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MOLYBDENUM ALLOY UNDER DYNAMIC TENSION LOADS

K. G. Hoge

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Some Mechanical Properties of Uranium-10 wt % Molybdenum Alloy
Under Dynamic Tension Loads¹

K. G. Hoge, Associate Member ASME

Mechanical Engineer
Lawrence Radiation Laboratory, University of California
Livermore, California

February 1, 1965

Abstract

Dynamic tensile tests were conducted on uranium-10 wt % molybdenum alloy in as-cast and wrought conditions. Carbon content was varied from 71 to 800 ppm. Heat treatment was also varied. Tests were conducted on a universal testing machine at strain rates of 4.8×10^{-5} and 0.057 in./in./sec and on a modified Dynapak at strain rates up to 100 in./in./sec. Tests were run from 75 to 600° F. Relatively low strain rates were used to determine the susceptibility to stress corrosion while the higher strain rates demonstrated the sensitivity of various mechanical properties to strain rate effects.

¹ Work performed under the auspices of the U. S. Atomic Energy Commission.

Nomenclature

UMC = cast specimens with high carbon content

UMC(LC) = cast specimens with low carbon content

UME = extruded specimens

UMR = hot rod rolled specimens

Introduction

The purpose of this investigation was to determine the mechanical properties of uranium-10 wt % molybdenum alloy under dynamic tensile loads. This particular alloy was selected because of its high strength, stability, and homogeneity. U-10 wt % Mo has a body-centered cubic structure which unlike alpha uranium is not subject to ratcheting. Also, this alloy is not subject to gross segregation. The microsegregation which does occur can be relieved by homogenization.

Various experimenters have shown that many metals will show an increase in tensile strength when loaded dynamically [1-4].² The effect of high strain rates on U-10 wt % Mo alloy is to increase its tensile strength through two independent mechanisms. First, as the strain rate is increased, the effect of stress corrosion is mitigated due to the decrease in time the load is applied. The stress corrosion effect predominates at relatively slow strain rates, i. e., less than one per second. At faster loading rates there is insufficient time for any appreciable corrosive action. Second, high strain rates tend to delay the yield phenomenon and increase the tensile strength. It is believed that this effect is associated with the movement of dislocations. It will predominate at relatively high strain rates.

² Numbers in brackets designate References at end of paper.

The alloy was tested in both as-cast and wrought conditions. Two batches of as-cast specimens were investigated, one of high carbon content and one of low carbon content. The wrought alloy tested was of two types: extruded and hot-rolled. Hot-rolled specimens were extremely notch sensitive. Several different heat treatments were used in an attempt to alleviate this adverse effect.

Tests were conducted at temperatures from 75 to 600° F and at strain rates from "static" tests were run on a Baldwin Lima Hamilton Universal Testing Machine of 120,000 lb capacity. Dynamic tests were conducted on a Dynapak, model 600, modified to produce tension loading.

Experimental Procedure

Instrumentation

Specimen stress was obtained from a load cell instrumentated with four C6-141-R2B strain gages forming a full bridge circuit. This circuit gives an amplification of about 2.6 over a single gage. Gages were attached with Eastman 910 cement and covered with a thin layer of Gagekote number 2 for insulation and protection.

Strain rates and specimen elongations were determined from displacement measurements between couplings threaded to the specimen ends. Coupling displacement was sensed by a Sanborn velocity transducer, model 3LVA5-N. As the transducer output is actually a measure of velocity, it is integrated by a Tektronix type O preamplifier to convert it to a displacement measurement.

Specimen temperatures were measured by iron-constantan thermocouples spot welded to the specimens. Thermocouple output was read on an L and N millivolt potentiometer.

Three Tektronix oscilloscopes were used to record data. Specimen load versus time was displayed on a model 555 scope. Strain versus time was recorded on a model 35A, and stress versus strain was recorded on a model 536. Permanent records of these oscilloscope readings were made by photographing the display with a Polaroid Land camera. No permanent record of specimen temperature was deemed necessary. The oscilloscopes were triggered by a bistable multi-vibrator trigger circuit which closed as the Dynapak ram contacted the ram plate. This circuit and the load and strain circuits are shown schematically in Fig. 1.

