Study on Non-Isothermal Transformation of Ti-6Al-4V in Solution Heating Stage

Hui Yu 1,2,*, Wei Li 1, Haibei Zou 1, Songsong Li 1, Tongguang Zhai 3 and Ligang Liu 1,*

1 College of Mechanical Engineering, Yanshan University, Qinhuangdao 066004, China
2 National Engineering Research Center for Equipment and Technology of Cold Steel Rolling, Qinhuangdao 066004, China
3 Department of Chemical and Materials Engineering, University of Kentucky, Lexington, KY 40506, USA
* Correspondence: yuhui@ysu.edu.cn (H.Y.); lgliu@ysu.edu.cn (L.L.); Tel.: +86-137-8035-0181 (H.Y.); +86-0335-804-7771 (L.L.)

Received: 18 July 2019; Accepted: 31 August 2019; Published: 3 September 2019

Abstract: In order to understand the non-isothermal transformation behavior of Ti-6Al-4V titanium alloy in the continuous heating stage of solution treatment, thermal dilatometry tests with heating rates of 0.1~0.8 °C/s were designed. The conversion between the expansion amount and the transformed volume fraction was realized by the lever principle, and the transformation characteristics of \( \alpha + \beta \rightarrow \beta \) were quantified based on the Kissinger-Akahira-Sunose (KAS) theory and the modified Johnson–Mehl–Avrami (JMA) model. The results show that the phase transformation kinetics curves present typical “S” patterns, and the element diffusion transformation controls the nucleation and growth of new grains during the transformation of \( \alpha + \beta \rightarrow \beta \). The phase transformation interval gradually moves to high temperature regions with the increase of heating rates, and the phase transformation activation energy is 445.5 kJ·mol\(^{-1}\). The phase transformation process is divided into three stages according to the relationship between the Avrami exponent \( n \) and the transformed volume fraction \( f_T \). These three stages correspond to different stages of grain nucleation or growth.

Keywords: Ti-6Al-4V; solution treatment; non-isothermal dilatometry; JMA model; transformation kinetics

1. Introduction

The most basic phases in titanium alloy are the room temperature stable \( \alpha \) phase (HCP) and the high temperature stable \( \beta \) phase (BCC). The coexistence of these two phases at room temperature can be realized through the addition of a stabilizing alloy element. The mutual transformation between the \( \alpha \) phase and the \( \beta \) phase is the basis of all phase transformations of titanium alloy [1]. Under the premise of consistent composition, phase structural changes lead to significant differences in the macroscopic mechanical properties of materials [2–5]. Therefore, a lot of research has been done to clarify the influence of thermomechanical and thermochemical treatment processes on the microstructure evolution and properties of titanium alloys [6–8]. The relationship between microstructure evolution and heat treatment process parameters can be quantitatively described by studying phase transformation kinetics.

Thermal dilatometry is an effective method for analyzing the phase transformation kinetics of solid metal. Shah et al. [9] studied the \( \alpha + \beta \rightarrow \beta \) phase transformation activation energy of Ti-6Al-4V by thermal dilatometry, and speculated that the diffusion of the stable element V from the \( \beta \) phase to the \( \alpha \) phase determines the transformation speed. In addition, Ming et al. [10], Katzarov et al. [11] and Elmer et al. [12] also described the effect of the V element on the phase transformation behavior of different types of titanium alloys. Zhou et al. [13] studied the \( \omega \) phase transformation kinetics of
TB-13 titanium alloy during continuous heating by non-isothermal dilatometry. It was found that the exponent $n$ changed continuously with the increase of the transformed volume fraction, while the nucleation and growth mechanism of the $\omega$ phase were analyzed based on the local Avrami exponent $n$. Hui et al. [14] investigated the transformation process of the metastable phase $\beta' \rightarrow \omega$ of near $\beta$ titanium alloy Ti-7333 at different heating rates by non-isothermal dilatometry. The activation energy and phase transformation kinetics models of $\beta' \rightarrow \omega$ were obtained based on the KAS theory and the modified JMA model. Regardless of the method used, the study of phase transformation kinetics has great practical significance for understanding the phase transformation mechanism, quantifying the phase transformation process and guiding technological optimization.

