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Fatigue and Fracture Behavior of
U-6 Wt. Pct. Nb

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May 21, 1993

Lawrence
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Work performed under the auspices of the U.S. Department of Energy by the Lawrence Livermore National Laboratory under Contract W-7405-Eng-48.

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UCRL-ID-110446

ABSTRACT

The fatigue and fracture properties of U6Nb were measured to provide the materials property data needed for structural designs in material processed by solution quenching and aging 200°C/2h. Limited testing was also performed on as-quenched U6Nb. We have extended the database on fatigue properties in U6Nb to include both crack initiation data (fatigue strengths) and crack propagation data (fatigue crack growth rates). The static load carrying capabilities have been characterized through fracture toughness and tensile property measurements. Using a rotating beam fatigue machine, a fatigue strength of 248 MPa was measured at 10^8 cycles for smooth bars at zero mean load. As is typical of nonferrous alloys, U6Nb does not exhibit a fatigue endurance limit. Reductions in fatigue strength for notched bars and for mean loads of 276 MPa (40 ksi) and 483 MPa (70 ksi) were also determined. The predominant sites for fatigue crack initiation were identified as niobium carbide and uranium oxide inclusion clusters and the distribution of these inclusions are presented.

Fatigue crack propagation rates were measured in the near-threshold regime using compact tension specimens. The fatigue threshold for crack growth rates below 10^{-7} mm/cycle were measured at both $R=0.1$, for which a fatigue threshold of $3.2 \text{ MPa}\sqrt{\text{m}}$ was measured, and for constant K_{max} cycles with K_{max} values of $14.6 \text{ MPa}\sqrt{\text{m}}$ and $30.5 \text{ MPa}\sqrt{\text{m}}$, for which the fatigue threshold was reduced to $0.9 \text{ MPa}\sqrt{\text{m}}$ and $0.6 \text{ MPa}\sqrt{\text{m}}$, respectively. Crack closure effects were present in near-threshold testing for $R=0.1$ based upon specimen compliance data, and an effective threshold ΔK was calculated as $2.1 \text{ MPa}\sqrt{\text{m}}$. Fatigue crack growth rates were measured for both as-quenched and quenched plus aged (200°C/2h) conditions, and for environments of laboratory air, dry nitrogen, and 100RH air. Crack branching was observed in all environments for ΔK 's below approximately

10 MPa $\sqrt{\text{m}}$. Reduced crack growth rates in dry nitrogen at $R=0.1$ and the K_{max} dependence of near-threshold behavior in air are ascribed to environmentally-assisted-fatigue.

ACKNOWLEDGEMENTS

The authors acknowledge D. Selway (LLNL), D. Wood (LLNL), , R. Riddle (LLNL), and I. Finnie (U. C. Berkeley) for contributions in directing the scope of this work, N. Nguyen for assistance with the DCPD system, R. Kershaw for metallographic preparations, E. Sedillo for SEM assistance, B. Westfall for assistance with the S-N tests, R. Oakes (Y-12) for the low temperature fracture toughness measurements, and L. Chapman (Y-12) for statistical data on inclusion distributions in U6Nb.

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INTRODUCTION

The utility of the depleted uranium alloy U-6 wt.% Nb (U6Nb) as a structural material is a result of the beneficial effects of Nb in improving the resistance to general corrosion and stress corrosion cracking while providing a substantial strengthening increment [1-3]. In the present investigation, the fatigue and fracture properties of U6Nb were measured to provide the materials property data needed for structural designs. Previous to this work, no fatigue crack propagation data has been published for this material and the only reported fatigue strengths are for smooth-bar tests of as-quenched U6Nb [2]. We have extended the fatigue property measurements of U6Nb to provide data for predictions of both crack initiation lifetime (fatigue strengths) and crack propagation lifetime (fatigue crack growth rates). The static load carrying capabilities have been characterized through fracture toughness and tensile property measurements.

Fatigue can be defined as the progressive localized damage produced by the cyclic application of tensile stresses, usually at levels below that required for bulk deformation. This damage can be separated into three regimes: crack initiation, crack propagation, and final failure. Crack initiation testing provides the most traditional fatigue design data. The total number of fatigue cycles required for failure is measured vs. the applied cyclic stress amplitude and the fatigue strength is determined based on the required cyclic lifetime. For high-cycle fatigue conditions, the total lifetime is principally governed by the number of fatigue cycles necessary to initiate a crack. The fatigue strength is therefore dependent on the surface condition and stress concentrations in the component. The fatigue strength can also be altered by mean tensile loads, test environment, and the metallurgical condition of the material. In this investigation, each of these variables were incorporated into the test matrix by using anticipated service

conditions and/or worst-case conditions for each variable as described in the experimental procedures.

An alternative fatigue design approach is to use fracture mechanics methods. The fatigue lifetime can be calculated for a given loading history by integrating the crack growth rates for the stress intensities produced between an initial and final crack length. This design method is more conservative than the previous approach since it assumes the pre-existence of a crack-like flaw in the material and allows calculation of the remaining cyclic lifetime. The cyclic stress intensity can be calculated for any component from the applied stresses and geometric factors which greatly facilitates the application of test data to specific component designs. At sufficiently low cyclic stress intensities a fatigue threshold is reached, below which crack growth rates decrease asymptotically to below 10^{-8} mm/cycle. The value of the fatigue threshold is expected to be sensitive to the mean stress level, the fatigue environment, and the metallurgical condition of the material [24]. Again, these variables have been evaluated as described in the experimental procedures.

The strength and corrosion properties of U6Nb are known to be sensitive to the thermal history of the material [1-3, 11, 18-21]. The two heat treatment conditions in most common use are the as-quenched condition in which the Nb remains in solid solution or a condition consisting of subsequent aging at 200 °C for 2 hr in which early stage precipitation of a Nb-rich phase occurs [1]. While the 200°C/2h heat treatment produces an under-aged microstructure, it avoids the severe degradation in corrosion resistance which accompanies the higher strength microstructures achieved with further aging [1-3]. The influence of metallurgical condition on the propagation behavior and fatigue threshold was evaluated in this investigation by testing both as-quenched, and under-aged U6Nb, though testing has focused on (under)aged material heat treated at 200°C/2h. The baseline

fatigue propagation behavior of U6Nb was determined for a cyclic load ratio, R (minimum load/ maximum load), of 0.1 in laboratory air. Comparison behavior was measured for conditions of high mean tensile loads and atmospheres of moisture-saturated air and dry nitrogen. The room temperature tensile properties were measured for both as-quenched and aged conditions. The fracture toughness of the aged material was measured at both room temperature and at $-55\text{ }^{\circ}\text{C}$ [4]. In addition, the distributions of inherent internal flaws in the form of oxide and carbide inclusions are summarized for use in fracture mechanics-based fatigue analyses. Finally, the methodology for fatigue lifetime predictions based on the property data reported here will be discussed.

EXPERIMENTAL METHOD

The U6Nb plates used in this investigation were produced at Martin Marietta's Y-12 Plant using standard production procedures. The plates were warm cross-rolled to a thickness of 16 mm (5/8 inch), and were received in the as-quenched condition (part number WR6854). The compositions of the two plates used in this study are listed in Table 1. Residual carbon in the alloy is an undesirable impurity which forms U and Nb carbide inclusions and the test compositions were selected to span the range of typical carbon contents. Plate 'A' contains a high carbon content of 100 ppm, which is near the maximum impurity level produced by Y-12. Plate 'B' contains only 65 ppm carbon. The tensile and fracture toughness properties were measured for each of the two plates. Fatigue measurements were made using material from the high carbon plate 'A', which should provide conservative property data due to the potentially detrimental effects of carbide inclusions.

Specimen preparation consisted of cutting specimens in the plane of the cross-rolled plates and rough machining to size. Where aging treatments were required, the rough-machined samples were heat-treated in an air furnace at 200 °C. A thermocouple was placed in contact with the parts and the time at temperature was controlled (commencing for a temperature of >195 °C). A typical aging treatment required a heating time of approximately 1 h. A more prolonged heating time was required for the vacuum furnace treatments performed on the first series of specimens which consisted of three tensile and six 0.5" thick compact tension specimens (#1AH to #6AH). These specimens were used for toughness testing and initial fatigue propagation tests. They are distinguished from the remaining aged samples by an elevated yield strength. Fatigue crack propagation measurements were performed initially on the vacuum treated specimens (constant $R=0.1$) but all the remaining samples were heat treated 200°C/2h in air. The influence of such heat treatment variations on the fatigue crack propagation behavior is expected to be minor based on the similar FCGR's measured in comparison tests of aged and unaged material, as discussed later. All subsequent specimens were heat treated in air at atmospheric pressure under standardized conditions. The test specimens were finish-machined using a standard series of cuts with decreasing depths of material removal simulating conventional machining schedules in component fabrication. Semi-finish cuts of 0.005" (0.13 mm), 0.002" (0.051 mm), and 0.002" (0.051 mm) were followed by a finish cut of 0.001" (0.025 mm) depth.

Static Mechanical Behavior

Tensile testing was performed using standard round tensile specimens with a 1.0" (25.4 mm) gage length and 0.25" (6.35 mm) gage diameter. Yield strength and elastic modulus values were measured using a clip-gauge extensometer. Total

elongation and reduction in area values were determined from final specimen dimensions. All specimens were tested in the ambient laboratory air environment.

Fracture toughness was measured using the single-specimen J_{IC} technique, following the recommended procedures in ASTM E813-87. Standard compact tension specimens, 0.5" (12.7 mm) thick, were used in fracture toughness testing of both as-quenched and aged (200 °C/2 hr) conditions. A clip gauge positioned at the crack mouth opening was used to measure the unloading compliance for crack length determinations. The fracture toughness was also measured at -55°C using the multiple specimen J_{IC} method with specimens 0.4" (10 mm) thick [4].

