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HIGH TEMPERATURE PROPERTIES OF URANIUM AND ITS ALLOYS

by

* T.J. Heal
A.B. McIntosh

Abstract

1. Data are presented on the mechanical and physical properties of high purity and normal uranium and for a range of dilute uranium alloys, for temperatures up to 700°C.
2. The results of thermal conductivity and thermal expansion determinations for a range of alloys are discussed; and some results on the latent heat of fusion and on the volume change on melting are presented.
3. The creep data on normal uranium can be interpreted in terms of the relation $\dot{\epsilon} = k \sigma^n$ where $\dot{\epsilon}$ is the strain rate and σ the applied stress. The best value of the parameter n , for the range of temperatures and stresses for which data are available, is arrived at and the data extrapolated to the low stresses obtained in service in the Calder Hall fuel elements. The relation between this extrapolated data and creep rates deduced from sagging bar and strut tests in the stress range of interest is discussed and reasons for the discrepancy advanced. The results of thermal cycling through the alpha/beta change point on a wide range of dilute alloys are examined in relation to current views on the mechanism of deformation and to tensile test data in the alpha and beta phases.
4. The results of creep tests on a range of dilute alloys are reported and are examined for correlation with the thermal cycling data and with available results on resistance to swelling under irradiation.

* U.K. Atomic Energy Authority, Industrial Group.

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HIGH TEMPERATURE PROPERTIES OF URANIUM AND ITS ALLOYS

A. B. McIntosh +
T. J. Heal +

1. INTRODUCTION

The present paper reports the results of experimental work largely confined to the measurement of properties relevant to the behaviour of uranium used as a fuel. Work is reported on the creep strength, the effect of thermal cycling, the tensile properties between 600° and 700°C, the thermal expansion and the thermal conductivity.

2. CREEP STRENGTH

In fuel elements vertically stacked as in the Calder reactors the fuel rod is loaded as a strut which will bow at a rate controlled by the effective creep strength of the fuel. Irradiation creep contributes to this effective strength but nevertheless out of pile creep data is of value in assessing the performance of present production uranium and in gauging the potential value of alloying additions.

2.1 CREEP OF MAGNESIUM-REDUCED URANIUM

The magnesium-reduced uranium studied has contained carbon and aluminium at levels of up to $1\frac{1}{2}$ and $\frac{1}{2}$ atomic per cent respectively in addition to significant quantities of iron and silicon. The creep behaviour of this low alloy material has been determined in the temperature range of interest on conventional high sensitivity tensile creep testing machines. At the minimum of the stress range used, secondary creep rates of the order of 10^{-6} per hour were obtained. Machines of greater sensitivity are under development to measure smaller creep rates. For uranium the minimum tensile stresses used

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to date are 0.1 tons per square inch at 500°C and 0.5 tons per square inch at 400°C. A typical operational stress is 0.1 tons per square inch. Bend test data on creep of uranium in the operational stress and temperature range have been obtained by the Springfields Laboratories⁽¹⁾ using end loaded strut and sagging bar tests. In order to correlate the two sets of data it is necessary to extrapolate the tensile creep data to lower stresses and to analyse the bend test data to obtain equivalent tensile creep rates.

Data on magnesium-reduced uranium of controlled grain size in the temperature range 400° to 600°C are given in Table 1. This data has been subjected to statistical analysis assuming that the secondary creep rate obeys a law of the form

$$\dot{\epsilon} = A e^{-Q/RT} \sigma^n \quad \dots \dots \dots (1)$$

where $\dot{\epsilon}$ is the strain rate, σ the applied stress in tons per square inch T is the temperature expressed in degrees Kelvin, Q is the activation energy for self diffusion, R the gas constant, and n and A are constants. The analysis first indicated that the data was self consistent in that the assumption that n is a constant independent of temperature and stress over the range considered is valid. Using the whole of the data the constants in equation (1) were determined to give

$$\dot{\epsilon} = 5.02 \times 10^6 \cdot e^{\frac{-40,100}{RT}} \sigma^{1.70} \quad \dots \dots \dots (2)$$

and using this equation in Figure 1 the 400°, 500°, and 600°C lines have been drawn in a plot of log $\dot{\epsilon}$ against log σ . The actual experimental points at 400°C, 500°C, and 600°C have been included and show the degree of correlation between equation (2) and the data from which it was derived.

TABLE 1

CREEP PROPERTIES OF MAGNESIUM REDUCED URANIUM

Stress ton/in ²	Temp. °C	Secondary creep rate $\times 10^{-6}$	Duration (hours)	Total strain(%)	Average grain size(mm)	*Heat treatment
5.0	400	7.05	1,440	1.56	-	2
2.0	400	0.856	2,100	0.24	-	2
2.0	400	0.83	2,030	0.26	-	1
1.0	400	0.33	2,380	0.11	-	2
1.0	400	0.46	1,490	0.11	0.16	1
2.0	500	85.7	75	1.00	-	2
2.0	500	64.0	255	1.89	0.15	1
1.0	500	21.5	1,170	2.55	0.17	2
1.0	500	21.0	960	2.14	-	1
0.75	500	9.1	1,170	1.21	-	1
0.5	500	5.6	1,290	0.76	0.19	2
0.50	500	6.6	1,340	0.95	0.15	1
0.30	500	2.1	830	0.37	-	1
0.2	500	2.66	1,300	0.47	-	2
0.20	500	0.625	960	0.07	-	1
0.20	500	2.50	1,500	0.45	0.13	1
0.5	600	100	260	3.30	-	2
0.2	600	28	970	2.95	0.16	2
0.20	600	18.6	1,100	2.13	-	1

- *1. Traverse β quenched α annealed
2. γ quenched, α annealed

The tensile creep data obtained by Springfields by analysis of their bend results is also included in Fig. 1, the creep rates obtained being approximately 3-5 times greater than the extrapolated tensile data.

Anderson(2) has obtained a factor of this order theoretically by considering the effect of continuous small thermal fluctuations on the creep rate of polycrystalline uranium at low stresses. For a given fluctuation he found the increase in creep rate increases with decreasing stress and the effect will therefore be of increased magnitude at the operational stress levels used in the bend tests, where additionally the degree of temperature control achievable is necessarily less precise than that in high sensitivity creep testing equipment. Experimental evidence in support of Anderson's theoretical analysis has been obtained and the results are given in Table 2, in which the secondary creep rates quoted are those after 1000 hours and the duration of each cycle is 4 hours. The main features of these results are

the increased effect of cycling as the magnitude of the cycle increases and more particularly as the stress decreases. The results are in good agreement with Anderson's theory and therefore lend confidence to predictions based on his theory for lower stresses. At 500°C with a thermal cycle of amplitude 2°C the theoretical increase in secondary creep rate is fivefold for an applied stress of 100 pounds per square inch, which, assuming such a temperature cycle on the bend tests, provides satisfactory correlation between the two sets of data.

