1. Introduction

Understanding the interaction between slip dislocations and grain boundaries (GBs) has a paramount importance on the mechanical response of metals [1,2]. Extensive research has been reported during the last decades on the strengthening effect introduced by blockage of partial or full slip dislocations at GBs. In particular, much interest has been devoted to the specific slip dislocation–GB reactions and the resulting slip transmission process across the GB. In this regard, early research focused on the details of these reactions at GBs utilized transmission electron microscopy (TEM) [3,4]. In particular Clark et al. [3] and Lee at al. [4] utilized in situ TEM images and characterized the possible outcomes of the dislocation reactions occurring at a GB. Several insights into the transmission of the incoming dislocation, and incorporation into the GB creating extrinsic (residual) dislocations were gained as a result of these studies.

Further experimental efforts are required to overcome the difficulties in correlating the results of these dislocation reactions with the associated strain fields across the GBs (on the meso-scale). A measure of the strengthening associated with a GB is decided based on whether dislocation strain fields undergo a continuous variation (full dislocation transmission), or whether large strains accumulate on one side of the GB (dislocation blockage). The level of strain heterogeneity measured across the GBs represents a measure of the capability of a GB to block or transmit dislocations. Therefore, a focused study on the experimental determination of the localized strain gradients across GBs is important and will provide considerable insight into dislocation transmission and GB contribution to hardening. In this paper, with special image correlation techniques, we study the local strain fields at the meso-length scales covering multiple grains in a Fe–Cr alloy. The strain measurements are used to establish the possible outcome of slip-GB interaction and capture the importance of the residual dislocation mediated plasticity. We extend the combination of local strain and grain orientation measurements on fcc alloys [5,6] introduced in early work to a bcc metal where a single slip system per grain is activated allowing a clear examination of the strain fields.

Grain boundaries induce heterogeneities in the deformation response of polycrystals. Studying these local variations in response, measured through high resolution strain measurement techniques, is important and can improve our understanding of fatigue damage initiation in the vicinity of grain boundaries and material hardening. In this work, strain fields across grain boundaries were measured using advanced digital image correlation techniques. In conjunction with strain measurements, grain orientations from electron back-scattered diffraction were used to establish the dislocation reactions at each boundary, providing the corresponding residual Burgers vectors due to slip transmission across the interfaces. A close correlation was found between the magnitude of the residual Burgers vector and the local strain change across the boundary. When the residual Burgers vector magnitude (with respect to the lattice spacing) exceeds 1.0, the high strains on one side of the boundary are paired with low strains across the boundary, indicating the difficulties for slip dislocations to penetrate the grain interfaces. When the residual Burgers vector approaches zero, the strain fields vary smoothly across the boundary due to limited resistance to slip transmission. The results suggest that the residual Burgers vector magnitude, which relates to the GB (Grain Boundary) resistance to slip transmission, enables a quantitative analysis of the accumulation of strain at the microstructural level and the development of strain heterogeneities across grain boundaries. The results are presented for FeCr bcc alloy which exhibits single slip per grain making the measurements and dislocation reactions rather straightforward. The work points to the need to incorporate details of slip dislocation–grain boundary interaction (slip transmission) in modeling research.

© 2013 Elsevier B.V. All rights reserved.

Keywords: Bcc polycrystal
Slip transmission
EBSD
Digital image correlation

Article history:
Received 27 March 2013
Received in revised form 31 July 2013
Accepted 26 August 2013
Available online 2 September 2013

Materials Science & Engineering A 588 (2013) 308–317

Luca Patriarca a,⁎, Wael Abuzaid b, Huseyin Sehitoglu b,1, Hans J. Maier c

a Politecnico di Milano, Department of Mechanical Engineering, Via La Masa 34, 1-20156 Milano, Italy
b Department of Mechanical Science and Engineering, University of Illinois at Urbana-Champaign, 1206W. Green St., Urbana, IL 61801, USA
c Institut für Werkstoffkunde, Leibniz Universität Hannover, An der Universität 2, D-30823 Garbsen, Germany