Nowadays, Ti-6Al-4V is the most suitable two-phase titanium alloy for application testing and industrial production. The improvement of industrial standards and the refinement of design concepts require the continuous optimization of the product performance of titanium alloy. In this paper, the non-isothermal expansion test was used to simulate the solution treatment heating process. The quantitative relationship between process parameters and the degree of transformation was obtained by conversion between the expansion amount $\Delta L$ and the transformed volume fraction $f_T$. The phase transformation kinetics model of Ti-6Al-4V during continuous heating was established based on the modified JMA model and the KAS theory. Finally, the nucleation and growth mechanism was analyzed by using the local Avrami exponent $n$. This paper aims to realize the quantitative characterization of the relationship between the phase transformation behavior and process parameters of Ti-6Al-4V alloy, and then to guide the optimization of technological parameters in the solution heating stage.

2. Materials and Methods

The tested material was the annealed Ti-6Al-4V rolled plate with a chemical composition (wt%) of 6.1 Al, 4.03 V, 0.12 Fe, 0.012 C and the remainder Ti. The optical microscope (OM) image of Ti-6Al-4V is shown in Figure 1. The original microstructure consisted of primary equiaxial $\alpha$ and intergranular $\beta$, and the average grain size of $\alpha$ was about 20 $\mu$m.

![Figure 1. Optical microscope (OM) image of Ti-6Al-4V alloy in annealed state.](image)

The sizes of the thermal expansion specimen were $\Phi$ 3 mm $\times$ 10 mm. The thermal expansion test was carried out on an L78RITA thermal expansion deformation/phase transformation tester (Linseis, Germany). The temperature was controlled precisely by the high-frequency induction heating and the gas spray cooling, and the temperature accuracy was controlled within $\pm$1 °C. The maximum heating and cooling rates were 100 °C/s. During the heating process, high-purity 99.999% argon gas was introduced as a shielding gas to prevent oxidation of the test specimen. The change of thermal expansion was measured by a special expansion measuring instrument.

The experimental scheme was as follows: Under the protection the state of inert gas, the sample was heated to 1150 °C and held for 10 min at a heating rate of 0.1, 0.3, 0.5 and 0.8 °C/s, separately, and then rapidly cooled to room temperature at a cooling rate of 80 °C/s. After the expansion test,
the samples were ground using different types of sandpaper. The polishing reagent contained SiO$_2$ particles with an average diameter of 0.05 µm, and the corrosion reagent was Kroll reagent with the proportion HF:HNO$_3$:H$_2$O = 2:4:94. The microstructures of the samples were observed with the optical microscope.

3. Theoretical Model of Non-Isothermal Transformation

The dilatation caused by phase transformation during heat treatment of polycrystalline titanium alloys is generally isotropic [13]. The titanium alloys usually exhibit a linear change in volume when the heating temperature is much lower than the phase transformation temperature. Solid-state phase transformation occurring in the internal microstructure of the material leads to the redistribution of the lattice types, lattice constants and elements, which results in the deviation of the volume variation rule from the linear relationship [15,16]. The corresponding relationship between the alloy volume change and the transformed volume fraction can be obtained by analyzing the thermal expansion curve with the lever principle [17].

