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Evolution of lamellar structure in Ti−47Al−2Nb−2Cr−0.2W alloy sheet

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Abstract: The Ti−47Al−2Nb−2Cr−0.2W alloy sheets were obtained by hot pack rolling. The as-rolled sheet has an inhomogeneous duplex microstructure composed of elongated gamma grains and lamellar colonies. Heat treatments were conducted on the as-rolled sheets. The results show that the microstructures with different sizes and grain boundary morphologies were developed after different heat treatments. A coarse fully lamellar structure can be refined if the heating time, together with the cooling rate, is appropriately controlled. The grain growth exponent is found to be approximately 0.2, and the activation energy of grain boundary migration of the alloy is around 225 kJ/mol.

Key words: TiAl alloy; heat treatment; lamellar structure; grain growth

1 Introduction

γ-TiAl alloys usually contain 45%−48% (mole fraction) aluminum with other elements such as chromium (Cr), niobium (Nb) and tungsten (W). LEE et al [1] revealed that adding Cr can enhance the tensile ductility of TiAl alloys at ambient temperature and improve the oxidation resistance at elevated temperatures. It has also been suggested that the creep and oxidation resistance of the alloy can be improved in the presence of Nb and W [2,3].

γ-TiAl alloys are considered candidate material for high temperature applications in aircraft engines and have already been used in automobile engines [4]. Currently, hot-rolled sheet material for large lightweight structural components in aerospace application is still extensively used [5−8]. This is due to the faith that γ-TiAl alloys have low density (3.9−4.2 g/cm^3), which is half of nickel base alloys, and γ-TiAl alloys possess favorable mechanical properties such as good oxidation resistance, high strength, and good creep resistance at elevated temperatures.

Most γ-TiAl alloys contain two-phase (γ+α) mixtures and some beta-phase particles. The mechanical properties of TiAl alloys are very much dependent on their microstructures. The microstructures can be controlled by appropriate thermomechanical process and heat treatments. There are four specific morphological features in γ-TiAl alloys: near gamma (NG), duplex (DP), nearly lamellar (NL) and fully lamellar (FL) [9]. It has been widely accepted that the near gamma and duplex microstructures have high ductility and low tensile strength. Fully lamellar structures are poor in tensile ductility. However, they tend to have higher tensile strength, fracture toughness, fatigue resistance, and high temperature creep resistance. A fully lamellar structure can be characterized by its colony size, interlamellar spacing, and type of grain boundary. Alloy with fine colony size and interlamellar spacing may have higher tensile ductility and fracture toughness [10]. Also its creep resistance can be improved in the presence of interlocked lamellar grain boundaries [11].

In order to obtain fine and homogeneous fully lamellar microstructure, heat treatment time should be long enough to ensure completely dissolving of the gamma phase, but not so long as to produce serious growth of the α grains. Cooling rate of the TiAl alloys is another important factor for refining the grain size and interlamellar spacing [12]. NOVOSELOVA et al [13] suggested that in order to reduce the lamellar colony size, once the lamellar structure is formed, the cooling rate

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should be as fast as possible.

There has been a great amount of study regarding the effect of heat treatments on microstructural evolutions of TiAl alloys which are mostly cast alloys [14] and forged ones [15], whereas the study on as-rolled TiAl alloy is still relatively rarely reported. The objective of this research is to present the results obtained regarding the effect of heat treatments on the microstructure development and grain boundary morphology in packed hot rolled Ti−47Al−2Nb−2Cr−0.2W alloy sheets.

2 Experimental

The alloy with a nominal chemical composition of Ti−47Al−2Nb−2Cr−0.2W (mole fraction, %) was attained by powder metallurgical method. The powders which passed through a 80 mesh sieve were put into a stainless steel can for hot isostatic pressing (HIP). A 250 mm×300 mm× 150 mm HIPed sample was obtained by HIP (175 MPa, 1250 °C for 2 h). Preforms of the TiAl in size of 40 mm×60 mm×8 mm were cleaned and canned after being cut from the HIPed sample by electro-discharge machining (EDM). The packs were soaked at 1300 °C for approximately 1 h and then hot rolled. Each pack was rolled for 10 to 15 passes, in which the nominal reduction per pass was about 10%. After rolling, the rolled sheets were slowly cooled to room temperature.