Fatigue Crack Initiation Testing

Fatigue stress vs. fatigue life (S-N) tests at zero mean load were performed on a single-cantilever, rotating-beam fatigue machine (Fatigue Dynamics RBF-200). Smooth-bar tests were performed using a tapered specimen with the dimensions shown in Figure 1a. The surface finish was characteristic of that produced by standard machining procedures, as outlined above. The specimens were coated with oil after final machining (Gulf No-Rust C), stored in the laboratory environment, and wiped dry prior to testing. The stresses were applied to the rotating specimens through bending loads which produce a sinusoidal waveform of fully reversed tensile and compressive stresses at a frequency of 50 Hz. The failure criterion for these tests was crack propagation sufficient to trigger a displacement-actuated microswitch. Typically, the remaining ligament at failure comprised less than 25% of the original diameter. Testing was interrupted upon exceeding 10^8 cycles without failure, with the exception of a limited number of tests which were extended further. Testing was performed in the ambient laboratory environment with an average humidity of between 40 and 60%.

The effect of stress concentrations on the fatigue life was evaluated by S-N testing of severely notched specimens. A V-notch with a 0.005" (0.13 mm) root radius (ρ) and depth (a) was cut into 0.5" (12.7 mm) diameter bars such that the remaining ligament was equivalent in diameter to the tapered smooth bar specimen, as shown in Figure 1b. This notch produces a theoretical stress concentration factor, K_t , of eleven, where $K_t = 1 + 2 \sqrt{a/\rho}$. Test conditions were identical to those used for smooth bars.

The effect of mean tensile stresses on the fatigue strength was evaluated using an axially-loaded specimen tested in a conventional tensile loadframe. The specimen was designed with a uniform cross-section over the center 0.5" of the reduced section as shown in Figure 1c. The failure criterion for fatigue life determinations was total failure. In addition, crack initiation lifetimes were monitored by novel application of a direct-current potential-drop system with current leads attached through threaded holes at the specimen ends, and voltage leads attached through mechanically-fastened copper collars at the shoulders of the gage section [15]. The voltage was automatically sampled at a rate of 1 Hz from a digital nanovoltmeter and stored on disk. The fatigue cycle was applied by hydraulic actuator at a frequency of 50 Hz using a square-wave function generator.

Fatigue Crack Propagation Testing

Fatigue crack growth rates (FCGR's) were measured vs cyclic stress intensity (ΔK) using compact tension specimens. A specimen thickness of 0.5" (12.7 mm) was used initially, then decreased to as thin as 0.1" (2.5 mm) for tests at constant maximum K to eliminate potential residual stress effects. Testing was performed exclusively at 20 Hz with a square wave applied by a hydraulic actuator on a 4-post MTS load-frame. The testing procedures followed ASTM standard E647-88 methods. Initial tests were performed under K -increasing conditions with

constant load cycles and constant stress ratio (R) of 0.1 (Note: K increases with increasing crack length). All near-threshold FCGR testing was performed under K -decreasing conditions using a stepped load shedding technique, with $d\Delta K/da$ limited by standard load decreases of 10% after a minimum of 0.02" (0.5 mm) of crack growth. Crack closure loads were determined from the unloading compliance measured using a clip gauge mounted on integral knife edges at the crack mouth opening. To prevent unwanted crack closure and to simulate service environments with mean tensile stresses, the FCGR's at constant maximum K values were also evaluated. The constant plastic zone size produced for conditions of constant maximum K allows increased rates of load shedding [5], but were maintained in general. Exceptions were overnight fatigue cycles at near-threshold conditions to determine the near threshold behavior.

The crack lengths were monitored by either physical examination with an optical microscope or by voltage changes in a direct-current potential drop (DCPD) system. For initial tests at $R=0.1$, the crack was wedged open at the minimum load, and optical surface measurements of crack length were performed to the nearest 0.0001" (0.025 mm) at a magnification of 100X. Later tests, including all environmental testing, utilized the in-situ DCPD method of crack length measurement. Procedures for the DCPD measurements are detailed in [6]. Environmental testing was performed in a vacuum tight chamber which was specifically designed for these tests. An atmosphere of saturated water vapor was achieved by bubbling inlet air through a water bath and verified by humidity measurements. Testing in dry nitrogen was performed by evacuating the test chamber and backfilling with ultrahigh-purity nitrogen gas passed through drierite. The chamber was continuously purged while maintaining a slightly positive internal gas pressure.

RESULTS

Static mechanical properties

The tensile properties of the two U6Nb compositions were measured in both the as-quenched condition and after aging at 200°C/2h. The first series of tensile specimens were heat treated in a vacuum furnace along with compact tension specimens 1AH to 6AH. The room temperature tensile properties for the series-1 heat treatments are summarized in Table II. The average values of yield strength, tensile strength, elongation to failure, and reduction-in-area at fracture for the as-received condition are 24.2 ksi (167 MPa), 122.6 ksi (845 MPa), 33 %, and 44 %, respectively, and for the aged condition 78.4 ksi (540 MPa), 129.3 ksi (892 MPa), 31 %, and 44 %, respectively. The difference in properties measured for compositions 'A' and 'B' differ by less than the specimen-to-specimen variations. The most significant difference between the as-received and aged conditions is the more than doubling of the yield strength (0.2 % offset criterion) after heat treatment. Conversely, the ultimate tensile strength increases mildly after heat treatment, and the tensile ductility is not significantly affected. The elastic modulus increases from 65 GPa (9.5 (10⁶) psi) in the as-received material to 77 GPa (11.2 (10⁶) psi) after 2 h at 200 °C.

The tensile stress/strain curves for the as-received and vacuum aged specimens are presented in Figure 2a,b. The shapes of the curves are complex, with initial yield regions followed by high rates of work hardening prior to a second yield region and with extensive uniform plastic deformation and significant localized plastic deformation (necking) prior to final failure. The principal effect of aging at 200 °C is to alter the material behavior at small strains. While aging more than doubles the 0.2% offset yield strength the second yield

stress increases only 16% from 725 MPa (105 Ksi) to 840 MPa (122 Ksi) and the tensile strength increases just 5%.

The sensitivity of the tensile properties to soaking time at 200°C was also evaluated by comparing 1 h and 2 h heat treatments in air using standardized methods. The stress vs. strain curves for specimens aged 1 h and 2 h at 200°C are compared in Figure 2c,d. As summarized in Table II, the yield strength increases from 63.3 ksi (436 MPa) after 1 hr at 200 °C to 67.7 ksi (467 MPa) after 2 hrs at 200 °C. The ultimate tensile strength, elongation to failure, and reduction-in-area at fracture were not measurably different.

The fracture toughness of the high-carbon composition (A) was measured in both the as-received and the aged (200 °C/2h) conditions using 0.5 inch thick compact tension specimens. With the exception of an aged specimen, the provisional values, J_Q , did not meet minimum specimen thickness requirements due to the high ratio of J_Q relative to the effective yield strength. These J_Q values therefore represent the resistance to extension of a sharp crack in section thicknesses less than or equal to 0.5 inch (12.7 mm) but not in larger section thicknesses. The J-integral toughness of the aged material exceeded that of the as-quenched material (1722 in-lbs/in² vs. 1215 in-lbs/in²) despite the higher 0.2% offset yield strength (79 ksi vs. 23 ksi).

The temperature dependence of the fracture toughness was evaluated through comparison tests at RT and -54°C [4]. While the toughness values are lower than those measured at LLNL, the toughness did not decrease between RT and -54°C. Subsequent tests at LLNL using newly implemented test methods resulted in J_Q values closer to those measured by Y-12 [4]. A summary of the fracture toughness measurements is presented in Table III. Despite the variations in toughness values, the resistance of U6Nb to static propagation of a sharp crack remained high

in all tests and the fracture surfaces were all ductile in appearance (see Microstructural Characterization section).

Fatigue Crack Initiation

The fatigue stress vs. fatigue life (S-N) characteristics of smooth-bar specimens aged at 200°C/2h are plotted in Figure 3. The fatigue strength decreases continuously for increasing fatigue lifetimes. This trend continues to at least 10^8 cycles with no indication of a fatigue endurance limit; fatigue failures were observed at lifetimes exceeding 10^8 cycles. Because of this absence of a fatigue endurance limit, the fatigue strength is reported for specific cyclic lifetimes. From Figure 3, the lower bound fatigue strength at 10^8 cycles is 36 ksi (248 MPa), increasing to 43.5 ksi (300 MPa) at 10^7 cycles. Average fatigue strengths are approximately 7 ksi (50 MPa) higher than the lower-bound values for these conditions.

The sensitivity of the fatigue strength to stress concentrations was evaluated for specimens containing a severe notch, and the test results are summarized in Figure 4. For a lifetime of 10^8 cycles, the notched-bar fatigue strength decreases to a lower-bound value of 11 ksi (75 MPa). The experimentally determined fatigue strength reduction factor due to the presence of the notch is therefore equal to 3.3. This is much lower than the theoretical stress concentration factor, K_t , of 11 calculated from elastic theory (see Exper. Methods Section).

The influence of mean tensile stresses on the smooth-bar fatigue strength were evaluated using axially-loaded specimens subjected to tension-tension cycles. For a mean load of 40 ksi (275 MPa), the fatigue strength at 10^7 cycles is reduced to 15 ksi (103 MPa), as shown in Figure 5. Under these conditions, the specimen-to-specimen variation in fatigue lifetime is much larger than for rotating beam tests at zero mean load. For a mean load of 70 ksi (483 MPa), which exceeds the bulk

0.2% offset yield strength of the material, the fatigue strength is further reduced to a minimum value of 6 ksi (41 MPa), as plotted in Figure 6. A summary of the influence of mean tensile loads on the fatigue strength of aged U6Nb for lifetimes of 10^7 cycles is plotted in Figure 7 and compared to three empirical relations developed for other material systems. The fatigue strength of 43.5 ksi (300 MPa) at zero mean load and 10^7 cycles is taken from the rotating beam data. Only the relation by Soderberg, which predicts zero fatigue strength at mean loads above the bulk yield strength of the material, is conservative in estimating the fatigue strength reductions due to mean stresses in U6Nb.

