Materials Research Express

PAPER

Design of high strength titanium alloy through finding a critical composition with ultra-fine $\alpha$ phase

Li-gang Zhang\textsuperscript{1,2} \textsuperscript{a}, Jing-ya Zhou\textsuperscript{1} \textsuperscript{a}, Zhen-yu Wang\textsuperscript{1} \textsuperscript{a}, Jing-lu Tang\textsuperscript{1} \textsuperscript{a}, Di Wu\textsuperscript{1,3,4} \textsuperscript{b} and Li-bin Liu\textsuperscript{1,3,4} \textsuperscript{b}

\textsuperscript{1} School of Materials Science and Engineering, Central South University, Changsha, Hunan, People’s Republic of China
\textsuperscript{2} Key Laboratory of Non-ferrous Metallic Materials Science and Engineering, Ministry of Education, Changsha, Hunan, People’s Republic of China
\textsuperscript{3} State Key Laboratory of Powder Metallurgy, Changsha, Hunan, People’s Republic of China
\textsuperscript{4} Authors to whom any correspondence should be addressed.

E-mail: wudi30@csu.edu.cn and lbliu@csu.edu.cn

Keywords: titanium alloy, diffusion couple, strength, thermodynamic, secondary $\alpha$ phase

Abstract

This study presents the design of a high strength titanium alloy based on the idea that the $\beta$ titanium alloy with critically stable composition could precipitate ultra-fine $\alpha$ phase. An efficient method was used to find the composition of TC21-xV alloy in which ultra-fine $\alpha$ phase could be obtained. It is found that TC21-5V alloy containing ultra-fine $\alpha$ phase and possessing highest hardness could be obtained after solution treatment in the $\beta$ single-phase region and subsequent aging. The morphology of primary and secondary $\alpha$ phases can be controlled by changing the solution treatment and aging temperatures. Calculation results of driving force and growth rate of secondary $\alpha$ phase show that the growth rate may be a decisive factor for the final size of secondary $\alpha$ phase after solution and aging treatments in the two-phase region. Finally, when the alloy was solution treated at 820 °C and aged at 500 °C, the highest strength of 1540 MPa was obtained. The alloy was solution treated at 820 °C and aged at 600 °C exhibited high strength of 1210 MPa with 11.4% elongation.

1. Introduction

Titanium alloys have been widely used in various industries due to their high strength, low density and strong corrosion resistance. Among them, high-strength titanium alloys play an important role in the manufacture of aerospace structural parts [1, 2]. TC21 titanium alloy is a high strength and high toughness titanium alloy developed on the basis of Ti-62222 alloy in the United States. It has a nominal composition of Ti-6Al-3Mo-2Nb-2Sn-2Zr-1Cr-0.1Si, strength of about 1100 MPa, elongation of about 8%, and fracture toughness of about 70 MPa.m$^{1/2}$ [3–5].

The main reason for the high strength of titanium alloy is the fine secondary $\alpha$ strengthening phase precipitated from $\beta$ matrix. This dispersed $\alpha$ phase forms a large number of $\alpha/\beta$ phase boundaries with $\beta$ matrix, which effectively block dislocation slip and increase the strength of the alloy. A large number of studies show that the strength of the alloy can be effectively improved by reducing the size of secondary $\alpha$ phase and increasing its volume fraction [6–8]. Whether the super-dispersed $\alpha$ structure can be precipitated in the $\beta$ matrix of the titanium alloy largely depends on the stability of the $\beta$ matrix and the precipitation driving force of $\alpha$ phase in the aging stage. The stability of $\beta$ matrix is mainly determined by the content of $\alpha$ stabilizing elements Al, O, etc and $\beta$ stabilizing elements Mo, Nb, Cr, V, Fe, etc [9]. When the content of $\beta$ stabilizing element in titanium alloy is low, $\beta \rightarrow \alpha'$ martensite transformation will take place during quenching. With the increase of the content of the $\beta$ stabilizing elements to a critical value, $\alpha'$ martensite transformation will be inhibited and a metastable $\beta$ phase (with or without very fine $\alpha'$, $\omega$) will be obtained [10, 11]. This metastable $\beta$ phase has very bad stability. In the subsequent aging stage, $\alpha$ phase precipitation driving force will be very high and ultra-fine $\alpha$ phase will be obtained [12, 13]. Therefore, it is feasible to obtain high-strength titanium alloy by designing around this critical composition.

