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DEVELOPMENT OF HIGH-STRENGTH TERNARY AND QUATERNARY THORIUM-URANIUM BASE ALLOYS

Final Report

By
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November 1, 1961

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A Division of American-Standard
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Final Report

DEVELOPMENT OF
HIGH-STRENGTH TERNARY AND QUATERNARY
THORIUM-URANIUM BASE ALLOYS

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FOREWORD

Development of ternary and quaternary alloys containing thorium and uranium suitable for power-reactor applications has been conducted under Contract No. AT(04-3)-250, Project Agreement No. 5, by Advanced Technology Laboratories (ATL), a Division of American-Standard, for the U. S. Atomic Energy Commission. The investigation was performed during the period December 1958 through September 1961.

SUMMARY

The addition of alloying elements to base alloys of thorium containing 5 and 10 weight percent uranium (Th-5 and 10 w/o U) was investigated with regard to alloy structures and elevated-temperature mechanical properties. The alloys developed possess elevated-temperature tensile and stress-rupture strengths of essentially twice those associated with binary Th-5 w/o and 10 w/o U alloys at 600 to 800°C. The greater strength of these alloys should result in improved resistance to radiation-induced swelling at high temperatures.

Noticeable improvement in elevated-temperature mechanical properties resulted from the addition of 2 w/o and 5 w/o zirconium to the Th-5 w/o U and Th-10 w/o U base alloys. Zirconium was found to be the most effective additive in improving elevated-temperature mechanical properties, although significant improvement was noted with additions of columbium, molybdenum, carbon, aluminum, and beryllium. Both solid-solution strengthening and dispersion strengthening of the Th-U base alloys were encountered.

It was found that a fine intragranular distribution of the secondary uranium phase improved the mechanical properties of the alloys studied. Structures with optimum properties were produced by careful control of the arc-melting procedure and by solution heat treating at 1350°C followed by aging at 850°C. Preliminary testing was performed to determine the elevated-temperature mechanical properties of thorium metal and also of the Th-U base alloys.

Irradiation specimens of the most promising alloys have been prepared. These specimens have been encapsulated and are ready for irradiation testing at temperatures up to 800°C and for subjection to burnups of 10,000, 20,000 and 30,000 mwd/tonne.

INTRODUCTION

The long-term importance of nuclear fuels for power production depends on our ability to use not only the natural fissionable isotope U^{235} but also the artificial fissionable isotopes U^{233} and Pu^{239} . In the process of breeding the artificial fissionable isotopes, many metallurgical and nuclear properties must be considered. The most important metallurgical requirement is dimensional stability of the fertile material at high temperatures for long exposures to a reactor environment. The use of metallic thorium as a matrix material and also as a fertile material for the production of U^{233} offers a system of outstanding potential. Dimensional stability of the cubic structure of thorium is essentially undisturbed by irradiation and high-temperature cycling. Alloying 5 to 20% by weight of uranium is sufficient to incorporate the reactivity needed for use of thorium as a fuel element and yet is not sufficient to deteriorate seriously the metallurgical structure of the thorium. Irradiation tests of this binary alloy have been conducted at both Argonne National Laboratory and Atomics International. At Argonne National Laboratory,¹ thorium specimens containing up to 20 weight % (w/o) uranium were subjected to an estimated 2.8 atomic % (a/o) burnup at 650°C with good dimensional stability resulting. Earlier, Atomics International² had irradiated thorium specimens containing 11% uranium to 1.1 a/o burnup at 550°C with the same result. An AEC task force in 1959 indicated³ that Th-U alloys capable of operating at central temperatures of about 800°C are necessary to provide low fuel-cycle costs. By strengthening the Th-U alloy for high-temperature service and conducting irradiation tests, the ability to attain this goal can be determined.

As part of the AEC Fuel Cycle Program, Advanced Technology Laboratories has undertaken a project to determine what beneficial gains can be made in the elevated-temperature strength of Th-U alloys by alloying additions and by heat treatment. A number of elements of low-thermal-neutron cross section have been alloyed with Th-U base alloys. These elements, added in concentrations up to 15 w/o, include zirconium, yttrium, lanthanum, columbium, molybdenum, carbon, aluminum, and beryllium. To attain optimum strength of the alloys, heat-treating procedures were incorporated in the fabrication of the test specimens. Elevated-temperature tensile and stress-rupture tests were conducted.

1. Superscripts indicate references, page 116.

Results of irradiation tests of these alloys should be of keen interest in fuel programs for the Experimental Breeder Reactor II, Advanced Epithermal Thorium Reactor, Enrico Fermi Power Reactor, Sodium Reactor Experiment, Hallam Nuclear Power Facility, and other high-temperature reactors that have need for a high-temperature metallic fuel capable of high burnups.

TECHNICAL BACKGROUND

A large amount of physical metallurgical effort has been devoted to the metal thorium because of its long history as a potential fertile nuclear material. Considerable information on the fabrication, heat treatment, mechanical properties, and corrosion behavior of thorium and binary thorium alloys is available. The Second Edition of the Reactor Handbook, Volume I, "Materials," contains a comprehensive summary and bibliography of thorium alloys.⁴

Although binary alloy systems have been explored extensively, very little information concerning ternary or quaternary systems has been developed. Also, most data available on the mechanical properties of thorium alloys are primarily for room temperature. A small amount of elevated-temperature tensile-test data at 500°C are available for Th-U alloys,⁵ and hot-hardness values to 900°C are reported for Th-U alloys with additions of zirconium, columbium, and molybdenum.⁶ Work published to date has established the following:^{*}

1. The 500°C ultimate tensile strength and yield strength of commercial-grade thorium are increased approximately 50% when the thorium is alloyed with uranium.⁵
2. Small additions of columbium and molybdenum are effective in stabilizing the gamma-uranium phase in supersaturated Th-U alloys.⁶
3. Zirconium is an effective hardener of thorium at temperatures to 600°C.⁶
4. Carbon is an effective alloying element for hardening and strengthening thorium and is mainly responsible for the age-hardening characteristics noted in bomb-reduced thorium.⁷
5. Aging heat treatments are found to improve the normal temperature strength of thorium-carbon and thorium-nitrogen alloys. Work hardening followed by aging produces the maximum strengthening effects observed.⁴
6. The transition elements produce mild strengthening, except that titanium, zirconium, molybdenum, and vanadium produce a softening in bomb-reduced thorium by removing interstitial carbon.⁴
7. Aluminum, beryllium, silicon, tin, and indium produce some strengthening of thorium.

* Unless otherwise indicated the effects listed were noted at room temperature.

