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THE METASTABLE GAMMA PHASE
IN URANIUM BASE MOLYBDENUM ALLOYS

by

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THE METASTABLE GAMMA PHASE
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Introduction

The rare earth element uranium exists in three allotropic forms. The low temperature phase, α , is orthorhombic with eight atoms per unit cell and is stable to 662°C. The β phase is tetragonal with 30 atoms per unit cell and is stable from 662° to 770°C. The high temperature phase, γ , is body-centered cubic and is stable to the melting point. Molybdenum is one of the body-centered cubic elements that tend to stabilize the γ phase.

The first thorough study of the uranium-molybdenum system was done by P. C. L. Pfeil¹ at Harwell. His results indicated that the molybdenum was soluble in γ uranium to the extent of about 23 w/o at 1280°C and that in the region of 10 w/o Mo the γ phase was stable to room temperature.

Further work by Pfeil and Browne² and by Tucker³ showed that equilibrium conditions had not been obtained and that the γ phase is metastable and transforms below about 600°C into a new tetragonal phase, γ' . Saller, Rough and Vaughan⁴ proposed that the decomposition of the γ phase was a eutectoid reaction and further that the new γ' phase was an ordered form of the body-centered γ phase. Pfeil's later constitutional diagram showed a complex behavior of the existence range of the γ' phase, with the existence range increasing with increasing temperature.

In an attempt to resolve the conflicting reports, work was initiated at the Bettis Plant to study various aspects of the uranium-molybdenum system. Most of the work was concentrated on those alloys in the range of 6 to 18 w/o molybdenum.

Phase Diagram

The results of this study of the uranium-molybdenum system are shown in Fig. 1. A more detailed diagram showing the composition of the samples used is given in Fig. 2. All material used was in the form of quadruple arc-melted buttons or induction melted ingots. Material from each button or ingot was analyzed by wet chemical and by X-ray fluorescence methods. The molybdenum content was determined to within 0.25 w/o. All alloys were homogenized by annealing at 900°C for 24 hours followed by a water quench. The heat treatment procedure consisted of sealing solid alloy samples in Vycor tubes, putting the

tubes in iron pipe jackets, then annealing in constant temperature salt for at least 3 months. Powdered samples were not used because of the difficulty of avoiding oxide and nitride formation during the long anneals.

The solid samples were examined before and after annealing by means of X-ray techniques utilizing a Norelco spectrometer with a specimen rotator. CoK_{α} radiation was employed. Any oxide or nitride formed during the anneal was removed by grinding prior to examination. The samples were polished metallographically in the conventional manner and subsequently polished electrolytically by using a chromic-acetic acid solution. During grinding operations X-ray examination revealed that the equilibrium $\alpha + \gamma'$ structure was transformed back to the high temperature γ phase. In view of this behavior, all metallographic samples were examined by X-ray diffraction methods before and after metallographic preparation to verify that the phases present after annealing had been retained.

The diagram shown in Fig. 2 agrees with that originally proposed by Saller, et al.⁴ The γ phase decomposes by a eutectoid reaction into the α and γ' phases. This reaction may be considered as two separate reactions, the precipitation of the α phase and the ordering of the γ' phase. The ordering reaction was originally postulated because the structure of γ' phase differs only slightly from that of γ , and thus only a small volume change occurs upon transformation. Further evidence for ordering was found by Pfeil and Browne² using electrical resistivity measurements. They proposed a phase diagram which differs somewhat from those of Battelle and Bettis. They determined the phases present by examining X-ray diffraction patterns from annealed powders. Their longest reported anneal was for 4 weeks. In the British diagram the eutectoid was replaced with a minimum in the γ field that occurred at 11.5 w/o Mo. The γ' field widened at higher temperatures with the $\alpha + \gamma'/\gamma'$ boundary located at about the position of the dotted line in Fig. 2.