A summary of the fatigue strength data is presented in Table IV. The ratio of fatigue strength to tensile strength varies from 0.28 for smooth bar specimens to 0.09 for notched bars ($K_t=11$) at zero mean load, and 0.05 for smooth bars at mean loads of 70 ksi (483 MPa).

In addition to measurements of total fatigue lifetime, several of the tension-tension fatigue specimens were instrumented with a direct-current potential-drop (DCPD) system for early crack detection. Under an imposed constant current, the potential between copper collars fixed to the specimen was continuously monitored during the tests. The initiation of a fatigue crack results in an increase in the potential drop above baseline values. After first detection of a crack, the rate of voltage rise increases continuously until final failure occurs. Typical DCPD initiation data is plotted in Figures 8 a,b for mean stresses of 40 ksi (276 MPa), and 70 ksi (483 MPa), respectively. The crack initiation lifetime constituted a minimum of 85% and as much as 99.9% of the total fatigue life. The crack propagation lifetime can be calculated as the difference between the total lifetime and the crack initiation lifetime, and is included in a summary of the DCPD data in Table V. For similar loading conditions, the specimen to specimen variation in total lifetime is approximately 70, while the calculated fatigue propagation

lifetimes vary by less than a factor of 2. The large differences in fatigue life between these specimen can be attributed to variations in crack initiation lifetime and emphasize the statistical nature of fatigue crack initiation.

The source of fatigue crack initiation was identified through fractographic analysis of fractured smooth-bar S-N samples. In every case examined, the crack initiation event could be traced to an inclusion near the specimen surface. Two types of inclusions were identified using energy-dispersive x-ray (EDX) analysis in the SEM; niobium carbides, and uranium oxides. On a typical fracture surface from a rotating-beam S-N test such as shown in Figure 9, the fracture surface markings clearly point to a thumbnail-shaped region near the specimen surface as the crack initiation source. Closer examination identified inclusions at the edge of the thumbnail-shaped region as the crack initiation site, as shown in Figure 10. The crack first propagated to the specimen surface to produce the thumbnail-shaped region, and then propagated across the sample leaving classical fatigue striations in some areas, as shown in Figure 11.

Fatigue Crack Propagation

The fatigue crack growth rates (FCGR's) of aged U6Nb (200 °C/2h) vs. cyclic stress intensity range are summarized in Figure 12 for tests in laboratory air at a constant R-ratio of 0.1. Measurements from five 0.5" (12.7 mm) thick specimens are presented in this plot. The data at high stress-intensity-range (ΔK) was determined using K-increasing methods and overlaps the data at lower ΔK determined using K-decreasing methods. The FCGR behavior in the overlap region is similar irrespective of which of the two methods were used. Overall, there are three characteristic regions present in this data as plotted in Figure 12. At cyclic stress intensity ranges, ΔK , exceedingly approximately $20 \text{ MPa}\sqrt{\text{m}}$, the FCGR's increase rapidly with increasing ΔK , reaching a growth rate of $3(10^{-3})$

mm/cycle at a ΔK of $54 \text{ MPa}\sqrt{\text{m}}$. This low-cycle fatigue region is bordered by a mid-growth-rate region at lower ΔK 's, which can be characterized as low- n power-law behavior:

$$da/dN = C(\Delta K)^n \quad [\text{Eq. 1}]$$

where da/dN is the cyclic crack growth rate, C is a constant, ΔK is the cyclic stress intensity range and n is the Paris-Law exponent. This exponent, n , decreases from a value of 3.9 in the low-cycle fatigue region to a value of approximately 1.5 in the mid-growth-rate region. With further decreases in ΔK , a threshold region is normally expected at which growth rates rapidly decrease to below 10^{-7} mm/cycle. The start of the threshold transition region is indicated at the lowest ΔK test values.

For comparison purposes, the FCGR behavior of as-quenched U6Nb was also evaluated. The FCGR's for constant $R=0.1$ and K -decreasing test methods are compared in Figure 13. The FCGR's of the as-quenched material are similar within the range of overlap but can be characterized as having a lower Paris Law exponent, n (see Equation 1). For a ΔK of $10 \text{ MPa}\sqrt{\text{m}}$, the FCGR's are slightly lower than for the aged condition, while the FCGR's are slightly higher in the near-threshold region. The threshold ΔK for as-quenched material can be estimated as no lower than $3.4 \text{ MPa}\sqrt{\text{m}}$, which is based on three data points defined by their maximum undetected growth rates.

Crack Closure Phenomena. The magnitude of the fatigue threshold is well known to be sensitive to many material and environmental variables, especially those which induce crack closure effects. Crack closure is a phenomenon which occurs upon unloading where-in previously open crack faces make contact at loads greater than the minimum cyclic load, even though the minimum cyclic load is tensile. Because this phenomenon reduces the cyclic load range at the crack tip, the effective ΔK at the crack tip is also reduced. Crack closure can be detected

by a change in the specimen compliance (slope of the load vs. displacement curve). The compliance of the as-quenched specimen was measured after 10^6 cycles at a ΔK of $3.1 \text{ MPa}\sqrt{\text{m}}$ and $R=0.1$ ($K_{\text{max}}=3.5$ and $K_{\text{min}}=0.35 \text{ MPa}\sqrt{\text{m}}$). Opening of the crack faces upon reloading was identified by a decrease in compliance at a load of 9.5 lbs or K of $1.1 \text{ MPa}\sqrt{\text{m}}$. Because this crack-opening K value is greater than the applied K_{min} , the effective ΔK is reduced to $K(\text{max}) - K(\text{opening})$ or $2.4 \text{ MPa}\sqrt{\text{m}}$.

To determine the influence of fatigue crack length on crack closure, the length of fatigue crack surfaces able to mechanically interact was incrementally shortened by widening the crack mouth with an abrasive blade. The length of crack behind the crack tip available for mechanical crack closure was therefore incrementally reduced. The original fatigue length of 0.784" (20 mm) was shortened to 0.284" (7.2 mm) and 0.173" (4.4 mm), successively. The corresponding crack opening loads decreased from 9.5 lbs ($K=1.13 \text{ MPa}\sqrt{\text{m}}$), to 5.5 lbs ($K=0.65 \text{ MPa}\sqrt{\text{m}}$), and finally 1.85 lbs ($K=0.22 \text{ MPa}\sqrt{\text{m}}$). This final K -opening value is below K_{min} , and no closure effects are expected in this condition. Subsequent FCGR's with the shortened effective crack length were similar to those measured at high crack opening loads, however, as shown in Figure 13.

One common source of crack closure relevant to these tests is that of crack surface roughening. The typical fatigue crack path in the near-threshold region is severely branched, as shown in Figure 14, and the fracture surface is highly roughened by the numerous changes in crack propagation direction. The large number of secondary fatigue cracks form a "lightning-bolt" crack morphology. The crack propagation direction is dominantly transgranular in nature but with preferred directions within each grain, as evidenced by numerous parallel cracks in some individual grains. From observations on several test specimens, the crack branching consistently initiates only when ΔK decreases to near $10 \text{ MPa}\sqrt{\text{m}}$.

Once started, the crack branching continues for all lower values of ΔK . In regions exhibiting crack branching, cracks emanating from inclusion particles are commonly observed, and the principal crack paths preferentially intersect these particles. The fracture surfaces are also darkened during fatigue testing and the extent of darkening increases with decreasing fatigue crack growth rates. This surface darkening is primarily attributed to surface oxidation during fatigue crack propagation.

The potential role of residual stresses within the processed plates on branching of the crack away from the direction of highest applied stress was evaluated by testing specimens with a reduced thickness of 0.1 inch (2.5 mm). Crack branching was again first observed for a ΔK of approximately $10 \text{ MPa}\sqrt{\text{m}}$ and continued with decreasing ΔK . The crack growth rate behavior is compared with that for thicker specimens in Figure 15. The fatigue threshold is reduced to approximately $3.2 \text{ MPa}\sqrt{\text{m}}$ with an accompanying plateau in growth rate at $4(10^{-6}) \text{ mm/cycle}$ for ΔK 's between approximately 3.5 and $6.5 \text{ MPa}\sqrt{\text{m}}$. This fatigue threshold value is close to that determined for as-quenched U6Nb with a specimen thickness of 0.5 inch (12.7 mm). Compliance measurements were performed after $2(10^6)$ cycles at a ΔK of $3.1 \text{ MPa}\sqrt{\text{m}}$ ($K_{\text{max}}=3.44$ and $K_{\text{min}}=0.344 \text{ MPa}\sqrt{\text{m}}$) with no measurable crack growth. A change in compliance at 10.3 lbs ($K=1.21 \text{ MPa}\sqrt{\text{m}}$) was determined from the compliance measurements (load/displacement), which identifies the crack opening load. Because K_{opening} is above the applied K_{min} , the effective fatigue cycle at the threshold is $(K_{\text{max}})-(K_{\text{opening}})$ or $2.2 \text{ MPa}\sqrt{\text{m}}$.

Constant K_{max} Fatigue Crack Propagation. While crack closure effects can increase the fatigue threshold, crack closure is often absent in service when short cracks or high load ratios (R) are present. In order to evaluate the fatigue crack propagation behavior for high mean tensile loads and without crack closure,

testing was performed for a loading cycle in which the maximum stress intensity K_{max} is kept constant. In this case, the minimum load increases as the cyclic stress intensity range decreases. Thus, as the fatigue threshold is approached, the minimum load increases toward the load at K_{max} , the R-ratio approaches unity, and crack closure is prevented.