ABSTRACT

Grain boundaries induce heterogeneities in the deformation response of polycrystals. Studying these local variations in response, measured through high resolution strain measurement techniques, is important and can improve our understanding of fatigue damage initiation in the vicinity of grain boundaries and material hardening. In this work, strain fields across grain boundaries were measured using advanced digital image correlation techniques. In conjunction with strain measurements, grain orientations from electron back-scattered diffraction were used to establish the dislocation reactions at each boundary, providing the corresponding residual Burgers vectors due to slip transmission across the interfaces. A close correlation was found between the magnitude of the residual Burgers vector and the local strain change across the boundary. When the residual Burgers vector magnitude (with respect to the lattice spacing) exceeds 1.0, the high strains on one side of the boundary are paired with low strains across the boundary, indicating the difficulties for slip dislocations to penetrate the grain interfaces. When the residual Burgers vector approaches zero, the strain fields vary smoothly across the boundary due to limited resistance to slip transmission. The results suggest that the residual Burgers vector magnitude, which relates to the GB (Grain Boundary) resistance to slip transmission, enables a quantitative analysis of the accumulation of strain at the microstructural level and the development of strain heterogeneities across grain boundaries. The results are presented for FeCr bcc alloy which exhibits single slip per grain making the measurements and dislocation reactions rather straightforward. The work points to the need to incorporate details of slip dislocation–grain boundary interaction (slip transmission) in modeling research.
As emphasized above, several important studies examining dislocation–GB interactions have been undertaken in the past \cite{7-10}. Recently, advanced simulation tools also provided further insight into energetics of the dislocation glide near grain boundaries \cite{11-13} and corroborated the experimental findings. On the other hand, an experimental quantification of the role of GBs on the local strain fields, e.g. the strain gradients following a specific slip–GB interaction, has received less attention. In this work, we combine microstructure characterization via electron back scatter diffraction (EBSD) and high resolution strain measurements from DIC \cite{5,6,14,15}. By combining these two techniques, we aim to shed further light into the localization of plastic strains due to dislocation–GB interactions. Before introducing the experimental results, we provide a brief overview of the slip–GB geometry and the interactions.

A schematic of a GB is shown in Fig. 1. Grain 1 contains the incoming slip system, while Grain 2 contains the outgoing slip system. In this schematic, the dislocations leave the GB in the second crystal (outgoing slip plane) as a result of the dislocation

\[ \vec{b}_1 = \vec{b}_2 + \vec{r}, \]

where \( \vec{b}_1 \) (incoming) and \( \vec{b}_2 \) (outgoing) are determined on the same coordinate basis. We calculate \( \vec{b}_1 \) using the slip directions of the incoming and outgoing slip systems in order to characterize the average slip dislocation–GB reaction that had occurred at the GB.

\[ \vec{b} = \vec{b}_1 (\text{LD} n_1) \]

\[ \vec{b} = \vec{b}_2 (\text{LD} n_2) \]

Fig. 1. Schematic of the dislocation–grain boundary interaction geometry. For both the incoming and outgoing slip planes are indicated the normal to the slip planes \((n_1\) and \(n_2\)) and the Burgers vectors of the dislocations \((b_1\) and \(b_2\)). \( \theta \) indicates the angle between the lines of intersection between the two slip planes and the GB plane. The Burgers vector of the residual dislocation left at the GB is determined from the equation \( \vec{b}_1 = \vec{b}_2 + \vec{r} \).
A 2 mm × 2 mm region on the sample’s surface was marked using Vickers indentation marks (the procedure is discussed in [19], also used in [5,6]). The underlying microstructure in the marked region of interest was characterized using EBSD. Fig. 2a shows the grain orientations for the analyzed region and Table 1 lists the magnitudes of the Euler angles in Bunge convention (φ₁, Φ, φ₂) for each detected grain. A total of nine grains were observed in the selected area, with an approximately mean grain size of about 1 mm. In Fig. 2b, the stereographic triangle along with the crystal orientations of each grain in the load direction are reported. The stereographic triangle is subdivided into five regions [20] that indicate the slip systems with the largest Schmid Factors (SFs) for bcc materials. In determining the slip system with the largest SFs in Fig. 2b, all of the (112), (011) and (123) slip planes were considered with (111) slip directions.

