Thermomechanical Process Optimization of U-10wt% Mo – Part 1: High-Temperature Compressive Properties and Microstructure

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ABSTRACT

Nuclear power research facilities require alternatives to existing highly enriched uranium alloy fuel. One option for a high density metal fuel is uranium alloyed with 10 wt% molybdenum (U-10Mo). Fuel fabrication process development requires specific mechanical property data that, to date has been unavailable. In this work, as-cast samples were compression tested at three strain rates over a temperature range of 400 to 800°C to provide data for hot rolling and extrusion modeling. The results indicate that with increasing test temperature the U-10Mo flow stress decreases and becomes more sensitive to strain rate. In addition, above the eutectoid transformation temperature, the drop in material flow stress is prominent and shows a strain-softening behavior, especially at lower strain rates. Room temperature x-ray diffraction and scanning electron microscopy combined with energy dispersive spectroscopy analysis of the as-cast and compression tested samples were conducted. The analysis revealed that the as-cast samples and the samples tested below the eutectoid transformation temperature were predominantly γ phase with varying concentration of molybdenum, whereas the ones tested above the eutectoid transformation temperature underwent significant homogenization.

1. Introduction

For several years, research to develop alternate fuels for nuclear research and test reactors has been underway to replace high enriched uranium fuel with low enriched uranium (LEU) fuel

[1, 2]. To achieve similar ^{235}U content with LEU, given reactor space constraints, a higher bulk density of fuel is needed, necessitating the use of metal alloy fuel. For this, an alloy that offers a good combination of irradiation performance, oxidation resistance, strength, and ductility is required for test-reactor applications. Various uranium alloys have been evaluated for many decades looking at mechanical properties and susceptibility to oxidation and corrosion [3-6]. Uranium alloyed with 10 wt% molybdenum (U-10Mo) shows improvement in these characteristics compared to similar alloys [1].

The current designs for LEU-10Mo fuel are thin foils (i.e., ranging from approximately 150 to 500 μm) bonded to zirconium and encapsulated in aluminum [7-9]. Given the high room temperature strength of U-10Mo and the need for a metallurgical bond to the zirconium, hot working of the cast plate is required. To optimize the fuel fabrication development, modeling of the hot working behavior is essential. The first step in modeling the process is to ascertain accurate mechanical property information at the appropriate conditions. For hot working prediction, in particular, the flow stress must be tested over a range of expected strain rates and temperatures. There is very limited mechanical data available in the literature for U-10Mo alloy for the expected process conditions (temperatures and strain rates). Generally, that which is available, such as Saller [6] and McGeary [5], is not in a relevant composition or processing condition, making precise predictions difficult. Although several studies have evaluated the properties of uranium alloys, none have looked specifically at the strain-rate sensitivity of U-10Mo [2, 3, 6, 7, 9-12].

The current work summarizes initial results conducted at the Pacific Northwest National Laboratory (PNNL) to determine the flow stress at conditions that represent hot working of U-10Mo. As-cast samples were tested in compression at temperatures between 400 and 800°C and at nominal strain rates between 1×10^{-1} and $1 \times 10^{-5} \text{ s}^{-1}$. The samples were then characterized by scanning electron microscopy (SEM), energy dispersive x-ray spectroscopy (EDS) and x-ray diffraction (XRD).

2. Experimental

2.1 Materials

The compression samples were machined from 38.1 mm-long by 6.35 mm-diameter pins that had been arc melted and cast into copper molds at the Idaho National Laboratory (INL),

Idaho Falls, USA. The feedstock used in the arc melting was in the form of pieces sectioned from a coupon that was machined from a vacuum-induction-melted and cast plate of U-10Mo. The plate was melted at 1330°C, cast into a graphite mold and homogenized in-mold at 1100°C for 4 h. All materials processing, other than that performed at INL, was conducted at the Y-12 National Security Complex, Oak Ridge, USA. The bulk chemical composition of the arc-cast samples was not measured; however, the composition of the coupon was determined to be U 88.7 wt% (99.8 wt% ^{238}U), Mo 10.1 wt %, C 0.08 wt%, Si 0.04 wt%, with Al, Co, W and Cu accounting for less than 1.5 wt%. The oxygen content in the samples was not measured and due to the homogenization of the feedstock material in the carbon crucible the final composition may contain a significant proportion of carbon.

