Research Article

An Atomic Study of Substructures Formed by Shear Transformation in Cast γ -TiAl

Hanlei Zhang, Hongchao Kou, Xiaolei Li, Bin Tang, and Jinshan Li

State Key Laboratory of Solidification Processing, Northwestern Polytechnical University, Xi'an 710072, China

Correspondence should be addressed to Hanlei Zhang; zhlnwpu@126.com

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Substructures and microsegregation of γ/γ laths are analyzed with HRTEM and HAADF-STEM. Results show that the substructures are generated during evolution of shear transformation on the $(11\bar{1})$ plane of γ lath. At the beginning, shear transformation evolves in a single γ lath, and a superstructure intrinsic stacking fault (SISF) forms in the γ lath. After the formation of the SISF, the shear transformation may evolve in two different ways. If the shear transformation evolves into neighboring γ laths, the SISF also penetrates into neighboring γ laths and a ribbon of SISFs forms. If shear transformation continues to evolve in the original lath, complex substructures begin to form in the original. If shear transformation in the original lath is homogeneous and complete, secondary twin forms which may further grow into twin intersection. Incomplete shear transformation could not form secondary twins but generates a high concentration of planar faults on the $(11\bar{1})$ plane. These planar faults may further penetrate the γ/γ lath interface, grow into adjacent laths, and form a ribbon of planar faults.

1. Introduction

γ -TiAl is an important low density, high strength structural alloy for application at high temperature (up to 900°C) [1, 2], which draws much attention in the field of aeronautics, astronautics, and automobiles [3, 4]. The most important microstructure of γ -TiAl is full lamellar structure, which consists of α_2 and γ laths. Fine, equiaxed full lamellar microstructure can serve excellent overall mechanical property at high temperature. The mechanical property of γ lath is heavily dependent on its substructures, and the most common substructures in the γ lath are superstructure intrinsic stacking fault (SISF), antiphase boundary (APB), and secondary twin [5]. The formation of the SISF and the secondary twin are both on the $(11\bar{1})$ plane, and it is explained by three different theories. Study [6] reveals that SISF and secondary twin form during shear transformation on the $(11\bar{1})$ plane. In a different theory, the SISF and the secondary twin may also be generated by movement of dislocations on the $(11\bar{1})$ plane [7]. Last but not least, SISF and secondary twin could form by subinterfaces. Considering all these 3 relative theories, the formation of SISF and the secondary

twin is a complex process. The complex microstructure of SISF and secondary twin should be characterized and compared, so their formation could be further explored; the growth mechanism of them can thus be studied, and an overall growth mechanism could be proposed.

In the current work, structure and segregation of SISF and secondary twin are analyzed with high resolution transmission electron microscopy (HRTEM) and high-angle annular dark-field scanning transmission electron microscopy (HAADF-STEM). Afterwards, the formation mechanism of SISF and secondary twin is proposed. With these works, the subsequent substructures of SISF that are twin intersection and a high concentration of planar faults are explored. The formation mechanisms of twin intersection and planar faults are thus discussed.

2. Experimental

Chemical composition of the alloy in this work is Ti-48Al-2Cr-2Nb (at%). The ingot was prepared by vacuum arc melting, and then the sample was prepared with investment pattern precision casting. Afterwards, the sample was

FIGURE 1: (a) A TEM graph of continuous γ laths. (b) A HAADF-STEM graph of the same region. (c) A schematic illustration of this region.

annealed in the atmosphere of argon gas at 1380°C for 1 h, followed by furnace cooling. Oxygen concentration of the sample is 200~300 ppm. Microstructure analysis, HRTEM, and HAADF-STEM were conducted with an FEI Tecnai G2 F30 transmission electron microscopy (TEM).

3. Results and Discussion

Figure 1(a) shows the TEM graph of three continuous γ laths, marked as M_1 , T_1 , and M_2 . The γ laths are continuous γ laths in the lateral direction, and no α_2 lath forms between the γ laths. The interfaces between the γ laths are γ/γ interfaces. The continuous γ laths form because, in γ -TiAl alloy, the fraction of γ lath (90 vol%) is much higher than α_2 lath. To have better understanding of the crystallographic relationship between the γ laths, fast Fourier transform (FFT) patterns of the three laths are obtained, shown in Figures 1(b), 1(c), and 1(d). Because of the symmetry in Figures 1(b), 1(c), and 1(d), it can be determined that M_1 , T_1 , and M_2 are true twins. The twinning plane is (111) plane.