2.2. Experimental set-up and strain measurements

The experiment was conducted at room temperature using a servo hydraulic load frame. The sample was deformed in displacement control at a mean strain rate of 5 × 10⁻⁵ s⁻¹. We used the strain data from in situ DIC [21–24] to construct the stress–strain curve using DIC field averages. The images were captured using an IMI model IMB-202 FT CCD camera (1600 × 1200 pixels) with a Navitar optical lens, providing a resolution of 3.0 μm/pixel. The speckle pattern for DIC was obtained using black paint and an Iwata Micron B airbrush. We captured a reference image of the sample surface at zero load, and deformed images of the same area every 2 s during loading.

We also used ex situ high resolution DIC for acquiring higher resolution strain measurements. In that case, the reference and
deformed images were acquired out of the load frame (see details for this procedure in [19]) using an optical microscope which enables capturing higher magnification images than the in situ set-up (0.18 and 0.87 μm/pixel for ex situ images, as opposed to in situ images which were obtained at 3.0 μm/pixel). A speckle pattern suitable for high resolution DIC was applied on the sample's surface after the initial EBSD scan. A set of 140 images covering the analyzed region was captured before the experiment (reference condition) and after loading the sample (deformed condition). The correlations were implemented on each pair of images (reference and deformed) and the results were successively stitched together. We overlaid the grain map (from EBSD) with the strain fields (from DIC) using the Vickers indentation markers which are visible in the EBSD grain orientation map and the optical microscope images. We note that the reported strain fields using ex situ DIC refer to the residual plastic strains (un-loaded sample).

3. Results

3.1. Stress–strain curve and DIC strain measurements

We provide the results of the stress–strain curve and the strain contours in this section. Fig. 3 shows the stress–strain curve for the sample deformed in compression. The reported strains (x axis) were obtained using DIC field averages (in situ). The insets in Fig. 3 show strain contour plots of the axial strain εyy in the load direction. The inset images, marked A’ and B’, represent the residual axial strain fields for the 2 mm × 2 mm region obtained via ex-situ DIC at two different levels of strain. Each strain field is a composition of 9 images (resolution 0.87 μm/pixel) captured outside the load frame and after unloading from points A and B on the stress–strain curve. Strain heterogeneities develop as a consequence of the local grain orientation. The contour plots also show clear strain bands that are associated with slip system activation, we then identified the observed slip systems using the grain orientations from EBSD and SF analysis. The activated slip planes were determined by observing the possible traces of all the slip planes on the well-known (112), (110), (123) plane families for bcc materials. The slip direction, which is required to define the slip system, was determined using Schmid factor analysis for each observed slip. The slip systems with largest Schmid factors are successively selected comparing the projected lines with the slip traces on the sample’s surface. Two regions are selected from the global strain fields: 1A and 2A after load step A’, and 1B and 2B after load step B’ (Fig. 3). In Fig. 4 the same regions are paired with the schematic of the slip plane geometries. The GB plane is drawn using the GB trace from the EBSD map assuming that the normal
lies on the plane of the sample surface. For the region marked $1^A$–$B$ (Fig. 4a and b), the observed incoming slip system is $[11\bar{1}]\langle 2\bar{3}1 \rangle$, while the outgoing slip system is $[\bar{T}1\bar{T}]\langle 2\bar{1}1 \rangle$. In that case, the residual Burgers vector magnitude is low $|\vec{b}\_r| < 1a$:

$$\frac{a}{2}[11\bar{1}]_{\text{Cr2}} - \frac{a}{2}[11\bar{1}]_{\text{Cr4}} + \vec{b}_r \Rightarrow |\vec{b}\_r| = 0.38a$$

(2)

The strain fields $1^A$ and $1^B$ clearly show an accumulation of strains on both sides of the GB which can be associated with slip transmission across the interface. For the second case, $2^A$–$B$ (Fig. 4c and d), we selected a GB for which the reaction occurring between the incoming and the outgoing slip systems leads to a high residual Burgers vector magnitude $|\vec{b}\_r| > 1a$:

$$\frac{a}{2}[11\bar{1}]_{\text{Cr2}} - \frac{a}{2}[11\bar{1}]_{\text{Cr4}} + \vec{b}_r \Rightarrow |\vec{b}\_r| = 1.28a$$