2.2 Compression Sample Geometry

A sample configuration of 6.35 mm in diameter and 12.5 mm in length, shown in Figure 1a, was selected based on the availability of INL's pin casting mold and an existing test fixture at PNNL (Figure 1b). The 2:1 length-to-diameter ratio was considered the maximum ratio allowable to avoid buckling and still provide an accurate flow stress.

As previously mentioned, no further thermomechanical processing or heat treatments were conducted on the as-cast samples prior to testing. The ends of the right-cylindrical specimens were examined and any cutting burrs were removed. Samples were measured to evaluate end-to-end perpendicularity; if required, minimal wet sanding was performed using a fixture to square the ends.

2.3 Mechanical Testing

Mechanical testing was performed in a servo-hydraulic Instron tensile testing machine equipped with a box-type furnace with an argon gas atmosphere. Before heating, the furnace was purged at an elevated Ar flow rate (estimated to be five furnace exchanges) for over 900 s (15 mins). During testing, a continuous Ar flow flushed oxygen away from the specimen to minimize oxidation; evidence of slight oxidation was observed on the sample surface.

A custom-designed compression-test fixture that applies compression with cross-head tension was used to minimize the effect of any misalignment and is shown in Figure 1b. Boron nitride was applied to the platens prior to testing to act as a lubricant and prevent any reaction

between the U-10Mo and the Ni-based alloy fixture. Heating rates were consistent throughout testing. Samples were heated from room temperature to 350°C at 10°C/min. After holding for 1800 s (30 min) at 350°C, the temperature was raised, typically at a rate of 5°C/min, to the desired test temperature. Once the test temperature was reached, the sample was allowed to soak for 15 min and then compressed. The tests were performed at 400, 500, 600, 700 and 800°C and at strain rates of 6.7×10^{-2} , 6.5×10^{-4} and $8.6 \times 10^{-6} \text{ s}^{-1}$. Due to the desire to measure flow stress at high strains and the test system configuration, compression tests were conducted without a method to directly measure strain from the sample, and the cross-head displacement was used as the measure of strain. Therefore, compressive load for each test was measured as a function of the linear voltage displacement transducer in the cross-head. To increase the accuracy of the stress-strain data, displacements were later adjusted to compensate for the compliance (elasticity) of the testing system by conducting compression tests using the test fixture without specimens at representative test temperatures. To calculate strain, the displacement within the system without a sample as a function of load was subtracted from the actual test data. One example of the stress-strain before and after correction is given here in Figure 2, for a sample strained at a strain rate of $6.5 \times 10^{-4} \text{ s}^{-1}$ at 400°C, and shows that the stress-strain is effectively moved to the left eliminating the “extra” strain of the system compliance. All the stress-strain data given in this report were adjusted for compliance.

In addition to stress-strain compliance, the actual true-strain rates ($\dot{\epsilon}$) for each test were determined from the slope of true strain (after compliance correction) versus time and were found to be 6.7×10^{-2} , 6.5×10^{-4} , and $8.6 \times 10^{-6} \text{ s}^{-1}$, for the nominal strain rates of 1×10^{-1} , 1×10^{-3} and $1 \times 10^{-5} \text{ s}^{-1}$, respectively. Tests were performed up to a true strain of 0.4 (40%) and were then stopped, except for the $8.6 \times 10^{-6} \text{ s}^{-1}$ tests which were run for a fixed 2 h period.