Figure 2(a) shows a local HRTEM graph of the three γ laths shown in Figure 1(a). Figure 2(b) shows a HAADF-STEM graph of the same region. Figure 2(c) is a schematic illustration of the same region. When the γ laths grow in

the lateral direction, the stack sequence may change (from ABC to ACB), so twinning occurs. Twinning is an important mechanism for lateral growth. For the three twin laths, a SISF forms on the $(11\bar{1})$ plane of T_1 . Driving force of the SISF is internal stress generated by phase transformation $\alpha \rightarrow \gamma$ [8]. On the T_1/M_2 interface, lattice distortion can be observed. On the M_1/T_1 interface there is almost no SISF, and this interface is almost a perfect twin interface. Figure 2(b) is a HAADF-STEM graph, so it manifests segregation of Ti and Al in the three laths. In Figure 2(b), the red region represents high Ti/Al ratio while the green region represents low Ti/Al ratio. As can be seen, in most parts of the γ lath, it is homogeneously green so the Ti/Al ratio is stable. Ti atoms segregate on the SISF and the T_1/M_2 interface, where the Ti/Al ratio is much higher than that inside the γ laths. On the M_1/T_1 interface, there is no clear segregation and the Ti/Al ratio is almost the same as that inside the γ lath. Considering the SISF and distortion (Figure 2(a)), it seems like that the Ti atoms tend to segregate where the lattice structure is changed.

Figure 2 characterized and discussed the structure and segregation of SISF. The formation of SISF is via propagation of $1/6[112]$ Shockley partials [9, 10]. The Shockley partial moves on the $(11\bar{1})$ plane and causes shear transformation on the plane. Because of the shear transformation, SISF

FIGURE 2: (a) A local HRTEM graph of the continuous γ laths. (b) A HAADF-STEM graph of the same region. (c) A schematic illustration of this region.

forms. Movement of 1 Shockley partial can form 1 SISF. Appel proposed that the $1/6[112]$ Shockley partials can propagate by decomposition of $1/3[111]$ dislocations [11]:

$$\frac{1}{3} [111] \longrightarrow \frac{1}{6} [112] + \frac{1}{6} [110]. \quad (1)$$

The reaction takes place on the γ/γ lath interface, where $1/3[111]$ partials are consumed and $1/6[112]$ and $1/6[110]$ partials propagate. In the cast alloy there may not be much $1/3[111]$ partials on the γ/γ lath interface [6], so $-1/3[111]$ partials may form to generate $1/3[111]$ partials. These $-1/3[111]$ partials segregate on the γ/γ lath interface, so there may be lattice distortion, as shown in Figure 2(a). After the reaction, $1/6[112]$ can travel across the γ lath, from one γ/γ interface to another. During its traveling, shear transformation may be caused on the $(11\bar{1})$ plane, and thus SISF may form. The $1/6[110]$ partials, however, could stay only at the nucleation region, so it may segregate and cause lattice distortion. In short, the $-1/3[111]$ and $1/6[110]$ partials

segregate on the nucleation region, and the $1/6[112]$ partials segregate on the other side of the γ lath. The segregation of Ti on the γ/γ lath interface and the SISF may be caused by movement of dislocations. Movement of $1/6[112]$ partials and $1/6[110]$ partials equals insert of (112) or (110) planes [12]. Considering that the (112) and (110) planes are made of one type of atoms, this dislocation movement may increase the Ti concentration on the slide plane, as shown in Figure 2(b). On the M_1/T_1 interface, no dislocation movement takes place, so the chemical composition remains unchanged. A schematic illustration of the reaction and movement of dislocations is shown in Figure 3.

The SISFs formed on the $(11\bar{1})$ plane may later penetrate into adjacent γ lath and engender SISFs in the adjacent lath, so a continuous ribbon of planar faults may exist and penetrate a few γ/γ lath interfaces, as shown in Figure 4(a). The ribbon is made of many planar faults, so it is accompanied with shear transformation on the $(11\bar{1})$ plane when it penetrates the γ laths and γ/γ lath interfaces. This shear transformation is

FIGURE 3: A schematic illustration of the reaction and movement of dislocations.

