The Effect of Heat Treatment on the Microstructure and Mechanical Properties of Powder Metallurgy Ti-48Al Alloy

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Abstract: Heat treatment is the critical step in achieving a refined microstructure and enhanced mechanical properties of TiAl-based alloys. This study investigated the influence of heat treatment temperature, cooling method, and heat treatment time on the microstructure and mechanical properties of an extruded powder metallurgy Ti-48Al alloy, and achieved the control of fully lamellar fine microstructures and the enhancement of performance through a simple heat treatment, rather than the traditional approach of homogenization followed by heat treatment. The results indicate that the heat treatment temperature determines the type of microstructure, while the cooling rate dictates the lamellar width. As the heat treatment temperature was increased from the two-phase region to the α single-phase region, the microstructure transitioned from duplex to near lamellar, and the alloy strength initially increased and then decreased, influenced by both the lamellar colony ratio and grain size. A rapid cooling rate (water quenching) induces a non-diffusive massive phase transformation, whereas a slow cooling rate (air cooling) gradually forms α2/γ lamellar colonies. Therefore, a suitable heat treatment regime for the powder metallurgy Ti-48Al alloy was determined to be 1340 °C/5 min/air cooling. The microstructure of the alloy was near lamellar, consisting of lamellar colonies approximately 50 µm and a small number of γ equiaxed grains of about 10 µm. Subsequently, the alloy exhibited a room temperature tensile strength of 784 MPa and a yield strength of 763 MPa, representing improvements of 17.0% and 38.7% over the extruded alloy, respectively. This research provides a reference for establishing a heat treatment process for powder metallurgy TiAl alloys.

Keywords: TiAl-based alloy; heat treatment; microstructure; mechanical properties

1. Introduction

TiAl-based alloys have a wide range of applications in aerospace, gas turbines, and other fields due to their low density, high specific strength, and excellent high-temperature properties [1,2]. At present, TiAl-based alloys have several industrial applications. Notably, more than 200 Boeing 787 series aircraft have been equipped with TiAl blades, and the engines of the C919 passenger jet also feature TiAl blades [3]. With the increasing thrust-to-weight ratio of the new generation of engines, there is a demand for high material properties, requiring TiAl-based alloys with outstanding comprehensive properties.

The principal method of enhancing the performance of TiAl alloys is through the rational design of composition and fabrication processes. For alloys with a fixed composition, it is essential to regulate their microstructure via suitable thermomechanical and
heat treatment processes, thereby endowing the alloy with high comprehensive properties [4]. The thermomechanical processing of γ-TiAl-based alloys is typically carried out in the (α + γ) two-phase region, which includes forging, extruding, and rolling. The microstructure of the alloy after thermomechanical processing transforms from a coarse fully lamellar microstructure in the as-cast state to a duplex microstructure composed of γ equiaxed grains and α2/γ lamellar colonies, along with some residual lamellae [5, 6]. At this stage, heat treatment is required to enhance the uniformity of the microstructure and eliminate residual lamellae, while also transitioning the microstructure to a finer fully lamellar microstructure, which is associated with superior mechanical properties. Some studies have indicated that TiAl-based alloys with an average volume percentage of α2/γ between 0.05 and 0.25, containing none or a small amount of γ grains, and lamellar colony sizes of approximately 50 µm, exhibit a better comprehensive performance [7].

The heat treatment of TiAl alloys produced by casting and thermo-mechanical processes can be classified into cyclic heat treatment and simple heat treatment. Cyclic heat treatment involves repeatedly heating the alloy to the α single-phase region to utilize multiple phase transformation cycles to refine the microstructure, which is effective but complex. Kim et al. reduced the grain size from 500 µm to 65 µm through homogenization heat treatment and multi-step cyclic heat treatment [8]. Yim et al. refined the microstructure of a Ti-48Al-2Cr-2Nb alloy through a multi-stage heat treatment process consisting of solution treatment, cyclic heat treatment, annealing, short-term heat treatment, and aging, which increased the tensile strength and elongation to 697 MPa and 2.1%, respectively [9]. Simple heat treatment aims to achieve the desired microstructure by adjusting process parameters, offering the advantages of a shorter process and lower costs. For instance, Wallgram et al. significantly refined the grain size and improved the comprehensive properties of the alloy by combining forging with a simple recrystallization heat treatment [10]. S. Bolz et al. enhanced the mechanical properties of a forged TNB-V5 alloy by heat treating it at 1270 °C followed by air cooling and annealing at 800 °C [11]. However, in TiAl-based alloys prepared by powder metallurgy, the initial microstructure differs significantly from that of cast alloys, and it is necessary to further explore the impact of heat treatment on the microstructure and mechanical properties of the alloy.