The FCGR's at a constant K_{max} of $14.6 \text{ MPa}\sqrt{\text{m}}$ are compared with those at constant R-ratio of 0.1 in Figure 16 for a 0.1 inch (2.5 mm) thick sample. For constant K_{max} conditions, the fatigue threshold is shifted to lower ΔK and the low-n plateau in growth rates between ΔK 's of $3 \text{ MPa}\sqrt{\text{m}}$ and $10 \text{ MPa}\sqrt{\text{m}}$ is shifted up to growth rates of approximately $6 (10^{-6}) \text{ mm/cycle}$. The growth rates begin to decrease again for ΔK 's below approximately $3 \text{ MPa}\sqrt{\text{m}}$ and the fatigue threshold (defined by growth rates of 10^{-7} mm/cycle) is decreased to $0.95 \text{ MPa}\sqrt{\text{m}}$. The onset of crack branching continues to occur for ΔK 's less than $10 \text{ MPa}\sqrt{\text{m}}$.

The FCGR behavior at higher mean loads was evaluated by increasing K_{max} from $14.6 \text{ MPa}\sqrt{\text{m}}$ to $30.5 \text{ MPa}\sqrt{\text{m}}$. The data is plotted in Figure 17. The FCGR's are similar for the two conditions in the mid-growth-rate regime but increase rapidly for the higher K_{max} condition when ΔK exceeds approximately $10 \text{ MPa}\sqrt{\text{m}}$.

Additional fatigue data was acquired at a K_{max} of $30.5 \text{ MPa}\sqrt{\text{m}}$ for a specimen thickness of 0.5 inch (12.7 mm). This and all subsequent FCGR data was acquired using a direct-current potential-drop (DCPD) system to measure the fatigue crack length in-situ. DCPD calibration data and test methods are included in a report on near-threshold fatigue testing [6]. As shown in Figure 18, the mid-growth-rate plateau has disappeared and the growth rates decrease at a nearly constant power law exponent of 2.2 for ΔK 's below approximately $10 \text{ MPa}\sqrt{\text{m}}$. The fatigue threshold is depressed from $0.95 \text{ MPa}\sqrt{\text{m}}$ for a constant K_{max} of $14.6 \text{ MPa}\sqrt{\text{m}}$ to below $0.5 \text{ MPa}\sqrt{\text{m}}$ for these conditions. Comparing data at a constant K_{max} of 30.5

$\text{MPa}\sqrt{\text{m}}$ to that measured at a constant R-ratio of 0.1 in Figure 19, the FCGR's are significantly higher for the conditions of higher mean stress (constant K_{max}) at all values of ΔK .

Environmental Effects. A final variable explored in this investigation is that of the fatigue environment. To evaluate the contributions of fatigue environment to FCGR's, comparison tests were performed in an inert environment of dry nitrogen, and an aggressive environment of air with 100% relative humidity (RH). Previous tests were all performed in laboratory air with RH ranging between 40 and 60%. The FCGR's in dry nitrogen are compared to lab air for constant K_{max} values of $14.6 \text{ MPa}\sqrt{\text{m}}$ in Figure 20. A FCGR plateau is clearly present in both environments, but at lower growth rates in dry nitrogen than in lab air, e.g. 1 to 2 (10^{-6}) mm/cycle vs. 7 (10^{-6}) mm/cycle, respectively. The fatigue threshold defined as a FCGR of 10^{-7} mm/cycle increases to $1.5 \text{ MPa}\sqrt{\text{m}}$ for dry nitrogen relative to $0.9 \text{ MPa}\sqrt{\text{m}}$ for lab air. However, the fatigue threshold does not asymptote at this value and decreases to $0.9 \text{ MPa}\sqrt{\text{m}}$ for growth rates of 10^{-8} mm/cycle in dry nitrogen.

At higher mean stresses with K_{max} equal to $30.5 \text{ MPa}\sqrt{\text{m}}$, the FCGR's increase relative to behavior at the lower K_{max} value. Like tests in lab air, the FCGR's in dry nitrogen decrease in a logarithmic manner with decreasing ΔK , without a clear mid-growth-rate plateau, as shown in Figure 21. Unlike in lab air, where no threshold region is observed, the FCGR's in dry nitrogen decrease rapidly from 3×10^{-7} mm/cycle at a ΔK of $0.9 \text{ MPa}\sqrt{\text{m}}$ to below 10^{-8} mm/cycle at approximately $0.8 \text{ MPa}\sqrt{\text{m}}$. For comparison, the FCGR in lab air is 3×10^{-7} mm/cycle at a ΔK of $0.55 \text{ MPa}\sqrt{\text{m}}$.

Fatigue crack propagation behavior in a more aggressive environment of moisture-saturated air (100RH) is plotted in Figure 22. The FCGR's decrease continuously following logarithmic behavior ($da/dN = C\Delta K^n$) where $n=3.3$ for a

constant K_{max} value of $14.6 \text{ MPa}\sqrt{\text{m}}$ and $n=2.9$ for a constant K_{max} value of $30.5 \text{ MPa}\sqrt{\text{m}}$. For both values of K_{max} there is no mid-growth-rate plateau and no indication of an asymptotic threshold stress intensity at FCGR's as low as 10^{-8} mm/cycle. Despite the absence of these features, the FCGR's in 100RH air are significantly lower than in standard lab air tests for ΔK 's less than $10 \text{ MPa}\sqrt{\text{m}}$.

As-quenched U6Nb was also tested in conditions of dry nitrogen and lab air at a constant K_{max} of $14.6 \text{ MPa}\sqrt{\text{m}}$ for comparison with aged material. The near-threshold fatigue behavior is similar for both environments with a threshold stress intensity of approximately $0.95 \text{ MPa}\sqrt{\text{m}}$ at growth rates of 10^{-7} or 10^{-8} mm/cycle, as shown in Figure 23. A mid-growth-rate plateau at FCGR's of 2 to 4×10^{-6} mm/cycle is more pronounced in the dry nitrogen environment, extending to ΔK 's above $10 \text{ MPa}\sqrt{\text{m}}$. The FCGR's in lab air increase above the plateau rates for ΔK 's greater than approximately $6 \text{ MPa}\sqrt{\text{m}}$. In dry nitrogen atmospheres, the fatigue crack propagation behavior of as-quenched material is remarkably similar to that of U6Nb aged $200^\circ\text{C}/2\text{h}$, as shown in Figure 24.

Microstructural Characterization

The general microstructure of U6Nb aged at 200°C for 2 h is shown in Figure 25. It consists of equiaxed prior- γ grains which transform to α'' martensite upon cooling from above approximately 640°C [7]. With proper illumination, additional substructure can be seen within the prior- γ grains. In some regions, a subgrain network is visible within the prior- γ grains, and in other regions the finely banded α'' martensite can be seen, as in Figure 26. In addition to the banded martensite, acicular martensite is observed in isolated areas at relatively low volume fraction. There are also two types of inclusion particles present in the material, uranium oxide and niobium carbide, both of which are visible in Figure 25. These are the same inclusions identified by Chapman in a quantitative evaluation of inclusion types and distributions in 57 heats of U6Nb produced at Y-

12 [8]. The maximum inclusion sizes reported by Chapman are similar to that observed in the material evaluated in this investigation. Carbides dominate the high end of the inclusion size distribution. The mean of the maximum size carbide found in each heat (from several 1.9 mm² areas selected for evaluation) was 200 square microns ($1.6(10^{-2})$ mm equivalent spherical diameter). The overall maximum sized carbide was 650 square microns ($2.9(10^{-2})$ mm equivalent spherical diameter), as can be seen in the plot of Figure 27. The carbide distributions were similar for both center and edge sections, as illustrated in the distributions plotted in Figure 28. The role of these inclusions as crack initiation sites make their sizes relevant to estimates of maximum initial flaw sizes inherent to U6Nb components.

A typical fracture surface produced by tensile overload is shown in Figure 29. The fracture mode is ductile void coalescence which is generally a high-energy-absorption process. The nucleation of voids at inclusion particles is verified by the presence of inclusion particles within the voids on the fracture surface. Many of these inclusions, especially the larger ones, are fractured. These fractured inclusions were previously shown to initiate fatigue failure in smooth-bar fatigue specimens.

The relationships between fatigue crack propagation directions and the microstructures were evaluated from fracture profiles and fracture surface examinations. Fatigue crack propagation occurs exclusively along transgranular paths for all testing conditions. The presence of numerous secondary cracks in the near threshold region at ΔK 's of less than $10 \text{ MPa}\sqrt{\text{m}}$ was commonly found. In Figure 14, the primary fatigue crack path as well as numerous secondary crack paths can be seen in the fracture profile of a specimen tested in lab air at a constant R-ratio of 0.1. Secondary cracks exist as both as connected branches to the primary crack and as disconnected segments. The disconnected segments are typically

parallel to the connected cracks. All fatigue cracks are characterized by jogs in the crack propagation direction, with several jogs within single grains being most common. These jogs are often parallel to each other within a single prior- γ grain, showing preferential crystallographic directionality. The crack paths also preferentially intersect large inclusion particles although the numerous crack branches and segments are not clearly accounted for solely by these particles. When the test atmosphere is changed from laboratory air to 100RH air or dry nitrogen, the crack morphologies remain similar and continue to contain secondary cracking, as shown in Figure 30 and 31, respectively.

The fracture surfaces of fatigued specimens differ substantially from those produced during tensile overload fracture. The fracture surfaces produced by these two loading conditions are compared in Figure 32 for tests in lab air and fatigue loading at a constant K_{max} of $30.5 \text{ MPa}\sqrt{\text{m}}$. The macroscopic crack branches present in the fatigue zone are extinguished upon application of a tensile overload and the fracture surface becomes relatively planar. At higher magnification, the fatigue fracture surface can be seen to be composed of numerous cleavage-like facets whereas the tensile overload is completely covered with ductile voids. The influence of test environment on the fatigue fracture surface characteristics are illustrated in Figure 33. The macroscopic crack surface roughness is reduced relative to that in 100RH air and increases in 100RH air with an increase in K_{max} from $14.6 \text{ MPa}\sqrt{\text{m}}$ to $30.5 \text{ MPa}\sqrt{\text{m}}$. The fracture surface characteristics in dry nitrogen at a constant K_{max} of $14.6 \text{ MPa}\sqrt{\text{m}}$ are further illustrated in Figures 34 and 35. Fine-scale features nearly perpendicular to the macroscopic fracture plane indicate crack branching and a prominent cracked inclusion particle is present in each figure. Back-scattered electron images more clearly reveal the extensive degree of secondary cracking in dry nitrogen as shown in Figure 36 for a constant K_{max} of $30.5 \text{ MPa}\sqrt{\text{m}}$.