FIGURE 4: (a) A HRTEM graph of a ribbon of planar faults. (b) A magnified HRTEM graph of the upper γ/γ lath interface. (c) A magnified HRTEM graph of the lower γ/γ lath interface.

large, that the γ lath is deformed and the γ/γ lath interface is broken into two pieces. Considering this, lattice distortion can be observed on the γ/γ interface. Figures 4(b) and 4(c) show the lattice distortion upper and lower to the ribbon. As shown in Figure 3, the upper distortion is generated by $-1/3[111]$ partials, so it is perpendicular to the interface; the lower is generated by $1/6[110]$ partials, so it is parallel to the interface. As discussed above, $1/6[112]$ is the most common shear transformation on the $(11\bar{1})$ plane, and it could form SISF on the $(11\bar{1})$ plane. However, other shear transformations may also take place, that is, 2 times of 3 times of $1/6[112]$, and thus form more complicated planar faults.

After formation of the ribbon, shear transformation may further develop and more complicated substructure

could form. Figure 5 shows two major possible substructures: twin intersection and high concentration of planar faults. Figure 5(a) illustrates a SISF formed on the $(11\bar{1})$ plane. A $1/6[112]$ shear has already taken place to form the SISF, so it could act as nuclei for further shear transformation [13]. If the following shear is complete and homogeneous, secondary twin may form on the $(11\bar{1})$ plane. Yoo and Hishinuma [14] calculated that perfect secondary twin could form by homogeneous shear of $2/3[112]$, namely, 4 times of $1/6[112]$ shear transformation. The formation of secondary twin may cause lattice distortion inside the γ lath. This distortion may be accommodated by glide of perfect $1/2\langle 110 \rangle$ dislocations [6]. After formation of the secondary twin, it may penetrate into adjacent γ laths. Since the adjacent γ has a true twin

FIGURE 5: (a) A schematic illustration of a SISF formed inside a γ lath. (b) A schematic illustration and a TEM graph of the twin intersection. (c) A schematic illustration and a HRTEM graph of a high concentration of planar faults.

relationship with the original lath, the twinning system is totally changed. The original $1/6[112](11\bar{1})$ system becomes $1/18[552](11\bar{5})$, which is immobile [15]. Because of this, the adjacent γ lath could provide an efficient barrier to the penetration of secondary twin, and the adjacent lath is thus called barrier twin [16]. Also, the growth of the secondary twin may be deflected at the twin interface, since twinning transformation takes place, and an intersection of twins thus forms. Figure 5(b) shows a schematic illustration and a TEM graph of the twin intersection.

If the shear transformation following the formation of SISF is incomplete or inhomogeneous, a high concentration of planar faults may form based on the SISF. As discussed above, the formation of secondary twin requires 4 times of $1/6[112]$ shear transformation. The 4 times of shear must act on the same $(11\bar{1})$ plane. If they act on different $(11\bar{1})$ planes, it could lead to formation of 4 SISFs. Also, shear of 2-3 times of $1/6[112]$ may also lead to planar faults. The formation of planar faults requires less shear transformation compared with the secondary twin, so the concentration of the planar fault is much higher than that of the secondary twin. The planar fault may also penetrate through the γ/γ lath interface. Because the formation of planar faults requires

less shear transformation, the penetration is much easier compared with the secondary twin, and the high concentration of penetrating planar faults can be also observed on the $(11\bar{1})$ plane of adjacent, twinning laths. The distortion caused by penetration distributes homogeneously on the γ/γ lath interface, so the over configuration of the interface is changed. It is not straight any more, but curved where the distortion is relatively strong. Figure 5(c) shows a schematic illustration and a TEM graph of the penetrating planar faults. As can be seen from the TEM graph, planar faults almost fill up the γ laths.

4. Summary

Shear transformation is a major mechanism for deformation mechanism for γ laths. In the current work, the shear transformation on the $(11\bar{1})$ plane of γ laths and its relative substructures are analyzed.

At the beginning of shear transformation, $1/6[112]$ Shockley partials are generated and SISF forms. As the shear transformation develops, the SISF can act as nuclei of more complex substructures. If the subsequent shear transformation is homogeneous and complete, secondary

twin could form. The secondary twin may penetrate the γ/γ lath interface and form a twin intersection. If the subsequent is inhomogeneous or incomplete, secondary could not form and planar faults form instead. The planar fault may also penetrate the γ/γ lath interface and engender planar faults on the $(11\bar{1})$ plane of adjacent γ laths.

Conflict of Interests

The authors declare that there is no conflict of interests regarding the publication of this paper.

Authors' Contribution

The authors claim that all authors have participated sufficiently in this work to take public responsibility for it. The authors claim that all authors have reviewed the final version of the paper and approve it for publication. The authors claim that neither this paper nor one with substantially similar content under our authorship has been published or is being considered for publication elsewhere.

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