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Published in:
Scripta Materialia

Link to article, DOI:
10.1016/j.scriptamat.2018.04.034

Publication date:
2018

Document Version
Publisher's PDF, also known as Version of record

Link back to DTU Orbit

Citation (APA):
In-situ investigation of the evolution of annealing twins in high purity aluminium

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Abstract

With focus on annealing twins, the microstructural evolution of cold rolled high purity aluminium was characterized in-situ during annealing using the electron backscatter diffraction technique. It was found that annealing twins developed during recrystallization. Many but not all of the twins were gradually removed during grain growth. The grain boundary energies of all the boundaries in a network associated with the twins are estimated and reasons why most twins disappear while a few remain are discussed.

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Keywords: Aluminium, Annealing twins, In-situ annealing, EBSD
annealing temperature of 350 °C and a total annealing time of approximately 6 h is illustrated in Fig. 1b. The microstructure of the sample at each annealing step was characterized by EBSD using a step size of 5 μm to compromise between a fast acquisition time (~5 min) and a large sampling area (~1 mm × 0.8 mm). Twin boundaries were identified having a maximum deviation of 1.5° from the Σ3, 60°/〈111〉 misorientation relationship, and are shown as white lines in Figs. 2–4.

Fig. 2a shows an EBSD map of the sample after annealing at 150 °C. Compared to the map in the deformed state, the observed area in the annealed state is shifted about 800 μm to reveal the entire grain D1. Both annealing twins and recrystallizing grains are found to form within D1, whereas the grains D2-D4 have not yet started to recrystallize. Within D1, two twin boundaries named TB1 and TB2 with lengths of tens to hundreds of micrometers, respectively, are observed separating recrystallizing grains. Two “regular twins” labeled T3 and T4 with approximately parallel straight boundary traces have developed within the recrystallized grain R5, with one end of the twins attached to the grain boundary of R5.

Several mechanisms may have led to the formation of these twins and twin boundaries: as can be seen from Fig. 2b, the two grains, R1 and R2, are twin orientation-related; and both grains have a 〈111〉 type misorientation to the deformed matrix. A similar orientation relationship is observed between the two grains R3 and R4 (separated by TB1) and the deformed matrix. In previous works [17,18], nuclei misoriented to the deformed matrix by a rotation around a common 〈111〉 axis have been reported, and in other works [12,13,19], it is observed that when a nucleus is formed, it may continue to grow by twinning. Both these observations may well explain the presence of TB1 and TB2 in our work. However, as each of the recrystallized grain separated by TB1 and TB2 shares a common 〈111〉 twinning plane with the deformed matrix, it is not possible to conclude which grain developed first.

The regular twins, T3 and T4, have an orientation within the spread of the deformed grains, D1 and D2, respectively (see Fig. 2c), and are contained within the recrystallized grain R5. There are also near twin relationships between R5 and both T3 and T4 as well as between R5 and both D1 and D2 (50–52°〈111〉). Because of the similarity in orientation between the twin and the deformed grain, it is very likely that the twins nucleated from the deformed grains and R5 formed by growth twinning.

As the main aim of the present work is to follow the evolution of twins as a function of recrystallization and grain growth, it is an advantage that we in the investigated area have twin boundaries separating recrystallized grain, i.e. TB1 and TB2 as well as boundaries surrounding “regular twins” i.e. T3 and T4.
Fig. 3 illustrates the microstructural evolution of the twin boundaries TB1 and TB2 upon further annealing. It should be noted that after the annealing step of 250 °C for 70 min (Fig. 3b), the local area is completely recrystallized. As Fig. 3 reveals, both twin boundaries TB1 and TB2 are essentially stationary during the annealing, which is as expected. They are however both removed because either one or both of the recrystallized grains separated by those stationary twin boundaries are ‘consumed’ by neighboring grains during grain growth. 

Thereby the total boundary energy (i.e. the energy per unit length of both the grain and the twin boundary) within the area is reduced. The reduction of energy per unit length, $\Delta E$, can be estimated using the following equation [20]:

$$
\Delta E = \sigma_{GB}\Delta l_{GB} + \sigma_{TB}\Delta l_{TB}
$$

where the energy of the grain boundary, $\sigma_{GB}$, and the twin boundary, $\sigma_{TB}$, of pure Al are 0.324 J/m$^2$ and 0.075 J/m$^2$ [20], respectively; and $\Delta l_{GB}$ and $\Delta l_{TB}$ are the change in lengths of the grain boundary and twin boundary, respectively. It is found that the reduction in boundary energy from Fig. 3b to f is $-1.01$ kJ/μm. Energy reduction can thus explain why these annealing twin boundaries are removed. In this respect, it is important to note that in Al the twin boundary energy is $-24\%$ of the high angle grain boundary energy [20], whereas in Cu it is only $-4\%$ [20]. This may explain why it is energetically favorable to keep many twins in Cu, while they are removed in Al during annealing.

Fig. 4 shows the microstructural evolution of the area containing the two twins T3 and T4. The local area near T3 and T4 is completely recrystallized after annealing at 250 °C for 25 min (Fig. 4b). D2 has been replaced by several recrystallized grains (R8–R10). But at this stage none of these grains seems to affect the twins T3 and T4. It is interesting to note that T3 is eliminated during the annealing while T4 actually grows to become larger. To the authors’ knowledge, such a continuous increase in length has not before been reported for annealing twins in Al during grain growth.

By relating to the fix points marked by red crosses in Fig. 4, it is clear that both the grain boundary GB1 and GB2, to which the twins are attached, as well as the tips of the twins migrate significantly during the annealing. The average migration velocities were measured from the...
series of EBSD maps obtained at a given annealing temperature and the results are reported in Table 1. As velocities are measured at different temperatures, activation energies may also be calculated using the equation [21]:

\[
Q = R \left( \frac{1}{T_1} - \frac{1}{T_2} \right) \ln \frac{v_2}{v_1}
\]  

(2)

where \(Q\) is the activation energy, \(R\) is the gas constant, \(T_1\) and \(T_2\) are the absolutely temperatures, and \(v_1, v_2\) are the migration velocities at \(T_1\) and \(T_2\), respectively. The data show a large variation in migration velocity and in activation energy. Some of the calculated activation energies are within the range of 84–142 kJ/mol generally reported for Al in literature [20], but also smaller values and even a negative activation energy is observed. This is not in accord with standard conception. However, synchrotron x-ray measurements have shown that the activation energies for migration of individual boundaries, have a very wide distribution.

In summary, the evolution of annealing twins in 25% cold rolled high purity Al has been characterized in-situ by EBSD. The results show that many annealing twins develop during early stages of recrystallization. Most of annealing twins disappear during grain growth and the migration velocities of both incoherent twin boundaries and individual grain boundaries vary a lot giving rise to a wide distribution of activation energies. One annealing twin formed on a boundary with a large curvature is observed to grow to a large size, resulting in a local boundary energy increase associated with the growing twin. However, it is found that the total boundary energies are reduced when all the boundaries in a not too small region around the twin (including neighboring grain boundaries and triple junction) are included in the calculation. For all the observed annealing twins, it is thus concluded that their evolution is strongly affected by the migration of the grain boundaries adjacent to the twin boundaries and is governed by a reduction in energy of the total boundary network.

Acknowledgments

This project was financially supported by the National Natural Science Foundation of China (NSFC, funding Nos. 51471039, 51327805 and 51421001). QYH thanks the Graduate Student Research Innovation Project of Chongqing University (No. CYB14007). DJJ thanks for the support of the 111 Project (B16007) by the Ministry of Education and the State Administration of Foreign Experts Affairs of China.

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