Effects of High Niobium Addition on the Microstructure and High-Temperature Properties of Ti-40Al-xNb Alloy

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The effect of Nb content on the microstructure and high-temperature properties of the TiAl alloy with low Al content has been investigated. The as-cast Ti-40Al-10Nb alloy consists of a Widmanstätten lath and a γ phase in a B2 matrix. The increased addition of Nb to the alloy inhibits the solidification path β → β + α. Therefore, the as-cast Ti-40Al-xNb (x = 15, 16) alloys are composed of primary β/B2 phases as the major constituent. Following heat treatment, the microstructure of the heat-treated Ti-40Al-10Nb alloy resembles that of the as-cast Ti-40Al-10Nb alloy. The homogenized Ti-40Al-15Nb alloy has a two-phase microstructure of B2 + γ, while the homogenized Ti-40Al-16Nb alloy has a four-phase microstructure of B2 + γ + α + σ. The creep response of the studied TiAl-Nb alloy is correlated closely with the tertiary creep, and is affected strongly by its microstructure. The creep resistance of the alloy depends on the resistance to the propagation of the cracks in the B2 matrix. In the Ti-40Al-15Nb alloy, the resistance to the extending of cracks is the most apparent among these three alloys. Therefore, the Ti-40Al-15Nb alloy has the highest creep life of the three tested alloys. The oxidation resistance of the investigated TiAl-Nb alloy is independent of the addition of Nb. The differences among the oxidation resistances of these three alloys arise from their various microstructures. The oxide scales of the Ti-40Al-10Nb alloy are composed mainly of TiO2, while the oxide scales formed on the Ti-40Al-15Nb alloy are dense Al2O3-rich oxides. The Ti-40Al-15Nb alloy has the lowest mass gain among these three alloys due to the existence of the Al2O3-rich oxides. The formation of Al2O3-rich oxides appears to be related to the high Nb content of the alloy. The fact that the Ti-40Al-16Nb alloy has the highest weight gain of these three alloys is attributed to the formation of oxide scales of σ phase. [doi:10.2320/matertrans.47.2588]

(Received July 20, 2006; Accepted August 21, 2006; Published October 15, 2006)

**Keywords:** titanium-aluminum-niobium intermetallic alloys, high Nb additions, microstructure, creep, isothermal oxidation

1. Introduction

γ-TiAl alloys have low density and high strength retention at elevated temperature. Hence, they are promising materials for high temperature applications in both aerospace and automobile industries. However, the practical applications of this alloy are limited by its inherent low ductility at ambient temperature and poor oxidation resistance. Adding a third element can improve the engineering properties of TiAl alloys for actual applications. Most work has been done on γ-TiAl-based alloys with 45 to 49 mol% Al, microalloyed with various third elements such as Cr, Mn, Ta, W, Si and B. Recently, TiAl alloys with large amounts of Nb have been developed. High Nb-containing TiAl alloys have good strength at elevated temperature and exhibit some ductility at room temperature with a compressive ratio of 12–14%. The oxidation resistance of the Nb-bearing TiAl alloys exceeds that of Nb-free TiAl alloys. Moreover, TiAl alloys with high Nb contents exhibit improved capabilities for hot-working. For industrial applications, the development of TiAl alloys that can be manufactured by conventional processes is necessary. Therefore, the potential of adding such large amounts of Nb to TiAl alloys is of interest.

As mentioned previously, adding Nb to TiAl alloys beneficially affects both mechanical properties and oxidation resistance. However, investigations of TiAl alloys with an Al content of less than 45 mol% and more than 10 mol% Nb are few. In this study, the effects of high Nb addition on the microstructure and high-temperature properties of the Ti-40Al-xNb (x = 10, 15, 16) alloy are investigated. For high temperature usage, properties such as creep and oxidation resistance are important. Accordingly, this study examines the effect of Nb content on the microstructures, creep and oxidation resistance of TiAl-Nb alloy. In this work, the creep resistance of the Ti-40Al-xNb (x = 15, 16) alloys is compared with that of hot-rolled Ti-40Al-10Nb alloy.}