8. Lanthanum, yttrium, and cerium are reported to have a high solid solubility in thorium. Uranium and zirconium have a room temperature solubility of approximately 2 w/o, increasing at higher temperatures.

9. Intermetallic compounds are formed with thorium by aluminum, beryllium, silicon, bismuth, carbon, nitrogen, copper, magnesium, nickel, selenium, iron, and manganese. Some of these may possess slight solid solubility at elevated temperatures.

The materials that would most likely improve the elevated-temperature mechanical properties of thorium were selected on the basis of the above information. The alloying elements selected for study were those which would provide either solid-solution strengthening or dispersion strengthening and were limited to those possessing low-thermal-neutron cross section. Because yttrium and lanthanum were known to have high solubility in thorium and because zirconium had shown both solubility in and hardening of thorium, these three elements were selected as solid-solution strengthening agents for the Th-U base alloy.

Development of a dispersion-hardening alloy was based on the formation of either carbides or intermetallic dispersions. Two carbide systems were selected:

- 1) Carbon forming a dispersion of thorium carbide throughout the matrix of the base alloy.
- 2) Zirconium and carbon forming a dispersion of zirconium carbide throughout the matrix of the base alloy.

Two intermetallic dispersion systems were selected:

- 1) A system containing up to 0.3 w/o aluminum that would exist as either thorium- or uranium-aluminum compounds.
- 2) A system containing up to 0.3 w/o beryllium that would exist as either thorium or uranium beryllides.

Heat-treating and aging studies were conducted with both the carbide and the intermetallic systems in order to obtain optimum dispersion in the matrix material.

The solubility of uranium in thorium is low at the temperatures of interest (approximately 1.5 w/o at 600 to 800°C), and the base alloy studied should be considered a dispersion alloy. The addition of uranium to thorium in an amount greater than that which is soluble has resulted in improved tensile properties.⁵ From this it is logical to conclude that the size and spacing of the uranium dispersion will further affect the elevated-temperature mechanical properties.

The effect of the size and distribution of the secondary uranium phase on swelling due to irradiation damage is also of interest. The gases xenon and krypton produced from the fission of uranium have a very low solubility limit in metals; hence, their presence in the matrix of the fuel creates internal stresses. These gas atoms can migrate by diffusion through the lattice of the fuel either until they are trapped by imperfections such as vacancies and inclusions or until they coalesce with other gas atoms. Eventually, both of these processes produce bubbles of gas in the fuel, and these bubbles exert a hydrostatic force on the surrounding matrix. To prevent swelling of the fuel, the strength of the matrix metal surrounding the bubble must exceed the stress created by the hydrostatic force of the fission gases. For this reason, it is important to increase the elevated-temperature mechanical properties of the matrix material.

Since there is a fission-damaged region surrounding each "island" of uranium, the size and spacing of this dispersion is also of interest. If the spacing diminishes to a certain value, there may be an overlapping of fission-damaged regions of the matrix, thereby either improving or degrading the strength of the restraining matrix material. Predicting the change in the strength of the matrix material due to fission damage is a complex problem, and a quantitative treatment of the subject would, at best, be open to question. Therefore, the approach taken in this investigation has been to obtain maximum elevated-temperature mechanical properties for the matrix material regardless of the size and distribution of the fissile dispersion. The optimum heat treatment resulted in a uranium dispersion with an average diameter of approximately 10^{-6} (1 μ) inch and an average spacing of approximately 10^{-5} (10 μ) inch.

EXPERIMENTAL PROCEDURES

A. Materials

Calcium-reduced, high-purity thorium was used for all alloys studied. The uranium added to the thorium was depleted reactor-grade material. Alloy elements used were of comparable quality. Typical analyses of the materials used are presented in Table I.

B. Physical Testing

The long-term, high-temperature service of a fuel element places the fuel under a condition of continued low strain which causes creep. For this reason, it is important to consider the resistance of the alloys to creep.

Because of the time required for creep testing and the large number of samples to be tested, an evaluation method of fairly short duration (less than 100 hours) that could reliably indicate the creep properties of an alloy was desired. A stress-rupture test was employed in which the minimum-creep-rate stress was plotted versus a parameter proposed by Larson and Miller.⁸

The Larson and Miller parameter, $P = T(20 + \log t)$, where T = test temperature (°Rankine), t = rupture time (hours), and 20 = constant of material, is based on the theory that creep at elevated temperatures is a rate-controlled process. For a given stress level the parameter will be constant, and by calculation various desired times and/or temperatures can be determined. When a plot of minimum-creep-rate stress versus the parameter has been determined, the relationship of temperature to time to rupture can be calculated for any stress level by finding the value of P corresponding to the stress level of interest and calculating from $P = T(20 + \log t)$ the desired values of time or temperature. In this way, it is possible to find the resulting rupture time for any stress level and temperature of interest.

In order to make precise predictions of rupture times from these curves, it would be necessary for the curves to be very precisely determined. Such precision would require an exceedingly large number of tests on each alloy. The curves presented in this report are the result of only a few tests on each alloy and, therefore, are not suitable for precise rupture-life predictions. However, the purpose of this study was to obtain relative information on the various alloys studied; and for this purpose the minimum-creep-rate stress versus Larson-Miller parameter curves presented are

TABLE I
TYPICAL ANALYSES OF METALS USED FOR
THORIUM-ALLOY PREPARATION
(Values in ppm)

Thorium

C - 160	Ca - 200	Mn - 0.8
N - 130	Cd - 0.2	Ni - 80
U - 11	Cr - 40	Zn - 50
Al - 10	Cu - 40	Dy - 2.5
B - 0.2	Fe - 550	Eu - 0.1
	Mg - 10	ThO ₂ - 2000

Uranium

C - 57	Al - 6	Mg - 4
N - 18	Ca - 3	Mn - 20
Cl - 2	Cu - 8	Ni - 25
Ag - 0.1	Fe - 30	Si - 20

Zirconium

Al - 54	H - 10	N - 27
B - 0.2	Fe - 690	O - 760
C - 95	Pb - 10	Si - 64
Cr - 72	Mg - 10	Ti - 20
Co - 5	Mn - 26	V - 20
Cu - 20	Mo - 10	
Hf - 67	Ni - 10	

Yttrium

O - 630	Co - 10	Ni - 30
N - 30	Cr - 100	Pb - 50
C - 80	Cu - 10	Si - 10
Al - 10	Fe - 75	Sn - 5
B - 20	Mg - 8	Sr - 100
Ba - 10	Mn - 8	Ti - 50
Ca - 40	Mo - 50	V - 10
Cb - 20	Nb - 100	Zr - 100

Lanthanum

Ce - 2,000	Ca - 100	Ni - 1,000
Al - 100	Fe - 500	O - 580
Si - 100	Cu - 100	

adequate. In the procedure presented by Larson and Miller, the rupture stress of the materials tested was used. In the testing of thorium alloys a better correlation was achieved by plotting minimum-creep-rate stress versus the Larson-Miller parameter.