The samples in size of 10 mm×6 mm×4 mm were cut from the rolled sheets by EDM for heat treatments. The alpha transus temperature ($\gamma+\alpha\rightarrow\alpha$) of this material was measured to be approximately 1300 °C by heat treatment and optical metallography. In the present work, two heat-treatment procedures were used. 1) The samples were heated to 1310, 1315, 1325, and 1350 °C, then annealed for different time. The heat-treating time at 1310 and 1315 °C were 5, 10, 30, 60, and 120 min; at 1325 and 1350 °C were, 2, 5, and 10 min. After which, all the samples were furnace cooled. 2) The samples were heated to 1310 °C for 5 min, and then quickly transferred to another furnace with a temperature of 900 °C for about 5 min, after which they were air cooled to room temperature.

The microstructure features after heat treatments were examined by optical microscopy. Metallographic samples were ground, polished and etched in a solution of 1 mL HF+3 mL HNO_3+96 mL H_2O. For transmission electron microscopy (TEM), 0.5 mm-thick foils were cut from the samples, and then polished to 80 μm in thickness. Discs with 3 mm in diameter were punched from the polished sections and were further thinned by a twin-jet polishing unit. The colony size was measured by a linear intercept method. In all micrographs, the horizontal direction is the rolling direction, and the vertical direction is the normal direction of the sheets.

3 Results and discussion

The HIPed samples of Ti−47Al−2Nb−2Cr−0.2W have a typical NG structure, consisting mainly of gamma phase with a few α_2 particles dispersing in the matrix. The alloy sheets, after being rolled at 1300 °C without any heat treatments, have an inhomogeneous duplex microstructure consisting of fine elongated gamma grains and lamellar grains with obvious morphological directionality along the rolling direction.

3.1 Grain growth kinetics and thermodynamics

Figure 1 shows the optical microstructures of the as-rolled sheets soaked at 1310 °C for 5, 10, 30, 60 and 120 min followed by furnace cooling. All the microstructures are fully lamellar and with increasing lamellar colony size, the annealing time increases. The average colony size varies from 325 μm at annealing time of 5 min, to 570 μm at annealing time of 120 min. All of the grains became more equiaxial, and the morphological directionality in as-rolled microstructure had been eliminated after the heat treatment.

The fully lamellar microstructure could be obtained by annealing the alloy in the α single phase region and subsequently furnace cooling. During the cooling process, gamma laths nucleate from α grains, and finally fully lamellar structure forms through the ledge mechanism. As shown in Fig. 2, many ledges can be observed in the interfaces between the γ lath and α lath after annealing. These ledges could be generated from Shockley partial dislocations, and their movements led to the growth of the γ lath with reducing of the α phase [11]. Assuming no grain growth took place during cooling process, the lamellar colony size can be presumed to represent the prior α grain size, and the average grain size D can be calculated using the following equation [16]:

$$D = c_1 \times t^n \tag{1}$$

where c_1 is a constant; t is the annealing time; n is the grain growth exponent.

Logarithmic equation (1) can be written as follows:

$$\ln D = \ln c_1 + n \ln t \tag{2}$$

The double logarithmic plot of average grain diameter against annealing time for as-rolled TiAl alloy is shown in Fig. 3. As shown from the results, all the fitted lines for the sheets annealed at 1310, 1315, and1325 °C have a similar linear slope of about 0.2. For ideal pure metal with uniformed grain structure, n can generally be equal to 0.5. If the metal contains impurities, which can hinder the grain boundary from moving, n