Figure 2 shows the dilatometric curve of the phase transformation from $\alpha + \beta \rightarrow \beta$ at a heating rate of 0.3 °C/s. When the heating temperature is in the AD range shown in Figure 2, the transformed volume fraction $f_T$ can be expressed as the following equation according to the law of leverage:

$$f_T = \frac{EO}{EF}$$  \hspace{1cm} (1)

![Figure 2. Dilatometric curve during heating process at a heating rate of 0.3 °C/s.](image)

The $\alpha + \beta \rightarrow \beta$ phase transformation of Ti-6Al-4V during continuous heating is a process of nucleation and growth, which can be characterized by the modified JMA model [18]:

$$f_T = 1 - \exp\left\{-\frac{[K(T - T_0)/H]^n}{RT}\right\}$$  \hspace{1cm} (2)

where, $f_T$ is the transformed volume fraction at the heating temperature $T$, $n$ is the Avrami exponent, $T_0$ is the initial transformation temperature of the $\alpha + \beta \rightarrow \beta$ transformation, $H$ is the heating rate and $K$ is a temperature-related parameter. $K$ can be represented as:

$$K(T) = K_0 \exp\left(-\frac{Q}{RT}\right)$$  \hspace{1cm} (3)

where, $K_0$ is a constant and $Q$ is the activation energy of the $\alpha + \beta \rightarrow \beta$ transformation.

The activation energy of the transformation is an important thermodynamic parameter for evaluating the difficulty of the $\alpha + \beta \rightarrow \beta$ transformation. The activation energy $Q$ for a single phase transformation process is usually a constant [19]. The transformation activation energy $Q$ of Ti-6Al-4V during continuous heating can be obtained from the expansion results at different heating rates. From
the point of view of calculation accuracy and simplification, the Kissinger-Akahira-Sunose (KAS) model is the best method for determining the activation energy of linear heating experiments [20,21]. The KAS model is as follows:

\[
\ln \left[ \frac{H}{T^2} \right] = - \frac{Q}{RT} + \ln \left( \frac{RQ_0}{K} \right)
\]

(4)

where, \( T_\beta \) is the temperature corresponding to the fastest transformation rate at a certain heating rate. \( K_0 \) is the pre-exponential factor of the Johnson-Mehl-Avrami (JMA) equation, which is a constant. \( R \) is the molar gas constant (\( R = 8.314 \text{ J} \cdot \text{mol}^{-1} \cdot \text{K}^{-1} \)).

The phase transformation activation energy \( Q \) can be obtained according to Equation (4), and the transformed volume fraction \( f_T \) can be obtained from Equations (2) and (3). The Avrami exponent \( n \) of the \( \alpha + \beta \rightarrow \beta \) transformation can be obtained by differentiation with the above transformed volume fraction \( f_T \):

\[
n = - \frac{\partial \ln[-\ln(1 - f_T)]}{\partial \left( \frac{Q}{RT} \right)}
\]

(5)

4. Results and Analysis

4.1. Microstructure Analysis

Figure 3 presents metallographic microstructures of samples at two heating rates of 0.3 and 0.8 °C/s after cooling to room temperature at a cooling rate of 80 °C/s. As can be seen from Figure 3, the microstructures under different heating rates are similar, and the original \( \beta \) grains are clearly visible. The larger size of the \( \beta \) grains at the heating rate of 0.3 °C/s can be attributed to the longer heating time, and the \( \beta \) grains themselves are highly susceptible to growth and coarsening (Figure 3a,b). The cooled microstructures consist of orthogonally distributed fine-needle martensite and white massive \( \alpha_m \) mainly distributed along the original \( \beta \) grain boundary. The fine-needle martensite preferentially nucleates at original \( \beta \) grain boundary and then grows toward the interior of the grain (Figure 3c,d). The formation of the white massive \( \alpha_m \) is related to the cooling rate. \( \alpha_m \) has the same crystal structure and element content as \( \alpha' \) martensite [22].

![Figure 3](image-url)
transformation. The starting and finishing temperatures of the transformation at different heating rates are different, as shown in Table 1.

When phase transformation occurs during continuous heating, the continued accumulation of the solid solution element and lattice distortion results in a non-linear expansion of the sample. The initial heating time is taken as the starting time, at which time the expansion amount of the sample is regarded as zero. The change in the expansion amount is indicated by \( \Delta L \).

