A LITERATURE REVIEW OF LOW TEMPERATURE ($<0.25T_{mp}$) CREEP BEHAVIOR OF α, α-β, AND β TITANIUM ALLOYS

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Prepared by

R. Ankem (Consultant)
T. Wilt

Center for Nuclear Waste Regulatory Analyses
San Antonio, Texas

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ABSTRACT

This study systematically reviews the low temperature creep deformation behavior of α, α-β, and β titanium alloys as a function of (i) stress level, (ii) temperature, (iii) chemical composition, and (iv) microstructure of phases for Titanium Grades 7 and 24 or their surrogates. Creep mechanisms discussed include slip, twinning, grain boundary sliding, α-β interface sliding, Widmanstatten colony boundary sliding, and stress-induced martensite. In addition to these mechanisms, various factors that can affect creep deformation include microstructure, texture, proportional limit, elastic interaction stresses, stress type, and cold work. The effect of creep on fracture toughness at low temperatures is also reviewed. It is known that α, α-β, and β titanium alloys are prone to sustained load cracking, and the literature review indicates that there is sufficient information to suggest that creep strain can affect fracture toughness.

The general description of the low temperature creep behavior is presented. The appropriate choice of constitutive relationships, which may be either logarithmic or parabolic and dependent on available data, is discussed. The associated material coefficients of these models governing long-term, low temperature creep behavior are identified. This report establishes a technical basis for the stress threshold needed to initiate low temperature creep of Titanium Grades 7 and 24, or reasonable surrogates, based on the information available in the public domain. The literature indicates that the threshold stress range is 35–60 percent and 25–50 percent of the yield stress for surrogates of Grade 7 and Grade 24, respectively.

The review determined that surrogates of both of these titanium grades may be vulnerable to creep failure when subjected to stresses close to the yield stress for 10,000 years, although this review also indicates that there is a lack of specific information on the creep behavior of Grade 7 and Grade 24 alloys to draw definitive conclusions. It is suggested that experiments may be conducted to determine threshold stresses and to determine the creep model constants for accurate predictions of the long-term creep behavior of the drip shield components.

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USEFUL UNIT CONVERSIONS

Force
1 lbf = 4.4482 N
1 kgf = 9.806665 N

Stress
1 ksi = 6.8948 MPa
1 dyne/cm² = 0.1 Pa
1 kgf/mm² = 9.806665 MPa

Toughness
1 ksi $\sqrt{in}$ = 1.089 MPa $\sqrt{m}$
1 MPa $\sqrt{m}$ = 1 $MNm^{-3/2}$

Energy
1 cal = 4.1840 J
1 ft*lbf = 1.3558 J
1 eV = 1.6022×10^{-19} J
1 cal/g = 4.1840×10^{3} J/kg
1 cal/g* °C = 4.1840×10^{3} J/kg*K

Temperature
$T_C = (T_F - 32)/1.8$
$T_C = T_K = 273.15$

Conversion wt% oxygen to at.% oxygen

at%(O) = (wt%(O) * 47.887)/(wt%(O) * 47.887 +wt%(Ti) * 15.999)

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QUALITY OF DATA, ANALYSES, AND CODE DEVELOPMENT

DATA: All CNWRA-generated data contained in this report meet quality assurance requirements described in the Geosciences and Engineering Division Quality Assurance Manual. Sources of other data should be consulted for determining the level of quality of those data.

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EXECUTIVE SUMMARY

The U.S. Department of Energy (DOE) has been studying the Yucca Mountain site in Nevada to determine whether it is suitable for building a potential geologic repository for the disposal of the nation's spent nuclear fuel and high-level radioactive waste. If this potential repository is approved, these radioactive materials will be encapsulated in waste packages and emplaced in tunnels (drifts). Subsequently, drip shields will be placed over the waste packages. The drip shield is an engineered barrier that limits the amount of water contacting the waste packages as a result of seepage into the near-field environment. The proposed drip shield design is to protect the waste packages from drift degradation. As a result, staff expects that DOE drip shield analyses will include all potential loading scenarios to ensure a complete and high-quality license application. Specifically, the drip shield should withstand rockfall impacts, as well as static and dynamic (seismic) loading pressures originated by the accumulated rockfall rubble. The drip shield evaluation should include the effect of high temperatures, corrosion, and low-temperature creep, etc. These analyses will provide a basis for evaluating whether the performance objectives for the repository are achieved after permanent closure 10 CFR 63.113 (Code of Federal Regulations, 2005).

Creep processes, which may lead to deformation of the drip shield in response to deviatoric (i.e., pressure independent) stress are excluded by DOE based on a low probability screening criterion (Bechtel SAIC Company, LLC, 2004). The temperature in the drip shield will be less than 300 °C [572 °F], and according to DOE, the deformation of many titanium alloys loaded to yield point does not increase with time for temperatures between 200 and 315 °C [392 and 599 °F] (Bechtel SAIC Company, LLC, 2004). Given that creep rates decrease at lower temperatures (i.e., $< 0.25\ T_{mp}$ where T_{mp} is the melting point of the material), DOE concludes that creep deformation of the drip shield will not occur to any appreciable extent under repository exposure conditions. Moreover, DOE indicates that the emplacement drifts will remain stable for 10,000 years under nominal scenario conditions (Bechtel SAIC Company, LLC, 2004). Thus, the drip shield will not be subjected to permanent stresses that may lead to creep processes.

Staff analyses, however, indicate that the proposed thermal loading could cause the drifts to degrade and fill with rubble within a few hundreds of years (Ofoegbu, et al., 2004). Therefore, Ibarra, et al. (2005) investigated the maximum vertical load that the drip shield can withstand when subjected to permanent static loading (i.e., the vertical load carrying capacity). The analyses of the current drip shield design indicate that structural instability may be caused by plastic buckling of the Titanium Grade 24 support beams (columns). The drip shield vertical load carrying capacity, however, may decrease due to creep processes once the stresses on the drip shield approach or exceed the material yield stresses. If the drip shield exhibits large creep deformation, mechanical interaction of the drip shield and the waste package outer barrier may take place, potentially causing stress concentrations on the waste package.

Because the conditions for creep processes may be present in the engineered barriers, this report presents survey of the available literature pertaining to the long-term, low temperature creep behavior of Titanium Grades 7 and 24, which are the materials used in the current drip shield design. Titanium alloys are α, α-β, or β depending on the amount of alloy element. Grade 7 is an α alloy and Grade 24 is an α-β alloy. The general factors that affect creep deformation are the stress levels, temperature, and microstructure of the material. Specifically, this report investigates the creep mechanisms of slip, twinning, grain boundary sliding,

α-β interface sliding, Widmanstatten colony boundary sliding, and stress-induced martensite. Parameters such as the microstructure of phases, chemical composition, texture, stress type, and cold work are also evaluated.

The results of the literature search indicate that no data is available for Titanium Grade 24, whereas the information about Titanium Grade 7 is not sufficient. Thus, appropriate surrogates are selected based on titanium alloys that exhibit similar chemical compositions, microstructure, and tensile properties. Based on these criteria, Titanium Grade 5 is selected as surrogate of Titanium Grade 24, and Titanium Grade 2 is used as surrogate for Titanium Grade 7. The main difference between the drip shield material and the corresponding surrogates is the palladium content. Palladium is added to improve corrosion resistance and may not have a significant effect on the mechanical behavior of titanium alloys in small quantities. The available literature indicates that the threshold stress for low temperature creep is in the range of 35 to 60 percent of the yield stress for Titanium Grade 2 and in the range of 25 to 50 percent of the yield stress for Titanium Grade 5. For stress levels close to the yield stress, these materials may experience creep failure after several thousands of years. In this report, the creep deformation rate for titanium material experiencing stresses larger than the yield stress has not been evaluated. These predictions, however, are preliminary because of the lack of complete experimental information of the drip shield materials.

The effect of creep on fracture toughness of these titanium alloys at low temperatures is also reviewed. It is generally accepted that α and α-β titanium alloys are prone to sustained load cracking, which involves subcritical crack growth under the absence of an aggressive environment. The available information suggests that creep strain can affect sustained load cracking behavior (i.e., creep can affect fracture toughness.)

The information collected in this extensive review demonstrates that creep deformation of the titanium drip shield components can occur at low temperatures ($T<0.25T_{mp}$) and at stress levels below the yield stress. This report has determined that there is limited data pertaining to the creep behavior of Titanium Grades 7 and 24. A detailed investigation of the microstructure of titanium in general is used to select surrogates of Grades 7 and 24 from which initial estimates are made of long-term creep deformation. The limited data serves to motivate the necessity for creep experiments to be conducted in order to accurately characterize the creep constitutive models presented in this report. The appropriate creep model will be used to estimate the amount of creep deformation in the titanium drip shield components.

References

Bechtel SAIC Company, LLC. "FEPs Screening of Processes and Issues in Drip Shield and Waste Package Degradation." Rev. 3. Las Vegas, Nevada: Bechtel SAIC Company, LLC. 2004.

Code of Federal Regulations. "Disposal of High-Level Radioactive Wastes in Geologic Repository at Yucca Mountain, Nevada." Title 10—Energy, Chapter 1—Nuclear Regulatory Commission, Part 63. Vol. 2. Washington, DC: U.S. Government Printing Office. 2005.

Ibarra, L., T. Wilt, G. Ofoegbu, and A. Chowdhury. "Structural Performance of Drip Shield Subjected to Static and Dynamic Loading." San Antonio, Texas: CNWRA. 2005.

Ofoegbu, G.I., B. Dasgupta, and K.J. Smart. "Thermally Induced Rock Stress and Implications for Degradation of Emplacement Drifts at the Potential Yucca Mountain Waste Repository." San Antonio, Texas: CNWRA. 2004.

1 INTRODUCTION

1.1 Background

The U.S. Department of Energy (DOE) is proposing to build a potential geological repository at Yucca Mountain in Nevada for the disposal of spent nuclear fuel and high-level waste which will be placed in waste packages and emplaced underground in tunnels (drifts). 10 CFR §63.113 requires that the geologic repository include an engineered barrier system, as well as natural barriers, so as to limit the release of radionuclides. These engineered barriers must be designed to provide protection for the performance period.

The titanium drip shield (Figure 1-1) was originally intended to protect the waste packages from water due to seepage in the near-field environment as well as to prevent rocks from directly impinging on the waste package. As a result, the U.S. Nuclear Regulatory Commission (NRC) required documentation of analyses pertaining to point loading due to rockfall (Schlueter, 2000). These issues were addressed in Bechtel SAIC Company, LLC (2003, Appendix K) which provides a response to key technical issues pertaining to Container Life and Source Term 2.02, 2.08, and 2.09. This appendix presents finite element results for point load rockfall analyses for both the waste package and drip shield. The primary finding was that the drip shield does not contact the waste package when impacted by the largest rock block used in the finite element analyses (Bechtel SAIC Company, LLC, 2003, Appendix K). Note that even though the drip shield does not immediately fail due to static or dynamic loads, the drip shield sides and crown will carry dead-weight loading due to rockfall. The result of this sustained loading is the possible occurrence of creep.

The load carrying capacity of the drip shield was recently investigated by Ibarra, et al. (2005), finite element analysis of the drip shield predicts failure due to plastic buckling of the

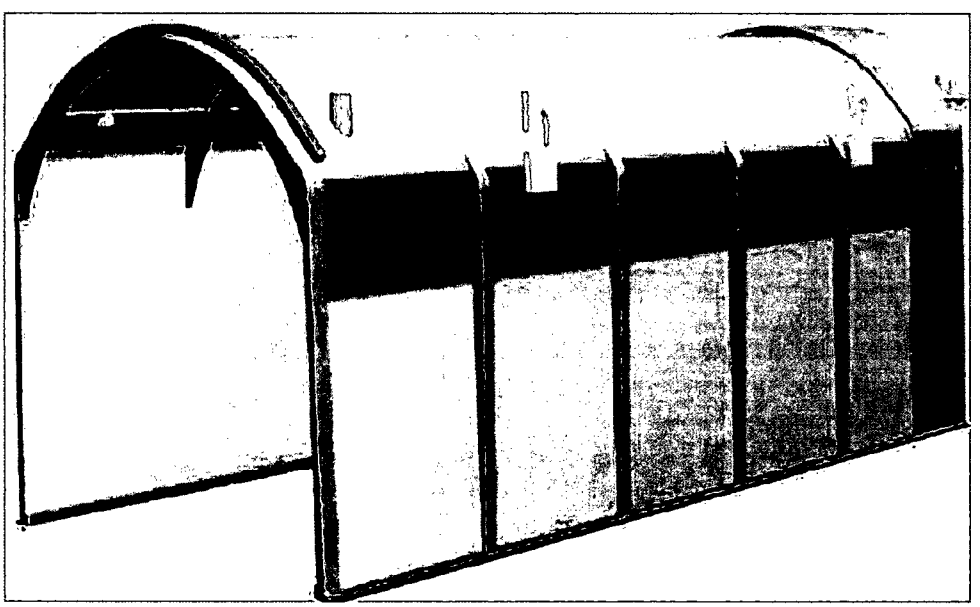

Figure 1-1. Illustration of Proposed Drip Shield Design (Neuberger, et al., 2002) (Reproduced With Permission From the Materials Research Society and Neuberger)

Titanium 24 support beams (columns). Note that the presence of plasticity may be a precursor to the development of creep in this location. In addition, due to the sustained rockfall load, creep may occur in the drip shield crown which consists of Titanium 7 shell and plate components and the Titanium 24 bulkhead and longitudinal stiffeners. If the drip shield begins to creep, the resulting deformation increases the possibility that the drip shield will have mechanical interaction with the outer barrier of the waste package which may cause localized stress concentrations. These areas of stress may lead to stress corrosion cracking of the waste package Alloy 22 outer barrier and result in the breach of the outer barrier.

Because the waste package is the heat source, the waste package surface temperature would be greater than the drip shield surface temperature (Bechtel SAIC Company, LLC, 2004). DOE believes that possible creep in the drip shield at expected temperatures of 200–315 °C [392–599 °F] will be screened out because titanium alloys are not expected to creep at these temperatures (Bechtel SAIC Company, LLC, 2004). NRC and the Center for Nuclear Waste Regulatory Analyses, citing Neuberger, et al. (2002), commented that long-term performance of the drip shield components may be affected by creep at stresses that are at a significant fraction of the yield stress, and DOE should consider the potential for low temperature creep of the drip shield. Therefore, this report investigates the long-term, low temperature creep behavior of Titanium Grade 7 and Grade 24 alloys used in the current drip shield design.

Titanium undergoes an allotropic phase transformation at 883 °C [1,621 °F]. This transformation temperature is known as β-transus. The titanium phase which exists above this temperature is known as β, and it has a body-centered cubic crystal structure. The lower temperature phase is known as α, and its crystal structure is hexagonal close packed. The titanium alloys can be broadly classified into α, α-β, or β alloys depending on the alloying elements. Grade 7 is an α alloy, and Grade 24 is an α-β alloy. It is known that titanium alloys can creep at low temperatures including room temperature. The creep deformation mechanisms of titanium alloys at low temperatures include time-dependent twinning (Ankem, et al., 1994; Hultgren, et al., 1999), interface sliding (Greene and Ankem, 1995), and slip (Ankem, et al., 1994; Hultgren, et al., 1999; Greene and Ankem, 1993; Greene, et al., 1995). Creep can directly affect component performance or indirectly affect material properties such as fracture toughness.

1.2 Objectives and Scope

The primary objective of this literature review is to systematically review α and α-β titanium alloys, Grades 7 and 24 in particular. The titanium alloys will be evaluated in terms of their creep deformation behavior, including stress level, temperature, chemical composition and microstructure of phases. A technical basis will be established for determining the stress threshold needed to initiate low temperature creep of Titanium Grades 7 and 24, or reasonable surrogates, based on the information available in the public domain. In addition, various creep constitutive models and their material coefficients will be identified to estimate creep strain as a function of time once the initiation stress threshold is exceeded for these materials. Sample predictions of creep strains after 10,000 years will be made based upon the appropriate creep models.

This report initially focuses (Chapter 2) on a general review of the low temperature creep behavior of titanium alloys and will include the forms of the creep models to be used and the corresponding activation energies for various low temperature creep processes. Various creep

deformation mechanisms will be discussed including slip, twinning, grain boundary sliding, α-β interface sliding, Widmanstatten colony boundary sliding, and stress-induced martensite. In addition, factors that can affect creep deformation such as chemistry, microstructure, texture, stress level and proportional limit, elastic interaction stresses, stress type, cold work, and creep test temperature will also be investigated. A brief discussion on the selection of the appropriate selection of surrogates of Titanium Grades 7 and 24 is given in Chapter 3. Chapter 4 of this literature review will focus on the low temperature creep of Titanium Grades 7 and 24 and their surrogates. Finally, the effect of low temperature creep on the fracture toughness of titanium alloys will be addressed (Chapter 5).

2 LOW TEMPERATURE CREEP OF TITANIUM ALLOYS—GENERAL

2.1 Constitutive Relationships/Equations

Strain versus time of pure metals and simple single-phase materials can be represented by three stages of creep deformation as shown in Figure 2-1: primary, secondary, and tertiary creep. Creep at low temperatures (< $0.25T_{mp}$) (T_{mp} is the melting point of the specific material) is generally transient, achieving creep exhaustion during stage 1 (primary creep) and never reaching stage 2 (secondary creep). Thermal energy required for dynamic recovery is not available in the temperature ranges of interest; therefore, the material experiences creep exhaustion or creep saturation.