TABLE 2

EFFECT OF SMALL THERMAL CYCLES ON THE CREEP RATE OF URANIUM -
COMPARISON OF THEORY WITH EXPERIMENT

Applied stress	10°C Cycles (500°-480°C)		5°C Cycles (500°-490°C)	
	$\dot{\epsilon}_i/\dot{\epsilon}_0^*$ experiment	$\dot{\epsilon}_i/\dot{\epsilon}_0^*$ theory	$\dot{\epsilon}_i/\dot{\epsilon}_0$ experiment	$\dot{\epsilon}_i/\dot{\epsilon}_0$ theory
0.20 t.s.i.	6.39	6.91		4.26
0.30 t.s.i.	4.66	5.21	2.65	3.20
0.75 t.s.i.	2.41	2.74	1.82	1.69

* $\dot{\epsilon}_i$ = creep rate with mean temperature 500°C and a cycling amplitude as indicated.

* $\dot{\epsilon}_0$ = creep rate with constant temperature 500°C

In addition to the increase in creep rate due to continuous small cycles consideration has been given to the influence of major temperature changes at less frequent intervals during test. Two specimens were creep tested at 400°C and 3 tons per square inch for approximately 1700 hours to establish the secondary creep rate. The specimens were then unloaded and the temperature allowed to fall to 100°C. The test were resumed at 3 tons per square inch at 400°C after a period of 8 hours. At intervals of 4 days this treatment was repeated with the acceleration in overall creep rate shown in Fig. 2. At approximately 2200 hours the period of the cycle on specimen KX1 was halved with a further resulting acceleration in overall creep rate.

2.2 EFFECT OF GRAIN SIZE ON CREEP RATE

The data from Table 1 used to derive Fig. 1 relates to material of reasonably uniform grain size. The effect of grain size on the properties of metals and alloys has been widely studied and much is known of the trend of properties with changing grain size. Rotherham and Tottle⁽³⁾ have pointed out that the treatments used to vary the grain size are capable of causing other structural changes which will affect the properties. The isolation of the grain size effect is therefore difficult to achieve unless high purity materials are used under carefully designed conditions to ensure the absence of structural differences, apart from that of grain size, in the preparation of the specimens. Cowan⁽⁴⁾ carried out such work on high purity solid solutions of magnesium in aluminium, but experienced difficulties due to lack of uniformity of solute concentration even in that case. With uranium the low solubilities in the α -phase of both impurities and deliberate alloying additions lead to the possibility of supersaturated solid solution or unstable precipitate structures, the stability of which depends on the heat treatments used to obtain the range of grain sizes. In some cases it would appear that time dependent changes of such metastable structures produce changes in creep strength large in relation to effects due to grain size. Thus in Fig. 3 where secondary creep rates at 400°C for a stress of 3 tons per square inch for magnesium and calcium-reduced uranium and for a number of alloys have been plotted against grain size, although the majority of the alloys show a common tendency for creep strength to increase with grain size the $\frac{1}{2}$ atomic per cent molybdenum alloys and the 1 atomic per cent aluminium alloys show a very marked departure from the general behaviour. Uranium worked at temperatures below 350°C shows twinning, indicating deformation principally within the grains, whereas above this temperature deformation is primarily at the grain boundaries. Thus at the lower temperatures a fine grain size should show enhanced creep strength, while at 400°C and above the reverse should be the case. The general tendency shown in Fig. 3 is thus in accordance with expectation. Details of the data on which this figure is based are given in Table 3, and some work on a 0.5 atomic per cent chromium alloy at 550°C and 1 ton per square inch is reported in Table 4. The relation between grain size and secondary creep rate for this alloy is plotted in Fig. 4. In order to achieve the recorded range of grain sizes and to attempt to limit other structural changes the specimens were taken from a single cast and given the α -phase isothermal heat treatment recorded in Table 4. The ratio of the increase in grain size to the increased creep strength is of the order of 4/3, which does not differ greatly from the general behaviour in Fig. 3 and suggests the effect of grain size does not vary greatly over the temperature range 400°C-550°C.

2.3 CREEP OF URANIUM ALLOYS

In addition to the results already given in Table 3 at 400°C and 3 tons per square inch, sorting tests have been carried out on a wide range of uranium alloys at 500°C and 2 tons per square inch. In these tests alloys containing the elements molybdenum, aluminium, vanadium, chromium, tantalum, titanium and thorium have shown the most promise of enhanced creep strength. Subsequently the creep resistance of alloys containing aluminium, molybdenum,

vanadium and chromium have been determined more precisely at 500°C and 2 tons per square inch (Fig. 5 and Table 5). Unless the effects of the considerable carbon and aluminium content of the base material, the limited solubility of the addition elements in α -uranium and the marked effect of grain size can be isolated and eliminated, high sensitivity creep results such as are given in Fig. 5 and Table 5 must present a complex picture. No relationship between the level of aluminium addition and the creep strength emerges, but the secondary creep rates of the molybdenum alloys do appear to be approximately related to the molybdenum content as Fig. 6 shows. Sorting tests on alloys containing vanadium additions up to 3 atomic per cent do not reveal a relation between the creep strength and the vanadium content.

TABLE 3

CREEP PROPERTIES OF URANIUM AND
URANIUM ALLOYS AT 400°C AND 3 TON/IN²

Alloy (at.%)	Grain size (mm) (measured after test)	Total creep strain (%)	Duration of test (hr)	Min. creep rate hr ⁻¹ x 10 ⁶
U - $\frac{1}{2}$ Mo.	Av. = 0.53 mm ($\rightarrow$ 1 mm)*	0.035	2190	0.05
U - $\frac{1}{2}$ Mo.	Fine	0.034	1312	0.23
U - $\frac{1}{2}$ Mo.	Av. = 0.44 ($\rightarrow$ 1)	0.035	1332	0.08
U - $\frac{1}{2}$ Mo.	Av. = 0.6	0.037	2623	0.015
U - $\frac{1}{2}$ Mo.	Av. = 0.59	0.049	914	0.05
U - $\frac{1}{2}$ Mo, - $\frac{1}{4}$ Fe.	Av. = 0.038	0.079	2641	0.08
U - $\frac{1}{2}$ Mo.	Av. = 0.72	0.059	1752	0.05
U - 1 Al.	Av. = 0.44 ($\rightarrow$ 1.0)	0.140	604	1.28
U - 0.9 Al.	Av. = 0.3	0.207	1097	1.13
U - $\frac{1}{2}$ Nb.	3-5 mm	0.027	1098	-
U - $\frac{1}{2}$ Nb.	Av. = 0.61	0.087	1036	0.12
U - $\frac{1}{2}$ Nb.	Av. = 0.34 ($\rightarrow$ 2)	0.090	521	0.47
U - $\frac{1}{2}$ V.		0.281	893	1.50
U - $\frac{1}{2}$ Zr.	Av. = 0.21 ($\rightarrow$ 0.5)	0.290	473	3.06
U - $\frac{1}{2}$ Fe.	Av. = 0.3 ($\rightarrow$ 3)	0.117	857	0.50
U - $\frac{1}{2}$ Fe.	Av. = 0.9	0.110	522	1.23
U - $\frac{1}{2}$ Cr.	Av. = 0.44	0.089	1098	0.15
U - $\frac{1}{2}$ Cr.	Av. = 0.15	0.233	1024	0.96
Mg, Red, U.	> 1 mm	0.220	1898	0.45
Ca, Red, U.	Av. = 1.4	0.082	642	0.66
Mg, Red, U.	Av. = 0.3	0.393	1710	1.70
Ca, Red, U.	Av. = 0.32	0.378	1637	1.70
Mg, Red, U.	Av. = 0.25	0.5058	1973	2.10
Mg, Red, U.	Av. = 0.30	0.423	1684	1.82
Ca, Red, U.	Av. = 0.32	0.378	1637	1.76
Mg, Red, U.	Av. = 0.30	0.393	1710	1.75
Ca, Red, U.	Av. = 0.35	0.324	1830	1.30
Mg, Red, U.	Av. = 0.50	0.335	1773	1.27

*Indicates maximum grain size measured.

