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Radiation-induced segregation and corrosion behavior on $\Sigma 3$ coincidence site lattice and random grain boundaries in proton-irradiated type-316L austenitic stainless steel

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Abstract

The behavior of radiation-induced segregation (RIS) and intergranular corrosion at random grain boundaries and $\Sigma 3$ coincidence site lattice (CSL) boundaries in proton-irradiated 316L stainless steel was examined. The frequency of the CSL boundaries was enhanced up to 86.6% by grain boundary engineering treatment prior to irradiation. Significant nickel enrichment and chromium depletion were induced at the random grain boundary owing to the RIS. At faceted $\Sigma 3$ CSL boundaries, chromium depletion occurred at the asymmetrical boundary facet plane whereas no RIS was observed at the coherent twin boundary. After the electrochemical etching test, an intergranular corrosion groove was found along the random grain boundaries because of the low chromium concentration (~12%) at the boundaries. At the faceted $\Sigma 3$ CSL boundaries, the discontinuous groove along the asymmetric facet plane was completely disrupted by the non-corrosive coherent twin boundary.

Keywords

Grain boundary engineering, coincidence site lattice, proton irradiation, radiation-induced segregation, intergranular corrosion
1. Introduction

Solute redistribution in concentrated alloys under irradiation results from the preferential interaction between the solute atoms and excess point defects migrating toward sinks, such as dislocations, surfaces and grain boundaries [1–3]. In austenitic stainless steels, which are used as core materials for the boiling water reactor (BWR) and pressurized water reactor (PWR), radiation-induced segregation (RIS) near the grain boundaries degrades their mechanical and chemical properties significantly [4–7]. The gravity of this problem has been considered as irradiation-assisted stress corrosion cracking (IASCC) caused by the RIS because of the depletion of the chromium concentration near the grain boundaries [8–10]. Recently, the correlation between the RIS and nature of the grain boundaries has been studied for proton- [11] and electron-irradiated [12] austenitic stainless steels. It has been indicated that the magnitude of the RIS depends upon the grain boundary type: significant RIS occurred at random (general) grain boundaries whereas the RIS was suppressed visibly at coincidence site lattice (CSL) boundaries. The relationship between the grain boundary nature and crack initiation by intergranular stress corrosion cracking (IGSCC) has also been discussed by West and Was [13]. They realized that the low angle and \( \Sigma 3 \) CSL boundaries were less susceptible to IGSCC than random grain boundaries in a supercritical water environment.

Regarding localized corrosion on the grain boundaries, many studies have clarified that the susceptibility of grain boundary corrosion depends strongly on the nature of the grain boundary
In general, low angle and CSL boundaries are classified by low energy boundaries while other high angle boundaries are termed random grain boundaries with high boundary energy. It is accepted that the low energy boundaries are highly resistant to intergranular corrosion. Accordingly, grain boundary engineering (GBE) processing has been developed [17–19] to improve significantly the material properties without changing the alloy composition or adding minor elements. Moreover, improvements in the grain boundary corrosion property can also be achieved in the austenitic stainless steels prepared using the GBE treatments [19].

In this study, we have investigated the behavior of RIS and intergranular corrosion on random grain boundaries and Σ3 CSL boundaries in 316L austenitic stainless steel after proton irradiation. A correlation among the RIS, corrosion property and grain boundary nature was established by transmission electron microscopy (TEM) and electrochemical etching. Availability of the GBE treatment for intergranular corrosion during irradiation will be discussed.

2. Experimental procedure

2.1 Sample preparation

The composition of the type-316L austenitic stainless steel is given in Table 1. The procedure of GBE for the 316L stainless steel was as follows. As-received 316L stainless steel was solution-annealed at 1373 K for 30 minutes. The sample was then subjected to 1% cold rolling and
finally was annealed again at 1260 K for 72 hours to increase the frequency of the CSL boundaries to over 80 %. Electron backscatter diffraction (EBSD) patterns were acquired using a scanning electron microscope (SEM; Philips XL-3) equipped with an EBSD instrument to confirm the CSL boundary frequency in the GBE sample. The sample was then polished mechanically to a thickness of 0.15 mm. The foil sample was punched out onto TEM disks of 3 mm diameter and electropolished in a 5 % HClO₄ + 95 % CH₃COOH solution.

