The phase transformation behavior between $\gamma$ lamellae and massive $\gamma$ in a Ta containing TiAl-based alloy

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The phase transformation behavior between massive $\gamma_m$ phase and $\gamma$ lamellae was studied through interrupted quenching from single $\alpha$-phase region and $\alpha + \gamma$ two-phase region in a Ti–48Al–3Nb–0.5Ta alloy. Massive $\gamma + \gamma_m$ phase transformation temperature $T_0$ is confirmed to be 1400 °C (in $\alpha + \gamma$ two-phase region). Combining with the thermodynamic analysis, it indicates that the formation of $\gamma_m$ phase is gradually inhibited with the decrease of quenching temperature, instead fine $\gamma$ lamellae are preferred. In this case EBSD (Electron back scattered diffraction) and HRTEM (high resolution transmission electron microscope) have shown that besides nucleating on $\alpha$ grain boundary, $\gamma_m$ phase can nucleate on defects in the interior of an $\alpha$ grain with Blackburn orientation relationship or nucleate on fine $\gamma$ lamellae along close-packed (111)$_\alpha$ with orientation relationship of [121]$_\gamma$/ [111]$_\alpha$/ [110]$_\gamma$ at their interface. Furthermore, the ledge growth mechanism of $\gamma_m$ phase is identified.

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1. Introduction

Cast TiAl-based alloys comprising $\gamma$-TiAl (L10) and $\alpha_2$-Ti$_3$Al (D0$_{19}$) intermetallic phases are attractive for applications in aerospace and automotive industries due to their high specific strength at elevated temperatures and a lower density (3.9–4.2 g/cm$^3$, depending on composition and constitution) than polycrystalline Ni-based superalloys [1–3]. Numerous studies have been conducted in the fields of solid-state phase transformation to control the microstructures of TiAl-based alloys [4–7]. The $\alpha_2 + \gamma$ lamellae and massive $\gamma$ ($\gamma_m$) are considered to be the most important phases, the recent interest in these three phases has been driven by the possibility of using the massive transformation to refine the microstructure and thus improve the properties of cast components [8]. To date, the nucleation and growth of $\gamma_m$ phase and $\alpha_2 + \gamma$ lamellae have been recognized [9–14]. For $\gamma$ lamellae, its nucleus is considered to have Blackburn orientation relationship (BOR) with its parent $\alpha$ grain [14,15], i.e., ([0001]$_\alpha$)//([111]$_\gamma$) and $<11\overline{2}0>$$_\alpha$/ $<11\overline{2}0>$$_\gamma$ [16,17]. It is now established that the formation of the lamellae structure does not occur through a eutectoid reaction but results from the precipitation of $\gamma$ lamellae in either a disordered $\alpha$ or an ordered $\alpha_2$ matrix [18]. During the growth process, $\gamma$ lamellae require the diffusion of atoms [19] and still have BOR with their parent $\alpha$ grain. Meanwhile, early works [14,15] had shown that highly-defected $\gamma_m$ phase through a partitionless and diffusionless transformation was initiated by relatively high cooling rates from the single $\alpha$ phase, either nucleating on grain boundary $\alpha$/ $\gamma$ lamellae formed during the first stage of cooling through the $\alpha + \gamma$ two phase field or that it formed directly on grain boundaries [9].

Recently, many studies have already been carried out to investigate Nb and Ta as solute atoms added in TiAl alloys to improve the properties of TiAl-based alloys [5,20–22]. The critical cooling rate to form $\gamma_m$ phase can be lowered by Nb and Ta through slowing down the diffusion controlled growth of $\gamma$ lamellae, and Ta as the slower diffuser than Nb in TiAl from its position in the Periodic Table of Elements, is considered to be the more effective slow diffuser solute atom than Nb to promote the formation of $\gamma_m$s so that Ta increases the range of cooling rates over which even an effectively 100% massive microstructure can be obtained in air cooling condition with addition of 8 at.% Ta, because they can inhibit the growth of lamellae [23,24]. The massive transformation itself is diffusionless, involving atomic movement across a high angle boundary but the fact that nucleation can occur on lamellae (which do require atomic diffusion) in a Nb-containing TiAl-based alloy [9]. It means the cooling conditions to control the formation of
massive gamma and lamellae are complex. Therefore, it is an important aspect to reveal the precipitation behavior of γm phase and γ lamellae of TiAl-based alloys containing heavy elements of Ta and Nb which would influence the final microstructure and thus the mechanical properties of samples.

