Effects of Powder Preparation and Sintering Temperature on Properties of Spark Plasma Sintered Ti-48Al-2Cr-8Nb Alloy

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Received: 9 July 2019; Accepted: 6 August 2019; Published: 7 August 2019

Abstract: TiAl alloy has become a key element in aerospace and automotive engine development due to its favorable high temperature mechanical properties and low density. In this paper, high performance TiAl alloy was prepared using atomized Ti-48Al-2Cr-8Nb powder by spark plasma sintering. This paper analyzed the variation of density, microstructure, Vickers hardness, and fracture strength of TiAl alloys prepared with spherical pre-alloyed powder (named as SP powder) and pre-alloyed powder after 12 h of ball milling (named as MP powder) at different sintering temperatures. The results indicate that the density, Vickers hardness, and room temperature (25 °C) bending strength of Ti-48Al-2Cr-8Nb alloy sintered using MP powder, are significantly higher than that of TiAl alloy sintered using SP powder. Specifically, the densification temperature of the MP powder sintered specimen is reduced by 100 °C, the Vickers hardness is increased by 15%, and the room temperature bending strength is increased by 51.9% at a sintering temperature of 1250 °C. The microstructure analysis shows that the Ti-48Al-2Cr-8Nb alloy has the best bending strength when it has a fine grain phase structure.

Keywords: spark plasma sintering; ball milling process; mechanical properties; fine grain strengthening; TiAl alloy

1. Introduction

TiAl alloys are ideal materials for aerospace and automotive engines because of their low density, good high temperature strength, creep resistance, and oxidation resistance [1–3]. However, poor plasticity at room temperature and relatively low strength at high temperature severely limit the practical application of TiAl alloys [4–6]. Recently, research on how to improve the properties of TiAl alloys has received wide attention.

Spark plasma sintering (SPS) is an emerging sintering method. During the past decade the SPS method was successfully applied for the synthesis of a number of innovative composite materials with metals use. These include Cu, stainless steel and Ti, Co, Nb, Zr-Co-Nb-Si-B rare-earth free permanent magnets, Nd-Ce co-doped Gd2Zr2O7 ceramics and Mo–Cu pseudo-alloys [7–12]. By means of pressurization and pulse current, the material can be densified in a short time, and its grain size is greatly reduced which improves the strength of the material [13,14]. However, due to heat radiation during the sintering process, the mold loses heat during the heat transfer process which causes the material to have a temperature gradient from the center to the surface [15]. The addition of Nb to TiAl alloy can further improve its high temperature yield strength, high temperature oxidation resistance, and high temperature creep resistance [16–18]. However, due to the high melting point of Nb (2468 °C) [19], high contents of Nb can lead to an increased melting point, and detrimental inhomogeneity and segregation may also appear in the microstructures of the TiAl alloy making
The addition of Nb also results in solid solution strengthening which increases the deformation resistance of the Nb-TiAl alloy. This reduces the plastic deformability of high Nb titanium alloys [21]. Therefore, understanding how to improve the plasticity of high Nb titanium-aluminum alloys is a key problem to be solved in the preparation process. Presently, the ways for improving the plasticity of TiAl alloys include microstructure modifications, grain size refinement, and alloying. The mechanical milling process can significantly improve the mechanical properties of TiAl alloys by reducing the sintering temperature, decreasing the size of the pre-alloyed particles, and increasing the reactivity and the degree of alloying. Guyon [22] found that after ball milling, the pre-alloyed powder has a high defect density, smaller particle size, and jagged particle surface. These results are due to the high strength plastic deformation causing the densification temperature to be 200 °C lower than that of the spherical pre-alloyed powder. Through tensile strength experiments, Wang [23] found that the fracture strength of the milled powder after sintering is higher than that of the non-milled powder. This lead to the conclusion that a fine grained and dense high performance material can be obtained by selecting suitable milling parameters. Sanjay [24] found that pre-alloyed powder after milling has a significantly accelerated diffusion process during sintering. This causes the coarse dendritic structure to disappear, forming a fine microstructure with a specific phase distribution, where the material becomes lamellar and equiaxed and has a higher fracture strength and ductility. It has been found that initial pre-alloy powder prepared using gas atomization has problems such as coarse particle size and microstructure segregation. This leads the microstructure to be coarse and unevenly distributed inside the material after SPS sintering. Therefore, satisfying the mechanical properties of materials for practical application is difficult.