2. Experimental

Alloys (mol%) with a nominal composition of Ti-40Al-xNb (x = 10, 15, 16) were prepared in an arc-melting under non-consumable tungsten electrode under protective argon gas. The Ti-40Al-10Nb alloy ingot was processed by hot rolling at 1473 K with a reduction of 40% and followed by air cooling. After the process of hot rolling, the rolled sheets were heated at 1623 K for 30 min and then air-cooled. The sheet was further annealed at 1273 K for 2 h and air cooled. The Ti-40Al-xNb (x = 15, 16) alloy ingots were homogenized at 1373 K for 72 h in Ar gas. Microstructures were observed in a SEM using BSE imaging. The creep test specimens machined from the heated-treated samples were in the form of plates with gauge dimensions of 12 × 3 × 2 mm3. Tensile creep tests were performed in air for the temperature of 1073 K under the constant load between 200 and 280 MPa. The microstructures of the creep specimens were examined by a SEM. For isothermal oxidation testing, the specimens had a size of 10 × 10 × 1 mm3. Isothermal oxidation tests were carried out at 1073 K in a box furnace in static air for 600 h. The specimens tested were weighted at intervals of 12–72 h to measure the change in its mass. Phase identification of oxide scales formed on the specimens was conducted by an XRD. A FEG-SEM was used to examine the surface morphology of the oxidized specimens.

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3. Results and Discussion

3.1 Microstructure of the Ti-40Al-xNb (x = 10, 15, 16) alloys

Figure 1(a) shows that the as-cast Ti-40Al-10Nb alloy is composed of gray Widmanstätten laths (α/γ) and dark γ phases in a white B2 matrix. Our early work described the sequence of phase transformations for the as-cast Ti-40Al-10Nb alloy during solidification and indicated that the Widmanstätten laths have a fine lamellar structure composed of α and γ phases, both of which are perpendicular to the long axis of the laths. According to the Ti-Al phase diagram, the binary Ti-40Al alloy has two phases, which are α and γ. Adding 10 mol% Nb to the Ti-40Al alloy forms an additional B2 phase, because Nb, which is a strong β/B2 stabilizer, can enlarge the β/B2 region and reduce the α field in the Ti-Al phase diagram. Hence, in the Ti-40Al-10Nb alloy, the high-temperature β/B2 phase can be retained at the room temperature during solidification, yielding the microstructure in Fig. 1(a). However, a unique feature of the as-cast Ti-40Al-xNb (x = 15, 16) alloys is the presence of dendritic microstructures at low magnification, as shown in Fig. 1(b). The EDS analysis reveals that the primary crystal of enriched Nb is the β/B2 phase. The inter-dendrite regions, indicated by a slightly dark contrast, contain dense γ phases. The dense γ phases are formed by retained Al-rich molten metal after the primary β/B2 phase solidification. Compared with the microstructure of the as-cast Ti-40Al-10Nb alloy, the as-cast Ti-40Al-xNb (x = 15, 16) alloys do not exhibit the structure of Widmanstätten laths, because of increased addition of Nb to the alloys. The primary β/B2 of enriched with Nb in the as-cast Ti-40Al-xNb (x = 15, 16) alloys can not transform completely to the γ phase due to the low diffusion rate of Nb under fast cooling rate during solidification. The solidification path \( \beta \rightarrow \beta + \alpha \) is inhibited by the relatively large amount of Nb added. Therefore, the microstructure of the as-cast Ti-40Al-xNb (x = 15, 16) alloys is composed of the primary β/B2 phase as the major constituent.

After heat treatment, the microstructure of the Ti-40Al-xNb alloy is very similar to that of the as-cast Ti-40Al-10Nb alloy, as shown in Fig. 2(a). The figure reveals that the γ phases originally in the B2 matrix become thick, and some fine γ phases precipitate at the interfaces between the Widmanstätten laths and the B2 matrix, suggesting that subsequent heat treatment induces a phase transformation to the stable γ phase. Figure 2(b) indicates that the microstructure of the Ti-40Al-15Nb alloy comprises a dark γ phase in a B2 matrix after homogenization. The homogenized Ti-40Al-15Nb alloy has a two-phase microstructure of B2 + γ. The microstructure of the homogenized Ti-40Al-16Nb alloy resembles that of the homogenized Ti-40Al-15Nb alloy, except for the appearance of α phases and a few γ phases, as shown in Fig. 2(c). The few α phases are surrounded by σ and γ phases, but the phase contrast of α makes identification therefore difficult. Table 1 summarizes the results of chemical compositions for the phases presented in the heat-treated Ti-40Al-xNb (x = 10, 15, 16) alloys. The EDS analysis reveals that the Nb content in each phase of the studied alloys is increased as Nb is added, indicating that Nb is highly soluble in the alloys. The partitioning of Nb between γ and α phases is almost equal; while the ratio of Nb to Ti to Al in the σ phase is around 1:1:1.