For single-phase materials, a number of equations describe the creep behavior of each stage (Bendersky, et al., 1985; Evans and Wilshire, 1985). Primary or transient creep is often described by the following empirical power law equation (Imam and Gilmore, 1979; Miller, et al., 1987; Suri, et al., 1999; Thompson and Odegard, 1973) for a given uniaxial stress σ, normalized with respect to yield stress

$$\varepsilon = At^n \tag{2-1}$$

where

ε	—	strain
A	—	a constant
n	—	time exponent
t	—	time

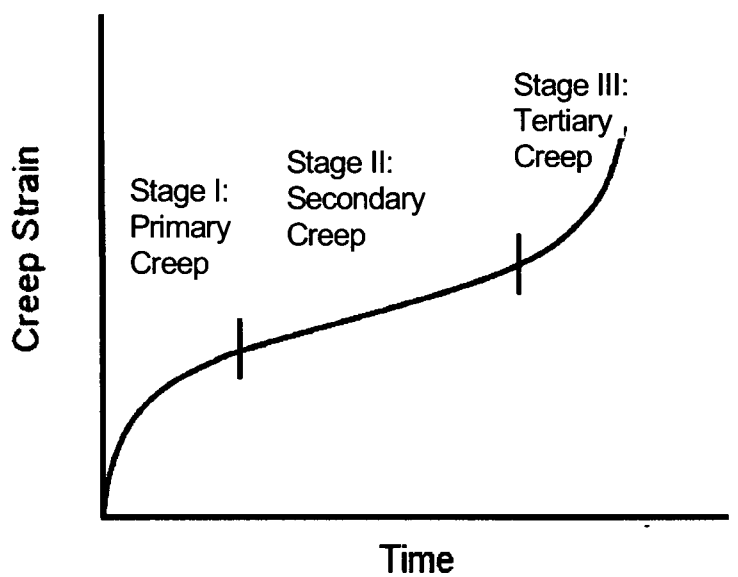

Figure 2-1. Typical Creep Curve Showing the Three Stages of Creep

The constant A is reportedly dependent on microstructure and temperature. Second stage creep is typically never achieved at low temperatures. The activation energies for non-steady-state creep at constant stresses were determined earlier by Thompson and Odegard (1973) by using Eqs. (2-2) and (2-3). Thompson and Odegard (1973) utilized the time derivative of the Eq. (2-1), $\dot{\varepsilon}$, to eliminate the time dependence of the strain rate. That is, the equation for the strain rate as a function of strain, ε, is given by

$$\dot{\varepsilon}(\varepsilon) = nA^{1/n}(\varepsilon)^{1-1/n} \tag{2-2}$$

Thompson and Odegard (1973) calculated the activation energy $Q(\varepsilon)$ at a given uniaxial stress σ from the equation

$$Q(\varepsilon) = -R\left[\frac{\Delta \ln \dot{\varepsilon}(\varepsilon)}{\Delta(1/T)}\right]_{const.\sigma} \tag{2-3}$$

where

R	—	gas constant
σ	—	uniaxial stress (normalized with respect to the corresponding yield stress at a given temperature)
ε	—	selected strain level at which $Q(\varepsilon)$ is calculated

Equation (2-3) implies that for non-steady-state primary creep, the $Q(\varepsilon)$ value may vary depending on the strain level ε chosen. To account for this variation, Thompson and Odegard (1973) have calculated $Q(\varepsilon)$ values for different strain levels and computed an average $Q(\varepsilon)$ value. The variation of $Q(\varepsilon)$ with the strains was found to be very small, suggesting that the averaging procedure is acceptable to determine the value for $Q(\varepsilon)$. Earlier, Cuddy (1970) utilized a similar procedure, determined the activation energies at different strain levels for steel specimens, and found that the activation energies thus obtained did not vary significantly with strains. For non-steady-state conditions, Cuddy further showed that for a given strain level, the strain rate, $\dot{\varepsilon}$, is proportional to σ^{n^*}, i.e.,

$$\dot{\varepsilon} \propto \left[\sigma^{n^*}\right]_{const.\,T,\varepsilon} \tag{2-4}$$

where

$\dot{\varepsilon}$	—	strain rate
n^*	—	stress exponent
σ	—	stress

Based on the results of Thompson and Odegard (1973) and Cuddy (1970), it is evident that the following equation can be used for the determination of Q for the non-steady state condition, with the understanding that an average Q value may have to be determined if Q changes significantly with strain level

$$\dot{\varepsilon}(\varepsilon) = \left[A\sigma^{n^*} \exp\left(\frac{-Q(\varepsilon)}{RT} \right) \right] \tag{2-5}$$

It is clear that Eq. (2-5) can be used for steady-state creep as well, because at steady-state, strain rate does not vary with strain. Miller, et al. (1987) studied the creep behavior of a near α two-phase titanium alloy, Ti-6Al-2Nb-1Ta-0.8Mo, at temperatures ranging from 25 to 600 °C [77 to 1,112 °F] and have adopted Eq. (2-5) to determine the activation energy for steady-state creep. Miller, et al. (1987) found that at temperatures much less than 515 °C [959 °F], the activation energies were much lower than that for self-diffusion of titanium. In addition, it was suggested that different creep mechanisms may be operating depending on the *Q* value, although these creep mechanisms were not clearly identified.

Sargent and Ashby (1982) have developed deformation maps for pure single-phase materials. A deformation map for pure titanium (Figure 2-2) shows that the deformation mechanisms depend on the normalized stress σ/μ and the homologous temperature T/T_{mp}, where μ is the shear modulus and T_{mp} is the melting point of the material. The boundaries between different mechanisms depend on the stress level, temperature, and grain size. In these maps, there is little information at the low stress and low temperature range of interest (< $0.25T_{mp}$). Therefore,

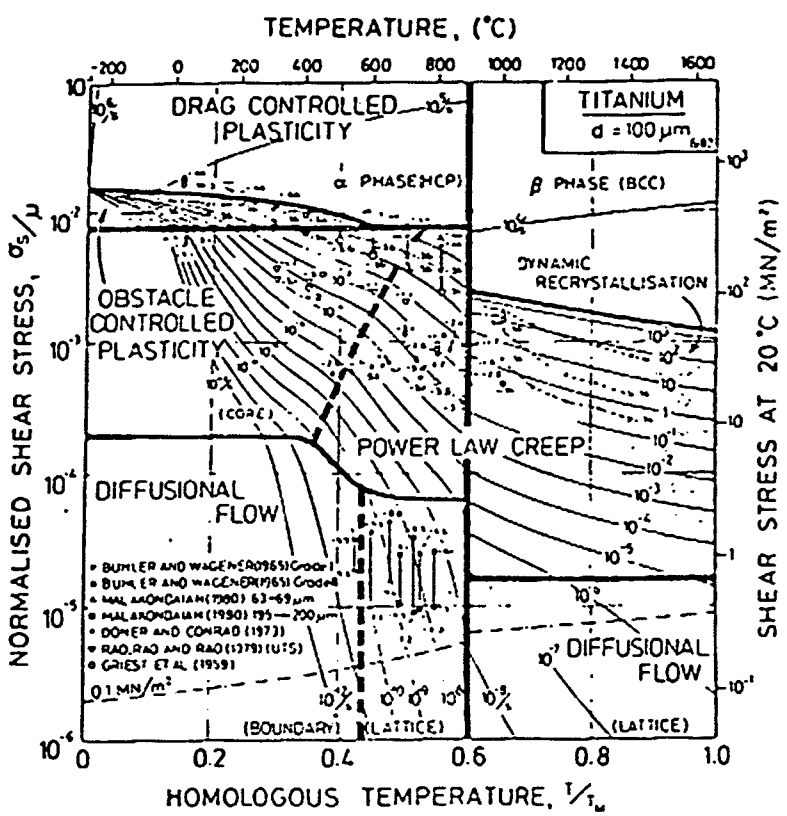

Figure 2-2. The Deformation Mechanisms Map for Commercially Pure Titanium With a Grain Size of 100 μm [3,900 μin] (Sargent and Ashby, 1982) (Reproduced With Permission From Elsevier)

it is difficult to predict the activation energies of the creep mechanisms that might be operating in the α, α-β, and β titanium alloys based on this map.

There are two creep mechanisms for polycrystalline single-phase materials, which are not explicitly identified in deformation map for pure titanium (Sargent and Ashby, 1982). The first is grain boundary sliding, which normally occurs at high temperatures, and the second is twinning, which has been observed at low temperatures. The activation energy for the creep process will determine which mechanism is active. For example, Boehlert and Miracle (1999) measured the activation energy for creep of orthorhombic + body-centered cubic titanium alloys at 650–750 °C [1,202–1,382 °F] as a function of stress level. It was determined that the activation energy increases and the creep mechanism changes with increasing stress level. Several theories of grain boundary sliding in polycrystalline materials have been proposed and they all result in expressions of the type

$$\dot{\varepsilon}_{gbs} - (\varepsilon) \propto \sigma^{\hat{n}} d^{-m} \exp\left(-Q(\varepsilon)/RT\right) \tag{2-6}$$

where

$\dot{\varepsilon}_{gbs}(\varepsilon)$	—	strain rate due to grain boundary sliding
$\hat{n}$ and m	—	integers greater than or equal to unity
$Q(\varepsilon)$	—	activation energy
σ	—	stress
d	—	grain size
R	—	gas constant
T	—	temperature

Grain boundary sliding (α-α or β-β) may not be a significant creep mechanism at low temperatures.

At ambient temperature 25 °C [77 °F], time-dependent twinning is a dominant mechanism for creep deformation in coarse-grained α and β titanium alloys (Ankem, et al., 1994; Hultgren, et al., 1999). Note that twinning was previously understood to be a diffusionless transformation where the twin growth normally occurs at the speed of sound (Dieter, 1986, p. 133; Reed-Hill and Abbaschian, 1992). The time-dependent twinning phenomenon reported by Ankem, et al. (1994) is the first time such a slow growth of twins was observed in bulk crystalline materials. However, recent studies by Murty (2003) showed no twinning for biaxial creep of commercially pure Ti and Ti 3Al-2.5 V. Due to the small grain size of the alloys studied, only slip is expected. Further, studies by Salem, et al. (2003) suggest that twinning is important for strain-hardening behavior. Twins form a barrier to dislocation motion and reduce the effective slip distance. Dislocation mechanisms for creep have been proposed by several investigators (Christian, 1965; Hirth and Lothe, 1982). Additionally, arrays of aligned $\langle 11\bar{2}0 \rangle$ screw dislocations were seen by Neeraj, et al. (2000) in a study of Ti-6Al. This contrasts with the twinning mechanism observed in coarse-grained Ti-1.6V and Ti-0.4Mn. This implies that aluminum promotes slip and hinders twinning. The steady-state creep rate due to twinning $\dot{\varepsilon}_{twin}$ at a given stress can be represented generally by

$$\dot{\varepsilon}_{twin}(\varepsilon) = A \exp(-Q/RT) \tag{2-7}$$

where

A — constant (constant stress)

For non-steady-state or transient creep, based on Eq. (2-5), it is reasonable to suggest that $\dot{\varepsilon}_{twin}$ can be represented by

$$\dot{\varepsilon}_{twin}(\varepsilon) = \left[A\sigma^{n^*} \times \exp\left(\frac{-Q_{twin}(\varepsilon)}{RT} \right) \right] \tag{2-8}$$

In the single-phase alloy Ti-13Mn, very little ambient temperature {25 °K [77 °F]} creep deformation was observed at 100 percent yield stress after 200 hours (Greene and Ankem, 1993). Nevertheless, some coarse and wavy slip lines were observed on the surface of polished specimens. Decreasing the stability of the phase by decreasing the weight percent of manganese resulted in the ambient temperature creep deformation mechanism of stress-induced plate (SIP) formation. Stress-induced plate formation was also observed in Ti-14.8V β Ti alloy (Ramesh, 1997). These stress-induced plate formations have been reported as {332} <113> twins in the Ti-14.8V alloy (Ramesh and Ankem, 1999). The amount of creep deformation was found to depend strongly on the stability of the β phase as well (Doraiswamy and Ankem, 2003).

It is well known (Dieter, 1986, p. 133) that the creep behavior of single phase metals and alloys depends on a number of factors: temperature, stress, grain size, and dislocation structure. For two-phase materials, morphology and volume fraction of the second phases are also significant (Evans and Wilshire, 1985; Miller, et al., 1987; Imam and Gilmore, 1979; Ankem, et al., 1989; Ankem and Seagle, 1985; McLean, 1980, 1985; Dyson and McLean, 1983; Henderson and McLean, 1983, 1985). For these reasons, most of the earlier creep models were concerned with simple models based on the law of mixtures (Cho and Gurland, 1988; Ankem and Margolin, 1986a; Chen and Argon, 1979; Elliot, 1965; Greene, 1994; Grewal and Ankem, 1990, 1989). For two-phase materials, there are two simple models based on the law of mixtures: the first model assumes that both phases have the same strain (isostrain, i.e., constant strain), and the second assumes that the stress is the same (isostress) in the two phases (i.e., constant stress, during the course of deformation). For the isostrain model it is assumed that $\dot{\varepsilon}_{\alpha-\beta} = \dot{\varepsilon}_{\alpha} = \dot{\varepsilon}_{\beta}$ for a given strain ε and additionally for the isostress model $\sigma_{\alpha-\beta} = \sigma_{\alpha} = \sigma_{\beta}$ for a given stress σ.

For two phase materials it has been generally shown that stress-strain relations and creep properties of two-phase materials cannot be predicted by assuming either constant strain (strain rate) or constant stress (Ankem and Margolin, 1986a,b; Ankem, et al., 1989; Chen and Argon, 1979) without modifications. In this regard, Kolluru and Pollock (1998) utilized finite element analysis to model creep deformation in discontinuous, unidirectional fiber-reinforced composites. The finite element model, however, did not take into account interface sliding, which can occur in titanium alloys. Further, Hasija, et al. (2003) modeled the creep deformation behavior of polycrystalline Ti-6Al, emphasizing on how heterogeneity of the crystal affects local response. In other words, the interface stresses caused by a mismatch in strength of adjacent crystals in a single-phase alloy were measured; however, interface sliding was not considered. The previous models also did not include the effect of twinning in α and its effect on

stress-induced martensite in β during creep. In addition, the models did not consider the effect of β stability on creep.

As noted before, transient creep, which is expected in Grade 7 and Grade 24 at temperatures below 0.25 T_{mp}, often follows a power-law equation, as described in Eq. (2-1). Transient creep may also follow a logarithmic fit of the type

$$\varepsilon = A' + B\ln(t) \tag{2-9}$$

where

ε	—	instantaneous plastic and creep strain
A' and B	—	creep constants
t	—	time measured in hours

Equation (2-9) is particularly suited when the creep strains are relatively small, as in the case of β titanium alloys reported by Doraiswamy and Ankem (2003) or at stresses much lower than the yield stress shown by Aiyangar, et al. (2005) and Drefahl, et al. (1984). It is extremely important to select the right equation to predict the short-term creep behavior, otherwise erroneous conclusions may be drawn, which will be demonstrated in a later section.

Equations (2-1) and (2-9) show that when time t is set equal to 1 hour, A and A' should be equal. As most creep deformation takes place within the first few hours, A and A' are equal to the creep strain.

Neeraj, et al. (2000) suggested that the constants in Eq. (2-1) can be obtained through constant strain rate tests via Holoman constants. Neeraj, et al. (2000) predict a constant time exponent n value for different stress levels. However, it has been shown (Aiyangar, et al., 2005; Imam and Gilmore, 1979; Miller, et al., 1987; Odegard and Thompson, 1974) that for different alloys, n normally decreases with a decrease in stress at stress levels that are much below the yield stress. Therefore, this method may not be suitable for long-term creep predictions.

When the applied stresses and/or the temperatures are high, fracture can occur due to creep, which is termed as stress rupture. At a given stress, the time to rupture and temperature are related by the Larsen-Miller parameter P (Dieter, 1986, p. 463) as

$$P = T(\log t + C) \tag{2-10}$$

where

t	—	time to rupture
T	—	absolute temperature (usually in degrees Kelvin)
C	—	constant

In this regard, Drefhal, et al. (1984) reviewed the literature and suggested that a value of 20 is suitable for the constant C. Further, based on their own and other experimental data, they plotted a graph between P and the stress σ. From this plot, it is possible to estimate the time to

rupture for a given stress. One has to be extremely careful using these plots, because they can vary depending on the microstructure (e.g., grain size) and the deformation mechanisms which may be different depending on the microstructure, stress level, and the temperature.

2.2 Creep Deformation Mechanisms

2.2.1 Slip

Slip is one of the main deformation mechanisms during low temperature deformation and creep behavior of titanium alloys. As mentioned before, the crystal structure of α titanium is hexagonal close packed with a c/a[1] ratio of 1.587; slip occurs primarily on the prism planes and to a lesser extent on the basal and pyramidal planes with $\langle 11\bar{2}0 \rangle$ as the slip direction.[2] Slip in α titanium, in general, is very fine as in the case of an α Ti-0.4 wt% Mn alloy with an oxygen content of 0.071 wt% as shown in Figure 2-3 (Ankem, et al., 1994). Comparison of Figure 2-3(a) with Figure 2-3(b) indicates that the grid lines are elongated in the direction of loading (which is in the vertical direction in the figure) after a creep strain of 0.61 percent. Because there are no visible slip lines in Figure 2-3(b), this indicates fine slip. Fine slip is also observed during room temperature creep of the α titanium alloy Ti-1.6wt%V with an oxygen content of 0.069 wt% (Aiyanger, et al., 2005), and planar slip is also present during low temperature creep deformation as shown in Figure 2-4a. Ti-6Al with an oxygen content of 0.08 wt% exhibits intense planar or coarse slip as shown in Figure 2-5 (Neeraj, et al., 2000). These dislocations were identified as "a-type" ($b = \langle 11\bar{2}0 \rangle$) screw dislocations and are evidence that coarse slip is an operative deformation mechanism in Ti-6Al. As shown in Figure 2-5, the dislocations tend to be found in aligned arrays, and isolated dislocations are not typically observed. This suggests that the slip behavior can be modified by the alloying elements in the phase.