The $\alpha + \gamma'$ field can be divided into two regions, the dotted line showing the approximate line of demarcation. In the low molybdenum region the rate of precipitation of the α phase exceeds the rate of ordering of the γ phase to γ' . In the high molybdenum region the rate of ordering exceeds the rate of α precipitation. A typical photomicrograph of the low molybdenum region is shown in Fig. 3. The dark plates are α and the white areas γ that has ordered to γ' . A typical photograph of the high molybdenum region of the $\alpha + \gamma'$ phase field is

shown in Fig. 4. This sample was annealed at 550°C and during the initial stages of transformation it ordered to γ' . At 550°C the long range order parameter S is considerably less than one and it is not possible to obtain perfect order. Thus, formation of a pure γ' phase having the correct stoichiometric ratio is difficult and it is equally as difficult for the uranium atoms to agglomerate and precipitate as α . After anneals in excess of 100 days at 550°C the bands in the γ' phase lose their original sharpness and α appears as a fine intragranular precipitate. These small amounts of α can also be detected through X-ray diffraction by step scaling rather than scanning across the diffraction lines. The most conclusive evidence for the simple eutectoid type of diagram is the detection of the α phase in six samples (ranging in composition from 14.0 to 15.5 w/o) that had been annealed at 550°C for over 100 days.

Water quenching is sufficient to retain the γ' phase if the molybdenum concentration exceeds about 7 w/o. For alloys in the composition range of 2 to 7 w/o, water quenching produces a diffusionless phase change.⁵ Because of the similarity of the X-ray powder patterns of this new phase to those of α uranium the phase has been designated as α' . In very rapidly quenched specimens the transformation can be induced by scratching the surface. A typical photomicrograph of the α' structure is shown in Fig. 5. The crystal structure of the α' phase has been determined from a single crystal of U-5 w/o Mo, water quenched from 900°C. The structure is base-centered orthorhombic with $a = 7.00 \text{ \AA}$, $b = 6.84 \text{ \AA}$, and $c = 3.42 \text{ \AA}$.

Annealing of alloys in the composition range 5 to 7 w/o molybdenum for periods of at least 30 days at temperatures below 400°C produces a new structure. The X-ray spectrometer pattern consists of three or four very broad and unresolved lines in the same region of 2θ values for which sharp lines of α and γ' normally appear. This type of pattern is typical of the structure that has been designated the distorted α structure. The crystal structure of this material is not known.

A typical photomicrograph of the $\alpha + \gamma$ structure is shown in Fig. 6. It can be seen that nodular growth of the plate-like structure occurs at the grain boundaries. The $\alpha + \gamma/\gamma$ boundary was determined by Pfeil using metallographic techniques. The transformation in this region occurs within a few days so that for the lower molybdenum concentrations this boundary is correct as originally determined.

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The appearance of the $\gamma + \gamma'$ structure is shown in Fig. 5. The exact relationship between the $\gamma + \gamma'$ and γ' field was difficult to derive by X-ray techniques because the detection of γ in the presence of γ' is complicated by the superposition of the 110 reflections. For randomly oriented γ' the integrated intensity of the 110 reflection should be 55% of the integrated intensity of the 103 reflection. This figure has been found to vary for different samples which metallographically were apparently all γ' . This intensity variation is presumably due to variations in composition and varying degrees of preferred orientations in the samples. For a rapid estimation of the amount of γ present in an ordered matrix, it was assumed that any integrated intensity of the 110 reflection in excess of 55% of the integrated intensity of the 103 reflection was due entirely to the residual γ phase.

Structure of the Gamma Prime Phase

When the γ phase transforms to the γ' phase the unit cell changes from a $10\bar{6}$ -centered cubic structure in which the uranium and molybdenum atoms are distributed at random on the lattice sites to a tetragonal unit cell⁶ in which the atoms occupy preferred sites. The ordered phase consists of warped layers of atoms with every third layer being molybdenum atoms. The new cell can be considered to consist of three pseudo cubes as shown in Fig. 8. The axial ratio for one of the pseudo cubes is 0.957, which permits easy twinning along the 103 planes. The metallographic appearance of the ordered phase is shown in Fig. 9.

The determination of the crystal structure of the γ' phase was first attempted by means of X-ray powder data from bulk samples that had been given an ordering heat treatment. Insufficient superlattice reflections were observed with this technique to determine uniquely the unit cell or the ordering arrangement. An accurate X-ray determination of the structure was made by utilizing a tricrystal of γ' obtained from a single crystal of γ that had been given an ordering anneal. A total of 23 superlattice reflections were successfully indexed in the single crystal pattern obtained from this tricrystal. The unit cell was found to be tetragonal with lattice parameters of $a_0 = 3.427 \text{ \AA}$, $c_0 = 9.834 \text{ \AA}$, $c_0/a_0 = 2.871$. The molybdenum atoms are located at the positions 000 and $1/2, 1/2, 1/2$. The uranium atoms are at the positions $00\bar{Z}$; $00\bar{Z}$; $1/2, 1/2, 1/2 + Z$; and $1/2, 1/2, 1/2 - Z$. By plotting the structure factor as a function of Z and taking into account the observed superlattice reflection intensities it was determined that the parameter Z should have a value of

0.328 ± 0.002 . This structure is the Cl1b MoSi_2 type and has been found in the following intermetallic phases: MoSi_2 , MoGe_2 , WSi_2 , ReSi_2 , MgHg_2 , and AlCr_2 .