In the Fig. 2(b), the coarse γ phase, present as the equiaxed and rod-like grains, results from the growth and coarsening of γ phases themselves during long-term heat treatment; the fine granular γ phase is formed by precipitation from the B2 matrix. In the Ti-40Al-16Nb alloy, the formation of the σ phase is consistent with the report that the σ phase is only observed in the sample that contains more than 15 mol% Nb. However, the existence of the α phase, accompanying the σ phase, may be correlated with the formation of the σ phase. During homogenization, the σ phase probably precipitates at the interface between the γ phase and the B2 matrix via heterogeneous nucleation. As the σ phase grows, it has to absorb Nb atoms from the neighboring B2 matrix with high Nb content, and discharge Ti atoms to yield the equilibrium composition of the σ phase; the fluctuation of Al concentration is reduced because the amount of Al is quite close to the equilibrium Al content in the σ phase. Based on the above, the region that surrounds the σ phase is certainly relatively Ti-rich and Nb-lean, corresponding exactly to the chemical composition of the α phase, as indicated in Table 1. Therefore, the homogenized Ti-40Al-16Nb alloy contains additional α phases. The microstructural analysis of Ti-40Al-xNb (x = 10, 15, 16) alloys demonstrates that adding a large amount of Nb to the alloys changes the constituent phases of

![Fig. 1](image-url)
the investigated TiAl alloys with low Al concentration. The various microstructures are expected to affect the high-temperature properties to a certain extent.

3.2 Creep properties

Figure 3(a) plots typical creep curves of three tested Ti-40Al-xNb (x = 10, 15, 16) alloys at 1023 K/200 MPa. The figure clearly shows that the Ti-40Al-15Nb alloy has the highest creep resistance. The result of creep test indicates that the microstructure of the alloy strongly affects its creep resistance. Moreover, the accumulated strain of the Ti-40Al-xNb (x = 10, 15, 16) alloys is high. The high creep fracture strain of the tested alloys is though to be associated with the B2 phase. The B2 phase has an open bcc structure, and is therefore softer and more easily deformed than the other phases in the alloy at elevated temperature. Therefore, the high creep fracture strain is correlated with the B2 phase, in which the deformation during creep is concentrated. Figure 3(b) plots a representative creep curve of creep rate against creep strain, for the Ti-40Al-15Nb alloy deformed
at 1073 K and under 200 MPa. The minimum creep rate of the two-phase alloy of B2 + γ is lower than that of TiAl alloys with a nearly-fully lamellar structure, but still higher than that of those with a fully lamellar structure,\textsuperscript{13} indicating satisfactory creep resistance for this B2 + γ microstructure. Additionally, all of the creep curves obtained in this study have the same basic shape, as displayed in Fig. 3(b). The curve exhibits a short primary creep stage, a very limited secondary creep region and a prominent tertiary creep range. The creep curves of the investigated alloys reveal a minimum creep region, instead of a steady-state creep region. Therefore, the creep response of the TiAl-Nb alloy is closely related to the tertiary creep.

With respect to the absence of an extended steady-state region, the reason for this phenomenon remains unclear. The literature offers two possible explanations of the absence of an extended steady-state region. The first is the lack of a distinct subgrain substructure in the ordered alloys.\textsuperscript{14} Our earlier work confirmed this suggestion and showed that the dislocations are uniformly distributed in the alloy during secondary creep, and do not tend to form subgrain substructures.\textsuperscript{15} Furthermore, the dislocation density of the B2 phase is higher than that of the γ phase, indicating that the creep deformation of the investigated TiAl-Nb alloy converges primarily in the B2 phase. The second explanation is the effect of oxidation.\textsuperscript{16} Owing to extensive cracking formed in the oxide layers, further oxidation and stress during creep aggravate the cracking, causing the cracks to propagate into the metal matrix, resulting in the early start of the tertiary stage, and thus the absence of the steady-state deformation. In the next section, the result of the isothermal oxidation test at 1073 K indicates that the highest weight gains among these three alloys are less than 1.4 mg/cm\textsuperscript{2} after a testing time of 600 h, whereas the lowest are 0.7 mg/cm\textsuperscript{2}. After the creep fracture time of the tested alloys, the highest weight gains are only less than 0.7 mg/cm\textsuperscript{2}. The results reveal that the formed oxide layer is very thin and the studied high-Nb containing TiAl alloys exhibit good oxidation resistance at 1073 K. Moreover, no oxide scales flake off of these three alloys after long term exposure for 600 h, indicating strong adherence between the oxide layer and the alloys. Accordingly, in this study, no effect of oxidation on the creep curves of the investigated alloys is assumed.