TABLE 4

CREEP PROPERTIES OF A URANIUM 0.5 ATOMIC PER CENT
CHROMIUM ALLOY AT 550°C AND 1 TON/SQ.IN.

Heat treatment	Average grain size (mm)	Min. creep rate hr ⁻¹
740°C for 30 min to 450°C for 30 min W.Q.*	0.056	59 x 10 ⁻⁶
740°C for 30 min to 550°C for 30 min W.Q.	0.10	37.8 x 10 ⁻⁶
740°C for 30 min to 600°C for 30 min W.Q.	0.19	13 x 10 ⁻⁶
740°C for 30 min to 650°C for 30 min W.Q.	0.20	16.5 x 10 ⁻⁶

*W.Q. Water quenched.

TABLE 5

CREEP PROPERTIES OF SOME URANIUM ALLOYS AT 500°C AND 2 TON/SQ.IN.

ALLOYING ADDITION ATOMIC PER CENT	AVERAGE GRAIN SIZE mm	CHEMICAL ANALYSIS (ATOMIC PER CENT) WHERE AVAILABLE	TIME IN HOURS FOR SPECIFIED PERCENTAGE CREEP (TIME DEPENDANT) STRAIN						MINIMUM SECONDARY CREEP RATE HR ⁻¹ x 10 ⁶	TOTAL CREEP STRAIN PER CENT	TOTAL HOURS
			0.05	0.1	0.2	0.5	1.0	2.0			
$\frac{1}{8}$ Mo	0.28	Mo 0.1, C 2, Al 0.4, Fe 0.1	-	-	2	16	52	180	77	9.1	1000
$\frac{3}{8}$ Mo	0.21	Mo 0.35, C 2, Al 0.4, Fe 0.1	-	-	9	60	450	1700	6.3	1.99	1700
$\frac{1}{2}$ Mo	0.28	Mo 0.6, C 2, Al 0.4, Fe 0.1	4	12	24	250			3.5	0.94	1050
0.25 Mo	0.55		32	110	313	852	1540	2420	4.95	2.4	2600
0.25 Mo	0.38		6	16	39	122	240	426	38.5	2.6	521
0.5 Mo	0.52		13	49	168	558	1120	1810	6.8	2.3	1990
0.5 Mo	0.30		28	90	260	805	1550		4.3	1.6	2370
4.0 Mo	0.06		100	415					1.05	175	1145
	0.25				4	18	50	120	155	11.67	500
1.0 Al	0.09		6	18	36	110	224	418	46.8	2.95	550
2.0 Al	0.40		380	1050	1690				0.85	0.6	2321
4.0 Al	0.05			20	60	170	260	380	83	3.5	549
0.5 Cr			10	25	65	185	385	740	25	2.14	785
0.5 Cr	0.24		15	67	192	720			5.2	0.67	977
28.0 Mo	.05		22	80	320	1610			1.57	0.75	3197
2.0 V			25	100	300	800	1100		4.3	1.25	1314

3. EFFECTS DUE TO THERMAL CYCLING THROUGH THE α/β TRANSFORMATION

The dimensional stability of a fuel alloy on thermal cycling through the α/β transformation and the effects of cycling on the physical and mechanical properties are of direct interest to reactor applications where such cycling may occur. Both uranium and a range of alloys have been examined for dimensional and property changes.

3.1 EXPERIMENTAL METHOD

Dimensional, density and surface distortion measurements were made on each specimen, together with chemical analysis and grain size determinations. Measurements were normally made before thermal cycling and repeated after each successive period of 25 cycles. Thermal cycling was discontinued after a hundred cycles except in those cases where little distortion had been produced. All thermal cycling was carried out at a pressure not exceeding 10^{-4} millimetres of mercury. The maximum temperature gradient along a specimen was normally less than 10°C rising to a maximum of 30°C at some stages of a cycle. A thermocouple attached to the surface of each specimen indicated its temperature as well as the arrest point when the alloy passed through the phase change. The temperature range for the cycle was adjusted to give an upper limit 25°C above the heating transition temperature and a lower limit 25°C below the cooling transition temperature. Heating and cooling rates were balanced so as to remain between 3°C and 5°C per minute.

3.2 EXPERIMENTAL RESULTS

The alloys with few exceptions showed an increase in diameter but length variations were more erratic⁽⁵⁾. Results from the duplicate specimens used show agreement except in a few cases where it is believed that some constitutional difference existed in the material. The means of these duplicate results of percentage increase in diameter plotted against the number of cycles give the typical results shown in Fig. 7 and 8. From the extrapolation of these curves to their intersection with the abscissa it is deduced that there is a minimum number of cycles for each alloy below which no distortion can be detected and above which progressive distortion occurs at a linear rate. These minima, together with the subsequent rates of distortion per cycle, are given in Table 6, which also records changes in grain size.

Results of the density measurements showed only negligible changes of up to 0.3%, scarcely different from the limit of measurement.

Surface distortion was determined as the difference between the maximum and minimum deviation from the mean diameter along the specimen, measured at 25 cycle intervals. In Table 7 the alloys are classified with respect to the extent of surface distortion produced by 100 thermal cycles.

3.3 EFFECT OF CONCENTRATION OF ALLOYING ELEMENT

The series of uranium-aluminium alloys examined contained respectively 0.6, 2.1 and 4.0 at. % aluminium and thus allowed some comparison of the effect of increasing amounts of this alloying element in uranium. The increased concentration is shown metallographically to result in an increase in the amount of precipitation. Each of the alloys was cycled in the beta-quenched condition. On thermal cycling, in addition to a slight variation in the minimum number of cycles for dimensional distortion the rate also varies inversely with the composition (Fig. 9). The general inference would therefore appear to be that the precipitation of UAl_2 imparts dimensional stability on thermal cycling. A distinct difference, Fig. 10, in the behaviour of duplicate specimens of the uranium - 2 at. % vanadium alloy during thermal cycling is of interest in this connection. Subsequent examination of the microstructures revealed that one specimen, removed from the upper portion of the cast bar, contained a coarse acicular second phase, believed to be a nitride absent from the other specimen. There seems little doubt that the coarse acicular precipitate present in one specimen enhanced the dimensional and surface stability of the alloy during thermal cycling. Fig. 11 (a) and (b) show the specimen after 100 and 200 cycles respectively. In each case the high nitrogen specimen is the left hand number of the pair.