(3)

The incoming slip system is $[11\bar{1}]\langle 2\bar{3}1 \rangle$, while the outgoing slip system is $[11\bar{1}]\langle 2\bar{1}1 \rangle$. In this case the strains preferentially accumulate on one side of the GB, indicating the difficulty for the incoming dislocations to be transmitted through the GB. The accumulation of strains on the GB side of the incoming slip system is observed for both the loading steps (Fig. 4c and d). More details on the slip–GB interaction for the cases analyzed in Eqs. (2) and (3) are given in Fig. 5. All the active slip systems have high SF values close to 0.5. The angle $\theta$ and the residual Burgers vector magnitude $|\vec{b}\_r|$ are higher for the second case ($|\vec{b}\_r| = 1.28, \theta = 48.5^\circ$) compared to the first case ($|\vec{b}\_r| = 0.38$) which showed easy slip transmission and strain accumulation across both sides of the GB.

3.2. High resolution DIC strain measurements

In this section we provide ex situ strain fields obtained capturing ex situ images at higher resolutions than reported in the previous section. Higher image resolutions are required in order to index all the slip systems activated in the selected EBSD region (Fig. 2a) and thus calculate the dislocation reactions. Fig. 6a displays the strain field obtained using images at high resolution (0.18 μm/pixel). A total of 140 images were captured before the experiment. As described in the previous section, after loading the sample (point A’ in Fig. 3), slip systems were indexed using grain orientations from EBSD. In Table 2 the slip systems for each grain along with the SFs are reported. For grains 2, 3, 4, 5, 7 and 8 only a single slip system is observed, while for grains 1 and 6 traces of secondary slip systems are visualized. In the schematic of Fig. 6b the $|\vec{b}\_r|$ values for each GB are also reported. We then marked the GBs on the DIC strain field with $T$ in case of slip transmission visualized as a strain continuity along the slip traces, while $B$ indicates the GBs for which no strain continuity is observed. For GBs 1–2, 4–6, 7–8 it is evident how the strains induced by slip continued almost unaltered through the interfaces, while for GBs 2–3, 4–5, 5–6, 2–7, 2–8, 6–4 the strains accumulate on one side of the GB. Strain accumulation is particularly evident for GBs 4–5 and 2–7. Each GB can be also characterized by the estimation of the $|\vec{b}\_r|$ magnitude due to slip transmission (see schematic in Fig. 6b). Table 3 gives a summary of the observed outcomes due to slip–GB interaction in the vicinity of GBs ($T$: slip transmission, $B$: slip accumulation) and the correlation with the $|\vec{b}\_r|$ magnitudes and the misorientation angles. From Table 3 it is clear that a low $|\vec{b}\_r|$ is associated with the slip transmission cases (geometrically smooth strain pathways), while slip accumulation is observed for the reactions resulting in high $|\vec{b}\_r|$ magnitudes (high strains on one side of the grain boundary). From Table 3 it is also evident that considering only the misorientation angle is not sufficient to explain the observed slip transmission and accumulation cases.

3.3. Strain measurements across grain boundaries

In order to quantify the magnitude of the strain change across the GB sides$, we provide two sets of strain measurements for low and high $|\vec{b}\_r|$ magnitudes. The inset image (a) in Fig. 7 shows the strain contour plot for a specific region consisting of grains 1 and 2 (see also Fig. 6). In this case a clear strain continuity associated with the incoming slip system $[\bar{T}1\bar{T}]\langle 1\bar{2}3 \rangle$ and outgoing slip system $[11\bar{1}]\langle 2\bar{3}1 \rangle$ is observed across the GB, and the residual Burgers vector magnitude is low $|\vec{b}\_r| < 1a$:

$$\frac{a}{2}[1\bar{1}1]_{\text{Cr1}} - \frac{a}{2}[1\bar{1}1]_{\text{Cr4}} + \vec{b}_r \Rightarrow |\vec{b}\_r| = 0.16a$$

(4)