2.2.2 Twinning

As shown originally by Ankem, et al. (1994), twinning can play a significant role in low temperature creep behavior of titanium alloys. Twinning, as shown in Figure 2-3c, is an important deformation mechanism in the tensile creep of this titanium. There are four common types of deformation twins observed in titanium: $\{10\bar{1}2\}$, $\{11\bar{2}1\}$, $\{11\bar{2}2\}$, and $\{10\bar{1}1\}$[3] Slip can precede twinning as shown in Figure 2-3, and twins can start instantaneously and grow with time causing creep as shown in Figure 2-6. Twins were also observed in an α Ti-1.6 wt% V alloy during an ambient temperature, tensile creep test conducted at a stress level of 95 percent of the yield stress (Figure 2-7) (Aiyangar, et al., 2005). The grain size of this specimen was 226 μm [8,814 μin], and the twins were identified as $\{10\bar{1}1\}$ twins—the same

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[1]The ratio c/a corresponds to the ratio of the "c" axis to that of the "a" axis in the hexagonal close packed crystal structure.

[2]The brackets $\langle 11\bar{2}0 \rangle$ represents a family of crystallographic directions in the hexagonal close packed crystal structure.

[3]The brackets $\{10\bar{1}2\}$ represent a family of crystallographic planes in the hexagonal close packed crystal structure.

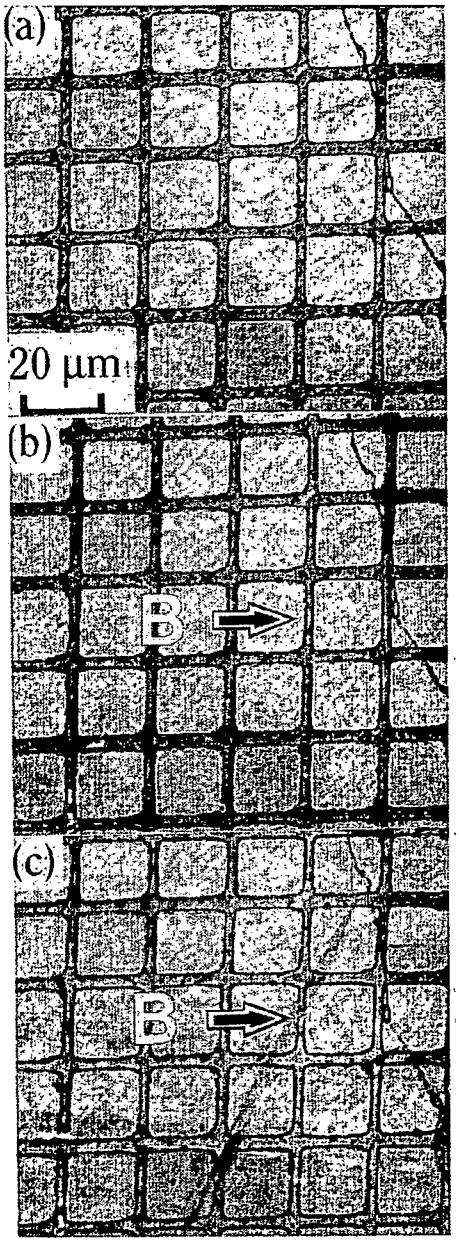

Figure 2-3. Optical Micrographs of a Different Area of Specimen B. (a) Undeformed, (b) Creep Strain = 0.61 Percent, Time = 0.96 Hour and (c) Creep Strain = 1.22 Percent, Time = 5.18 Hour. Note: Elongation of Grid Lines in Loading Direction at "B" in (b), (c) Shows Formation of Twin at "B." The Grain Size of the Specimen Was 500 µm [19,500 µin] and the Tensile Loading Direction Is Vertical (i.e., Along the Left Vertical Edge of the Micrographs (Ankem, et al., 1994) (Reproduced With Permission From Elsevier).

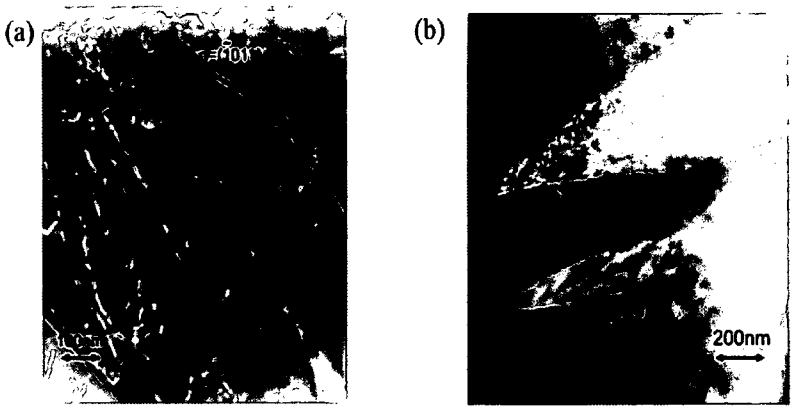

Figure 2-4. For a Large-Grained Ti-1.6V Creep Tested at 95 Percent Yield Stress: (a) Dark Field Transmission Electron Microscope Micrograph Showing Planar Slip in $\{10\bar{1}0\}$ Prism Planes. Dislocations Are $< 11\bar{2}0 >$ "a" Type Dislocations in Screw Orientation, (b) Bright Field TEM Micrograph Showing Parallel Twins (Aiyangar, et al., 2005) (Reproduced With Permission From Metallurgical Transactions).

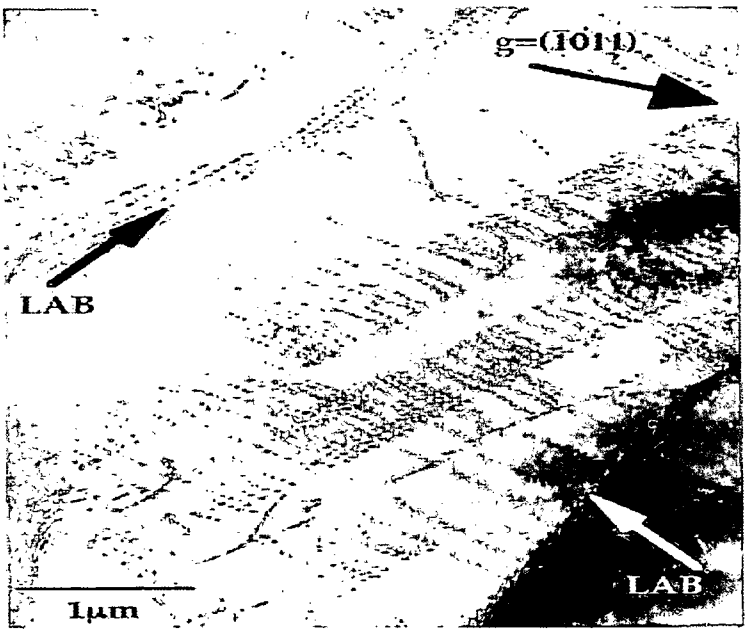

Figure 2-5. Planar Slip in $\{10\bar{1}0\}$ Prism Planes Observed in a Creep-Deformed, {σ = 716 MPa [104 ksi]}, c = 0.9 Percent) Titanium-Aluminum Sample. The Dislocations Seen Are 1/3 $< 11\bar{2}0 >$ "a" Type Dislocations Primarily in Screw Orientation. The Grain Size Was 500 μm [19,500 μin] and the Loading Was Compression (Neeraj, et al., 2000) (Reproduced With Permission From Elsevier).

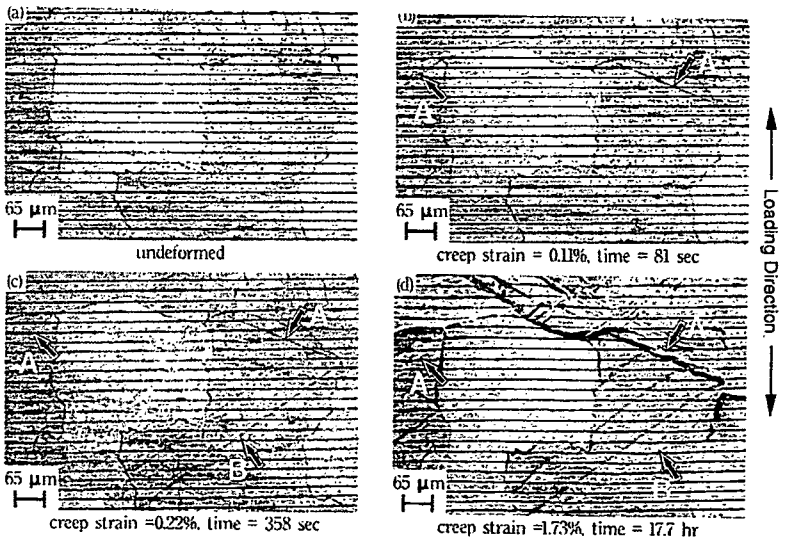

Figure 2-6. Optical Micrographs of (a) Undeformed and (b), (c), and (d) Deformed α Ti-0.4 Mn Alloy. Horizontal Lines Are Gold Fiducial Lines, (b) Shows Instantaneous Twins at "A," (c) Shows Nucleation of New Twins at "B," and (d) Shows Growth of Twins at "A" and "B" and Formation of New Twins at "C" (Ankem, et al., 1994) (Reproduced With Permission From Elsevier).

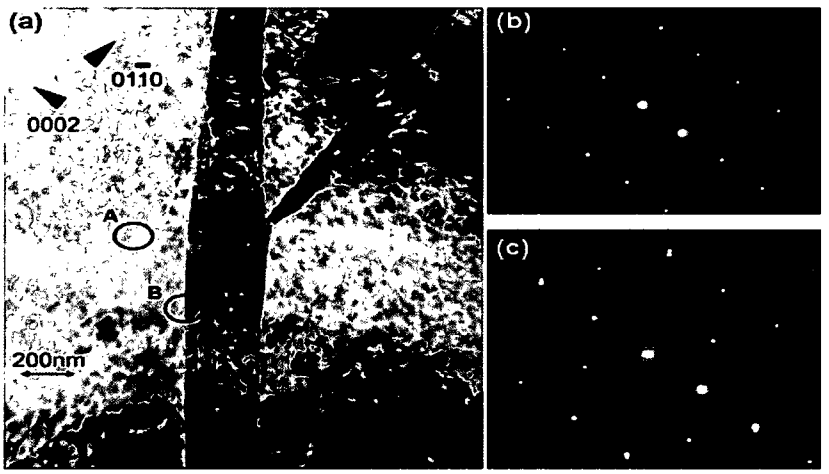

Figure 2-7. Deformation Twinning in a Large-Grained Ti-1.6V Alloy Crept to 95 Percent Yield Stress: (a) Transmission Electron Microscope Micrograph Showing Twins, (b) Selected Area Diffraction Pattern of Matrix Taken at Location "A," and (c) Selected Area Diffraction Pattern of Matrix/Twin Interface at Location "B." Note: Extra Spots Corresponding to $\{10\bar{1}1\}$ Twin (Aiyangar, et al., 2005) (Reproduced With Permission From Metallurgical Transactions).

type as those previously identified in the creep of the Ti-0.4Mn alloy (Ankem, et al., 1994), suggesting that the formation of this type of twin is characteristic of αtitanium alloys with small amounts of alloying elements. Further, the twin orientation is similar to that of the Ti-0.4Mn alloy, in which the twins in both alloys are often found parallel to one another (Figure 2-4b). Hultgren, et al. (1999) also observed twins during room temperature compression creep of an α Ti-0.4 wt Mn alloy with a grain size of 500 μm [19,500 μin]; however, the twins were found to be of the type $\{11\bar{2}2\}$. These results clearly suggest that twinning is an extremely important phenomena in coarse-grained α titanium alloys with small amounts of minor alloying elements with an oxygen content of 0.1 wt%.

2.2.3 Grain Boundary Sliding

Grain boundary sliding is an important deformation mechanism particularly at high temperatures $T > 0.25\ T_{mp}$, thus it is not expected to play a significant role at the low temperatures of interest (i.e., $T < 0.25\ T_{mp}$).

2.2.4 α-β Interface Sliding

Normally α-β interface sliding generally is considered to be a high temperature phenomena. However, by putting fiducial gridlines on the specimens, Ankem and Margolin (1983) have shown for the first time that α-β interface sliding can occur at low temperatures. α-β interface sliding was reported in Miller, et al. (1987) to enhance low temperature creep of a Ti-6Al-2Nb-1Ta-0.8Mo near-α titanium alloy. This mechanism can also occur during the creep deformation of Grade 24 alloy or its surrogates.

2.2.5 Widmanstatten Colony Boundary Sliding

Depending on the heat treatment, the microstructure of near-α and α-β titanium alloys can result in colony-type Widmanstatten microstructures, somewhat similar to the lamellar microstructure shown in Boyer, et al. (1994). Miller, et al. (1987) reported that sliding along the colony boundaries during low temperature creep deformation can result in higher creep strains and thus may exist during the low temperature creep deformation of Grade 24 or its surrogates.

2.2.6 Stress-Induced Martensite

Depending on the temperature from which the titanium alloys are cooled and the rate of cooling, the β of the α-β alloys can be in the metastable state. If the metastable β phase is highly unstable, then this phase can transform under stress and result in creep. For example, in an α-β Ti-8.1 wt% V alloy, α deforms either by slip or twinning, but the metastable β deforms by stress-induced martensite as shown in Figure 2-8.[4] Such stress-induced transformations can occur in β phase of Grade 24, or its surrogates, during low temperature creep.

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[4]Jaworski, A. and S. Ankem. "Influence of the Second Phase on the Room Temperature Tensile and Creep Mechanisms of Alpha-Beta Titanium Alloys: Part II-Creep Deformation." Accepted for Publication. 2005.

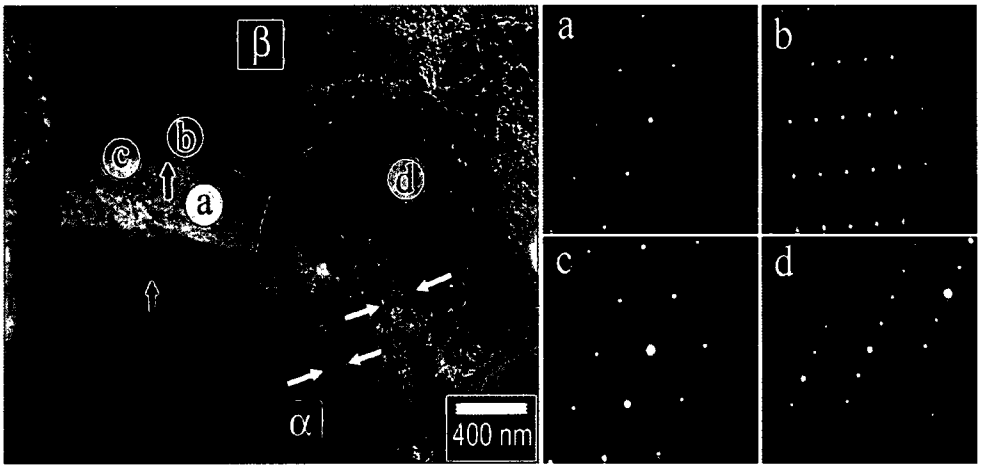


Figure 2-8. Bright Field Transmission Electron Microscope Micrograph of Stress-Induced Hexagonal Martensite (α′) Plates in the α Phase and Twins in the β Phase of Ti-8.1V, Indicated by Arrows. To the Right, Selected Area Diffraction Patterns Taken From (a) β Phase (b) α′ Plate (c) Across the α′/β Interface and (d) α Phase. The α′ Plate and the α Phase Both Have a $\{10\bar{1}1\}$ Twin Relationship. Zone Axis Is $[1\bar{2}10]_\alpha$ / /$[1\bar{1}1]_\beta$ (Reproduced With Permission From Metallurgical Transactions).

2.3 Factors Affecting Creep Deformation

2.3.1 Chemistry

The chemical composition of the alloy determines how much of the alloy contains α and β phases, their morphologies, and whether or not the phases are stable. The α stabilizers raise and the β stabilizers reduce the β-transus temperature. In addition, in a given phase, the alloying elements can alter the deformation mechanisms. For example, Paton, et al. (1973) reported that the addition of aluminum to titanium in significant quantities reduces the propensity for twinning in the α phase. This may explain why twinning was not observed by Neeraj, et al. (2000) for a Ti-6wt.% Al alloy, although extensive twinning was reported in Hultgren, et al. (1999) for another α alloy (without aluminum) under similar conditions. However, it is not clear at this time if twinning occurs in the α phase of the α-β Ti- 6Al-4V alloy.

Similar to aluminum, oxygen is also an α stabilizer. Oxygen affects the deformation mechanism of low temperature creep behavior of α-titanium or β-titanium. The effect of oxygen on twinning is shown in Figure 2-9.[5] Note that a titanium alloy with an oxygen concentration of 2,500 ppm in weight is considered as commercial purity titanium (Figure 2-9) and approximately corresponds to the oxygen content in Grade 7. Figure 2-9 also shows that increased oxygen content from 250 to 2,500 ppm significantly reduces the extent of twinning at lower temperatures, such as

[5]Ibid.