In our previous research, we prepared a Ti-48Al alloy with a fine duplex microstructure using powder metallurgy combined with encapsulated extrusion. The alloy had a tensile strength of 670 MPa and a yield strength of 550 MPa at room temperature [12]. Based on this, the current study systematically investigated the effects of heat treatment temperature, cooling method, and heat treatment time on the microstructure and mechanical properties. A suitable heat treatment regime was adopted to adjust the microstructure of the alloy, thereby enhancing the mechanical properties of the γ-TiAl-based alloy.

2. Materials and Methods

The powder metallurgy sintered Ti-48Al alloy was used as the raw material, and the sintering temperature was 1250 °C [12]. The Ti-48Al alloy in an extruded state was prepared by hot extrusion at 1300 °C with an extrusion speed of 50 mm/s. The 304 stainless steel was used as the jacket material and asbestos was used as the insulation. The alloy diameter after extrusion was approximately 30 mm and the compression ratio was approximately 9:1. Samples of the extruded alloy were cut into φ 10 × 80 mm for the exploration of the heat treatment processes. The appropriate heat treatment temperature was determined within the range of 1300–1360 °C. The effect of cooling rate on the microstructural properties of the alloy was investigated by employing different cooling methods such as water quenching and air cooling. A suitable heat treatment time was established within a range of 3–20 min.

The microstructure of the Ti-48Al alloy was analyzed using a scanning electron microscope (SEM, Zeiss, Oberkochen, Germany) in backscattered electron (BSE) imaging mode. Phase composition was characterized by X-ray diffraction (XRD, D/max-RB, Rigaku, Tokyo, Japan) analysis using Cu Kα radiation at 50 kV and 200 mA, with the diffraction angle 2θ ranging from 10° to 90° at a scan rate of 5°/min. Electron backscatter diffraction
(EBSD) analysis was conducted with an accelerating voltage of 20 kV and a step size of 120 nm, with EBSD data interpreted by the HKL Channel 5 analysis software package. Tensile samples conforming to the ASTM-E08 standard were prepared as M6 standard specimens [13], and room temperature tensile tests were conducted on the hot isostatically pressed samples using an AGI-250KN testing machine.

3. Results

3.1. Microstructure and Properties of Extruded Ti-48Al Alloy

The microstructure and room temperature mechanical properties of the extruded alloy are shown in Figure 1. The microstructure of the extruded alloy consisted of a fine duplex microstructure composed of equiaxed γ grains and α2/γ lamellar colonies, with the equiaxed grains sized between 5 and 10 µm and the lamellar colonies ranging from 10 to 30 µm. The stress–strain curve of the extruded alloy is presented in Figure 1b. At room temperature, the alloy exhibited a tensile strength of 670 MPa and a yield strength of 550 MPa, with an elongation of 1.0%.

![Figure 1](image1.png)

Figure 1. Microstructure and mechanical properties of the extruded Ti-48Al alloy. (a) SEM image, (b) stress–strain curve.

3.2. Effect of Heat Treatment Temperature on the Microstructure and Mechanical Properties of Ti-48Al Alloy

Heat treatment of TiAl-based alloys at different temperatures, particularly in different phase regions, results in various types of microstructures. The cooling rate mainly affects the phase composition, while the holding time mainly affects the size and grain size of lamellar colonies [10,11,14]. Figure 2 shows the microstructure of the Ti-48Al alloy after holding at different heat treatment temperatures for 10 min. After holding at 1300 °C for 10 min, the structure remained duplex, with equiaxed grains of about 10 µm and lamellar colonies of about 20–30 µm, as shown in Figure 2a. The alloy heat-treated at 1320 °C still presented a duplex structure, with no significant change in the size of the equiaxed grains compared to those at 1300 °C, but their proportion had noticeably reduced, as shown in Figure 2b. Figure 2c shows the sample treated at 1340 °C, where the microstructure was near-lamellar, with many γ equiaxed grains transforming into α2/γ lamellar colonies. In the duplex microstructure, a significant number of γ grains were pinned at the interface with the lamellar colonies, restricting grain growth. Therefore, when the temperature was raised to 1340 °C, the γ grains decreased, the pinning effect weakened, and the lamellar colonies rapidly grew in size to about 50 µm. When the heat treatment temperature was further increased to 1360 °C, the structure became fully lamellar, as shown in Figure 2d. At this higher temperature, which exceeded the Tₓ temperature, the size of the lamellar colonies exceeded 100 µm.
The microstructure of the Ti-48Al alloy after holding at different heat treatment temperatures for 10 min: (a) 1300 °C, (b) 1320 °C, (c) 1340 °C, and (d) 1360 °C.