Heating Unit

A Master Heat Gun was used to heat specimens. This gun blows hot air through a duct into a metal shell surrounding the specimen. A model HG301 gun was used for 200 and 300° F runs and a model HG601 for the 600° F tests. Insulation was placed around the outer shell surface to prevent undue heating of the load cell and velocity transducer.

Impact Machine

A model 600 Dynapak, modified to produce tension loading, was used to impact test specimens. This machine is pictured in Fig. 2.

Dynapak is basically a gas gun with nitrogen as the working fluid. The nitrogen, stored at pressures up to 2000 psi, is allowed to rapidly expand against a 6 in. diam piston. This piston drives the ram at high velocities. The ram impacts the ram plate, moving it and the reverse tension plate at a high velocity. The specimen, placed between the reverse tension plate and the bolster plate, is thus given a tensile load at a high strain rate. The strain rate is controlled by varying combinations of fire pressure (pressure driving the ram), ram weight, and distance between the ram and ram plate before

nitrogen release. To eliminate eccentric loading due to misalignment, the specimen is connected to both the bolster and reverse tension plates through spherical seats.

Test Specimens

Test specimens were made of U-10 wt % Mo alloy by Union Carbide Nuclear Company to the configuration of method 211.1 type R-3 of Federal Test Method Standard No. 151a. Figure 3 shows the specimen dimensions. Specimens can be broken into four batches or groups as follows:

- (a) Extruded.
- (b) As-cast, high carbon content.
- (c) Hot-rolled.
- (d) As-cast, low carbon content.

The fabrication record of each of these four groups will be discussed separately.

Extruded specimens. The U-10 wt % Mo alloy was first cast in the shape of a cylindrical ingot 3.250 in. in diameter and 20.250 in. long. The casting was then homogenized at a temperature of 1650° F for 24 hr under a vacuum. The ingot was then machined into two billets 3.125 in. in diameter and 10 in. long. These billets were heated to 1650-1740° F in an inert gas (argon) furnace for 2 hr and extruded to a diameter of 1.750 in. The extrusion press ram rate was approximately 0.333 in./sec. Three hundred mesh borate glass and flake graphite were used as a lubricant. The extruded rods were then heat treated at 1400° F for 1 hr. Specimens were made with their axes in the extruded direction. Chemical analysis shows:

Molybdenum	- 9.4%	Nickel	- 15 ppm
Carbon	- 228 ppm	Manganese	- 10 ppm
Iron	- 115 ppm	Copper	- 10 ppm
Magnesium	- 2 ppm	Cesium	- 10 ppm

As-cast, high carbon content specimens. These specimens were machined from billets which were cast and machined in the same way as the extruded specimens. However, no homogenization process was used. Chemical analysis of these specimens is:

Molybdenum - 10%	Nickel - 300 ppm
Carbon - 800 ppm	Manganese - 20 ppm
Iron - 225 ppm	Copper - 15 ppm
Magnesium - 4 ppm	Cesium - 100 ppm

Hot-rolled specimens. The U-10 wt % Mo alloy was cast and machined into a 6-7/8 in. diam ingot. The ingot was then homogenized at 1650° in a 20 micron vacuum for 24 hr. The ingot was heated in a globular electrical resistance furnace to approximately 1800° F and then rolled into a billet having the standard billet cross section specified by Hanford Atomic Products Operation (2.034 by 2.562 in.). In the rolling operation 19 passes were made. Only a slight decrease in temperature was noted during rolling. The billet was then annealed at 1380° F for 18 hr in a vacuum to relieve residual stresses. Chemical analysis is:

Molybdenum - 10.4%	Nickel - 6 ppm
Carbon - 98 ppm	Manganese - 4 ppm
Iron - 30 ppm	Copper - 6 ppm
Magnesium - 2 ppm	Cesium - 100 ppm
Phosphorous - 100 ppm	Hafnium - 100 ppm

As-cast, low carbon specimens. The alloy was cast as for extruded specimens. However, several steps were taken to minimize carbon pickup. First, a protective coating of ZrO_2 was applied to both the crucible and mold. Second, as carbon can be picked up only when the alloy is in the molten state, hold times were scheduled at times before the alloy melts. During the casting operation

and for 1/2 hr thereafter the furnace remained under a vacuum. The ingot was homogenized for 24 hr at 1650° F in a 20 micron vacuum. Chemical analysis of these specimens is:

Molybdenum -	9.9%	Nickel -	6 ppm
Carbon -	71 ppm	Manganese -	3 ppm
Iron -	25 ppm	Copper -	20 ppm
Magnesium -	2 ppm	Cesium -	100 ppm
Phosphorous -	100 ppm	Hafnium -	100 ppm

Discussion and Results

General

The modified Dynapak machine and the measurement instrumentation used for this series of tests formed a system that produced valid dynamic data. However, it must be realized that because of the many complications associated with testing at high rates of strain all such test systems require some compromise between the optimum requirements [5]. The Dynapak produces dynamic forces which load the test specimen with a uniaxial tensile stress. The load cell indicates this stress with sufficient accuracy, provided that the loading time is slow enough so that the stress wave propagation effects can be neglected. The required strain rate can be obtained by properly setting the Dynapak fire pressure, ram weight, and distance between the ram and ram plate. Values for ram force, velocity, and energy in terms of various combinations of the above settings have been determined on an analog computer. The ram velocity at impact will determine the initial strain rate. To maintain this strain rate at a constant value the resistive force of the specimen should equal the other external forces acting on the ram. As both the external forces and specimen resistance will vary during a test, it is usually impossible to maintain a

constant rate of strain. However, conscientious settings will usually produce a strain rate that varies only a few percent up to the yield point. During plastic flow, larger velocity changes may occur. A velocity transducer measures the average strain rate referred solely to the gage length. This strain rate is furthermore averaged from impact to peak load and from peak load to rupture. The load cell and velocity transducer and stress wave propagation effects are discussed in detail in the appendices of reference [6].

Tests conducted on each batch of specimens can be divided into two categories. The first consists of "static" tests conducted on a universal testing machine at strain rates of 0.057 and 4.8×10^{-5} per second. These data were principally used to evaluate the alloy's sensitivity to stress corrosion by noting the decrease in tensile strength between the fast to slow rate. Data show this alloy is susceptible to stress corrosion, or stress cracking as it is sometimes called [7-9]. However, the circumferential cracking noted in reference [10] was not prevalent in these tests. It is believed that the stress corrosion is the result of oxygen attack, particularly in the grain boundaries.

The second category of tests were those conducted on the Dynapak at strain rates greater than one per second. Data from both "static" and dynamic tests are presented in tabular form in Tables 1 to 4 and are illustrated by graphs of peak stress versus strain rate in Figs. 4 to 7. Peak stress is defined as the first stress level where sufficient plastic strain has occurred so that the tangent to the stress-strain curve is parallel to the strain axis. For this alloy the peak stress is the material's ultimate engineering strength. However, it is also approximately the yield strength of the alloy for strain rates faster than ten per second. Even on "static" tests the ultimate strength never exceeded the 0.2% yield strength by more than 4%. As strain rate is increased, there is a tendency toward delaying plastic flow, thus causing the

yield stress to approach the ultimate. This delaying action is most notable in materials that have a definite yield point [11]. This alloy does have a definite yield point.

