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INFLUENCE OF THE CRYSTALLOGRAPHY ON THE INCORPORATION OF LATTICE DISLOCATIONS IN GRAIN BOUNDARIES

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I - INTRODUCTION

Under equilibrium conditions, the interaction between a lattice dislocation and a grain boundary free of extrinsic dislocations is most probably controlled by the long range elastic properties of this dislocation so long as the dislocation is situated relatively far from the planar interface [1, 2]. But, when the dislocation meets or is located at the very close vicinity of the interface it is highly likely that crystallographical parameters influence this interaction. Some authors have analyzed the absorption processes implying the dissociation of the trapped lattice dislocation [3] and experiments were generally performed on FCC materials [4, 5].

The main purpose of this study is to determine what are the geometrical parameters of the system constituted by the two crystals and their interface that play a role in the capability of incorporation of a lattice dislocation in a grain boundary in a BCC Fe - Cr alloy. The investigated parameters are:
- the grain boundary misorientation
- the grain boundary plane
- the possible occurrence and, if so, the characteristics of the preexisting extrinsic dislocations (EBGD's)
- the lattice dislocation line and Burgers vector.

Other factors as the grain boundary chemistry, the strain level of the sample, the temperature involved in the interaction between a dislocation and a grain boundary will be analyzed in further work.
II - EXPERIMENTAL

The investigated material is an iron-9% chromium alloy containing 70 ppm of sulfur. Small ingots elaborated by unidirectional solidification were cold rolled by 90%, annealed at 1180°C for 24 hours to get an homogenized austenite. The specimens are cold rolled again by 60%, annealed at 780°C for 72 hours then quenched in order to obtain a well equilibrated α'-microstructure with an homogeneous sulfur distribution checked by the micro-Baumann technique.

Crystallographical analyses by TEM allow us to describe the geometry of the "grain boundary / lattice dislocation" system. The experimental description is obtained with an accuracy of ± 1° on the misorientation and ± 4° on the grain boundary plane. In addition to the previous macroscopic parameters, the deviation $\Delta \theta$ from the closest CSL misorientation $\Sigma$ is given and the grain boundary is classified as special or general according to the Brandon criterion: $\Delta \theta < 15 \Sigma^{1/2}$.[7]

Contrast analyses using the MARUKAWA method permit the determination of the Burgers vectors.[8]

We must emphasize that nearly all the dislocations observed in the thin foil most probably result of the annealing treatment for two reasons:
- Careful handling of the foil is not able to promote plastic deformation of a very ductile material as α'-iron,
- Thinning of the foil is performed so that no hole is formed at its center, thus there is no stress induced by crack.

III - RESULTS AND DISCUSSION

We have analyzed two kinds of situation obtained on annealed specimens without subsequent deformation:
- The grain boundary contains EGBD's and there are lattice dislocations at least in one of the neighbouring crystals.
- The lattice dislocations do not enter the grain boundary.

Table I gives the crystallographical characteristics of the investigated systems.

<table>
<thead>
<tr>
<th>Grain boundary i/j</th>
<th>$\theta / \Sigma$</th>
<th>$\Delta \theta / \Sigma$</th>
<th>G.B. plane in $i$</th>
<th>G.B. plane in $j$</th>
<th>Lattice dislocation</th>
<th>EGBD's</th>
</tr>
</thead>
<tbody>
<tr>
<td>1/2</td>
<td>37° / [772]</td>
<td>0° / $\Sigma$ 9 G</td>
<td>(168)</td>
<td>(747)</td>
<td>No</td>
<td>Yes</td>
</tr>
<tr>
<td>3/4</td>
<td>42° / [621]</td>
<td>0° / $\Sigma$ 23 G G</td>
<td>(211)</td>
<td>(131)</td>
<td>Yes</td>
<td>Yes</td>
</tr>
<tr>
<td>5/6</td>
<td>34° / [961]</td>
<td>0° / $\Sigma$ 27 G G</td>
<td>(111)</td>
<td>(213)</td>
<td>Yes</td>
<td>No</td>
</tr>
<tr>
<td>6/7</td>
<td>49° / [754]</td>
<td>0° / $\Sigma$ 21 G G</td>
<td>(141)</td>
<td>(115)</td>
<td>Yes</td>
<td>No</td>
</tr>
<tr>
<td>8/9</td>
<td>4° / [180]</td>
<td>0° / $\Sigma$ 1 L.A.</td>
<td>(111)</td>
<td>(647)</td>
<td>Yes</td>
<td>No</td>
</tr>
<tr>
<td>8/10</td>
<td>9° / [310]</td>
<td>0° / $\Sigma$ 1 L.A.</td>
<td>(232)</td>
<td>-</td>
<td>Yes</td>
<td>No</td>
</tr>
<tr>
<td>8/11</td>
<td>48° / [331]</td>
<td>0° / $\Sigma$ 29 G G</td>
<td>(031)</td>
<td>(681)</td>
<td>Yes</td>
<td>Yes</td>
</tr>
</tbody>
</table>
Despite the fact that no lattice dislocation is visible in the case 1/2, the two first situations may be analyzed similarly. Figure 1 shows the different dislocation sets which occur in the 3/4 boundary; a detailed analysis of the system is represented on a stereographic plot (Fig. 2). EGBD's may result of the incorporation of lattice dislocations at the annealing temperature. Dislocations of M type in crystal 4 may have slipped in the (110) plane and enter the grain boundary along the direction labelled Z preserving their Burgers vectors \( \frac{1}{2}[\overline{1}11] \) or its opposite. Then the dislocations of type Z may have moved in the grain boundary plane by a glide-climb process to give rise to the \( V_1, V_2, X \) EGBD's families. The position of the lines of the dislocation set labelled \( U \) may be interpreted by the same way.