The growth of grains at the alloy interfaces after heat treatment was observed, as shown in Figure 3. For the sample treated at 1300 °C, there were many equiaxed grains at the grain boundaries. The originally fine recrystallized grains at the grain boundaries grew gradually, and the lamellar colonies found it difficult to grow under the pinning of the equiaxed grains [15], showing no significant change in size. As the temperature increased to 1320 °C, the volume fraction of γ grains decreased, the pinning effect weakened, and the lamellar colonies gradually grew in size. However, the lamellar spacing did not change significantly. Interlocking lamellar growth at the interfaces of the lamellar colonies occurred, which was due to the slower cooling rate of air cooling, allowing lamellar growth into adjacent grains.

The microstructure of the Ti-48Al alloy held at different heat treatment temperatures for 10 min: (a) 1300 °C, (b) 1320 °C, and (c) 1340 °C.
When the heat treatment temperature was further increased to 1340 °C, the original α grains grew with the increase in temperature, and there was a significant change in the size of the lamellar colonies, as shown in Figure 3c. The lamellar spacing remained largely unchanged, suggesting that the heat treatment temperature was not the main factor affecting the lamellar spacing. At this point, equiaxed grains were no longer observed at the grain boundaries, indicating that the equiaxed grains had grown and transformed into the α phase at high temperatures, and then transformed into α2/γ lamellar colonies during the cooling process.

The mechanical properties of the Ti-48Al alloy samples after heat treatment at different temperatures were tested at room temperature, and the results are shown in Figure 4. The tensile strength of the alloy treated at 1300 °C was 626 MPa, at 1320 °C it was 713 MPa, at 1340 °C it reached 773 MPa, and at 1360 °C it was 747 MPa. Below the Tα temperature, the strength of the alloy increased with the temperature. This is mainly because as the temperature increases, the proportion of lamellar colonies rises. The more interfaces there are, the greater the obstruction to dislocations [9,16]. At this point, the strengthening effect of the lamellar interfaces on the alloy is greater than that of the fine grain strengthening, hence the increase in alloy strength. However, when the temperature was higher than the Tα temperature, the structure transitioned to a fully lamellar structure, and the size of the lamellar colonies increased rapidly. At 1340 °C, the proportion of lamellar colonies in the structure was already high; further increases in the proportion of lamellar colonies were limited while the size increased substantially, reducing the obstruction to dislocations and thus decreasing the tensile strength. Moreover, when the temperature was below the Tα temperature, the elongation was about 0.70%, while above the Tα temperature, the elongation dropped to about 0.55%. The decrease in elongation can be attributed to two factors: the reduction in the number of equiaxed grains and the increase in the size of lamellar colonies. Therefore, the optimal heat treatment temperature was determined to be 1340 °C for a further exploration of cooling methods and heat treatment durations.

![Figure 4](image_url)

**Figure 4.** The stress–strain curves of the Ti-48Al alloy held at different heat treatment temperatures (1300–1360 °C) for 10 min.