Although the casting in graphite of the alloys used has resulted in a variable carbon content which is of alloying proportions when computed on an atomic per cent basis, some evidence has been obtained that increased quantities of precipitate due to the specific alloying addition impart enhanced resistance to thermal cycling. (Table 6) The explanation could be the simple one that there is an increased volume of material on which the thermal cycle has no effect.

TABLE 6

EFFECT OF THERMAL CYCLING THROUGH THE α/β PHASE CHANGE ON URANIUM
AND URANIUM ALLOYS

SPECIMEN IDENTIFICATION	CHEMICAL ANALYSIS ATOMIC %		INITIAL GRAIN SIZE (mm)		MINIMUM NO. OF CYCLES FOR DISTORTION	DISTORTION RATE % PER CYCLE	FINAL GRAIN SIZE (mm)	
	ALLOY	CARBON	AV.	MAX.			AV.	MAX.
Ca - Reduced		~0.9	-	-	10	0.024	1.5	4.0
Mg - Reduced		~2.2	-	-	21	0.063	1.5	4.0
$\frac{1}{2}$ Ge	Ge 0.49	1.9	0.14	0.3	20	0.070	1.7	4.0
$\frac{1}{2}$ Al	Al 0.6	2.6	0.10	-	19	0.050	1.0	3.5
$\frac{1}{2}$ Al	Al 0.6	2.6	-	-	16	0.037	0.95	2.0
$\frac{1}{2}$ Al	Al 0.6	1.8	0.64	1.3	10	0.066	0.4	2.0
2 Al	Al 2.09	4.7	0.10	0.5	9	0.058	1.5	3.0
4 Al	Al 4.01	2.8	0.07	2.0	17	0.046	1.8	4.0
$\frac{1}{2}$ Fe	Fe 0.6	2.2	0.14	0.5	11	0.053	1.05	
$\frac{1}{2}$ Fe	Fe 0.6	2.2	-	-	8	0.030	0.34	2.1
2 Fe	Fe 2.11	1.5	0.23	0.7	10	0.060	0.69	4.5
1 Ag	Ag 0.25	2.1	0.67	3.0	0	0.047	1.5	3.0
2 V	V 2.08	1.8	-	-	58	0.034	0.94	2.0
$\frac{1}{2}$ Nb	Nb 0.58	1.7	0.25	2.5	25	0.018	1.4	5.0
4 Mo	Mo 4.20	1.9	0.02		0	0.009	0.03	0.1
$\frac{1}{2}$ Mo	Mo 0.47	2.5	0.27	0.5	0	0.002	0.51	1.3
$\frac{1}{2}$ Cr	Cr 0.55	1.7	0.05		0	-0.004	0.34	2.0
$\frac{1}{2}$ Pt	Pt 0.49	1.4	0.10	0.3	0	-0.0045		
$\frac{1}{4}$ Ru	Ru 0.27	1.3	0.20	0.5	34	0.023	0.41	1.0
$\frac{1}{2}$ Si	Si 0.52		0.19	0.4			0.44	1.5
1 Sn	Sn 0.98	1.9	0.41	1.3	0	0.022	0.32	1.3
$\frac{1}{2}$ Ta	Ta 0.55	1.3	0.42	1.5	11	0.033	0.39	1.1
1 Th	Th 0.82	6.3	0.89	3.5	16	0.054	0.42	1.5
$\frac{1}{4}$ Ti	Ti 0.24	1.5	0.58	3	0	0.030	0.84	2.0
$\frac{1}{2}$ Ti	Ti 0.48	1.3	0.40	1.7	20	0.102	0.17	2.0
$\frac{1}{2}$ V	V 0.55	1.4	0.41	1.7	13	0.110	0.85	2.7
2 Ge	Ge 2.0	1.3	0.59	3	12	0.066	1.5	5
3 Pd	Pd 3.04	1.9	0.12	0.3	2	0.098	0.80	2.6
1 Pd	Pd 0.98	1.1	0.15	0.4	2	0.081	0.82	2.5

TABLE 7

SURFACE DISTORTION OF URANIUM AND URANIUM ALLOYS AFTER
100 THERMAL CYCLES THROUGH THE α/β PHASE CHANGE

SURFACE DISTORTION Difference between max. and min. diameters $\times 10^3$ ins.				
GOOD < 25	REASONABLE 25 - 50	MODERATE 50 - 75	BAD 75 - 100	VERY BAD > 100
4 Mo $\frac{1}{2}$ Cr $\frac{1}{2}$ Mo 2 V $\frac{1}{2}$ Pt $\frac{1}{4}$ Ru	$\frac{1}{2}$ Fe 2 Fe $\frac{1}{2}$ Ta 1 Th $\frac{1}{4}$ Ti 1 Sn 1 Pd	4 Al $\frac{1}{2}$ Fe 2 Al $\frac{1}{2}$ Ti 2 Ge $\frac{1}{2}$ V1	$\frac{1}{2}$ Al $\frac{1}{2}$ Ge Ca - U $\frac{1}{2}$ Si 3 Pd	$\frac{1}{2}$ Al 1 Ag $\frac{1}{2}$ Al $\frac{1}{2}$ Nb Mg - U

3.4 THEORIES OF PHYSICAL DAMAGE DUE TO CYCLING PHASE TRANSFORMATIONS

Buckley, Harding and Waldron⁽⁶⁾ have recently surveyed the previous theories advanced to explain the pattern of deformation of uranium after repeated passage through the alpha-beta phase transformation and have advanced their own theory, which, although it involves extensive simplifying assumptions to permit mathematical analysis, is a considerable advance on previous theories. They assume that the plastic flow occurring during a single α - β - α -cycle depends on three factors:-

- (1) The transformation volume change
- (2) The relative strength of the two phases
- (3) The direction and shape of the phase interface

The particular case of a plane phase interface may be evaluated and the results applied qualitatively to more complex phase fronts. In order to examine the validity of the above theory the relative volumes of the alpha and beta phases and their relative strengths are required.

3.5 VOLUME CHANGES DURING THE α/β TRANSFORMATION

Data on the percentage dilation during the α/β transformation for calcium and magnesium reduced uranium and for some chromium and molybdenum alloys of

uranium are given in Table 8. The variation is relatively small and is insufficient to explain the marked difference in thermal cycling behaviour of the materials listed.

3.6 RELATIVE TENSILE STRENGTHS OF THE α AND β PHASES

Data on the relative tensile strengths of the α and β phases has been obtained by Taylor⁽⁷⁾ in a temperature range on either side of the α/β transformation, for four alloys chosen for their widely differing resistance to thermal cycling, Fig. 12. The compositions of the alloys are given in Table 9. The tests were carried out in vacuo in two Hounsfield Tensometers at strain rates of either 4.2 inches per hour or 3.2 inches per hour (chromium alloys only). Each specimen was annealed for one hour inside the tensometer furnace in a single phase region (β or γ) before slow cooling ($1\frac{1}{2}$ - $3^\circ\text{C}/\text{min}$) to the test temperature at which it was annealed for one hour prior to test to ensure that the equilibrium structure was obtained. For specimens other than the 4 atomic per cent molybdenum alloy the annealing was carried out at 700°C . For this alloy the annealing temperature was raised to 740°C to avoid the β/γ transformation which occurs at 705°C - 710°C ⁽⁸⁾. In all cases the Ultimate Tensile Strength, Limit of Proportionality, Breaking Stress, percentage Elongation and percentage Reduction of Area were determined. The magnesium reduced uranium and the molybdenum alloy were tested as cast, and the vanadium alloy as homogenised. Though it was not considered likely that the initial grain structure of the material would greatly affect the results, this was verified using the chromium alloy in both the as-cast and the isothermally transformed conditions with grain sizes of 0.5 and 0.05 mm respectively.