In this paper, a spherical pre-alloyed powder (SP powder) and a mechanical milled 12 h pre-alloyed powder (MP powder) were used to prepare Ti-48Al-2Cr-8Nb alloy using spark plasma sintering. Parameters such as particle size distribution, surface morphology, and microstructure of the two pre-alloyed powders were compared and analyzed. The resulting differences in density, Vickers hardness, microstructure, and cross-sectional morphology of the Ti-48Al-2Cr-8Nb alloys prepared using the two different kinds of pre-alloyed powders, and at different sintering temperatures, were studied. Additionally, the bending strength and fracture mechanism of the prepared Ti-48Al-2Cr-8Nb alloys at different bending test temperatures were analyzed to see the effect of the powder state on these properties.

2. Materials and Methods

In the present study, a pre-alloyed powder of 80–250 mesh produced using gas-atomization with a nominal composition of Ti-48Al-2Cr-8Nb was used (the powder was developed by ourselves). In the following text, this material will be referred to as spherical pre-alloyed powder (named as SP powder). A portion of the SP powder was ground using a planetary ball mill (QM-3SP2, Nanda, Nanjing, China). The specific parameters of the ball milling were as follows: The ball milling time was 12 h, the rotation speed was 250 rpm, the grinding ball was made of WC (tungsten carbide), and the ball to powder ratio was 5:1. In the remainder of the paper, the ball-milled powder will be referred to as the MP powder. The two powders were poured into a graphite mold with a bore size of 20 mm × 20 mm and were then sintered using a spark plasma sintering system (LABOXTM-650F, Sinter Land INC., NAGAOKA, Japan). The sintering vacuum pressure was less than 6 Pa, and the sintering pressure was 30 MPa. In order to prevent the occurrence of over-burning, the temperature was raised to 50 °C below the target sintering temperature at a heating rate of 100 °C/min and then raised to the target sintering temperature at a rate of 50 °C/min. The molds were then taken out of the furnace and cooled to below 100 °C after a holding time of 10 min. After cooling, the graphite paper on the sample surface was removed, and the sample density was measured using the Archimedes drainage method. The etching solution used in the present work was Kroll reagent (5% HF + 15% HNO₃ + H₂O), and the microstructure was observed using a scanning electron microscope (SEM, Quant250FEG, FEI, Hillsboro, OR., USA) and optical microscopy (OM, BX53M, OLYMPUS, Tokyo,
Japan). Phase composition analysis was performed using a Bruker D8 X-ray diffractometer (XRD, D8 Advance, Bruker, Leipzig, Germany). The bending mechanical properties tests were performed using a universal testing machine (UTM5105-G, Hengsisheng, Jinan, China).

3. Results and Discussion

3.1. Characterization of the As-Atomized SP Powder and MP Powder

Figures 1 and 2 show the morphology and particle size distribution of the Ti-48Al-2Cr-8Nb pre-alloyed powder, respectively. It can be seen that the particle size of the SP powder is primarily distributed within the range of 60–180 μm. Because of the gas atomization process, the shape of the powder is nearly spherical, and the surface of the powder particles is a dendritic crystal structure. Some small-sized particles stick to the molten surface of large-sized particles due to the high cooling rate during the curing process (upper right corner in Figure 1a). Figure 1b shows that the surface morphology of the MP powder exhibits an irregular fracture shape. This is primarily because the powder undergoes multiple reciprocating deformations, fracture, and cold welding during the high energy ball milling process [25]. Compared with the SP powder, the irregular and small-sized powder formed by the mechanical ball milling process has better packing properties during the contact of the powder particles. In addition, the accumulated defects and fine grained structure in the milled powder particles can enhance the diffusion kinetics during subsequent sintering, resulting in improved sinterability and chemical homogeneity [21]. Figure 2b shows that the particle size distribution of the MP powder after 12 h of high energy ball milling is predominantly between 10–20 μm. This is 6–8 times lower than prior to the ball milling.