2.2 Proton irradiation

Proton irradiation was carried out for the foil sample using a multi-beam high-voltage electron microscope (MBHVEM; JEOL JEM-ARM-1300) equipped with 300 and 400 kV ion accelerators. The H⁺ ions with acceleration energies of 280 kV were irradiated at a dose of 2.7×10¹⁷ ions/cm² at 673 K. The damage profile calculated by SRIM software [20] is shown in Fig. 1. The average irradiation dose was 0.2 dpa at a depth of 250 nm.

2.3 TEM observations

After proton irradiation, the chemical compositions near random grain boundaries and Σ3 CSL boundaries were analyzed using a field-emission gun TEM (FEG-TEM; JEOL JEM-2010F) equipped with an energy-dispersive X-ray spectroscope (NORAN Inc. VOYAGER). After EDS analysis, the sample was etched electrochemically in 10 % oxalic acid solution at a current density of 0.07 A/cm² for 1 minute. The sample was then inserted into the FEG-TEM and energy-filtered TEM
(EF-TEM) images were acquired by Gatan Imaging Filter (Gatan Inc. Model 678) to reveal the mass loss at the grain boundaries after etching.

3. Results and discussion

The grain boundary character distribution obtained by SEM-EBSD analysis of as-received and GBE 316L stainless steels is presented in Fig. 2. The black and gray lines indicate the random grain boundaries and CSL boundaries ($\Sigma \leq 29$), respectively. Prior to the GBE treatment, the frequency of the CSL boundaries was measured to be 64.2% in the as-received 316L. The frequency of the CSL boundaries in the GBE 316L was increased significantly to 86.6% and a network of random grain boundaries was divided perfectly by the CSL boundaries. This is a common feature among the GBE materials which show high resistivity for material degradation along the grain boundaries. Here, the ratio of each CSL boundary in the usual GBE austenitic stainless steel is 75% for the $\Sigma 3$ CSL boundary, 10% for the $\Sigma 9$ CSL boundary and 15% for the other CSL boundaries. This is the reason for us addressing the behavior on the $\Sigma 3$ CSL boundaries. Additionally, the solute concentration profiles across the grain boundaries in the GBE 316L were measured by the TEM/EDX as shown in Fig. 3. No visible segregation was detected on the random grain boundaries and twin boundaries.

A typical microstructure of the proton-irradiated GBE sample is shown in Fig. 4(a). A random grain boundary was stubbed by a $\{111\} \Sigma 3$ CSL boundary (coherent twin boundary) indicated by the
yellow arrow. The measured solute concentration profiles across the random grain boundary and coherent twin boundary are shown in Fig. 4(b) and (c), respectively. Remarkable nickel enrichment and chromium depletion owing to the RIS occurred on the random grain boundary. Particularly, the concentration of the chromium was less than 12 mass % at and near the grain boundary. The RIS of the minor elements (silicon enrichment and molybdenum depletion) was also indicated at the grain boundary. Conversely, no RIS was observed on the coherent twin boundary because of its low sink strength as described in [21].

Almost all of the Σ3 CSL boundaries can be faceted into two individual boundaries. The faceted Σ3 CSL boundary is always altered into a pair of the twin boundary and another asymmetric Σ3 CSL boundary as shown in Fig. 5(a). The solute concentration profiles across the asymmetric Σ3 CSL boundary and the coherent twin boundary are shown in Fig. 5(b) and (c), respectively. Similarly in the above mentioned, no RIS was induced near the coherent twin boundary, whereas remarkable compositional changes were seen at the asymmetric Σ3 CSL boundary. The chromium concentration at the asymmetric boundary was approximately 11 mass %, which was the same level as that at the random grain boundary. This indicates that the asymmetric Σ3 CSL boundary is not classified to a low energy boundary anymore; it should have a much higher grain boundary energy than that in the coherent twin boundary. The matching or indices of the boundary planes is therefore one of the important parameters, as for the Σ value, describing the nature of the grain boundary. In fact, it has
been reported by experimental [22] and theoretical approaches [23] that the boundary energy of the
\{111\} \Sigma 3 CSL (coherent twin) is always lower than that of the \{112\} \Sigma 3 CSL (incoherent twin) and
other CSL boundaries in face-centered cubic (FCC) metals. Thus, the asymmetric \Sigma 3 CSL
boundaries are classified to a high energy boundary, which acts as a strong sink site for the point
defects.

