Ti–6Al–4V: Deformation map and modelisation of tensile behaviour

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Abstract

Tensile tests have been performed on a superplastic titanium alloy. Different strain rates from $5 \times 10^{-4}$ to $5 \times 10^{-2}$ s$^{-1}$ and various temperatures up to 1050 °C were explored. The results of these tests, supplemented with microstructural investigations allow, as a function of temperature and strain rate, the determination of mechanical properties such as flow stress, strain hardening and fracture strain as well as the identification of mechanisms such as grain boundary sliding, dynamic recrystallization and dynamic grain growth. A modified Norton–Hoff law is proposed to characterize the tensile behaviour of this alloy. All temperature–strain rate combinations have been grouped into a number of domains, each corresponding to a particular deformation mechanism. For each domain the Norton–Hoff parameters have been determined. Close agreement between the predicted and experimental stress–strain relationships has been achieved.

Keywords: Non-ferrous metals and alloys (A); Mechanical (E); Plastic behaviour (F)

1. Introduction

The use of titanium alloys in industrial applications increases from day to day especially in the aeronautical sector. This can be explained by their remarkable properties such as strength, toughness, fatigue behaviour, corrosion resistance, and biocompatibility. The aim of the present contribution is to study the deformation mechanisms of an aeronautical quality Ti–6Al–4V as a function of the strain rate and of the temperature [1]. Under particular conditions, this alloy can be superplastic. Superplastic materials are polycrystalline solids which have the ability to undergo large strains prior to failure. As a general rule, the deformation temperature has to be superior to half of the melting temperature (in K), while the strain rate must be situated in the range of $10^{-4}$ to $10^{-2}$ s$^{-1}$ [2]. The alloy studied here is known to be superplastic approximately between 750 °C and 950 °C, and at strain rates between $10^{-4}$ and $5 \times 10^{-3}$ s$^{-1}$ [3]. In the present study the tensile properties, the microstructural evolution and the crystallographic texture change caused by deformation between room temperature (RT) and 1050 °C have been investigated. The strain rate was chosen between $5 \times 10^{-2}$ s$^{-1}$ and $5 \times 10^{-4}$ s$^{-1}$. Particular attention was paid to the evolution of grain size as a function of temperature and deformation. The results of the tensile tests together with microstructural observations were exploited in order to define a deformation map. Finally the Norton–Hoff equation for viscoplastic flow was chosen and adapted in order to have a single expression allowing for the fitting of all experimental results.

2. Material and method

The material used for the current study is a commercial titanium alloy, TIMETAL 6-4 Titanium Aero. The as received state was annealed after hot and cold rolling into sheet of 1 mm thickness, resulting in an initial grain size of 8 μm. Tensile tests were performed on a Zwick T-050 tensile machine equipped with a furnace with three independent heating zones. This furnace has a constant temperature zone of 30 cm and offers the possibility to control the atmosphere. Flat samples with a gauge length of 10 mm long and 6 mm width were cut by...
water jet streaming. The tensile machine was programmed to operate at a constant true strain rate. The experimental setup and condition did not allow the use of an extensometer, so sample deformation was determined from the mobile traverse displacement. The oxidation was avoided by the introduction of a nitrogen gas flow that prevent the formation of a case at the specimen surface. After each test, samples were quenched to room temperature by the injection of nitrogen gas inside the furnace. Using this method, the cooling rate between 1000 °C and 700 °C was approximately 30 °C/s. Samples were prepared for automated electron backscattered diffraction (EBSD) measurements as proposed by Katrakova et al. [4]. These measurements were carried out in a Philips XL30 microscope and analyzed with the TSL (EDAX) software (OIM data collection 4 and OIM analysis 4.5). Grain size measurements were performed according to the mean linear intercept method as proposed in ASTM E-112.