© 2020 The Author(s). Published by IOP Publishing Ltd
It has been found that the microstructure characteristics of the alloy, such as volume fraction, thickness, and aspect ratio of secondary α phase, are very sensitive to heat treatment, and determine the mechanical properties of the alloy. However, it is very difficult to achieve precise control of the microstructure. For example, the secondary α phase is the main factor that determines the strength of the alloy, and the size and volume fraction of the phase is mainly controlled by changing the aging temperature. Appolaira et al [14], assuming secondary α phase mainly nucleates at the grain boundary α, proposed a transformation model of sympathetic nucleation for near β titanium alloys. They also described the growth of grain boundary α with the assumption of no overlapping around precipitates [15]. But it is not clear whether nucleation or growth or other factors are the decisive factors of the final size of secondary α phase.

Based on the current requirements for light weight and high strength of new generation aircraft, the purpose of this paper is to design high strength titanium alloy by relying on the idea that near-β titanium alloy with critical stable composition can precipitate ultra-fine α phase. In this experiment, diffusion couple method is used to obtain a continuously changing alloy composition on a sample. Combined with the morphology and hardness tests, the critical composition of the alloy can be found quickly and efficiently. In the later stage of heat treatment, thermodynamics and kinetics calculation are combined to unravel the factors controlling the microstructure of the alloy. More importantly, the factors that determine the final morphology of secondary α phase are discussed.

2. Materials and methods

TC21 alloy and TC21-20V alloy were used to prepare a diffusion couple. The actual compositions of the two alloys are shown in table 1. The diffusion couple in this experiment was manually pressed with a steel fixture, so there were no specific compaction parameters. Based on the contact surface area of the diffusion couple being 100 square millimeters and the hardness of the two-part alloy, the pressure value was estimated to be between 50 and 100 MPa under the condition that the two-part alloy can be compacted. To fully bond the two alloys, the compacted diffusion couple was placed in a high vacuum (≈10⁻⁴ Pa) annealing furnace and annealed at 1000 °C for 10 h. After that, the pressed diffusion couple was taken out from the jig, the surface oxide scale removed, and it was sealed into a vacuum quartz tube. Then it was annealed again at 1000 °C for 600 h and the elements in the alloy fully diffused. Thereafter, the diffusion couple was solution treated at 1050 °C for 6 h, quenched, and aged at 600 °C for 6 h. JEOL-JSM 7001 F field emission scanning electron microscope was used to observe microstructure of the alloy under different compositions. The hardness of the alloy was measured by a microhardness tester, and finally we determined the TC21-5V, which with finest microstructure and highest hardness under this treatment, as the candidate alloy composition for further research. The α/β phase transition temperature of the TC21-5V alloy was determined by the metallographic method and was approximately 850 °C.

The TC21-5V alloy was double melted by vacuum arc melting. The ingot of 160 mm diameter was forged at around 1050 °C to a rod at a deformation of approximately 50%. The rod was then forged for the second time at around 900 °C to a square rod with dimensions of 1000 × 80 × 40 mm. It was then cut into blocks with dimensions of 80 × 40 × 10 mm and forged for the third time at around 850 °C to a square rod with dimensions of 320 × 10 × 10 mm. The final forging temperature of the alloy is 30-40°C below the α/β phase transition point of the alloy. The heat treatment of the alloy was carried out in a vacuum tube furnace. The samples of the as-forged alloy were treated in a preheating furnace at the α+β two-phase region and the β single-phase region (820°C and 900°C) and held for 0.5 h before air-cooling. The air-cooled (AC) samples were then aged at 500°C and 600°C for 6 h in a preheated furnace and subsequently air cooled. Samples for the optical microscope were mechanically polished by standard metallographic methods. The microstructure analysis of the samples was performed in a JEOL-JSM 7001 F field emission scanning electron microscope and a Tecnai G2 20 ST transmission electron microscope.

The Gibbs free energy of α phase and metastable β phase of TC21-5V alloy at different temperatures was calculated based on the thermodynamic database established previously by our group [16].

