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Local strain distributions in partially recrystallized copper determined by *in situ* tensile investigation

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**Abstract.** A partially recrystallized copper sample produced by cold-rolling and annealing was deformed *in situ* by uniaxial tension in a scanning electron microscope, and electron backscatter diffraction data were collected before and after deformation to certain strains. The local strain distributions are quantified using digital image correlation. Distributions of the normal strain along the tensile direction ($\varepsilon_{xx}$) are shown in this paper. The recrystallized grains are found to deform more than the remaining unrecrystallized matrix. When $\varepsilon_{xx}$ is averaged along lines perpendicular to the tensile direction, significant variation are observed, which may be related to the local recrystallized volume fraction.

**1. Introduction**

Nanostructured and ultra-fine-grained metals produced by plastic deformations to high strains often have high strength but poor ductility, which limits their practical application. Introduction of recrystallized grains into the deformed matrix to create a ‘composite’ structure has been suggested to improve ductility of such materials. However, to achieve certain ductility, the recrystallized volume fraction ($V_V$) significantly varies for different samples. For example, for copper deformed by dynamic plastic deformation (DPD), the ductility does not improve much until $V_V$ is larger than 80% [1], whereas, if the sample was rolled after DPD, the ductility is observed to improve when $V_V$ is ~20% [2]. The different “efficiency” of the recrystallized material for improving the ductility may be related to heterogeneities within the sample [3], which leads to different local strain distributions. Digital image correlation (DIC) has recently been used to map local strains [4]. The correlation can be based on painted or deposited patterns [e.g. 5], or on samples’ own microstructure [e.g. 6,7]. The latter has the advantage of further linking the strains to the microstructure. In this work, we investigated a ~45% partially recrystallized copper produced by cold-rolling and annealing. The microstructural evolution was followed by electron backscatter diffraction (EBSD) during *in situ* tensile tests, based on which local strain distributions were determined using DIC. The strains in the recrystallized grains and in the remaining unrecrystallized matrix are compared and discussed for understanding the effects of heterogeneities on the local strain distribution.

**2. Experimental**

Oxygen free high conductivity copper (purity 99.95%) was rolled at room temperature to a thickness reduction of 90%, corresponding to a von Mises strain of 2.7. A tensile specimen (shape and dimensions shown in Fig. 1a) was cut from the rolled plate with the tensile...
direction parallel to the rolling direction (RD). The specimen was annealed at 230 °C for 30 min, after which it was 45% recrystallized according to previous EBSD measurements [8]. The specimen was ground and electro-polished before tensile testing, and the sample thickness after electro-polishing was 1.58 mm.

The in situ tensile test was conducted in a TESCAN MIRA 3 XM scanning electron microscope (SEM) using a Gatan Microtest 2000EW tension stage. The cross head velocity was 0.4 mm/min. Before the tensile test, an area of 120×150 µm² was mapped by EBSD on the rolling section (defined by RD and the transverse direction (TD)) using a step size of 0.2 µm. During the tensile test, the deformation was interrupted at several strains, and the same region was characterized by EBSD repeatedly using the same step size.

EBSD maps before and after straining were compared using a MATLAB-based DIC algorithm [9], from which local strains were calculated. Various maps can be generated based on the EBSD data, and we used four types of maps for DIC: band contrast (BC) maps (Figs. 1 b and c), maps colored according to inverse pole figure of RD (IPF-RD maps, Figs. 1 d and e), of TD (IPF-TD maps) and of the normal direction (IPF-ND maps). For each type of maps, displacements at a grid of 59×71 points with a spacing of 10 pixels were determined using DIC, and for each grid point a subset size of 40×40 pixels (8×8 µm²) was used. In this paper, we only compare the map at the maximum load with the map before loading. At the maximum load, the engineering stress is about 330 MPa. The engineering strain is unknown for the in situ experiment, but according to previous conventional tensile tests of a similar sample, the engineering strain is about 10% [8].

3. Results and discussion
3.1. Image correlation and displacements
Fig.1 shows the microstructures at no load and the maximum load. The recrystallized volume fraction in the mapped area is 29%, smaller than the overall Vv (45%), which is related to the
sample heterogeneity. After straining, individual recrystallized grains and subgrains of the unrecrystallized matrix can still be recognized, with only small changes of shapes. The microstructure can thus be used directly for DIC. The intercept length (boundaries >2°) is 2.2 µm for the recrystallized grains when twin boundaries are considered as normal boundary, and is less than 1 µm for the subgrains in the unrecrystallized matrix. Therefore, a subset of 8×8 µm² contains enough features for DIC.

