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Advances in Low-Temperature (<0.25T_m) Creep Behavior of Single and Two-Phase Titanium Alloys

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Abstract

Not too long ago, the authors found that twinning can be a very slow process (i.e. many orders of magnitude lower than the speed of sound) and contribute to low-temperature (<0.25T_m) creep deformation behavior in single-phase HCP titanium alloys. Since then, it has been shown that this time-dependent twinning behavior can also be found in BCC titanium alloys. Furthermore, the time-dependent twinning was also found to play a significant role in the two-phase α + β titanium alloys. The extent of twinning and its contribution to creep behavior was found to depend on the grain size in the single-phase titanium alloys. In regard to single-phase beta alloys, the stability of the beta-phase was also found to play a significant role in twinning behavior. With respect to the two-phase α + β titanium alloys, twinning was found to depend on the morphology and volume fraction of the phases and stability of the beta phase. These developments are reviewed and the ramifications in regard to the design and selection of titanium alloys for various applications are presented.

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1. Introduction

Single-phase and two-phase titanium (Ti) alloys are technologically important. Ti alloys are used in various areas such as aerospace [1], biomedicine [2-4], naval applications[5], and energy production [6]. The superior specific strength, excellent corrosion resistance, and biocompatibility of these materials is making them more and more popular [7]. The uses for titanium alloys are likely to grow in the 21st century. It is therefore critical to understand the microstructure and mechanical behavior of Ti alloys.

The deformation behavior of two-phase alloys depends upon the properties of the component phases, the morphology and volume fraction of the component phases and the crystallographic relationships between the phases. When two-phase alloys are subjected to a stress, the individual phases deform differently, resulting in elastic, elasto-plastic, and plastic deformation [8]. Therefore, deformation in these two-phase alloys is a process that cannot be modeled just on individual phase deformation and the rules of mixing.

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One characteristic of titanium alloys is that they are prone to creep at low temperatures (<0.25T<sub>m</sub>) and at low stresses [9-11]. Creep can cause failure in components by a change in dimension which exceeds the allowable tolerance[11-18]; oftentimes, components that depend heavily on titanium, such as aircraft landing gear, require precise part tolerances so predicting low temperature creep becomes very important. Creep in α-titanium alloys has been primarily explained by slip [15]. However, in comparison with α-titanium, studies of low temperature creep in β and near-β alloys have been very limited with a few exceptions [19-25], and the deformation mechanisms consist of primarily slip with some twinning in large grain sized alloys. Creep studies of two-phase titanium alloys have primarily concerned commercial alloys [11, 12, 14-18, 26], and have been attributed to slip in the individual phases as well as across the interphase boundaries.

Over the years, considerable research has been performed examining and seeking to improve the mechanical properties (including low temperature creep) of two-phase alloys [27]. In many of the applications where use of titanium is attractive, low temperature creep is an important factor, where low temperature is defined as less than 25% of the melting temperature (1941 K for Ti). Creep is an important consideration in high-pressure vessels, aerospace components, structural components [28] and artificial joints such as hips or knees [4]. Given the importance of reliability in these cases, it is vital to understand the deformation mechanisms of Ti alloys under these conditions. In this sense, the tensile strength and creep resistance of the two-phase α + β Ti alloys are dependent upon a number of parameters such as microstructure, the properties of the component phases, interactions between phases, and the β-phase stability.

In addition to creep deformation leading to creep failure, creep deformation can also affect other failure modes such as fatigue failure and stress corrosion cracking [9]. This is due to the fact that low temperature creep deformation can result in such features as twins and martensite [19, 22, 29-33], which are not seen in the respective single-phase alloys of equal composition, and which can act as fatigue crack initiation sites. To address some of these issues, Ankem et al. have made significant advances in understanding the low temperature creep deformation behavior of two-phase α + β titanium alloys [29, 31-37]. It has been shown that a decrease in grain size results in a decrease in time-dependent twinning and hence a reduction in creep strain in single phase α and β alloys [22, 29, 32], a trend opposite to that observed during high temperature creep [38, 39]. It was also shown that a fine-equiaxed microstructure is more resistant to creep strain at low temperatures than a Widmanstätten microstructure in two phase α + β alloys, which is also contrary to observations during high temperature creep [40]. Most recently, the authors found that an increase in the stability of the metastable β phase will result in a decrease in creep strain [41] and that whether the β phase deforms by twinning or by martensite formation depends on the volume fraction of the β phase. These developments are briefly reviewed below.

2. Single-phase alpha titanium alloys

When an α-Ti-1.6wt% V alloy with a coarse grain size (225μm) was creep deformed in tension at a stress level of 95% yield stress at 298K, time-dependent {101 ¯ 2} type twinning was observed [29, 33]. Similar phenomena were also observed earlier in a coarse grained α-Ti-0.4wt% Mn alloy [30]. This phenomenon is contrary to the conventional understanding of twinning where the twin growth rates are expected to approach the speed of sound. The activation energy for twin growth is often much less than the activation energy for nucleation so twins have been previously thought to grow to their final lamellar thickness almost instantaneously[42, 43].