In an atmosphere of 100RH air, cleavage-like fracture surfaces are produced in fatigue at constant K_{max} values of $14.6 \text{ MPa}\sqrt{\text{m}}$ and $30.5 \text{ MPa}\sqrt{\text{m}}$, as shown in Figures 37 and 38, respectively. The facet size and the macroscopic planarity of the fracture surfaces both increase at the higher K_{max} testing condition. Again, prominent fractured inclusion particles are present in each Figure.

The presence of chemical heterogeneity in the form of Nb banding has been reported in U6Nb by other investigators [9,10]. In this study, the distribution of Nb perpendicular to the primary fatigue crack and crossing secondary cracks was examined by electron probe microanalysis (EPMA). This direction is parallel to the original rolling plane during plate fabrication. As shown in Figure 39, no Nb segregation was detected in this direction over a distance of 60 microns.

DISCUSSION

Static Mechanical Properties

The alloy U6Nb has a number of characteristics which are desirable for a structural material. Among these are a moderately high tensile strength and good toughness in the as-quenched or underaged conditions (e.g. quenched and aged $200^{\circ}\text{C}/2\text{h}$). The yield strength is raised substantially with quench and age heat treatments which cause fine-scale precipitation of a Nb-rich phase [1]. The optimum material condition depends on the application, but for maintenance of good corrosion resistance, U6Nb is normally used either in the as-quenched condition or in the underaged condition. The underaged heat treatment limits the amount of Nb precipitation and the resulting strengthening increment but prevents large decreases in corrosion resistance. Underaging also adds

dimensional stability to machined components by preventing the α''_a to γ_0 transformation [3].

The most common aging heat treatment for U6Nb is 200°C/2hrs. This heat treatment typically produces properties which meet the requirements of D.O.E. specification RM254952 which a yield strength between 52.2 ksi (360 MPa) and 70.4 ksi (485 MPa), a minimum tensile strength of 111.5 ksi (769 MPa), and minimum elongation of 25% at room temperature are specified. All of the material tested in this investigation meets these requirements with the exception of high yield strengths in the initial series of heat treatments in vacuum. As summarized in Table I, the yield strength more than doubles after 1 hr at 200°C and increases another 7% after 2 hrs at 200°C. Vacuum heat treatments are inherently difficult to control at low temperatures due to negligible convective heat transfer and the resulting extension of heating and cooling times resulted in extended aging and associated strengthening. Although the yield strength increased for the vacuum treated specimens, the tensile strength, elongation, and reduction-in-area at fracture all remained largely unaffected. It is of interest to note that the major changes in the tensile stress/strain curves after heat treatment at 200°C are limited to the flow stresses at low-strains. As substantiated in Figure 2, low temperature aging inhibits initial yielding but does not largely affect the flow stress at strains greater than 5 percent. Similar results have been found by others [11].

Fracture toughness testing has verified the high resistance to unstable crack extension at RT in both the as-quenched and in the aged conditions and at -54°C for aged material. Although uniaxial tensile data predicts a large decrease in ductility between RT and -50°C [12], the toughness values determined here at RT and -54°C are nearly equivalent [4]. In Charpy impact tests between +100°C and -100°C by Anderson [13], the absorbed impact energy decreased with decreasing test temperature but remained relatively high and the fracture mode remained ductile

void coalescence at the lowest test temperature. Between RT and -50°C, the absorbed impact energy decreased from 34 J to 27 J. These results confirm that the resistance of aged U6Nb to unstable crack propagation remains high at -54°C.

Valid fracture toughness measurements were only obtained for the aged 0.5 inch (12.7 mm) thick samples due to dimensional requirements for small-scale yielding in J-integral tests (ASTM E813). The specimens used in the Y-12 tests at RT and -54°C were 0.4 inch (10 mm) square bars. While specimens of greater thickness may exhibit a lower resistance to unstable crack extension, the behavior of components of equal or lesser constraint (e.g. thickness) is confirmed. The early test measurements which gave the higher toughness values (1A, 2A, and 1AH) used a procedure which has subsequently been identified in round-robin testing as overestimating the toughness by approximately 25%. Modified test procedures were used for specimen 2AH, and the toughness value more closely matches the results by Y-12, as shown in Table IV.

A typical application of the fracture toughness data is to calculate the critical crack length required for unstable crack extension. For a through-thickness crack from the edge of a semi-infinite plate, the critical crack length can be calculated from the relation:

$$K_c = 1.12\sigma\sqrt{\pi a_c} \quad [\text{Eq. 2}]$$

where K_c is the critical stress intensity, σ is the applied stress, and a_c is the critical crack length. Using a worst-case applied load equal to the yield strength of aged U6Nb (68 ksi (470 MPa)), the critical crack length is 0.35 inch. Such critical crack length determinations can also be useful for prediction of the extent of available fatigue crack growth prior to unstable crack extension for a given maximum applied stress.

Fatigue Crack Initiation Behavior

Stress vs. life fatigue tests have been used to determine the fatigue strength for both smooth and notched specimens, to evaluate the effects of mean tensile loads on the fatigue strength, to identify the source of fatigue crack initiation, and to measure crack initiation lifetimes through novel use of the direct-current potential-drop (DCPD) technique.

It is evident from the smooth-bar data in Figure 3 that U6Nb does not exhibit a fatigue endurance limit. The fatigue strength continues to decrease for increasing cyclic lifetimes to beyond 10^8 cycles, unlike steels for which the fatigue strength is typically constant beyond approximately 10^6 cycles. This lack of a fatigue endurance limit in U6Nb is similar to the behavior of several other nonferrous materials such as aluminum and copper alloys [14]. The fatigue strength used in design decisions must therefore be made based on the cyclic lifetime requirements. For 10^8 cycles, the lower-bound fatigue strength is 36 ksi (248 MPa). The ratio of fatigue strength to tensile strength under these conditions is 0.28, which is well below the trend for steels of 0.5.

A microstructural factor with a significant influence on the fatigue strength is that of the niobium carbide and uranium oxide inclusion particles. These particles have been identified as the source of fatigue crack initiation in smooth-bar fatigue specimens, and are also void initiation sites in tensile overload failures. Although mixed inclusion types are often clustered together, the niobium carbides appear to be the most common source of crack initiation. The niobium carbides also constitute the largest inclusion particles found in this material. For these reasons, high carbon contents can be expected to degrade the mechanical properties, especially with respect to fatigue crack initiation. The bulk of the testing in this investigation was performed on an alloy with a carbon

content of 100 ppm (wt.) which is near the maximum level produced by Y-12 as WR6854 plate [16] even though D.O.E. specification RM254952 allows up to 200 ppm carbon.

Stress concentrations in the form of a machined notch also caused significant reductions in fatigue strength. The test results for a severe notch with theoretical stress concentration factor of 11 are summarized in Figure 4. The effective reduction in fatigue strength of 3.3 relative to levels for smooth-bar specimens is much lower than that predicted by elastic theory. This can be attributed in part to the low yield to tensile strength ratio and the role of plastic deformation in limiting the stress concentration. An additional factor is the presence of inclusions which provide intrinsic stress concentration sites and lower the smooth bar fatigue strength which is used as a baseline value.

The sensitivity of the fatigue strength to mean tensile stresses was evaluated for mean stresses of 40 ksi (275 MPa), or approximately two-thirds of the 0.2% offset yield strength, and just above the yield strength at 70 ksi (483 MPa). The comparisons of mean loading effects in U6Nb with empirical models proposed by Gerber, Goodman, and Soderberg (Figure 7A) show that U6Nb is relatively sensitive to mean loading. For most material, the fatigue properties fall between the Gerber and Goodman curves [17]. The U6Nb data compares closest with the Soderberg curve in which the fatigue strength approaches zero at the yield strength of the material. The tests at 40 ksi (275 MPa) had more scatter than tests in the other conditions, with specimens either failing at less than 10^6 cycles or beyond 10^7 cycles. This could be related to early crack initiation at high mean loads for specimens containing statistically large inclusions.

The crack initiation lifetime is expected to be statistical in nature when the initiation event occurs at worst-case flaws which are randomly distributed, as for inclusion particles. A distinct measurement of crack initiation lifetime vs. total

lifetime was performed for several of the same axially-loaded tensile specimens used to measure mean loading effects. A novel use of a DCPD system was applied as a method of early crack detection; crack initiation was detected from an increase in electrical potential due to a local reduction in conducting cross-section associated with a fatigue crack [15]. The maximum undetected crack length can be conservatively estimated from the calibrations performed on the compact tension (CT) specimens (Note: the conducting cross sections of the CT specimens are larger than for the axially-loaded fatigue specimens). For the CT specimens, a voltage resolution of approximately 5 microvolts yields a minimum crack size detectability of 2×10^{-4} inch ($5(10^{-3})$ mm) and this voltage increase was used as a basis for the crack initiation determinations. The time to initiate a detectable crack accounts for up to 99.9 percent of the total lifetime in the mean tensile loaded specimens. It is of interest to note from Table V that although the total fatigue lifetimes vary by a factor of 70 for identical loading conditions the calculated crack propagation lifetime varies by less than a factor of 2, as expected. Due to a requirement for tension-tension load cycles, crack initiation lifetimes could not be measured for specimens tested at zero mean load.

Because of the statistical nature of crack initiation lifetimes, the lower bound values for fatigue strengths are recommended as the appropriate design criteria for high confidence levels. The maximum observed inclusion sizes of approximately $3(10^{-2})$ mm (0.0012 inch) equivalent spherical diameter also give a basis for comparison of worst-case intrinsic flaws relative to flaws introduced by material processing or machining. The surfaces of the smooth bar specimens evaluated here were prepared by methods typical of machining practice for component fabrication as detailed in the procedures section and should therefore represent actual service conditions. A more conservative approach to fatigue design than the use of S-N test data is to assume the presence of a worst-case pre-existing flaw

in the material and to calculate the fatigue lifetime based on fatigue crack propagation data as discussed below.