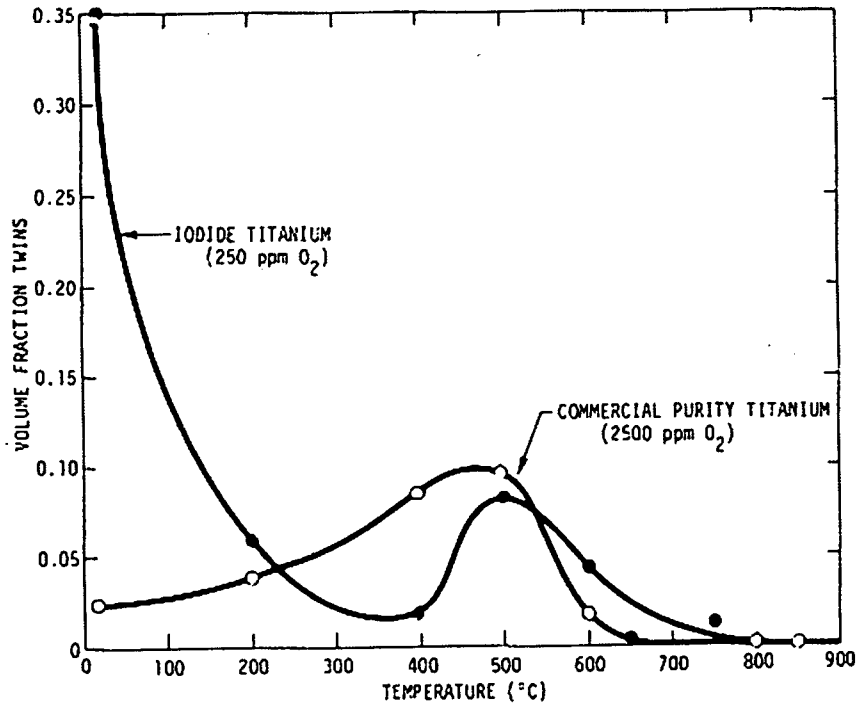


Figure 2-9. Influence of Deformation Temperature on the Volume Fraction of Twins After 5 Percent Strain in Titanium of Two Different Purities (Oxygen Concentration Weight Parts Per Million) (Paton, et al., 1973) (Reproduced with Permission From Springer)

room temperature, but at 200 °C [392 °F] the twinning is similar for both oxygen contents. It is not clear at this time the effect of twinning during creep of Grade 7 titanium at different temperatures of interest ($T < 0.25T_{mp}$), and twinning particularly needs to be studied for large-grained Grade 7 Titanium.

Another interstitial element of interest is hydrogen, which is a β-stabilizer. There is considerable interest in the effect of hydrogen on low temperature creep behavior as well as its effect on sustained load cracking. Of interest here is whether dissolved hydrogen can affect the low temperature creep of titanium alloys. In this regard, Gao and Dexter (1987) have shown that an increase in hydrogen content in the Ti-6Al-4V alloy increases creep strain as shown in Figure 2-10, attributed to hydrogen-induced softening and embrittlement at different stages of deformation. Gao and Dexter (1987) have suggested that at the beginning of deformation, the dissolved hydrogen softens metals and alloys due to a decrease in activation volume and an increase in dislocation mobility. Significant localized dislocation motion at the beginning of deformation is due to a "sweeping" effect, thereby causing hydrogen embrittlement in a later stage of deformation. There are other impurity elements such as carbon and nitrogen, but normally their concentrations are similar in different alloys and are well controlled.

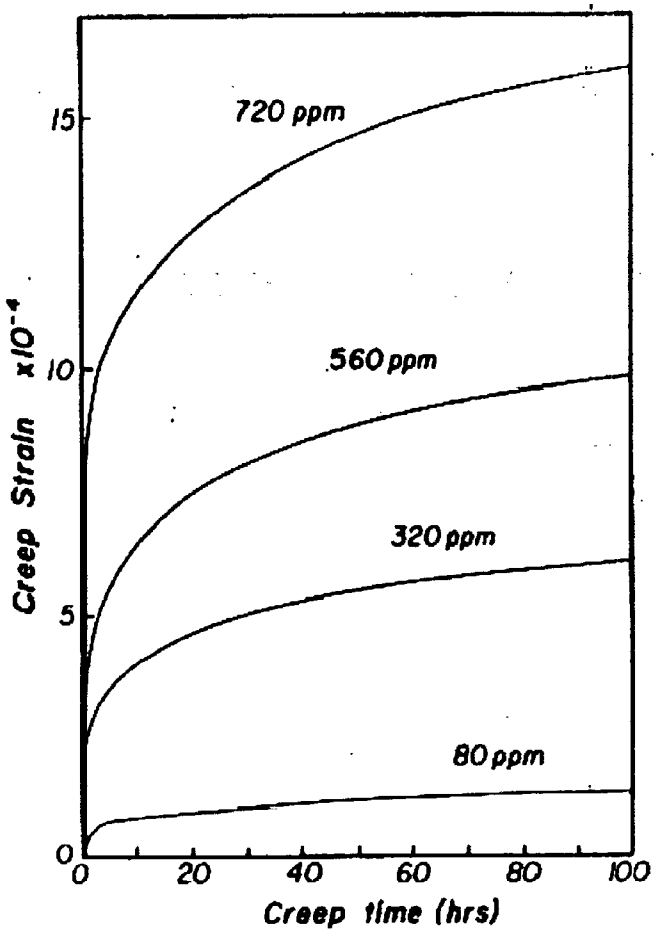


Figure 2-10. Room Temperature Creep Strain Versus Creep Time Curves of Ti-6Al-4V Alloy Containing Different Hydrogen Concentrations at a Stress of 72 kg/mm^2 [706 MPa] [102 ksi] (Gao and Dexter, 1987) Note: ppm Denotes Parts Per Million (Reproduced With Permission From Metallurgical Transactions)

2.3.2 Microstructure

For a given single phase alloy such as Grade 7 or its surrogates, the grain size can significantly vary and the mechanical properties, including low temperature creep, strongly depend on grain size. For example, Ankem, et al. (1994) and Aiyangar, et al. (2005) have shown that at a given normalized stress, the α titanium alloys with a larger grain size due to twinning creep more than those with a smaller grain size. For α-β alloys such as Grade 24 and its surrogates, the microstructure can vary significantly due to heat treatments (Boyer, et al., 1994). If the Ti-6Al-4V alloy is fast cooled from a temperature above the β-transus, the resultant microstructure consists of α′, and as shown in Figure 2-11, the martensite can twin during deformation. It is not clear at this time whether the Ti-6Al-4V alloy creeps more with the martensitic structure, which can twin. Note that these types of martensitic structures are

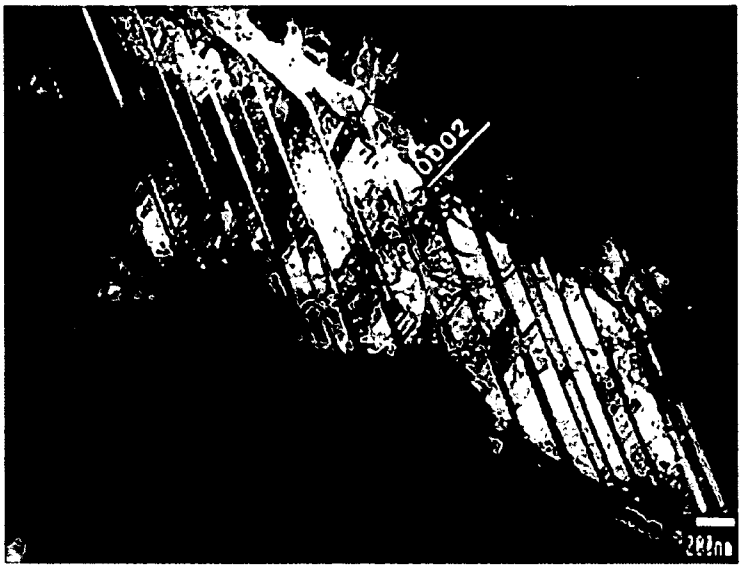


Figure 2-11. Transmission Electron Microscope Micrograph Showing Large {10$\bar{1}$1} Twins Within the α′ Martensitic Plates (Manero, et al., 2000) (Reproduced with Permission of Elsevier)

possible when large structures are welded. The effect of the microstructure of this alloy on low temperature creep is discussed further in later sections.

2.3.3 Texture

As demonstrated above, titanium alloys deform either by slip or twinning during creep deformation and depend on the type of loading on specific planes and directions. It is possible to process titanium alloys so that most of the grains can be oriented in a specific crystallographic direction; that is, the alloy has a preferred orientation or texture. Anisotropic materials such as titanium and its alloys are prone to texture, particularly in sheets and wires as compared to plates. Texture, which can affect the mechanical properties including low temperature creep, can be reduced or eliminated through thermo-mechanical processing.

2.3.4 Stress Level and Proportional Limit

The extent of creep deformation depends on stress level. In the studies of low temperature creep deformation, the applied stress is expressed in terms of the 0.2 percent yield stress. Unfortunately, this does not relate to the proportional limit where the plastic deformation starts. Thus, it is quite possible that two alloys may have the same yield stress but different proportional limits. In addition, proportional limit is usually not reported because it is difficult to measure. Note that the proportional limit depends on local stresses, particularly the interaction stresses. These interaction stresses are the subject of the following section.

2.3.5 Elastic Interaction Stresses

Due to compatibility requirements at grain boundaries for single phase materials and at phase interfaces for multiphase materials, elastic interaction stresses develop and can be either positive or negative. The effective stress for initiation of plastic deformation can be written as

$$\tau_{effective} = \tau_{applied} + \tau_{interaction} \tag{2-11}$$

When $\tau_{effective}$ reaches a critical value for either slip or twinning, plastic deformation occurs. In general, the elastic modulus of α titanium is greater than that of a β phase. Therefore, due to elastic interactions, the effective stress in the α phase near an α-β interface is higher. When the effective stress reaches a critical stress, plastic deformation initiates even though the applied stress may lower than the critical resolve of either α or β phases. The extent of elastic interaction stresses strongly depends on the morphology of the α and β phases. For example, Greene and Ankem (1995) have calculated elastic interaction stresses for α-β titanium-manganese alloys. Their results show that, due to elastic interactions, the stress in the α phase near the interface is higher, and plastic deformation can initiate at an applied stress much lower than the applied stress necessary to cause plasticity in an alloy having only an α phase.

2.3.6 Stress Type

Creep tests are performed under constant load and given that the dimensions change during creep, the true stress is different from the engineering stress which is based on the original cross-sectional area. Based on the assumption that the volume of the material remains constant during deformation, it can be shown that (Dieter, 1986, p. 439)

$$\sigma_T = \sigma_e(1+e) \tag{2-12}$$

where

σ_T — true stress
σ_e — engineering stress
e — engineering strain

After creep deformation starts, the true stress increases in tension and decreases in compression. Hultgren, et al. (1999) has suggested this is one reason why the time exponent n values were lower for compression loading as compared to tensile loading during room temperature creep of an α Ti-0.4 wt% Mn alloy.

2.3.7 Cold Work

Depending on the thermo-mechanical processing, a significant amount of cold work may be left in the material in the form of higher dislocation density and internal stresses. This is particularly true in the case of cold-worked sheet materials. If the material is not annealed, at least for commercially pure α titanium, the cold-worked material seems to have higher threshold stresses for creep and lower creep rates (Adenstedt, 1949; Kiessel and Sinnott, 1953); however, the strain to fracture is considerably reduced.

2.3.8 Creep Test Temperature

The last factor that can affect creep of titanium alloys is the test temperature. As the temperature is increased, the kinetics of various deformation mechanisms such as slip, twinning, and sliding increase, thereby increasing the creep strain rates.

3 SELECTION OF SURROGATES FOR GRADES 7 AND 24

The result of this rather extensive literature review has determined that there are no readily available published articles, in the open literature, that present information with regards to low temperature creep of Titanium Grades 7 and 24. As an alternative, surrogates for these titanium grades need to be identified. This chapter briefly discusses the bases for the appropriate selection of these surrogates.

ASTM International Designation B 381-03 indicates that Grade F7 is an unalloyed titanium with 0.12–0.25 percent palladium, and Grade F24 is titanium alloy, 6 percent aluminum and 4 percent vanadium with 0.04–0.08 percent palladium. Unalloyed titanium has four grades: 1, 2, 3, and 4 which are based on oxygen content. Note that the oxygen content as specified by different countries in the world is not necessarily the same. For the purpose of this report, only the U.S. compositions are considered. The oxygen content for Grades 1, 2, 3, and 4 are 0.1–0.18 percent, 0.2–0.25 percent, 0.3–0.35 percent, and 0.4 percent, respectively, and the oxygen content in Grade 7 is 0.2 percent. Therefore, with respect to oxygen content, Grades 7 and 2 are similar except for the difference in palladium. ASTM B 381-03 indicates that tensile properties for Grades 2 and 7 are the same and that the tensile properties for Grades 5 and 24 are the same. Therefore, the closest surrogate for Grade 7 is Grade 2 and for Grade 24 is Grade 5. The main difference between Grades 7 and 2, and Grades 24 and 5 is the palladium content. Palladium is added to improve corrosion resistance and may not have a significant effect on the general mechanical behavior when added in small quantities. Figure 3-1 shows that palladium is a β-stabilizer but has some solubility in the α phase. Therefore, additions of palladium in small amounts does not significantly alter the amounts of β phase. The microstructure of Grade 24 is expected to be similar to Grade 5, and the microstructure of Grade 7 is expected to be similar to Grade 2. However, it is not clear at this time if palladium effects the creep and fracture toughness of Grades 7 and 24.

Based on the discussion above, Grade 2 is the closest surrogate to Grade 7, and Grade 5 is the closest surrogate to Grade 24. Note that the oxygen contents in the commercially pure alloys used in the creep studies to be discussed in the following sections have oxygen contents close to Grade 1. For this reason, most of the α alloys are considered as surrogates for Grade 7. Similarly, for Grade 5, most of the following investigations deal with Ti-6Al-4V which has oxygen contents similar to Grade 24; thus, most of the α-β titanium alloys are considered as surrogates for Grade 24.

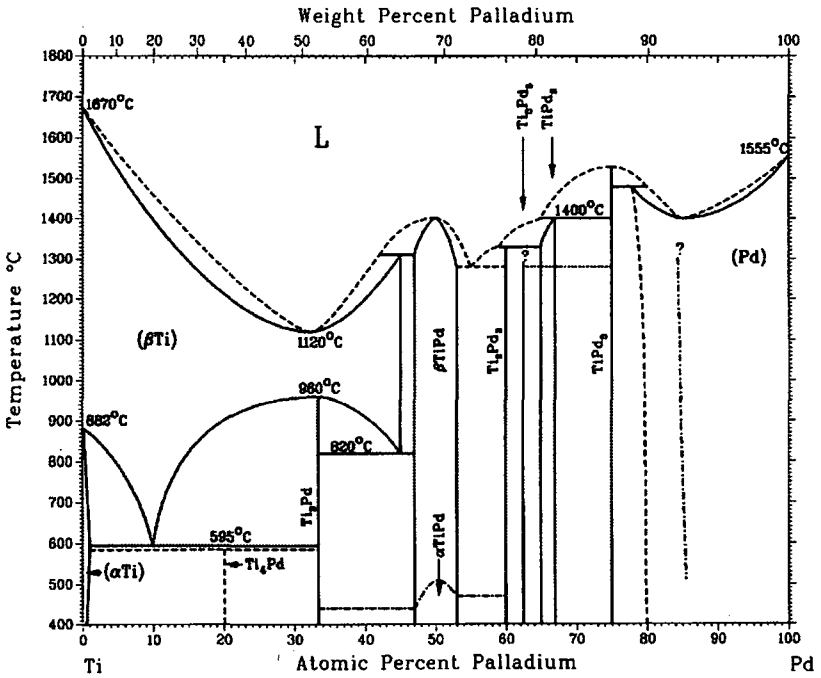


Figure 3-1. Assessed Titanium-Palladium Phase Diagram (Murray, 1987) (Reproduced With Permission From ASM International)

4 LOW TEMPERATURE CREEP OF GRADES 7 AND 24 SURROGATES

The previous chapter discussed the rational for choosing the appropriate surrogates for Titanium Grades 7 and 24. This chapter presents the low temperature creep behavior for the chosen titanium surrogates. A general discussion of creep for each surrogate is given as related to microstructural quantities (e.g., grain size). Finally, based on the creep data available, predictions of creep strains corresponding to a time of 10,000 years are made.

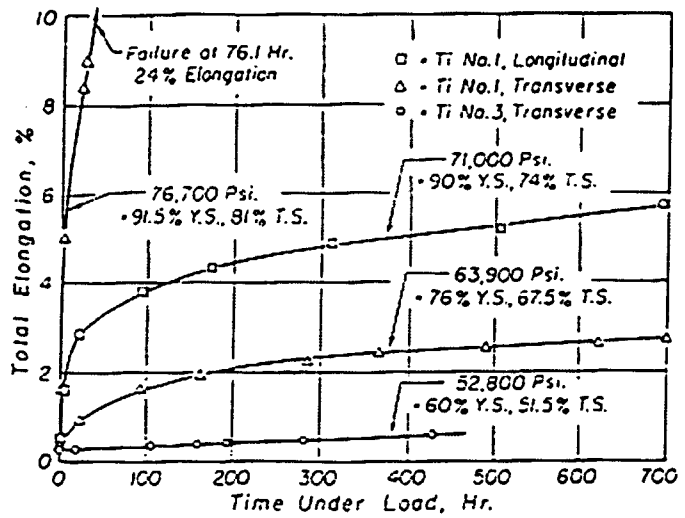
4.1 Creep of Grade 7 Surrogates

4.1.1 General

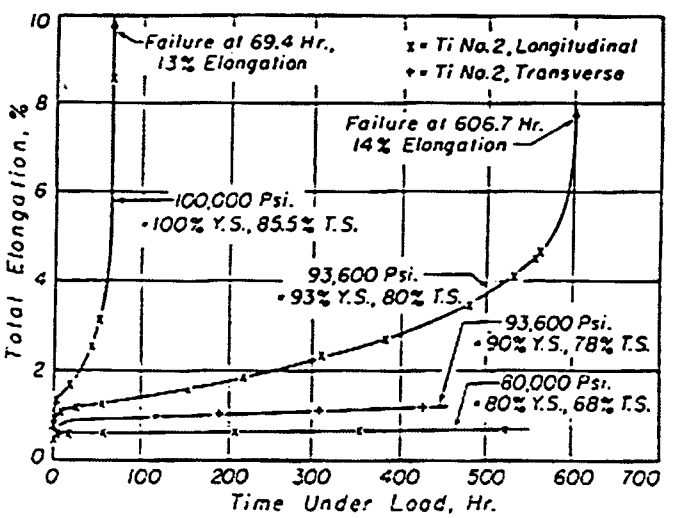
Adenstedt (1949) initiated work on room temperature creep behavior of titanium alloys. Most earlier studies (Adenstedt, 1949; Kiessel and Sinnott, 1953) dealt with commercially pure titanium sheets. The results of Adenstedt (1949) are shown in Figure 4-1 and indicate that room temperature creep depends on cold work and that creep can occur at stresses as low as 60 percent yield stress, at least in sheet materials. The oxygen content was not reported, but it is stated that the titanium was a commercially pure titanium alloy. It is not clear what the grain size of this alloy was, but given that it is a sheet material, the grain size is expected to be small.