Specimen history, including fabrication, heat treatment, and impurities has a very important effect on peak stress versus strain rate curves as well as other mechanical properties. The peak stress versus strain rate plot for as-cast, high carbon specimens, shown in Fig. 5, illustrates a curve that is concave upwards for rates up to forty per second. No doubt at some higher rate this curve will pass through an inflection point and become concave downward. Such changes in curvature are not uncommon in stress versus strain rate graphs [3]. The graphs for extruded (Fig. 4) and hot rolled specimens (Fig. 6) show a very slight negative change in slope at strain rates faster than ten per second. Hot rolled specimens indicate a slightly greater increase in peak stress for corresponding changes in strain rate than do extruded specimens. Whether or not the stress versus strain rate curves of high carbon, as-cast specimens would cross or become coincident to the corresponding curves for wrought specimens at some high strain rate is a matter of conjecture. Although, to the author's knowledge, no relationship between fatigue and high strain rate loading has been established, it is interesting to note that modified S-N curves for cast and wrought specimens fabricated similar to those tested dynamically, do cross. These curves are illustrated in Fig. 8. There were not sufficient as-cast, low carbon specimens available to completely define their peak stress versus strain rate curve given in Fig. 7. However, the mechanical properties of this group appear to be almost as good as the wrought specimens.

Theoretical

It would be advantageous if a mathematical relationship for stress in terms of strain, strain rate, and temperature could be established.

$$\sigma = f(\epsilon, \dot{\epsilon}, T) \quad (1)$$

However, for this alloy such a relationship is difficult to obtain. Due to the large and unpredictable effect of stress corrosion at the lower strain rates, the material's strain rate sensitivity at rates below one per second cannot be determined. It should be noted that strain rate sensitivity refers to the effect of strain rate on the dislocation mechanisms associated with plastic flow. As dynamic tests were conducted at strain rates below 100 per second there is not a sufficient range in data to formulate a valid logarithmic-type solution. But, a relatively simple empirical formula is derived with the following assumptions and limitations:

1. Peak stress versus strain rate curves, Figs. 11 to 14, are approximated by straight lines.
2. For a particular batch of specimens these straight line approximations (constant temperature tests) are parallel.
3. Temperatures are limited to values from 75 to 600° F.
4. Strain rate effects are limited to regions covered by experimental data.

Assumptions one and two appear to be valid in view of the stress-strain rate curves. However, for higher temperatures there may be an augmenting effect on the strain rate sensitivity [5]. Therefore, limitation three is stated. As the tensile strengths of many metals show a dependency on the logarithm of strain rate, it would be unwise to extrapolate stress-strain rate curves beyond data points. Hence the necessity for limitation four.

For any batch of specimens, the straight line approximations of the stress-strain rate curves are assumed parallel so that stress can be obtained as the sum of functions of two independent variables $\dot{\epsilon}$ and T .

$$\sigma_{\text{peak}} = g(\dot{\epsilon}) + h(T) \quad (2)$$

As the function $g(\dot{\epsilon})$ is linear,

$$\sigma_{\text{peak}} = P\dot{\epsilon} + Q + h(T) \quad (3)$$

where

$$h(T) = 0 \quad \text{at} \quad T = 75^\circ \text{F}$$

and

$$\sigma_{\text{peak}} = \sigma_0 \quad \text{at} \quad T = 75^\circ \text{F}, \dot{\epsilon} \rightarrow 0$$

Therefore, equation (3) becomes

$$\sigma_{\text{peak}} = \sigma_0 + P\dot{\epsilon} + h(T) \quad (4)$$

The above end conditions are chosen for convenience. The stress at 75°F and zero strain rate is now equal to the constant Q and can be termed the theoretical room temperature static tensile strength. The constant P is the slope of the line approximation.

$$P = \Delta\sigma/\Delta\dot{\epsilon} \quad (5)$$

This slope is essentially the materials sensitivity to strain rate effects. For U-10 wt % Mo alloy the peak stress occurs at approximately the same value of strain, i. e., 2%. Equation (4) can be applied to stresses at other values of strain. However, the values of the constants would be different.

Under the end conditions chosen (at $T = 75^\circ \text{F}$, $h(T) = 0$), $h(T)$ can be determined by horizontally translating the peak stress versus temperature curves, shown in Fig. 9, so stress equals zero at 75°F . The function, $h(T)$, is different for each group of specimens.