In the disordered γ phase each atom is surrounded by eight random atoms at a distance of $2.93 \text{ \AA}$. In the ordered γ' phase each molybdenum atom is surrounded by eight uranium atoms at a distance of $2.96 \text{ \AA}$; each uranium atom is surrounded by four molybdenum atoms at $2.96 \text{ \AA}$ and four other uranium atoms at $2.87 \text{ \AA}$. The interatomic distance for pure molybdenum is $2.72 \text{ \AA}$. Thus during the ordering process the larger uranium atoms become closer together and attain an arrangement of pairs of atoms similar to that of α uranium. After ordering, the smaller molybdenum atoms are surrounded by atoms at a slightly greater distance than in the disordered state.

The question of validity of the postulate that γ' is an ordered arrangement of γ was tested by using calorimetric and X-ray intensity measurements.⁷ Early work by Pfeil² with electrical resistivity measurements indicated the possibility of an ordering reaction. The electrical resistivity of previously homogenized samples of 25 a/o molybdenum decreased 5% when isothermally annealed at 500°C for 200 hours. A further anneal of this sample at 400°C produced a further decrease in the resistivity of 7%. As samples given similar anneals showed very little precipitation of α , the resistivity decrease was attributed to the ordering of the atoms to a higher degree of equilibrium order at 400°C than at 500°C .

Calorimetric Measurements

When a sample that is in a state of high order is heated below the critical temperature the long-range order is partially destroyed by thermal agitation and the energy required for this process appears as an increase in the magnitude of the specific heat over the value expected from the simple Debye theory. This excess specific heat depends upon the destruction of long-range order and thus approaches a maximum at the critical temperature. Above the critical temperature there is still some degree of local order which is destroyed by heating, but the specific heat rapidly returns to the value predicted by the Debye theory. Thus the appearance of an excess specific heat spread over a range of temperature is a positive indication of a disordering mechanism rather than a simple phase change or a change in the relative concentration of two phases.

A modification of a calorimetric method devised by C. S. Smith⁸ was used to obtain the specific heats and transformation temperatures of the well ordered

U-Mo alloys. Installation of the calorimeter assembly in a high vacuum was necessary to prevent excess gaseous contamination of the easily oxidized sample. The method consists of placing the test specimen in a refractory container of low thermal conductivity and then allowing a constant heat flux to flow through the walls of the container. The time the sample requires to increase in temperature is proportional to the heat absorbed. A sample of spectrographically pure aluminum was used to calibrate the apparatus. A sample of Cu-6 w/o Be, which was presumed to be the eutectoid composition, was run as a test specimen because its eutectoid temperature of 608°C is close to that of 575° found in the U-Mo alloys. The time versus temperature record for this material is shown in Fig. 10 along with a similar record for an ordered U-21.5 w/o Mo alloy. At the eutectoid temperature of the Cu-Be sample the slope actually becomes infinite and the transition occurs over a small temperature interval of about 3°C. The sensitivity of the apparatus is indicated by the fact that a slight deviation from the eutectoid composition existed in the sample and an additional change in slope occurred when the sample passed through the two-phase field. For the U-Mo sample the phase change began gradually and was spread over an interval of at least 31°C.

A series of U-15.2 w/o Mo samples was given an ordering treatment in which the samples were annealed at 590°C for 7 days and subsequently slowly cooled to an ordering temperature of 550, 500, or 400°C where they were held for times of 76, 44, or 25 days. The results of the specific heat measurements are shown in Fig. 11. The complete heating curve for the 500°C sample is shown in Fig. 12. The excess energy required for disordering begins to appear at about 540, 525, and 497°C, which agrees qualitatively with the final ordering temperature. The sample whose final annealing temperature was 400°C was apparently cooled too rapidly through the 500°C range and therefore insufficient time was spent at the lower temperatures to develop any additional degree of order. Additional samples of U-7, 8, 9, 10.5, 11.3, and 12 w/o Mo were also given long-term ordering anneals. The excess specific heat appeared in the temperature range of 450 to 500°C. Thus, there is no calorimetric evidence for the presence of $\alpha + \gamma'/\gamma'$ boundary in the region of the dotted line of Fig. 2. Pfeil and Browne's results can be best explained on the basis of nonequilibrium conditions as their samples were isothermally transformed for times of only 28 days. In an earlier determination of the phase diagram at Bettis Plant, when similar heat treatment times of 35 days were used, results similar to those of Pfeil and Browne were obtained.