TABLE 8

VOLUME CHANGE DURING THE α - β PHASE TRANSFORMATION IN
URANIUM AND SOME URANIUM ALLOYS

MATERIAL	DILATION α - β 10^{-4} in.
CALCIUM-REDUCED URANIUM MAGNESIUM-REDUCED URANIUM	) 34
0.6 at.% CHROMIUM-URANIUM 1 at.% CHROMIUM-URANIUM	27 35
0.5 at.% MOLYBDENUM-URANIUM 4.2 at.% MOLYBDENUM-URANIUM	34 27

TABLE 9

ANALYSIS OF URANIUM ALLOYS USED TO INVESTIGATE THE
RELATIVE TENSILE PROPERTIES OF THE α AND β PHASES

MATERIAL	C	Si	Mo	V	Cr
Mg-reduced Uranium	1,400*	30	-	-	-
U-4 at. % Mo	2,200	25	4.07 at. %	-	-
U-2 at. % V	1,900	27	-	2.04 at. %	-
U- $\frac{1}{2}$ at. % Cr	900	31	-	-	0.49 at. %

*Values quoted are parts per million (p.p.m.) unless otherwise stated.

The results obtained, which are given in Tables 10-14 and Figs. 13-17, show notable differences in the temperature dependence of the tensile properties of the four alloys examined. The magnesium reduced uranium had low relative strength in the α phase. By extrapolation to the change point in Fig. 13 the following figures were obtained:-

	<u>U.T.S.</u>	<u>L. of P.</u>	<u>% E.</u>
α (667°C)	2.3	1.6	58
β (667°C)	7.4	5.9	2 $\frac{1}{2}$

For the vanadium alloy where a narrow two phase region exists Fig. 14 gives

α (652°C)	4.5	8.4	72
β (660°C)	9.0	7.4	11

For the molybdenum alloy, Fig. 15, the properties are essentially unchanged at the phase transformation but the two phase $\alpha + \gamma$ structure is considerably stronger than the $\beta + \gamma$ structure when assessed over a temperature range. Finally the chromium alloys show a two phase region of enhanced strength (Figs. 16 and 17) separating α and β regions of roughly equivalent strength. The differences between the chromium alloys are marginal, which confirms the assumption that the initial structural differences would not persist.

TABLE 10

HIGH TEMPERATURE TENSILE PROPERTIES OF MAGNESIUM REDUCED URANIUM

Test Temp. °C	Phase(s) present	U.T.S. (t.s.i.)	L of P (t.s.i.)	Breaking stress (t.s.i.)	% E	% R.A.
700	β	6.6	5.5	6.6	4	1
690	β	7.0	5.7	7.0	3	1
680	β	7.2	5.8	7.2	3	1
660	α	2.5	1.7	0.35	55	*
640	α	2.9	2.0	0.35	48	*
620	α	3.4	2.6	0.35	42-46	*

*Specimens exhibited a knife edged fracture with highly irregular contours, making it impossible to obtain a reliable figure for reduction in area.

TABLE 11

HIGH TEMPERATURE TENSILE PROPERTIES OF URANIUM 2 AT.% VANADIUM

Test Temp. °C	Phase(s) present	U.T.S.	Limit of proportion- ality	Breaking stress	% E	% R.A.†
700	β	8.2	6.4	1.2	32%	55-25
685	β	8.3	6.4	2.4	24%	70-20
670	β	8.7	7.0	1.8	16%	38-25
655	$\beta + \alpha$	12.8	10.2	1.6	23%	60-43
655*	$\beta + \alpha$	11.1	8.9	-	22-28%	65-22
640	$\alpha + \delta$	4.6	3.4	0.2	62%	76
620	$\alpha + \delta$	4.9	3.6	0.75	30%	66-38
600	$\alpha + \delta$	5.4	4.2	3.0	35%	66-30

*This specimen exhibited double necking.

†Nearly all specimens were heavily distorted with an elliptical cross-section at the point of fracture. Two reduction in area figures, measured at right angles, are quoted.

TABLE 12

HIGH TEMPERATURE TENSILE PROPERTIES OF URANIUM 4 AT.% MOLYBDENUM

Test Temp. °C	Phase(s) present	% γ	U.T.S.	L of P	Breaking stress	% E	% R.A.
700	$\beta + \gamma$	71	1.8	1.3	-	225*	97*
685	$\beta + \gamma$	58	2.2	1.4	0.15	133	95
670	$\beta + \gamma$	48	3.8	2.6	0.2	167	93
650	$\alpha + \gamma$	41	7.2	5.7	0.5	25-29	55
630	$\alpha + \gamma$	31	9.0	6.5	6.1	18	25
610	$\alpha + \gamma$	25	12.1	9.4	10.4	10	16

*Specimen unbroken; values quoted are those at which deformation was halted.

TABLE 13

HIGH TEMPERATURE TENSILE PROPERTIES OF URANIUM $\frac{1}{2}$ AT.% CHROMIUM (AS CAST)

Test Temp. °C	Phase(s) present	U.T.S.	L of P	Breaking stress	% E	% R.A.
700	β	5.2	4.1	2.2	15	20
680	β	7.8	6.2	7.7	5	6
670	β	7.7	5.8	7.4	8	7
660	-	6.1	4.7	2.2	20	62
650	-	6.4	5.2	2.7	18	66
642	-	5.5	4.8	0.9	20	62
640	-	8.3	7.4	4.1	16	58
640	-	6.4	5.2	2.9	19	63
637*	-	6.8	6.1	4.6	14	42
620	α	6.6	5.4	3.2	24	50
600	α	8.4	7.6	4.7	16	35

*Specimen broke at large hole remote from neck.