The aim of the present work was to clarify orientation relationship, nucleation and growth behavior of $\alpha_2$, γ lamellae and γm phase in a Ta-containing medium-Nb-containing cast TiAl alloys which is designed to be used at a temperature of 800 °C or higher. The massive $\alpha \rightarrow \gamma_m$ phase transformation temperature $T_0$ was confirmed. In addition, the morphologies and crystallographic orientation relationships of intragranular $\gamma_m$ phase, γ lamellae and their parent $\alpha_2$ phase were characterized in detail. Finally, the ledge growth mechanism of γm phase was identified.

2. Experiments

An alloy with nominal compositions of Ti–48Al–3Nb-0.5Ta was selected for investigation, which was prepared with commercial purity titanium (99.99 wt%), aluminum (99.99 wt%), Ti–Nb binary alloy (52.47 wt% Nb), and tantalum (99.96 wt%). The button ingots with weight of 30 g were obtained by non-consumable vacuum arc melting. To ensure homogeneity, the ingots were remelted four times. Then samples with sizes of $10 \times 10 \times 10$ mm$^3$ were cut from the ingots. The samples were placed in ceramic crucibles coated with $\mathrm{Y_2O_3}$ and held at 1500 °C (in single $\alpha$ domain) for 40 min under a constant flow of argon. Then they were cooled to 1450 °C, 1410 °C, 1400 °C and 1390 °C at 5 °C/min and held for 20 min respectively, after that the samples were quenched into water.

The microstructures of the samples were observed by scanning electron microscope (SEM) using Tescan VEGA3 LUM equipped with an electron backscattering diffraction (EBSD) system and crystallographic information about $\alpha_2$ and $\gamma$ phases was obtained by EBSD using Channel 5 software from HKL. Small step size (300 nm) was used to be able to measure very minute orientation details. The resolution of EBSD is not able to differentiate between the six variants of $\gamma_m$ phase (which is possible using TEM), neither through the position of their substructure bands nor through the rotation of their Kikuchi patterns (deviate only by 113°) [25]. Instead, one can only differentiate them as a set of two cubic variants, twin related to each other (including true-twin and pseudo-twin). Hence, during EBSD mapping, the $\gamma_m$ phase was considered to be faced-centered cubic (Fcc) crystal, rather than a faced-centered tetragonal.

The elemental distributions of the alloy were observed by a field-emission electron probe microanalyzer (EPMA) using Shimadzu EPMA-1720 operated at an accelerated voltage of 15 kV. The high resolution transmission electron microscopy (HRTEM) and high angle annular dark field (HAADF) investigations were carried out on a FEI Themis Z double Cs Corrector transmission electron microscope operating at 300 kV.

3. Results

3.1. Morphology evolution of $\gamma_m$ phase and γ lamellae

Typical SEM-BSE micrographs of the Ti–48Al–3Nb-0.5Ta alloy taken from samples cooled at 5 °C/min from 1500 °C to 1450 °C, 1410 °C, 1400 °C and 1390 °C holding for 20 min respectively and water quenched are shown in Fig. 1. After quenching from 1450 °C, a large amount of $\gamma_m$ phases are observed as shown in Fig. 1(a) which indicates that the alloy is in the single $\alpha$-phase region at 1450 °C. Our previous study has proved [23] that small amount of (0.5 at%) Ta in TiAl-based alloys can promote the formation of $\gamma_m$ phase after quenching from single $\alpha$-phase region through decreasing the interfacial energy between $\gamma_2$ and $\gamma$. As quenching temperature drops to 1410 °C, it is clear that not only coarse γ lamellae (widths are about 0.5–2 μm) have formed during holding at 1410 °C but also significant amounts of $\gamma_m$ phases (formed during quenching) between these coarse $\gamma$ lamellae are visible (Fig. 1(b)), thus at 1410 °C, this alloy is in the $\alpha$+γ two-phase region. In samples quenched from a lower temperature of 1400 °C, the main difference between the structures is the formation of fine γ lamellae, their widths are only about 10–60 nm (also can be measured by HRTEM image in Fig. 4). These fine γ lamellae are unable to form during holding process at 1400 °C because of the high driving force for them to thicken [19]. Meanwhile, fine γ lamellae formed from quenching are always associated with massive regions that formed in the remaining regions of $\alpha$ grain. This association is obvious in Fig. 1(c) where individual groups of massively transformed γ are visible at the growth front of the fine lamellae. Finally Fig. 1(d) shows a typical microstructure of samples quenched from 1390 °C. The main difference between the structures seen in this sample from that cooled 10 °C higher is that the formation of $\gamma_m$ phase will be completely inhibited, and the lamellae will occupy the whole microstructure.