As shown in Figure 4-2, the work of Kiessel and Sinnott (1953) also suggests that prior cold work on commercially pure titanium alloy sheets increases creep resistance. It was further shown that recovery can take place in these materials at room temperature depending on the amount of cold work. Minimum secondary creep rates at a given stress were found to increase with an increase in temperature but not in a linear fashion (Kiessel and Sinnott, 1953). Most of the recent studies did not report minimum creep rates, because most of the creep observed was primary or stage 1 creep. Therefore, it appears that the observation of secondary creep is due to small sheet thickness and heavy cold work in the earlier investigations.

Drefahl, et al. (1984) conducted low temperature creep tests for as long as 27 years at a temperature of 20 °C [68 °F] on commercially pure titanium. These tests were conducted on 8 mm [0.3 in] bars with constant loads, and although not clear, appear to be tensile. Some of the results are shown in Figures 4-3 and 4-4. These properties correspond to a commercially pure titanium with an oxygen content of 0.077 wt% and a grain size of 1,000 μm [39×10^3 μin]. The yield strength of the alloy used in Drefahl, et al. (1984) is not given, but based on the oxygen content, should be close to that of Grade 1. Based on ASTM standards (ASTM B 381-03), the minimum yield strength is 170 MPa [25 ksi]. Therefore, the stress levels shown in Figures 4-3 and 4-4 are near or slightly below the yield stress. Figure 4-3 indicates that the creep strain decreases as the applied stress is decreased. Drefahl, et al. (1984) have indicated that the creep curves followed a logarithmic equation of the form given in Eq. (2-9) up to a creep strain of 1 percent and a parabolic equation of the form given in Eq. (2-1) for a creep strain range of 1–15 percent. The time exponents (n values) used in the equations are in the range 0.16–0.19. Drefahl, et al. (1984) indicated that the variation was within the experimental error and believed that these n values are constant for various stress levels; however, the A values increased with an increase in stress level as observed by other investigators. Drefahl, et al. (1984) also studied other commercially pure titanium alloys by simultaneously varying the composition and grain size and concluded that grain size may not have a significant effect on low temperature creep. This conclusion is highly questionable because oxygen



(a)



(b)

Figure 4-1. Creep Curves for Commercially Pure Titanium at Room Temperature (a) Annealed Sheet and (b) Cold-Rolled Sheet (Adenstedt, 1949) (Reproduced With Permission From ASM International)

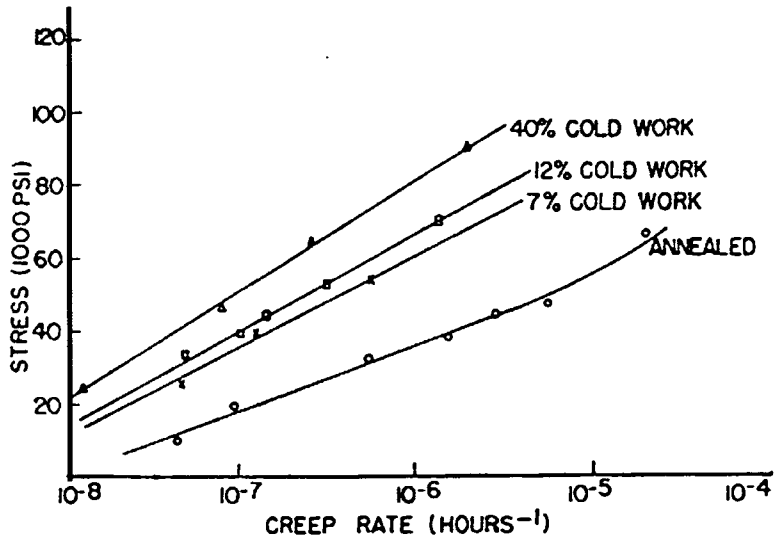


Figure 4-2. Secondary Stage Creep Rates at Room Temperature in Commercially Pure Titanium in Various Metallurgical States (Kiessel and Sinnott, 1953) (Reproduced With Permission From The Minerals, Metals & Materials Society)

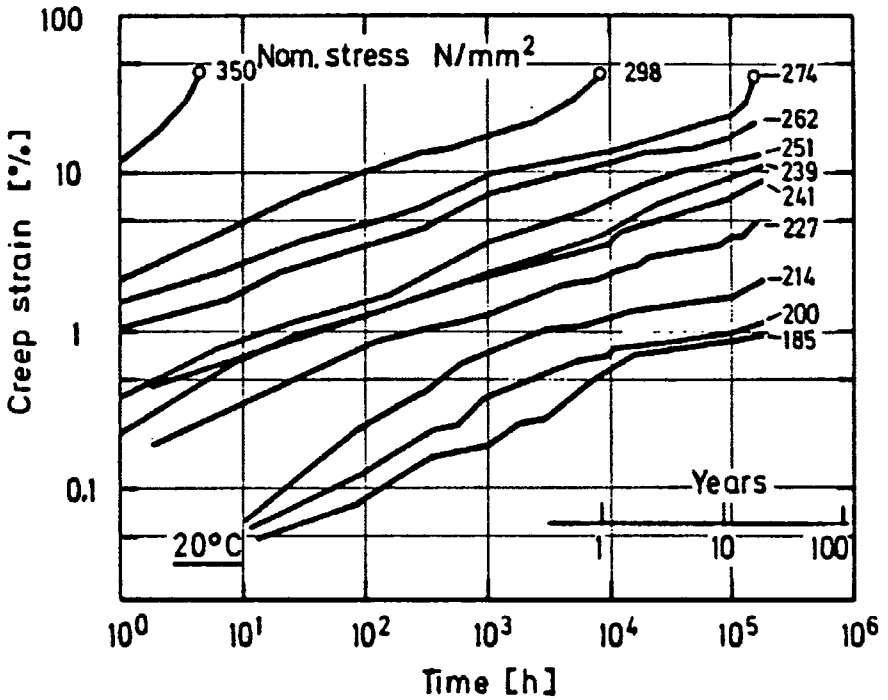


Figure 4-3. Creep Curves for Commercially Pure Titanium at 20 °C [68 °F] (Drefahl, et al., 1984) (Reproduced With Permission From Deutsche Gesellschaft für Metalllkunde)

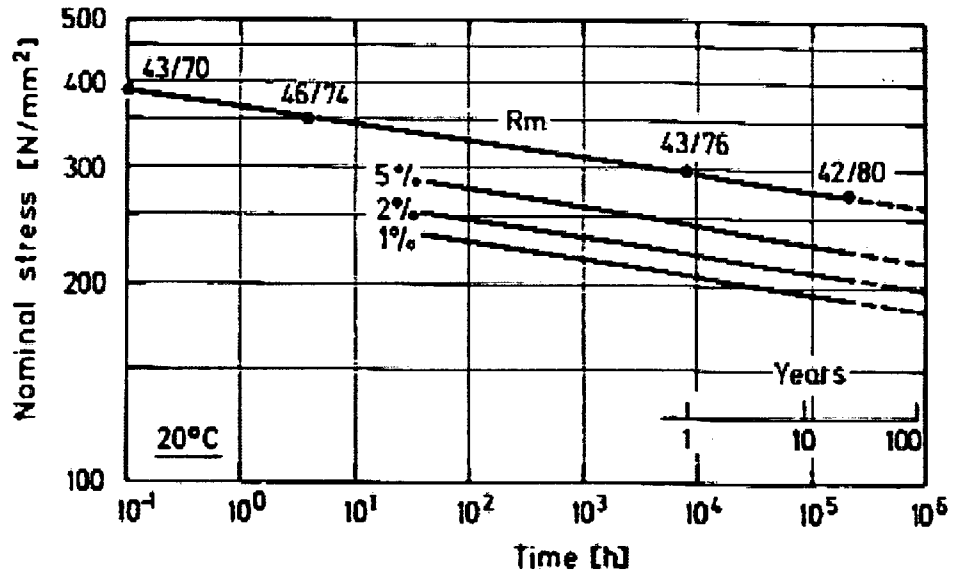


Figure 4-4. Creep Rupture Strength (Curve Marked Rm) and Strain Limits (Curve Marked 5 Percent, 2 Percent, and 1 Percent) for Commercially-Pure Titanium at 20 °C [68 °F]. The Figures on the Rupture Curve Represent Percent Elongation/Percent Reduction in Area (Drefahl, et al., 1984) (Reproduced With Permission of Deutsche Gesellschaft für Metalllkunde)

content can also affect creep behavior as demonstrated by Oberson and Ankem (2005) in β titanium alloys. Moreover, systematic studies (Ankem, et al., 1994; Greene and Ankem, 1993; Greene, et al., 1995; Hultgren, et al., 1999; Doriaswamy and Ankem, 2003; Aiyangar, et al., 2005) clearly demonstrate that grain size has a significant effect on low temperature creep behavior of titanium alloys.

Ankem, et al. (1994) and Greene, et al. (1995) carried out systematic studies on the room temperature creep behavior of Ti-0.4 wt% Mn alloy. The tests were carried out on round bar specimens loaded in tension and with a grain size from 26 to 500 μm [1,014 to 19,500 μin]. The oxygen content of the alloys was 0.071 wt%, which is close to Grade 1 commercially pure titanium alloy oxygen content. Ankem, et al. (1994) and Greene, et al. (1995) performed these tests at a stress level of 95 percent of the yield stress. Some of the results of Ankem, et al. (1994) are shown in Figure 4-5.

The results show that grain size has a significant effect on creep and can be represented by Eq. (2-1) in which both constants A and n vary with grain size

$$A = (0.1)d^{0.29} \tag{4-1}$$

$$n = (0.1)d^{0.17} \tag{4-2}$$

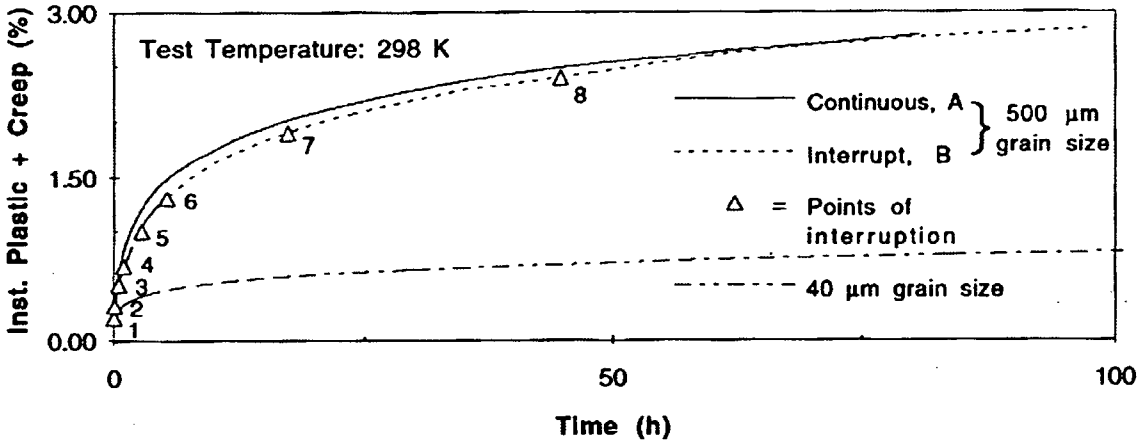


Figure 4-5. Instantaneous Plastic Plus Creep Strain Versus Time for α Ti-0.4Mn Alloy. Creep Stress: 95 Percent Yield Stress. The Triangles and the Associated Numbers Correspond to Test Interruptions (Ankem, et al., 1994) (Reproduced With Permission of Elsevier)

where

d — grain size in µm

Ankem, et al. (1994) and Greene, et al. (1995) have attributed the effect of grain size to increased twinning activity as shown in Figures 2-3 and 2-6.

The effect of grain size and stress level on creep was recently confirmed by Aiyangar, et al. (2005) for room temperature creep tests on round bar specimens loaded in tension for a Ti-1.6 wt% V alloy with an oxygen content of 0.069 wt%. This oxygen content is close to that of Grade 1 commercially pure titanium. Aiyangar, et al. (2005) measured the effect of grain size at a creep stress level of 90 percent of the yield stress and for a single grain size of 226 µm [8,814 µin] loaded at various stress levels. The results are shown in Figures 4-6 to 4-8 and indicate that both grain size and stress level have a strong effect on creep, confirming the earlier results of Ankem, et al. (1994) and Greene, et al. (1995) on Ti-0.4% Mn alloy. Aiyangar, et al. (2005) attributed the lower creep resistance of coarse-grained material to the increased twinning activity.

Several investigators have conducted low temperature creep studies on highly alloyed α (Thompson and Odegard, 1973; Neeraj, et al., 2000) and near-α alloys (Miller, et al., 1987). Some of the results of Thompson and Odegard (1973) on low temperature creep behavior of Ti-5 Al- 2.5 Sn alloy are shown in Figure 4-9. These results clearly show that processing conditions (and their corresponding microstructure), stress level, and temperature have an effect on low temperature creep. For a given microstructure, the creep strain increased with an increase in stress level and temperature as expected. However, the results have shown that the time exponents (i.e., *n* values) are not consistent. Equation (2-1) reveals that for a given microstructure and stress level, if the *n* value is lower at a higher temperature, then there is always a time *t* above which the creep strain will be lower at a higher temperature irrespective of the *A* values. This appears incorrect barring any changes with temperature such as phase transformations. Similarly, for a given microstructure and temperature, if the *n* value in Eq. (2-1)

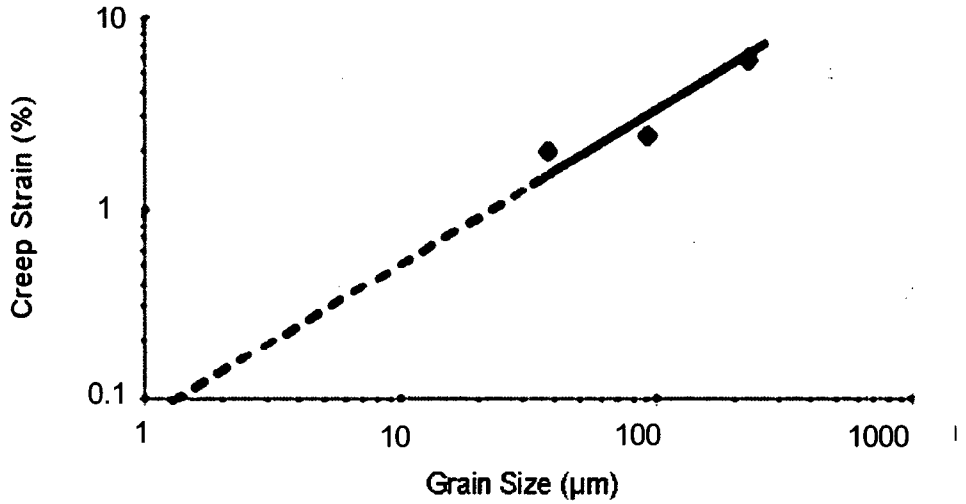


Figure 4-6. (a) Creep Strain Versus Time at Various Stress Levels for a Ti-1.6V Alloy at Grain Size of 226 μm [8,814 μin] and (b) Creep Strain Versus Time for α Ti-1.6V Alloy for Various Grain Sizes at Stress Level of 90 Percent Yield Stress (Aiyangar, et al., 2005) (Reproduced With Permission From Metallurgical Transactions)

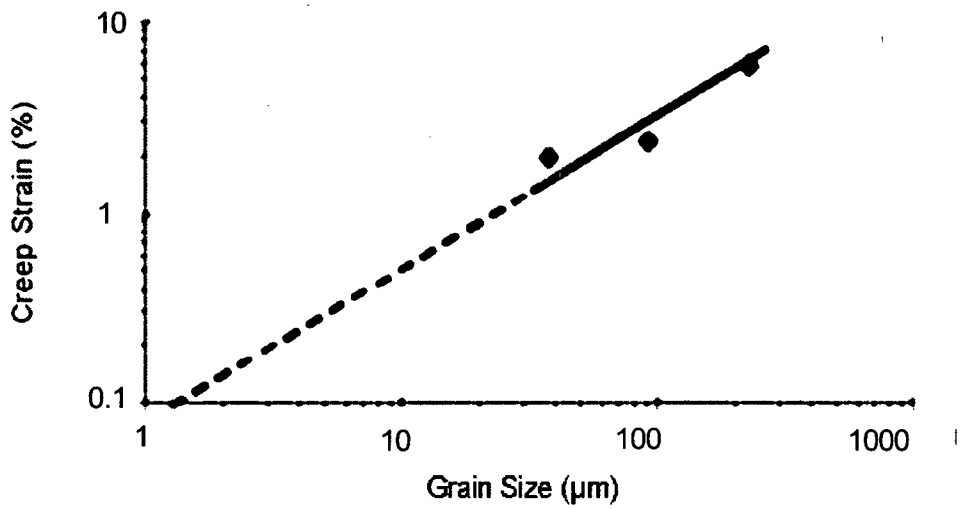


Figure 4-7. For Ti-1.6V Alloy, Creep Strain Versus Grain Size at 90 Percent Yield Stress at 25 °C [77 °F] for 200 Hours (Aiyangar, et al., 2005) (Reproduced With Permission From Metallurgical Transactions)

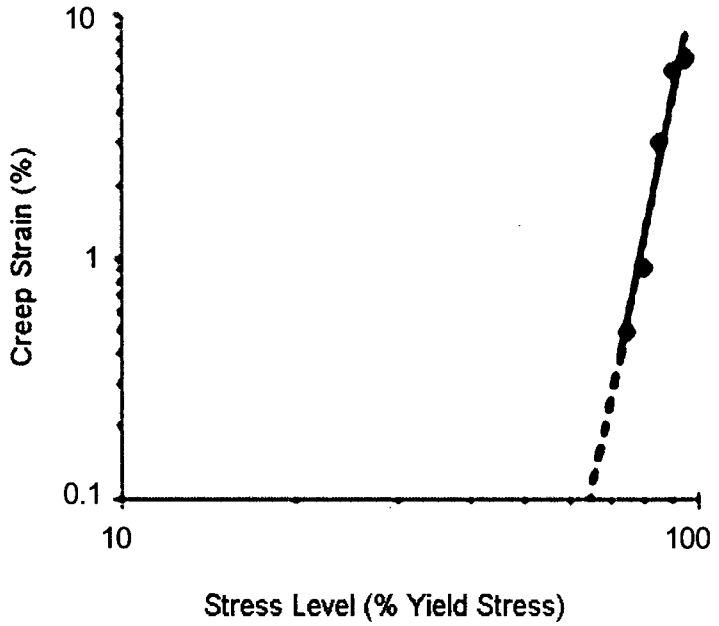


Figure 4-8. For Ti-1.6V Alloy, Creep Strain Versus Stress Level for 226 μm [8,814 μin] Grain Size After 200 Hours (Aiyangar, et al., 2005) (Reproduced With Permission From Metallurgical Transactions)

is higher at a lower stress, then there is a time *t* after which the creep strain is higher at a lower stress even though the corresponding *A* value at the lower stress is lower. Therefore, sound judgment must be applied when using the empirical equations, which were obtained from short term creep tests, to predict long-term behavior as required for the drip shield application. If Eq. (2-1) does not give meaningful values for the constants, then either Eq. (2-9) or a combination of these two equations should be employed.