In order to study the growth kinetics of secondary α phase in TC21-5V titanium alloy, the relationship between the chemical composition of α phase and β phase and temperature was obtained by using JMatPro10.0

### Table 1. Chemical composition of the alloys to manufacture diffusion couple (wt%).

| Element | Al | Mo | Cr | Nb | Sn | Zr | V | Ti |
|---------|----|----|----|----|----|----|---|----|
| TC21    | 5.97 | 3.01 | 1.12 | 1.96 | 1.83 | 1.92 | 0 | Bal. |
| TC21-20V| 5.92 | 3.11 | 1.06 | 1.87 | 1.88 | 1.93 | 19.24 | 20.26541 Li-gang Zhang et al |
software [17]. Then the relationship between the growth rate of secondary $\alpha$ phase and temperature was obtained.

As mentioned in other literatures, the velocity $v$ of $\alpha/\beta$ interface movement can be expressed as follows [14]:

$$v = \frac{d l}{d t} = \frac{D}{C_{\beta} - C_{\alpha}} \frac{\Delta C}{L}$$  \hspace{1cm} (1)

Where: $L$ is the diffusion distance, $L = K\cdot r$, $K$ is a constant, $K \approx 1$, $r$ is the radius of curvature (it was assumed $r = 1 \mu m$ in this paper); $D$ is the atomic diffusion coefficient; $\Delta C$ is the difference in the content of solute atoms at the interface, $\Delta C = C_{\beta} - C_{\alpha}$.

Because TC21-5V titanium alloy is a multi-component titanium alloy, it contains many solute components, which is more complicated in theoretical treatment. In order to simplify the calculation, the content of the new phase and the equilibrium content of the interface were expressed in terms of molybdenum equivalent, and the expression was [18]:

$$[\text{Mo}]_{\text{eq}} = \%\text{Mo} + \%\text{Ta}/4 + \%\text{Nb}/3.3 + \%\text{W}/2$$

$$+ \%\text{V}/1.4 + \%\text{Cr}/0.6 + \%\text{Fe}/0.5 + \%\text{Co}/0.9$$

Where: $[\text{Mo}]_{\text{eq}}$ is the molybdenum equivalent.

The equilibrium molybdenum equivalent of the metastable $\beta$ phase obtained after quenching of TC21-5V titanium alloy at room temperature was represented by $C_{\beta}$, $C_{\beta} = 7.233\%$ (mass fraction). When the transformation amount of precipitated $\alpha$ phase reaches equilibrium state after isothermal aging treatment, the equilibrium molybdenum equivalent of $\alpha$ phase was expressed by $C_{\alpha}$, and the equilibrium molybdenum equivalence of $\beta$ phase was expressed by $C_{\beta}$.

The relationship between diffusion coefficient and temperature generally obeys Arrhenius equation [14]:

$$D = D_0 \exp \left(-\frac{Q}{RT}\right)$$  \hspace{1cm} (3)

Where: $D_0$ is the diffusion constant; $T$ is the thermodynamic temperature; $Q$ is the diffusion activation energy, $Q = 147.353 \text{ kJ mol}^{-1}$; $R$ is the gas constant, $R = 8.314 \text{ J mol}^{-1} \text{ K}^{-1}$.

Finally, the growth rate of $\alpha$ phase can be calculated by substituting the obtained data into equation (1). In this experiment, the room temperature tensile test was carried out on the MTS810 tensile test system. The test sample was flaky. The dimensions of the sample was according to the national standard GB/T 228–2002 (ASTM E8-01e1), and the tensile speed was 2 mm min$^{-1}$. At least, three samples of each condition were tested for an average value of ultimate tensile strength ($R_m$), yield strength ($R_{p0.2}$), elongation ($A$) and elastic modulus ($E$).

3. Results

3.1. Alloy composition determination

3.1.1. Effect of $V$ element content on TC21 alloy in $\beta$ solution state

Figure 1 is a low magnification backscattering picture and composition distribution curve of the diffusion couple uniformly distributed in the diffusion region. After that, the electron probe was used to determine the composition near the indentation dots. It can be seen that V atom diffuses into TC21 alloy and mainly replaces Ti atom, and a relatively smooth composition gradient curve is obtained. The diffusion distance of V in TC21 alloy is around 4500 $\mu m$.