### 3.3. Effect of Cooling Method on Microstructure and Properties

Based on the results, the appropriate heat treatment temperature of 1340 °C was chosen to study the effect of different cooling methods on the microstructure and properties of the alloy. Figure 5 shows the phase composition of the Ti-48Al alloy under different cooling conditions. The phase composition of the microstructure with air cooling was similar to that observed in the as-extruded condition. In the water-quenched microstructure, the α2 phase peaks dominated, with α2 (200) and α2 (201) being the main peaks, while the
Figure 5. The phase composition of the Ti-48Al alloy under different cooling methods. After heat treatment at 1340 °C, the grain size for both the water-quenched and air-cooled samples remained essentially the same. Both conditions resulted in large blocky grains with small grains present at the boundaries, as shown in Figure 6a,c. Comparing the microstructures of the air-cooled (Figure 6b) and water-quenched (Figure 6d) alloys, the microstructure after being air-cooled contained a large number of lamellar colonies, while the microstructure after being water-quenched contained either no lamellae or only a small amount. This difference occurred because in the (α + γ) phase region, due to the rapid cooling rate, the α→γ phase transformation was hindered, and the α phase mainly transformed directly into the α2 phase through ordering reactions [17,18].

The rapid cooling rate of water quenching was effective in preserving the high-temperature microstructure, facilitating the study of phase composition in the (α + γ) phase field and the lamellar formation process. Electron backscatter diffraction (EBSD) analysis was performed on the sample that was heat-treated at 1340 °C for 10 min and then water-quenched. The results are presented in Figure 7. The microstructure contained large blocky grains of α2 phase, with the γ phase primarily present as smaller grains at the boundaries of α2 large grains. The α2 phase was the dominant phase with a volume fraction of 69.5%, while the γ phase had a volume fraction of 28.6%. The α2 phase was a low-temperature ordered phase of the α phase, and some γ lamellae precipitated from within the α phase, indicating the initial stage of lamellae formation. During slow cooling, the γ lamellae would sequentially precipitate and spread throughout the entire α grains. Subsequently, the remaining α phase transforms into the α2 phase through ordering, leading to the formation of α2/γ lamellar colonies [6,19]. The orientation distribution of the γ phase in the microstructure is shown in Figure 7c. At this point, the orientation distribution of grains was relatively uniform, which was different from the extruded state where continuous orientation changes within grains were not observed. This suggests that recrystallization has largely completed, and the subsequent growth of recrystallized grains and phase transformation will primarily occur. The kernel average misorientation (KAM)
map also showed that the overall strain in the sample was low, with a small amount of strain mainly concentrated at defects such as interfaces, as depicted in Figure 7d.

**Figure 6.** The microstructure after being heat-treated at 1340 °C for 10 min followed by different cooling methods (a,b) air cooling, and (c,d) water quenching.

**Figure 7.** The electron backscatter diffraction (EBSD) results of the alloy after a heat treatment of 1340 °C for 10 min followed by water quenching. (a) The band contrast (BC) image, (b) phase distribution map, (c) inverse pole figure (IPF) for the TiAl phase, and (d) kernel average misorientation (KAM) map.
In Figure 8a, partial lamellae can be observed, initiating growth from the boundaries of blocky grains. When combined with the phase distribution map, it is evident that the lamellae preferentially nucleating from the α phase were γ lamellae, and there was no observation of the formation of α2 phase lamellae. In the IPF, the γ lamellae preferentially grew along the (110) direction. Additionally, the interfaces, especially the grain boundaries, exhibited a high density of dislocations. These dislocations and their entanglements provide the necessary energy for the nucleation and growth of the γ phase [20], explaining why the γ-phase lamellae preferentially grew at the α phase grain boundaries. Furthermore, the γ equiaxed grains within the α2 phase grains and at the grain boundaries showed a uniform orientation distribution, with fine grain sizes. These grains represent the residual γ phase that remained after the transformation from γ to α in the (α + γ) phase field during the heat treatment, hence why they were uniformly distributed throughout the alloy matrix.

Figure 8. The EBSD results from the grain boundary area of the sample in Figure 7: (a) band contrast (BC) image, (b) phase distribution map, (c) inverse pole figure (IPF) for the TiAl phase, (d) kernel average misorientation (KAM) map.

In TiAl-based alloys, the transformation of the α phase to γ phase is closely related to the cooling rate. When the cooling rate is fast such as with water quenching, the phase transformation is a non-diffusional massive transformation, that is, a transformation from the disordered hcp phase (α phase) to the ordered L10 (γ phase). When the cooling rate is slower such as with air cooling, the γ phase precipitates from the α phase in a lamellar form and progressively fills the parent phase, thus forming α2/γ lamellar colonies.

