A MASSIVE TRANSFORMATION MECHANISM IN Ti-48Al ALLOY①

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ABSTRACT The structure transformation mechanism in Ti-48Al(mole percent) intermetallic alloy was investigated when it was cooled rapidly from single α phase region. The results showed that the massive γ was obtained by an $\alpha \rightarrow \gamma$ massive transformation during the quenching process in this alloy. A large number of defects, such as anti-phase boundaries (APBs), stacking faults, microtwins and dislocations, were observed to be within the massive γ. Studies indicate that the massive γ nucleates with coherent interface and grows in a diffusion controlled ledge growth mechanism.

Key words TiAl alloy massive transformation diffusion controlled ledge growth mechanism

1 INTRODUCTION

The massive transformation was investigated intensively in the early 1970s[1,2]. This structure can be formed at a fast cooling rate with a product phase behaving a short and coarse morphology (in a massive shape), so this transformation is called massive transformation[3]. In the past, the massive microstructures were found and studied primarily in Cu-Zn, Cu-Al, Cu-Ge, Ag-X, Fe-Ni, and Ti-X alloy systems etc[4,5]. Unfortunately, the mechanism of nucleation and growth of massive phase is far from being of the same view[5-7]. In this study, the formation of the massive microstructure in TiAl alloy was investigated, and the massive transformation mechanism was discussed based on the experimental observation and result analysis.

2 EXPERIMENTAL PROCEDURES

An alloy of nominal composition Ti-48Al used in the tests was prepared by plasma arc melting using 99.8% Ti sponge and 99.99% high purity Al. The buttons were inverted three times during melting and then homogenized at 1473 K for 48 h followed by slow cooling. The samples with 5 mm in diameter and 8 mm in length were spark-cut from the buttons.

Samples for heat treatment were wrapped in tantalum foils and encapsulated in quartz capsules to a vacuum of about 1×10^{-3} Pa and back-filled with argon to 1×10^{5} Pa. They were solution-treated at 1683 K for 1 h, and then quenched into brine, in the mean time, the quartz capsules were broken.

Metallographic samples were examined with optical metallograph, Hitachi S530 scanning electron microscope(SEM), H-800 and Philips CM12 transmission electron microscope(TEM).

Specimens for TEM observation were produced by cutting slices from quenched samples and then ground down to a thickness of about 100 μm. Final electropolishing was performed in a twin-jet polisher using an electrolyte of 65% methanol, 30% *n*-butyl alcohol and 5% perchloric acid, with a voltage 35 V at − 35 ℃.

3 RESULTS

3.1 As-cast microstructure of Ti-48Al alloy

The microstructure of as-cast sample consists of equiaxed grains γ and γ/α_2 lamellae. Structure analysis by X-ray diffraction confirms

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the existence of γ and α_2, as shown in Fig. 1.

Fig. 1 X-ray diffraction spectrum of Ti-48Al alloy

3.2 As-quenched microstructure of Ti-48Al alloy

The as-quenched microstructure of samples solution-treated at 1683 K for 1 h is exhibited in Fig. 2. It can be noted that there are black, deeply etched patchy structures distributing in the bright, featureless matrix. These patchy structures usually nucleate at or near the boundaries, and grow into the grain interior rapidly in the massive way. The detailed observation under the TEM confirms that the dark massive structure is massive γ, while the bright matrix is α_2 ordered structure, and there is no obvious orientation relationship between the massive γ and the surrounding matrix. A large number of defects are found in the massive γ, including microtwins, dislocations, stacking faults and antiphase boundaries (APBs), as shown in Fig. 3. The lattice parameters of massive γ determined by TEM are $a = 4.06$ Å, $c = 4.138$ Å and $c/a = 1.009$, in comparison to that of stoichiometric γ whose lattice parameters are $a = 4.00$ Å, $c = 4.06$ Å and $c/a = 1.015$, which indicates that during the quenching process, equilibrium phases were suppressed due to a fast cooling rate, and replaced by a new phase with a slightly smaller retragonality: massive phase.

The chemical compositions of the massive γ and the surrounding matrix were determined by the electron microprobe in the range of 500 μm perpendicular scanning distance at the two sides of the boundary with 20 μm scanning interval, which indicates that the compositions of these two phases are practically identical, as shown in Fig. 4. This evidence shows that the massive transformation produces a change in the crystal structure without a change in the composition, which indicates that no long-range diffusion of Al atoms occurs during the phase transformation.

4 ANALYSIS AND DISCUSSION

4.1 Nucleation of massive transformation

Massalski[4] pointed out that the massive transformation involved a process of diffusion nucleation and growth, in which the crystal structure of parent phase took a change without change in compositions, the nucleation of the new phase usually occurred at grain boundaries, and the growth proceeded by a thermally activated process. However, Porter[5] argued that the massive transformation was civilian transformation without diffusion. In this study, it was found that in the early stage of the nucleation and growth of the massive transformation, the perfect crystallographic relationship between product phase and parent phase was observed, and the interface between them was confirmed to be coherent, as shown in Fig. 5. Fig. 5(a) illustrates the massive particles at the early stage of growth with an irregular appearance, whose sizes are 20 ~ 500 μm. Massive particles with regular triangular shape were also found, as shown in Fig. 5(b). Selected area diffraction pattern at the interface of the two phases shows

Fig. 2 Optical micrograph of Ti-48Al alloy (1683 K, 1 h, quenched in brine)

Fig. 3 TEM image of Ti-48Al alloy after solution-treatment at 1 683 K for 1 h and quenched in brine