To clarify the relationship between the RIS and intergranular corrosion at the random and \Sigma 3 CSL
boundaries, the electrochemical etching test was applied to the unirradiated and proton-irradiated
samples. The microstructure of a region including the interconnection of a random grain boundary
and coherent twin boundaries in unirradiated GBE316L after the electrochemical etching is shown in
Fig. 6. As shown in the TEM micrograph and the EF-TEM analysis, any visible selective etching on
the grain boundary was not seen in both the random grain boundary and twin boundaries in the
present etching condition. A similar microstructure of the proton-irradiated sample after the
electrochemical etching is shown in Fig. 7(a). Unlike a previous result, it is very obvious that the
random grain boundary was corroded by the etching test while no structural change was observed at
the twin boundary. The EF-TEM image was then acquired to estimate the thickness changes at the
grain boundary. A thickness map near the random grain boundary and corresponding thickness
profile across the grain boundary are given in Fig. 7(b) and (c), respectively. Since a thicker region
appears as a brighter contrast in the thickness map, it is shown clearly that the intergranular
corrosion groove along the grain boundary was formed by the etching. In the thickness profile, the
depth and the width of the groove were measured to be approximately 40 and 10 nm, respectively.
The measured width of the groove was almost equal to the width of the RIS at a typical random
grain boundary, as shown in Fig. 4(b). This implies that the intergranular corrosion is caused by
crchromium depletion along the grain boundary. In fact, we could never find this type of grooving in
the unirradiated sample. Similarly, no groove is observed along the twin boundary because the RIS
never occurs there.

The corrosion behavior along the faceted Σ3 CSL boundary was also investigated as shown in Fig.
8. The evident intergranular corrosion groove appeared along the asymmetric Σ3 CSL boundary with
the same depth and width levels as those in the random grain boundary, however, the groove
appeared to be discontinuous at several places. In these regions, the non-corroded coherent twin
boundaries have always existed between two separated grooves. The length of the coherent twin
boundary shown in Fig. 8 was approximately 50 nm, which is two or three times larger than the RIS
width at the asymmetric Σ3 CSL boundaries. This result is not in conflict with the RIS behavior on
the faceted Σ3 CSL boundary as shown in Fig. 5. It is therefore expected that the penetration of the
corrosion along the Σ3 CSL boundary will be stopped perfectly when the corrosive asymmetric Σ3
CSL boundaries meet the non-corrosove coherent twin boundary. In other words, a faceted Σ3 CSL
boundary effectively prevents corrosion penetration although localized corrosion will be generated
on the one boundary facet plane with an asymmetrical structure.

In Fig. 9(a), we show a panoramic TEM photograph of the proton-irradiated 316L sample after the electrochemical etching test. Some grain boundaries were attacked preferentially because of the chromium depletion caused by the RIS. The interconnections of the corroded and non-corroded grain boundaries are indicated by the green arrows in the photograph and a schematic illustration of the grain boundary network is shown in Fig. 9(b). As predicted from the previous result, the corrosive network along the grain boundaries is disrupted completely by the non-corrosive grain boundaries such as the coherent twin. Although the evaluation of the corrosive network in our study is limited only to the sample surface, it is expected that the network of random grain boundaries is divided perfectly in 3-dimensions by the CSL boundaries in the GBE materials with a CSL frequency of 80 % [24]. Thus, the corrosion propagation through a corrosive network caused by the RIS at grain boundaries would be prevented effectively in the GBE austenitic stainless steels. This will be one of the essential features for the prevention of the IASCC in the BWR and PWR environment. Further experiments with neutron irradiation will be required to confirm the above speculation. This will also help us to understand the effect of the CLS boundaries on crack prevention.