Figure 9 shows the stress–strain curves of different microstructures at room temperature. The room temperature tensile strength of the Ti-48Al alloy after air cooling was 773 MPa with an elongation of 0.7%. The tensile strength at room temperature of the water-quenched alloy was 710 MPa with an elongation of about 0.55%. The strength of the air-cooled sample was 8.8% higher than that of the water-quenched sample. The absence of lamellar strengthening in the water-quenched microstructure resulted in lower strength. Studies have shown that the hardness of the α2 phase is higher, about 50% higher than that of the γ phase, and that there is partial lattice mismatch in the atoms of the massive phases.
Therefore, the massive $\alpha_2$ phase is often considered a strengthening phase that can increase the hardness of the alloy. However, in this study, the massive $\alpha_2$ phase had a larger size and lacked internal lamellae, which greatly reduced the interface strengthening effects, and thus its strength was lower than that of the near-lamellar structure. Additionally, due to the fast cooling rate of water quenching, both the high internal stress and the hard massive $\alpha_2$ phase contributed to the reduced plasticity of the alloy, which is why the elongation of the water-quenched sample was lower than that of the air-cooled alloy.

![Stress-strain curves](image)

**Figure 9.** The stress-strain curves of the Ti-48Al alloy after being held at 1340 $^\circ$C for 10 min.

3.4. Effect of Holding Time on Microstructure and Properties

In our previous work, the appropriate heat treatment temperature was determined to be 1340 $^\circ$C, with air cooling as the cooling method. This section will explore the suitable holding time for the alloy. Since the heat treatment temperature is quite high, the grain of the extruded alloy can easily grow, making the control of the holding time crucial. Figure 10 shows the microstructure of the alloy after holding at 1340 $^\circ$C for 3–20 min. When held for 3 min, many equiaxed $\gamma$ grains were still present at the boundaries of blocky grains, indicating that the holding time was too short for many $\gamma$ phases to have had sufficient time to diffuse and transform into the $\alpha$ phase. At 5 min of holding time, the equiaxed $\gamma$ grains in the microstructure decreased rapidly due to the $\gamma \rightarrow \alpha$ phase transformation. The newly formed $\alpha$ phase grew at high temperatures and was constrained by the original $\alpha$ phase, limiting the growth of lamellar colonies within a short time. Moreover, the grain boundaries became jagged and uneven due to the intergrowth of lamellae. Between 10 to 20 min of holding time, the proportion of equiaxed grains in the microstructure was almost consistent, indicating that phase equilibrium had been reached at the current temperature. As the holding time increased, new grains continued to engulf the surrounding grain structures, causing the grains to grow gradually. At 20 min, the phenomenon of lamellae growing through grain boundaries into adjacent lamellar colonies was quite evident, and the grain growth rate at this stage was very fast.

Room temperature tensile properties of the samples treated at different holding times were tested, and their stress–strain curves are shown in Figure 11. The tensile strengths for the samples held for 3 min, 5 min, 10 min, and 20 min were 774 MPa, 784 MPa, 773 MPa, and 718 MPa, respectively. As the holding time increased, the strength of the alloy initially rose and then decreased, reaching its peak at 5 min. This was due to the initial increase in lamellar content leading to a strengthening effect, followed by a decrease in strength caused by grain growth and the resultant weakening of fine-grain strengthening effects as the holding time increased. Concurrently, the ductility gradually decreased with extended
holding times. This was partly due to the reduction in equiaxed crystals and partly due to the increase in the size of the lamellar colonies. Studies in the literature have indicated that lamellar colonies with sizes between 50 and 80 µm and interlamellar spacings less than 1 µm, having the jagged grain boundaries of fully lamellar structures, exhibit optimal mechanical properties, consistent with the results of this section’s research. The finalized heat treatment process determined to be most appropriate was 1340 °C/holding for 5 min/air cooling.

Figure 10. The microstructures after different holding times at 1340 °C with air cooling. (a,e) 3 min, (b,f) 5 min, (c,g) 10 min, and (d,h) 20 min.

Figure 11. The stress–strain curves after different holding times at 1340 °C with air cooling.
The tensile strength of the extruded alloy at room temperature was 670 MPa, and the yield strength was 550 MPa; after heat treatment, the tensile strength was 784 MPa, and the yield strength was 763 MPa. The tensile strength and yield strength of the heat-treated alloy increased by 17% and 38%, respectively.