Experiments were performed by Ankem & Oberson [33] to determine the activation energy of the creep process from all deformation mechanisms which contribute to creep. The activation energy was found to increase from a value of approximately 38 kJ/mol at low strain (<0.2%) to approximately 47 kJ/mol at higher strain (≥4%) [33]. This was the first time that the activation energy for creep has been calculated for an α-titanium alloy that deforms by twinning. The change in activation energy suggests a potential change in the rate-controlling creep deformation mechanism, from predominately slip at the beginning of the deformation process to a mixed mode consisting of slip and twinning [33]. This suggestion is consistent with the fact that the value of the activation energy for deformation of α-titanium when slip is the predominant deformation mechanism was found to be 30 – 40 kJ/mol [33].
The question is: why are the twin boundaries moving so slowly? It has been traditionally believed that twins grow at speeds approaching the speed of sound in the material. To understand this phenomenon, a crystallographic model has been developed by Oberson & Ankem [36]. Some of the results of this crystallographic model are shown in Fig. 1. The model in Fig. 1 is an extension of the model proposed by Song and Gray [44, 45], where the movement of octahedral interstitial sites are included, in addition to the movement of the parent atoms. This figure shows that none of the octahedral sites are conserved, i.e. they are all displaced. This means that if atoms such as oxygen are present at these interstitial sites, the oxygen atoms will interfere with the twin propagation, and the oxygen must be diffused away from the interface either into the matrix or into the newly formed twin. Therefore, originally a solely shear and shuffle process for the formation of the twin is now controlled by the diffusional process of oxygen atoms. If indeed the oxygen diffusion controls the twin growth, then the activation energy for the twin growth would correspond to the activation energy of oxygen diffusion in α-Ti. The activation energy of twin growth in α-Ti was calculated by measuring individual twin growth rates. This activation energy was found to be 66 kJ/mol [33]. This activation energy was compared to the activation energy of diffusion of oxygen in α-Ti. At 298K, the experimentally measured activation energy for oxygen in α-titanium ranges from about 65 – 200 kJ/mol [46]. The low end of this range is close to the value of the model, suggesting that the growth of twins is controlled by the diffusion of oxygen. It is also possible that the slightly lower activation energy can be related to the fact that dislocation motion or slip is also involved in the creep deformation process simultaneous with twinning and the activation energy for the slip process is lower than that for twinning.

Fig. 1. Schematic illustration of {101̅2} twinning in the HCP lattice of α-titanium. Projection of the lattice is onto the {11̅20} plane. (a) Shear of the lattice in the <110̅1> direction moves only A-type Ti atoms (large, corner circles) directly to twinned positions. (b) Arrows give the shuffles required to move the Ti atoms in B-type sites (large, center circles) to their twinned B-type or C-type positions. Reorientation of the lattice eliminates the octahedral sites (small circles with “x” mark) where oxygen could reside. (c) Schematic illustration of a completed {101̅2} twinning event. The completed positions for B-type Ti atoms and the corresponding possible octahedral sites and completed positions for C-type Ti atoms are indicated by the arrows. These models are an extension of earlier twinning models by Song and Gray [44, 45], but now include the corresponding movements of octahedral sites where oxygen may reside [36].
3. Single-phase beta titanium alloys

Similarly to α-titanium alloys, it was found that time dependent twinning and slow twin growth also plays a significant role in the low temperature creep behavior of some coarse-grained β-titanium alloys. Doraiswamy & Ankem studied the creep behavior of a coarse-grained β-Ti-14.8wt% V alloy with a grain size of 350 μm and deformed at 95% yield stress and 298K[19]. It was found that slip as well as [332]<113> twinning are the major deformation mechanisms, and the amount of twinning and hence creep strain increases with increasing grain size and decreasing beta phase stability; the creep data was used to calculate the activation energy for the deformation process[35]. The activation energy was shown to increase with increasing strain, from 39 kJ/mol to 112 kJ/mol. This variation is due to a change in the dominant creep deformation mechanisms, from slip at the beginning of the creep process to twinning at the end. This suggestion is consistent with the work of Ramesh & Ankem [22] who have observed slip at low strains and twinning at high strains in similar β-Ti alloys.

Similarly to α-titanium alloys, the slow growth of twins was attributed to the movement of octahedral interstitial oxygen atoms away from the twin-matrix interface. Detailed crystallographic models were developed for the [332]<113> twinning process which involves both shear and shuffles as well as the destruction of the octahedral interstitial sites at the twin-matrix interface [35]. Some of the details of this crystallographic model are shown in Fig. 2. It is suggested that twinning shear alone does not create a valid twin and an additional shuffle is required to move Ti atoms to their final positions for this particular twin. The reorientation of the atoms eliminates the space where the interstitial oxygen atoms can reside (octahedral interstitial site), and they are required to diffuse to an acceptable site. Based on the measured twin growth rates in the β alloy, the activation energy for the creep deformation process was calculated as 105 kJ/mol [35]. This is close to the value for the diffusion of oxygen in β-Ti of 122 kJ/mol determined by Claisme and Koenig [47]. These results suggest oxygen atoms play a significant role in controlling the growth of twins; thereby controlling the extent of creep strain.