Fatigue Crack Propagation

The fatigue crack propagation (FCP) behavior of U6Nb shares many of the characteristics common to typical engineering materials. First, the dependence of fatigue crack growth rates (FCGR's) on the cyclic stress intensity range has three characteristic regimes. These regimes can be compared by fitting the data to Eq. 1 and determining the power-law exponent, n . For standard test conditions of $R=0.1$ and laboratory air, an n -value of 3.9 was measured in the low-cycle fatigue region at high ΔK 's (region III), decreasing to an n -value of approximately 1.5 in the mid-growth regime (region II). As ΔK decreased further, the growth rates decreased asymptotically, corresponding to a large n -value. Secondly, the FCGR's for these conditions are placed within the trends for other structural materials for modulus compensated ΔK 's, as shown in Figure 40. Thirdly, the near-threshold fatigue behavior is modified by the presence of crack closure under standard test conditions. Crack closure decreases the effective ΔK and therefore increases the apparent fatigue threshold. This investigation has shown that the fatigue threshold is not single-valued, but that it is influenced by both load ratio (R -value) and test environment as summarized in Table VI.

Mean Loading Effects. The effects of mean tensile loads on the fatigue crack propagation behavior was evaluated through the use of loading cycles with constant K_{max} values at two levels. With a constant value of K_{max} , the R -ratio increases toward unity as the imposed ΔK decreases and should therefore prevent or limit the extent of crack closure. For a constant K_{max} of $14.6 \text{ MPa}\sqrt{\text{m}}$, the R -ratio increases from an initial value of 0.1 to a value of 0.93 at a ΔK of $1 \text{ MPa}\sqrt{\text{m}}$. The fatigue threshold decreases from $3.2 \text{ MPa}\sqrt{\text{m}}$ for constant $R=0.1$ to $0.9 \text{ MPa}\sqrt{\text{m}}$

for a constant K_{max} of $14.6 \text{ MPa}\sqrt{\text{m}}$, as shown in Figure 16. An increase in K_{max} to $30.5 \text{ MPa}\sqrt{\text{m}}$ resulted in a further decrease of the fatigue threshold to below $0.6 \text{ MPa}\sqrt{\text{m}}$. This behavior cannot be explained solely on the basis of crack closure and suggests the contribution of an environmentally-assisted fatigue mechanism at high mean tensile loads.

Environmental Effects. A contribution of environmentally-assisted fatigue (EAF) crack propagation was confirmed through comparative tests in inert and aggressive environments. The fatigue properties in an inert environment of dry nitrogen can be considered to provide baseline fatigue behavior in the absence of EAF. While EAF has not been previously investigated in U6Nb, a number of investigators have evaluated the static susceptibility to environmentally-assisted cracking (EAC) which is also known as stress corrosion cracking (SCC) [2, 18-21].

There is general agreement among these investigators that the as-quenched material possesses the best environmental resistance. The general corrosion resistance remains high for material aged $200^\circ\text{C}/2 \text{ hrs}$ but decreases substantially when aged to peak strength. Likewise, the resistance to EAC is decreased by aging. The critical stress intensity for stress corrosion cracking, K_{IEAC} in 100% RH air has been reported [20] to decrease from $24 \text{ MPa}\sqrt{\text{m}}$ for as-quenched U6Nb to $17 \text{ MPa}\sqrt{\text{m}}$ after aging at $200^\circ\text{C}/2 \text{ hrs}$, and to $9 \text{ MPa}\sqrt{\text{m}}$ after aging at $360^\circ\text{C}/1.2 \text{ hrs}$. The maximum susceptibility to EAC is reported to occur for aging treatments between 400°C and 450°C [21]. The susceptibility to EAC is also highly dependent on the environmental conditions. K_{IEAC} in as-quenched material has been reported as $80 \text{ MPa}\sqrt{\text{m}}$ in 50% RH air, decreasing to $30 \text{ MPa}\sqrt{\text{m}}$ in 100% RH air [19]. While increasing moisture contents increase EAC susceptibility in H_2 or O_2 environments, EAC is reported to be absent in N_2 environments [20].

Based upon the data in the literature, EAC is only expected in the fatigue tests performed in 100RH air at K_{max} values of $30.5 \text{ MPa}\sqrt{\text{m}}$. As shown in Figure 22,

the FCGR behavior is similar for K_{max} values of both $30.5 \text{ MPa}\sqrt{\text{m}}$ and $14.6 \text{ MPa}\sqrt{\text{m}}$ in 100RH air and both exhibit an absence of a clear fatigue threshold. However, the FCGR's are reduced relative to values in inert atmospheres within the mid-growth-rate region. The source of these reduced growth rates can only be speculated but may be related to the increased severity of off-axis crack growth and associated crack tip stress reductions or the presence of crack closure due to increased amounts of roughening and oxide debris.

The stress intensities during fatigue testing in all other conditions evaluated in this investigation were below the reported critical stress intensity, K_{IEAC} . The increase in FCGR's in 50RH and 100RH air relative to dry nitrogen are attributed to true environmentally assisted fatigue (EAF) in which a synergism occurs between the environmental and fatigue contributions to crack growth. Similar behavior has been observed in aluminum alloys which are susceptible to EAC such as alloy 7039-T651 [22]. In these alloys fatigue crack propagation is accelerated in environments that accelerate EAC even though the stress intensities are below K_{IEAC} .

Conversely, there is no clear environmental effect on the near-threshold fatigue behavior of as-quenched U6Nb for atmospheres of dry nitrogen and lab air at a constant K_{max} of $14.6 \text{ MPa}\sqrt{\text{m}}$ although the growth rates do increase in lab air at the highest ΔK values tested (see Figure 23). The differences in FCGR's for aged and as-quenched material are also minor for these test conditions, as illustrated by comparing Figures 20 and 23.

Common to much of the FCGR data is a plateau region where the FCGR is relatively independent of ΔK and "n" approaches unity. This behavior occurs in a region between the fatigue threshold and ΔK 's of approximately $10 \text{ MPa}\sqrt{\text{m}}$. The exceptions are for aggressive environments at moderate values of K_{max} ($14.6 \text{ MPa}\sqrt{\text{m}}$) and for all environments at high values of K_{max} ($30.5 \text{ MPa}\sqrt{\text{m}}$) for

which the plateau is absent. While a decrease in the fatigue threshold occurs at high values of K_{max} and in aggressive environments, the plateau behavior is replaced by FCGR's that are below those for conditions exhibiting plateau behavior. As discussed above, this decrease in FCGR's in aggressive environments is contrary to traditional behavior and may be related to changes in crack path or due to crack closure from oxide debris.

The source of crack branching in these alloys is also open to speculation. It is generally agreed that microstructural effects become important for ΔK 's in which the cyclic plastic zone size decreases below the size of a controlling microstructural feature. Similar to the behavior observed here, crack branching occurs in a titanium alloy below a critical ΔK [17]. In the case of the titanium alloy, crack branching occurs when the cyclic plastic zone size decreases below the Widmanstätten packet size. For U6Nb, the onset of crack branching occurs at ΔK 's of approximately $10 \text{ MPa}\sqrt{\text{m}}$, which corresponds to a cyclic plastic zone size of 0.02 mm, calculated from the relation [17]:

$$r_c = 0.033 (\Delta K_T / \sigma_{ys})^2 \quad [\text{Eq. 3}]$$

where r_c is the cyclic plastic zone length above the Mode I plane, ΔK_T is the ΔK value at the fracture transition, and σ_{ys} is the cyclic yield strength, estimated here as equivalent to the monotonic yield strength of 68 ksi (469 MPa). Both the prior- γ grain size (0.06 mm mean linear intercept grain size) and the mean inclusion spacing (estimated as 0.07 mm) are significantly larger than r_c . The fracture surfaces indicate, however, that while jogs in crack path occur at martensite boundaries within the prior- γ grains (such as those shown in Figure 26), the major crack branching occurs on a scale larger than the prior- γ grain size in U6Nb. The obvious microstructural feature present on a scale where major crack branches occur (e.g. between 0.1 and 1 mm) is macrosegregation of Nb. The presence of Nb macrosegregation in U6Nb castings is not avoidable [9,10] and has been measured

with a band spacing of 5 mm in the as-cast condition and reduced to 0.6 mm after rolling into plate [9]. It is speculated that although Nb banding was not detectable within the rolling plane as measured in this study, that it still exists in the through thickness direction, and causes the preferential crack paths due to associated variations in strength and environmental resistance and from carbide banding.

Fatigue Design

The application of fatigue crack propagation data to component design may follow either of two general approaches. The first approach is to limit the applied cyclic stress intensity range to levels below the fatigue threshold. In making this assessment, the experimental test conditions used in determining the fatigue threshold value must incorporate the anticipated service conditions of the component. Thus for conditions where high mean tensile loads (high R-ratios) or short cracks limit crack closure, threshold measurements made under these same conditions should be used. The fatigue threshold measurement represents a stress intensity range at which the fatigue crack growth rate is below a critical value (reported here as 10^{-7} mm/cycle unless otherwise noted). In most test conditions, an asymptotic threshold behavior was measured but in some cases, such as in 100RH air, the power-law behavior extends to below crack growth rates of 10^{-8} mm/cycle. In using the fatigue threshold as a design parameter it is important to consider potential crack growth increments due to the finite growth rates, especially with non-asymptotic thresholds. The potential for accelerated propagation of short cracks should also be incorporated [23]. In addition, environmentally assisted fatigue crack growth rates are typically dependent on the cyclic waveform and frequency. Accelerated environmental damage is expected with an increase in the period at high stress and with low loading frequencies.