Figure 2 is a backscattered image of TC21–TC21-20V diffusion couple in different positions after solution treatment and water-cooling. The composition of the alloy shown in the figure, TC21-xV(x ≈ 1, 2, 3….) implies that, besides other alloying elements of TC21, the alloy also contains x percent of V. As shown in figures 2(a) and (b), TC21 and TC21-1V alloys consist of coarse elongated martensite $\alpha'$ and $\beta$ matrix. As the V content increases to 2-3 wt%, the thickness of martensite $\alpha'$ lamella becomes thinner [figures 2(c) and (d)]. When the content of V increases to 4 wt%, it can be seen that martensite $\alpha'$ becomes fine needle-like [figure 2(e)]. Interestingly, when the content of V element increases to 5 wt% or more, only $\beta$ matrix can be observed under SEM. This indicates that the martensite transformation temperature has dropped below room temperature, and martensite transformation cannot take place during the rapid cooling process, and thus the $\beta$ matrix is completely retained [figure 2(f)]. Therefore, the critical V content for hindering $\alpha'$ martensite transformation in V modified TC21 alloy is around 5 wt%.

Figure 3 is the micromorphology of the diffusion couple solution treated at 1050 °C in $\beta$ single-phase region. The corresponding nanoindentation results are shown in the figure. As the diffusion distance increases, the content of V in the alloy gradually decreases the microhardness of the alloy increases first and then decreases.
When the content of V element is about 5 wt%, the microhardness of the alloy is approximately 5.25 GPa, which is about 0.75 GPa more than that of TC21 alloy.

3.1.2. Effect of V element content on TC21 alloy in solution plus aging state

Figure 4 shows the backscattered images of the TC21–TC21-20V diffusion couple at different positions after solution treatment at 1050 °C for 0.5 h and subsequent aging at 600 °C for 6 h. As shown in figures 4(a), (b) and (c), when the V content is 0-2 wt%, coarse lath-like α phase is distributed in the β matrix. When the V content is increased to 3-4 wt%, the thickness of the α phase in the alloy and its volume fraction decrease [figures 4(d) and (e)]. In figure 4(f), it is observed that the finest α phase is distributed in the β matrix, at which point the content of V is about 5 wt%. But when the V content is further increased to about 8 wt%, the thickness of α phase begins to increase again, although the volume fraction continues to decrease.

Figure 5 shows the variation of the composition and the microhardness after aging with the diffusion. As the diffusion distance increases, the content of V in the alloy decreases, and the hardness of the alloy increases first and then decreases. The TC21-5V alloy, which has the most diffuse α phase, has a hardness of 5.9 GPa, which is 1.4 GPa higher than that of TC21. At this condition, the thickness of α phase in the alloy is the thinnest and interlaced on the β matrix.

The microstructure analysis at different positions of TC21–TC21-20V diffusion couple shows that the hardness of the alloy has a close relationship with the size of α phase. The thickness of α phase in the TC21-5V alloy is only about 50 nm. This fine α phase produces a large number of α/β phase boundaries, which can
Figure 2. SEM, backscattered electron images of TC21xV alloys after solution treatment taken from TC21–TC21-20V diffusion multiple.

Figure 3. Composition–hardness relationship of the series TC21xV alloys after solution treatment.
Figure 4. SEM, backscattered electron images of TC21xV alloys after aging taken from TC21–TC21-20V diffusion multiple.

Figure 5. Composition-hardness relationship of the series TC21xV alloys after aging.
effectively hinder the movement of dislocations. Thus it was chosen TC21-5V as the candidate alloy composition for further study.