4. Discussion

In common hot working processes, the lamellar structure of $\alpha_2 + \gamma$ in TiAl alloys decomposes through dynamic recrystallization. However, the $\alpha_2/\gamma$ lamellar structure is difficult to completely destroy through thermomechanical deformation, resulting in residual lamellae. Lin et al. [5] reported that under all deformation conditions of hot compression tests, residual $\alpha_2/\gamma$ lamellar colonies still existed. Moreover, the residual lamellae have a negative effect on strength and ductility [6]. Therefore, the first step in common heat treatment is usually homogenization heat treatment, where the residual lamellae are fully transformed into lamellar colonies and equiaxed grains in the $\alpha + \gamma$ two-phase region after prolonged heat treatment. A subsequent simple heat treatment or cyclic heat treatment is then carried out to achieve alloy microstructure control and performance improvement. The extruded microstructure in this study consisted of fine and uniform dual-phase structures composed of $\gamma$ equiaxed grains (5–10 µm) and $\alpha_2/\gamma$ lamellar colonies (10–30 µm). There were no residual lamellae in the alloy, allowing for direct microstructure control through the heat treatment, thereby enhancing the alloy preparation efficiency.

During heat treatment, different phase regions of TiAl alloys can be controlled for microstructure regulation. Heat treatment near the eutectoid point followed by cooling to room temperature yields near-gamma microstructures. Heat treatment in the $\alpha + \gamma$ two-phase region slightly above the eutectoid point results in a common dual-phase structure, which is the microstructure obtained after extrusion in this study. Heat treatment in the $\alpha + \gamma$ two-phase region slightly below the $\alpha$ phase transition temperature $T_\alpha$ produces near-lamellar microstructures, with the lamellar colony content increasing and the $\gamma$ equiaxed grain content decreasing as the temperature approaches $T_\alpha$. Heat treatment above the $T_\alpha$ temperature results in a fully lamellar microstructure [14]. However, grain growth occurs rapidly, resulting in larger lamellar colony sizes. Table 1 illustrates that as the temperature increased, the proportion of equiaxed grains gradually decreased while the proportion of lamellar colonies increased from 64% to 100%. The $T_\alpha$ transition temperature of Ti-48Al alloy is around 1360 °C. At this temperature, the alloy transforms into a fully lamellar microstructure. Additionally, with an increase in the heat treatment temperature, the average grain size of the $\gamma$ equiaxed grains decreased from 12 µm to 8 µm, while the average size of the lamellar colonies increased from 18 µm to 57 µm. When the alloy transformed into a fully lamellar microstructure, the $\gamma$ equiaxed grains completely disappeared, and the average size of lamellar colonies rapidly increased to 107 µm. Therefore, the size of the lamellar colonies is not only related to temperature, but the proportion of $\gamma$ equiaxed grains also greatly influences the size of the lamellar colonies. In addition, the heat treatment temperature has an important effect on the spacing of the lamellar colonies. As 1300–1340 °C is located in the ($\alpha + \gamma$) two-phase region, and there was no significant difference in the spacing of the lamellar colonies in this temperature interval. However, the spacing of the lamellar colonies increased from 445 nm to 701 nm when the temperature rose above the $T_\alpha$ (1360 °C). Therefore, the phase transition has a crucial effect on the size and spacing of lamellar colonies.
Table 1. Quantitative statistics of the microstructure under different temperatures.

<table>
<thead>
<tr>
<th>Temperature/°C</th>
<th>1300</th>
<th>1320</th>
<th>1340</th>
<th>1360</th>
</tr>
</thead>
<tbody>
<tr>
<td>Equiaxial grain size/µm</td>
<td>12</td>
<td>11</td>
<td>8</td>
<td>-</td>
</tr>
<tr>
<td>Lamellar colonies size/µm</td>
<td>18</td>
<td>28</td>
<td>57</td>
<td>107</td>
</tr>
<tr>
<td>Lamellar interspacing/nm</td>
<td>468</td>
<td>476</td>
<td>445</td>
<td>701</td>
</tr>
<tr>
<td>Lamellar colonies ratio/%</td>
<td>64</td>
<td>79</td>
<td>92</td>
<td>100</td>
</tr>
</tbody>
</table>