A second design method is to calculate the fatigue life by integrating a growth rate law fit to the measured data. The most commonly used growth rate law (Eq. 1) assumes power-law behavior with a constant exponent, n . Of course, the data can be fit to many additional mathematical forms which may more closely match the data over the desired range of ΔK . To assess the fatigue life using this method, one must make estimates of the initial crack length, the loading history, the stress intensity due to component and crack geometry, the appropriate growth rate law, and the critical stress intensity for unstable crack growth. A simple example is given below.

First we assume a component and crack geometry which approximates a single-edge cracked plate for which:

$$K = \sigma \sqrt{\pi a} \quad [\text{Eq. 4}]$$

where K is the Mode I stress intensity, σ is the applied stress, and " a " is the crack length. Next we assume an alternating stress of $\Delta\sigma = \sigma_{\max} - \sigma_{\min}$ and the growth rate law of Eq. 1. Substituting for ΔK and integrating from an initial crack size a_0 to a final crack size a_f for a total life of N_f fatigue cycles, then:

$$\int_{a_0}^{a_f} \frac{da}{a^{n/2}} = C \Delta\sigma^n \pi^{n/2} \int_0^{N_f} dN \quad [\text{Eq. 5}]$$

and,

$$N_f = \frac{2}{(n-2) \Delta\sigma^n \pi^{n/2}} \left[\frac{1}{a_0^{(n-2)/2}} - \frac{1}{a_c^{(n-2)/2}} \right] \quad [\text{Eq. 6}]$$

The initial crack size can be estimated from intrinsic defects observed in metallography, from the largest undected flaw size during non-destructive evaluation, or from proof testing. The final crack size may be determined from some allowable shape change or from the critical stress intensity for unstable crack growth (K_{IC}).

CONCLUSIONS

The fatigue properties of U6Nb have been measured to support design evaluations based upon either the fatigue strengths determined from stress vs. life (S-N) testing or fracture mechanics methods using fatigue crack propagation rate data. The three stages of fatigue: crack initiation, crack propagation, and final overload failure have each been characterized in this investigation. While most of the testing was performed on material aged at 200°C for 2 h, tensile properties and fatigue crack propagation rates were also measured in as-quenched material. A number of specific conclusions are listed below.

Stress vs. fatigue life characteristics (aged U6Nb in lab air):

As is typical of nonferrous alloys, U6Nb does not exhibit a fatigue endurance limit. The smooth bar fatigue strength decreases with increasing lifetime requirements and failures continue to occur at lifetimes beyond (10^8) cycles.

The smooth bar fatigue strength is 248 MPa (36 ksi) and 300 MPa (43.5 ksi) for fatigue lifetimes of 10^8 cycles and 10^7 cycles, respectively, using a lower-bound fit to the data.

The fatigue strength is reduced by a factor of 3.3 for a severe notch with a theoretical (elastic) stress concentration factor of 11. The notched bar fatigue strength is 75 MPa (11 ksi) for a fatigue lifetime of 10^8 cycles.

For mean tensile loads of 275 MPa (40 ksi), the fatigue strength is reduced to 103 MPa (15 ksi) for a fatigue lifetime of 10^7 cycles. For mean loads of 483 MPa (70 ksi), the fatigue strength is approximately 40 MPa (6 ksi) for a fatigue lifetime of 10^7 cycles.

Fatigue crack initiation occurs preferentially at inclusions identified as mixed niobium-carbide and uranium-oxide. The largest inclusions observed in the

fatigue test plates were niobium carbides with a maximum diameter of $1.6(10^{-2})$ mm. Fractured inclusions are present in enhanced density along fatigue crack paths and dominate as sources of ductile void initiation sites during overload fracture.

Crack initiation lifetimes of 85 to 99.9 percent of the total lifetime were measured for smooth bar specimens subjected to high mean tensile loads. A novel application of a direct-current potential-drop technique was used for these measurements with an estimated crack length detectability of $5(10^{-3})$ mm (2×10^{-4} inch).

Fatigue crack propagation rates:

Fatigue crack growth rates (FCGR's) of aged U6Nb in lab air decrease rapidly with decreasing cyclic stress intensity (ΔK) to approximately $10 \text{ MPa}\sqrt{\text{m}}$, below which a plateau in growth rates occurs in which the FCGR's are relatively insensitive to ΔK , and at lower ΔK decrease asymptotically to a threshold ΔK .

The fatigue threshold is $3.2 \text{ MPa}\sqrt{\text{m}}$ for FCGR's below 10^{-7} mm/cycle at an R-ratio of 0.1. This threshold value was determined to be influenced by crack closure which occurs above $K_{\min}$. The effective threshold ΔK , calculated as the difference between the $K_{\max}$ and K at crack opening is reduced to $2.1 \text{ MPa}\sqrt{\text{m}}$.

With an increase in mean tensile loading imposed by constant $K_{\max}$ fatigue cycles, the fatigue threshold decreased further. At a $K_{\max}$ of $14.6 \text{ MPa}\sqrt{\text{m}}$, the fatigue threshold decreased to $0.9 \text{ MPa}\sqrt{\text{m}}$ and at a $K_{\max}$ of $30.5 \text{ MPa}\sqrt{\text{m}}$, power-law growth continued to below $0.6 \text{ MPa}\sqrt{\text{m}}$ with no asymptotic decreases in FCGR's.

Fatigue crack propagation below ΔK 's of approximately $10 \text{ MPa}\sqrt{\text{m}}$ resulted in severe crack branching with multiple crack fronts. This crack branching persisted for reductions in specimen thickness from 12.7 to 2.5 mm, for high R-ratios, and for dry nitrogen (inert) environments. Macroscopic crack path tortuosity, as

observed on the fracture surfaces, increased in an aggressive environment of 100RH air and with increased mean tensile loads imposed by constant K_{max} fatigue cycles.

In an inert environment of dry nitrogen, the FCGR's decreased relative to those in lab air but maintained similar characteristics. An exception is the presence of asymptotic threshold behavior at both K_{max} values in the nitrogen environment. The fatigue threshold was measured as $1.5 \text{ MPa}\sqrt{\text{m}}$ and $0.8 \text{ MPa}\sqrt{\text{m}}$ for K_{max} values of $14.6 \text{ MPa}\sqrt{\text{m}}$ and $30.5 \text{ MPa}\sqrt{\text{m}}$, respectively.

In an environment of 100RH air, power law growth continues to growth rates below (10^{-8}) mm/cycle for constant K_{max} fatigue cycles of both $14.6 \text{ MPa}\sqrt{\text{m}}$ and $30.5 \text{ MPa}\sqrt{\text{m}}$. FCGR's are reduced in this environment with respect to lab air or dry nitrogen while the fracture path becomes increasingly tortuous. A clear microstructural correlation with the crack path deviations could not be confirmed although Nb banding is expected to exist on the appropriate microstructural scale. Environmentally-assisted cracking is expected to be operative at a K_{max} of $30.5 \text{ MPa}\sqrt{\text{m}}$ in high humidity environments based upon previous investigations.

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Table I: Chemical Composition of U6Nb Test Plates

Plate ID	Ingot Location	U (wt%)	Nb (wt%)	C (wt ppm)
A	top center	93.32	6.39	103
	top edge	-	-	97
	bottom center	94.3	5.91	-
B	top center	93.82	5.85	72
	top edge	-	-	58
	bottom center	93.97	5.9	-

Table II: U6Nb Tensile Properties

Condition	Composition	Yield Strength MPa (ksi)	Tensile Strength MPa (ksi)	Elongation %	Reduction in Area %	Elastic Modulus# GPa (psi*10^6)
As-Quenched	A	164 (23.8)	847 (123)	33	43	64 (9.3)
	B	170 (24.6)	843 (122)	33	45	65 (9.5)
Aged 200°C/1h	A	436 (63.3)	880 (128)	31	36	77 (11.2)
Aged 200°C/2h	A	467 (67.7)	883 (128)	33	37	77 (11.2)
Aged 200°C/2h(v)	A	547 (79.4)	900 (130.5)	30.5	42	78 (11.3)
	B	534 (77.4)	883 (128)	31.5	45	78 (11.3)

(v) vacuum heat treatment.
from extensometer data.

Table III: U6Nb Fracture Toughness

Specimen ID	Material Condition	Temperature °C	Test Method	Thickness in.	(J)Q in-lb/in ²	(K)Q MPa*m ^{0.5}
1A, 2A	As-Quenched	23	compact tension	0.5	1215*	114*
1AH	Aged 200C/2h (vacuum)	23	" "	0.5	1722*	137*
Y-12	Aged 200°C/2h	23	multi-specimen bend	0.394	675	97
Y-12	Aged 200°C/2h	-54	" "	0.394	655	96
2AH	Aged 200°C/2h	23	compact tension	0.5	642*	95

* valid J(IC), K(IC)

Table IV: Fatigue Strength Reduction Factors

Material Condition	Specimen Type	Mean Stress MPa (ksi)	Fatigue Strength (10⁸ cycles) MPa (ksi)	Tensile Strength MPa (ksi)	Fatigue/Tensile Strength Ratio %
quenched	smooth RBF	0	240 (35)*	845 (123)	28
aged 200°C/2h	smooth RBF	0	248 (36)	882 (128)	28
aged 200°C/2h	notched RBF (K _t =11)	0	75 (11)	882 (128)	9
aged 200°C/2h	smooth axial	275 (40)	103 (15)	882 (128)	12
aged 200°C/2h	smooth axial	483 (70)	41 (6)	882 (128)	5

* from P. M. Finnerty, S.J. Hull, and P. A. Mackett; JOWOG 22C, June 1987.