1. Tensile Testing

All tests were run in vacuum at temperatures up to 800°C. The equipment used is shown in Figure 1. The specimen was held within the vacuum furnace mounted on the tensile machine.

- a. A standard strain rate of 0.050 inch/inch/minute was used for all tests.
- b. Test specimens were of rectangular cross section, having a 2-inch gauge length and a cross-sectional area of approximately 0.025 square inch.

2. Stress-Rupture Testing

All tests were run in vacuum at temperatures between 600 and 800°C. The pair of stress-rupture furnaces used are shown in Figure 2. Instrumentation control and readout are on the right-hand side of the equipment.

Test specimens for the stress-rupture tests were identical to those used for tensile testing. Stress levels were selected to keep the time to rupture below 100 hours in most tests. In general, the alloys were tested under the following conditions:

<u>Stress</u> <u>(psi)</u>	<u>Temperature</u> <u>(°C)</u>
8000	600
1000	800
2000	800

C. Metallography

Standard methods of mechanical polishing with diamond abrasives were used. Thorium metal and its alloys are normally difficult to etch by either chemical or electrolytic means. However, an electrolytic etch composed of 130 milliliters of ethyl alcohol and 8 milliliters of perchloric acid was used successfully. A different voltage and time were necessary for each alloy series. The following conditions were found to provide the most satisfactory results:

1. Th-U, Y and La alloys - 70 volts, 1/2 sec, followed by 50 volts, 1 sec.
2. Th-U, Zr alloys - 25 volts, 2 to 10 sec.
3. Th-U-C alloys - 40 volts, 1 sec.
4. Th-U, base alloys - 30 volts, 2 to 10 sec.

ELEVATED-TEMPERATURE, VACUUM TENSILE TESTING EQUIPMENT

FIGURE 1

ELEVATED-TEMPERATURE,
VACUUM STRESS-RUPTURE TESTING EQUIPMENT

FIGURE 2

D. Alloy and Specimen Preparation

1. Melting Procedure

All alloys were prepared in the nonconsumable tungsten-electrode arc-melting furnace shown in Figure 3. A rectangular-shaped, water-spray-cooled, 25-cubic-centimeter copper crucible was used for ingot melting. An argon atmosphere was achieved in the furnace by evacuation and purging.

It was found that the tensile properties and stress-rupture properties were dependent on the structure of the alloy. For example, the Th-5 w/o U base alloy specimens when tensile tested at 800°C showed strengths ranging from approximately 2000 to 4000 psi. An investigation of this variation revealed that the strength of the alloy was dependent upon the dispersion of uranium throughout the thorium matrix. Metallographic examination of the structures disclosed a very fine uranium dispersion in some specimens and in others large particles of uranium both within the grain and as a discontinuous-grain-boundary phase. Tensile properties of the alloys possessing the fine uranium dispersion were definitely superior to those containing large particles of uranium. A standard arc-melting procedure was thus formulated, and alloys produced by this method thereafter consistently yielded optimum properties. The basic metallurgical requirement is to attain a uniform solution of uranium and thorium in the liquid state and then to preserve this condition by a rapid quench from the molten state. Since thorium and uranium exhibit little solid solubility at lower temperatures, it is necessary to contain the finely dispersed liquid uranium after it is entrapped by the solidified thorium matrix and to freeze it quickly before it can be expelled from solution and form large particles or grain-boundary phases. By this procedure the amount of the pure uranium-grain boundary phase is greatly diminished, and a fine uranium dispersion can be obtained.

Following is the standard melting procedure established:

- a. Melt at as low an arc voltage as practical. In the equipment used, the lower limit was found to be 8 to 10 volts.
- b. Melt at a high current to achieve vigorous stirring action and follow by an abrupt power shutdown. The equipment used allowed a current of 900 amperes.
- c. Use a high rate of coolant flow on the copper crucible to achieve a rapid quench.

NONCONSUMABLE TUNGSTEN-ELECTRODE ARC-MELTING EQUIPMENT

FIGURE 3

2. Test Specimen Fabrication Methods

The progression of shapes in the process of obtaining the test specimens is shown in Figure 4. The ingots were melted as described in the standard procedure. A neutral salt bath was then used for heating the ingots to 700°C. The ingots were given four reductions of 0.020-inch per pass and reheated for 3 minutes. The rolling procedure was repeated until a final thickness of 0.100 inch was achieved. For alloys containing carbon or 0.3 w/o aluminum additions, it was found that a 0.010-inch reduction per pass was a maximum; greater reductions resulted in severe edge cracking. Two fabrication methods following this procedure were used.

a. Method A

The ingots were rolled from a thickness of 0.650 to 0.100 inch, approximately an 85% reduction in area. No additional intermediate heat treatments were introduced.

b. Method B

This method involved the introduction of a severe heat treatment into the fabrication procedure. The ingots were reduced 50% in area by rolling at 700°C. They were then heat treated for 24 hours at 1000°C in vacuo. This was done in an attempt to produce a homogenous distribution of uranium throughout the ingot and was not successful. Finally the ingots were rolled at 700°C to 0.100-inch thickness.

Method B yielded a recrystallized grain size nearly the same as that produced by Method A. However, the physical properties of the alloys had noticeably deteriorated. Examinations of the microstructures showed that this heat treatment caused the dispersed uranium particles to agglomerate resulting in a drop in tensile and stress-rupture properties.

Regardless of whether Method A or B was employed, slight edge cracking was usually encountered with certain alloys; however, in some this edge cracking could be reduced by decreasing the reduction per pass during rolling. The relative fabrication properties of all alloys studied are presented in Table II.