The presence of long-range order is most conclusively demonstrated by plotting the change with temperature of the long-range order parameter. The samples used for the calorimetric measurement were examined in an X-ray spectrometer with a sample rotator. CuK_α radiation was used before and after calorimetric runs. The value of the long-range order parameter was determined from the 105 and 110 normal lattice reflection and the 112 superlattice reflection. The X-ray results obtained on the series of 15.2 w/o molybdenum samples used for this specific heat measurement are shown in Fig. 13. The temperatures plotted are the final ordering temperatures prior to quenching. The value of S for the sample whose final temperature was 400°C appears quite low, most probably due to the excess time required to reach equilibrium at these low temperatures. This sample is from the series previously mentioned and had spent the least time at the lowest final ordering temperature. The remaining three samples show a monotonic decrease of S with increasing temperature in a fashion similar to that of AB_2 alloys.

Thermal Dilation

Numerous dilation curves for binary uranium-molybdenum alloys have been determined⁹ for temperatures up to 900°C . Autographic records produced by an X-Y recorder for U-9, 12, 15, and 18 w/o Mo and for natural uranium are reproduced in Fig. 14. Alloy samples were 0.600 inch long; the uranium sample was 0.724 inch long. All alloy samples were transformed by annealing at 525°C for 15 days before being heated in the dilatometer. The 9 and 12 w/o samples expanded 0.068 and 0.053% upon decomposition of the γ' phase whereas the 15 and 18 w/o samples contracted 0.034 and 0.041%. These curves also show that the disordering occurs rapidly upon heating and that the γ phase is retained upon cooling at 100°C per hour.

Mechanical Properties of U-Mo Alloys

The mechanical properties of uranium-base alloys containing up to 14 w/o molybdenum have been determined at temperatures up to 800°C as a function of both temperature and prior thermal history.⁹

The relationship between tensile strength, elongation, reduction in area, and heat treatment for U-12 w/o Mo alloys is shown in Table I. The relationship between temperature, tensile strength, and strain rate for these alloys is shown in Fig. 15. A maximum in ductility at each strain rate was noted with increasing

temperature of test; the faster the strain rate the higher the temperature at which maximum ductility is obtained. It has been observed that for samples tested at temperatures that resulted in maximum ductility for a given strain rate, the load-elongation curves show an increase in load between the yield point and the breaking load, whereas at other temperatures, little or no work hardening was observed. Serrations with amplitudes of ± 10 to 25 pound load and a frequency of 1500 to 1800 per inch of extension were also observed in this portion of the curves, indicating strain-induced transformation at elevated temperatures.

Low temperature testing of the 12 w/o molybdenum alloy indicated an abrupt ductile-brittle transition occurs between -15°C and -31°C when the alloy is tested at 0.05 inches per minute strain rate.

Young's modulus was determined at room temperature and found to be

$$E = 14.3 \times 10^6 \text{ psi.}$$

The impact properties, as determined from unnotched Izod specimens, for a U-12 w/o Mo alloy annealed at 900°C for 24 hours and subsequently water quenched are shown in Table II.

Electrical Resistivity

The electrical resistivity of U-Mo alloys was used by Pfeil and Browne² to study ordering in the γ phase and by Saller, et al,⁴ and Van Thyne and Peterson¹⁰ to follow the decomposition of the metastable γ phase. All of these measurements were made at room temperature or at temperatures between 20° and 500°C . The first low temperature measurements were made by Eichenberg¹¹ who observed that the resistance at liquid nitrogen temperature was higher than at room temperature. This anomalous result prompted a more detailed study of the temperature dependence of the electrical properties of U-Mo alloys.