In Fig. 7b we also report the strain measurements calculated across the grain boundary reported in Fig. 7a and characterized by the dislocation reaction in Eq. (4). Fig. 7b reports the averaged strains approaching the grain boundary sides. The difference of the strains measured across the GB provides a direct quantification of the slip accumulated on one side of the grain boundary. Each point on the strain plot in Fig. 7b is calculated averaging a selection of strain values captured from the DIC strain field (Fig. 7a). The selected strain values are contained in a rectangular selection oriented along the GB side with an approximate size of 40 μm × 400 μm. The coordinate $x$ is the distance between the center of the rectangular selection and the GB. For the grain boundary between grains 1 and 2 (Fig. 5) the dislocation reaction (Eq. (4)) indicates a low residual Burgers vector magnitude $|\vec{b}\_r| = 0.16a$, and the difference of the strain magnitudes approaching the GB is equal to $|\Delta \varepsilon_{\text{GB}}| = 0.09%$.

The grain boundary between grains 2 and 6 is analyzed in Fig. 8 (see also Fig. 6). Strain bands associated with the incoming slip system $[11\bar{1}]\langle 2\bar{3}1 \rangle$ and the outgoing slip system $[\bar{T}1\bar{T}]\langle 1\bar{2}3 \rangle$ are still observed to propagate continuously across the GB. The DIC strain
Fig. 6. (a) High resolution ex situ DIC strain field measurements. A total number of 140 reference images and 140 deformed images were captured outside the load frame (ex-situ), correlated and successively stitched. Slip systems were indexed using crystal orientations from EBSD selecting the systems with slip traces displaying the largest SFs. (b) Schematic displaying grain numbers and residual Burgers vector magnitudes. Depending on the magnitude of the residual Burgers vector, dislocations gliding on the active slip system in one grain can be transmitted through the GB or can accumulate at the GB. Different strain fields in the proximity of the GBs result from different dislocation–grain boundary interactions. Strains are transmitted in case of low $|\mathbf{b}| < 1a$, leading to high strain values measured at both sides of the GBs. In case of slip accumulation (high $|\mathbf{b}| > 1a$), high local strains can be observed on one side of the GB (see for example GB 2–7).

Table 2
Activated slip systems and SFs.

<table>
<thead>
<tr>
<th>Grain</th>
<th>1</th>
<th>2</th>
<th>3</th>
<th>4</th>
<th>5</th>
<th>6</th>
<th>7</th>
<th>8</th>
</tr>
</thead>
<tbody>
<tr>
<td>Slip</td>
<td>$[\text{111}][0\text{11}]$</td>
<td>0.46</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>$[\text{111}][2\text{3}1]$</td>
<td>0.48</td>
<td>0.49</td>
<td>0.47</td>
<td>0.45</td>
<td>0.42</td>
<td>0.49</td>
<td>0.49</td>
</tr>
<tr>
<td></td>
<td>$[\text{111}][\text{1}\text{2}\text{1}]$</td>
<td>0.45</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
field displays also intermediate magnitudes of strain bands (green color) between the incoming slip bands on the left side of the GB. These additional strain bands developing in proximity of the GB can be associated with the partial dislocations left in the GB having a residual Burgers vector magnitude of \( \frac{a}{2} \). For this grain boundary (\( |b_r| = 0.28a \)), the difference in the strain magnitudes approaching the GB is higher than the previous case (\( |b_r| = 0.16a \)) and equal to \( |\Delta \varepsilon_{GB} - \varepsilon_{GB}^{\text{low}}| = 0.22\% \). The higher strain change measured in this case is paired with a higher residual Burgers vector magnitude.

Table 3
Comparison between the observations on the DIC strain field on the slip mechanism (T: slip transmission, B: slip accumulation) with the residual Burgers vector magnitude \( |b_r| \) and misorientation angle \( \varphi \).