Thompson and Odegard (1973) also reported creep deformation at stresses as low as 40 percent of the yield stress but cautioned that these measurements were close to the resolution of their experimental measurements.

Miller, et al. (1987) studied low temperature creep of a near-α Ti-6Al-2Nb-1Ta-0.8Mo alloy. Figure 4-10 shows that room temperature creep strongly depends on the microstructure. In addition, for a given microstructure, the cyclic creep strain is higher than the static creep strain which is attributed to recovery processes during unloading.

Adenstedt (1949) observed room temperature creep at 60 percent of the yield stress in commercially pure titanium sheets. Given that the materials to be used in the drip shield most probably are plates, the grain size is expected to be slightly larger than those in sheets. The net elastic interaction stress gradients in large-grained materials are expected to be larger than those in smaller grain sizes, and therefore, the proportional limit is expected to be lower. The

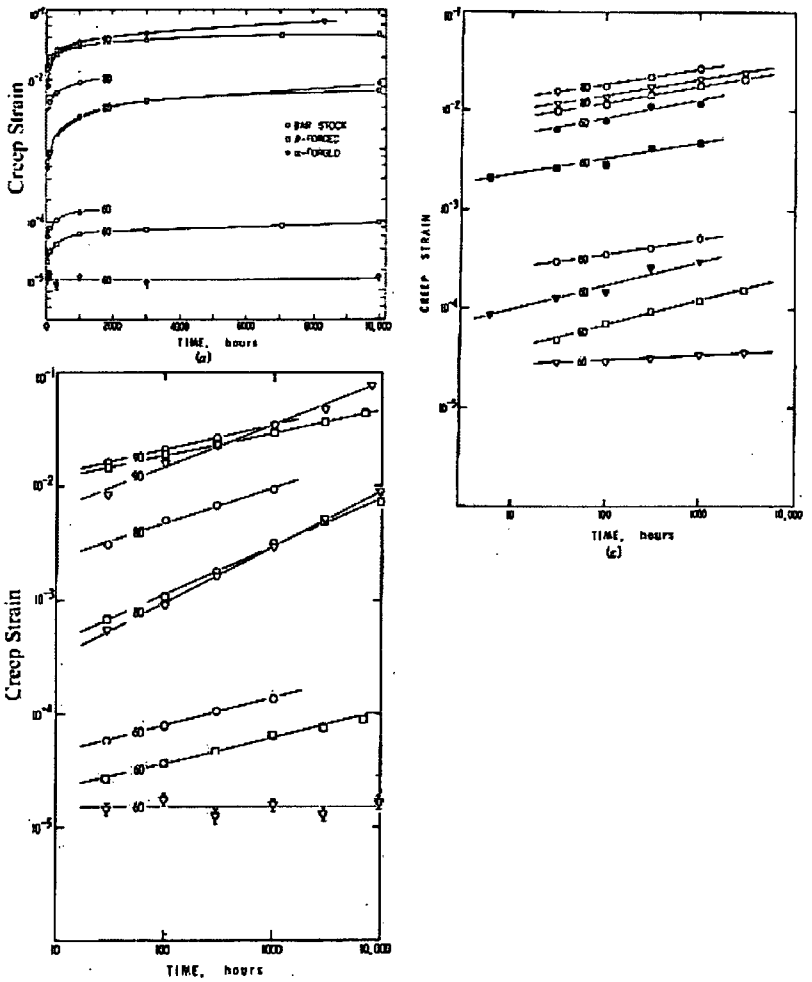


Figure 4-9. (a) Creep Curves for Ti-5Al-2.5Sn in Three Metallurgical States (Bar Stock, β-Forged, α-Forged). Curves Are Labeled With the Creep Stress as a Percent of Yield Stress. (a) Tests at 26 °C [79 °F], (b) Log-Log Plot for Tests at 26 °C [79 °F], (c) Log-Log Plot for Tests at 66 °C [151 °F] (Open Symbols) and 149 °C [300 °F] (Filled Symbols) (Thompson and Odegard, 1973) (Reproduced With Permission From Metallurgical Transactions).

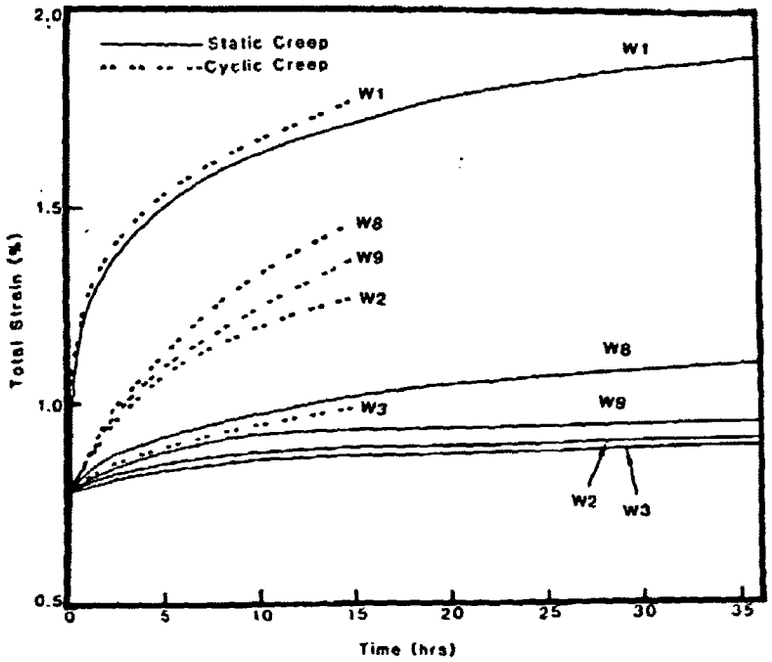


Figure 4-10. Comparison of Cyclic Creep With Static Creep at Room Temperature in Ti-6Al-2Nb-1Ta-0.8Mo (W1, W2, etc., Refer to Different α-β Microstructures) σ = 684 MPa [100 ksi] for Static Creep and σ = 304 MPa [44 ksi] (Stress Amplitudes) for Cyclic Creep (Miller, et al., 1987) (Reproduced With Permission From Metallurgical Transactions)

recent work of Aiyangar, et al. (2005) suggests that this can lead to threshold stresses slightly lower than 60 percent yield stress. The results shown in Figure 4-8 are replotted in Figure 4-11 to include lower strains and stresses. If it is assumed that a strain of 10^{-4} percent is the lowest measurable strain, then the corresponding threshold stress is approximately 35 percent of the yield stress. Note that the grain size of the drip shield plates may be smaller, the oxygen content of Grade 7 is larger than that in this α titanium alloys, and both of these factors can increase the threshold stress. Therefore, the threshold stress for Grade 7 is expected to range from 35 to 60 percent of the yield stress. Experiments are required on either Grade 2 or Grade 7 commercially pure titanium to determine the threshold stress.

4.1.2 Sample Predictions of Creep Strains After 10,000 Years

Neuberger, et al. (2002) have assembled various empirical constants for low temperature creep of Grade 7 surrogates which are shown in Table 4-1. Note that this table does not include the results of Aiyangar, et al. (2005); however, Tables 4-2 and 4-3 show constants from their most recent results. As discussed before, the time exponent *n* values have to be carefully selected, and one should not simply average the values obtained by various investigators.

The grain size of the drip shield plates is not known, but if it is assumed that the grain size is 20 μm [780 μin], then using Eq. (4-1), (4-2), and (2-1), the creep strain in 10,000 years at

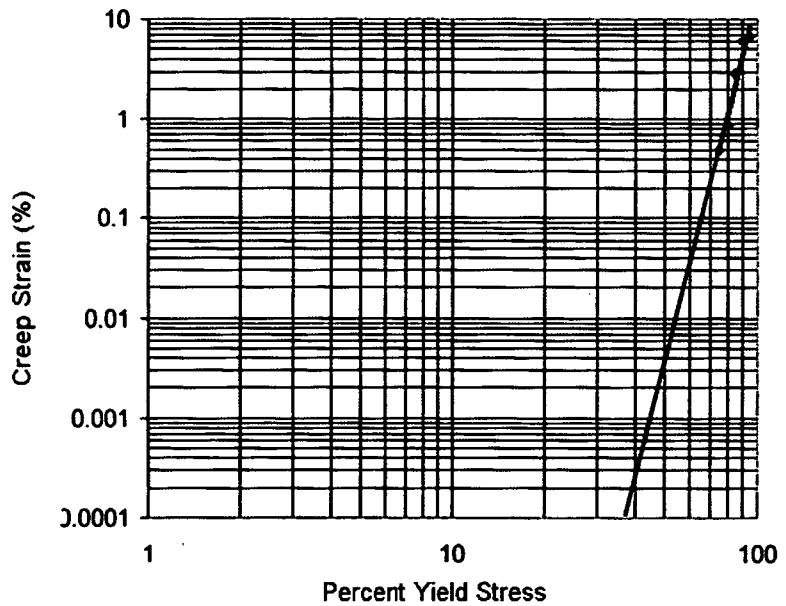


Figure 4-11. A Replot of Data Shown in Figure 4-8 (Aiyangar, et al., 2005). Room Temperature Creep Strain Versus Percent Yield Stress (YS). The Creep Strain Corresponds to 200 Hours for a Ti-1.6 wt.% V Alloy With a Grain Size of 226 μm [8,814 μin].

95 percent yield stress is approximately 5 percent at room temperature for a Ti-0.4% Mn alloy (Ankem, et al., 1994; Dieter, 1986, p. 439). However, a similar calculation performed using the data in Table 4-2 for the Ti-1.6% V alloy (Aiyangar, et al., 2005) predicts a creep strain of approximately 81 percent in 10,000 years at a stress level of only 90 percent yield stress. Furthermore, at higher temperatures {e.g., 150 °C [302 °F]} as suggested by Neuberger, et al. (2002), the *A* values are approximately doubled, which lead to double the creep strains as compared to room temperature. Based on ASTM B 381-03 standards for Grade 2, the minimum total elongation is 20 percent. That means the uniform elongation after which instability develops is even lower than 20 percent. Therefore, barring further studies on Grade 2 or Grade 7 which might show otherwise, it is reasonable to suggest that Grade 7 is vulnerable to creep failures when operating stresses are close to the yield stress.

4.2 Creep of Grade 24 Surrogates

4.2.1 General

A number of investigators (Drefahl, et al., 1984; Gao and Dexter, 1987; Imam and Gilmore, 1979; Odegard and Thompson, 1974; Reimann, 1971) have carried out low temperature, particularly room temperature, creep studies on Ti-6Al-4V alloy. Reimann (1971) conducted room temperature creep studies on Ti-6Al-4V alloy both in tension and torsional loading conditions. Some of the results related to torsional loading are shown in Figure 4-12. At higher

Table 4-1. Creep Power Law Coefficients for Alpha and Near-Alpha Titanium Alloys* (Reproduced With Permission of Material Research Society and Neuberger)

Titanium Alloy	Grain Size (μm) [μin]	Percent Yield Stress Tested	*A*	*n*
Ti-0.4 Mn†	500 [19,500]	95	1.01	0.24
Ti-0.4 Mn Compressive Stress‡	500 [19,500]	95	1.29	0.1
Ti-0.4 Mn§	72 [2,808]	95	0.65	0.21
Ti-0.4 Mn‖	40 [1,560]	95	0.34	0.19
Ti-0.4 Mn§	26 [1,014]	95	0.39	0.18
Ti-5Al-2.5Sn¶	50 [1,950]	80	0.11	0.31
Ti-5Al-2.5Sn {66 °C [151 °F]}¶	50 [1,950]	80	0.9	0.15
Ti-6Al#	500 [19,500]	94	0.41	0.24
Ti-6211**	—	92	0.26	0.18
Ti-6211††	45 [1,765]	85	1.02	0.16
Ti-6211 {180 °C [356 °F]}††	45 [1,765]	85	2.48	0.13

*Neuberger, B., C. Greene, and D. Gute. "Creep Analyses of Titanium Drip Shield Subjected to Rockfall Static Loads in the Proposed Geologic Repository at Yucca Mountain." Proceedings of the Materials Research Society. Symposium Proceedings 716. pp. JJ11.7.1–JJ11.7.8. 2002.

†Ankem, S., C.A. Greene, and S. Singh. "Time Dependent Twinning During Ambient Temperature Creep of Alpha Ti-Mn Alloys." *Scripta Materialia*. Vol. 30, No. 6. pp. 803–808. 1994.

‡Ankem, S., C.A. Green, A.K. Aiyangar. "Recent Developments in Growth Kinetics of Deformation Twins in Bulk Metallic Materials." Proceedings of the Symposium on Advances in Twinning. S. Ankem and C.S. Parde, eds. Warrendale, Pennsylvania: The Minerals, Metals & Materials Society. pp. 127–143. 1999.

§Greene, C.A., S. Ankem, and S. Singh. "The Effect of Morphology and Alloying Elements on the Ambient Temperature Creep of Alpha and Alpha-Beta Titanium Alloys." Proceedings of the Eighth World Conference on Titanium. Vol. II. P.A. Blenkinshop, W.J. Evans, and H.M. Flower, eds. Birmingham, United Kingdom: Woodhead Publishing. pp.1,315–1,322. 1995.

‖Greene, C.A. and S. Ankem. "Ambient Temperature Tensile and Creep Deformation Behavior of α and β Titanium Alloys." *Beta Titanium Alloys in the 1990's*. D. Eylon, R.R. Boyer, and D.A. Koss, eds. Warrendale, Pennsylvania: The Minerals, Metals & Materials Society. pp. 309–319. 1993.

¶Thompson, A.W. and B.C. Odegard. "The Influence of Microstructure on Low Temperature Creep of Ti-5Al-2.5Sn. *Metallurgical Transactions A*. Vol. 4, No. 4. pp. 899–908. 1973.

#Neeraj T., D.H. Hou, G.S. Daehn, and M.J. Mills. "Phenomenological and Microstructural Analysis of Room Temperature Creep in Titanium Alloys." *Acta Materialia*. Vol. 48. pp. 1,225–1238. 2000.

**Chu, H.P. "Room Temperature Creep and Stress Relaxation of a Titanium Alloy." *Journal of Materials*. Vol. 5, No. 3. pp. 633–642. 1970.

††Miller H., Jr., R.T. Chen, and E. A. Starke, Jr. "Microstructure, Creep and Tensile Deformation in Ti-6Al-2Nb-1Ta-0.8Mo." *Metallurgical Transactions A*. Vol. 18A. pp. 1,451–1,468. 1987.

Table 4-2. Creep Strain *A* and *n* for Various Grain Sizes at a Stress Level of 90 Percent Yield Stress* (Reproduced With Permission From Metallurgical Transactions)				
Grain Size (μm)	**Creep Strain After ~200 Hours (%)**	***A***	***n***	**R^2**
226	5.95	1.18	0.32	0.98
62	2.37	0.48	0.31	0.99
38	1.92	0.36	0.33	0.99

*Aiyangar, A.K., B.W. Neuberger, P.G. Oberson, and S. Ankem. "The Effect of Stress Level and Grain Size on the Ambient Temperature Creep Deformation Behavior of an Alpha Ti-1.6 Wt Pct V Alloy." *Metallurgical and Materials Transactions A: Physical Metallurgy and Materials Science*. Vol. 36A. pp. 637–644. 2005.

Table 4-3. Creep Strain *A′* and *B* at Various Stress Levels for Grain Size of 226 μm [8,814 μin]* (Reproduced With Permission From Metallurgical Transactions)				
Stress Level (%)	**Creep Strain After ~200 Hours (%)**	***A′***	***B***	***R^2***
95	6.67	1.08	0.99	0.99
90	5.95	0.99	0.85	0.99
85	2.88	0.17	0.46	0.95
80	0.88	0.04	0.13	0.9
75	0.49	0.02	0.07	0.82

*Aiyangar, A.K., B.W. Neuberger, P.G. Oberson, and S. Ankem. "The Effect of Stress Level and Grain Size on the Ambient Temperature Creep Deformation Behavior of an Alpha Ti-1.6 Wt Pct V Alloy." *Metallurgical and Materials Transactions A: Physical Metallurgy and Materials Science*. Vol. 36A. pp. 637–644. 2005.

stresses, the solution treated and aged appears to have higher creep resistance, and when stresses are lower than 85 percent of the yield stress, no significant creep occurred. However, in reverse loading, significant creep was noticed similar to a Bauschinger effect (i.e., the property in tension and compression is not the same) (Dieter, 1986, p. 72).