3. Results and discussion

Considering all results from the experiments mentioned above, it is possible to classify the deformation behaviour of this Ti–Al–V alloy into essentially four main domains. A first domain exists from room temperature up to about 650 °C. In this range (irrespective of the strain rate), no dynamic recrystallization nor grain boundary sliding could be detected. This is the domain of classical cold and warm deformation. A second domain appears at temperatures between 725 °C and 950 °C and strain rates lower than \(5 \times 10^{-3} \text{s}^{-1}\). In this domain the dominant deformation mechanism is grain boundary sliding. The material shows enhanced deformability (superplasticity). A third domain is situated at temperatures above 950 °C and low strain rates (below \(5 \times 10^{-3} \text{s}^{-1}\)). Dynamic grain growth is causing the disappearance of superplasticity; Ashby [5] proposed that dislocation creep is responsible for the deformation.

Finally, a fourth domain is situated above 750 °C and at high strain rate. In this domain, \(\sigma-\varepsilon\) curves are of the steady-state type and classical hot deformation mechanisms combined with dynamic recrystallization are believed to take place. An overview of these domains is presented in Fig. 1.

3.1. RT – 650 °C at all strain rates

At lower temperature, work hardening is important and the fracture strain is relatively limited (Fig. 2). With increasing temperature dynamic recovery becomes more important, resulting in softening. The ductility rises with increasing temperature. In this domain the elongation of individual grains is parallel to the macroscopic (tensile) deformation (Fig. 3). Dynamic recovery becomes more prominent with increasing temperature, but no dynamic recrystallization occurs. The influence of the strain rate on the tensile strength is weak, while softening increases when the strain rate decreases. An overview of the true stress–true strain curves as a function of temperature and strain rate are shown in Fig. 4.

![Fig. 1. Deformation map for temperatures between RT and 1000 °C and for strain rates between \(5 \times 10^{-3} \text{s}^{-1}\) and \(5 \times 10^{-2} \text{s}^{-1}\) (DGG: dynamic grain growth; GBS: grain boundary sliding; DRX: dynamic recrystallization).](image1)

![Fig. 2. True stress–true plastic strain curves of a Ti–6Al–4V flat specimens tested between RT and 650 °C and a strain rate of \(5 \times 10^{-4} \text{s}^{-1}\).](image2)

![Fig. 3. Microstructure of a Ti–6Al–4V sample deformed at 500 °C at a strain rate of \(5 \times 10^{-4} \text{s}^{-1}\) until 30% engineering strain. The tensile direction is horizontal. Grains are stretched in the tensile direction.](image3)
3.2. 725–950 °C and strain rates between $5 \times 10^{-4}$ s$^{-1}$ and $5 \times 10^{-3}$ s$^{-1}$

The evolution of the $\sigma$–$\varepsilon$ curves as a function of temperature is presented in Fig. 5 for a strain rate of $5 \times 10^{-4}$ s$^{-1}$. Starting from 725 °C, grain boundary sliding begins to be dominant and the material behaves superplastically (true plastic strain higher than 1.3). The stress increase, visible in the tail of the $\sigma$–$\varepsilon$ curves, is not due to work hardening but due to dynamic grain growth. The integrity of the material during superplastic deformation is assured by the intervention of grain boundary accommodation mechanisms; these mechanisms are known to be intragranular diffusion, grain boundary diffusion and dislocation climb [6]. When the grain size increases, the effectiveness of those mechanisms will be more difficult because the volume to be accommodated becomes larger, thus resulting in an increasing flow stress as can be seen on Fig. 5. The maximum elongation is observed at temperatures between 800 °C and 850 °C and a strain rate of $5 \times 10^{-4}$ s$^{-1}$. Looking at the details of the evolution of the grain size for these particular test conditions (see Fig. 6), different particularities can be observed. At the early stages of the deformation the sample first undergoes a decrease in grain size due to dynamic recrystallization (quite similar with the one

Fig. 4. Experimental $\sigma$–$\varepsilon$ charts for tensile test carried out at 300, 400, 500, 600 °C and at strain rates of $5 \times 10^{-2}$ s$^{-1}$, $5 \times 10^{-3}$ s$^{-1}$, $5 \times 10^{-4}$ s$^{-1}$ (indicated by $\Delta$). Results calculated with the Norton–Hoff law are shown with $\bigcirc$ and cover nearly completely the experimental curves.