Resistivity measurements over an extended temperature range for a series of gamma quenched U-Mo alloys were performed by Chandrasekhar, Flinn, and Halteman.¹² The values of R/R_0 as a function of absolute temperature for a 9 w/o sample are shown in Fig. 16. R_0 is the resistivity at 300°K . The samples containing 3 to 15 w/o molybdenum showed a behavior similar to that of the 9 w/o sample, i.e., a resistivity practically independent of temperature; the resistivity never varied by more than 5% from the room temperature value. All of the alloys tested exhibited an anomalous negative temperature coefficient in

From 4°K to about 400°K. All of the alloys showed a positive temperature coefficient above 900°K in the $\alpha + \gamma$ or γ phase field. The 1 w/o sample did not retain the γ phase upon quenching and thus showed a behavior similar to that of pure uranium, i.e., a positive coefficient at room temperature.

Measurements were made by Bleiberg, Jones, and Lustman¹³ of the temperature variation of resistivity of transformed specimens. The results for a U-9 w/o Mo sample gamma quenched and subsequently annealed for 13 days at 525°C are also shown in Fig. 17. It can be seen that the temperature dependence of the transformed samples is that of a normal metal with a positive temperature coefficient.

The anomalous electrical behavior of these alloys is similar to that found in Ti-Nb alloys by Ames and McQuillan¹⁴ and by Brotzen, Harmon, and Troiano¹⁵ in the Ti-V system. The resistivity anomaly is most probably associated with electron scattering between the unfilled "d" band and the overlapping "s" band of the transition metal. From consideration of the $\alpha + \gamma/\gamma$ phase boundary, Pfeil and Browne² showed that uranium behaves as a quadravalent element in the U-Nb and U-Mo systems. Thus one would also expect these alloys to show anomalous resistivity effects inasmuch as uranium is a rare earth metal with two unfilled bands, the "5f" and "6d" bands, which are overlapped by the "7s" band and between which multiple electron scattering could occur.

Effect of Neutron Irradiation

Neutron bombardment can have a considerable effect on the properties of a metal. The severity of these effects is greatly increased when fissionable materials are irradiated because of the production of heavy, highly energetic fission fragments. A carefully controlled experiment was performed by Bleiberg, et al,¹³ to determine the effects of neutron irradiation on the metastable U-Mo γ phase and its decomposition products. Homogenized and gamma quenched specimens were prepared from natural U-9, 10.5, 12, and 13.5 w/o Mo. Alpha transformed specimens were obtained by annealing the gamma quenched material for 13 days at 525°C. Control specimens in the above conditions were held at reactor ambient temperatures for times equal to those of the irradiated samples.

Each specimen was examined by X-ray diffraction before and after annealing to be sure the desired transformations had occurred. Electrical resistivity measurements were made over the temperature range of -196°C to +150°C. Density, hardness measurements, and microstructure were obtained for each sample.

Measurements were taken with those specimens containing the γ' phase before disordering by mechanical working.

The results of the post irradiation examination showed that the stable room temperature phases had reverted to the metastable γ phase during irradiation. The results of Bleiberg, et al,¹³ are in agreement with the results reported by Konobeevsky, Pravdyuk, and Kutaitsev¹⁶ on U-9 w/o Mo alloy. The density changes upon transformation are very small, about 0.06%, so that it is difficult to determine conclusively from density measurements whether a phase change has occurred. However the resistivity changes upon irradiation are very pronounced because the negative temperature coefficient of resistivity of the metastable γ phase reverts to a normal positive coefficient when the alloy is isothermally transformed to its room temperature stable phases. Thus the magnitude and size of the coefficient is indicative of the progress of transformation. The effects of irradiation upon the resistivity of the U-9 w/o Mo alloy are shown in Fig. 17. After neutron irradiation the resistivity of the gamma-quenched sample showed a small decrease whereas that of the transformed samples showed a relatively large increase. In this alloy the resistance at -196°C decreased 1.2% in the gamma-quenched condition but increased 125% and 114% for two different neutron exposures for the transformed samples.

X-ray diffraction patterns were also obtained on the irradiated specimens. The results clearly showed that the $\alpha + \gamma'$ structure had been converted back to the high temperature, metastable γ phase. The X-ray line-broadening observed on this specimen indicated that the material had been homogenized on a microscopic scale to within ± 0.5 w/o of the nominal molybdenum concentration. The hardness results also indicated that transformation had occurred as a result of irradiation.