### 3.2. Microstructure and mechanical properties of forged TC21-5V alloy

The actual composition of TC21-5V alloy obtained by melting and casting is shown in table 2. The microstructure of the alloy after solution treatment in single-phase region of 900 °C plus aging treatment at 600 °C is shown in figure 6(a). The diffraction spots show that the alloy is composed of β matrix and α precipitation phase. Different from the α phase of TC21-5V alloy obtained by diffusion couple, there is no obvious variant selection of α phase (distributed at a certain angle), because a large number of defects (dislocation, sub grain boundary, etc) are introduced into the alloy during forging [19, 20]. Finer acicular α structure can be seen after aging at lower temperatures (500 °C, figure 6(b)) because lower temperatures produce greater degree of under cooling, which provides greater nucleation driving force for α phase precipitation. At the same time, the low temperature limits the diffusion of alloying elements, thus inhibiting the growth of α phase [21, 22]. The microstructure of the alloy consists of primary α phase, secondary α phase and β matrix after solution treatment in the two-phase region plus aging [figure 6(c)]. The most obvious difference between this microstructure and figure 6(a) is the primary α phase, which is produced in solution treatment stage, with larger size and thickness up to micrometer level (∼1 μm). The quantity and size of the phases depend on the alloy composition, solution-treatment temperature and forging conditions [23, 24]. In addition, it should be noted that when the aging temperature is the same, the size of secondary α phase of the alloy solution-treated in the two-phase region (820 °C) is smaller than that solution-treated in the single-phase region. The detailed reasons will be discussed in the next section.

The tensile properties of the alloy are shown in table 3. As expected, the alloy shows high tensile and yield strengths after solution treatment in the single-phase region plus aging, even better than Ti-5553 alloy [23].

| Element | Al | Mo | Zr | Nb | Sn | Cr | V | Ti |
|---------|----|----|----|----|----|----|---|----|
| Mass%   | 5.95 | 2.84 | 2.12 | 1.67 | 1.93 | 1.40 | 5.33 | Bal |

**Table 2.** Chemical composition of the TC21-5V alloy (wt%).
(~1314 MPa) under similar heat treatment conditions (i.e., all aged at 600 °C). This is attributable to the fine α phase dispersed in the alloy. However, the elongation of the alloy is only about 2%, and it breaks just after reaching the yield point in the process of drawing. The alloy shows obvious brittle fracture characteristics, with no uniform deformation or necking. The strength of the alloy after solution treatment in the two-phase region (820 °C) plus aging at 500 °C reached 1540 MPa. When the aging temperature is 600 °C, the alloy experiences appreciable improvement in ductility, with an elongation of 11.4%. Even at an aging temperature of 500 °C, the elongation is still higher than that of the alloy solution treated in the single-phase region.

Figure 7 is the tensile fracture morphology of each alloy under different heat treatment conditions. It can be seen from figures 7(a) and (b) that the alloy obtained after solution treatment above the transformation point plus aging has high brittleness, and the samples exhibited mainly brittle intergranular fracture. There are many cracks in the fracture surface of the alloy; smooth facets and dimple inclusions are also formed at the same time, as shown by the arrow in the figure. Because of the high local stress and strain in the triangle region, the alloy cracks in the triangle region of the grain boundary; and since the small-angle grains in the triangle region bear more load than the other two grains, the grains are separated at the small angle, thereby revealing the poor plasticity of the alloy. Figure 7(c) shows a typical ductile fracture, with a large number of shallow dimples in the shape of equal axis distributed in a network. The white ridge-like area around the dimples shows that the deformation took place uniformly in all the grains, and the alloy has good plasticity. The samples in figure 7(d) have a mixed fracture mode with ductile fracture as the main mode and brittle fracture as the auxiliary mode. There are two different types of dimple distribution on the fracture surface, one is the deeper dimple distributed in the plane, the other is the mesh-shallow dimple on the facets. In this case, the plasticity of the alloy is still low.
4. Discussion

4.1. Microstructure evolution of TC21-xV alloy

The composition of the alloy was designed by the diffusion couple method. In the solution treatment stage, with the increase of V content, the martensite lamellae first becomes thinner, and then a single $\beta$ phase is obtained [figure 2]. This is because V is a $\beta$ stabilizing element. The addition of V enhances the stability of $\beta$ phase and inhibits the occurrence of $\beta \rightarrow \alpha'$ transformation [25]. As the V content increases, the initial and end temperatures of the martensite transformation decrease during the quenching process, and the growth (roughening) time of the martensite layer is shortened, resulting in a decrease in the thickness of the sheet. As the V content increases above 5 wt%, $\beta \rightarrow \alpha'$ transformation is completely inhibited, and the initial temperature of martensitic transformation is reduced to below room temperature; the alloy obtained is a single $\beta$ structure.