![Figure 1](image1.png)

**Figure 1.** The shape of Ti-48Al-2Cr-8Nb pre-alloyed powder. (a) SP powder and (b) MP powder.

![Figure 2](image2.png)

**Figure 2.** Particle size distribution of Ti-48Al-2Cr-8Nb powder. (a) SP powder and (b) MP powder.

Figure 3 shows the XRD patterns of the Ti-48Al-2Cr-8Nb pre-alloyed powder. It can be seen from Figure 3a that the atomized SP powder mainly consists of α2 (Ti3Al) phase, γ (TiAl) phase, a small...
amount of TiAl₂ phase, and TiAl₃ phase. During the atomization process, the powder undergoes a transition from a liquid phase to a solid phase which is accompanied by a microstructure transformation process. Due to the different cooling rates, the phase transformation sequence of small-sized particles (<100 μm) during solidification was suggested as [26]:

\[
L \rightarrow L + \beta \rightarrow \beta \rightarrow \alpha \rightarrow \alpha_2
\]  

(1)

Due to the segregation of the Al element in the TiAl alloy bar during the casting process, a small amount of TiAl₂ phase, and TiAl₃ phase were found in the XRD after atomization. Figure 3(b) shows the XRD pattern of the MP powder. Compared with the XRD analysis results of the SP powder, the observed waveform of the MP powder was significantly changed. In particular, the amplitude of the wave was significantly reduced, and severe broadening was observed. The main reason for the broadening of the XRD peak can contribute to the accumulation of defects [24]. During the process of mechanical milling, large amounts of lattice defects such as dislocations and point defects accumulated in the powder particles, resulting in the broadening of the XRD peak after ball milling. The microstructure of the Ti₂Al phase was observed after identifying the phase of the MP powder. The presence of Ti₂Al phase demonstrated that the pre-alloyed powder produced a chemical reaction during the mechanical ball milling process, which resulted in a change of the powder microstructure.

\[
\begin{align*}
\gamma &- \text{TiAl} \\
\alpha_2 &- \text{Ti₃Al} \\
\beta &- \text{TiAl}_2 \\
\alpha &- \text{TiAl}_3 \\
\gamma &- \text{Ti} \\
\end{align*}
\]

Figure 3. XRD patterns of Ti-48Al-2Cr-8Nb pre-alloyed powder. (a) SP powder and (b) MP powder.

For large-sized particles the phase transformation sequence of large-sized particles during solidification was suggested as [27]:

\[
L \rightarrow L + \beta \rightarrow \beta \rightarrow \alpha \rightarrow \alpha_2 + \gamma
\]  

(2)

where \(L\) denoted the liquid phase, \(\beta\) represented the \(\beta\)-Ti phase, \(\alpha\) denoted the \(\alpha\)-Ti phase, \(\gamma\) denoted the TiAl phase, and \(\alpha_2\) denoted the Ti₃Al phase.

Due to the segregation of the Al element in the TiAl alloy bar during the casting process, a small amount of TiAl₂ phase and TiAl₃ phase were found in the XRD after atomization. Figure 3b shows the XRD pattern of the MP powder. Compared with the XRD analysis results of the SP powder, the observed waveform of the MP powder was significantly changed. In particular, the amplitude of the wave was significantly reduced, and severe broadening was observed. The main reason for the broadening of the XRD peak can contribute to the accumulation of defects [24]. During the process of mechanical milling, large amounts of lattice defects such as dislocations and point defects accumulated in the powder particles, resulting in the broadening of the XRD peak after ball milling. The microstructure of the Ti₂Al phase was observed after identifying the phase of the MP powder. The presence of Ti₂Al phase demonstrated that the pre-alloyed powder produced a chemical reaction during the mechanical ball milling process, which resulted in a change of the powder microstructure.