TABLE 14

HIGH TEMPERATURE TENSILE PROPERTIES OF URANIUM $\frac{1}{2}$ AT.% CHROMIUM
(HEAT TREATED)

Test Temp. °C	Phase(s) present	U.T.S.	L of P	Breaking stress	% E	% R.A.
700	β	6.1	4.7	2.7	19	35
694	β	6.4	5.2	5.0	8	10
680	β	6.4	4.9	5.6	5	6
660	-	8.3	6.6	6.9	9	18
660	-	5.8	4.3	4.3	10	40
655	-	9.8	7.8	8.9	7	11
650	-	8.9	6.6	7.6	10	13
645	-	6.5	5.0	3.0	14	61
640	-	6.4	5.2	2.7	19	68
620	α	6.2	5.1	2.5	18	57
600	α	7.2	6.2	4.0	20	49

A feature of the results obtained for magnesium reduced uranium was the low ductilities and the brittle fractures of specimens broken in the high strength beta phase and the higher ductilities and irregular deformations of specimens tested in the low strength alpha phase. The evidence suggested transcrystalline fracture in the alpha phase and intergranular fracture in the beta phase. The alpha phase might therefore be expected to be more susceptible to solid solution strengthening by addition elements. The evidence in the case of the 2 at. % vanadium and $\frac{1}{2}$ at. % chromium alloys supports this, the alpha phase being strengthened more than the beta phase, particularly in the case of the chromium alloy, where there is little difference in strength between the two phases. Thus for natural uranium and these two alloys the thermal cycling resistance (Fig. 12) can be roughly correlated with the net difference in strength between the alpha and beta phases, in good agreement with the hypothesis of Buckley, Harding and Waldron⁽⁶⁾. There is no single phase region in the temperature range studied for the 4 at. % molybdenum alloy: at 658°C the two phase beta plus gamma region becomes as a result of reaction alpha plus gamma. No abrupt change in tensile properties occurs at 658°C, which suggests that the continuous presence of the gamma phase (Table 12) exerts a large influence on the properties. It also serves to reduce the effective volume change on transformation. On both these counts the very high degree of thermal cycling resistance, (Fig. 12) would be predicted by Buckley et al.

A further notable feature of the results is the sharp increase in strength in the alpha + beta phase region for both the vanadium and to a lesser degree the chromium alloys. As the alpha + beta regions exist only over a narrow temperature range, the relative proportions of alpha and beta must vary rapidly with temperature. It seems clear that the strength of one phase (alpha or beta) is considerably affected by the presence and dispersion of a proportion of the other and the deformation leading to fracture in one phase is constrained by the presence of the other phase.

4. PHYSICAL PROPERTIES

4.1 LINEAR THERMAL EXPANSION OF URANIUM

The thermal expansion of a polycrystalline bar of anisotropic material such as alpha uranium will be dependent on the degree of preferred orientation present and the relation between the pattern of preferred orientation and the direction of measurement. With material of substantially random orientation such as the beta-quenched alpha-annealed material used in the U.K.A.E.A. Calder type reactors there is a meaningful mean coefficient of linear thermal expansion along the axes of the uranium bars. However, because of the slight permanent change in the length of the bars on thermal cycling residual differences in length of about $\pm 0.1\%$ may occur between successive expansion determinations. Changes in the mean expansion coefficient of about 1.5×10^{-6} also occur between determinations.

As has been pointed out earlier magnesium-reduced uranium is essentially an alloy material and as a result the properties are sensitive to thermal

history. The values given in Table 15 are considered to be typical and the linear coefficient of thermal expansion can be represented by the equation

$$L_t = L_0 (1 + \alpha t + \beta t^2)$$

$$\text{where } \alpha = 14.8 \times 10^{-6}$$

$$\text{and } \beta = 5.5 \times 10^{-9}$$

and t is expressed in degrees Centigrade.

4.2 LINEAR THERMAL EXPANSION OF URANIUM ALLOYS

The remarks made above in 3.1 relate to the expansion behaviour of all alloys when in the alpha phase. Fig. 18 shows the variation of the expansion coefficient with temperature for magnesium reduced uranium and a number of alloys. Tables 16 - 22 give the values of the mean coefficient over the range 0° to $t^\circ\text{C}$, α_0 , the mean coefficient over the range 18° to $t^\circ\text{C}$, α_{18} , and the coefficient at $t^\circ\text{C}$, α_t . In all cases the expansion coefficients are lower than those of magnesium reduced uranium.

4.3 THERMAL CONDUCTIVITY OF URANIUM

The best straight line for 6 determinations of the thermal conductivity of magnesium reduced uranium in the range $117^\circ - 574^\circ\text{C}$ is given by

$$K (\text{cals/cm.sec.}^\circ\text{C}) = 0.067 + 2.1 \times 10^{-5}t$$

values obtained from this equation are given in Table 23 and Fig. 19 shows the results (A) in comparison with what is considered the best available data on U.K.A.E.A. calcium reduced uranium (B) and with American data as summarised by Smith(9), who gives an upper (C) and lower limit (D) for the scatter of American values and selects a composite curve which in his view represents the best selection of the American data.

4.4 THERMAL CONDUCTIVITY OF URANIUM ALLOYS

Fig. 20 and Table 23 give the results of determinations carried out on alloys containing the following additions - 4 Al, 2 Al, 4 Mo, $\frac{1}{2}$ Mo, 2 V and $\frac{1}{2}$ Cr, all quantities being given in atomic per cent. For a number of the alloys a higher thermal conductivity has been found at the higher temperatures than that of the base material, magnesium reduced uranium. No explanation has been found for this anomaly.

5. CONCLUSIONS

Certain features of the wide range of data presented in this paper merit special comment.

1. The satisfactory agreement which exists (Fig. 1) between the extensive creep data on controlled grain size magnesium-reduced uranium and the equation derived from that data holds over a range of creep rates from 10^{-4} to 10^{-6} , over a stress range of 0.1 to 5.0 tons per square inch and over a temperature range from 400° to 600°C .
2. The predictions of Anderson's theory of the effect of continuous small thermal cycles on the creep rate of polycrystalline uranium agree well with the actual experimental values. The implications of this theory are considerable. It provides an explanation of the discrepancy between creep rates at low stresses obtained by extrapolation of the high sensitivity tensile creep data and creep rates deduced from low stress bend tests. It makes clear the need for close temperature control in all studies of deformation at low stresses and suggests that a significant contribution to the scatter observed amongst measurements of the swelling of uranium under irradiation⁽¹⁰⁾, may arise from variations in the degree of temperature control.
3. The high temperature tensile properties of the alloys chosen for examination show the extent of permanent distortion on repeated α/β cycling to be related to the relative strengths of the α and β phases.
4. Other noteworthy features of the high temperature tensile data are the high relative strength of the β -phase in magnesium-reduced uranium, the marked increase in strength of the α -phase for the alloying additions investigated and the strength maxima occurring in two phase $\alpha + \beta$ temperature ranges. The occurrence of these maxima indicates that fuels which can be maintained in the two phase $\alpha + \beta$ condition will possess a high resistance to deformation at tensile rates.

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TABLE 15

LINEAR THERMAL EXPANSION OF MAGNESIUM REDUCED URANIUM

Temp. (°C)	$\alpha_0 \times 10^6$	$\alpha_1 \times 10^6$	$\alpha_{18} \times 10^6$
100	15.3 ₄	15.8 ₈	15.4 ₄
200	15.8 ₈	16.9 ₈	15.9 ₈
300	16.4 ₃	18.0 ₇	16.5 ₃
400	16.9 ₈	19.1 ₆	17.0 ₈
500	17.5 ₂	20.2 ₆	17.6 ₂
600	18.0 ₇	21.3 ₅	18.1 ₇

Temperature range 19° to 619°C

TABLE 16

LINEAR THERMAL EXPANSION OF U-4 ATOMIC PER CENT Al

Temp. (°C)	$\alpha_0 \times 10^6$	$\alpha_1 \times 10^6$	$\alpha_{18} \times 10^6$
100	14.1 ₆	14.5 ₆	14.2 ₃
200	14.5 ₆	15.3 ₇	14.6 ₃
300	14.9 ₆	16.1 ₇	15.0 ₄
400	15.3 ₇	16.9 ₇	15.4 ₄
500	15.7 ₇	17.7 ₈	15.8 ₄

Range of temperatures:- 23° to 500°C.