In the aging stage, the size of $\alpha$ phase becomes thinner and then coarsens as the V content increases [figure 4]. The finest $\alpha$ phase is obtained at V content of 5 wt%. Comparing figures 2 and 4, it is obvious that when martensitic transformation occurs in the solution quenching stage, the morphology and lamellar thickness of the $\alpha$ phase in the aging stage depends on the morphology of martensite. When a single $\beta$ phase (V content is more than or equal to 5 wt%) is obtained in solution quenching stage, the thickness of $\alpha$ lamellae in aging stage mainly depends on the phase transformation driving force of $\alpha$ phase. However, the driving force of $\alpha$ phase transformation is related to the alloy composition [25]. When the content of V is low, the driving force of $\beta \rightarrow \alpha$ transformation is great, resulting in a higher nucleation rate, and a finer $\alpha$ structure can be produced. With the increase of V content, the driving force of $\beta \rightarrow \alpha$ transformation decreases, and the nucleation rate of $\alpha$ phase decreases. Thus, when the V content is 5 wt%, the solution quenched $\beta$ phase is of critical stability. During aging, $\alpha$ phase has the largest nucleation driving force thus producing the thinnest $\alpha$ phase [figure 4(f)]. When the V content is further increased to about 8 wt%, the nucleation driving force decreases and the thickness of $\alpha$ phase begins to increase again.

4.2. Effect of solution temperature on the growth of secondary $\alpha$ phase during aging

As shown in figure 6, with the increase of solution temperature (aging at 500 °C), the size of secondary $\alpha$ phase increases, and the number of $\alpha$ phase (quantity density) in unit area decreases. According to the classical ‘nucleation growth’ theory of new phase formation, the change of the morphology (size and quantity density) of secondary $\alpha$ phase is controlled by the ‘nucleation’ and ‘growth’.

In order to find out whether the process of nucleation or growth has more influence on the morphology of secondary $\alpha$ phase during aging, the composition of the residual $\beta$ phase after the precipitation of the primary $\alpha$ phase, the driving force and growth rate of secondary $\alpha$ phase transformation after different solution temperatures and at different aging temperature need to be calculated.

It is considered that the nucleation and growth of the secondary $\alpha$ phase occur in the $\beta$ matrix of the alloy, while the remaining $\beta$ matrix composition of the alloy changes after treatment at different solution temperatures. The composition of the residual $\beta$ phase of the alloy at different temperatures is shown in figure 8. The driving force of phase transformation in titanium alloy is the free energy difference of two phases. The variation of the driving force and growth rate of the secondary $\alpha$ phase with the solution temperature were calculated. As shown in figure 9, with the increase of solution temperature, the driving force and growth rate of secondary $\alpha$ phase transformation increase. The increase of driving force indicates that the alloy has a higher nucleation rate, which usually means that smaller size and larger density of $\alpha$ phase can be obtained. This result is contrary to the observed trend of $\alpha$ phase [figure 6]. It shows that the nucleation rate of $\alpha$ phase is not the main factor that determines the final morphology of secondary $\alpha$ phase. The calculation results of the growth rate of secondary $\alpha$ phase show that with the increase of solution temperature, the growth rate of $\alpha$ phase increases significantly, which may be the main factor affecting the morphology of secondary $\alpha$ phase. Meng et al [26] found that the growth of secondary $\alpha$ phase is controlled by the diffusion of alloy elements, which is realized by the aggregation of Al elements and the elimination of Mo, V and other elements. According to the residual $\beta$ matrix composition of the alloy calculated at different solution temperatures, with the increase of solution temperature, the content of Al in $\beta$ matrix increases, while the content of Mo, V and other stable elements decreases. The higher Al content and the lower Mo, V content can provide more Al elements for the growth of $\alpha$ phase. In addition, the primary $\alpha$ phase may also hinder the grown of secondary $\alpha$ phase. As shown in figures 6(c) and (d), when the alloy is solution treated below the transformation point, the primary $\alpha$ phase divides the $\beta$ matrix into different regions. The primary $\alpha$ phase is perpendicular to the expansion direction of secondary $\alpha$ phase, which hinders the growth of secondary $\alpha$ phase. Therefore, the ‘growth’ process of secondary $\alpha$ phase, which is influenced by the remaining $\beta$ matrix composition and the obstruction of primary $\alpha$ phase, may be the decisive factor affecting the morphology of secondary $\alpha$ phase at different solution temperatures.
4.3. Effect of aging temperature on the morphology of secondary α phase
With the increase of aging temperature, the secondary α phase in the alloy coarsened obviously and the amount decreased [figure 6]. The driving force and growth rate of secondary α phase transformation at different aging temperatures were calculated [figure 10]. The results show that with the increase of aging temperature, the driving force of α phase transformation decreases and the growth rate increases. It can be seen that the decrease of the nucleation rate of α phase and the increase of the growth rate jointly cause the secondary α phase to coarsen as the aging temperature increases.