3.2. Effect of Sintering Temperature on Density and Vickers Hardness of Ti-48Al-2Cr-8Nb Alloy

The two primary factors resulting in SPS sintering densification are the sintering pressure and sintering temperature. In the present experiment, a constant pressure sintering method was used to analyze the influence of sintering temperature on the densification process. Figure 4 shows the density and Vickers hardness of the Ti-48Al-2Cr-8Nb sintered material as a function of sintering temperature. It can be seen from Figure 4a that the density of the SP powder shows an increasing trend with increased sintering temperature. The trend becomes more subtle when the sintering temperature reaches 1300 °C, and when the relative degree of densification of the sintered specimen reaches 99.3%. Similar to the SP powder, the density of the MP powder sintered specimen increases with rising temperature and subsequently becomes more subtle above 1200 °C. At this temperature, the relative densification degree
of the sintered specimen reaches 99.8%. By comparing the sintering densification temperatures of the two powders, it was found that the sintering densification temperature of the MP powder was approximately 100 °C lower than the SP powder. This result is primarily caused by the refinement and crushing of the ball milling process. Because of this process, the size of the powder after refinement is significantly reduced, and the particle size is uneven. The resulting smaller sized particles are able to occupy the gaps between large sized particles before the particles are compacted under pressure. Moreover, since the shape of the SP powder is a regular sphere, a small amount of the sintered neck can be formed at the adjacent contact points as the sintering temperature increases. Furthermore, it rapidly grows and the material is densified under the driving force of the sintering. However, since more contact points or even contact faces are produced among the MP powder particles under pressure, the densification process can be completed in a shorter time.

![Figure 4](image.png)

**Figure 4.** Density, Vickers hardness and average grain size of Ti-48Al-2Cr-8Nb alloy as a function of sintering temperature. (a) Density and (b) Vickers hardness and average grain size (densation greater than 97%).

It can be seen from Figure 4b that the Vickers hardness gradually increases with increasing density. For example, the density and Vickers hardness simultaneously increase when the sintering temperature of the SP and MP powders are increased from 1200 °C to 1300 °C and from 1100 °C to 1250 °C, respectively. After the densification is essentially complete, the resulting Vickers hardness of the sintered specimen is directly related to its internal structure [28]. Figure 5 shows that the microstructure of the SP powder sintered specimen is predominantly a grain phase structure when the sintering temperature reaches 1300 °C. At this point, the corresponding Vickers hardness is 290 HV. When the temperature rises to 1350 °C, its internal structure becomes a duplex (DP), and the resulting grain coarsening causes a decrease in Vickers hardness. The Vickers hardness of the MP powder sintered specimens does not change significantly with temperature. When the temperature is 1250 °C, the Vickers hardness reaches a maximum value of 333 HV, which is 15% higher than the maximum Vickers hardness prior to ball milling. In general, when the sintered specimen is densified, the Vickers hardness is primarily affected by the microstructure and grain size.

### 3.3. Bending Strength and Fracture Mechanism

The bending strength of Ti-48Al-2Cr-8Nb material prepared with different pre-alloyed powders as a function of sintering temperature and bending test temperature is shown in Figure 6. It can be seen from Figure 6a that the bending strength of the Ti-48Al-2Cr-8Nb material prepared by the SP powder increases first and then decreases at room temperature. It is well known that the mechanical properties of TiAl alloys are related to their internal microstructure.
Figure 5. The microstructure of specimens sintered by the SP powder at 1300 °C and 1350 °C. (a) At 1300 °C and (b) at 1350 °C.

Figure 6. Curve of bending strength of Ti-48Al-2Cr-8Nb alloy with sintering temperature and ambient temperature. (a) Sintered by the SP powder and (b) sintered by the MP powder.