<table>
<thead>
<tr>
<th>GB</th>
<th>1–2</th>
<th>2–3</th>
<th>3–4</th>
<th>4–5</th>
<th>5–6</th>
<th>2–6</th>
<th>2–7</th>
<th>2–8</th>
<th>7–8</th>
<th>7–6</th>
<th>6–4</th>
</tr>
</thead>
<tbody>
<tr>
<td>Slip mechanism</td>
<td>T</td>
<td>B</td>
<td>B</td>
<td>T</td>
<td>B</td>
<td>T</td>
<td>B</td>
<td>B</td>
<td>T</td>
<td>T</td>
<td>B</td>
</tr>
<tr>
<td>(</td>
<td>b_r</td>
<td>) (divided by a)</td>
<td>0.16</td>
<td>1.07</td>
<td>1.14</td>
<td>0.38</td>
<td>1.14</td>
<td>0.64</td>
<td>0.28</td>
<td>1.28</td>
<td>1.31</td>
</tr>
<tr>
<td>Misorientation angle [( \varphi )]</td>
<td>24.9</td>
<td>41.5</td>
<td>31.36</td>
<td>42.3</td>
<td>42.6</td>
<td>17.0</td>
<td>21.6</td>
<td>18.2</td>
<td>24.9</td>
<td>6.9</td>
<td>13.0</td>
</tr>
</tbody>
</table>

Fig. 7. Strain measurements across a grain boundary in case of low \( |b_r| = 0.16a \). For this case slip transmits almost unaltered through the GB (a). The strains measured across the interface (b) display the small strain change across the grain boundary.

Fig. 8. Strain measurements across a grain boundary in case of low \( |b_r| = 0.28a \). For this case a small step on the strain field (b) is observed which represents preferential slip accumulation on the left side of the GB (a). The residual Burgers vector magnitude calculated is slightly higher than in the previous case shown in Fig. 7.

\[
\frac{a}{2}[1\overline{1}1]_{\text{Grain} 2} - \frac{a}{2}[\overline{1}1\overline{1}]_{\text{Grain} 6} + \vec{b}_r \Rightarrow |\vec{b}_r| = 0.28a
\]  

In Figs. 9 and 10 two cases of grain boundaries displaying high \( |b_r| > 1a \) magnitudes are shown. The first case refers to the strain measurements across grains 4 and 5 (see also Fig. 6). The incoming dislocations glide on the [11\overline{1}][\overline{1}3\overline{1}] slip system, while the outgoing dislocations glide on the [\overline{1}1\overline{1}][3\overline{2}1] slip system. In this case, the slip transmission process is characterized by a dislocation...
reaction that results in a high residual Burgers vector magnitude (1.14a):

\[
\frac{a}{2}[\{11\bar{1}\}^{\text{Grains} 4}] \rightarrow \frac{a}{2}[\{11\bar{1}\}^{\text{Grains} 5} + \vec{b}_r] \Rightarrow |\vec{b}_r| = 1.14a
\]  

(6)

From the DIC strain plot (Fig. 9b) it is clear that as a consequence of the high |\vec{b}_r|, strains accumulate on the right side of the GB leading to a high strain gradient equal to |Δε_{GB} | = 0.47% which is higher than the cases previously reported in Eqs. (4) and (5). The last case analyzed (Fig. 10) has been already introduced in Fig. 4c and d (strain field 2\text{A}–\text{B}) using lower image resolution (0.87 versus 0.18 μm/pixel). The dislocation reaction between the incoming \( [11\bar{1}][2\bar{3}1] \) and the outgoing \( [11\bar{1}][2\bar{3}1] \) slip systems (see Eq. (2)) indicates a residual Burgers vector magnitude of |\vec{b}_r| = 1.28a. From the strain measurements obtained, the strain change across the GB is particularly high |Δε_{GB} | = 1.29%. The results presented clearly indicate that the level of strain heterogeneity (measured as the difference between the averaged strains approaching the GB sides) increases with the residual Burgers vector magnitude |\vec{b}_r|.

4. Discussion

Considerable research efforts have been devoted to incorporating dislocation slip at the crystal level to predict the overall response of metals. Substantial progress has been gained in predicting crystal orientation effects, strain hardening of the slip systems [25], slip–twin interactions [26], and change in crystallographic texture [27,28] upon deformation. Grain boundaries have been treated as a contributor to geometric hardening and the obstacle length has been incorporated in the crystal plasticity models [26]. These models typically allow for predicting the overall macroscopic stress–strain response upon use of various homogenization schemes. Further advances in these models should encompass developments on grain boundary specific effects. The grain boundaries represent different levels of strengthening depending on the incoming and outgoing slip because of the different residual Burgers vectors that remain at the boundary.