The mechanism for the phase reversal has been qualitatively explained by Bleiberg, et al, on the basis of the displacement spike mechanism described by Brinkman.¹⁷ This mechanism provides a model wherein atoms may be moved relatively large distances and localized high temperature areas are produced momentarily and subsequently rapidly quenched. As the α transformed material consists of thin lamella of α and γ' spaced about 5000 Å apart, neutron irradiation readily disorders the γ' phase, leaving a structure of $\alpha + \gamma$. The high temperatures and motion produced by the Brinkman displacement spike thus provide a mechanism for

re organization of the sample so that after exposure to irradiation the α phase dissolves and only the resultant metastable γ phase remains.

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TABLE I

MECHANICAL PROPERTIES OF U-12 w/o Mo ALLOYS TESTED AT 0.05 IN./MIN STRAIN RATE

Condition	Temp. °C	0.5% Yield (1000 psi)	Strength (1000 psi)	% Total Elongation	% Reduction in Area
γ_2 (200°C/24 hr/WQ)	25	166	166	9.0	41.9
γ_2 + 200°C - 2 days	25	156	157	10.4	32
γ_2 extruded	260	114	120	6.1	15.4
γ_2	260	102	104	6.8	50
γ_2 - 500°C - 4 hr	260	113	116	5.2	37
γ_2	316	106	107	3.8	23.3
γ_2 + 500°C - 2 days	316	99	100	4.0	32

TABLE II

IMPACT PROPERTIES OF A U-12 w/o Mo ALLOY

Test Temp. °C	Absorbed Energy ft-lb	Fracture
25	2.0	Brittle
66	2.6	Brittle
66	8.9	Ductile
71	14.5	Ductile

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FIGURE TITLES AND NEGATIVE NUMBERS

Fig. No.	Title	Neg. No.
1	The Uranium-Molybdenum Constitutional Diagram	19253-1
2	Detail of the Uranium-Molybdenum Constitutional Diagram	19253-2
3	$\alpha + \gamma'$ Phase Field of U-7.1 w/o Mo Isothermally Annealed 93 Days at 550°C (Bright Field 250X)	19598-7
4	$\alpha + \gamma'$ Phase Field of U-14.9 w/o Mo Isothermally Annealed 93 Days at 550°C (Bright Field 250X) (Polarized Light 250X)	19598-5 19598-2
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Fig. 1 The Uranium-Molybdenum Constitutional Diagram

24- 016

Fig. 2 Detail of the Uranium-Molybdenum Constitutional Diagram

441 - C17

(Bright Field 250X)

Fig. 3 $\alpha + \gamma'$ Phase Field of U-7.1 w/o Mo Isothermally
Annealed 93 Days at 550°C

64-018

(Bright Field 250X)

(Polarized Light 250X)

Fig. 4 $\alpha + \gamma'$ Phase Field of U-14.9 w/o Mo
Isothermally Annealed 93 Days at 550°C

441 C19

(Bright Field 250X)

Fig. 5 α' Structure Produced by Water Quenching
U-4 w/o Mo from 900°C

(Bright Field 250X)

Fig. 6 $\alpha + \gamma$ Phase Field of U-8 w/o Mo
Isothermally Annealed 59 Days at
570°C

441 020

(Polarized Light 250X)

Fig. 7 $\gamma + \gamma'$ Phase Field of U-12 w/o Mo
Isothermally Annealed 59 Days at
570°C

441 021

Fig. 8 Crystal Structure of γ and γ' Uranium-Molybdenum
(Solid circles are molybdenum; open circles are uranium)

(Bright Field 250X)

Fig. 9 γ' Phase Field of U-16.2 w/o Mo
Isothermally Annealed 59 Days at
550°C

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Fig. 10 Heating Curve for Cu-6 w/o Be and U-21.5 w/o Mo

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Fig. 11 Heating Curve for U-15.2 w/o Mo after Final Ordering Temperatures of 400°C, 500°C, and 550°C

Fig. 12 Complete Heating Curve for U-15.2 w/o Mo Ordered at 500°C

Fig. 13 Long Range Order Parameter vs. Final Ordering Temperature for U-15.2 w/o Mo

Fig. 14 Thermal Dilation of Natural Uranium and Uranium-Molybdenum Alloys after Annealing at 550°C for 16 Days

Fig. 15 Temperature Dependence of Tensile Strength of U-12 w/o Mo

Fig. 16 Relative Resistance vs. Temperature

Fig. 17 Resistance vs. Temperature for U-9 w/o Mo