Malvern [12] has used equation (6) to show the dependency of material strength on the logarithm of strain rate.

$$\sigma = \sigma_0 + M \ln \dot{\epsilon} \quad (6)$$

The constant M is a function of temperature and to some extent a function of strain rate [5]. Equation (6) is very similar to equation (4) except for the logarithm dependence. As stated previously, this approach was not used because only one decade of strain rate data was available for a dependency analysis. However, as more data become available Malvern's logarithmic formula will be checked.

Extruded Specimen Tests

This batch of specimens displayed the most consistent mechanical properties of those tested. All extruded specimens failed in the gage length. On peak stress versus strain rate curves all data points plotted within 5% of the mean curve except for test number 7. In this test the fractured surface was unusually rough and perhaps this specimen should not be considered representative of the group. Data are tabulated in Table 1 and illustrated in Fig. 4. A typical oscilloscope record of a ductile failure is pictured in Fig. 10.

As-Cast, High Carbon Specimen Test

The carbon content of these specimens was about 800 ppm. This was the only batch of specimens which was not homogenized. This results in an uneven distribution of carbon and an irregular grain size. In Fig. 11 the microstructure of specimens UMC-2 and 8 is shown. Note the difference in grain size of the two specimens. Both intergranular and transgranular cracks can be observed. Considerable scatter of data resulted as shown in

the stress versus strain rate curves (Fig. 5). One specimen fractured in the threaded area in a preliminary calibration test and specimen UMC-16 failed in the fillet radius during a 75° F test. All 600° F tests resulted in ductile failures. At 300° F and below both ductile and brittle failures were observed. Fig. 12 pictures the oscilloscope record of a typical brittle fracture. Test data are recorded in Table 2 and peak stress versus strain rate curves are plotted in Fig. 5.

Hot-Rolled Specimen Tests

In 75 and 300° F tests these specimens were extremely notch sensitive. At 75° F only one specimen failed in the gage length. Fig. 13 shows the oscilloscope record of a typical brittle failure in the threaded section. The microstructure of a thread fracture is pictured in Fig. 14. Cracks are mostly intergranular. Note the lack of twinning as compared to Fig. 11. At 300° F UMR-6 fractured in the threads in a "static" test and UMR-16 and 17 failed in the threads in dynamic tests. The four other 300° F tests resulted in normal ductile failures in the gage length. At 600° F specimen UMR-4 fractured in the fillet radius. All other specimens failed in a ductile manner with elongations of about 11%. A typical ductile failure is pictured in Fig. 15. Stress versus strain rate curves for this batch are shown in Fig. 6. The reason for these brittle failures in the threads has not been determined. Possible causes are stress corrosion, residual stresses, or changes in microstructure due to the long heat treatment period. However, because these specimens showed excellent mechanical properties, except for their notch sensitivity, and because the rolling process is simpler than the extrusion process, further investigation was conducted. Several specimens were given special heat treatments. UMR-18, 19, and 20 were heat treated at 1380° F for 15 hr in a vacuum

of 1 micron. UMR-18 was loaded twice as the first load produced a coupling failure. The second dynamic load resulted in a thread failure. UMR-19, tested dynamically, also failed in the threads. UMR-20 was tested "statically" and failed in a ductile manner in the gage length. Final elongation was 17%. Specimens UMR-21 to 25 were heat treated at 1400° F for 1 hr in a 10 micron vacuum. UMR-22 was tested "statically" and showed a normal ductile fracture with 23% elongation. UMR-21 failed in a normal ductile manner in a dynamic test with 6.2% elongation. This is approximately the same ductility exhibited by extruded specimens. The other three specimens failed in the threaded area in dynamic tests. UMR-23 was machined from the center of the billet and showed an unusually low rupture strength. All the other specially heat treated specimens that fractured in the brittle mode failed at a higher stress than did the original specimens. Thus the special heat treatments helped although it did not completely alleviate the notch sensitiveness of this material. Long-time heat treatments appear particularly detrimental. All data for this group of specimens are tabulated in Table 3.