As stated in the preceding paragraph, the creep response of the alloys tested is closely related to tertiary creep and the creep resistance is strongly affected by its microstructure. Microstructural observation after creep fracture can reveal the strengthening effect of the microstructure. Figure 4 displays representative images of these three TiAl-Nb alloys after creep. Figure 4(a) shows that the fracture mode of the Ti-40Al-10Nb alloy is transgranular fracture and cracking proceeds by directly traversing the B2 matrix and the Widmanstätten laths. Furthermore, some Widmanstätten laths are slightly curved in tensile direction. In the Ti-40Al-15Nb alloy, the propagation of cracks through the B2 matrix by cleavage is apparent, as shown in Fig. 4(b). However, in the Ti-40Al-15Nb alloy, the situations in which cracks are deflected and shielded by γ phases greatly differ from those in the Ti-40Al-10Nb alloy. In the Ti-40Al-15Nb alloy, the γ phase is an important strengthening phase in the B2 matrix. By contrast, in the T-40Al-10Nb alloy, the retardation of the propagation of the cracks by the dense Widmanstätten laths is not evident. The result indicates that the morphology of the constituent phases in the investigated alloys markedly influences the strengthening. The observed microstructures of the tested alloys show that the incompatibility of the plastic behavior between the major phase and the B2 matrix of alloys causes cracking to originate at an interface, such as...
the \( \gamma \)/matrix or the Widmanstätten laths/matrix interface, during creep deformation. During tertiary creep, the creep strain gradually increases, and cracks propagate either along the interface or through the cleavage in the B2 matrix. As cracks progress into the B2 matrix, they pass the Widmanstätten laths in the Ti-40Al-10Nb alloy, whereas their propagation is hindered by the \( \gamma \) phases in the Ti-40Al-15Nb alloy, as shown in Figs. 4(a) and 4(b).

The cracks proceed in the Ti-40Al-16Nb alloy as do those in the Ti-40Al-15Nb alloy, but the existence of the \( \sigma \) phase does not impede their progress, as shown in Fig. 4(c). The brittleness of the \( \sigma \) phase causes most of the cracks to run through it immediately, indicating no resistance to their propagation. Therefore, the creep life of the Ti-40Al-16Nb alloy is shorter than that of the Ti-40Al-15Nb alloy, but still longer than that of the Ti-40Al-10Nb alloy. The interaction between the \( \sigma \) phase and the crack in the Ti-40Al-16Nb alloy is similar to that of the \( \gamma \) phase and the crack in the Ti-40Al-15Nb alloy, as presented in Fig. 4(c). However, the beneficial effect of the \( \sigma \) phase is negligible because it is present in a very small amount in the alloy. In conclusion, the creep resistance of the investigated TiAl-Nb alloy depends on the resistance to the propagation of cracks. The creep life of the alloy increases with resistance to the extending of the cracks.

For the reason discussed above, the Ti-40Al-15Nb alloy with the B2 + \( \gamma \) microstructure has the highest creep life of the three tested alloys.

### 3.3 Isothermal oxidation

Figure 5 plots the weight gain curves of the Ti-40Al-\( x \)Nb (\( x = 10, 15, 16 \)) alloys as a function of oxidation time at 1073 K. The figure shows that the Ti-40Al-15Nb alloy with the B2 + \( \gamma \) microstructure has the lowest weight gain of these three alloys after long-term exposure for 600 h, whereas the Ti-40Al-16Nb alloy has the highest weight gain. The oxidation resistance of the investigated TiAl-Nb alloys is superior to that of the \( \gamma \)-TiAl based alloys under the same test conditions at 1073 K,\(^{17}\) indicating that the high Nb-containing TiAl alloys has a satisfying resistance against oxidation. Moreover, the weight gain data reveal that the oxidation resistance of the TiAl-Nb alloys is independent of the addition of Nb, suggesting that the differences among the oxidation resistances of these three alloys arise from their various microstructures.