Therefore, we can find that $\gamma_m$ phase is sensitive to the temperature. $T_0$ (the lowest temperature at which $\gamma_m$ phase could form in $\alpha$-$\gamma$-phase region of the specific composition) [26] is confirmed to be 1400 °C for the Ti–48Al–3Nb-0.5Ta alloy where generating γ lamellae and $\gamma_m$ have the same free energies. In other words, when the quenching temperature is lower than 1400 °C, $\gamma_m$ phase will be completely inhibited, no matter what kinds of cooling process takes place, lamellae would be preferred.

Fig. 2 shows the typical SEM-BSE images and corresponding EPMA maps of Ti–48Al–3Nb-0.5Ta alloy quenched from 1410 °C. According to the EPMA maps, Al concentration in coarse γ lamellae is higher than that in $\alpha$ phase, Ti and Nb concentration in coarse γ lamellae are lower than those in $\alpha$ phase, and the Nb concentration has a homogeneous distribution in coarse γ lamellae, $\alpha_2$ phase and $\gamma_m$ phase. This indicates the formation of γ lamellae requires the diffusion of the elements Ti, Al and Ta. But the composition of $\gamma_m$ phase is completely the same as the $\alpha$ parent grain, which indicates the transformation from $\alpha \rightarrow \gamma_m$ is diffusionless. Therefore, although the γ lamellae and $\gamma_m$ phase have the same L1_0 crystal structure, their compositions are completely different. Moreover, the formation of $\gamma_m$ phase strongly depends on the Al, Ti and Ta partitioning during the formation of γ lamellae. If the remaining $\alpha_2$ phase is heavily depleted in Al (especially lower than 44 at. %) and enriched with Ti and Ta during γ lamellar precipitation, it plays an important role in suppression of massive transformation [27] at the subsequent quenching process.

3.2. Nucleation and growth of $\gamma_m$ phase and γ lamellae

In order to gain further insight into the relationships among $\alpha_2$, $\gamma_m$ and γ lamellae as well as their early stages of nucleation and growth, EBSD technique was used to illustrate the micrographs and diffraction patterns in $\alpha$ phase quenched at $T_q$ 1400 °C which is in accordance with Fig. 1(c). The EBSD image in Fig. 3(a) shows that $\gamma_m$ domains are divided in several $\gamma_m$ subgrains presenting different crystallographic orientations and the size of each $\gamma_m$ domain is very large. The analyzed orientation and their corresponding pole figures (PFs) shown in Fig. 3(b) reveal that in this area only one $\gamma_2$ grain is identified, and (0001) of $\gamma_2$ grain is parallel with [111] of the fine γ lamellae, that BOR relationship (0001) $\gamma_2$// [111]$\gamma$ lamellae and $\{1120\}^{\perp} \gamma_2$// $\{110\}^{\perp} \gamma$ lamellae is strictly obeyed by $\gamma_2$ and fine γ lamellae. Meanwhile, under the cubic assumption described in experiment part, two variants of $\gamma_m$ nuclei: $\gamma_{m1}$ in the area marked as a-1 and $\gamma_{m2}$ in the area marked as a-2 can be identified.
as nucleation stages which are both initiated in the $\alpha_2$ grain. As for $\gamma_m$ nuclei, from its PF we confirm that $\{111\}$/$\{111\}$ $\gamma$ lamellae and $\gamma_m$ do not have BOR with the $\alpha_2$ grain. It indicates that in some cases the specific fine $\gamma$ lamellae can act as the nucleus for $\gamma_m$ since the parallel orientation relationship is identical with that of the fine lamellae. Therefore, $\gamma_m$ nuclei might be expected to have the twin-related relationship with fine $\gamma$ lamellae. As for $\gamma_m$ nuclei, the orientation relationship of $\{111\}$/$\{1000\}$ $\alpha_2$ is confirmed which strictly obey the BOR. It means $\gamma_m$ phase could nucleate directly in $\alpha_2$ grain with BOR and have the same crystallographic relationship with their adjacent $\gamma$ lamellae. More complex orientation relationships can be generated during growth of the $\gamma_m$ phase along $\{111\}$ $\gamma$ which makes various crystallographic orientation of $\gamma_m$ phase.
Typical HRTEM observations are carried out to further characterize the relationship of $\alpha_2$ fine lamellae and $\gamma_m$ phase shown in Fig. (4). In Fig. 4(a), two adjacent $\gamma_m$ variants: $\gamma_m1$ and $\gamma_m2$ are identified. The SADPs taken from the fine lamellar regions indicated that $\gamma_m1$ phase shares a {111} plane with $\gamma_m2$, thereby they have 120° rotational relationship with each other which indicates the growth of $\gamma_m$ phase proceeds by successive twinning steps over {111} planes along all possible six variants.