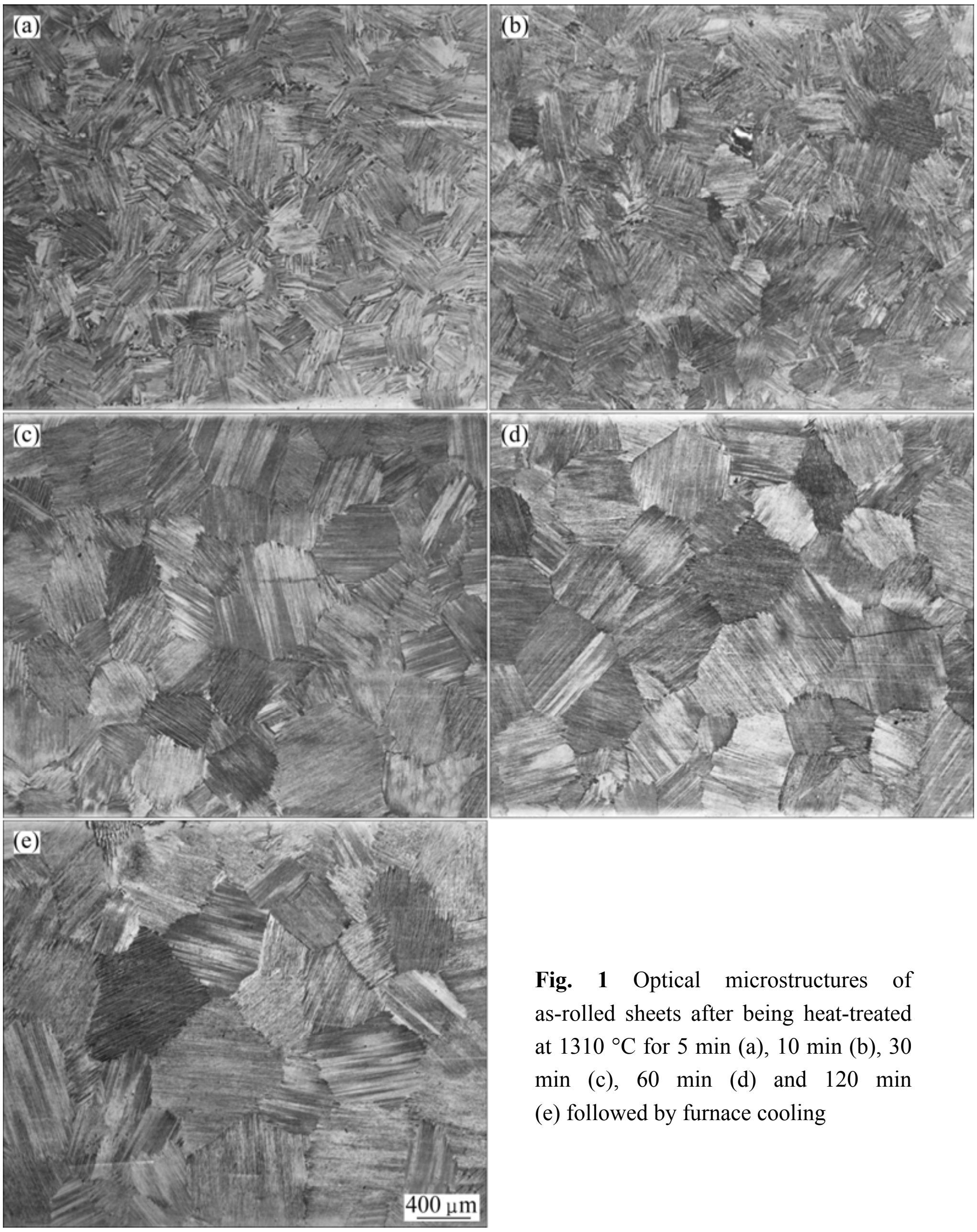

Fig. 1 Optical microstructures of as-rolled sheets after being heat-treated at 1310 °C for 5 min (a), 10 min (b), 30 min (c), 60 min (d) and 120 min (e) followed by furnace cooling

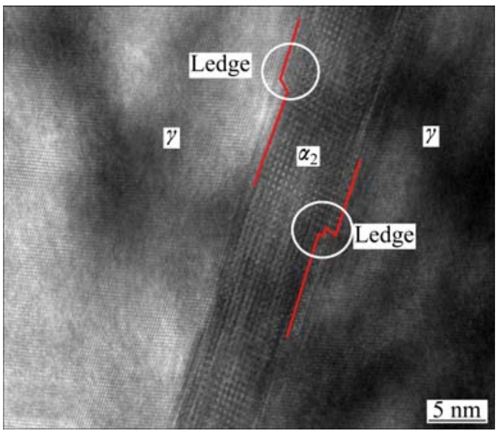

Fig. 2 HREM image showing ledges at lamellar interfaces in as-rolled TiAl sheets after annealing