Fig. 5. True stress–true plastic strain curves of Ti–6Al–4V flat specimens tested between 750 °C and 950 °C and a strain rate of $5 \times 10^{-4}$ s$^{-1}$.

Fig. 6. Grain size evolution as a function of strain for samples tested at 700, 800, 925 °C and at a strain rate of $5 \times 10^{-4}$ s$^{-1}$.
observed in commercially pure titanium by Zhu et al. [7]). But due to the high temperature, kinetics of grain growth are rather fast and grain growth becomes preponderant. The influence of strain on grain growth can be illustrated by thermal treatments. At 800 °C a static anneal (without previous deformation) showed only a very small increase in the grain size (less than 10%). This led to the conclusion that the stored deformation energy triggers the recrystallization and accelerates grain growth during the tensile test.

The suppression of grain boundary sliding with raising temperature is caused by the grain size evolution. When the temperature rises, dynamic grain growth becomes more intense. At 925 °C the increase of the grain size is so important that the dynamic recrystallization is only observable during the first 20% of strain (see Fig. 6).

Since the prevalent mechanism active in this domain is grain boundary sliding, the influence of strain rate is important. The shape of the domain in a temperature vs. strain rate graph is Gaussian. At the higher and the lower temperature the superplasticity is only present at the lowest strain rate. A representation of this domain is presented in Fig. 1.

3.3. 950–1050 °C and strain rates between 5 × 10⁻⁴ s⁻¹ and 5 × 10⁻³ s⁻¹

Above 950 °C dynamic grain growth creates grains too large to allow grain boundary sliding. The evolution of the β-phase fraction in this domain plays also an essential role. At 950 °C already 65% of the volume consists of β-phase and at 995 °C the α-β transformation is complete. The large grain size and the high β-phase content, prevent the occurrence of superplasticity. Due to the limited industrial importance of this domain, the modelisation of the σ-ε curves was not attempted.

3.4. 750–900 °C for strain rates above 5 × 10⁻³ s⁻¹

For this domain, the mechanisms of deformation have not been analyzed in detail but the fitting with the Norton–Hoff law will be attempted, based on the results of the tensile tests.

4. Modelisation

In each of the aforementioned temperature–strain rate domains, the tensile stress strain behaviour of the Ti–6Al–4V alloy was modeled using a modified Norton–Hoff law [8]. The Norton–Hoff law is an elasto-viscoplastic law with four parameters and is written in one dimension as presented in Eq. (1), where σ is the true stress, ε the true strain, the strain rate, p₁, p₂, p₃ and p₄ are four materials dependant parameters

\[ \bar{\sigma} = \bar{\varepsilon}^{p_2} \exp(-p_1 \varepsilon) p_5 \] \[ \theta^3 \sqrt{\bar{\varepsilon}}^{p_3}. \]  

In the superplastic domain, the grain size is an important additional parameter. The original Norton–Hoff law was adapted to take this into account. The adapted version of the Norton–Hoff equation used in the present study, is shown in Eq. (2), where σ₁ is the yield stress, d is the grain size and p₅ the grain size exponent

\[ \bar{\sigma} = \sigma_1 + \bar{\varepsilon}^{p_2} \exp(-p_1 \varepsilon) p_2 \bar{\varepsilon}^{p_3} \theta^{p_5} \] \[ d. \]  

The most popular law describing superplasticity is Eq. (3) where n is the hardening exponent, m the strain rate sensitivity coefficient and p the grain size exponent

\[ \sigma = k \bar{\varepsilon}^{m} d^{p} \] \[ d. \]

Comparing Eqs. (2) and (3), some correspondence is evident:

- p₂ corresponds to k and is a scaling factor;
- p₃ corresponds to m and is the strain rate sensitivity parameter;
- p₄ corresponds to n and is the hardening parameter;
- p₅ corresponds to p and is the grain size parameter;
- p₁ has no homologue in the classical law but considering Eq. (2), it will deal with softening.