Meanwhile, the relationships of $\alpha_2$, fine $\gamma$ lamellae and $\gamma_m1$ are shown in Fig. 4(b)(c) by their FFT image. $\gamma$ lamellae have BOR with $\alpha_2$ that [1120] $\alpha_2$/[101] $\gamma$ lamellae, and have relationship with $\gamma_m1$ that [101] $\gamma$ lamellae//[010] $\gamma_m1$. It indicates $\gamma_m1$ has true-twin relationship with fine $\gamma$ lamellae and does not have BOR with $\alpha_2$. These results are in accordance with the EBSD analysis mentioned above that fine $\gamma$ lamellae can act as nucleus as $\gamma_m$ phase. In other words fine $\gamma$ lamellae are the only available nucleation sites for the precipitation of these $\gamma_m$ phases. According to their atomic structure at interface (Fig. 4(d)(e)), $\gamma$ lamellae and $\gamma_m1$ have coherent interface with $\alpha_2$ grain, and they have crystallographic relationship [12T] $\gamma$ lamellae//[111] $\gamma_m1$/[10T] $\alpha_2$ at the interface. Therefore, according to their crystallographic relationship, the possible atomic schematic among these phases are shown in Fig. 4(f). Interestingly, it can be noticed that the growth of $\gamma_m$ phase which is along all possible six $\gamma$ variants is the reason to make $\gamma_m$ phase deviate from BOR with $\alpha_2$ grain.

The atomic resolution HAADF image at $\gamma_m1$ and $\alpha_2$ interface
which is illustrated in Fig. 4(d) is carried out in detail shown in Fig. 5(a). It reveals a darker contrast about 3 atoms layer, indicative of lack of heavy elements along the interface, at this situation, the interface is found to be coherent. In addition, the ledge can be clearly identified at the interface, and each step is accompanied by the direction: ½ 12/T shear vector, and it is seen to be composed of terraces parallel to the (111)γ plane. It is obvious that the atom spacing along the two sides is not the same, thus, these ledges could be formed to reduce the distortion caused by misfit of these two phases.

High resolution EDX of the interface shows that the darker contrast, in the HAADF image, is caused by the lack of heavier element Ti, and other elements Al, Ta and Nb are uniform distributed. Previous studies[13] and our EPMA results have shown that γm phase has the same composition with α2 phase in a large scale, but in atomic scale it is identified by the high resolution EDX that there is no significant difference in composition between the γm and α2, but at the interface, there may exist a lack of Ti area about 3 atom layers.