Figure 4 shows the thermal dilatometric curves of the Ti-6Al-4V at different heating rates. It is shown that: (1) The variation of the thermal expansion curve at different heating rates is similar. Before and after transformation, the expansion amount \( \Delta L \) is proportional to the temperature. However, there is a non-linear relationship between the expansion amount and the temperature during phase transformation. (2) The starting and finishing temperatures of the transformation at different heating rates are different, as shown in Table 1.

**Figure 3.** OM images of two kinds of samples after the rapid cooling at 80 °C/s: (a,c) correspond to the sample processed at the heating rate of 0.3 °C/s; (b,d) correspond to the sample processed at the heating rate of 0.8 °C/s.

4.2. Phase Transformation Kinetics during the Heating Process

When phase transformation occurs during continuous heating, the continued accumulation of the solid solution element and lattice distortion results in a non-linear expansion of the sample. The initial heating time is taken as the starting time, at which time the expansion amount of the sample is regarded as zero. The change in the expansion amount is indicated by \( \Delta L \).

Figure 4 shows the thermal dilatometric curves of the Ti-6Al-4V at different heating rates. It is shown that: (1) The variation of the thermal expansion curve at different heating rates is similar. Before and after transformation, the expansion amount \( \Delta L \) is proportional to the temperature. However, there is a non-linear relationship between the expansion amount and the temperature during phase transformation. (2) The starting and finishing temperatures of the transformation at different heating rates are different, as shown in Table 1.
Therefore, the phase transformation mode of Ti-6Al-4V alloy during continuous heating is a process controlled by nucleation growth diffusion [23]. The "S" curve gradually moves to a high temperature region with the increase of the heating rate, indicating that the transformation starting time and the phase transformation degree are affected by heating temperatures and heating rates.

The relationship between the transformed fraction and heating temperatures at different heating rates can be obtained by Equation (1), as shown in Figure 5. It can be seen from Figure 5 that the phase transformation is completed within a temperature range during the continuous heating process. The relationship between the transformed volume fraction and temperature at different heating rates presents a typical “S” pattern. Therefore, the phase transformation mode of Ti-6Al-4V alloy during continuous heating is a process controlled by nucleation growth diffusion [23]. The “S” curve gradually moves to a high temperature zone with the increase of heating rates, indicating that the transformation starting time and the phase transformation degree are affected by heating temperatures and heating rates.

![Dilatometric curves of Ti-6Al-4V at different heating rates](image)

**Figure 4.** Dilatometric curves of Ti-6Al-4V at different heating rates of (a) 0.1 °C/s, (b) 0.3 °C/s, (c) 0.5 °C/s and (d) 0.8 °C/s.

| Heating Rate (°C/s) | 0.1 | 0.3 | 0.5 | 0.8 |
|---------------------|-----|-----|-----|-----|
| Starting temperature (°C) | 873 | 885 | 893 | 902 |
| Finishing temperature (°C) | 1072 | 1086 | 1095 | 1103 |

**Table 1.** The phase transformation temperatures with a non-linear change in the expansion amount at different heating rates.

![The relationship between the transformed fraction and heating temperatures](image)

**Figure 5.** The relationship between the transformed fraction and heating temperatures at different heating rates.

It is can be known from Table 1 and Figure 5 that the phase transformation temperature range of Ti-6Al-4V gradually moves to a high temperature region with the increase of the heating rate, and the transformation temperature range is about 200 °C. The transformation mode of $\alpha + \beta \rightarrow \beta$ during the continuous heating process belongs to the elemental diffusion type that controls the nucleation and...
growth of β grains. When the heating rate is low, it takes a long time to heat from room temperature to 1150 °C, so that the element has sufficient time for diffusion. Therefore, the phase transformation begins at 873 °C when the heating rate is 0.1 °C/s. During the heat treatment, the phase change may occur only when the local element concentration reaches a critical value. The higher heating rate results in a reduction in the time taken for the heating phase such that the elemental concentration does not reach the critical concentration required for the transformation, causing the starting and finishing transformation temperatures to move to the high temperature zone.