4.4. Relationship between microstructure and properties of alloy
The alloy exhibits high strength and poor plasticity after solution treatment in the single-phase region plus aging treatment. This is because the uniform deformation of the alloy is mainly due to the slip and entanglement of
dislocations in α phase and β matrix, and the twin deformation may also occur in the secondary α phase in the region of high stress [27]. However, a large number of α/β interfaces in the alloy severely inhibit the dislocation slip, and the twin deformation inside the acicular α phase is also difficult to provide great convenience for the deformation of the alloy, because the size of twin is related to the size of α phase. Therefore, when the slip and twinning are completely suppressed, the local stress concentration in the material will make the weaker β grain boundary region become the point of crack initiation, and finally cause the intergranular brittle fracture of the material [28].

Compared with the solution treatment in β region, there are two reasons for the sharp increase of the plasticity of the alloy after solution treatment in α + β region. First of all, the increase of plasticity is due to the existence of primary α phase. The primary α phase is soft, and the β matrix filled with acicular secondary α phase is hard. Shekhar et al [12] studied the tensile deformation of Ti-5553 alloy, and found plane slip, deformation twin and tangle of dislocations in the primary α phase particles, which indicated that the primary α phase has enough plastic deformation ability, can coordinate the deformation between layered structures, and increase the ductility of the alloy. Due to the limitation of a large number of α/β interfaces, the sliding space of dislocations in β matrix covered by acicular secondary α phase is severely restrained. Secondly, the β grain size of the two-phase region solution treatment is smaller than that of the single-phase solution treatment. Although there is no evidence in the microstructure image [figure 6] of this experiment that the alloy has smaller β grain size after solution treatment in the two-phase region (due to the smaller field of view of TEM image). However, previous studies have shown that the primary α phase at the β boundary can effectively inhibit the growth of β phase and produce smaller β grains [29]. The small grain size can increase the ductility of the alloy, because it can make the cooperative deformation between grains easier and enhance the crack growth resistance. Therefore, the alloy treated in α + β phase region shows better plasticity than that treated in the single-phase region.

5. Conclusion

In this study, composition-microstructure-performance relationship of V modified TC21 alloy was studied by diffusion couple method, and the relationship between the microstructure and properties of target composition TC21-5V alloy was analyzed in detail. The conclusions are as follows:

(a) In the V modified TC21 alloy system, 5 wt% V is the critical composition which can hinder α’ martensite transformation and generate metastable β with the highest instability during quenching, which can produce ultra-fine α phase and the highest hardness during aging.

(b) When the TC21-5V alloy was solution treated at 820 °C for 0.5 h and aged at 500 °C for 6 h, the strength reached 1540 MPa, and the elongation, 4.7%; but when the alloy was solution treated at 820 °C for 0.5 h and aged at 600 °C for 6 h, the strength reached 1210 MPa, and the elongation, 11.4%.

(c) After solution and aging treatment in the two-phase region, the growth rate of secondary α phase was found to play a decisive role in its final size.

Figure 10. (a) Variation of the precipitation driving of the secondary α phase force with the aging temperatures; (b) Variation of the growth rate of the secondary α phase with the aging temperatures.
Acknowledgments

The authors acknowledge financial support from the National Key Technologies R & D Program of China (Grant No.2016YFB0701301), National Natural Science Foundation of China(Young Scientists Fund, Grant No. 51901251), National Natural Science Foundation of China (Grant No.51671218 and 51501229), State Key Laboratory of Powder Metallurgy Independent Project of China (Grant No.621021907).