For each test, the true flow stress, σ_f , was determined at a true offset strain of 0.1 (10%). The flow stress and true-stress/true-strain curves for each test were used to calculate strain-rate sensitivity, m , strain-hardening exponent, n , and strain-hardening coefficient, K (Equations 1 and 2). [13]

$$\sigma = C \dot{\epsilon}^m |_{T, \epsilon} \quad (1)$$

$$\sigma = K\epsilon^n|_{T,\dot{\epsilon}} \quad (2)$$

2.4 Microstructural Characterization and Phase Analysis

Microstructural characterization was performed on samples that were strained (up to 0.4 strain), as-cast and heated with the same heating and cooling rate used in the compression test to compare the effect of strain on thermal cycling. All microstructural samples were characterized transverse to the compression direction approximately midway between the ends (5 to 6 mm). The actual strain in the samples used for microstructural characterization was close to 0.4, although due to the barrel shape (after compression testing) and polishing differences the strain may have varied slightly. For the cross sectional analysis, the compression tested samples were cold mounted in an epoxy resin and then cross sectioned at approximately half the length or at the region that experienced maximum strain. The mounted samples were successively polished down to 600 grit with SiC polishing papers and were further polished using 6 μm and 1 μm diamond slurries, followed by a final colloidal silica polish. Upon polishing the samples were carbon coated for SEM analysis. The detailed technique used to prepare the samples for characterization can be found elsewhere. [14] For purposes of comparison in this paper, only results from the $6.5 \times 10^{-4} \text{ s}^{-1}$ strain rate are presented. Microstructural characterization was performed using an optical microscope as well as using a JEOL JSM-7600F SEM equipped with an Oxford Instruments X-Max 80 EDS detector. The EDS analysis was performed using the INCA Microanalysis Suite software, version 4.15.

The XRD pattern/phase analysis on the sectioned metallographic samples was performed on a Panalytical X'Pert Pro Alpha-1 machine that uses an Xcelerator detector. The source was a ceramic copper tube which was operated at 45 kV and 40 mA. The data was collected at a step size of 0.02° for 35.05 s at each step.

3. Results

3.1 Mechanical Testing

The true stress- true strain curves for the samples compression tested at different strain rates and temperatures are provided in Figure 3a-c. It was observed that at a particular strain rate the

samples that were tested below the eutectoid temperature (550°C) showed significantly higher yield stress and flow stress as compared to those that were compression tested above the eutectoid temperature by over 50%. Figure 4 presents the flow stress at 10% strain as a function of temperature and strain rate. It can be observed that at a particular temperature, as the strain rate increases the flow stress increases both above and below the eutectoid transformation temperature. As mentioned earlier, at a particular strain rate the drop in flow stress above the eutectoid temperature is nonlinear as can be observed from Figure 4. At $6.7 \times 10^{-2} \text{ s}^{-1}$ strain rate the drop in flow stress from 500 to 700°C is nearly 50%, whereas at 6.5×10^{-4} and $8.6 \times 10^{-6} \text{ s}^{-1}$ the drop in flow stress was over 65%. (DELETED A SECTION ON YOUNG'S MODULUS)

Compressive true-stress/true-strain curves at various temperatures and strain rates, like those given in Figure 3, were used to determine n , m , and K . The values for n , m and K determined from the true-stress/true-strain curves have been summarized in Tables 1 and 2. The as-cast U-10Mo exhibited a strain-rate sensitivity as a function of temperature, and samples tested at 400°C exhibited no strain-rate sensitivity ($m = 0.0$) while the maximum strain-rate sensitivity, $m = 0.22$, was observed at 800°C. The strain-hardening exponent, n , and strength coefficient, K , were calculated using Equation 2 and are given in Table 2 and show that n varied from positive values at high strain rates and low temperatures to negative values at low strain rates and high temperatures and, as expected, K decreased with temperature and the corresponding flow stress. This trend is consistent with the marked drop in flow stress observed previously between temperatures above and below the eutectoid temperature.