3. Annealing

It is important that no structural changes take place during the actual use of this material as a reactor fuel. Therefore, the recrystallization temperatures of the alloys were studied. All alloys exhibited recrystallization properties similar to those

ALLOY TEST SPECIMEN HISTORY

FIGURE 4

TABLE II
FABRICATION PROPERTIES OF THORIUM ALLOYS

Alloy (w/o)	Ability to Hot Roll (700°C)	Machinability
Th-5 U	Good	Good
Th-5 U-2 Zr	"	"
Th-5 U-5 Zr	"	"
Th-5 U-10 Zr	"	"
Th-5 U-15 Zr	"	"
Th-5 U-2 Y	"	"
Th-5 U-5 Y	"	"
Th-5 U-10 Y	"	"
Th-5 U-15 Y	"	"
Th-5 U-2 La	Fair	"
Th-5 U-5 La	"	"
Th-5 U-10 La	"	"
Th-5 U-15 La	Poor	"
Th-5 U-0.2 C	Fair*	Difficult
Th-5 U-0.25 C	"*	"
Th-5 U-0.3 C	Poor*	Not machinable
Th-5 U-0.5 C	Not rollable	"
Th-5 U-5 Zr-0.1 C	Good	Good
Th-5 U-5 Zr-0.2 C	"	"
Th-5 U-5 Zr-0.3 C	"	"
Th-5 U-5 Zr-0.5 C	"	"
Th-5 U-1 Al	Not rollable	-
Th-5 U-0.5 Al	Partially rollable	-
Th-5 U-0.3 Al	Good	Good
Th-5 U-0.2 Al	"	"
Th-5 U-0.1 Al	"	"
Th-10 U	"	"
Th-10 U-2 Cb	"	"
Th-10 U-1.5 Mo	"	"
Th-10 U-2 Zr	"	"
Th-10 U-5 Zr	"	"

* Notch sensitivity in these alloys requires that the surfaces be machined prior to rolling.

of the base alloys. At 700°C the alloys exhibited partial recrystallization, but a higher temperature was required for complete recrystallization. Since applications for this fuel might require 800°C, the selection of 850°C as the annealing temperature satisfied both the need for complete recrystallization and for a stable structure. Time variations of 1 to 6 hours at 850°C showed that after 2 hours a stable structure was produced. Specimens fabricated by both Methods A and B displayed nearly identical recrystallization properties. Therefore, a standard vacuum-annealing procedure of 2 hours at 850°C was used for all alloys studied.

4. Solution Heat Treatment

It was found that most of the alloys tested were amenable to a solution heat treatment to control the dispersion of uranium in the microstructure of the alloys. By use of the solution heat treatment followed by an aging process, the mechanical properties of the alloys were further enhanced.

The solution-heat-treatment procedure consisted of heating the machined specimens in vacuo at 1350 to 1360°C for $3\frac{1}{2}$ hours. The specimens were then removed from the hot zone of the furnace and were quickly quenched to room temperature by a rapid flow of helium gas. The specimens were then aged for 2 hours at 850°C prior to testing.

EXPERIMENTAL RESULTS

As an aid to the reader in utilizing the numerous illustrations and tables presented in this section, the following guide is offered:

1. The alloy or alloys studied are identified in the title.
2. Unless otherwise indicated, the condition of the alloy(s) should be assumed to be rolled and annealed.
3. "Heat-Treated" means solution heat treated and "Not Heat Treated" means rolled and annealed.
4. Unless otherwise indicated, aging is for 2 hours.
5. Unless otherwise indicated, the fabrication method was Method A (Experimental Procedures).
6. Figures relating to each alloy system studied are grouped in the following general order after the description of the alloy results:
 - a) Phase diagram of the alloy system if available.
 - b) Microstructure evaluation.
 - c) Ultimate tensile strength versus temperature.
 - d) Yield strength versus temperature.
 - e) Effect of alloy concentration versus ultimate tensile strength.
 - f) Effect of alloy concentration versus yield strength.
 - g) Larson-Miller parameter versus minimum-creep-rate stress for various concentrations of alloy within the same system.
 - h) Microhardness versus temperature.
7. Specific test conditions and results of the individual tensile tests and stress-rupture tests are listed in Appendices A and B.

A. Thorium-Uranium Base Alloy

The use of thorium as a fuel material depends on the incorporation of a certain amount of uranium to provide the required reactivity. The range of uranium contents to meet most requirements for reactor cores and for irradiation experiments has been from 5 to 10 w/o. This program of alloy development has used Th-5 w/o U as the principal base alloy. The Th-10 w/o U was tested to find if the best alloys developed were affected by the higher uranium content. As found by Bentle⁵ and shown in Figure 5,

EFFECT OF URANIUM CONTENT ON
TENSILE PROPERTIES OF COMMERCIAL-GRADE THORIUM

FIGURE 5

the increased strength of the binary Th-U alloy as a result of uranium additions has already been achieved in the Th-5 w/o U alloy; and further additions (to 8 w/o U) have little effect on the strength.

As shown by the phase diagram of the Th-U system in Figure 6, alloys containing 5 to 10 w/o uranium are supersaturated alloys at room temperature. However, by heating to 1350°C, a percentage of the uranium can be put into solution and with a rapid quench held in solution. The dispersion characteristics of the uranium in the thorium can be controlled by aging.

The structural factors desired are a very fine uranium dispersion and the absence of any grain-boundary phase of high uranium content. Since pure uranium loses its mechanical strength rapidly above 750°C, it is very important that no continuous uranium phase exists at the grain boundaries. The alloy would then tend to possess the strength of uranium and suffer from grain-boundary failure. As noted earlier, these conditions were suitably controlled by establishment of appropriate arc-melting procedures.

Figure 7 shows the microstructures resulting from the two different fabrication techniques described in the section on Experimental Procedures. Method-A structure shows a finely dispersed uranium phase. Method B was intended to homogenize the structure by heat treatment of 24 hours at 1000°C; however, the actual effect was over-aging, thus causing agglomeration of the secondary uranium dispersion phase. Although both photomicrographs show a dark gray phase (thorium oxide), another phase is clear in the structure fabricated by Method B. This phase, slightly lighter in appearance than the oxide phase, is the agglomerated uranium. The deleterious effect on tensile properties of the uranium phase being in an agglomerated condition is shown in Figure 8.

The solution heat treatment of the base alloy was found to produce a homogeneous (nearly single-phase) structure. Subsequent aging of the heat-treated structure resulted in the precipitation of a very fine secondary phase of uniformly dispersed uranium particles. Figure 9 shows the effect of the solution heat treatment and aging on the microstructure. By proper solution heat treatment, the uranium dispersion of the alloys prepared by Method B became indistinguishable from those fabricated by Method A.

Elevated-temperature tensile testing of the solution-heat-treated and aged base alloy indicated improved tensile strength (Figure 10). At a test temperature of 700°C, a more pronounced improvement resulted from heat treatment and aging than at 600 and 800°C.

PARTIAL THORIUM-URANIUM PHASE DIAGRAM⁵

FIGURE 6

a. Fabricated by Method A.

b. Fabricated by Method B.

**MICROSTRUCTURES OF ROLLED AND ANNEALED
Th-5 w/o U BASE ALLOY FABRICATED BY METHODS A AND B
(500×)**

FIGURE 7

EFFECT OF FABRICATION TECHNIQUE ON
TENSILE PROPERTIES OF Th-5 w/o U BASE ALLOY

FIGURE 8

a. Annealed.

b. Solution Heat Treated.

c. Solution Heat Treated and Aged.