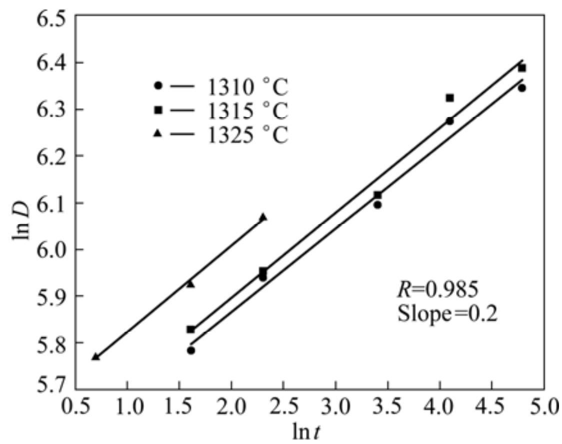

Fig. 3 Double logarithmic plots of average grain size and annealing time for alloy annealed at 1310, 1315 and 1325 °C

may be less than 0.5. Several works had been done to understand the growth of TiAl alloys containing alloying elements such as Nb, W, Cr, and Mn [17,18]. The values of the grain growth exponent in these alloys were regarded to vary from 0.2 to 0.45.

It is suggested that the as-rolled alloys exhibit a smaller grain growth exponent than other TiAl alloys, which indicates a slower rate of grain growth. The rate of grain growth depends on the composition because different alloying elements have different effects on grain growth. The growth rate also depends on the states of the alloys since different states contain different internal energies which are the driving force of grain growth. In the present study, the as-rolled alloys contain high deformation energy and high grain boundary energy. Thus, the α grains will grow dramatically in a short time when annealed in the α single phase, consuming most of the alloy's internal energy, after which the grain growth may slow down in the later annealing period between 5 min to 120 min.

The annealing temperature is also an important factor because the grain growth is a thermally activated process. The higher the annealing temperature is, the faster the grains grow. The relationship between the grains growth and annealing temperature T can be described by the following empirical equation:

$$D = c_2 \times t^n \times \exp\left(\frac{-Q_\mathrm{m}}{RT}\right) \tag{3}$$

where c_2 is a kinetic constant; R is the gas constant; Q_m is the activation energy for grain boundary migration; n is a constant (here n is 0.2).

Equation (3) can be changed into

$$\ln\left(\frac{D}{t^n}\right) = \ln c_2 - \frac{Q_\mathrm{m}}{RT} \tag{4}$$

Putting the experimental data into Eq. (4), we can obtain the result shown in Fig. 4. Q_m can be calculated to be approximately 225 kJ/mol. We know that the self-diffusion activation energy values of Ti in TiAl alloy

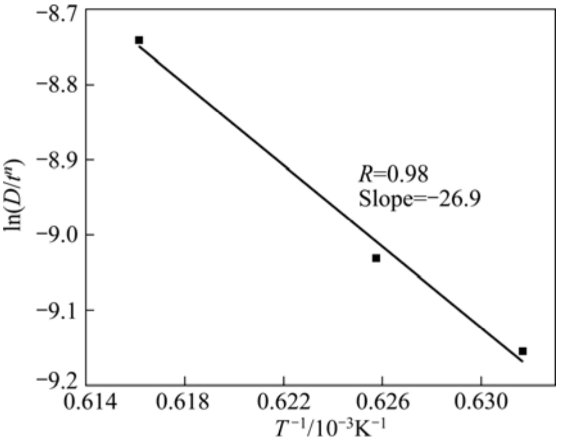

Fig. 4 Plot of $\ln(D/t^n)$ versus $1/T$ with annealing time of 10 min

and Al in TiAl alloy are ~250 kJ/mol [19] and ~360 kJ/mol [19], respectively. The value of activation energy for grain boundary migration is closer to but slightly less than the self-diffusion activation energy value of Ti in TiAl alloy. So the grain growth of α grain may be controlled by the diffusion of Ti.