Figure 12 illustrates the mechanism diagram of microstructure evolution during the heat treatment process. According to the phase diagram, the Ti-48Al alloy was in the γ phase region below 1255 °C, in the α + γ two-phase region between 1255 and 1360 °C, and in the α single-phase region above 1360 °C. The heat treatment experiments in this study were conducted between 1300 and 1360 °C. The temperature range of 1300–1340 °C falls within the α + γ two-phase region, while 1360 °C lies within the α single-phase region. The temperature range of 1300–1320 °C corresponds to a duplex microstructure. At 1340 °C, the microstructure transitions to a near-lamellar structure, as indicated by the phase diagram, during which the proportion of α phase increases while the proportion of γ phase decreases. During the holding period, the α₂ phase and some γ phase transform into the α phase. Subsequently, during the cooling process, γ lamellae preferentially nucleate in the disordered α matrix, forming α₂/γ lamellar structures [21]. Consequently, the proportion of lamellar colonies increases while the proportion of equiaxed grains decreases. At 1360 °C, the temperature falls within the α single-phase region, where the α₂ phase and all γ phase transform into the α phase. Upon cooling, α₂/γ lamellar structures form, ultimately resulting in a fully lamellar microstructure.
The heat treatment temperature is a crucial factor determining the type of microstructure, while the cooling method determines the shape and width of lamellar colonies. Different cooling rates result in different decomposition products of the α phase; the fastest cooling rate can directly yield a single-phase structure (α2 phase) through ordering, while quenching produces blocky structures, and air cooling favors lamellar structures. The blocky α2 phase has larger sizes and lacks lamellae internally, significantly weakening the interfacial strengthening effects. Table 2 presents the quantitative statistics of the microstructures at different heat treatment times at 1340 °C. With prolonged holding time, there was minimal variation in the size of the equiaxed grains but a significant decrease in the quantity. From a holding time of 3 min to 20 min, the size of the lamellar colonies increased from 31 μm to 75 μm, and their proportion rose from 81% to 94%. During the holding process at 1340 °C, a large number of equiaxed grains transformed into α2/γ lamellar colonies.

<table>
<thead>
<tr>
<th>Holding Time/°C</th>
<th>Equiaxial Grain Size/μm</th>
<th>Lamellar Colonies Size/μm</th>
<th>Lamellar Colonies Ratio/%</th>
</tr>
</thead>
<tbody>
<tr>
<td>3</td>
<td>10</td>
<td>31</td>
<td>81</td>
</tr>
<tr>
<td>5</td>
<td>8</td>
<td>43</td>
<td>88</td>
</tr>
<tr>
<td>10</td>
<td>8</td>
<td>57</td>
<td>92</td>
</tr>
<tr>
<td>20</td>
<td>8</td>
<td>75</td>
<td>94</td>
</tr>
</tbody>
</table>

Figure 13 depicts the growth process of lamellar colonies during the heat treatment. Lamellar colonies grow into adjacent γ equiaxed grains with similar orientations, gradually engulfing them. As a result, the size and number of equiaxed grains progressively decrease. At the same time, due to the lack of pinning effect from the equiaxed grains, the growth rate of the lamellar colonies increases. Ultimately, when the proportion of equiaxed grains and lamellar colonies reaches a balance, the growth of lamellar colonies stops.

Figure 13. Lamellar colony growth in heat treatment.

A significant portion of interfaces in TiAl-based alloys serves as one of the primary strengthening mechanisms. However, in Ti-48Al alloys after heat treatment, the effects of dislocation strengthening, solid solution strengthening, and precipitation strengthening are relatively weak. Therefore, the strengthening effects resulting from internal interfaces in the alloy can be described using the Hall–Petch relationship, as shown in Equation (1):

\[ \sigma_y = \sigma_0' + k_y \cdot d^{-1/2} \]  

(1)