Starting from the experimental curves, the following operations were necessary in order to calibrate the model. First, the parameters of the Norton–Hoff law were determined in each of the domains described above, by fitting of Eq. (2) to the experimental curves and by minimizing the error by the least squares method. Secondly, the evolution of the parameters p₁–p₅ as a function of the strain rate and of the temperature was studied and some expressions relating these evolutions were determined (see for example Eqs. (5)–(9)). These expressions were then introduced in the main expression of the Norton–Hoff law, giving a new equation with a new series of parameters only function of the material. Once more the new parameters were determined using the minimization of the least squares method.

4.1. RT – 650 °C, all strain rates

Using the method described above, the expression presented in Eq. (4) was established, where A₁ . . . A₁₀ are material constants

\[ \sigma = \sigma_1 + \bar{\varepsilon}^{(A_1 \theta^2 + A_2 \theta + A_3 + A_4 \bar{\varepsilon}) A_5 \bar{\varepsilon}^{A_6} \theta^{A_7 + A_8 + A_9 \ln(\bar{\varepsilon})}}. \]  

By a fitting procedure the values of the different A-parameters were determined and the results are shown in Table 1. Comparing Eq. (4) with the modified Norton–Hoff law (2), the different p parameters can be expressed as a function of the temperature and the strain rate as shown:

\[ p_1 = A_1 \theta^2 + A_2 \theta + A_3 + A_4 \bar{\varepsilon} \] \[ p_2 = A_5 \bar{\varepsilon} + A_6 \] \[ p_3 = A_7 \theta \] \[ p_4 = T^{A_8} + A_9 + A_{10} \ln(\bar{\varepsilon}) \] \[ d. \]
developed by Kocks and Mecking [9] for FCC metals and qualitatively transposable to HCP and BCC metals because basically the same mechanisms act in cubic and in hexagonal metals.

4.1.5. p5 (Grain size)

Below 600 °C the grain size is stable; therefore p5 is equals 0 in this domain.

The fitting is qualitatively and quantitatively good and the evolution of the different parameters is consistent with the physical meaning of these parameters. The established model can be extrapolated reasonably well down to RT but the results are better for the highest strain rate. At $5 \times 10^{-3} \text{s}^{-1}$ and just before necking, the difference between the experimental and the calculated curves is approximately 12%.

The evolution of the shape of the $\sigma$-e curves in Fig. 4 implies that softening increases with rising temperature and at the same time the hardening gradually decreases. The softening term also increases when the strain rate decreases and it’s the opposite for the hardening term. The softening is caused by dynamic recovery which is the basic mechanism that leads to annihilation of pairs of dislocations during straining. When temperature rises and strain rate decreases, the dislocation density decreases because of easier cross slip, climb, and unpinning at nodes of the dislocations [10].

4.2. 725–950 °C and strain rates between $5 \times 10^{-4} \text{s}^{-1}$ and $5 \times 10^{-3} \text{s}^{-1}$

In this domain, the grain size is an important parameter. In a first approach, fitting of the experimental curves was attempted without the contribution of the grain size but the results were unacceptable. It was thus necessary to collect data relating to the grain size evolution as function of the strain, the strain rate and the temperature in this domain. Based on a number of experimental grain size measurements it was possible to extrapolate an expression describing this evolution. Eq. (9) represents a good approximation of the grain size for true strains below 2. By the
same method as described in previous paragraph and using (9) to calculate \( d \), the expression given by Eq. (10) was identified as the best compromise to model the evolution of the true stress–true strain curves as a function of strain rate and temperature, where \( B_1 \ldots B_{14} \) material constants