Table V: Crack Initiation Lifetimes

Mean Load, MPa (ksi)	Cyclic Amplitude, MPa (ksi)	Total Lifetime, # cycles	Initiation Lifetime, # cycles (% of total)	Propagation Lifetime, # cycles
276 (40)	121 (17.5)	4.1 E5	3.6 E5 (89)	4.5 E4
276 (40)	121 (17.5)	5.3 E5	4.5 E5 (86)	7.4 E4
276 (40)	121 (17.5)	2.9 E7	2.9 E7 (99.87)	3.8 E4
276 (40)	129 (18.8)	2.0 E5	1.7 E5 (86.5)	2.7 E4
483 (70)	52 (7.5)	1.9 E6	1.9 E6 (99.88)	2.3 E3

Table VI: Fatigue Thresholds for U6Nb Aged 200°C/2h

Load Ratio (R) or constant Kmax value (MPa*m ^{0.5})	Test Environment	Threshold Stress Intensity Range (MPa*m ^{0.5})
R=0.1	air (50RH)	3.2
14.6	air (50RH)	0.9
30.5	air (50RH)	<0.6
14.6	dry nitrogen	1.5
30.5	dry nitrogen	0.8

(a) Smooth-Bar Rotating Beam Fatigue Specimen

DIA. D ^{+.001} _{-.000}	A	B	C
.250	5/8	2	3/4
.375	✓	3	1
.500	1 1/4	4	1 1/2

(b) Notched-Bar Rotating Beam Fatigue Specimen

(c) Axial Mean Load Fatigue Specimen

Figure 1. Test specimen geometry for stress/life fatigue (S-N) tests: (a) smooth bar specimen, (b) notched-bar specimen, (c) mean axial-load specimen.

Figure 2. (a, b) Stress vs strain curve for U6Nb: (a) as-received wrought condition, (b) vacuum heat treated 200 °C/2 h.

(c)

(d)

Figure 2. (c, d) Stress vs strain curve for U6Nb: (c) air furnace treated 200 °C/1h, (d) air furnace treated 200 °C/2h.

U6Nb Aged 200C/2h:
Smooth-Bar Rotating-Beam Fatigue

Figure 3. Fatigue strength of smooth-bar specimens.

U6Nb Aged 200°C/2h:
Notched-Bar Rotating-Beam Fatigue
($K_t=11$)

Figure 4. Fatigue strength of notched-bar specimens.

Figure 5. Axial bar fatigue strength for 40 Ksi (275 MPa) mean load.

Figure 6. Axial bar fatigue strength for 70 Ksi (483 MPa) mean load.

Figure 7. Fatigue strength vs. mean tensile load (10^7 cycles to failure).

Figure 8. (a) Direct-current potential-drop (DCPD) crack initiation data for a mean stress of 40 Ksi (276 MPa).

Figure 8. (b) Direct-current potential-drop (DCPD) crack initiation data for a mean stress of 70 Ksi (483 MPa).

Fatigue crack initiation in aged U-6Nb;
Rotating-Beam-Fatigue Test
2.5 (10^6) cycles to failure at 400 MPa (57 ksi) peak stress

Figure 9. (a) Typical fracture surfaces of smooth-bar bend specimens loaded to an amplitude of 57 Ksi (400 MPa).

20 μm

5 μm

Figure 9. (b) Typical fracture surfaces of smooth-bar bend specimens loaded to an amplitude of 50.8 Ksi (350 MPa).

Fatigue crack initiation in aged U-6Nb;
Rotating-Beam-Fatigue Test
2.5 (10^6) cycles to failure at 400 MPa (57 ksi) peak stress

Niobium-carbide inclusion (marked)
adjacent to uranium-oxide inclusion

Figure 10. Fatigue initiation site identified as a cluster of niobium carbide and uranium oxide inclusions.

Fatigue striations in aged U-6Nb; Rotating-Beam-Fatigue Test
2.5 (10^6) cycles at 400 MPa peak stress

Figure 11. Fatigue striations on a smooth bar bend specimen loaded to an amplitude of 57 Ksi (400 MPa).

Figure 12. Fatigue crack growth rates of U6Nb aged 200 °C/2 hr.; R = 0.1 in laboratory air environment.

**As-Quenched vs Aged (200°C/2h) U6Nb:
Lab Air, R=0.1**

Figure 13. Fatigue crack growth rates of as-quenched vs. aged U6Nb; R = 0.1 in laboratory air environment.

Fatigue crack-growth-rate specimen profile;
Aged U-6Nb

Figure 14. Fatigue fracture cross-section illustrating the path of fatigue crack propagation for aged U6Nb; $R = 0.1$ in laboratory air environment.

U6Nb Aged 200°C/2h: Lab Air, R=0.1

Figure 15. Fatigue crack growth rates in 2.5 mm and 12.7 mm thick aged specimens; R = 0.1 in laboratory air environment.

U6Nb Aged 200°C/2h: Lab Air

Figure 16. Fatigue crack growth rates at constant $K_{max} = 14.6 MPa\sqrt{m}$ and at constant $R=0.1$; 2.5 mm thick, aged specimens, tested in laboratory air environment.

Figure 17. Fatigue crack growth rates at constant $K_{\max} = 14.6 \text{ MPa} \sqrt{\text{m}}$ and $30.5 \text{ MPa} \sqrt{\text{m}}$; 2.5 mm thick specimens, aged, tested in laboratory air environment.

Figure 18. Fatigue crack growth rates at constant $K_{max} = 30.5 \text{ MPa}\sqrt{\text{m}}$ from 2.5 mm and 12.7 mm thick specimens; aged, tested in laboratory air environment.

Figure 19. Fatigue crack growth rates at constant $R = 0.1$ and constant $K_{max} = 30.5 \text{ MPa}\sqrt{\text{m}}$; aged specimens, 12.7 mm thick, tested in laboratory air environment.

Figure 20. Fatigue crack growth rates in dry nitrogen and lab air: constant $K_{max} = 14.6 \text{ MPa}\sqrt{\text{m}}$, aged specimens.

Figure 21. Fatigue crack growth rates in dry nitrogen and lab air: constant K_{max} = 30.5 MPa√m, aged specimens, 12.7 mm thick.

Figure 22. Fatigue crack growth rates in moisture-saturated air at constant $K_{\max}$ = 14.6 and 30.5 $\text{MPa}\sqrt{\text{m}}$; aged U6Nb, 12.7 mm thick.

Figure 23. Fatigue crack growth rates in lab air and dry nitrogen; constant $K_{max} = 14.6 \text{ MPa}\sqrt{\text{m}}$, as-quenched U6Nb, 12.7 mm thick.

Figure 24. Fatigue crack growth rates of as-quenched plate vs. plate aged U6Nb; dry nitrogen atmosphere, constant $K_{max} = 14.6 \text{ MPa}\sqrt{\text{m}}$, 12.7 mm thick specimens.

20 μm
└───┘

Figure 25. Optical micrograph of U6Nb aged 200 °C/2h.

50 μm

20 μm

Figure 26. Fracture profile of U6Nb aged 200 °C/2h showing martensitic substructures and secondary fatigue cracks.

Maximum Individual Carbide Sizes in each of 57 heats of U6Nb
(from 1.9 mm² random areas of ingot center and edge samples) [8].

Figure 27. Maximum carbide sizes observed from center and edge sections of each of 57 heats of U6Nb [8].

**Inclusion Characterization of U6Nb
(from 57 heats produced at Y-12 [8])**

Figure 28. Mean carbide size distributions at edge and center sections of 57 heats of U6Nb [8].

Figure 29. Fracture surface of aged U6Nb produced by tensile overload.

Figure 30. Fracture profile showing secondary fatigue cracks in test atmospheres of 100 RH air at constant $K_{\max} = 30.5 \text{ MPa}\sqrt{\text{m}}$.

Figure 31. Fracture profile showing secondary fatigue cracks in test atmospheres of dry nitrogen at constant $K_{\max} = 14.6 \text{ MPa}\sqrt{\text{m}}$.

Figure 32. Fracture surfaces from fatigue crack propagation in lab air at constant $K_{max} = 30.5 \text{ MPa}\sqrt{\text{m}}$ (lower region in (a) and photo (c)) and from tensile overload (upper region in (a) and photo (b)).

1 mm

PRECRACK
LAB AIR

DRY NITROGEN
 $K_{max}=14.6 \text{ MPa}\cdot\text{m}^{0.5}$

100RH AIR
 $K_{max}=14.6 \text{ MPa}\cdot\text{m}^{0.5}$

1 mm

100RH AIR
 $K_{max}=14.6 \text{ MPa}\cdot\text{m}^{0.5}$

100RH AIR
 $K_{max}=30.5 \text{ MPa}\cdot\text{m}^{0.5}$

OVERLOAD
FAILURE

Figure 33. SEM fatigue fracture surfaces from a single specimen in several test atmospheres. Crack propagation direction is from top left to bottom right.

20 μm

5 μm

Figure 34. SEM fatigue fracture surface from tests in dry N₂ at constant $K_{max} = 14.6 \text{ MPa}\sqrt{\text{m}}$. Crack propagation direction is left-to-right.

100 μm

20 μm

Figure 35. Fatigue fracture surface from tests in dry N₂ at constant $K_{max} = 30.5 \text{ MPa}\sqrt{\text{m}}$. Crack propagation direction is left-to-right.

Figure 36. SEM fatigue fracture surface imaged with backscattered electrons from tests in dry N₂ at constant $K_{max} = 14.6 \text{ MPa}\sqrt{\text{m}}$. Crack propagation direction is left-to-right.

Figure 37. Fatigue fracture surface from tests in 100 RH air at constant $K_{max} = 14.6 \text{ MPa}\sqrt{\text{m}}$. Crack propagation direction is left-to-right.

500 μm

10 μm

Figure 38. Fatigue fracture surface from tests in 100 RH air at constant $K_{max} = 30.5 \text{ MPa}\sqrt{\text{m}}$. Crack propagation direction is left-to-right.

Figure 39. Electron microprobe line profile of Nb and U concentrations parallel to the original rolling plane and perpendicular to a secondary fatigue crack.

Figure 40. Near threshold fatigue crack propagation behavior for several engineering materials. (Data from P. K. Liaw, T. R. Leax, and W. A. Logsden, *Acta Metallurgica*, 31, 1581 (1983).)

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