reorientation occurs and the SSBs are relatively in favour during the cyclic deformation.

3.2. Micro orientation distribution in a fatigued [001]//LD grain

The micro orientation map of the [001]//LD grain is visualized in Fig. 4a with a local amplification including a wide shear band shown on the right. The colour is based on the coordinates in the [001] pole figure as pictured in Fig. 4c. The misorientation across the grain is no more than 0.2°. A sharp misorientation is observed on the left edge of the shear band, which is consistent with the low-angle grain boundary model with edge dislocations aligned as shown in Fig. 4b. Maps of misorientation showing grain subdivision are consistent with the observed deformation substructure (Fig. 3a).
Figure 4. (a) Orientation map of a fatigued [001]//LD grain with a local amplification on the right. (b) Model of low angle grain boundary with edge dislocations aligned. (c) Pole figure of {001} plane with an amplification showing detailed orientation distribution (d) on the right.

3.3. Damage source of planar slip stainless steel: Lomer-Cottrell locks

Our former work [13] has confirmed that the interaction of SSBs is the main damage source during cyclic deformation of AL6XN SS. The LCLs are forming by the dislocation reaction in Burger vectors of $\frac{1}{2}[10\bar{1}] + \frac{1}{2}[011] = \frac{1}{2}[110]$, in the crossing of SSBs, as shown in Fig. 5. The sessile dislocations deeply promote the dislocation pile-ups, and huge local stress gradients raise around the pile-ups. With the increase in the number of cycles, these locations finally develop into micro cracks [13, 14].

Figure 5. TEM showing the interactions of SSBs in the AL6XN SS after 29,000 cycles.

According to the above characterizations and analyses on the fatigue damage processing, the further development of SSBs through different slip systems within one grain will lead to a quicker local damage and earlier fatigue failure. Analyses on the formation and development of SSBs in various orientation grains are important for predicting the fatigue life. For simplification, the analysis below will expand based on the classical theory of plastic deformation in a qualitative manner due to the complexity of strained conditions of various grains.
It is extensively accepted that (111)[110] slipping is the dominating deformation mode of FCC materials in small/cyclic deformation. The Schmidt factor is often used in the determination of activated slip systems in high cycle fatigue. But focusing on the particular situation of ALXN SS during cyclic deformation that the interaction of mature SSBs is a precursor of grain damage. In another word, grain with the largest maximum Schmidt factor, that generally means it is the easiest one to deform, is not necessarily the weakest grain in fatigue. More factors and LCLs should also be taken into account. In the following, we establish a simplified double SSBs model to analysis the orientation dependent localized deformation substructure.

The model of double SSBs is established on the assumption that mature SSBs are allowed to develop on no more than two (111) planes among grains during cyclic deformation. On the classical theory of slipping deformation, the distributions of maximum Schmidt factor and second-maximum Schmidt factor in IPF along LD are drawn in Fig. 6a and 6b, respectively. Because our motivation is to evaluate the forming probability of LCLs in grains with various orientations and no interaction of SSBs occurs if dislocations slipping on one (111) plane, so what need to be clarified is that the slip systems having the maximum and second-maximum Schmidt factor should not be a same (111) plane.

In Fig. 6a and 6b, the red parts present the orientations easiest and second-easiest to be activated, respectively. Fig. 6c is the sum of Fig. 6a and 6b, from which we can evaluate the forming probability of double SSBs in simple way. The grey part in Fig. 6c is the orientation with no $\frac{1}{2} [101] + \frac{1}{2} [011] = \frac{1}{2} [110]$ LCLs forming on the above assumption that only two slip systems are allowed to be activated. This means that the grains in grey may be in favourable orientations to activate two slip systems simultaneously, but dislocations on these planes cannot be locked on the SSBs via dislocation reactions. The SSBs of these grains will be thinner and distribute in a homogeneous way. The analysis is in good consistent with the above-mentioned observation of deformation substructures. It can be see clearly from the model that double SSBs with Lomer-Cottrell locks are most likely to form in the near [001]/LD grains. It still need further investigations in order to reveal the underlying mechanisms of orientation-dependent crystal damage in a quantitative way, which is now in progress.

4. Conclusions
Our μXRD experiments illustrated the orientation-dependent residual strains in bulk fatigued stainless steel. The orientation-dependent local damage of fatigued planar slip metals is analysed by the classical slip theory in a qualitative manner. The typical double SSBs with LCLs are the potential damage spots, which are most likely to form in the near [001]/LD grains. The experimental findings and theoretical analyses of orientation-dependent properties will contribute to the understanding of fatigue failure mechanism, and further promote the design of materials. As demonstrated in this work, the spatially resolved 3D μXRD mapping of orientation and lattice stress field at submicron scale provides a new opportunity to probe the microscopic damage and failure mechanisms in plastically deformed bulk materials.
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