EFFECT OF HEAT TREATMENT AND AGING ON
THE MICROSTRUCTURE OF Th-5 w/o U BASE ALLOY
(500 $\times$)

FIGURE 9

EFFECT OF HEAT TREATMENT AND AGING ON
TENSILE PROPERTIES OF Th-5 w/o U BASE ALLOY

FIGURE 10

However, the stress-rupture test results at 600 and 800°C showed a very significant improvement in rupture life resulting from subjecting the alloy to heat treatment and aging (Figure 11). When tested at 600°C and 8000 psi, the heat-treated and aged specimen exhibited a rupture life approximately 100 times that of the rolled and annealed (i. e., not heat treated) specimen; when tested at 800°C and 2000 psi, the improvement in rupture life was approximately 20 times. This improvement in stress-rupture properties is attributed to the formation of large homogeneous grains with low-angle boundaries generally free of any continuous uranium phase.

B. Thorium-Uranium-Zirconium System

1. Rolled and Annealed Structure

The effect of up to 15 w/o zirconium additions to the Th-5 w/o U base alloy was investigated. Typical rolled and annealed structures of these alloys are shown in Figure 12. The optimum mechanical properties exhibited by these alloys at elevated temperatures in the rolled and annealed condition are shown in Figures 13 through 16 (tensile properties). From these results, it appeared that, above 600°C, 2 and 5 w/o zirconium additions provided the greatest improvements in the mechanical properties of the Th-5 w/o U base alloy. Stress-rupture properties of these annealed alloys are presented in Figure 17. To supplement the data obtained, a hot-hardness study was performed on the 2 and 5 w/o zirconium alloys. The results of this study are presented in Figure 18. There was no obvious correlation between the hot-hardness values and the tensile properties of those alloys. This was particularly true at 800°C, where the hot-hardness values for both alloys were nearly the same, and the tensile tests indicated that the 5 w/o zirconium alloy had approximately 3000 psi greater ultimate tensile strength and 4000 psi greater yield strength than the 2 w/o zirconium alloy. Hot-hardness testing appeared to be of little value in studying the elevated-temperature tensile properties of these alloys.

The effect of the uranium dispersion on the strength of the alloys containing zirconium is tabulated in Table III. In this case also, it is apparent that the agglomerated structure produced by Method B provided a deleterious effect.

Since noticeable improvement in tensile properties resulted from the addition of 2 and 5 w/o zirconium to the Th-5 w/o U base alloy, the effects of zirconium on thorium were studied. By determination of the elevated-temperature tensile properties of Th-Zr, Th-U and Th-U-Zr, the singular and combined effects of alloying constituents might be revealed.

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
FOR Th-5 w/o U BASE ALLOY

FIGURE 11

a. 2 w/o Zr.

b. 5 w/o Zr.

c. 10 w/o Zr.

d. 15 w/o Zr.

MICROSTRUCTURE OF
ROLLED AND ANNEALED Th-5 w/o U-Zr ALLOYS
(500 $\times$)

FIGURE 12

ULTIMATE TENSILE STRENGTH VS. TEMPERATURE FOR
Th-5 w/o U-Zr ALLOYS

FIGURE 13

YIELD STRENGTH VS. TEMPERATURE FOR
Th-5 w/o U-Zr ALLOYS

FIGURE 14

EFFECT OF ZIRCONIUM CONTENT ON
ULTIMATE TENSILE STRENGTH OF Th-5 w/o U-Zr ALLOYS

FIGURE 15

EFFECT OF ZIRCONIUM CONTENT ON YIELD STRENGTH OF
Th-5 w/o U-Zr ALLOYS

FIGURE 16

a. Th-5 w/o U-2 w/o Zr Alloy

b. Th-5 w/o U-5 w/o Zr Alloy

c. Th-5 w/o U-10 w/o Zr Alloy

d. Th-5 w/o U-15 w/o Zr Alloy

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
 FOR Th-5 w/o U-Zr ALLOYS

FIGURE 17

MICROHARDNESS VS. TEMPERATURE FOR
THORIUM AND Th-5 w/o U-Zr ALLOYS

FIGURE 18

TABLE III
COMPARISON OF TENSILE PROPERTIES FOR Th-5 w/o U-Zr ALLOYS
FABRICATED BY METHOD A AND METHOD B

Test Temperature (°C)	Th-5 w/o U		Th-5 w/o U-2 w/o Zr		Th-5 w/o U-5 w/o Zr		Th-5 w/o U-10 w/o Zr	
	A	B	A	B	A	B	A	B
	Ultimate Tensile Strength (psi)							
400	22,165	17,675	29,660	25,875	37,085	30,680	53,750	43,000
600	11,440	11,060	19,578	17,830	22,000	18,260	21,450	18,800
800	3,930	2,325	7,930	<1,000	10,800	<1,000	10,600	4,250
	Yield Strength (psi)							
400	9,450	6,695	14,240	9,950	18,016	12,908	34,500	26,200
600	8,065	5,425	11,420	9,330	13,900	11,120	19,236	15,300
800	3,325	1,140	5,820	<1,000	9,950	<1,000	9,950	3,500

According to the phase diagram⁹ of thorium-zirconium, Figure 19, zirconium has a limited solubility (approximately 3 w/o) in thorium at equilibrium conditions. Specimens of Th-Zr were prepared using essentially the same techniques that had been used for the Th-U-Zr specimens.

The tensile properties of Th-2 w/o Zr and Th-5 w/o Zr alloys in the rolled and annealed condition are shown in Figures 20 and 21. The binary Th-2 w/o Zr alloy exhibited nearly the same ultimate tensile strength and yield strength from 400°C through 800°C as the ternary Th-5 w/o U-2 w/o Zr alloy (Figures 13 and 14). Apparently the zirconium was almost entirely responsible for the improvement in elevated-temperature tensile properties of the Th-5 w/o U-2 w/o Zr alloy. However, this was not true for alloys containing 5 w/o zirconium. The Th-5 w/o Zr binary alloy had lower tensile properties than the Th-5 w/o U-5 w/o Zr ternary alloy (Figures 13 and 14). Apparently the presence of both the zirconium and uranium contributed to the improved elevated-temperature properties.

Hot-hardness data on these two binary alloys (Th-2 w/o Zr and Th-5 w/o Zr) were also obtained. As shown in Figure 22, from room temperature to 500°C there was a gradual decline in hardness values. However, over this same temperature range, the yield strength and ultimate tensile strength of the alloys changed only slightly. Above 500°C, the trend of hot-hardness and tensile properties was similar. Any definite correlation between tensile properties and hot-hardness values at the temperatures of interest appears unlikely.