The level of strengthening associated with the grain boundaries can be quantified by measuring the energetics of the slip transfer process. Using molecular dynamics simulations, Sangid et al. established the energy barrier levels for different grain boundaries.
in fcc metals. In particular, analyzing different twin and grain boundaries, the authors show that high energy barriers for slip transmission correspond to high residual Burgers vector magnitudes. The influence of the grain boundary specifics on the slip transmission–GB interactions can also be experimentally studied on the mesoscopic scale, since the slip transmission process influences the local deformation behavior at the grain boundaries. In that regard, the results reported in this work present a unique characterization of the local strain fields across grain boundaries. The high strain discontinuities measured for certain grain boundaries derive from the difficulty for slip to propagate through the grain boundaries. In these cases, the activation of the slip system on the second grain (outgoing slip system) is paired with an high residual Burgers vector magnitude of the dislocation that remains on the grain boundary. The largest contribution to the strengthening is then provided by the grain boundaries that display high strain discontinuities, concomitant with the high energy barriers that are measured for slip dislocation–GB reactions and high residual Burgers vector magnitudes [12]. The local strain heterogeneities across the grain boundaries have also a paramount importance in damage nucleation. Damage nucleation has been observed mostly at grain boundaries where imperfect slip transfer takes place and strain heterogeneities arise; then in locations where high local strains are measured [29,30]. It is therefore worthwhile to recognize the correlation between the grain boundary specifics with the measured local strain heterogeneities, and thus identify the potential damage initiation sites.

As shown along the present work, grain boundaries are characterized with the associated residual Burgers vector following slip dislocation–grain boundary interaction. The concept of residual Burgers vector has been established in earlier works and its importance is well known in the materials science community [2,31]. In particular, in many experimental studies different authors already recognize the importance of the residual Burgers vector in characterizing the slip process through grain boundaries [4,5,12,32]. What has been lacking is a quantitative illustration of the link between the residual Burgers vector and the local strain fields, particularly for a bcc material. This became possible with the development of digital image correlation techniques, and special codes in this study, written for the purpose of analyzing the strains along the slip lines as the slip approaches the boundary, and emanates or gets blocked at the boundary. Using this methodology, in Section 3 we provided different types of strain fields across selected GBs which display different residual Burgers vector magnitudes.

Along the work we make the fundamental assumption that the strain fields are directly correlated with the mechanism of interaction. Other potential sources of strains across GBs are then neglected. For example, elastic grain anisotropy induces the generation of compatibility stresses which can activate additional slip systems in the proximity of the GBs [1]. In some cases these contributions have to be considered in the analysis. In the present study, however, we only refer to the strain heterogeneities induced by partial dislocation transmission (including blockage and full transmission cases), since for the material considered only one a single slip system was active across each analyzed GB. In the present treatment we observe that the accumulation of residual dislocations at the grain boundary induce a strain discontinuity across the interface that is proportional to the \(|\vec{b}\!|\) magnitude. In case of full transmission, the Burgers vector conservation implies that no residual dislocations are left at the GBs [2], for this case no substantial strain gradients are expected to occur across the grain interface. In our experiments, for GBs displaying low \(|\vec{b}\!|\) we observed that the measured strains vary almost unaltered across the interface (Fig. 7a, \(|\vec{b}\!|=0.16a\)). Low \(|\vec{b}\!|\) GBs are then expected to provide the lowest strengthening contributions. In the general case of partial transmission of dislocations, a residual dislocation is left on the grain boundary, and another dislocation glides in the other grain. The presence of residual dislocation at GBs creates dislocation impingements on the side of the incoming dislocations. In this case, higher local strains are expected on that side of the GB, generating a general discontinuity in the strain field. For these common cases characterized by intermediate residual Burgers vector magnitudes \(|\vec{b}\!|=0.2a\) to 1a, the measured strains accumulate on one side of the grain boundary depending on the \(|\vec{b}\!|\) magnitude (Fig. 8a, \(|\vec{b}\!|=0.28a\)). The highest values of strain change across GBs are then measured for the largest residual Burgers vector magnitudes \(|\vec{b}\!|>1a\) (Fig. 9a, \(|\vec{b}\!|=1.14a\), and Fig. 10a, \(|\vec{b}\!|=1.28a\)). These results can be utilized to illustrate the significant role that grain boundaries play in the slip transfer process, in particular they can be useful in further modeling efforts. The \(|\vec{b}\!|\) can then be used as a parameter to (i) provide an indication of the level of strain heterogeneity at the grain boundary which indicates the possibility to accumulate damage and (ii) quantify the level of strengthening due to the single grain boundary.