4. Discussion

4.1. The possible nucleation and growth of γm phase during interrupted quenching

It is encouraging for the meaningfulness of the relating EBSD, HRTEM and HAADF results that the nucleation sites can be measured directly. From the results in the current study and elsewhere[14], three kinds of nucleation sites of γm phase can be concluded: at α2 grain boundaries, in the interior of an α2 grain and at fine γ lamellae. One may recognize that γm phase is easy to nucleate at α2 grain boundaries as the cooling rate is rapid enough, generally the nuclei of γm phase is coherent with its parent α grain in low-index BOR which only spread into neighboring α grains[14]. Secondly, as for γm phase nucleate in the α grain, Dey et al. believed some of intragranular γ grains nucleated with particular α2 twins, in a BOR[10], and the faceted growth of intragranular γm phase is constrained by the surrounding matrix[28]. But our study reveals that intragranular γm are available to nucleate in α2 grain directly with BOR which have the same orientation relationship with γ lamellae. In this connection the inference in earlier work has demonstrated that γ lamellae could nucleate in α2 grain on the defects[19], it is reasonable to assume that direct nucleation visible γm nuclei is also caused by the defects in α2 grain in rapid cooling process. Last, fine γ lamellae can act as nuclei for the γm phase, rather than at the α2 lamellae, which is confirmed to have coherent interface with the α2 grain which is [111]γm/[100]α2. Due to [111] γm1 direction and [110]α2 direction are not exactly parallel with each other in which atomic spacing along these two sides are not completely the same. Consequently, these non-parallelism of [111] γm1 and [110]α2 together with the difference interatomic distances may lead to lattice mismatches. Thereafter the lattice distortion caused by these mismatches could increase the stress at α2/γm interface.

Furthermore, whatever the nucleation process is, further growth of the γm phase always proceeds by successive twinning steps over close-packed {111}γ planes. During this process, twinning or 120° rotational domain may occur on the every six possible {111}γ planes and this would result in the loss of any simple orientation between the massively transformation γ and α2 into which it is growing. This mechanism probably explains the observation that γm phase nucleated on γ lamellae within α2 does not appear to have the BOR with the α2 grain (Figs. 3 and 4).

Although, consensus has been reached that the transformation of disorder h.c.p α phase to f.c.c. γ phase is diffusionless[29], the detail of γm phase growth process along [111] is still subject to controversy. Generally speaking, two mechanisms were proposed but need direct experimental confirmation. The first mechanism suggested of γm phase growth is involves thermally activated short-range jumps of individual atoms across an incoherent interface[29,30]. The second mechanism, namely, the ledge mechanism[28,31], assumed that the structure of the α2/γm interface can be of the terrace-ledge-kink type which contains a number of super-ledges with different sizes. Our present observations directly support the ledge mechanism that ledges are directly identified by the HAADF α2/γm interface image which contains several super-ledges with different sizes and each step is accompanied by one direction: ½ 12/T shear vector (Fig. 5(a)). Due to it is difficult to move an entire Shockley partial-like disconnection instantaneously under rapid cooling condition, the super-ledges are able to form. During the shear process, the atoms at γm interface front are not match very well with parent α phase which needs Ti atoms rearrangement from Ti3Al to TiAl. Although previous studies have shown massive transformation is diffusionless, there will be a composition fluctuation at α2/γm growth interface in atomic scale that the lack of Ti about three atoms layers. The lack of Ti and lattice

Fig. 5. (a) An atomic resolution HAADF image of γm1 and α2 at the interfaces, (b) corresponding elemental maps of Al, Ti, Nb and Ta in the rectangle in (a) by the high resolution EDX.
distortion at α/γm interface would increase the interface energy, in other words, the development of massive transformation can be driven by the motion of interfaces with high energy density.

4.2. Phase transformation behavior between γ lamellae and γm phase

Except the effect of structural features, the temperature is the most important factor to control the formation of γ lamellae and γm phase. To better understand this, consider the energetic model for the nucleation rate of γm phase of γ-TiAl, the nucleation model for the precipitation of γm phase, in the notation of [19,31,32] is:

\[
j_{\gamma_m}^{\nu} = j_{\gamma_m}^{SO} S_V s_a \exp \left( -\frac{\Delta G_{\gamma_m}^{\text{nuc}}(X_a, T)}{kT} \right) \exp \left( -\frac{Q_{\alpha m}}{kT} \right) \tag{1}\]