\[
d = \left( -6.54 \times 10^{-3} T - 2190.67 \varepsilon + 2.23 \varepsilon T + 8.36 \varepsilon^2 \right) \varepsilon^2 \\
+ \left( 0.047 T - 870.68 \varepsilon - 1.43 T - 43.72 \varepsilon + 8.1 \right) \varepsilon + 8.1 \\
\sigma = \sigma_y + (B_1 T^2 + B_2 T + B_3 + B_4 \varepsilon) e^{(B_5 T^2 + B_6 T + B_7 + B_8 \varepsilon)} \\
\times e^{(B_9 + B_{10} \varepsilon)} d e^{(B_1 T^2 + B_2 T + B_3 + B_4 \varepsilon)}.
\]

After fitting, values for the different \( B \) coefficients were obtained and are presented in the Table 1. Fig. 9 exposes the comparison between experimental and calculated stress–strain curves. The result is good as well qualitatively as quantitatively. The different \( p \) parameters can be calculated following

\[
P_1 = 0 \\
P_2 = (B_1 T^2 + B_2 T + B_3 + B_4 \varepsilon) \\
P_3 = (B_1 T^2 + B_6 T + B_7 + B_8 \varepsilon) \\
P_4 = (B_9 + B_{10} \varepsilon) \\
P_5 = (B_1 T^2 + B_12 T + B_{13} + B_{14} \varepsilon).
\]

4.2.1. \( p_1 \) (Softening)

In this domain, GBS is active and \( p_1 \) equals 0. During superplastic deformation, GBS goes together with accommodation mechanisms, mainly dislocation glide or diffusion. If the activity of dislocation involve the generation of dislocations and hence recovery, \( p_1 \) should be different from 0. Obviously the fitting seems to suggest grain boundary diffusion as the principal accommodating mechanism. Similar findings were reported in Ref. [11].

Fig. 9. Experimental \( \sigma-\varepsilon \) charts (grey) and fitting (black) with the Norton–Hoff law for tensile tests at different temperatures between 750 °C and 925 °C and different strain rates between \( 10^{-3} \) s\(^{-1} \) and \( 5 \times 10^{-3} \) s\(^{-1} \).

Fig. 10. Evolution of the parameter expressing the strain rate sensitivity in function of the temperature and the strain rate.
4.2.2. $p_2$ (Scaling factor)

The global flow stress decreases as a function of increasing $T$ and the influence of the strain rate is limited. The calculated $p_2$ reflects this evolution indeed.

![Graph showing the evolution of $p_2$ as a function of temperature and strain rate.](image)

Fig. 11. Evolution of the parameter expressing the grain size sensitivity in function of the temperature and of the strain rate.

4.2.3. $p_3$ (m)

The evolution of $p_3$ as a function of $\dot{\varepsilon}$ and $T$ is presented in Fig. 10. This evolution is in agreement with the theory of superplasticity. The value of $m$ is larger than 0.3, reaches its maximum where the maximum elongation is observed and decreases with increasing strain rate.

4.2.4. $p_4$ (Hardening)

$p_4$ is independent of the temperature, and slightly rises with increasing strain rate. The superplastic character of the deformation is at the origin of this evolution. The grains glide along each other without any significant hardening in the grain volume. Grain boundary sliding does not create real work hardening. Normally $p_4$ should be equal to zero but it was mathematically impossible to keep the grain hardening only in the $p_5$ term. In fact $p_4$ forces the flow stress to rise almost linearly in function of the strain since $p_4$ is fitted almost equal to 1 (in comparison with the power

![Graph showing the evolution of $\sigma-\varepsilon$ plots with Norton-Hoff law fitting for tensile tests at temperatures between 750°C and 900°C and strain rates of 10^-2 s^-1 and 5 \times 10^-2 s^-1.](image)

Fig. 12. Experimental $\sigma-\varepsilon$ charts (grey) and fitting (black) with the Norton-Hoff law for tensile tests at temperatures between 750°C and 900°C and at strain rates of 10^-2 s^-1 and 5 \times 10^-2 s^-1.
law, \( n = 1 \) means a totally elastic material) and \( p_5 \) will be responsible for the curved shape of the hardening observable on the \( \sigma-e \) curves.