Two types of structures predominate in the ternary Th-U-Zr alloys: a solid-solution structure with a fine uranium dispersion (2 w/o and 5 w/o Zr) and a typical eutectoid type of structure (10 w/o and 15 w/o Zr). Examination of the Th-Zr binary phase diagram in Figure 19 shows that this type of behavior can be expected in these ternary systems if the effects of the uranium content are relatively unimportant. The structure of the Th-5 w/o U-2 w/o Zr alloy appeared to be one of complete solid solution in thorium except for the typical second phase of finely dispersed uranium. The 5 w/o zirconium alloy was similar to the 2 w/o except that there was a significant amount of a third phase present. This third phase was believed to be the result of a limited amount of Th-Zr eutectoid being formed. This condition also is indicated by the Th-Zr phase diagram.

THORIUM-ZIRCONIUM PHASE DIAGRAM⁹

FIGURE 19

ULTIMATE TENSILE STRENGTH VS. TEMPERATURE FOR
THORIUM AND Th-Zr ALLOYS

FIGURE 20

YIELD STRENGTH VS. TEMPERATURE FOR
THORIUM AND Th-Zr ALLOYS

FIGURE 21

MICROHARDNESS VS. TEMPERATURE FOR
THORIUM AND Th-Zr ALLOYS

FIGURE 22

A close examination of both of these structures showed that a discrete, but nearly continuous, grain-boundary phase sometimes exists throughout the material. Only by careful control of the melting procedure or by heat treating was it possible to eliminate this condition. Tensile testing at 800°C indicated extreme sensitivity of the alloys to the presence of this grain-boundary phase. Those ingots without this phase showed good tensile strength at 800°C, while those with it had only 10 to 20% the maximum tensile strength attainable. Testing at 800°C resulted in brittle failure of an intergranular type due to the presence of the weak secondary phase in the grain boundaries. Although metallographic examination supports this theory, the secondary phase has not been positively identified. It is probably a uranium-rich phase expelled by the thorium during solidification. It is suspected that the zirconium in forming a solid solution with the thorium might have changed the kinetics of the thorium-uranium system.

It has been found that tensile testing at 700°C resulted in a transgranular ductile failure and at 800°C an intergranular brittle failure occurred. These are significant factors when viewed from the standpoint of the tensile strength of uranium. It is known that the uranium does exhibit very poor tensile strength at 800°C. Therefore, when an alloy with a nearly continuous uranium-rich grain-boundary phase is tested, poor strength can be expected at 800°C. Failures in this range would most likely be intergranular.

2. Solution-Heat-Treated Structure

Based on the premise that a uranium-rich grain-boundary phase is expelled from the Th-Zr matrix, a heat-treatment study was initiated. In alloys with up to 12 w/o uranium content, there is a region in the binary Th-U diagram (Figure 6) where extensive solid solubility exists. For the Th-5 w/o U base alloy, this region is from about 1200 to 1300°C in the α -thorium region and from 1350 to 1450°C in the β -thorium region. By subjecting the alloys to proper solution heat treatments in one of these areas and then giving them a suitable quench, the uranium should be retained in solid solution. After the solution heat treatment, the uranium should be precipitated as a fine dispersion by an aging treatment. An investigation to check the feasibility of this procedure was performed. The best results were obtained by using the solution-heat-treating procedure described in the section on Experimental Procedures.

Figure 23 is a graph of ultimate tensile strength versus zirconium content for specimens that had been heat-treated and aged. A comparison with Figure 15 shows that the alloying, heat treating, and aging more than tripled the ultimate tensile strength of the base alloy at 800°C. Essentially the same was true of yield strength (cf. Figures 24 and 16). It is also significant that these results were easily reproduced. Improvements in strength and in consistency of results of the heat-treated specimens were attributed to establishment of a more homogeneous structure, reduction of continuous weak grain boundary phases, and redistribution of the secondary phases. The effects of heat treating and aging on the microstructures of the Th-5 w/o U-2 w/o Zr and Th-5 w/o U-5 w/o Zr alloys are shown in Figures 25 and 26.

Stress-rupture properties of the heat-treated and aged 2 and 5 w/o Zr alloys were determined at temperatures of 600 and 800°C. The results of these tests are shown on Figure 27. The rupture life was improved considerably by the heat-treatment and aging process. From Figure 27b (5 w/o Zr alloy), the calculated rupture life for a stress level of 4000 psi and temperature of 800°C is approximately 36 times greater as a result of alloying and heat treating. Further evaluation of the effects of alloying and heat treating showed that for a testing temperature of 800°C and a rupture life of 0.35 hour, the non-heat-treated base alloy could withstand only 4000 psi, and the 5 w/o Zr alloy in the heat-treated condition could withstand 11,500 psi. The improvement in stress-rupture properties resulting from heat-treatment and aging operations on these alloys was attributed in part to the use of the fabrication procedure developed.

3. High Concentrations of Zirconium

The 10 and 15 w/o zirconium alloys consisted of a mixture of the eutectoid of zirconium and thorium (70 w/o Zr-30 w/o Th) in a matrix of pro-eutectoid α thorium. As the zirconium content increased, so did the amount of eutectoid in the structure. These alloys possessed high ductility at all test temperatures, but at elevated temperatures their tensile properties were inferior to those of alloys of lower zirconium content. The results of physical testing on these alloys in a rolled and annealed condition are included in Figures 13, 14, 15, 16, 17c, and 17d. The transformation of a metastable, complete solid solution of the Th-5 w/o U-15 w/o Zr alloy obtained by solution heat treatment was studied to determine what improvement, if any, might be available for elevated-temperature operation. The results of this study are shown in

EFFECT OF ZIRCONIUM CONTENT ON
ULTIMATE TENSILE STRENGTH OF
HEAT-TREATED AND AGED
Th-5 w/o U-Zr ALLOYS

FIGURE 23

EFFECT OF ZIRCONIUM CONTENT ON YIELD STRENGTH OF
HEAT-TREATED AND AGED Th-5 w/o U-Zr ALLOYS

FIGURE 24

a. Annealed.

b. Solution Heat Treated.

c. Solution Heat Treated and Aged.

EFFECT OF HEAT TREATMENT AND AGING ON
THE MICROSTRUCTURE OF Th-5 w/o U-2 w/o Zr ALLOY
(250 $\times$)

FIGURE 25

a. Annealed.

b. Solution Heat Treated.

c. Solution Heat Treated and Aged.