It can be seen from the Ti-Al-Nb ternary phase diagram [29] in Figure 7 that when the sintering temperatures are 1200, 1250, 1300 and 1350 °C, the materials are located in the γ single-phase region, the bottom of the α + γ phase region, the middle of the α + γ phase region and the top of the α+γ phase region, respectively. Figure 8 shows XRD patterns of TiAl alloys sintered at 1200, 1250, 1300 and 1350 °C. The XRD patterns confirm that the phases in the alloys sintered using SP powder consist of γ-TiAl, TiAl3 and TiAl2, which are the same as that using MP powder. Meanwhile, the intensity of the diffraction peaks of the alloys sintered with MP powder increases significantly compared with those sintered with SP powder. The difference mainly comes from the reduced grain size and the accumulated defects in MP powders [23].

Figure 9a shows that there are numerous pores in the sintered specimen and the powder boundary is still visible (the red area) when the sintering temperature is 1200 °C. At this temperature, the sintered specimen has a low bending strength. When the sintering temperature is 1250 °C, the microstructure of the sintered specimen, as shown in Figure 9b, is primarily composed of fine grain (region A) and spherical grain phases (region B). At this temperature, the bending strength reaches a maximum value of 644 MPa. As shown in Figure 9c, when the sintering temperature is raised to 1300 °C, the bending strength of the material is 598 MPa. At this temperature, the size of grain phase becomes larger than that at 1250 °C which leads to a slightly lowered bending strength. As shown in Figure 9d, when the sintering temperature is 1350 °C the microstructure of the material is mainly lamellar with a small amount of fine and spherical grains. At this temperature, the average grain size of the specimen was significantly increased causing the bending strength to be remarkably lowered. This result follows the Hall–Petch relationship.
As a high temperature alloy, TiAl alloy has good mechanical properties in a high temperature environment. It can be seen from Figure 6a, that when the ambient temperature is between 600 °C and 800 °C, the bending resistance of the Ti-48Al-2Cr-8Nb material prepared using the SP powder is significantly higher than at room temperature. Compared to RT (room temperature, 25°C) and 800 °C, the material achieves the best overall performance with respect to bending at an ambient temperature of 600 °C. It is worth mentioning that the trend of the temperature dependence of the bending strength of the Ti-48Al-2Cr-8Nb material prepared at different sintering temperatures is changed when the sintering temperature is 1350 °C. Compared with the room temperature environment, the bending strength of the Ti-48Al-2Cr-8Nb material decreased significantly at a sintering temperature of 1350 °C. However, the lamellar structure obtained at the sintering temperature of 1350 °C has better bending resistance than the grain phase structure obtained at 1300 °C.

It can be seen from Figure 6b that the bending strength of Ti-48Al-2Cr-8Nb material prepared using the MP powder under different bending test temperatures generally decreases first, then increases and finally decreases with the increase of sintering temperature. This is primarily related to the resulting changes in the microstructure of the material. It can be seen from Figure 10 that the microstructure of the sintered specimen is primarily grain phase when the sintering temperature is between 1200 °C and 1300 °C. In this range, the size of the grain phase undergoes a process of refining and then growing, causing the bending strength to decrease and then increase. When the sintering temperature is 1350 °C, the internal structure of the sintered specimen is a large-sized lamellar structure, which causes the bending strength at room temperature to be significantly reduced. Similarly, for the SP powder
sintered specimen, the bending strength of the material is significantly increased at the bending test temperatures of 600 °C and 800 °C, and reaches a maximum at 600 °C. It is worth noting that the lamellar structure has a higher bending strength at 800 °C than the bending test temperature of 600 °C. By comparing Figure 6a,b, it can be concluded that the mechanical ball milling process can greatly improve the bending strength of TiAl alloy.

![Figure 9. Microstructure of Ti-48Al-2Cr-8Nb alloy prepared by the SP powder with different sintering temperatures: (a) 1200 °C, (b) 1250 °C, (c) 1300 °C, and (d) 1350 °C.](image)