### 4.2.5. \( p_5 \) (Grain size)

\( p_5 \), as seen in Fig. 11, undergoes almost the same evolution as \( p_3 \). The influence of the grain size is the highest when the superplastic character of the deformation is maximum.

### 4.3. 750–900 °C and strain rates above \( 5 \times 10^{-3} \) s\(^{-1}\)

In this domain the \( \sigma-e \) curves were of the steady-state type. Such curves indicate that the mechanisms of softening are sufficiently fast to balance the rate of work hardening and are suggestive of mechanisms like dynamic recrystallization or dynamic recovery occurring at very high rate \[12\]. The strain rate is too high to allow for the occurrence of accommodation mechanisms necessary for superplastic deformation.

Because of the lack of grain size data, it was decided to attempt the Norton–Hoff fitting without taking the grain size into account. Using the same procedure as before, the Eq. (16) was determined, where \( C_1 \ldots C_{14} \) are material constants

\[
\sigma = \sigma_0 + \epsilon^{-(C_1 + C_2 T + C_3 \ln(i) + C_4 T^2 + C_5 \ln(i)^2 + C_6 T \ln(i) + C_7 T^3)} \times \left(C_8 T + C_9 \right)^{-\frac{1}{C_10}} \left(C_11 T + C_{12} \right)^\frac{1}{C_13} \left(C_{14} \right)^\frac{1}{C_{15}}.
\]

(16)

The different \( C \) values are presented in Table 1. The resulting fitting shown in Fig. 12 is less good than those for the other domains, but given the few data it is still acceptable. Looking at the evolution of the different \( p_r \), it is clear that softening is rising with temperature and is strongly influenced by strain rate. The scaling factor is mainly dependent of the strain rate, the strain rate sensitivity is only function of the temperature and remains below 0.3 being the theoretical lower limit for superplasticity. Finally the hardening term is slightly influenced by the strain rate but almost not by the temperature looking at the results for the different models at the transition between this domain and the low temperature domain, the transition between the two domains is especially visible for the conditions 650 °C and \( 5 \times 10^{-2} \) s\(^{-1}\), as presented in the Fig. 13. Both models could not well fit the experimental curve. At the early stage of the deformation the experimental results are in agreement with the high temperature high strain rate model but above a strain of approximately 0.008 the experimental curve will be more accurately fitted by the low temperature all strain rates model.

### 5. Conclusions

- Four different domains have been observed in the range of temperature (RT – 1000 °C) and for the strain rates between \( 5 \times 10^{-4} \) and \( 5 \times 10^{-2} \) s\(^{-1}\).
- The deformation mechanisms acting in these domains have been identified and confirmed by microstructural investigations for three of this domains.
- For temperatures below 600 °C, strain hardening and dynamic recovery occur, without dynamic recrystallization.
- At higher temperature and for strain rates below \( 10^{-3} \) s\(^{-1}\), dynamic recrystallization and dynamic grain growth occurred, determining together the optimal conditions for superplasticity by GBS.
- Above 750 °C and for strain rates larger than \( 10^{-3} \) s\(^{-1}\) at these temperatures softening is sufficiently fast to balance the rate of work hardening and mechanisms like dynamic recrystallization and dynamic recovery seems occurring at very high rate.
- A modified Norton–Hoff law was used to fit the experiments in all three domains. The fitting to the experimental data is rather satisfying and the evolution of the \( p \) parameters is in agreement with physically based expectations and data from the literature.
- In the superplastic domain, the introduction of a grain size related factor was necessary in order to get acceptable results.

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