(a) —twins, BF; (b) — Stacking faults, BF; (c) —Anti-phase boundaries, BF; (d) —[110]$_\gamma$ SADP

Fig. 4 Chemical compositions of massive γ and α_2 phases

that these particles yield coherent interfaces and preserve perfect crystallographic relationship with parent phase, as shown in Fig. 5(c)

4.2 Growth of massive microstructure

Fig. 6 illustrates the morphology of the growing head of a massive γ. It can be noted that there exist a large number of ledges at the right and left sides in the head of the massive γ (marked *A* and *B*), whose height is about 5 nm, as shown in Fig. 6(a). A ledge was also found on the right wide face of the massive γ (marked *C*), which is just the edge of an anti-

Fig. 5 Massive γ particles at beginning of growth process

(a) —Free-morphological massive γ particles;
(b) —Regular triangular massive γ particles;
(c) —Diffraction pattern at the interface B~ $[01\bar{1}]\ \gamma \parallel [11\bar{2}0]\ \alpha_2$

phase boundary, so its height is equal to the displacement vector of the anti-phase boundary.

The locally enlarged image of the massive γ at the right-hand head in Fig. 6(a) is exhibited in Fig. 6(c). It can be seen clearly that, during its growth, the γ plate proceeds forward when ledges sweep over the parent phase, and anti-phase domain sections with a series of ledge heights are preserved after the anti-phase boundaries are cut by the ledges, which indicates that the anti-phase boundaries play a significant role on the massive growth.

In addition to the edges of anti-phase boundaries, other defects, such as stacking faults, microtwins and dislocations, were observed to be the sources of the massive growth ledges.

During phase transformation, once the nuclei have been formed, growth takes place by consuming the parent phase through the interface migration. In general case, the growth rate of the product phase depends upon the driving force of the phase transformation(ΔG) and the movement of the thermally activated atoms. The latter is closely associated with the situation of the interface between the product phase and the parent phase. If the compositions of the product phase and parent phase are identical, the growth rate of the product phase is controlled by the short-range diffusion of atoms which transfer across the interface from the parent phase to the product phase.

The classical ledge-growth theory is applied to explain the product phase growth in gas-solid or liquid-solid transformation. In this theory[8], ledges which can grow up are thought to exist on the wide face of product phases, when they precipitate from the parent phase. These ledges are called growth ledges. The terrace which is coherent or semicoherent parallels to the wide face of the product phase. In addition, it consists of structural ledges, misfit dislocations and completely coherent regions. While the riser behaves non coherent interface (see Fig. 7), the atoms' transfer from the parent phase to the product phase only occurs on these risers, and the interface migration takes place by their movement along it. After the risers movement, the interface has migrated in the direction perpendicular to the terrace with a distance equal to the ledge height.

Fig. 6 Morphological image of massive γ plate

(a) —BF; (b) —DF; (c) —Locally enlarged image of massive γ plate

Christian[8] suspected that applying the classical theory describing the gas-solid and liquid-solid transformation to solid transformation might give rise to questions, because growth ledges or superledges with a few or tens of nanometers yield large Bergus vectors, which seems to be impossible in the view of the energy. On the basis of experimental facts, Aaronson put forward a new ledge theory[9]: The riser of the superledge on the wide face of the product precipitate is coherent or semicoherent, which makes the riser behave low interfacial energy. Thus, similar to the terrace, the riser can exist without violating the law of the lowest energy in thermodynamics. Furthermore, because of its low energy, the superledge riser can not migrate by diffusion. A new theory on diffusion controlled ledge growth was also suggested by Aaronson[10], i.e. ledge-

Fig. 7 Schematic model for diffusion-controlled ledge growth of massive γ

ledge or ledge-kink model, in which, the superledge is thought to be comprised of a series of smaller ledges or kinks whose risers are non-coherent, and can migrate by short-range diffusion of thermally activated atoms.

Based on the facts observed in experiment, a ledge growth model suitable for the massive transformation can be suggested as follows.

The wide face and the edge of the massive γ contain a series of ledges with different sizes, when it precipitates. The sources of these ledges are the defects within the massive γ plate, such as the edges of anti-phase boundaries, stacking faults, microtwins and dislocations. These ledges vary from small ledges or kinks to superledges formed by a series of small ledges and kinks piling up. The small ledges or kinks behave non-coherent interface, which can migrate by the short-range diffusion of the thermally activated atoms, and massive γ plate growth takes place by their migration. When their migration is hindered, they will pile up to form superledges. This growth process can be schematically shown in Fig. 7.

During the massive transformation, the density of defects within the massive is extremely high because of a fast cooling rate, especially at anti-phase boundaries produced by the ordering transformation, which suspends in the crystal interior to result in the lattice displacement, which is the primary source of the small ledges and kinks.

Although the classical extinction criterion $\theta = 2\pi \boldsymbol{g}\boldsymbol{R} = 0$ or 2π (θ is the phase factor, $\boldsymbol{g}$ is the reflecting vector and $\boldsymbol{R}$ is the displacement vector) is not satisfied by any superlattice reflection of the $L1_0$ structure, an investigation of the contrast of all the APBs under $\theta = \pi$ suggests that the displacement vector $\boldsymbol{R}$ is $1/2\langle 101\rangle + 5\% 1/2\langle 101\rangle$. So there exists a ledge with the height $|\boldsymbol{R}|$ at the end of APB, which is just the growth ledge of the massive γ. Above all, the massive transformation growth mechanism should be a diffusion-controlled ledge growth mechanism.

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