The influence of the ball milling process on the material bending resistance is presented in Table 1. The grain size after ball milling is approximately 50% smaller than prior to ball milling when the sintering temperature is 1300 °C. However, the room temperature bending strength is drastically increased from 598 MPa to 1018 MPa demonstrating that material properties can be significantly improved using the ball milling process to refine grain size. These results can be explained from the perspective of crack expansion. It can be seen from the white areas of Figure 9b,c that, the coarse spherical crystals first generate cracks and subsequently expand along the fine grained structure under pressure. The ball milling process can refine the size of the spherical crystals so that the cracks do not easily grow in the small-sized spherical crystal, thereby increasing the strength of the sintered specimen. It is worth mentioning that the lamellar size of the Ti-48Al-2Cr-8Nb material prepared using the MP powder is 2–3 times smaller than that of the SP powder when the sintering temperature is 1350 °C. However, the bending strength is only increased by 2%, indicating that the effect of refining the lamellar size on the bending strength of TiAl alloy is not significant. By comparing the microstructure transformation process of the SP and the MP powders at different sintering temperatures, it can be found that the microstructure of the two powders was mainly grain phase at temperatures lower than 1300 °C. On the other hand, the Ti-48Al-2Cr-8Nb material prepared by the SP powder had a duplex structure (with a dominant phase being the lamellar phase) when the sintering temperature reached 1350 °C. Conversely, the material prepared using the MP powder was a near-lamellar structure, which illustrates that the ball milling process can reduce the phase transition temperature, reduce the time required for sintering, shorten the process of grain growth, and further increase the strength of the material.
The influence of the ball milling process on the material bending resistance is presented in Table 1. The grain size after ball milling is approximately 50% smaller than prior to ball milling when the sintering temperature is 1300 °C. However, the room temperature bending strength is drastically increased from 598 MPa to 1018 MPa demonstrating that material properties can be significantly improved using the ball milling process to refine grain size. These results can be explained from the perspective of crack expansion. It can be seen from the white areas of Figures 9(b) and (c) that, the

The fracture surface morphologies of the Ti-48Al-2Cr-8Nb materials prepared using the SP and MP powders at different bending test temperatures and a constant sintering temperature of 1250 °C are shown in Figure 11. It shows that the microstructure of the specimen prepared using the SP powder is primarily grain phase with large size. The main fracture modes are transgranular and intergranular fractures. It can also be clearly observed, that area A has a visible river pattern, and its fracture mode is transgranular. Area B is flat with traces of grain breakage and extraction, and the fracture modes are a mixture of both transgranular and intergranular fractures. In Figure 11c, an enlarged view of the highlighted red region is shown in the upper right corner. When the bending test temperature is 600 °C, the fracture mode of the Ti-48Al-2Cr-8Nb material prepared using the SP powder becomes different. At this condition, the main fracture mode is a transgranular fracture. Correspondingly, the bending strength is increased by 18.7% compared with room temperature. These observations show that the high temperature environment can effectively change the fracture mode of the material, thereby improving the relevant mechanical properties. Figure 11b shows that the grain size of the Ti-48Al-2Cr-8Nb material prepared using the MP powder is greatly reduced, and the crystal grains are more closely bonded. The room temperature fracture mode is dominated by transgranular and

| Temperature (°C) | Time (h) | Density (g/cm³) | Size (µm) | Vickers Hardness (HV) | T_{room} Bending Strength (MPa) | Microstructure |
|------------------|----------|----------------|-----------|-----------------------|-------------------------------|----------------|
| 1200             | 0        | 4.080          | /         | 252.6                 | 440.5                         | Grain phase    |
| 1200             | 12       | 4.250          | 15–20     | 324.6                 | 635                           | Grain phase    |
| 1250             | 0        | 4.181          | 10–15     | 287.9                 | 644                           | Grain phase    |
| 1250             | 12       | 4.254          | 3–6       | 333.2                 | 1025                          | Grain phase    |
| 1300             | 0        | 4.226          | 8–13      | 290.5                 | 598                           | Grain phase    |
| 1300             | 12       | 4.239          | 5–8       | 331.5                 | 1018                          | Duplex (less lamellar) |
| 1350             | 0        | 4.244          | 200–300   | 286.5                 | 485                           | Duplex (more lamellar) |
| 1350             | 12       | 4.252          | 50–100    | 326.4                 | 494                           | Near lamellar  |
intergranular fractures. When the crack expands to a half position, the crack propagation path is deflected due to the strong grain bonding force. This effect leads to the appearance of the valley at arrow C shown in the figure. When the MP powder was sintered at 1250 °C, the room temperature bending strength of the sintered specimen was 1025 MPa, which was 59.1% higher than that of the SP powder at the same temperature. Figure 11d shows that when the bending test temperature is 600 °C, the fracture mode of the Ti-48Al-2Cr-8Nb material prepared using SP powder is an intergranular fracture. During the crack extension process, the crack is deflected at the grain of the intergranular fracture. This extends the crack propagation path and absorbs the expansion energy of the crack to a certain extent, thereby improving the critical stress of the material fracture [30]. At this temperature, the bending strength of the material was 1140 MPa.