with

\[
\Delta G_{\gamma_m}^{\text{nuc}}(X_a, T) = j_0^{\gamma_m} \frac{16\pi \sigma^3_{\gamma \gamma}}{3 \left( \Delta G_{V}^{\gamma_m}(X_a, T) \right)^2} \tag{2}\]

where \(j_{\gamma_m}^{\nu}\) is the nucleation rate of γm phase, \(j_{\gamma_m}^{SO}\) is a prefactor expressing in the lineal density of potential nucleation sites and \(S_V\) represents the surface density of grain boundaries, \(s_a\) denotes a weighting factor which accounts for the surface fraction of grain boundary that is still available for nucleation and \(Q_{\alpha m}\) is the activation energy for the atomic mobility. \(\Delta G_{\gamma_m}^{\text{nuc}}(X_a, T)\) is driving pressure of the precipitation reaction, \(j_0^{\gamma_m}\) is the wetting factor of the heterogeneous nucleation, \(\sigma_{\gamma \gamma}\) is the interfacial energy between \(\sigma_2\) and γ nuclei, and \(X\) is the concentration of Al in the matrix.

The model describing the lamellar microstructure is based on a similar approach as the massive model. The nucleation rate is calculated with the following expression:

\[
j_{\gamma_m}^{\nu} = j_0^{\gamma_m} s_a \exp \left( -\frac{\Delta G_{\gamma_m}^{\text{nuc}}(X_a, T)}{kT} \right) \exp \left( -\frac{Q_{\alpha m}}{kT} \right) \tag{3}\]

Where \(j_0^{\gamma_m}\) is a prefactor expressing the number of potential nucleation sites per unit area of grain boundaries. The activation energy for the formation of a stable atom cluster, \(\Delta G_{\gamma_m}^{\text{nuc}}\) is given by:

\[
\Delta G_{\gamma_m}^{\text{nuc}}(X_a, T) = f_0^{\gamma_m} \frac{16\pi \sigma^3_{\gamma \gamma}}{3 \left( \Delta G_{V}^{\gamma_m}(X_a, T) \right)^2} \tag{4}\]

It can be noticed the volumetric driving pressure \(\Delta G_{V}^{\gamma_m}(X_a, T)\) and \(\Delta G_{V}^{\gamma_m}(X_a, T)\) are the most important parameters to control the nucleation rate of γm and γ lamellae. Under a certain composition, these driving pressures are sensitive to the temperature. As for γ lamellae according to Eq. (4), in a higher temperature \(T\), it is obvious that \(\Delta G_{V}^{\gamma_m}(X_a, T)\) would be decreased resulting in the increase of \(\Delta G_{V}^{\gamma_m}(X_a, T)\). As indicated in Eq. (3), the nucleation rate of γ lamellae \(j_{\gamma}^{\nu}\) can be decreased by a higher value of \(\Delta G_{\gamma_m}^{\text{nuc}}(X_a, T)\), which means the lamellae structure is preferred with the decreased temperature, it is also verified by a phase-field study [33]. However, \(\Delta G_{V}^{\gamma_m}(X_a, T)\) which is controlled by diffusionless transformation is also sensitive to the temperature, associated with reported studies in the literature [22], it can be noticed with the decrease of temperature, \(\Delta G_{V}^{\gamma_m}(X_a, T)\) will be reduced, until to \(T_0\) the \(\Delta G_{\gamma_m}^{\text{nuc}}(X_a, T)\) will drop to zero. It means with the decreased of temperature, the γm phase will be gradually inhibited. As a result,

![Fig. 6. The microstructure evolutions of Ti–48Al–3Nb–0.5Ta alloy taken from samples cooled at 5 °C/min from 1500 °C to (a) 1410 °C (b) 1400 °C (c) 1390 °C holding for 20 min respectively and water quenched.](image-url)
the coverage of the nucleation sites by lamellae is not the dominant mechanism for the suppression of the massive transformation. Temperature may be the key to affect the nucleation behavior of \( \gamma \) lamellae and \( \gamma_m \) phase.