As-Cast, Low Carbon Specimen Tests

These specimens, containing only 71 ppm carbon, exhibited tensile strengths similar to extruded and hot-rolled specimens. However, they showed less ductility than did the wrought specimens. These specimens, when compared to the high carbon as-cast batch, indicate that carbon causes a decrease in both tensile strength and ductility. There also seems to be less scatter in test results for these specimens than for the high carbon group. This is probably due to the homogenization treatment given to the low carbon batch. Fig. 7 shows the peak stress versus strain rate curves. These curves are incomplete due to the limited number of specimens available for testing. Table 4 records the test data.

Conclusions and Observations

Considering the data from the four groups of specimens tested, the following conclusions are drawn:

1. Dynamic loading increases the tensile strength and lowers the ductility of U-10 wt % Mo alloy in both the as-cast and wrought conditions.
2. U-10 wt % Mo alloy is sensitive to stress corrosion.
3. Extruded specimens are best suited for room temperature applications and in cases where notch sensitivity may be a problem.
4. Hot-rolled specimens are superior at relatively high temperatures, i. e., around 600° F. At lower temperatures they are extremely notch sensitive.
5. As-cast specimens do not display as consistent mechanical properties as wrought specimens. As-cast specimens are also less ductile.

The following observations are also made. However, these observations are not conclusive due to the many variables present.

1. Carbon content should be kept low, below 200 ppm. High carbon content tends to reduce both ductility and tensile strength.
2. The heat treatment period should be kept low. A period of about 1 hr seems to be best. A vacuum of at least 20 microns should be maintained during the heat treatment period.

Acknowledgment

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Table 1 Test data on uranium-10 wt % molybdenum alloy, extruded condition

Test no.	Temp., ° F	$\dot{\epsilon}$, ^a in. / in. / sec	σ_{peak} , ksi	$\dot{\epsilon}$, ^b in. / in. / sec	$\epsilon_{\text{rupture}}$, %	σ_{rupture} , ksi	Failure mode
UME-1	75	34	200 ^c	33	6.0	--	Ductile
UME-3	75	--	215	--	--	135	Ductile
UME-4	75	39 ^c	206	35	7.5	139	Ductile
UME-5	75	35	201	34	6.0	135	Ductile
UME-6	75	19	190	22	6.0	143	Ductile
UME-7	75	11	190	10	5.8	140	Ductile, rough fracture
UME-8	75	16	191	16	6.2	130	Ductile
UME-9	75	37	201	37	6.2	138.5	Ductile
UME-10	75	65 ^c	221	--	--	129	Ductile
UME-11	75	95	230	280	6.3	125	Ductile
UME-12	600	40	127	36	8.0	61.5	Ductile
UME-13	600	22 ^c	121	20	--	69	Ductile
UME-14	600	17	114	22	9.0	68	Ductile
UME-15	75	4.8×10^{-5}	129	4.8×10^{-4}	14.0	101	Ductile
UME-16	75	0.057	168	0.057	--	108	Ductile
UME-17	75	4.8×10^{-5}	136	4.8×10^{-4}	18.0	107	Very ductile
UME-18	600	0.057	80.5	0.057	23.0	65	Very ductile
UME-19	300	0.057	106	0.057	16.0	81	Ductile
UME-20	300	42	156	60	6.3	90	Ductile
UME-21	600	89	164	120	9.3	79	Ductile
UME-22	75	50	209	48	4.5	150	Rough fracture

^a Average value to peak stress.^b Average value from peak to rupture stress.^c Approximate value.