Odegard and Thompson (1974) studied the room temperature creep behavior of Ti-6Al-4V alloy in the solution treated and aged condition and in the as-welded condition at various stress levels (Figures 4-13 and 4-14), and it was determined that creep resistance depends on microstructure and prestrain. In addition, creep resistance of the weld fusion zone is much lower than that of the aged material. Measurable creep occurs at stresses of 50 percent yield stress in the as-welded condition. Based on these results, it is recommended that heat treatment of the weld zone be used to improve creep resistance. Imam and Gilmore (1979) also studied room temperature creep behavior of Ti-6Al-4V alloy. Some of their results, which are accumulated creep strains, are shown in Table 4-4. Measurable creep was observed at stresses at 25 percent of the yield stress when tested in torsional loading. Imam and Gilmore (1979) concluded that unlike Reimann (1971), creep under torsional loading at low stresses is observed due to a strain-measuring technique which can resolve very low strains.

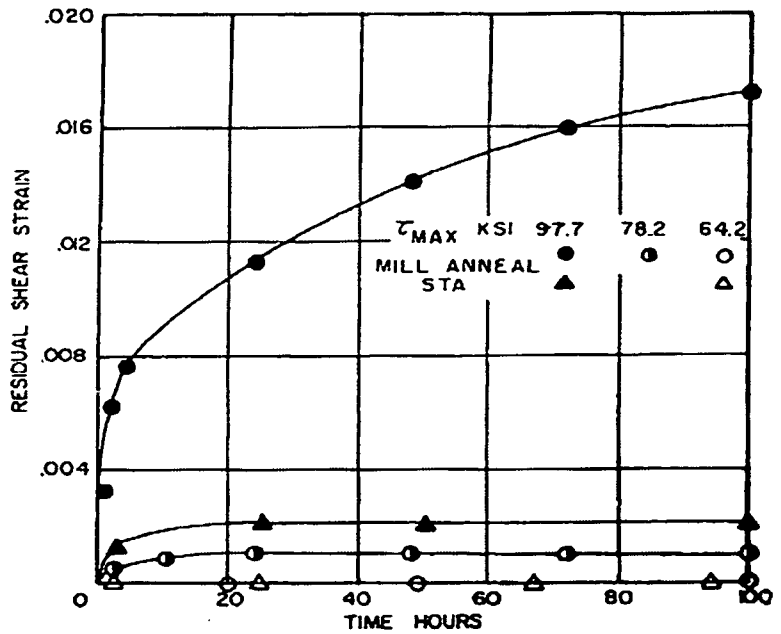


Figure 4-12. Typical Torsional Creep Curves of Solid Specimens in the Mill-Annealed and Solution-Treated and Aged Condition (Reimann, 1971) (Reproduced With Permission From ASTM International)

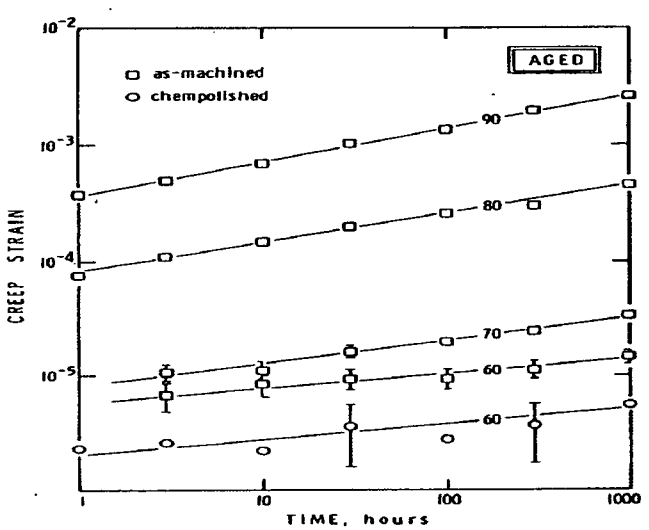


Figure 4-13. Creep Results for Aged Material. Curves Are Labeled With Creep Stresses as Percentages of σ_Y (Measurement Precision Shown Where Error Is Larger Than the Size of Corresponding Plotting Symbol; Two Examples Shown for Chempolished Tests) (Odegard and Thompson, 1974) (Reproduced With Permission From Metallurgical Transactions)

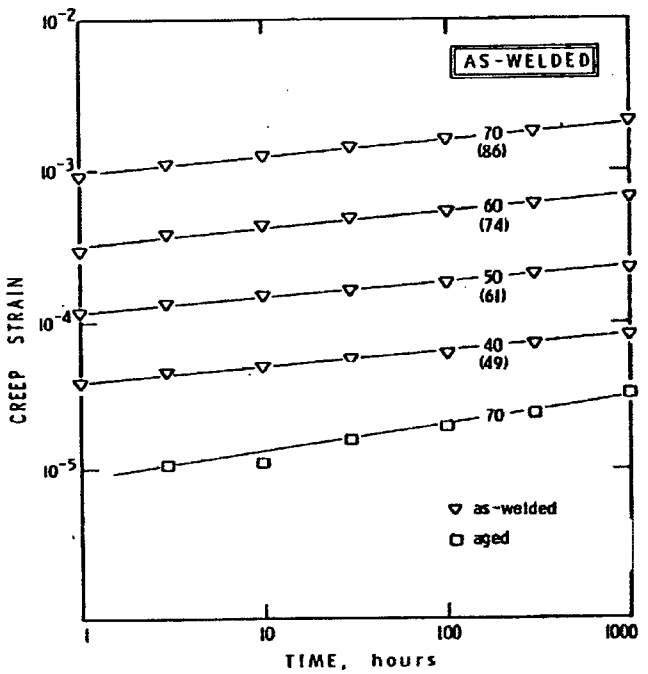


Figure 4-14. Creep Results for As-Welded Specimens, Labeled as in Figure 4-13. Values in Parenthesis Are Creep Stresses Expressed as Percentages of Welded σ_Y. Aged Results Included for Comparison (Odegard and Thompson, 1974) (Reproduced With Permission From Metallurgical Transactions).

Gao and Dexter (1987) studied the effect of hydrogen on the creep behavior of Ti-6Al-4V alloy at room temperature (Figure 2-10). From Figure 2-10, it appears that increased hydrogen enhances primary creep. This effect is due to softening of the material by enhancing the mobility of dislocations. Gao and Dexter (1987) further believe that the early dislocation movement results in a strong localized hydrogen buildup due to a "sweeping effect," thereby causing hydrogen embrittlement at a later stage in the deformation process.

Brandes and Mills (2004) observed static recovery in α (Ti-6wt% Al) and α-β (Ti-6242) alloys. This recovery occurred at low temperatures including room temperature and appears to be thermally activated. This recovery process is related to the pileups of planar arrays of dislocations at grain and phase boundaries. As a result, if a specimen is reloaded after static recovery from a previous creep loading, the specimen again deforms as if it is deforming for the first time. This recovery effect needs to be considered in drip shield design.

4.2.2 Threshold Stress

Based on the available information presented in the preceding section, the threshold stress for Grade 5 appears to be range from 25 to 50 percent yield stress. It is possible that the actual threshold is closer to the upper range, but this needs to be experimentally verified. There is currently no significant information to prove or disprove the lower threshold of 25 percent yield stress as predicted by Imam and Gilmore (1979).

Table 4-4. Total Accumulated Creep Strain During Forward Loading* (Reproduced With Permission From Metallurgical Transactions)				
	Anneal Treatment			
	RA†	α-βA‡		
Stress	Total Creep Strain × 10^3		βA§	
131 MPa [18.99 ksi]	0.065	0.065	0‖	0¶
147 MPa [21.36 ksi]	#	0.119	#	#
164 MPa [23.74 ksi]	0.25	#	0	0.025
245 MPa [35.60 ksi]	0.561	0.238	0.044	0.037
327 MPa [47.78 ksi]	1.106	0.412	0.154	0.087
409 MPa [59.34 ksi]	2.257	0.705	0.81	0.511
491 MPa [71.21 ksi]	31.064	21.057	8.447	#

*Imam, A. and C.M. Gilmore. "Room Temperature Creep of Ti-6Al-4V." *Metallurgical Transactions A.* Vol. 10A. pp. 419–425. 1979.
†RA—Recrystallization annealed
‡α-βA—α-β annealed.
§βA—β annealed.
‖Continuously increasing load test.
¶Load-unload test.
#Indicates this stress was not applied to this specimen.

4.2.3 Sample Predictions of Creep Strains After 10,000 Years

Neuberger, et al. (2002) have assembled creep constants corresponding to Eq. (2-1) for various α-β titanium alloys. As noted earlier, it is not advisable to average the creep constants for the long-term prediction, because the creep behavior strongly depends on the amount of α and β phases, the phase composition, and morphology.

For example, in Table 4-5, compare the Ti-6Al-4V alloy constants obtained by Odegard and Thompson (1974) for the aged and as-welded conditions. Their own conclusions and Figure 4-14 clearly indicate that at a given percent of the yield stress, the creep strain of welded microstructure is much greater than that of the aged microstructure. Yet, since the *n* value is less for the weld microstructure, there is a time after which the creep strain of the welded structure is lower than that of the aged structure. In addition, *n* values of Odegard and Thompson (1974) varied from 0.10 to 0.52. Note that while these constants may predict the short-term creep behavior reasonably well, inconsistencies occur when long term creep predictions are made based upon short-term data.

As an example, the specific constants obtained by Odegard and Thompson (1974) given in Table 4-5 are selected. For the microstructure of Ti-6Al-4V alloy at 90 percent yield stress, the

Table 4-5. Creep Power Law Coefficients for Various Alpha-Beta Titanium Alloys* (Reproduced With Permission From Materials Research Society and Neuberger)			
Titanium Alloy	**Yield Stress (%)**	***A***	***n***
Ti-6242†	95	0.4	0.21
Ti-8.1V‡	95	0.28	0.12
Ti-6.0Mn‡	95	0.037	0.29
Ti-6Al-4V aged§	90	0.037	0.29
Ti-6Al-4V as welded§	86	0.097	0.12

*Neuberger, B., C. Greene, and D. Gute. "Creep Analyses of Titanium Drip Shield Subjected to Rockfall Static Loads in the Proposed Geologic Repository at Yucca Mountain." Proceedings of the Materials Research Society. Symposium Proceedings 716. pp. JJ11.7.1–JJ11.7.8. 2002.

†Neeraj, T., D.H. Hou, G.S. Daehn, and M.J. Mills. "Phenomenological and Microstructural Analysis of Room Temperature Creep in Titanium Alloys." *Acta Metallurgica*. Vol. 48. pp. 1,225–1,238. 2000.

‡Greene, C.A., S. Ankem, and S. Singh. "The Effect of Morphology and Alloying Elements on the Ambient Temperature Creep of Alpha and Alpha-Beta Titanium Alloys." Proceedings of the Eighth World Conference on Titanium. Vol. II. P.A. Blenkinshop, W.J. Evans, and H.M. Flower, eds. Birmingham, United Kingdom: Woodhead Publishing. pp.1,315–1,322. 1995.

§Odegard, B.C. and A.W. Thompson. "Low Temperature Creep of Ti-6Al-4V." *Metallurgical Transactions A*. Vol. 5A, No. 5. pp. 1,207–1,213. 1974.

constant *A* is 0.037, and the constant *n* is 0.29. Substitution of these constants, along with a time period of 10,000 years in Eq. (2-1), results in a creep strain of about 7.44 percent at room temperature. Again, due to lack of data at higher temperatures such as 150 °C [302 °F], if it is assumed that the *A* value is doubled but *n* remains constant, the creep strain is 15 percent. ASTM B 381-03 suggests that the minimum total elongation for Grade 5 is 10 percent which means the uniform elongation, after which instability occurs in tension, is much lower than 10 percent. Therefore, based on these creep strain estimates, there is vulnerability for creep failure at stress levels close to the yield stress. Neuberger, et al. (2002) reached a similar conclusion. As a result, careful experiments are needed to determine the threshold stresses and the creep strains in the temperature range of 25 to 150 °C [77 to 302 °F].

5 THE EFFECT OF CREEP ON FRACTURE TOUGHNESS AT LOW TEMPERATURES

The last item to be addressed in this review is to determine whether low temperature creep can affect fracture toughness. In this regard, the phenomenon known as sustained load cracking observed in many titanium alloys is of interest. Sustained load cracking is subcritical crack growth which occurs in the absence of an aggressive environment including inert atmospheres. A number of investigators have studied this phenomena (Boyer and Spurr, 1978; Pardee and Paton, 1980; Yeh and Huang, 1998; Margolin, 1976; Pao, et al., 1995; Krafft, 1974; Gu and Hardie, 1997; Williams, 1976, 1975; Kennedy, et al., 1993; Yoder, et al., 1974; Meyn, 1974) and believe that somehow hydrogen is involved. As pointed out by Meyn (1974), sustained load cracking can occur in titanium alloys including commercially pure titanium. As an example, the stress intensities of fast crack growth and slow growth for Ti-6Al-4V are shown in Figure 5-1.

Note that hydrogen at low concentrations initially increases then reduces the slow crack growth stress intensity factor. Williams (1976) found that hydrogen at low concentration levels reduces the secondary creep rate by raising the slow crack, growth-rate stress intensity. At higher hydrogen levels, as pointed out by Gao and Dexter (1987), hydrogen increases the primary creep rate, which can promote embrittlement and leads to a lower stress intensity. While it is not clear why hydrogen reduces secondary creep rate but increases primary creep rate at higher concentration levels, there is sufficient information to suggest that creep can influence fracture toughness of titanium alloys at low temperatures.

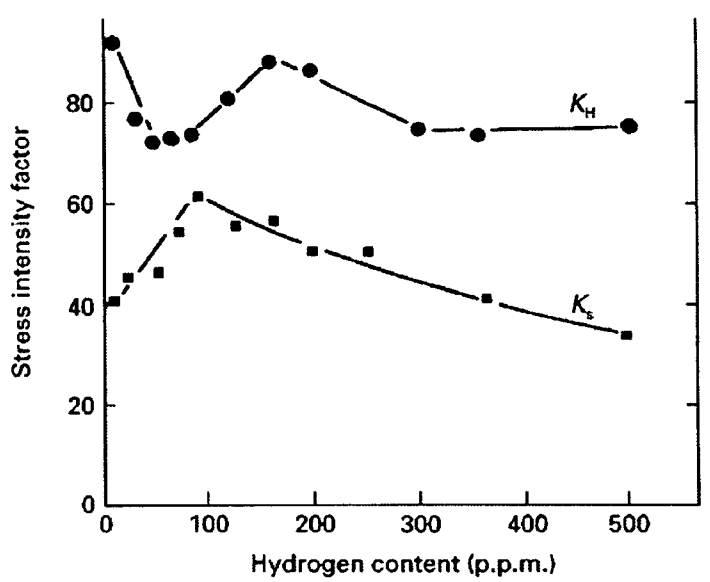


Figure 5-1. Effect of Hydrogen on the Threshold Stress Intensity Factor for Slow Crack Initiation K_s and the Critical Stress Intensity Factor for Fast Cracking K_H (Gu and Hardie, 1997) (Reproduced With Permission From Springer)

6 SUMMARY

This study systematically reviewed the creep deformation behavior of α, α-β, and β titanium alloys as a function of stress level, temperature, chemical composition, and microstructure of phases in general, and Grade 7 and Grade 24 or their surrogates in particular. The study also established a technical basis for the stress threshold needed to initiate low temperature creep of Titanium Grades 7 and 24, or reasonable surrogates, based on the information available in the public domain. In addition, various creep models and applicable coefficients, are used to estimate the creep strain.

A general description of the low temperature creep behavior of titanium alloys was presented. In particular, constitutive relationships governing low temperature creep behavior of various titanium alloys was presented. Two types of equations were found, either a logarithmic or a parabolic form, that can be used to calculate creep strain depending on the data. The descriptions included the activation energies for various low temperature creep processes. The section also dealt with various creep deformation mechanisms including slip, twinning, grain boundary sliding, α-β interface sliding, Widmanstatten colony boundary sliding, and stress-induced martensite. In addition to the various mechanisms, Section 2 also included various factors that can affect creep deformation: chemistry, microstructure, texture, stress level and proportional limit, elastic interaction stresses, stress type, cold work, and creep test temperature.

In this literature review, no specific creep data was found for Titanium Grades 7 an 24 which are the materials to be used in the DOE proposed drip shield design. As a result, because the object of this report is to determine if creep may lead to significant deformation of the drip shield, surrogates for these two grades of titanium were selected. This selection was based on the chemical composition of the titanium surrogates (Section 3).

In Section 4, the creep behavior of the surrogates of 7 and 24 were reviewed. The review included general creep behavior, threshold stress, and sample predictions of creep strains after 10,000 years. Based on the literature, it was concluded that the threshold stress range for surrogates of Grade 7 is 35–60 percent yield stress and for surrogates of Grade 24, the threshold stress range is 25–50 percent yield stress. Surrogates of both of these grades were found to be vulnerable for creep failure when subjected to stresses close to the yield stress for 10,000 years.

The effect of creep on fracture toughness at low temperatures was reviewed in Section 5. It is known that α and α-β titanium alloys are prone to sustained load cracking. The open literature indicated that there is sufficient information to suggest that creep strain can affect sustained load cracking behavior (i.e., creep can affect fracture toughness).