The DIC analysis determines the displacements for 81%, 86%, 78% and 87% of the grid points using BC, IPF-RD, IPF-TD and IPF-ND maps, respectively (see Figs. 2 a and b for the displacement field determined from BC and IPF-RD maps). It is found that: 1) except a few incorrectly indexed points, the displacements calculated from different maps are similar; 2) the unindexed points are not in the same locations. Therefore, it is possible to optimize the displacement field by combining the calculations from the four different maps (i.e. BC/IPF-RD/IPF-TD/IPF-ND maps). For each grid point, if the displacement is indexed only in one type of maps, use it directly, otherwise, use the average value from the two nearest calculations. The combined maps are shown in Fig. 2c, and displacements of 95% grid points are indexed. The displacements are then smoothed using a Gaussian kernel smoother of a size of 5×5 points (Fig. 2d).

Fig. 2 Maps showing displacements along the tensile direction (i.e. RD) at the maximum load. (a) DIC using BC maps, (b) DIC using IPF-RD maps, (c) Combining displacements from all the four types of maps (BC/IPF-RD/IPF-TD/IPF-ND). d) Smoothed displacements. The white parts in a)-c) are unindexed grid points. In b), a few isolated points, which are incorrectly indexed, have totally different colors from their neighboring points. The incorrectly indexed points were removed during the smooth process.

3.2. Strain distribution in the recrystallized grains and unrecrystallized matrix

Strains were calculated by interpolating displacements using bi-cubic finite element shape functions [10]. In this work, we only discuss the normal strains along the tensile direction $\varepsilon_{xx}$, which are shown in Fig. 3. The average $\varepsilon_{xx}$ is 8.8%, close to the macroscopic elongation reported in [8] for a similar sample. This value is also in good agreement with the manual calculations based on the distance change between two points (usually two triple junctions), which confirms that the strain calculated with the automatic DIC method is reliable. The average $\varepsilon_{xx}$ of the recrystallized grains and of the unrecrystallized matrix is 10.6% and 8.0%, respectively, i.e. the recrystallized part has been deformed slightly more. “Hot spots” with strains larger than 15% are observed in the strain map (red parts in Fig. 3). Most of the “hot spots” are associated with recrystallized grains, but a few “hot spots” appear in the unrecrystallized matrix, such as that marked by an arrow in Fig. 3. The “hot spots”, i.e. the highly deformed regions in Fig. 3, tend to be elongated along TD. The $\varepsilon_{xx}$ is therefore averaged along lines parallel to TD. As shown in Fig. 4, the average $\varepsilon_{xx}$ significantly varies along RD. These variations are at some positions closely related to $V_V$, as lines with large average $\varepsilon_{xx}$ also have large $V_V$. However, this relationship is far from complete, and it is not yet clear why some lines with high $V_V$ in Fig. 4 have relatively small local strains (may be related to the fact that $V_V$ is only measured at the surface, while the $\varepsilon_{xx}$ reflects bulk behavior). It is interesting to point out that the “hot spots” in the unrecrystallized matrix
(marked in Fig. 3) locate at lines with high $V_V$, which suggests that the variation of $V_V$ may have strong effects on the local strain distribution, and the “efficiency” of the recrystallized grains may be related to their spatial distribution.

Fig. 3 Maps showing the normal strains parallel to the tensile direction $\epsilon_{xx}$. Boundaries of the recrystallized grains are shown by black lines on top of the strain map. The arrow marks a region where large local deformation occurs in the unrecrystallized matrix.

Fig. 4 $\epsilon_{xx}$ averaged over lines parallel to TD, i.e. perpendicular to the tensile direction. $V_V$ of the lines are also plotted.

4. Conclusions
We investigated the local strain distributions of a partially recrystallized copper sample after in situ tensile deformation. It is demonstrated that microstructures from EBSD maps can be used as input for DIC and for determination of the local strains. The recrystallized grains deform on average more than the unrecrystallized matrix, but some unrecrystallized matrix is also deformed to relatively high strains, which may be related to the recrystallized grains in the vicinity. Local “hot spots” of highly deformed unrecrystallized matrix may be important for the mechanical properties of the sample. Detailed analysis of the link between the microstructure, texture and local strains is planned for further work.

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References
[1] Li Y S et al 2008 Scr Mater 59 475