3.2 Lamellar grain boundary morphology

Figure 5 shows the effects of cooling route on the grain boundary morphology, grain size and interlamellar spacing of the alloy. When the as-rolled alloy was heat-treated at 1310 °C for 5 min, then quickly transferred to another furnace with a temperature of 900 °C for about 5 min, followed by air-cooling, the lamellar structure in Figs. 5(a) and (c) has an average grain size of 165 μm and interlamellar spacing of 200 nm. When the as-rolled alloy was heat-treated at 1310 °C for 5 min and then furnace cooled, the average grain size increases to 325 μm with an interlamellar spacing of 400 nm, as shown in Figs. 5(b) and (d).

The γ laths preferentially nucleate and grow from the grain boundaries, which possess high energy, during cooling process [20]. Fast cooling with large undercooling can apply enough activation energy for γ laths to nucleate at most α grain boundaries. So, the grain boundaries can be kept planar from high temperature to room temperature due to the stability of the lamellar structures. On the contrary, only a part of the sites can be used for γ laths to nucleate and grow into lamellar grains by furnace cooling without the enough activation energy. During this cooling process, the parent lamellar would grow into their neighboring grains which could still be undecomposed α phase, and result in the interlocked structure.

The effects of different annealing time on lamellar grain boundary morphology in as-rolled alloys are shown in Fig. 6. It can be noted from Fig. 6 that the lamellar grain boundary became more planar with increasing the annealing time from 5 min to 120 min. This demonstrates that a longer annealing time is beneficial to obtaining planar lamellar grain boundary, which is similar to study the by PRASAD and CHATURVEDIM [17]. But the reasons for this are still not very clear. DUDZINSK et al [21] suggested that lowering grain boundary energy level is conducive to form planer boundary. If prolonging the annealing time, the α grains will grow with the consumption of internal energy and the grain boundary energy. When an α grain transforms into the lamellar structure while its neighboring grains may also have transformed at the same time, so more energy has been consumed. The whole system would become more stable with a lower system energy. Thus, the alloy annealed for a longer period tends to have planar boundary grains.