This review indicated that there is a lack of information on the creep behavior of Grade 7 and Grade 24 alloys to draw definitive conclusions. Therefore, recommendations have been made at appropriate places in this study to carry out specific experiments; in particular, experiments to determine the threshold stresses and meaningful creep constants which can be used to predict long-term creep behavior have been suggested.

7 REFERENCES

Adenstedt, H. "Creep of Titanium at Room Temperature." *Metal Progress*. Vol. 56, No. 5. pp. 658–660. 1949.

Aiyangar, A.K., B.W. Neuberger, P.G. Oberson, and S. Ankem. "The Effect of Stress Level and Grain Size on the Ambient Temperature Creep Deformation Behavior of an Alpha Ti-1.6 Wt Pct V Alloy." *Metallurgical and Materials Transactions A: Physical Metallurgy and Materials Science*. Vol. 36A. pp. 637–644. 2005.

Ankem, S. and H. Margolin. "Modeling Deformation in Two Phase Alloys." *Journal of Metals*. No. 4. pp. 25–29. 1986a.

———. "A Rationalization of Stress—Strain Behavior Of Two-Ductile Phase Alloys." *Metallurgical Transactions A*. Vol. 17A. pp. 2,209–2,226. 1986b.

———. "Alpha-Beta Interface Sliding in Ti-Mn Alloys." *Metallurgical Transactions A*. Vol. 14 A. pp. 500–503. 1983.

Ankem, S. and S.R. Seagle. "The Detrimental Effects of Iron on Creep of Ti-6242S Alloy." Proceedings of the Fifth International Conference on Titanium, Munich, Germany, September 10–14, 1984. Vol. 4. Munich, Germany: Deutsche Gesellschaft für Metallkunde eV. pp. 2,411–2,418. 1985.

Ankem, S., C.A. Greene, and A.K. Aiyangar. "Recent Developments in Growth Kinetics of Deformation Twins in Bulk Metallic Materials." Proceedings of the Symposium on Advances in Twinning. S. Ankem and C.S. Parde, eds. Warrendale, Pennsylvania: The Minerals, Metals & Materials Society. pp. 127–143. 1999.

Ankem, S., C.A. Greene, and S. Singh. "Time Dependent Twinning During Ambient Temperature Creep of Alpha Ti-Mn Alloys." *Scripta Materialia*. Vol. 30, No. 6. pp. 803–808. 1994.

Ankem, S., J.G. Shyue, M.N. Vijayshankar, and R.J. Arsenault. "The Effect of Volume Percent of Phases on the High Temperature Tensile Deformation of Two Phase Ti–Mn Alloys." *Materials Science and Engineering A*. Vol. A111. pp. 51–61. 1989.

Bechtel SAIC Company, LLC. "FEPs Screening of Processes and Issues in Drip Shield and Waste Package Degradation." Rev. 3. Las Vegas, Nevada: Bechtel SAIC Company, LLC. 2004.

———. "Technical Basis Document No. 6: Waste Package and Drip Shield Corrosion." Rev. 1. Las Vegas, Nevada: Bechtel SAIC Company, LLC. 2003.

Bendersky, L., A. Rosen, and A.K. Mukherjee. "Creep and Dislocation Substructure." *International Metals Review*. Vol. 30. pp. 1–15. 1985.

Boehlert, C. and D. Miracle. "Part II. The Creep Behavior of Ti-Al-Nb O+BCC Orthorhombic Alloys." *Metallurgical and Materials Transactions A: Physical Metallurgy and Materials Science.* Vol. 30A. pp. 2,349–2,367. 1999.

Boyer, R.R. and W.F. Spurr. "Characteristics of Sustained Load Cracking and Hydrogen Effects in Ti-6Al-4V." *Metallurgical Transactions A.* Vol. 9A. pp. 23–29. 1978.

Boyer, R., G. Welsch, and E.W. Collings, eds. *Materials Properties Handbook: Titanium Alloys.* Materials Park, Ohio: ASM International. 1994.

Brandes, M. and M.J. Mills. "Static Recovery in Titanium Alloys at Lower Temperatures." *Materials Science and Engineering A.* Vol. 387–368. pp. 570–575. 2004.

Chen, I.W. and A.S. Argon. "Steady-State Power Law Creep In Heterogeneous Alloys With Coarse Micro-Structures." *Acta Metallurgica.* Vol. 27. pp. 785–791. 1979.

Cho, K. and J. Gurland. "The Law of Mixtures Applied to the Plastic Deformation of Two Phase Alloys of Coarse Microstructures." *Metallurgical Transactions A.* Vol. 19A. pp. 2,207–2,040. 1988.

Christian, J.W. *The Theory of Transformations in Metals and Alloys.* New York City, New York: Pergamon Press. p. 789. 1965.

Chu, H.P. "Room Temperature Creep and Stress Relaxation of a Titanium Alloy." *Journal of Materials.* Vol. 5, No. 3. pp. 633–642. 1970.

Code of Federal Regulations. "Disposal of High-Level Radioactive Wastes in Geologic Repository at Yucca Mountain, Nevada." Title 10—Energy, Chapter 1—Nuclear Regulatory Commission, Part 63. Vol. 2. Washington, DC: U.S. Government Printing Office. 2005.

Cuddy, L.J. "Internal Stresses and Structures Developed During Creep." *Metallurgical Transactions A.* Vol. 1. pp. 395–401. 1970.

Dieter, E. "*Mechanical Metallurgy.*" 3rd Edition. New York City, New York: McGraw-Hill Book Company. 1986.

Doraiswamy, D. and S. Ankem. "The Effect of Grain Size and Stability on Ambient Temperature Tensile and Creep Deformation in Metastable Beta Titanium Alloys." *Acta Metallurgica.* Vol. 51. pp. 1,607–1,619. 2003.

Drefahl, K., P. Wincierz, U. Zwicker, and P. Delarbre. "230000 h Creep Properties of Titanium Produced from Electrolotic and Sponge Material and TiAl6V4 Alloy at 20 °C: Titanium Science and Technology." Proceedings of the Fifth International Conference on Titanium, Munich, Germany, September 10–14, 1984. Vol. 4. G. Lutjering, U. Zwicker, and W. Bunk, eds. Munich, Germany: Deutsche Gesellschaft für Metallkunde eV. pp. 2,387–2,394. 1984.

Dyson, F. and M. McLean. "Particle Coarsening, σ_0, and Tertiary Creep." *Acta Metallurgica.* Vol. 31. pp. 17–27. 1983.

Elliot, P. "Constitution of Binary Alloys." First Supplement. New York City, New York: McGraw-Hill, Inc. p. 617. 1965.

Evans, R.W. and B. Wilshire. *Creep of Metals and Alloys*. London, England: Institute of Metals. 1985.

Gao, G.Y. and S.S. Dexter. "Effect of Hydrogen on Creep Behavior of Ti-6Al-4V Alloy at Room Temperature." *Metallurgical Transactions A*. Vol. 18A. pp. 1,125–1,130. 1987.

Greene, C.A. "Fundamental Studies on the Ambient Temperature Creep Deformation Behavior of Alpha, Beta, and Alpha-Beta Titanium Alloys." Ph.D. thesis. University of Maryland. College Park, Maryland. p. 92. 1994.

Greene, C.A. and S. Ankem. "Modeling of Elastic Interaction Stresses in Two-Phase Materials by FEM." *Materials Science and Engineering A*. Vol. 202A. pp. 103–111. 1995.

———. "Ambient Temperature Tensile and Creep Deformation Behavior of α and β Titanium Alloys." *Beta Titanium Alloys in the 1990's*. D. Eylon, R.R. Boyer, and D.A. Koss, eds. Warrendale, Pennsylvania: The Minerals, Metals & Materials Society. pp. 309–319. 1993.

Greene, C.A., S. Ankem, and S. Singh. "The Effect of Morphology and Alloying Elements on the Ambient Temperature Creep of Alpha and Alpha-Beta Titanium Alloys." Proceedings of the Eighth World Conference on Titanium. Vol. II. P.A. Blenkinshop, W.J. Evans, and H.M. Flower, eds. Birmingham, United Kingdom: Woodhead Publishing. pp.1,315–1,322. 1995.

Grewal, G. and S. Ankem. "Particle Coarsening Behavior of α-β Titanium Alloys." *Metallurgical Transactions A*. Vol. 21A. pp. 1,645–1,654. 1990.

———. "Isothermal Particle Growth in Two-Phase Titanium Alloys." *Metallurgical Transactions A*. Vol. 20A. pp. 39–53. 1989.

Gu, J. and D. Hardie. "Effect of Hydrogen on the Tensile Ductility of Ti-6Al-4V." *Journal of Materials Science*. pp. 609–617. 1997.

Hasija, V., S. Ghosh, M. Mills, and D. Joseph. "Deformation and Creep Modeling in Polycrystalline Ti–6Al Alloys." *Acta Metallurgica*. Vol. 51. pp. 4,533–4,549. 2003.

Henderson, P.J. and M. McLean. "Micro-Structural Changes During High Temperature Creep of a Directionally Solidified Nickel Base Superalloy." *Scripta Materialia*. Vol. 19. p. 99–104. 1985.

———. "Micro-Structural Contributions to Friction Stress and Recovery Kinetics During Creep of the Nickel Base Superalloy IN738LC. *Acta Metallurgica*. Vol. 31. pp. 1,203–1,219. 1983.

Hirth, J.P. and J. Lothe. *Theory of Dislocations*. New York City, New York: John Wiley & Sons, Inc. pp. 818–825. 1982.

Hultgren, C.A., C.A. Greene, and S. Ankem. "Time Dependent Twinning During Ambient Temperature Creep of Alpha Ti-0.4% Mn Alloy." *Metallurgical Transactions A*. Vol. 30A. pp. 1,675–1,679. 1999.

Ibarra, L., T. Wilt, G. Ofoegbu, and A. Chowdhury. "Structural Performance of Drip Shield Subjected to Static and Dynamic Loading." San Antonio, Texas: CNWRA. 2005.

Imam, A. and C.M. Gilmore. "Room Temperature Creep of Ti-6Al-4V." *Metallurgical Transactions A*. Vol. 10A. p. 419. 1979.

Kennedy, J.R., P.N. Adler, and H. Margolin. "Effect of Activity Differences on Hydrogen Migration in Dissimilar Titanium Alloy Welds." *Metallurgical Transactions A*. Vol. 24A. pp. 2,763–2,771. 1993.

Kiessel, W.R. and M.J. Sinnott. "Creep Properties of Commercially Pure Titanium." *Journal of Metals*. Transactions AIME. p. 338. 1953.

Kolluru, D. and T. Pollock. "Numerical Modeling of the Creep Behavior of Unidirectional Eutectic Composites." *Acta Metallurgica*. Vol. 46. pp. 2,859–2,876. 1998.

Krafft, J.M. "An Interpretation of the Yonder-Griffis-Crooker Observations of Sustained-Load Cracking in Ti-6Al-4V." *Journal of Engineering Materials and Technology*. pp. 275–282. 1974.

Manero, J.M., F.J. Jil, and J.A. Panell. "Deformation Mechanisms of Ti-6Al-4V Alloy With a Martensitic Microstructure Subjected to Oligocyclic Fatigue." *Acta Metallurgica*. Vol. 48. pp. 3,353–3,359. 2000.

Margolin, H. "Stress, Hydrogen Segregation, and Fracture in α-β Titanium Alloys." *Metallurgical Transactions A*. Vol. 7A. pp. 1,233–1,235. 1976.

McLean, M. "On the Threshold Stress for Dislocation Creep in Particle Strengthened Alloys." *Acta Metallurgica*. Vol. 33. pp. 545–556. 1985.

———. "The Contribution of Friction Stress to the Creep Behavior of the Nickel-Base in situ Composite γ-γ'-Cr_3C_2." *Proceedings of the Royal Society of London Series A*. Vol. 373A. pp. 93–109. 1980.

Meyn, D.A. "Effect of Hydrogen on Fracture and Inert-Environment Sustained Load Cracking Resistance of Alpha-Beta Titanium Alloys." *Metallurgical Transactions A*. Vol. 5A. pp. 2,405–2,414. 1974.

Miller, H., Jr., R.T. Chen, and E.A. Starke, Jr. "Microstructure, Creep and Tensile Deformation in Ti-6Al-2Nb-1Ta-0.8Mo." *Metallurgical Transactions A*. Vol. 18A. p. 1,451–1,468. 1987.

Murray, J.L., ed. "Phase Diagrams of Binary Titanium Alloys." Materials Park, Ohio: ASM International. pp. 239–241. 1987.

Murty, K.L. "Effects of Alloying and Cold-Work on Anisotropic Biaxial Creep of Titanium Tubing." Proceedings of the Tenth World Conference on Titanium, Hamburg, Germany, July 13–18, 2003. G. Lutjering and J. Albrecht, eds. Vol. III. pp. 2,033–2,040. 2003.

Neeraj, T., D.H. Hou, G.S. Daehn, and M.J. Mills. "Phenomenological and Microstructural Analysis of Room Temperature Creep in Titanium Alloys." *Acta Metallurgica*. Vol. 48. pp. 1,225–1,238. 2000.

Neuberger, B., C. Greene, and D. Gute. "Creep Analyses of Titanium Drip Shield Subjected to Rockfall Static Loads in the Proposed Geologic Repository at Yucca Mountain." Proceedings of the Materials Research Society. Symposium Proceedings 716. pp. JJ11.7.1–JJ11.7.8. 2002.

Oberson, P.G. and S. Ankem. "Why Twins Do Not Grow at the Speed of Sound All the Time." *Physical Review Letters*. Vol. 95. pp. 165501-1–165501-4. 2005.

Odegard, B.C. and A.W. Thompson. "Low Temperature Creep of Ti-6Al-4V." *Metallurgical Transactions A*. Vol. 5A, No. 5. pp. 1,207–1,213. 1974.

Ofoegbu, G.I., B. Dasgupta, and K.J. Smart. "Thermally Induced Rock Stress and Implications for Degradation of Emplacement Drifts at the Potential Yucca Mountain Waste Repository." San Antonio, Texas: CNWRA. 2004.

Pao, P.S., D.A. Meyn, R.A. Bayles, and C.R. Feng. "Effect of Ripple Loads on Sustained-Load Cracking in Titanium Alloys." *Scripta Metallurgica*. Vol. 33. pp. 497–501. 1995.

Pardee, W.J. and N.E. Paton. "Modeling of Sustained Load Cracking by Hydride Growth in Titanium Alloys." *Metallurgical Transactions A*. Vol. 11A. pp. 1,391–1,400. 1980.

Paton, N.E, J.C. Williams, and G.P. Rauscher. "The Deformation of Alpha-Phase Titanium." Titanium Science and Technology: Proceedings of the Second International Conference on Titanium, Cambridge, Massachusetts, May 2–5, 1972. Vol. 2. R.I. Jaffee and H.M. Burte, eds. New York City, New York: Plenum Press. pp. 1,049–1,069. 1973.

Ramesh, A. "Effect of Stability and Grain Size on the Ambient Temperature Tensile and Creep Behavior of Beta Titanium Alloys." M.S. thesis. University of Maryland. College Park, Maryland. p. 100. 1997.

Ramesh, A. and S. Ankem. "Stress—Induced Products in a Ti-14.8%V Alloy Deformed in Tension." *Metallurgical and Materials Transactions A: Physical Metallurgy and Materials Science*. Vol 30A. pp. 2,249–2,251. 1999.

Reed-Hill, R. and R. Abbaschian. *Physical Metallurgy Principles*. 3rd Edition. Boston, Massachusetts: PWS-Kent. p. 538. 1992.

Reimann, W.H. "Room Temperature Creep in Ti-6Al-4V." *Journal of Materials*. Vol 6. pp. 926–940. 1971.

Salem, A., S. Kalidindi, and R. Doherty. "Strain Hardening of Titanium: Role of Deformation Twinning." *Acta Metallurgica*. Vol. 51. pp. 4,225–4,237. 2003.

Sargent, P.M. and M.F. Ashby. "Deformation Maps for Titanium and Zirconium." *Scripta Materialia.* Vol. 16. pp. 1,415–1,422. 1982.

Schlueter, J. "U.S. Nuclear Regulatory Commission/U.S. Department of Energy Technical Exchange and Management Meeting on Container Life and Source Term (September 12–13, 2000)." MOL.20010731.0161. Letter (October 4) to S. Brocoum, DOE. Washington, DC: NRC. 2000. <www.nrc.gov/waste/hlw-disposal/public-involvement/mtg-archive.html#KTI>

Suri, S., G.B. Vishwanathan, T. Neeraj, D.H. Hou, and M.J. Mills. "Room Temperature Deformation and Mechanisms of Slip Transition in Oriented Single-Colony Crystals of an α-β Titanium Alloy." *Acta Metallurgica.* Vol. 47, No. 3. pp. 1,019–1,034. 1999.

Thompson, A.W. and B.C. Odegard. "The Influence of Microstructure on Low Temperature Creep of Ti-5Al-2.5Sn." *Metallurgical Transactions A.* Vol. 4, No. 4. pp. 899–908. 1973.

Williams, D.N. "Effects of Hydrogen in Titanium Alloys on Subcritical Crack Growth Under Sustained Load." *Materials Science and Engineering A.* Vol. 24. pp. 53–63. 1976.

———. "Effect of Specimen Thickness on Subcritical Crack Growth Under Sustained Load." *Materials Science and Engineering A.* Vol. 18. pp. 149–155. 1975.

Yeh, M.-S. and J.-H. Huang. "Hydrogen-Induced Subcritical Crack Growth in Ti-6Al-4V Alloy." *Materials Science and Engineering A.* Vol. A242. pp. 96–107. 1998.

Yoder, G.R., C.A. Griffis, and T.W. Crooker. "The Cracking of Ti-6Al-4V Alloys Under Sustained Load of Ambient Air." *Transactions of the ASME.* New York City, New York: The American Society of Mechanical Engineers. pp. 268–274. 1974.