Figure 2: Stereographic plot of the 3/4 system:

- - - - : 001 trihedron of grain 4
- - - - : slip plane trace
- - - - : GB plane trace

The lattice dislocations labelled M have entered the grain boundary in Z, then have moved in the GB plane to lie along \( X_1, V_1, V_2 \) direction.

EGBD's labelled U may have their origine from lattice dislocations gliding in the (011) plane.

In the last case 8/11, EGBD's lines are the traces of a (211) slip plane of crystal 8 on the grain boundary plane (Fig. 3 a). Note that the mobility in the grain boundary plane and the activity of (211) slip plane support the hypothesis that observed situations result from annealing.

Among the four grain boundaries in which EGBD's do not occur the 8/10 one must be treated apart as the dislocation in crystal 8 lies parallel and far from the grain boundary and simultaneously is parallel to the foil surfaces. This position may result from an equilibrium between the images forces applied on the dislocation arising from the three interfaces. The three other cases require special interest as lattice dislocations meet boundaries without any obvious reaction. These analyses are reported on Table II and TEM micrographs corresponding to the 8/9, 5/6 and 6/7 grain boundaries are given on figures 3b, 4 and 5 respectively.
Figure 1: TEM Micrographs of grain boundary 3/4 with different EGBD's sets (for explanation see figure 2 and text).

Figure 3: TEM micrographs of the 8/9/11 system showing:

a) the incorporation of lattice dislocations in the grain boundary 8/11

b) The no-incorporation of a lattice dislocation impinging on a ledge/dislocation in the grain boundary 8/9 (note the presence of an intrinsic dislocation network in this low angle grain boundary)
TABLE II: Crystallographical analysis of three cases of no incorporation of a lattice dislocation in a grain boundary.

<table>
<thead>
<tr>
<th>Grain boundary</th>
<th>( \mathbf{b}_L )</th>
<th>( \mathbf{u}_L )</th>
<th>Slip Plane (PS)</th>
<th>Geometrical Features</th>
</tr>
</thead>
<tbody>
<tr>
<td>5/6</td>
<td>( \frac{1}{2} [\overline{1}11] _5 )</td>
<td>[251] 5</td>
<td>(\overline{2}11) 5</td>
<td>( U, \angle_{\text{Foil}} = 80^\circ )</td>
</tr>
<tr>
<td>6/7</td>
<td>( \frac{1}{2} [\overline{1}11] _6 )</td>
<td>[574] 6</td>
<td>( [\overline{0}11] _6 )</td>
<td>( U \parallel \text{Foil}, \angle &lt; 10^\circ )</td>
</tr>
<tr>
<td>8/9</td>
<td>( \frac{1}{2} [\overline{1}11] _8 )</td>
<td>[\overline{1}11] 8</td>
<td>(\overline{0}11) 8</td>
<td>( U \parallel \text{Foil}, \angle &lt; 10^\circ )</td>
</tr>
</tbody>
</table>

The equilibrium conditions in the three cases are fulfilled as follows:

- The most possible explanation of the 5/6 and 6/7 configurations is the dissociation of a very small portion of the lattice dislocation impinging on the grain boundary according to:

\[
\mathbf{b}_L + \sum_{\text{invisible}} b_{\text{DG}} = 0
\]

The necessary components for equilibrium at the node point are not resolved as they have negligible Burgers vectors in a general boundary.

- In the last case 8/9, the lattice dislocation terminates on a special low angle boundary more resistant to absorption. Furthermore, a ledge most probably associated to a preexisting EGBD is located to the emergence point of the dislocation in the grain boundary.

The no incorporation of lattice dislocations in the grain boundaries 5/6 and 8/9 may be explained as follows: in the two cases, the effective slip plane is perpendicular to the grain boundary plane, that may induce a stable equilibrium position of a straight dislocation near the interface [9]; furthermore, in the grain boundary 5/6, the lattice dislocation is purely sessile in the grain boundary plane.

But there is no obvious reason for the impediment to incorporation in the case of the interface 6/7. Thus, crystallographical considerations are not sufficient to predict a possible reaction between a lattice dislocation and a grain boundary even for a dislocation piercing the interface. It appears highly likely that the influence of the boundary on the anisotropic elastic field of the dislocation must be taken into account; this is not straightforward in the cases of complex systems "grain boundary/ dislocation" in polycristals.

IV - CONCLUSION

This study is a first attempt to understand the reason why a lattice dislocation enters or not a grain boundary in a B.C.C. metal. Crystallography may explain most cases, thus is probably the preponderant factor but some "special" configurations exist that need a thorough analysis implying elastic considerations.
Figure 4: Bright field images showing the incorporation of a lattice dislocation in the grain boundary 5/6: 
a) \( \mathbf{g} = [121] \) b) \( \mathbf{g} = [721] \)

Figure 5: Bright field images of the "lattice dislocation / grain boundary" 6/7 system: no EGBD's are visible at the emergence point of the lattice dislocation in the interface 
a) \( \mathbf{g} = [200] \) b) \( \mathbf{g} = [712] \)

REFERENCES
3. R.C. Pond, D.A. Smith, Phil. Mag. 36 (1977) 353