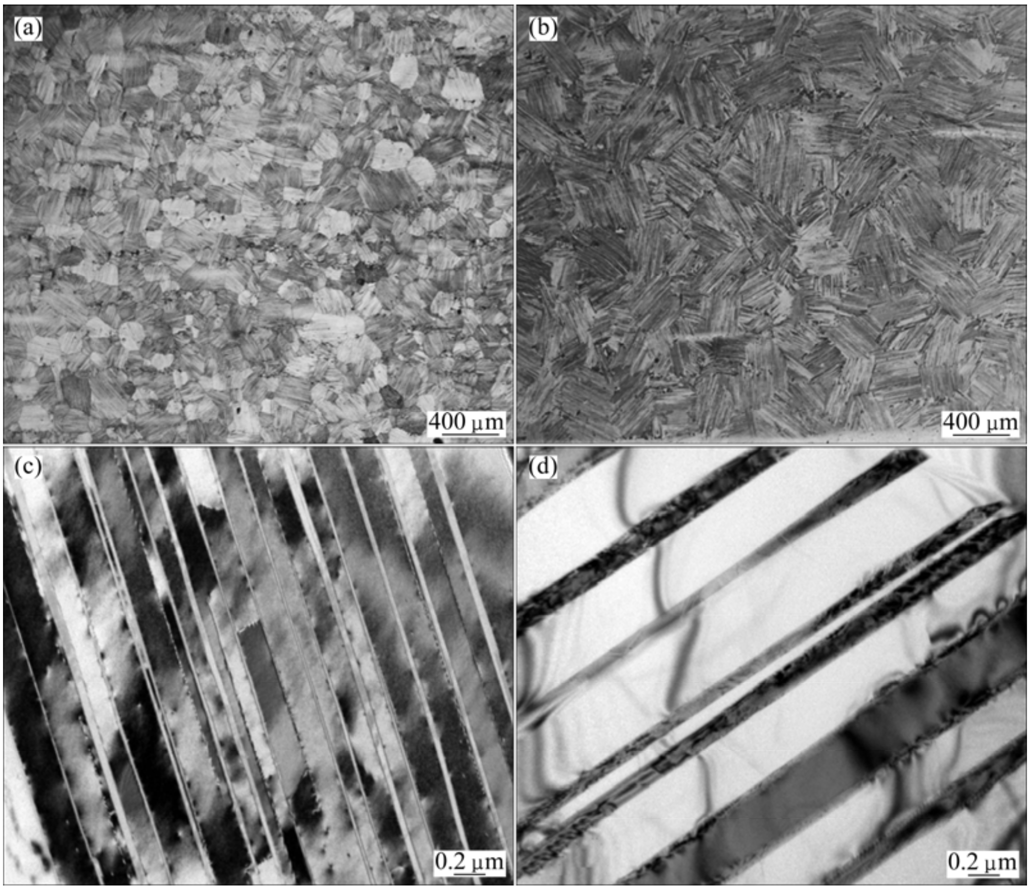

Fig. 5 TEM images (a, b) and optical micrographs (c, d) of as-rolled alloy after heat-treatment at 1310 °C for 5 min followed by different cooling routes: (a), (c) Air cooling; (b), (d) Furnace cooling

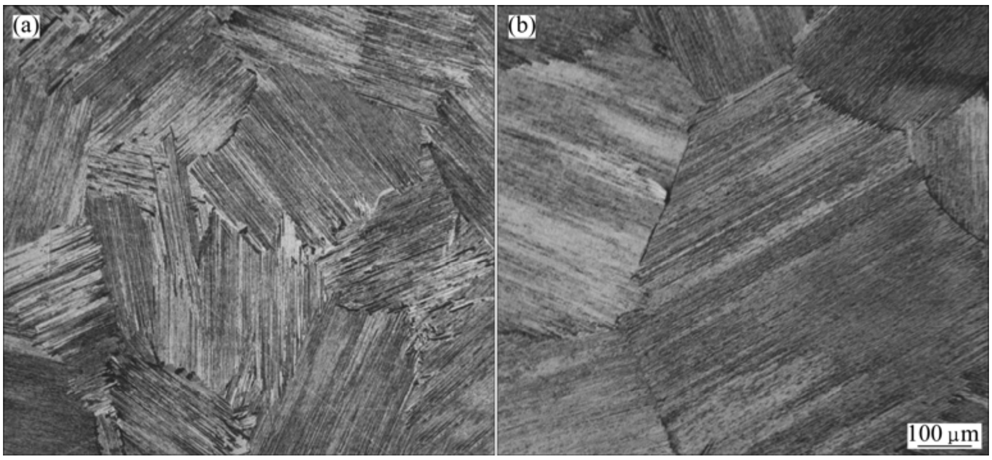

Fig. 6 High magnification optical micrographs of as-rolled sheets annealed at 1310 °C for 10 min (a) and 120 min (b)

4 Conclusions

1) Heat treatment of the as-rolled Ti−47Al−2Nb−2Cr−0.2W alloy sheets in α single phase area leads to the formation of a fully lamellar structure with no morphological directionality. The lamellar grains can be refined via controlling the cooling rate and annealing time.

2) The grain growth exponent of the alloy at temperature around α-transus is approximately 0.2, and the activation energy of grain boundary migration is around 225 kJ/mol.

3) The lamellar grain boundary morphology depends on the cooling route and annealing time. A fast cooling rate and a relatively long annealing time result in

planar grain, while a slow cooling rate tends to form interlocked lamellar structure.

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Ti−47Al−2Nb−2Cr−0.2W 合金板材中层片结构的演化

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摘 要：通过包覆热轧得到 Ti−47Al−2Nb−2Cr−0.2W 合金薄板。板材合金组织是由细小的 γ 相颗粒和层片状的晶团组成的双态组织。然而，热轧后合金板材组织不均匀，沿轧制方向具有明显方向性。研究在 α 相区热处理轧后的板材，结果显示通过不同热处理方式可得到不同晶粒尺寸及晶界形貌的全层片组织；平均晶粒尺寸随着保温时间的延长而增大，晶粒长大因子为 0.2，晶粒长大激活能约为 225 kJ/mol。粗大的层片状晶团可以通过控制冷却方式及保温时间而得到有效细化。

关键词：TiAl 合金；热处理；层片结构；晶粒长大

(Edited by Hua YANG)