ORCID iDs

Li-gang Zhang @ https://orcid.org/0000-0002-6198-0781
Jing-ya Zhou @ https://orcid.org/0000-0001-9812-1297
Zhen-yu Wang @ https://orcid.org/0000-0001-8984-0345
Jing-lu Tang @ https://orcid.org/0000-0002-9953-0516
Di Wu @ https://orcid.org/0000-0002-3673-196X
Li-bin Liu @ https://orcid.org/0000-0003-0399-7582

References

[1] Wu C and Zhan M 2019 J. Alloys. Compounds. 805 1144
[2] Sadeghpour S, Abbasi S M, Morakabati M and Bruschi S 2017 Mater. Des. 121 24
[3] Wen X, Wan M, Huang C, Tan Y, Lei M, Liang Y and Cai X 2019 Mater. Des. 180 107898
[4] Shi Z, Guo H, Zhang J and Yin J 2018 Trans. Nonferr. Metal. Soc. 28 2440
[5] Tan C, Sun Q, Xiao L, Zhao Y and Sun J 2018 Mater. Sci. Eng. A 725 33
[6] Sun H, Yu L, Liu Y, Zhang L, Liu C, Li H and Wu J 2019 Trans. Nonferr. Metal. Soc. 29 59
[7] Wang Y, Lu Y, Peng B, Kloenne Z T, Fraser H L, Chiu Y L and Loretto M H 2019 Acta Mater. 173 242
[8] Meng M, Yang H, Fan X G, Yan S L, Zhao A M and Zhu S 2017 J. Alloys. Compounds. 691 67
[9] Weng W, Biesiekierski A, Li Y and Wen C 2019 Materials 6 100323
[10] Sabih M, Teixeira J, Germain L, Lamielle E, Gey N and Aeby-Gautier E 2013 Acta Mater. 61 3758
[11] Wu D, Zhang L, Liu L, Bai W and Zeng L 2018 Trans. Nonferr. Metal. Soc. 28 1714
[12] Shekhar S, Sarkar R, Kar S K and Bhattacharjee A 2009 Mater. Des. 66 596
[13] Kar S K, Ghosh A, Fulzele N and Bhattacharjee A 2013 Mater. Charact. 81 37
[14] Appolaire B, Hericher L and Aeby-Gautier E 2005 Acta Mater. 53 3001
[15] Zhu S, Yang H, Guo L G and Fan X G 2012 Mater. Charact. 70 101
[16] Cheng K, Zhou H, Hu B, Du Y, Zhang L, Liu S, Xu H and Liu L 2014 Metall. Mater. Trans. A 45 2708
[17] Malinov S, Markovsky P, Sha W and Guo Z 2001 J. Alloys. Compounds. 314 122
[18] Du Z, Xiao S, Xu L, Tian J, Kong F and Chen Y 2014 Mater. Des. 55 183
[19] Rhodes C G and Williams J C 1975 Metall. Trans. A
[20] Sun T, Liu Y, Li S and Li J 2019 Acta Metall. Sin. (Engl. Lett.) 32 869
[21] Ahmed M, Gazder A A, Savvakin D G, Ivasishin O M and Pereloma E V 2012 J. Mater. Sci. 47 7013
[22] Li C, Lee D, Mi X, Ye W, Hui S and Lee Y 2013 J. Alloys. Compounds. 549 152
[23] Wang K, Bao R, Jiang B, Wu Y, Liu D and Yan C 2018 Int. J. Fatigue 116 535
[24] Wei Z, Peng G, Yongqing Z, Shewei X, Qian L, Jun C, Siyuan Z and Chaowen H 2017 Rare Met. Mat. Eng. 46 2852
[25] Liang S X, Peng Z H, Yin L X, Liu X Y and Liu R P 2015 J. Alloys. Compounds. 664 11
[26] Meng M, Fan X G, Yang H, Guo L G, Zhan M and Gao P F 2017 J. Alloys. Compounds. 714 294
[27] Shi Y, Zhang G, Ming L, Guo D and Zhang X 2016 J. Alloys. Compounds. 665 1
[28] Sadeghpour S, Abbasi S M, Morakabati M and Karjalainen L P 2019 J. Alloys. Compounds. 808 151741
[29] Wang X, Lv F, Shen L, Liang H, Xie D and Tian Z 2019 Acta Metall. Sin. (Engl. Lett.) 32 1173