3.2 Microstructural Analysis:

3.2.1 As-cast samples:

XRD analysis of the as-cast sample was performed and, as shown by Figure 5, confirmed that the material was predominantly γ phase, with small amounts of α and UO_3 phases. SEM of the as-cast samples was performed and indicated a relatively equiaxed grain size of less than 25 μm that was uniform across the entire cross-section, as shown by the image in Figure 6a. Higher magnification SEM, shown in Figure 6b-c, revealed the presence of at least two distinct additional phases, a globular cross-shaped particle located within the grain and a needle-like phase located predominately at the grain boundaries. EDS dot and map analysis, shown in Figure 6b-c, was performed on the “globular” and “needle” phases and indicated the globular phase was

rich in yttrium and oxygen and the needle phase was rich in carbon when compared to the uranium-molybdenum matrix. No structure representative of the eutectoid transformation was observed in the microstructure.

Elemental mapping by EDS indicated that there was micro-segregation across the entire sample for carbon, oxygen, molybdenum and silicon, as shown in Figure 7. In addition to the EDS mapping, the atomic number contrast (Z contrast) was observed in the backscattered-electron (BSE) images and indicated that the molybdenum content within the grain varied. From the molybdenum distribution (as indicated by EDS and visual Z contrast) it was determined that the difference in the molybdenum concentration in the intra-dendritic region and inter-dendritic region was approximately 2 wt%, i.e. the intra-dendritic regions were rich in molybdenum and near cast composition. The overall cast microstructure appears to have been developed by dendrite nucleation from oxide phase. The molybdenum segregation may have occurred during the peritectic transformation [15] due to the difference in solidus and liquidus and this also resulted in the displacement (pushing) of the carbides to form a near continuous network of carbides. Because there were many nuclei and a relatively high solidification rate, normal columnar dendrite growth was interrupted resulting in a relatively fine (25 μm), nearly equiaxed grain size. The carbides in this case may have been formed in the feedstock, which was cast in the graphite molds (Section 2.1), whereas the yttria was probably incorporated into the structure from the die mold coating.

3.2.2 Samples compression tested at 500°C, at $6.5 \times 10^{-4} \text{ s}^{-1}$ and undeformed sample:

The XRD analysis, shown in Figure 5, of the sample compression tested at 500°C and at a strain rate of $6.5 \times 10^{-4} \text{ s}^{-1}$ revealed the presence of only γ phase and no other peaks corresponding to other phases were observed. The BSE SEM image of the same sample revealed a microstructure similar to the cast microstructure (dendrites, inter-dendritic regions, carbon-rich needles and oxygen-rich phases) and is shown in Figure 8a. EDS analysis of these regions indicated chemistry similar to the as-cast samples. The overall increase in the grain size was approximately two times of that of the as-cast sample (between 50 and 60 μm). Apart from the aforementioned phases, a cellular/lamellar structure was also observed in these samples. These were primarily located at the grain boundaries as shown in Figure 8b and c. Some of these lamellae had a distorted structure, likely due to the compression testing. The interlamellar

spacing was less than 50 nm and hence it was difficult to discern the chemistry of the individual phases via EDS. However an overall molybdenum concentration of the interlamellar/ cellular phase was between 5 and 7 wt%. The composition and the volume fraction of the sample were low enough to detect the new phases, if any, by the XRD and hence did not show in the analysis.

The sample that underwent the same thermal profile as that of the compression tested sample at 500°C without strain had a microstructure similar to the as-cast sample and its strained counterpart. The BSE SEM image is shown in Figure 9a and indicated a grain size of approximately 35–50 μm with the same four phases, but in this case the lamellar structure was more pronounced and formed predominantly along the grain boundaries as shown in Figure 9b. The lamellar structure varied in size from less than 1 μm to over 10 μm with interlamellar spacing of less than 100 nm. The determination of phases was beyond the scope of current work; however, based on the Z-contrast imaging, the U-Mo phase diagram and the reported [16-20] time-temperature-transformation (TTT) curves, one can say that the structure is probably composed of $\alpha + \text{MoU}_2$.