Figure 6 presents the morphology of oxide scales formed on the surface of the Ti-40Al-\( x \)Nb (\( x = 10, 15, 16 \)) alloys after isothermal oxidation. The XRD analysis confirms that the oxide scales are composed of TiO\(_2\) and \( \alpha \)-Al\(_2\)O\(_3\) for all specimens. However, in this study, none of the investigated alloys forms nitrides and oxides of niobium, perhaps because the peaks of those oxides are masked by the overlapping of strong peaks derived from rutile and alumina. Figure 6(a) shows the morphological features of the surface oxide scales formed on the Ti-40Al-10Nb alloy. The pillar-like grains are the typical microstructure of TiO\(_2\), and are the major oxide scales of the Ti-40Al-10Nb alloy. In contrast, the oxide scales on the surface of Ti-40Al-15Nb alloy exhibit various morphologies that depend on the constituent phases of the alloy. After a long oxidation test, the contour of the \( \gamma \) phase can still be identified and the morphologies of the oxide scales of the B2 matrix and \( \gamma \) phase differ, as shown in Figs. 6(b) and 6(c), respectively. The oxide scales of the B2 matrix are lenticular and joined together, while those of the \( \gamma \) phase are present as fine particles and are not close to each other. The EDS analysis shows that both the oxide scales are Al-rich with an Al content of about 30 mol%. The Nb content in the oxide scales of the B2 matrix and the \( \gamma \) phase is in the range 1–2 mol%. The Ti content in the oxide scales of the B2 matrix is about 4 mol%–lower than that in the oxide scales of \( \gamma \) phase, which is about 7 mol%. Perhaps the morphologies of the oxide scales differ because the TiO\(_2\) quantities in the oxides scales of the B2 matrix and the \( \gamma \) phase differ.

Microstructural observations show that the morphological features of oxide scales of Ti-40Al-15Nb alloy differ greatly from those of the Ti-40Al-10Nb alloy, which fact is believed to be related to the difference between the microstructures of these two alloys. In Ti-40Al-10Nb alloy, the presence of Widmanstätten laths as the major precipitates causes the formation of TiO\(_2\) during oxidation. The literature\(^{18}\) demonstrates that the solubility of oxygen in the \( \sigma \) phase is around 20 mol%, whereas it is only 500–1000 ppm in the \( \gamma \) phase, suggesting that oxygen tends to be absorbed preferentially by the \( \sigma \) phase. Because TiO\(_2\) grows mainly by oxygen diffusion, the circumstance that oxygen tends to be absorbed preferentially by the \( \sigma \) phase promotes the formation of TiO\(_2\) during oxidation. Therefore, the oxide scales of the Ti-40Al-10Nb alloy are mostly composed of TiO\(_2\), which is a non-protective oxide product, resulting in a greater weight gains than that of the Ti-40Al-15Nb alloy. The oxide scales formed on the Ti-40Al-15Nb alloy are dense Al\(_2\)O\(_3\)-rich oxides, which differ greatly from the non-compact TiO\(_2\) that is always the main oxide product of TiAl alloy during oxidation process. The reason for the dense Al\(_2\)O\(_3\)-rich oxides formed on the Ti-40Al-15Nb alloy is believed to be associated with the high content of Nb in the alloy. The increased amount of Nb increases oxidation resistance. The improvement of the oxidation resistance of TiAl alloy by adding Nb has been reported and various mechanisms have been proposed to interpret the beneficial effect of Nb on oxidation.\(^{5,19–24}\) Two possible mechanisms, which are the suppression of rutile growth and the enhanced formation of alumina, can be used.

![Fig. 5 Isothermal oxidation curves for Ti-40Al-xNb (x = 10, 15, 16) alloys in air at 1073 K.](image-url)
to account for the improvement of oxidation resistance by the addition of Nb in this study. The former mechanism is the so-called doping effect, by which oxygen vacancies in TiO$_2$ are reduced by the Nb cations which are incorporated to TiO$_2$ and substituted for Ti cations. The effect results in the reducing of oxygen diffusion and then the suppressing of the growth of rutile. The latter mechanism is the improvement of Al activity by the addition of Nb in the TiAl alloy, inducing the formation of Al$_2$O$_3$-rich oxides in the alloy during oxidation. The results of EDS analysis and microstructural observation of the oxides scales formed on the Ti-40Al-15Nb alloys reveal the effects of these two mechanisms, increasing the oxidation resistance above that of the Ti-40Al-10Nb alloy. However, the results of the analyses do not reveal which mechanism dominates the increase in the oxidation resistance of the alloy. Overall, the increased Nb content in the alloy apparently improves the oxidation resistance, as displayed in Fig. 5.