Table 2 Test data on uranium-10 wt % molybdenum alloy, as-cast condition

Test no.	Temp., ° F	$\dot{\epsilon}$, ^a in./in./sec	σ_{peak} , ksi	$\dot{\epsilon}$, ^b in./in./sec	$\epsilon_{\text{rupture}}$, %	σ_{rupture} , ksi	Failure mode
UMC-2	75	34	192	34	2.8	192	Brittle - fine fracture
UMC-3	75	20	145	20	1.8	145	Brittle - coarse fracture
UMC-4	75	4.8×10^{-5}	120.4	4.8×10^{-4}	12.5	107	Ductile
UMC-5	75	4.8×10^{-5}	118.9	4.8×10^{-4}	5	111	Ductile
UMC-6	75	4.8×10^{-5}	116.8	4.8×10^{-4}	5	106	Ductile
UMC-7	200	30	127.3	30	1.0	127.3	Brittle - coarse fracture
UMC-8	200	36	178.8	42	7.5	102	Ductile - fine fracture
UMC-9	75	30	174.2	30	1.5	174.2	Brittle - coarse fracture
UMC-10	200	42	152	59	2.0	131.2	Brittle - ductile - fine fracture
UMC-11	200	49	179.2	49	2.5	163	Brittle - ductile - fine fracture
UMC-12	200	60	145.3	117	2.5	136.3	Brittle - ductile - fine fracture
UMC-13	300	46	160	81	2.0	135.3	Brittle - ductile - fine fracture
UMC-14	300	40	146.5	117	2.5	137.8	Brittle - coarse fracture near threads
UMC-15	75	22	142.3	22	1.0	142.3	Brittle
UMC-16	75	28	163	28	1.0	163	Brittle - coarse fracture near threads
UMC-17	600	9	91	21	--	75	Ductile - fine fracture
UMC-18	600	4	84	18	--	60	Ductile - fine fracture
UMC-19	600	6.5	90.2	16	11.5	72	Ductile - fine fracture
UMC-20	600	7	96	14	11.7	64	Ductile - fine fracture
UMC-21	200	4.8×10^{-5}	96.2	4.8×10^{-4}	12.0	94	Ductile
UMC-22	300	4.8×10^{-5}	90.4	4.8×10^{-4}	12.0	90.2	Ductile
UMC-23	600	4.8×10^{-5}	62.7	4.8×10^{-4}	14.0	62.7	Ductile
UMC-24	600	4.8×10^{-5}	62.8	4.8×10^{-4}	16.0	62.7	Ductile
UMC-25	600	5.7×10^{-2}	74	5.7×10^{-2}	17.0	61	Ductile
UMC-26	75	36	103	37	14.0	67	Ductile

^a Average strain rate to peak stress.^b Average strain rate from peak to rupture stress.

Table 3 Test data on uranium-10 wt % molybdenum alloy, hot-rolled condition

Test no.	Temp., ° F	$\dot{\epsilon}$, ^a in./in./sec	σ_{peak} , ksi	$\dot{\epsilon}$, ^b in./in./sec	$\epsilon_{\text{rupture}}$, %	σ_{rupture} , ksi	Failure mode
UMR-1	75	4.8×10^{-5}	141.8	4.8×10^{-4}	23	96	Ductile
UMR-2	75	0.057	165 ^c	0.057	--	165	Brittle, fracture in threads
UMR-3	300	0.057	129	0.057	22.5	81	Ductile
UMR-4	600	4.8×10^{-5}	88	4.8×10^{-4}	--	88	Fracture in fillet radius
UMR-5	600	0.057	91	0.057	23	76	Ductile
UMR-6	300	0.057	98 ^c	0.057	--	98	Brittle, fracture in threads
UMR-7	75	33	171 ^c	33	0.4	171	Brittle, failed in threads
UMR-8	75	19	179 ^c	19	0.5	179	Brittle, failed in threads
UMR-9	600	99	171	98	9.0	73	Ductile
UMR-10	600	73	157	70	11.0	70	Ductile
UMR-11	600	45	137	45	11.8	71	Ductile
UMR-12	600	27	125	29	12.6	72	Ductile
UMR-13	300	23	162	24	9.6	91	Ductile
UMR-14	300	31	166	31	7.5	93	Rough but ductile
UMR-15	300	75	189	75	6.8	115	Slightly rough
UMR-16	300	--	164 ^c	--	0.5	164	Fracture in threads
UMR-17	300	65	170 ^c	65	0.4	170	Fracture in threads
UMR-18 ^d	75	45	190 ^c	45	0.3	190	Fracture in threads
UMR-19 ^d	75	23	185 ^c	23	0.3	185	Fracture in threads
UMR-20 ^d	75	0.057	162	0.057	17	90.5	Ductile
UMR-21 ^e	75	11	196	14	6.2	105	Ductile
UMR-22 ^e	75	4.8×10^{-4}	145	4.8×10^{-4}	23	140	Ductile
UMR-23 ^e	75	25	141 ^c	25	0.5	141	Fracture in threads
UMR-24 ^e	75	19	186 ^c	19	0.5	186	Fracture in threads
UMR-25 ^e	75	15	189 ^c	15	0.3	189	Fracture in threads