3.2.3 Samples compression tested at 700°C, at $6.5 \times 10^{-4} \text{ s}^{-1}$ and undeformed sample:

The sample that was compression tested above the eutectoid temperature in the γ -phase field at 700°C and at a strain rate of $6.5 \times 10^{-4} \text{ s}^{-1}$ revealed only the γ phase as shown by the XRD analysis in Figure 5. The BSE SEM image of the same, shown in Figure 10, indicated fewer dendritic/popcorn-like structures and certain grains appeared to have completely homogenized, unlike the as-cast samples or the samples strained or heated at 500°C. The grain size was approximately 80 to 100 μm . The same four phases and the lamellar phase as found in the as-cast and 500°C samples was observed. However, the lamellar phase, shown in Figure 10b, for the 700°C test was not deformed and may have formed during the sample cooling. It was observed that the average concentration of molybdenum in the inter-dendritic space increased by approximately 2 wt% and tended more towards homogenization.

As compared to the compression-tested sample, the sample that underwent the same thermal history as that of the compression tested one showed a microstructure similar to the as-cast microstructure as shown by Figure 11. The homogenization in this case was minimal and the dendritic regions retained their form and structure. The grain size in this case was approximately 20–30 μm , comparable to the as-cast samples. It is noteworthy that unlike the compression-tested

sample, this sample had no lamellar phase. The EDS analysis of the sample revealed that the composition of the molybdenum increased by nearly 1 wt% as compared to the as-cast sample.

3.2.4 Samples compression tested at 800°C, at $6.5 \times 10^{-4} \text{ s}^{-1}$ and undeformed sample:

The XRD pattern of the samples compression tested at 800°C and at a strain rate of $6.5 \times 10^{-4} \text{ s}^{-1}$ revealed the presence of γ phase similar to the previous samples, indicating concentration of other phases less than 5 wt%. The BSE SEM images, shown in Figure 12a, revealed significant homogenization of the microstructure as compared to the previous samples. The dendritic structure (dark gray areas imaged by the Z contrast in the BSE mode) is no longer visible but appears as a hazy structure in some of the grains, indicating nearly complete homogenization. The grain size in these as-tested samples increased by six times, to 150 μm , as compared to the as-cast structure. Nearly 50% of the grains were homogenized completely and no segregation was observed. The EDS analysis revealed a variation of molybdenum concentration from 8–9 wt% in the bright gray areas and between 10 and 11 wt% in the hazy gray areas. In these samples, as compared to the samples tested at 500 and 700°C, the lamellar phases were finer, as shown in Figure 12b. The uranium-rich carbides and oxides of yttrium were present and had retained their form and shape. These precipitates were homogeneously dispersed along the entire cross-section, indicating that grain growth had occurred beyond the original carbide-rich boundaries.

In contrast to the compression-tested sample, the untested sample that underwent the same thermal profile had retained the dendritic structure similar to the as-cast sample as shown by the BSE Z contrast in Figure 13. The variation in molybdenum concentration improved slightly. The grain size was approximately 15–20 μm in diameter and was decorated by the uranium-rich carbide phases. The lamellar phase was not observed in this case, similar to the previous sample and the as-cast sample.

4. Discussion

In the current work it was observed that the mechanical properties of the samples compression tested below the eutectoid temperature were significantly higher than those tested above the eutectoid region in the single-phase γ field. A continuity in the linear relationship of the mechanical properties as a function of temperature was interrupted as the tests transitioned

above and below the eutectoid temperature (Figures 3–4). The as-cast U-10Mo sample was more strain-rate sensitive as the temperature increased and the strain-hardening exponent decreased rapidly as the temperature increased. The strain hardening observed in the samples compression tested below the eutectoid temperature can be attributed to the presence of micro-segregation and presence of other phases however the lack of the same above the eutectoid temperature maybe attributed to competing mechanism between homogenization, dislocation motion, grain boundary effects and the role of carbides in the strength of the same. Thus, the results indicated that the microstructure played a vital role in the deformation mechanism.