General equation

$$L_t = L_0 (1 + \alpha t + \beta t^2)$$

$$\alpha = 13.7_6 \times 10^{-6}, \beta = 4.0 \times 10^{-9}$$

TABLE 17

LINEAR THERMAL EXPANSION OF U-4 ATOMIC PER CENT Mo

Temp. (°C)	$\alpha_0 \times 10^6$	$\alpha_1 \times 10^6$	$\alpha_{18} \times 10^6$
100	13.7 ₇	14.5 ₀	13.9 ₀
200	14.5 ₀	15.9 ₆	14.6 ₃
300	15.2 ₃	17.4 ₂	15.3 ₆
400	15.9 ₆	18.8 ₈	16.0 ₉
500	16.6 ₉	20.3 ₄	16.8 ₂

Range of temperatures:- 20° to 504°C.

General equation:-

$$L_t = L_0 (1 + \alpha t + \beta t^2)$$

$$\alpha = 13.0_3 \times 10^{-6}, \beta = 7.3 \times 10^{-9}$$

TABLE 18

LINEAR THERMAL EXPANSION OF U-2 ATOMIC PER CENT V

Temp. (°C)	$\alpha_0 \times 10^6$	$\alpha_1 \times 10^6$	$\alpha_{18} \times 10^6$
100	13.4 ₄	14.3 ₂	13.6 ₀
200	14.3 ₂	16.0 ₈	14.4 ₈
300	15.2 ₀	17.8 ₄	15.3 ₆
400	16.0 ₈	19.6 ₀	16.2 ₄
500	16.9 ₆	21.3 ₆	17.1 ₂
600	17.8 ₄	23.1 ₂	18.0 ₀

Range of temperatures:- 20° to 611°C.

General equation:-

$$L_t = L_0 (1 + \alpha t + \beta t^2)$$

$$\alpha = 12.5_6 \times 10^{-6}, \beta = 8.8 \times 10^{-9}$$

TABLE 19

LINEAR THERMAL EXPANSION OF U - 2 ATOMIC PERCENT Al

Temp. (°C)	$\alpha_0 \times 10^6$	$\alpha_1 \times 10^6$	$\alpha_{18} \times 10^6$
100	14.7 ₈	15.4 ₆	14.9 ₀
200	15.4 ₆	16.8 ₁	15.5 ₈
300	16.1 ₃	18.1 ₆	16.2 ₅
400	16.8 ₁	19.5 ₁	16.9 ₃
500	17.4 ₈	20.8 ₆	17.6 ₀
600	18.1 ₆	22.2 ₁	18.2 ₈

Range of temperature:- 19° to 603°C

General equation:-

$$L_t = L_0 (1 + \alpha t + \beta t^2)$$

$$\alpha = 14.1_1 \times 10^{-6}, \beta = 6.7 \times 10^{-9}$$

TABLE 20

LINEAR THERMAL EXPANSION OF U - 1 ATOMIC PERCENT Al

Temp. (°C)	$\alpha_0 \times 10^6$	$\alpha_1 \times 10^6$	$\alpha_{18} \times 10^6$
100	14.0 ₀	14.5 ₈	14.1 ₀
200	14.5 ₈	15.7 ₄	14.6 ₈
300	15.1 ₆	16.9 ₀	15.2 ₆
400	15.7 ₄	18.0 ₅	15.8 ₄
500	16.3 ₂	19.2 ₁	16.4 ₂
600	16.9 ₀	20.3 ₇	17.0 ₀

Range of temperature:- 36° to 577°C

General equation:-

$$L_t = L_0 (1 + \alpha t + \beta t^2)$$

$$\alpha = 13.4_2 \times 10^{-6}, \beta = 5.8 \times 10^{-9}$$

TABLE 21

LINEAR THERMAL EXPANSION OF U - $\frac{1}{2}$ ATOMIC PERCENT Mo

Temp. (°C)	$\alpha_0 \times 10^6$	$\alpha_1 \times 10^6$	$\alpha_{18} \times 10^6$
100	13.4 ₂	14.3 ₇	13.6 ₀
200	14.3 ₇	16.2 ₇	14.5 ₅
300	15.3 ₂	18.1 ₇	15.5 ₀
400	16.2 ₇	20.0 ₇	16.4 ₄
500	17.2 ₂	21.9 ₇	17.3 ₉
600	18.1 ₇	23.8 ₇	18.3 ₄

Range of temperature:- 18° x 603°C

General equation:-

$$L_t = L_0 (1 + \alpha t + \beta t^2)$$

$$\alpha = 12.47 \times 10^{-6}, \beta = 9.5 \times 10^{-9}$$

TABLE 22

LINEAR THERMAL EXPANSION OF U - $\frac{1}{2}$ ATOMIC PERCENT Cr

Temp. (°C)	$\alpha_0 \times 10^6$	$\alpha_1 \times 10^6$	$\alpha_{18} \times 10^6$
100	12.9 ₇	14.0 ₁	13.1 ₆
200	14.0 ₁	16.0 ₇	14.1 ₉
300	15.0 ₄	18.1 ₄	15.2 ₃
400	16.0 ₇	20.2 ₀	16.2 ₆
500	17.1 ₁	22.2 ₇	17.2 ₉

Range of temperature:- 15° to 580°C

General equation:-

$$L_t = L_0 (1 + \alpha t + \beta t^2)$$

$$\alpha = 11.9_4 \times 10^{-6}, \beta = 10.3 \times 10^{-9}$$

TABLE 23

THERMAL CONDUCTIVITY OF URANIUM AND URANIUM ALLOYS

Thermal Conductivity K (cals/cm. sec. °C)						
Temp. Alloy	100	200	300	400	500	550
Mg. reduced U	0.069 ₂	0.071 ₃	0.073 ₄	0.075 ₅	0.077 ₆	0.078 ₇
U-4 ^a / _o Al	-	0.061 ₁	0.069 ₂	0.077 ₂	0.085 ₂	-
U-4 ^a / _o Mo	-	0.055 ₅	0.065 ₂	0.074 ₉	0.084 ₆	-
U-2 ^a / _o V	0.061 ₉	0.066 ₁	0.070 ₂	0.074 ₄	0.078 ₅	0.080 ₆
U-2 ^a / _o Al	-	0.060 ₀	0.064 ₅	0.069 ₀	0.073 ₅	0.075 ₈
U-1 ^a / ₂ Mo	-	0.064 ₄	0.071 ₅	0.078 ₅	0.085 ₅	0.089 ₁
U-1 ^a / ₂ Cr	0.064 ₆	0.069 ₃	0.074 ₀	0.078 ₆	0.083 ₃	0.085 ₆

