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Physical metallurgy of single crystal gamma titanium aluminide alloys: orientation control and thermal stability of lamellar microstructure

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Abstract

This paper summarizes our recent work on the development of cast single crystal gamma TiAl alloys (PST) focusing on both lamellar orientation control of PST crystal in process and the thermal stability of the lamellar microstructure in use at elevated temperatures. PST crystals can easily be produced by unidirectional solidification regardless of the kinds of primary solidification phases of bcc β-Ti or hcp α-Ti solid solutions. The lamellar orientation control can be achieved by controlling the orientation of α single crystal existing just underneath the liquid/solid interface. Seeding with Ti-rich PST crystal does not work because of random nucleation of new grains due to the reverse phase transformation of γ to α during heating. However, an Al-rich γ single crystal with no solid/solid phase transformation up to the melting point is a promising seed when the average composition of the seed and alloys to grow is in the region of α solidification. During exposure at elevated temperatures interface reaction of energetically unstable variant interface takes place, resulting in the coarsening of γ plates and eventually leading to the collapse of the lamellar microstructure unless α_2 plates exist. Thus, thermodynamically stable α_2 plates play an important role in pinning the coarsening of γ plates across the lamellae, which is responsible for the high thermal stability of the lamellar microstructure.

Keywords: A. Titanium aluminides, based on TiAl; B. Crystallography; B. Phase transformations; B. Thermal stability; C. Single-crystal growth

1. Introduction

There are two ways to develop γ-TiAl based alloys; one is for wrought to be used below 1073 K, and the other is for cast alloy targeting above 1073 K. Because of light-weight combined with excellent oxidation resistance [1], wrought gamma alloys are promising for applications where heat-resisting steels are currently used. Recently, we have successfully made it possible to hot-forge and toughen the alloys using high-temperature bcc β-Ti phase [2–4], based on our phase diagram study of Ti–Al–M ternary systems [5–7]. The details have been described elsewhere in this issue [8]. On the other hand, the development of cast gamma alloys has been undertaken for more than a decade in order to replace nickel-base superalloys, and only a few polycrystalline alloys have been realized for turbochargers in commercial automobiles [9]. However, further improvement is needed, in order to use the alloys at higher temperatures, such as turbine blades for aircraft jet engine.

The best way to improve the high temperature creep strength is to use single crystals, just like nickel-base superalloys. The fully lamellar single crystal (PST) can be obtained with no difficulty by unidirectional solidification (UDS) [10], and it shows superior creep strength to that of a polycrystalline sample when the lamellar plates are aligned parallel to the loading axis [11,12]. Thus, controlling the lamellar orientation is extremely important from the engineering viewpoint. However, no one has ever successfully established the method to control the lamellar orientation of the PST crystal. In addition, by taking the practical use into consideration, the lamellar microstructure should be maintained during long time exposure at elevated temperatures. PST crystal consists of not only α_2/γ interfaces but also three different types of γ/γ interfaces; variant interface, perfect-twin and pseudo-twin boundaries [13]. Therefore, we have to identify the role of each γ/γ interface as well as α_2/γ interface in the stability of lamellar microstructure.

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In this paper, we summarize our recent studies on the development of cast single crystal gamma alloys from the viewpoints of both lamellar orientation control of PST crystal in process and the stability of the lamellar microstructure in use at elevated temperatures.

2. Lamellar orientation control

Lamellar orientation of a PST crystal can be achieved by controlling the orientation of the α crystal. Fig. 1 schematically illustrates a vertical section of partially melted grown crystal during the UDS process, together with Ti–Al binary phase diagram [14], based on the microstructure analysis of frozen liquid zone of a Ti–48Al PST crystal during crystal growth [15]. The alloy shows β solidification with cellular morphology at the solid/liquid (S/L) interface, and because of the temperature gradient away from the interface partial melting occurs during the UDS process; the β phase transforms to α through the peritectic reaction (L + β → α) by following the orientation of the α phase already grown to form a single crystal in the α single phase region. As the process proceeds, the α crystal is further cooled into the α + γ two-phase region, resulting in precipitation of γ plates with Blackburn's orientation relationship [16], which is responsible for the formation of PST crystal. Thus, even in β solidification alloys, the grown crystal becomes PST.