EFFECT OF HEAT TREATMENT AND AGING ON
THE MICROSTRUCTURE OF Th-5 w/o U-5 w/o Zr ALLOY
(250 $\times$)

FIGURE 26

a. Th-5 w/o U-2 w/o Zr Alloy

b. Th-5 w/o U-5 w/o Zr Alloy

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
 FOR HEAT TREATED AND AGED Th-5 w/o U-Zr ALLOYS

FIGURE 27

Figure 28. Microhardness measurements indicated no apparent improvement in strength resulted from transformation of this metastable structure. Generally, from the structures and elevated-temperature properties observed in the investigation of zirconium additions, it appeared that alloys containing more than 5 w/o zirconium content had a deterioration of properties.

4. Aging Studies

Major improvements in the mechanical properties of the alloys are attributed to the structure resulting from the heat treating and aging. This structure must be stable if the alloys are to retain the greater strengths at elevated temperatures for long exposures. Therefore, in evaluating these alloys for reactor use, the thermal stability of the structure was determined at the highest temperatures expected in reactor service.

Specimens of both Th-5 w/o U-2 w/o Zr and Th-5 w/o U-5 w/o Zr alloys were heat-treated and aged from 100 to 200 hours at 800 and 850°C, and then tensile tested at 800°C. The results of this investigation are presented in Table IV. It was apparent that aging 100 hours at 800°C was required to produce a stable structure in the Th-5 w/o U-2 w/o Zr alloy, with a reduction of the 800°C tensile strength to approximately 35 to 40% of that when aged 2 hours at 850°C. There was no further loss in tensile strength when the aging time was extended to 200 hours at 800°C or 170 hours at 850°C. Apparently, the standard aging treatment of 2 hours at 850°C was not adequate to produce a structure that would be stable at 800°C for this 2 w/o zirconium alloy, whereas aging for 100 hours at 800°C was adequate.

On the other hand, the 5 w/o zirconium alloy did not show a noticeable loss in tensile strength when the duration of the aging treatment was increased beyond the standard 2 hours. As shown in Table IV, there was a slight variation in the tensile strength of the 5 w/o zirconium specimens when subjected to different aging treatments. This variation was generally within the experimental range. Apparently the standard aging procedure of 2 hours at 850°C for the 5 w/o zirconium alloy produced a structure that was stable for at least 170 hours at 850°C.

Since reactor applications of these alloys are actually aimed at a fuel temperature of 400 to 800°C, these aging tests merely indicated that at the upper end of this temperature range the 2 w/o zirconium alloy was not stable with the standard aging treatment.

a. Annealed 2 Hours at 850°C.

b. Water Quenched from 1100°C.

c. Water Quenched and Aged
120 Hours at 800°C.

d. Water Quenched and Aged
500 Hours at 800°C.

TRANSFORMATION OF METASTABLE SOLID-SOLUTION
Th-5 w/o U-15 w/o Zr ALLOY AGED AT 800°C
(250×)

FIGURE 28

TABLE IV
EFFECTS OF AGING TREATMENTS
ON THE THERMAL STRUCTURAL STABILITY
OF HEAT-TREATED Th-5 w/o U-Zr ALLOYS

Aging Conditions	Ultimate Tensile Strength *	Yield Strength *
	(psi)	(psi)
<u>Th-5 w/o U-2 w/o Zr Alloy</u>		
Reference: 2 hours at 850°C	12,900	11,190
100 hours at 800°C	7,700	7,250
170 hours at 850°C	8,200	7,530
2 hours at 850°C plus 200 hours at 800°C	8,350	7,585
<u>Th-5 w/o U-5 w/o Zr Alloy</u>		
Reference: 2 hours at 850°C	13,230	11,730
100 hours at 800°C	11,800	4,800
170 hours at 850°C	14,115	12,460
2 hours at 850°C plus 200 hours at 800°C	12,050	10,400

* Testing temperature was 800°C.

However, in some applications at less than 800°C, the 2 w/o zirconium alloy may be stable and actually preferred to the 5 w/o zirconium alloy. Sufficient time was not available to explore the thermal structural stability of the solution-heat-treated and aged 2 w/o zirconium alloys at temperatures less than 800°C.

Microstructures of the Th-5 w/o U-2 w/o Zr and Th-5 w/o U-5 w/o Zr alloys are shown in Figures 29 and 30. These photomicrographs show the basic differences in the structure of these two alloys and also illustrate the overaging of the 2 w/o zirconium alloy.

C. Thorium-Uranium-Zirconium Alloys with Columbium Additions

1. Rolled and Annealed Structure

A further improvement of the alloys containing zirconium was attempted by the addition of 1 w/o columbium. This effort was limited to the alloys with small concentrations of zirconium. It was expected that columbium would alloy principally with the uranium, forming a uranium phase with a cubic structure that would be more resistant to radiation-induced swelling. The decreased amount of swelling of the uranium should lessen the forces exerted by the uranium on the matrix material. The amount of columbium added should be sufficient to form a uranium phase containing 20 w/o columbium.

Microstructures of the alloys are shown in Figure 31. In the rolled and annealed condition, the structures were similar to those of the Th-U-Zr alloys without the columbium additions. The results of tensile testing are shown in Figure 32. It is significant that these alloys, when tested at 800°C, did not have the structural sensitivity exhibited by the annealed Th-5 w/o U-2 w/o Zr alloys without columbium additions.

Apparently the columbium improved the 800°C strength of the uranium secondary phase. The stress-rupture results of the Th-U-Zr-Cb alloys tested are shown in Figure 33. No general statements can be made regarding the beneficial effect of the columbium addition, except that columbium appeared to improve the properties of the high uranium phase in the microstructure and as a result removed the structural sensitivity of the 2 w/o zirconium alloy.

None of these alloys, quaternary or ternary, in the rolled and annealed condition possessed 800°C-stress-rupture properties as high as those of the Th-5 w/o base alloy. This indicated that a uranium-rich grain-boundary phase probably still was present in the structure. Metallographic examination of the alloy structures in the rolled and

a. Heat Treated and Aged 2 Hours at 850°C.

b. Heat Treated and Aged 100 Hours at 800°C.

c. Heat Treated and Aged 170 Hours at 850°C.

EFFECT OF AGING TIME AND TEMPERATURE ON THE
STRUCTURAL STABILITY OF A HEAT-TREATED Th-5 w/o U-2 w/o Zr ALLOY
(250×)

FIGURE 29

a. Heat Treated and Aged 2 Hours at 850°C.

b. Heat Treated and Aged 100 Hours at 800°C.

c. Heat Treated and Aged 170 Hours at 850°C.