The as-cast structure of U-10Mo consisted of a cored dendritic structure with predominant molybdenum-lean and molybdenum-rich γ phase regions. The dendritic regions contained approximately 10–11 wt% molybdenum, whereas the inter-dendritic regions were molybdenum lean and contained anywhere between 5 and 7 wt% molybdenum. Apart from that, the grain boundaries were rich in carbides of uranium. The high strength of the materials tested below the eutectoid temperature can be attributed to the wide distribution of these phases. The samples retained the as-cast structure with an exception of formation of lamellar structure in the inter-dendritic region. The chemical inhomogeneity and the presence of these aforementioned phases may have acted as significant barriers to dislocation motion, thus increasing the strength of the material. The formation of the lamellar phase in the molybdenum-lean regions (inter-dendritic) below the eutectoid temperature, i.e., without undergoing any eutectoid transformation, can be correlated back to the TTT curves [17–20]. The low diffusion rates [21–22] at the low temperatures tested prevented significant homogenization of the U-10Mo structure, thus conserving the molybdenum-lean and molybdenum-rich areas from the as-cast structure. A close look at the TTT curves shows that the time required to transform a 5 wt% molybdenum-rich alloy into $\alpha + \text{MoU}_2$ is approximately 1800 s (0.5 h), whereas the time required to transform the 10 wt% molybdenum-rich region is of the order of 3.6×10^4 s (10 h), thus explaining the formation of the lamellar structure in the molybdenum-lean regions/inter-dendritic regions. However a detailed set of experiments need to be conducted to prove this hypothesis and was beyond the scope of the current work. The distorted lamellar structure after the testing proves the hypothesis that they were formed rapidly and added to the strength of the material tested below the eutectoid transformation temperature.

The lowered flow stress and the strain hardening above the eutectoid transformation temperature can be explained by the significant homogenization occurring during the tests. It should be noted that, in contrast to the tested samples, the non-tested samples showed significant inhomogeneity in the structure. These results indicated that the stress may have played a significant role in homogenizing the sample and vice-versa.

5. Conclusions

This work evaluated the effect of temperature on the mechanical properties of U-10Mo under compression. The data generated will be beneficial to successfully model the rolling and other forming processes of this alloy. The following trends were observed:

- 1) An increase in temperature caused a drop in the flow stress and the strain-hardening exponent whereas the strain rate sensitivity increased drastically.
- 2) Above the eutectoid transformation temperature, the drop in flow stress was prominent and showed a strain-softening behavior, especially at lower strain rates of testing.
- 3) As the test temperature increased, the material became increasingly strain-rate sensitive.
- 4) The microstructure of the sample was not homogeneous and it is possible that the trends in properties observed were related to the evolving microstructure due to the high-temperature exposure during testing.
- 5) The development of accurate predictions for hot working will require knowledge of the temperatures and conditions.

To further the understanding of the effect of temperature on mechanical properties, detailed characterization of the U-10Mo at temperatures between 400 and 800°C needs to be conducted including:

- 1) Effect of time and temperature on the transformation. High-temperature XRD is needed to better understand the phases present when testing (or hot working).
- 2) Effect of hot deformation temperature on the recrystallization behavior and final microstructure.
- 3) Effect of stress at elevated temperature on the U-10Mo phase transformations.