We note the judicial choice of FeCr polycrystals with relatively large grain sizes, and most importantly the activation of a single slip system in each grain facilitated such observations. In the presence of two or more activated slip systems and also twinning, the interpretation of DIC strain fields becomes more complex. Multiple slip and twin systems also introduce additional strengthening effects. Our previous experiments conducted on single crystals of the same material and conditions (sample geometry, heat treatment and strain rate) indicate that very low hardening is observed (for low deformations \(\epsilon<3\%) when only one or two slip systems are activated [33]. It follows that the contribution to the hardening observed in the present case (\(h=0.014E\), see Fig. 3) is provided by partial or full blockage of the dislocations at the grain boundaries. Therefore, the isolated single slip system results for the present polycrystal sample shed light into the mechanism very clearly, hence represent unique findings in this work.

The strain resolution adopted by means of the DIC ex situ set-up offers the possibility to capture the strain heterogeneities introduced by the single slip trace. Experimentally, these results were achieved by the utilization of very fine speckles (which enabled to acquire images at resolutions up to 0.18 \(\mu\)m/pixel) and using a polycrystal material with a large average grain size (about 1 mm). The adopted experimental set-up can also be used for materials with smaller grain sizes. Further advancements can also be achieved for materials that plastically deform by twinning. Twin–GB and twin–slip interactions, which provide additional strengthening contributions, can be quantitatively analyzed through the same methodology based on the measure of strain heterogeneities across twin or grain boundaries. An important issue to be addressed is the improved resolution for in situ strain measurements, since for now all the strain fields reported are obtained acquiring images with the optical microscope in the un-loaded condition (ex-situ). The challenge is to reach the same strain resolutions adopted in the ex-situ DIC set up (0.18 \(\mu\)m/pixel) also for the in-situ set-up. Such high resolution real-time strain measurements potentially allow to capture the dynamics of slip and twinning during deformation, and thus provide further insight into validity of the present results at larger deformation levels when multiple slip/twin systems activate simultaneously.

In summary, we illustrated the considerable promise of the digital image correlation method when utilized in conjunction with EBSD in gaining insight on the strain fields at the grain boundaries. Other techniques such as TEM can be used in conjunction with these results presented here as well. These present results allow analysis at the mesoscale allowing a rapid
assessment of the microstructure. The methodology forwarded in this study can be utilized to check the confidence of crystal plasticity calculations as well as the simulations conducted with molecular dynamics methods which provide a better description of grain boundaries.

5. Conclusions

In this study we investigated dislocation–grain boundary interactions for a FeCr alloy using strain fields determined by digital image correlation. Strain fields across GBs provide a direct quantification of the GB capability to transmit or block slip dislocations. The study elucidates the role of the residual Burgers vector magnitude in predicting full/partial slip transmission, or slip blockage. We provide the strain fields across GBs displaying different residual Burgers vector magnitudes. In particular, for low \( |\mathbf{b}_r| \rightarrow 0 \) slip is observed to transmit unaltered across the interface, the resulting strain field is continuous and the GBs are then expected to provide low strengthening contributions. For intermediate \( |\mathbf{b}_r| < 1a \), a step on the strain field is observed on the interface depending on the \( |\mathbf{b}_r| \) magnitude. The strain change represents the slip accumulation on the GB side of the incoming slip system. Finally, for high \( |\mathbf{b}_r| > 1a \), slip accumulates at the grain boundary, and high strains are measured on one side of the grain. The results clearly show a direct correlation of the strain change across the interfaces with the \( |\mathbf{b}_r| \) magnitude thus indicating the possibility to use the \( |\mathbf{b}_r| \) as a parameter for predicting the slip transmission capability of the grain boundaries in a polycrystal material.

Acknowledgments

The work was supported by the National Science Foundation, NSF CMMI-113003. This support is gratefully acknowledged.

References