LIST OF TABLES

<u>Table Number</u>	<u>Title</u>
1	Creep properties of magnesium reduced uranium.
2	Effect of small thermal cycles on the creep rate of uranium - comparison of theory with experiment.
3	Creep properties of uranium and uranium alloys at 400°C and 3 ton/sq.in.
4	Creep properties of a uranium $\frac{1}{2}$ at. % chromium alloy at 550°C and 1 ton/sq.in.
5	Creep properties of some uranium alloys at 500°C and 2 ton/sq.in.
6	Effect of thermal cycling through the α/β phase change on uranium and uranium alloys.
7	Surface distortion of uranium and uranium alloys after 100 thermal cycles through the α/β phase change.
8	Volume change during the α/β phase transformation in uranium and some uranium alloys.
9	Analyses of uranium alloys used to investigate the relative tensile properties of the α and β phases.
10	High temperature tensile properties of magnesium-reduced uranium.
11	High temperature tensile properties of uranium 2 atomic per cent vanadium.
12	High temperature tensile properties of uranium 4 atomic per cent molybdenum.
13	High temperature tensile properties of uranium $\frac{1}{2}$ atomic per cent chromium (as cast).
14	High temperature tensile properties of uranium $\frac{1}{2}$ atomic per cent chromium (heat treated).
15	Linear thermal expansion of magnesium reduced uranium.
16	Linear thermal expansion of uranium 4 atomic per cent aluminium.
17	Linear thermal expansion of uranium 4 atomic per cent molybdenum.

<u>Table Number</u>	<u>Title</u>
18	Linear thermal expansion of uranium 2 atomic per cent vanadium.
19	Linear thermal expansion of uranium 2 atomic per cent aluminium.
20	Linear thermal expansion of uranium 1 atomic per cent aluminium.
21	Linear thermal expansion of uranium $\frac{1}{2}$ atomic per cent molybdenum.
22	Linear thermal expansion of uranium $\frac{1}{2}$ atomic per cent chromium.
23	Thermal conductivity of uranium and uranium alloys.

LIST OF FIGURES

<u>Figure Number</u>	<u>Caption</u>
1	Tensile Creep data for uranium.
2	The effect of large thermal and load cycling on the creep behaviour of uranium.
3	Creep rate against grain size for uranium and uranium alloys at 400°C and 3 tons per in ² .
4	Variation of secondary creep rate with grain size for a uranium $\frac{1}{2}$ atomic per cent chromium alloy.
5	Creep strength of uranium alloys at 500°C and 2 tons per in ² .
6	Effect of alloying on the creep strength of uranium molybdenum alloys.
7	Distortion of uranium and uranium alloys during α/β thermal cycling.
8	Distortion of uranium alloys during α/β thermal cycling.
9	Effect of aluminium content on the dimensional stability of uranium under α/β thermal cycling.

<u>Figure Number</u>	<u>Caption</u>
10	Comparison of the dimensional stability of two specimens of 2.1 at. % vanadium uranium alloy.
11(a) & 11(b)	Effect of (a) 100 (b) 200 thermal cycles through the α/β phase change on two specimens of 2.1 at. % vanadium uranium alloy. In each case the specimen containing the coarse acicular nitride phase is on the left.
12	Dimensional stability of uranium alloys thermally cycled through the α/β transformation.
13	Temperature variation of the tensile properties of magnesium reduced uranium.
14	Temperature variation of the tensile properties of a uranium 2 at. % vanadium alloy.
15	Temperature variation of the tensile properties of a uranium 4 at. % molybdenum alloy.
16	Temperature variation of the tensile properties of a uranium $\frac{1}{2}$ at. % chromium alloy (as cast).
17	Temperature variation of the tensile properties of a uranium $\frac{1}{2}$ at. % chromium alloy (heat treated).
18	Thermal Expansion of uranium and uranium alloys.
19	Thermal conductivity of uranium compared with American data.
20	Thermal conductivity of Uranium Alloys.

FIG.1 TENSILE CREEP DATA FOR URANIUM

FIG.2 THE EFFECT OF LARGE THERMAL AND LOAD CYCLING ON THE CREEP BEHAVIOUR OF URANIUM.(400°C. 3TON/IN.²)

FIG. 3.
CREEP RATE AGAINST GRAIN SIZE FOR
URANIUM & URANIUM ALLOYS (400°C. -
3 tons per in^2)

FIG. 4 VARIATION OF SECONDARY CREEP RATE WITH GRAIN SIZE FOR A URANIUM $\frac{1}{2}$ ATOMIC PERCENT CHROMIUM ALLOY.

505 36

FIG.5.
CREEP STRENGTH OF URANIUM ALLOYS AT 500°C. 2 TONS/IN.².

FIG. 6. EFFECT OF ALLOYING ON CREEP STRENGTH OF URANIUM-MOLYBDENUM ALLOYS.

595 138

FIG. 8 DISTORTION OF URANIUM ALLOYS DURING α - β THERMAL CYCLING.

FIG. 9 EFFECT OF ALUMINIUM CONTENT ON THE DIMENSIONAL STABILITY OF URANIUM UNDER α - β THERMAL CYCLING.

FIG. 10 COMPARISON OF THE DIMENSIONAL STABILITIES OF TWO SPECIMENS OF 2.1 at. % VANADIUM - URANIUM ALLOY.

Fig. 11(a). Effect of 100 thermal cycles through the α/β phase change on two specimens of 2.1 at.% vanadium uranium alloy. The specimen containing the coarse acicular nitride phase is on the left.

Fig. 11(b). Effect of 200 thermal cycles through the α/β phase change on two specimens of 2.1 at.% vanadium uranium alloy. The specimen containing the coarse acicular nitride phase is on the left.

FIG.12 DIMENSIONAL STABILITY OF URANIUM ALLOYS
THERMALLY CYCLED THROUGH THE α - β PHASE TRANSFORMATION.

FIG. 13. TEMPERATURE VARIATION OF THE TENSILE PROPERTIES OF Mg REDUCED URANIUM.

FIG. 14 TEMPERATURE VARIATION OF THE TENSILE PROPERTIES OF A URANIUM 2 at.% VANADIUM ALLOY.

FIG. 15 TEMPERATURE VARIATION OF THE TENSILE PROPERTIES OF A URANIUM 4 at. % MOLYBDENUM ALLOY

FIG. 16 TEMPERATURE VARIATION OF THE TENSILE PROPERTIES OF A URANIUM $\frac{1}{2}$ at. % CHROMIUM ALLOY (AS CAST.)

FIG. 17 TEMPERATURE VARIATION OF THE
TENSILE PROPERTIES OF A URANIUM $\frac{1}{2}$ at.%
CHROMIUM ALLOY (HEAT TREATED)

FIG. 18. THERMAL EXPANSION OF URANIUM & URANIUM ALLOYS

FIG.19. THERMAL CONDUCTIVITY OF URANIUM COMPARED WITH AMERICAN DATA.

FIG. 20 THERMAL CONDUCTIVITY OF URANIUM ALLOYS.