The temperature $T_β$, corresponding to the maximum transformation rate at different heating rates, can be obtained by differentiating the thermal expansion curve, as shown in Table 2. The data in Table 2 are used with Equation (4) to obtain the scatter diagram of $\ln(H/Tβ^2)-(1/T_β)$, and its linear fitting results are shown in Figure 6. The fitting result reliability is 97.44%. The slope–$Q/R$ of the fitted line is $-53,583$, and the intercept $\ln(RK_0/Q)$ is 28.32. Thus, the total activation energy $Q$ of the $α + β → β$ transformation during the continuous heating process is 445.5 kJ·mol$^{-1}$, and the pre-exponential factor $K_0$ of the Johnson-Mehl-Avrami (JMA) equation is $1.067 \times 10^{17}$. The above results are in the same order of magnitude as the phase transformation activation energy of Ti-6Al-4V alloy measured by the electric resistance method during the continuous heating process [9].

Table 2. The temperature corresponding to the maximum transformation rate at different heating rates.

| Heating Rate (°C/s) | 0.1 | 0.3 | 0.5 | 0.8 |
|---------------------|-----|-----|-----|-----|
| $T_β$ (°C)          | 925 | 951 | 970 | 975 |

Figure 6. $\ln(H/Tβ^2)-(1/T_β)$ linear fitting curve of Ti-6Al-4V during the heating process.

4.3. Transformation Mechanism during the Heating Process

The $\ln[−\ln(1−f_T)]$ and $−Q/(RT)$ curves at different heating rates can be drawn according to Equation (5). The relationship between the Avrami exponent $n$ and the transformed volume fraction $f_T$ at different heating rates can be obtained by differentiating the curves, as shown in Figure 7. During the allotropic transformation of titanium alloy, the nucleation and growth mechanism of the new phase in the process of phase transformation can be analyzed by the numerical variation of the Avrami exponent $n$ [24]. It is shown in Figure 7 that the Avrami exponent $n$ has a consistent trend with the change of the transformed volume fraction at different heating rates.
The phase transformation kinetics curve of Ti-6Al-4V during the continuous heating process is studied by non-isothermal dilatometry. The main conclusions obtained are as follows:

5. Conclusions

According to the JMA theory, the curve of the Avrami exponent $n$ of Ti-6Al-4V is roughly divided into three stages. Taking the $n$-value curve at a heating rate of $0.3 \, ^\circ\text{C}/\text{s}$ as an example, it can be known that: (1) When $0 < f < 0.01$, the value of $n$ is greater than 4, indicating that the initial stage of phase transformation is the three-dimensional nucleation dominated by elemental diffusion, and the nucleation rate increases with time. The $\beta$ grains nucleate at the residual $\beta$ phase grain boundaries, dislocations, etc., and grow by consuming the surrounding $\alpha$ phase [25]. (2) When $0.01 < f < 0.02$, $n$ decreases from 4 to 2, indicating that a two-dimensional or three-dimensional nucleation rate at this stage gradually decreases to zero. The decrease in the nucleation rate is mainly due to the nucleation and growth of the $\beta$ phase leading to a continuous decrease in the content of the $\beta$ stabilizing elements in the remaining phase. The elemental concentration is lower than the critical concentration required for nucleation, resulting in a continuous decline in the nucleation rate [12]. (3) When $0.02 < f < 0.96$, $n$ decreases slowly to the minimum value of 0.35 with the increase of the transformed volume fraction $f$, indicating that the nucleation has reached saturation. At this stage, $\beta$ grains grow rapidly, driven by the reduction of interface-free energy and the increase in the diffusion coefficient [26]. When the transformed fraction $f > 0.96$, $n$ rapidly increases to about 1.5. This phenomenon also occurs in previous references [13,18]. The relationship between the transformed volume fraction $f$ and the temperature $T$ is not linearly synchronous. For the whole transformation process, it is not suitable to simply rely on mathematical methods. Therefore, when the transformed volume fraction is below 95%, the JMA equation is suitable for application.