^aAverage value to peak stress.^bAverage value from peak to rupture stress.^cFractured in threads.^dGiven special high vacuum heat treatment.^eGiven special short time heat treatment.

Table 4 Test data on uranium-10 wt % molybdenum alloy, as-cast condition with low carbon content

Test no.	Temp., ° F	$\dot{\epsilon}$, ^a in./in./sec	σ_{peak} , ksi	$\dot{\epsilon}$, ^b in./in./sec	$\epsilon_{\text{rupture}}$, %	σ_{rupture} , ksi	Failure mode
UMC(LC)-1	75	21	194	22	2.5	179	Brittle
UMC(LC)-2	75	27	201	28	2.5	174	Brittle
UMC(LC)-3	75	39	207	32	2.5	192	Brittle
UMC(LC)-4	75	70	185	70 ^c	2.0	170	Brittle
UMC(LC)-5	75	65	205 ^c	65	0.4 ^c	205	Brittle, fracture in threads
UMC(LC)-6	600	12	118	11	8.5 ^c	69	Ductile, rough
UMC(LC)-7	600	62	150	50	3.0	72	Slightly ductile
UMC(LC)-8	600	37	128	56	10.0	66	Ductile
UMC(LC)-9	600	0.057	78	0.057	26	76	Very ductile
UMC(LC)-10	600	4.8×10^{-5}	72.5	4.8×10^{-4}	18	71	Ductile

^a Average value to peak stress.

^b Average value from peak to rupture stress.

^c Approximate value.

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Fig. 13 Oscilloscope record of hot-rolled specimen UMR-8 tested at room temperature. Specimen failed in a brittle manner in threads.

(a) Load versus time. (b) Integrated velocity versus time.

(c) Fractured specimen

Fig. 14 Microstructure of impacted hot-rolled specimen UMR-7. $\times 200$.

Specimen was tested at room temperature at a strain rate of 33 in./in./sec and failed in the threads. Note the intergranular crack beginning at the thread root.

Fig. 15 Oscilloscope record of hot-rolled specimen UMR-11 tested at 600° F.

Specimen failed in a ductile manner. (a) Load versus time.

(b) Integrated velocity versus time. (c) Fractured specimen

Fig. 2-U L-12357
Author: K. G. Hoge

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Fig. 3-UCRL-12357
Author: K. G. Hoge

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Fig. 8 - UCRL-12357
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TIME - 1msec/cm

(a)

TIME - 1msec/cm

(b)

INTEGRATED VELOCITY - 0.2volt/cm

(c)

(d)

(a)

(b)

TIME - 0.2 msec/cm
(a)

TIME - 0.2 msec/cm
(b)

INTEGRATED VELOCITY -
0.05 volt/cm
(c)

(d)

TIME - 0.5 msec/cm
(a)

TIME - 0.5 msec/cm
(b)

(c)

TIME - 0.5 msec/cm
(a)

TIME - 0.5 msec/cm
(b)

(c)