Seeding by using a Ti-rich PST crystal does not work because polycrystallization takes place due to the reverse phase transformation of γ → α during heating to melt [15]. Fig. 2 shows an outlook of a piece of Ti–48Al PST seed heated up to just above (1673 K) and below (1623 K) the α transus, T_α, at rate of 4 K/min, followed by slow cooling to room temperature [17]. The sample heated below T_α maintains the original crystal morphology with the same lamellar orientation, whereas the one heated above T_α shows several grains with random lamellar orientations. The result clearly indicates that random nucleation of α grains takes place just above T_α. The detail mechanisms of the polycrystallization will be described elsewhere [17]. Since the grown crystal follows an orientation of one of the α grains, it is out of control when polycrystallization of the seed takes place. We found that the polycrystallization caused by the reverse transformation can be avoided at extremely slow heating rate of 0.1 K/min or less, but from engineering viewpoint this is not the way.

Al-rich γ single-phase single crystal seed with no solid/solid phase transformation up to melt (γ solidification), on the other hand, is promising for lamellar orientation control of Ti-rich TiAl alloys when the average composition of the seed and alloys is in the region of α solidification. In this case, the primary α phase formed on the top of the γ seed at the initial stage of the crystal growth follows the seed orientation with $(0001)_\alpha // \{111\}_{seed}$ [15,18]. We have already proven the effectiveness of this method for Ti–48Al PST using a Ti–57Al single crystal seed [15,18]. This seeding method can be extended to multi-component PST crystal. Fig. 3 shows a calculated liquidus projection of Ti–Al–Nb ternary system [19,20], showing the primary α phase region expanding toward equi-aluminum concentration line from the binary edge up to about 15 at% Nb. Thus, there exists a wide composition range shown by hatching where the grown crystal follows the seed orientation if Ti–57Al was selected as a seed. One of the clear evidence is shown in Fig. 4, where Ti–48Al–8Nb with β solidification was grown with 57Al single crystal seed [20]. As shown in Fig. 4(a), there are several lamellar grains formed just above the contact interface between the seed and grown crystals, but a right-hand grain was selectively grown to become a PST crystal. The (111) and (110) pole figures obtained from the growth direction of the seed and PST crystals by EBSD (Fig. 4(b)) clearly show that the grown PST crystal follows the seed orientation parallel to

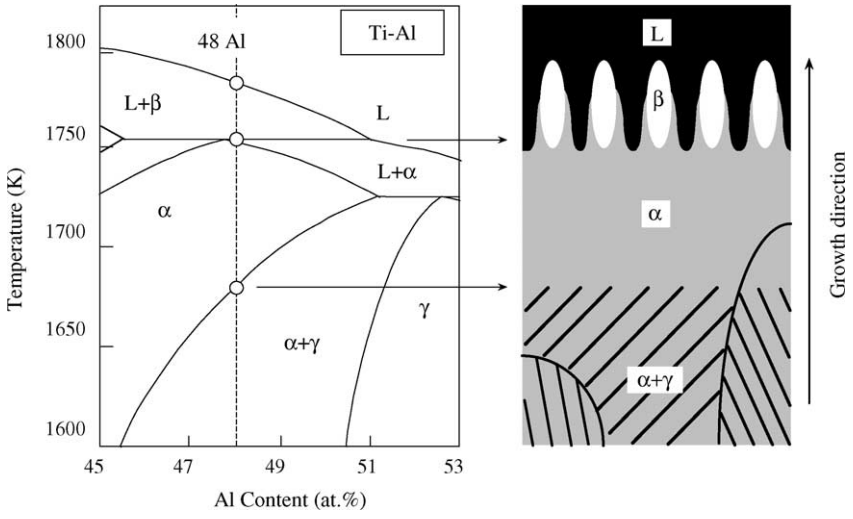

Fig. 1. Schematic illustration of a vertical section of partially melted grown crystal during UDS, together with Ti–Al binary phase diagram.