4. Two-phase alpha + beta titanium alloys

In regard to two-phase α + β titanium alloys, it was found that the deformation mechanisms could be quite different from those observed in the respective single-phase α and β alloys. For example, in the case of a 51%α – 49% β Ti-8.1wt% V alloy, with Widmanstätten α+ β microstructure, the creep deformation mechanisms were found to be twinning in the α phase and stress-induced martensite in the β phase [32, 34]. There were at least two surprising results here. First, only slip was found to be a deformation mechanisms in single phase α-Ti-1.6wt% V where the grain size was small (<60μm) [29]; however, in the case of this two phase alloy, twinning was observed in the α phase even though the α plate thickness is about 3.0μm. Secondly, in the case of the

![Fig. 2. Projection of BCC substitutional and octahedral interstitial atoms onto the (110) plane. (a) Untwinned structure showing direction of [113] twinning shear, and (b) Final twinned structure showing mirror symmetry across twin/matrix interface [35].](image-url)
single-phase β alloy discussed in Section 3, twinning along with slip were found to be the major mechanisms; however, in this two-phase alloy, stress induced martensite was found to be a major creep deformation mechanism in the β phase. It is to be noted that the chemical composition of the beta phase in both single phase and two phase alloys is the same. Additionally, twinning can occur in small equiaxed α grains in a two-phase Ti-8.1 wt% V alloy, suggesting that the interactions seen between phases with Widmanstätten microstructures also play a role in equiaxed microstructures[32]. These interesting results have been explained on the basis of elastic interaction stresses [27, 32, 34], which can oftentimes be much higher than the applied stress, triggering additional mechanisms.

During high temperature creep deformation, a Widmanstätten microstructure is known to have a greater resistance to creep strain due to the orientation relationship which exists between the α and β phases [38-40]. This platelet structure hinders interphase interface sliding which is a primary deformation mechanism during high temperature creep. However, as Figure 3 shows, at low temperatures (298 K), a Widmanstätten microstructure exhibits about 13% greater creep strain than an equiaxed microstructure for an equivalent testing time. Figure 4 demonstrates the reason for this unexpected trend [32]. As evidence by the figure, coarse deformation products (specifically twins in the α phase and stress induced martensite in the β phase) can span across phase boundaries and across large distances. It is the same Burgers orientation relationship that helps reduce creep strain during high temperature creep that provides an easy pathway for the transmission of deformation mechanisms between phases during low temperature creep, thus increasing creep strain as opposed to an equiaxed microstructure. Whether martensite instead of twinning occurs in two-phase alloys was also found to depend on the volume fraction of phases.

![Test Temperature: 298K](image)

Fig. 3. Creep curves of Ti-8.1 wt% V alloys, tested at 298 K and 95% YS. Note the increased creep strain of the Widmanstätten microstructured alloy over a similarly grain sized equiaxed alloy.

![TEM and SEM images](image)

Fig. 4. (a) TEM micrograph and (b) SEM image showing connection between twinning in α-phase and stress induced martensite in β-phase. Note the connected deformation features can span many grains and cover large distances, leading to increased creep strain [32].
5. Summary

Given that the diffusion of oxygen, and hence the rate of twin growth in α and β titanium alloys, depends on the number of interstitial sites occupied by oxygen, these results suggest a variation in the amount of oxygen in the alloy can have a significant effect on the amount of room temperature creep deformation. Further, the twin/matrix boundary can act as a nucleation site for fatigue cracks and other deformation mechanisms which would have another direct effect on the structural integrity of titanium components. Since this behavior is similar in both HCP and BCC crystal structures, this suggests that the models presented here are broadly applicable to describe the effect of interstitial impurities on twinning during creep of many crystalline materials such as metals, intermetallics, ceramics, and high-temperature super-conductors with a variety of crystal structures; where impurities can play roles in deformation mechanisms.

In two-phase $\alpha + \beta$ Ti alloys, many factors contribute to the possible deformation mechanisms during low temperature creep; including grain size, $\beta$-phase stability, elastic interaction stresses, and crystallographic phase orientation relationships. The mechanisms in the two-phase $\alpha + \beta$ alloys included slip and twinning in the $\alpha$-phase, and slip, stress induced martensite, and twinning in the $\beta$-phase. It was shown that new deformation mechanisms not seen in the respective single-phase alloys, such as stress induced martensite, can be explained by the interaction stresses between the two phases and the increased creep strain of Widmanstätten microstructures can be explained by these new deformation mechanisms as well as by the orientation relationship which exists between the two phases. In the future, the most important factor to predict and design multi-phase alloys for increased creep resistance, will be the control of the interactions between phases, which in turn depend on all of the factors discussed.

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References