where \( \sigma_y \) is the yield strength, \( \sigma_0' \) is the stress portion unaffected by grain size, \( k_y \) is the material constant reflecting the difficulty of dislocation slip, and \( d \) is the average grain size. In alloys with duplex and near-lamellar microstructures, relevant structural parameters include the grain size of equiaxed α2 and γ phases, lamellar colony size, and lamellar spacing. Yoo et al. found that \( k_y = 0.91 \text{ MPa} \cdot \text{m}^{-1/2} \) for equiaxed microstructures of TiAl-based alloys [22] while Dimiduk et al. found \( k_y = 2.7 \text{ MPa} \cdot \text{m}^{-1/2} \) for the fully lamellar microstructures of Ti-45.5Al-2Cr-2Nb. Additionally, the \( k_y \) at the lamellar interface
was approximately 0.45 MPa·m$^{-1/2}$, and the lamellar spacing was inversely proportional to $\sigma_y$ [23]. Hence, the strengthening effect of the lamellar colonies is higher than that of equiaxed grains, and the smaller the lamellar spacing, the more significant the strengthening effect. During the process of increasing the heat treatment temperature, the proportion of lamellar colonies continuously increases, leading to an enhancement in the mechanical properties of the alloy. However, when the temperature further increases beyond $T_\alpha$, the lamellar spacing significantly enlarges, resulting in a decrease in the mechanical properties of the alloy.

Under the mechanism described, this study achieved near fully lamellar microstructure control at the parameters of $1340 \, ^\circ C$, holding for 5 min, and air cooling. The average size of the lamellar colonies was $57 \, \mu m$, and the room temperature tensile strength reached 784 MPa. Compared to the performance reported in the literature, the alloy performance in this study was excellent, as shown in Table 3.

### Table 3. Tensile properties of the $\gamma$-TiAl alloys at room temperature.

<table>
<thead>
<tr>
<th>Alloy Composition</th>
<th>Process</th>
<th>Tensile Strength (MPa)</th>
<th>Yield Strength (MPa)</th>
<th>Elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti-48Al (this work)</td>
<td>HT</td>
<td>784</td>
<td>763</td>
<td>0.8</td>
</tr>
<tr>
<td>Ti-48Al-2Cr-2Nb [9]</td>
<td>CHT</td>
<td>697</td>
<td>-</td>
<td>2.1</td>
</tr>
<tr>
<td>Ti-48Al-2Cr-2Nb [24]</td>
<td>Forging</td>
<td>474</td>
<td>357</td>
<td>1.64</td>
</tr>
<tr>
<td>Ti-48Al-2Cr-2Nb [25]</td>
<td>HT</td>
<td>477</td>
<td>577</td>
<td>1.2</td>
</tr>
<tr>
<td>Ti-44.5Al-1.0Cr-2.5V-2.0Mo [26]</td>
<td>HE</td>
<td>540</td>
<td>481</td>
<td>-</td>
</tr>
</tbody>
</table>

HT—heat treatment, CHT—cyclic heat treatment, HE—hot extrusion, EBM—electron beam selective melting.

### 5. Conclusions

This study focused on the Ti-48Al alloy in its extruded state, investigating the effects of heat treatment temperature, cooling method, and holding time on the alloy’s microstructure and properties, and achieved the control of fully lamellar fine microstructures and the enhancement of performance through a simple heat treatment, rather than the traditional approach of homogenization followed by heat treatment. The main conclusions are as follows:

(1) The heat treatment temperature determines the type of microstructure, while the cooling rate determines the width of the lamellae. As the heat treatment temperature increases from the two-phase region to the $\alpha$ single-phase region, the microstructure transitions from dual-phase to near-lamellar. The strength of the alloy initially increases and then decreases, which is determined by both the number of lamellae and the grain size. With rapid cooling (water cooling), non-diffusive type blocky phase transformations occur, while slower cooling (air cooling) gradually forms $\alpha_2/\gamma$ lamellar colonies.

(2) The heat treatment process identified as suitable for powder Ti-48Al alloy was $1340 \, ^\circ C$, holding for 5 min, followed by air cooling. Under these conditions, the microstructure of the alloy as near-lamellar, consisting of lamellar colonies (about 50 $\mu m$) and a small amount of fine $\gamma$ equiaxed grains (about 10 $\mu m$). The room temperature tensile strength was 784 MPa, and the yield strength was 763 MPa, which represents improvements of 17.0% and 38.7%, respectively, compared to the extruded state.

### Author Contributions:

Conceptualization, M.Y.; Writing—original draft preparation, M.Y.; Funding acquisition, M.Y., H.Z. and H.F.; Writing—review and editing, H.Z., H.F. and Z.H.; Methodology, H.Z. and H.F.; Data curation, Z.H.; Investigation, Y.G. and F.Y.; Formal analysis, Y.G. and F.Y. All authors have read and agreed to the published version of the manuscript.
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Conflicts of Interest: The authors declare no conflicts of interest.

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