4. Conclusion

By the GBE thermomechanical treatment, the frequency of the CSL boundaries in the 316L
austenitic stainless steel increased from 64.2 % to 86.6 %. Nickel enrichment and chromium depletion owing to the RIS were induced at random grain boundaries and asymmetric Σ3 CSL boundaries by proton irradiation, whereas no RIS occurred at the {111} Σ3 CSL (coherent twin) boundary. After the etching test, intergranular corrosion grooves were present along the random and asymmetric Σ3 CSL boundaries where the chromium concentration was lower than 12 %. At the faceted Σ3 CSL boundaries, the discontinuous groove along the asymmetric facet plane was disrupted completely by the non-corrosive coherent twin boundary, thus the faceted Σ3 CSL boundary would effectively prevent corrosion penetration. It is concluded that the corrosion propagation through a corrosive network caused by the RIS at grain boundaries would be prevented effectively in GBE austenitic stainless steel having high dense CSL boundaries.

Acknowledgements

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References


Table captions

Table 1. Bulk composition of 316L austenitic stainless steel in mass percent.

Figure captions

Fig. 1  Calculated damage profile of 280 keV proton irradiation to 2.7×10^{17} ions/cm^2 as a function of depth.

Fig. 2  Grain boundary character distribution obtained by SEM/EBSD for as-received and grain boundary engineered 316L stainless steels. Black and gray lines indicate random and CSL boundaries (Σ < 29), respectively.

Fig. 3  Solute concentration profiles in unirradiated 316L stainless steel across (a) random grain boundary and (b) twin boundary.

Fig. 4  (a) TEM micrograph of grain boundary network consisting of random grain boundary and coherent twin boundary in proton-irradiated 316L stainless steel and solute concentration profiles across (b) random grain boundary and (c) twin boundary.

Fig. 5  (a) TEM micrograph of faceted Σ3 CSL boundary in proton-irradiated 316L stainless steel and solute concentration profiles across (b) asymmetric Σ3 CSL boundary and (c) twin boundary.

Fig. 6  (a) TEM micrograph of random grain boundary and twin boundaries in unirradiated 316L
stainless steel after electrochemical etching, (b) thickness map of random grain boundary acquired by EF-TEM and (c) thickness profile across the boundary.

Fig. 7 (a) TEM micrograph of random grain boundary and twin boundary in proton-irradiated 316L stainless steel after electrochemical etching, (b) thickness map of random grain boundary acquired by EF-TEM and (c) thickness profile across the boundary.

Fig. 8 (a) TEM micrograph of faceted Σ3 CSL boundary in proton-irradiated 316L stainless steel after electrochemical etching, (b) thickness map of random grain boundary acquired by EF-TEM and (c) thickness profiles across asymmetric Σ3 CSL and twin boundaries.

Fig. 9 (a) Panoramic TEM photograph of proton-irradiated 316L stainless steel after electrochemical etching test and (b) schematic illustration of grain boundary network.
Table 1

<table>
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Figure 1

Range of TEM observation and EDS analysis
Figure 2

- **as received 316L**
  - CSLB ratio: 64.2%

- **grain boundary engineered 316L**
  - CSLB ratio: 86.6%
Figure 3

(a) Random GB Twin boundary

(b) Twin boundary

Concentration (mass\%) vs. Distance from GB (nm)
Proton irradiation at 673 K to 0.2 dpa.

Figure 4

(a) Proton irradiation at 673 K to 0.2 dpa.

(b) Concentration (mass%) vs. Distance from GB (nm) for Cr, Ni, Mo, and Si.

(c) Concentration (mass%) vs. Distance from GB (nm) for Cr, Ni, Mo, and Si.
Proton irradiation at 673 K to 0.2 dpa.
Figure 6

(a) EF-TEM image with highlighted random GB and coherent twin.

(b) EF-TEM (thickness map) showing a coherent twin and random GB.

(c) Thickness profiles for coherent twin and random GB.
Figure 7

(a) {111} Σ3 CSLB (coherent twin) random GB

(b) EF-TEM (thickness map)

(c) Thickness (nm) vs. Distance (nm)
Figure 8

(a) (b) EF-TEM (thickness map)

{111} Σ3 CSLB (coherent twin)

asymmetric Σ3 CSLB

(c)

{111} Σ3 CSLB

Asym. Σ3 CSLB
Figure 9

(a) [Images with annotations indicating corrosion]

(b) [Images with annotations indicating corrosion]