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Table 1

Temperature	Flow Stress at 0.1 Strain, MPa			<i>m</i>
	$6.7\times10^{-2}\text{ s}^{-1}$	$6.5\times10^{-4}\text{ s}^{-1}$	$8.6\times10^{-6}\text{ s}^{-1}$	
400°C	765.3	861.8	772.2	0.00
500°C	765.3	737.7	413.7	0.07
600°C	620.5	324.0	117.2	0.18
700°C	408.2	158.6	117.2	0.14
800°C	268.9	89.6	34.5	0.22

Table 2

Temp °C	<i>n</i>			<i>K</i>		
	$6.7\times10^{-2}\text{ s}^{-1}$	$6.5\times10^{-4}\text{ s}^{-1}$	$8.6\times10^{-6}\text{ s}^{-1}$	$6.7\times10^{-2}\text{ s}^{-1}$	$6.5\times10^{-4}\text{ s}^{-1}$	$8.6\times10^{-6}\text{ s}^{-1}$
400	1.03	0.62	0.62	5521.1	3480.3	3276.4
500	0.71	0.56	0.15	3373.4	2591.8	597.4
550	0.82	–	–	3347.5	–	–
600	0.54	–0.20	–1.33	1764.1	203.5	9.8
700	0.50	–0.12	–1.08	917.9	120.2	15.4
800	0.63	0.17	–1.61	664.6	130.4	0.6

(a)

(b)

Figure 1

Figure 2

Figure 3

Figure 4

1

Figure 5

(a)

(b)

(c)

Figure 6

Figure 7

Figure 8

Figure 9

Figure 10

Figure 11

Figure 12

Figure 13

Table and Figure Captions

Table 1. Compressive Flow Stress and Strain-Rate Sensitivity as a Function of Temperature and Strain Rate for U-10Mo.

Table 2. Strain-Hardening Exponent, n , and Strength Coefficient, K , for U-10Mo Compression Tests

Figure 1. U-10Mo Samples Shown (a) Prior to Testing and (b) Inside the Test Fixture

Figure 2. Engineering Stress-Strain Plots Showing the Difference With and Without the Compliance Adjustment for a Cross-Head Velocity of $6.5 \times 10^{-4} \text{ s}^{-1}$ at 400°C .

Figure 3. Stress-Strain Curves of the U-10Mo for Selected Individual Tests as a Function of Strain Rate (i.e., 6.7×10^{-2} , 6.5×10^{-4} and $8.6 \times 10^{-6} \text{ s}^{-1}$) and Temperature

Figure 4. Effect of Flow Stress as a Function of Temperature and Strain Rate on U-10Mo.

(DELETED YOUNGS MODULUS FIGURE)

Figure 5. XRD Patterns of Samples As-Cast and After Compression Testing at Elevated Temperatures.

Figure 6: Backscattered SEM image of As-Cast Samples. (a) Low magnification image showing the grain size; (b) The oxygen rich phases; (c) and needle-like carbide precipitates.

Figure 7: EDS Map of the As-Cast Sample of the Image Shown in Figure 7c.

Figure 8: Backscattered SEM Image of the Sample Compression Tested at 500°C at a Strain Rate of $6.5 \times 10^{-4} \text{ s}^{-1}$. (a) Low magnification image showing the grain size and different phases; (b) and (c) cellular structure at the grain boundaries.

Figure 9: Backscattered SEM Image of the Sample Heat Treated to 500°C Following the Same Thermal Profile as that of the Compression Tested Sample (a) Low magnification image showing the grain size and different phases; (b) high magnification image showing the grain structure and the cellular structure.

Figure 10: Backscattered SEM Image of the Sample Compression Tested at 700°C at a Strain Rate of $6.5 \times 10^{-4} \text{ s}^{-1}$. (a) Low magnification image showing the grain size and different phases; (b) lamellar/cellular phase.

Figure 11: Backscattered SEM Image of the Sample Heat Treated to 700°C Following the Same Thermal Profile as that of the Compression Tested Sample. Low magnification image showing the grain size and different phases.

Figure 12: Backscattered SEM Image of the Sample Compression Tested at 800°C at a Strain Rate of $6.5 \times 10^{-4} \text{ s}^{-1}$. (a) Low magnification image showing the grain size and different phases; (b) magnified image of a region within the grain.

Figure 13: Backscattered SEM Image of the Sample Heat Treated to 800°C Following the Same Thermal Profile as that of the Compression Tested Sample. Low magnification image showing the grain size and different phases.