In conclusion, as shown in Fig. 6(a) when the holding temperature is at 1410 °C (10 °C higher than \( T_0 \)) in \( \alpha+\gamma \) two-phase region, \( \gamma \) lamellae begin to nucleate and thicken where all the regions that can possibly be reached by \( \gamma \) lamellae growing. With such a definition, after quenching, only \( \gamma_m \) phase is preferred which could well nucleate and grow inside the coarse lamellae colony as \( \alpha \) may be only slightly depleted in Al, enriched in Ti and Ta as illustrated in Fig. 6(a). When the holding temperature drops to 1400 °C at \( T_0 \), the nucleation rate of \( \gamma \) lamellae dramatically increase. The coarse lamellar spacing would be decreased with only 10 °C lower (Fig. 6(b)) and the \( \alpha \) matrix is not much depleted in Al and enriched in Ti and Ta. After quenching, because of the same free energy of lamellae and \( \gamma_m \) phase at \( T_0 \), the driving force for the nucleation of lamellae and \( \gamma_m \) both remains high. Consequently, \( \gamma_m \) phase are available to nucleate together with the fine lamellae and these fine \( \gamma \) lamellae have only limited time to thicken. When the holding temperature drops to 1390 °C (10 °C lower than \( T_0 \)), the nucleation rate of \( \gamma \) lamellae will further increase, parent \( \alpha \) phase transforms progressively into lamellar colonies, these colonies also include some \( \alpha \) phase which is no longer at nominal concentration since \( \gamma \) precipitation depletes \( \alpha \) phase in Al (especially lower than 44 at. %) and enriches \( \alpha \) phase in Ti and Ta, \( \gamma \) lamellae thicken much slower than they lengthen. After quenching, \( \gamma_m \) phase would be completely inhibited by the \( \alpha \rightarrow \alpha + \gamma \) fine lamellae precipitation as shown in Fig. 6(c).

Therefore, \( T_0 \) is determined to be 1400 °C of Ti-48Al-3Nb-0.5Ta alloy, which is much higher than that of binary Ti-47.5 at.%Al alloy [34] and nearly the same with high Nb content about 8 at.% TiAl-based alloy [35]. It indicates only small content of Ta in TiAl-based alloys can increase the phase transformation temperature which has the great potential to refine the microstructure of TiAl-based alloys to improve the mechanical properties.

5. Conclusions

In this study, the precipitation of \( \gamma_m \) phase coexisting with \( \gamma \) lamellae and their orientation relationships in Ti-48Al-3Nb-0.5Ta alloy are analyzed. The following conclusions can be drawn:

1. With the decrease of quenching temperature from single \( \alpha \) domain to \( \alpha + \gamma \) domain, formation of \( \gamma_m \) phase is gradually inhibited, instead \( \gamma \) lamellae is preferred. The massive \( \alpha \rightarrow \gamma_m \) phase transformation temperature \( T_0 \) is confirmed to be 1400 °C. At rapid cooling process, the \( \alpha \rightarrow \gamma_m \) massive transformation is reached as the temperature higher than \( T_0 \) that the \( \gamma \) lamellae have only limited time to nucleate and grow. As temperature lower than \( T_0 \), only lamellae are preferred.

2. Except nucleate from the grain boundary of \( \alpha \) phase, two kinds of intragranular \( \gamma_m \) phase are identified: one can nucleate on fine \( \gamma \) lamellae along close-packed [111]\( \gamma \) which have crystallographic relationship with \( [\overline{2}T \gamma] \) lamellae/[\( \overline{1}T \gamma_m ] \) \( \{\overline{1}0\overline{1}\} \) and a coherent interface, the other one can nucleate in the interior of an \( \alpha \) grain on defects with strictly BOR.

3. The achieved experimental results indicate that the growth of \( \gamma_m \) phase always proceeds over close-packed [111], planes along all possible six variants which reveals a ledge growth mechanism, each step is accompanied by one direction: \([ 1 \overline{1}2 \overline{2}] \) shear vector. During the growth process, there will be a redistribution area at interface with lack of Ti atoms about 3 atom layers.

Author contribution

Rui Hu and Bingqian Xu supervised the research, Keren Zhang performed and analyzed all experiments. Rui Hu and Jieren Yang provide the TiAl-based alloys. Keren Zhang and Bingqian Xu wrote the paper. All authors contribute to the overall scientific interpretation and edited the manuscript.

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