5. Conclusions

The $\alpha + \beta \rightarrow \beta$ phase transformation kinetics of Ti-6Al-4V during the continuous heating process are studied by non-isothermal dilatometry. The main conclusions obtained are as follows:

(1) The thermal dilatometry test is an effective method to study the phase transformation process. The expansion curve and its first derivative curve can directly reflect the starting, developing and finishing of phase transformation.

(2) The phase transformation kinetics curve of Ti-6Al-4V during the continuous heating process presents a typical “S” pattern, indicating that the phase transformation mechanism is the nucleation and growth of grains controlled by elemental diffusion. The activation energy $Q$ of the diffusion transformation during the continuous heating process is 445.5 kJ mol$^{-1}$.

(3) The phase transformation process of Ti-6Al-4V in the continuous heating stage is mainly divided into three stages: (1) Three-dimensional nucleation and growth, and the rate of nucleation increases. (2) Consumption of $\beta$ stable element leads to a decrease in the nucleation rate. (3) Nucleation saturation and $\beta$ grains growth.
Author Contributions: H.Y. wrote the paper; W.L. translated the paper; H.Z. analyzed the experiment data; S.L. performed the experiments and validated the results of experiments; T.Z. conceived and designed the experiments; L.L. coordinated test arrangements.

Acknowledgments: This work was supported by the Natural Science Foundation of Hebei Province of China No. E2016203217 and the National Natural Science Foundation of China No. 51205342.

Conflicts of Interest: The authors declare no conflict of interest.