Figure 6(d) shows the morphology of the oxide scales formed on the Ti-40Al-16Nb alloy. The figure reveals that new oxide scales are formed on the Ti-40Al-16Nb alloy, except for the presence of the oxide scales of the B2 matrix and the $\gamma$ phase. The microstructure of the Ti-40Al-16Nb alloy before oxidation shows that the oxides scales are the oxide products of the Nb-enriched $\sigma$ phase. The EDS analysis indicates that the amount of Nb in the oxides of the $\sigma$ phase is approximately 8 mol%, while the Al and Ti contents are 20 and 9 mol%, respectively. The results reveal that the oxide scales of the Nb-rich $\sigma$ phase are still Al$_2$O$_3$-rich oxides, suggesting that niobium can indeed promote the formation of alumina during oxidation. The high Nb content in the oxide is such that Nb$_2$O$_5$ is expected to be formed, even though it is not detected by XRD. The oxide scales are present as a dense structure perhaps because of the mix of Nb$_2$O$_5$ and other oxides, such as TiO$_2$ and Al$_2$O$_3$. Additionally, as seen from Fig. 6(d), the height of the oxide scales of the $\sigma$ phase exceeds that of the other oxide scales, indicating that the growth rate of the oxide scales of the $\sigma$ phase is fast during oxidation, resulting in the largest weight gains of the three tested alloys. Therefore, the fact that the Ti-40Al-16Nb alloy has the largest weight gains among these three alloys is attributed to the formation of oxide scales of the $\sigma$ phase. Microstructural observations of the investigated TiAl-Nb alloys after oxidation demonstrate that the oxidation resistance of the alloys depends on the constituent microstructures, and the formation of Al$_2$O$_3$-rich oxides appears to be related to the high Nb content in the alloys.

4. Conclusions

(1) The as-cast Ti-40Al-10Nb alloy consists of gray Widmanstätten laths and dark $\gamma$ phases in a white B2 matrix; while the as-cast Ti-40Al-15Nb ($x = 15, 16$) alloys comprise the primary crystal of $\beta$/B2 as the main constituent, and the $\gamma$ phases. After heat treatment, the microstructure of the Ti-40Al-10Nb alloy is quite
similar to that of the as-cast Ti-40Al-10Nb alloy. The differences between microstructures of the as-cast alloy and the heat-treated alloy are only the coarsening of γ phases. The homogenized Ti-40Al-10Nb alloy has a two-phase microstructure of B2 + γ. The microstructure of the homogenized Ti-40Al-15Nb alloy resembles that of the homogenized Ti-40Al-15Nb alloy, except for the presence of phases and a few α phases. Adding a large amount of Nb to the investigated TiAl alloy with low Al content changes the constituent phase of the alloys.

(2) The Ti-40Al-15Nb alloy has the highest creep resistance among these three alloys. The creep curve of the studied TiAl-Nb alloy does not reveal a steady-state creep region, but shows a prominent tertiary creep range. The creep response of the alloy is closely related to the tertiary creep. The microstructure of the alloy strongly affects its creep resistance. The microstructure of the alloy strongly affects its creep resistance. The microstructure of the alloy strongly affects its creep resistance. The microstructure of the alloy strongly affects its creep resistance. The microstructure of the alloy strongly affects its creep resistance. The microstructure of the alloy strongly affects its creep resistance. The microstructure of the alloy strongly affects its creep resistance. The microstructure of the alloy strongly affects its creep resistance. The microstructure of the alloy strongly affects its creep resistance. The microstructure of the alloy strongly affects its creep resistance.

(3) After isothermal oxidation test at 1073 K for 600h, the Ti-40Al-15Nb alloy has the lowest weight gain of these three alloys, whereas the Ti-40Al-16Nb alloy has the highest weight gain. The oxidation resistance of the studied TiAl-Nb alloy depends on the constituent microstructures. The oxide scale of the Ti-40Al-10Nb alloy mainly comprises TiO₂, while the oxide scale of the Ti-40Al-xNb (x = 15, 16) alloys consists of Al₂O₃-rich oxide. The formation of Al₂O₃-rich oxide seems to be associated with the high Nb content in the alloys. The Ti-40Al-16Nb alloy has the largest weight gains among these three alloys due to the formation of oxide scales of the α phase.

Acknowledgements

The author would like to thank National Science Council of the Republic of China for financially supporting the research under Contract No. NSC92-2216-E-002-022.

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