![Fracture surface morphology of Ti-48Al-2Cr-8Nb alloy at sintering temperature of 1250 °C.](image)

**Figure 11.** Fracture surface morphology of Ti-48Al-2Cr-8Nb alloy at sintering temperature of 1250 °C. (a) Alloy sintered by the SP powder under room temperature, (b) alloy sintered by the MP powder under room temperature, (c) alloy sintered by the SP powder under 600 °C, and (d) alloy sintered by the MP powder under 600 °C.

### 4. Conclusions

In this work, high performance TiAl alloy was prepared using atomized Ti-48Al-2Cr-8Nb powder by spark plasma sintering. The variation of density, microstructure, Vickers hardness, and fracture strength of TiAl alloys prepared with spherical pre-alloyed powder (SP powder) and pre-alloyed powder after 12 h of ball milling (MP powder) at different sintering temperatures were analyzed. The findings of this work can provide guidance for the pre-treatment of atomized powder and the sintering process of TiAl alloy. From the experimental results, the following conclusions can be made.

1. The particle size of the MP powder is predominantly distributed in the range of 10–20 µm, which is approximately 6–8 times smaller than that of the SP powder. The MP powder morphology is changed from a spherical to a broken shape. The SP powder is mainly composed of γ(TiAl), α2 (Ti3Al), TiAl2, and TiAl3 phases, and the MP powder mainly consists of γ(TiAl), α2 (Ti3Al), TiAl2, TiAl3, and Ti2Al phases.
(2) When the SP powder was sintered at 1300 °C, the material was close to densification and the Vickers hardness reached a maximum of 290 HV. The material of the MP powder was close to densification when the sintering temperature was 1200 °C. The Vickers hardness for the MP powder material reached a maximum value of 333 HV when the sintering temperature was 1250 °C.

(3) The phases in the alloys sintered at 1200, 1250, 1300 and 1350 °C using SP powder consisted of γ-TiAl, TiAl₃ and TiAl₂, which are the same as that by using MP powder.

(4) The SP powder and the MP powder have the best room temperature bending strength at a sintering temperature of 1250 °C. Compared to RT and 800 °C, the mechanical properties of these alloys are best at a bending test temperature of 600 °C. The dominant fracture modes of the sintered SP powder sintered at 1250 °C are mainly transgranular and intergranular fractures at room temperature. At 600 °C, the transgranular fracture is dominant. The dominant fracture modes of the MP powder sintered specimen at room temperature are mainly transgranular and intergranular fractures. Additionally, the MP powder sintered material shows an obvious intergranular fracture phenomenon at a bending test temperature of 600 °C.

Author Contributions: Conceptualization, Z.W.; methodology, Z.W.; validation, Z.W., H.S., Y.D. and J.Y.; investigation, H.S. and Z.W.; data curation, H.S. and Y.D.; writing—original draft preparation, H.S. and Z.W.; resources, Z.W.; writing—review and editing, Z.W. and H.S.; supervision, Z.W.; project administration, Z.W.; funding acquisition, Z.W.

Funding: This work is supported by National Natural Science Foundation of China (grant numbers 51571116 and 51775280) and the Jiangsu Provincial Six Talent Peaks Project (grant number 2016-HKHT-019).

Conflicts of Interest: The authors declare that they have no conflict of interest.

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