References
1. Nabil, K.; Mabrouk, B.; Riad, B. Beta to alpha transformation kinetics and microstructure of Ti-6Al-4V alloy during continuous cooling. *Mater. Chem. Phys.* 2016, 181, 462–469.
2. Tarzimoghadam, Z.; Sandlöbes, S.; Pradeep, K.G. Microstructure design and mechanical properties in a near-α Ti-4Mo alloy. *Acta Mater.* 2015, 97, 291–304. [CrossRef]
3. Terlinde, G.T.; Duerig, T.W.; Williams, J.C. Microstructure, tensile deformation, and fracture in aged Ti-10V-2Fe-3Al. *Metall. Trans. A* 1983, 14, 2101–2115. [CrossRef]
4. Li, C.L.; Zou, L.N.; Fu, Y.Y. Effect of heat treatments on microstructure and property of a high strength/toughness Ti-8V-1.5Mo-2Fe-3Al Alloy. *Mater. Sci. Eng. A* 2014, 616, 207–213. [CrossRef]
5. Li, C.; Chen, J.; Li, W. Study on the relationship between microstructure and mechanical property in a metastable β titanium alloy. *J. Alloys Compd.* 2015, 627, 222–230. [CrossRef]
6. Yu, W.X.; Li, M.Q.; Jiao, L. Effect of deformation parameters on the precipitation mechanism of secondary phase under high temperature isothermal compression of Ti–6Al–4V alloy. *Mater. Sci. Eng. A* 2010, 527, 4201–4217. [CrossRef]
7. Jie, F.; Hua, D.; Yi, H. Influence of phase volume fraction on the grain refining of a Ti-6Al-4V alloy by high-pressure torsion. *J. Mater. Res. Technol.* 2014, 4, 1–6.
8. Peng, X.N.; Guo, H.Z.; Wang, T. Effects of β treatments on microstructures and mechanical properties of Ti-6Al-4V-DT titanium alloy. *Mater. Sci. Eng. A* 2012, 533, 55–63. [CrossRef]
9. Shah, A.K.; Kulkarni, G.J.; Gopinathan, V. Determination of activation energy for α + β→β transformation in Ti-6Al-4V alloy by dilatometry. *Scr. Metall. Mater.* 1995, 32, 1353–1356. [CrossRef]
10. Ming, L.C.; Manghmani, M.H.; Katahara, K.W. Phase Transformations in the Ti-V system under high pressure up to 25 GPa. *Acta Metall.* 1981, 29, 479–485. [CrossRef]
11. Katzarov, I.; Malinov, S.; Sha, W. Finite element modeling of the morphology of β to α phase transformation in Ti-6Al-4V alloy. *Metall. Mater. Trans. A* 2002, 33, 1027–1040. [CrossRef]
12. Elmer, J.W.; Palmer, T.A.; Babu, S.S. In situ observations of lattice expansion and transformation rates of α and β phases in Ti-6Al-4V. *Mater. Sci. Eng. A* 2005, 391, 104–113. [CrossRef]
13. Zhou, Z.B.; Lai, M.J.; Bin, T. Non-isothermal phase transformation kinetics of phase in TB-13 titanium alloys. *Mater. Sci. Eng. A* 2010, 527, 5400–5104. [CrossRef]
14. Hui, Q.; Xue, X.; Kou, H. Kinetics of the ω phase transformation of Ti-7333 titanium alloy during continuous heating. *J. Mater. Sci.* 2013, 48, 1966–1972. [CrossRef]
15. Elmer, J.W.; Palmer, T.A.; Babu, S.S. Low temperature relaxation of residual stress in Ti-6Al-4V. *Scr. Mater.* 2005, 52, 1051–1056. [CrossRef]
16. Feng, S.; Li, J.; Kou, H. β phase transformation kinetics in Ti60 alloy during continuous cooling. *J. Alloys Compd.* 2013, 576, 108–113.
17. Charpentier, M.; Hazotte, A.; Daloz, D. Lamellar transformation in near-γ TiAl alloys—Quantitative analysis of kinetics and microstructure. *Mater. Sci. Eng. A* 2008, 491, 321–330. [CrossRef]
18. Blazquez, J.S.; Conde, C.F.; Conde, A. Non-isothermal approach to isokinetic crystallization processes: Application to the nanocrystallization of HITPERM alloys. *Acta Mater.* 2005, 53, 2305–2311. [CrossRef]
19. Wang, Y.; Kou, H.; Chang, H. Phase transformation in TC21 alloy during continuous heating. *J. Alloys Compd.* 2009, 472, 252–256. [CrossRef]
20. Budrugeac, P.; MusAt, V.; Segal, E. Non-isothermal kinetic study on the decomposition of Zn acetate-based sol-gel precursor. *J. Therm. Anal. Calorim.* 2007, 88, 699–702. [CrossRef]
21. Starink, M.J. The determination of activation energy from linear heating rate experiments: A comparison of the accuracy of isoconversion methods. *Thermochim. Acta* 2003, 404, 163–176. [CrossRef]
22. Ahmed, T.; Rack, H.J. Phase transformations during cooling in $\alpha+\beta$ titanium alloys. *Mater. Sci. Eng. A* 1998, 243, 206–211. [CrossRef]

23. Christian, J.W. The theory of transformations in metals and alloys. *Mater. Today* 2003, 6, 53–53. [CrossRef]

24. Jiang, X.D.; Zhang, H.W.; Wen, Q.Y. Crystallization kinetics of CoNbZr amorphous alloys thin films. *Mater. Chem. Phys.* 2004, 88, 197–201. [CrossRef]

25. Gil, F.J.; Planell, J.A. Growth order and activation energies for grain growth of Ti6Al4V alloy in $\beta$ phase. *Scr. Metall. Mater.* 1991, 25, 2843–2848. [CrossRef]

26. Gil, F.J.; Planell, J.A. Behaviour of normal grain growth kinetics in single phase titanium and titanium alloys. *Mater. Sci. Eng. A* 2000, 283, 17–24. [CrossRef]