EFFECT OF AGING TIME AND TEMPERATURE ON THE
STRUCTURAL STABILITY OF A HEAT-TREATED Th-5 w/o U-5 w/o Zr ALLOY
(250×)

FIGURE 30

a. Th-5 w/o U-2 w/o Zr-1 w/o Cb
Annealed.

b. Th-5 w/o U-2 w/o Zr-1 w/o Cb
Solution Heat Treated and Aged.

c. Th-5 w/o U-5 w/o Zr-1 w/o Cb
Annealed.

d. Th-5 w/o U-5 w/o Zr-1 w/o Cb
Solution Heat Treated and Aged.

MICROSTRUCTURES OF Th-5 w/o U-Zr ALLOYS
CONTAINING 1 w/o Cb ADDITIONS
(500 $\times$)

FIGURE 31

TENSILE PROPERTIES VS. TEMPERATURE FOR
Th-5 w/o U-Zr ALLOYS CONTAINING 1 w/o Cb ADDITIONS

FIGURE 32

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
 FOR Th-5 w/o U-Zr ALLOYS CONTAINING 1 w/o Cb ADDITION

FIGURE 33

annealed condition showed that portions of the grain boundaries were quite broad. These broadened grain boundaries appeared to contain a uranium-rich phase. If this is the case, solution heat treatment would be helpful for the purpose of homogenization. In summary, the addition of 1 w/o columbium to the 2 w/o and 5 w/o zirconium alloys was only slightly beneficial to specimens in the rolled and annealed condition.

2. Solution-Heat-Treated Structures

Since the Th-U-Zr alloys exhibited superior properties when solution heat treated at 1350°C, this temperature was also selected for the alloys containing 1 w/o columbium additions. The same procedure was used as previously described, and the alloys were tested over the temperature range of 600 and 800°C. Tensile testing results are shown in Figures 34 and 35, and the stress-rupture test results are shown in Figure 33. Heat treatment and aging resulted in a noticeable improvement in the elevated-temperature properties.

Structures of the heat-treated and aged alloys are shown in Figure 31. The effects of heat treatment on the quaternary alloy structure were best shown by the Th-5 w/o U-5 w/o Zr-1 w/o Cb alloy. Essentially the effect was one of producing a heavy residual substructure. The boundaries of the large grains were filled with a continuous grain-boundary network of a eutectoid-type structure. This grain-boundary condition was generally deleterious to high-temperature properties. However, the mechanical properties exhibited by these alloys when heat treated were essentially as good as those obtained from the corresponding heat-treated ternary alloys.

D. Thorium-Uranium-Yttrium System

The range of compositions studied in the yttrium system was from 2 to 15 w/o yttrium in the Th-5 w/o U base alloy. These alloys all showed complete solid solubility of yttrium in thorium, as indicated by the binary thorium-yttrium phase diagram,¹⁰ Figure 36. As in both the base alloy and the 2 and 5 w/o zirconium systems, the structure showed a very fine dispersion of uranium throughout the alloy matrix. The yttrium alloys studied do not exhibit the extreme structural sensitivity found in the Th-5 w/o U-2 w/o Zr alloys. Typical Th-U-Y structures are shown in Figure 37. As the concentration of yttrium increased, etching of the material became very difficult.

The elevated-temperature tensile properties of the rolled and annealed yttrium alloys are shown in Figures 38, 39, 40, and 41. Although some increase in strength

EFFECT OF HEAT TREATMENT AND AGING ON
ULTIMATE TENSILE STRENGTH OF Th-5 w/o U-Zr ALLOYS
CONTAINING 1 w/o Cb ADDITIONS

FIGURE 34

EFFECT OF HEAT TREATMENT AND AGING ON
YIELD STRENGTH OF Th-5 w/o U-Zr ALLOYS
CONTAINING 1 w/o Cb ADDITIONS

FIGURE 35

THORIUM-YTTRIUM PHASE DIAGRAM¹⁰

FIGURE 36

a. 2 w/o Y.

b. 5 w/o Y.

c. 10 w/o Y.

d. 15 w/o Y.

MICROSTRUCTURES OF Th-5 w/o U-Y ALLOYS
(500 $\times$)

FIGURE 37

ULTIMATE TENSILE STRENGTH VS. TEMPERATURE FOR
Th-5 w/o U-Y ALLOYS

FIGURE 38

YIELD STRENGTH VS. TEMPERATURE FOR
Th-5 w/o U-Y ALLOYS

FIGURE 39

EFFECT OF YTTRIUM CONTENT ON
ULTIMATE TENSILE STRENGTH OF Th-5 w/o U-Y ALLOYS

FIGURE 40

EFFECT OF YTTRIUM CONTENT ON
YIELD STRENGTH OF Th-5 w/o U-Y ALLOYS

FIGURE 41

resulted at all concentrations of yttrium, the most effective level was 15 w/o. The high level of yttrium concentration necessary for a noticeable improvement of high-temperature strength is most likely explained by the extreme similarity of the yttrium and thorium atoms. Since both have identical atomic radii of 1.80 Å, lattice distortion effects are probably of little significance. Any improvement of the elevated-temperature tensile properties of the base alloy by additions of yttrium probably resulted from the combined effects of various minor solid-solution hardening mechanisms.

Stress-rupture properties improved significantly with the addition of 2 w/o yttrium and decreased as the concentration of yttrium increased beyond 2 w/o. The Larson-Miller curves for these alloys, Figure 42, show this trend clearly. A noticeable decrease in the stress-rupture properties of the 5 w/o yttrium alloy was observed when compared with the 2 w/o yttrium alloy. Further additions beyond the 5 w/o level were only slightly more deleterious.

The effect of uranium agglomeration on the mechanical properties of the yttrium alloys is shown in Figures 42 and 43. Specimens fabricated by Method B were inferior to those fabricated by Method A.

Significant tensile strengthening of the thorium matrix by additions of yttrium was achieved only at an alloy composition of 15 w/o yttrium. Below 15 w/o, only a slight general improvement was realized. Improvement of stress-rupture properties was not achieved by the addition of large concentrations of yttrium - 2 w/o being the optimum concentration for this parameter. Irradiation of the alloys would be necessary to determine which alloy composition would be best for power-reactor-fuel applications.

E. Thorium-Uranium-Lanthanum System

The range of concentrations studied in the lanthanum system was from 2 to 15 w/o lanthanum in the Th-5 w/o U base alloy. The initial work on melting of these alloys revealed that segregation of lanthanum was a problem. This problem was solved by reducing the volume of the standard-25-cubic-centimeter ingot by 50%. By exercising caution in the fabrication treatment, suitable test specimens were produced with 2, 5, and 10 w/o lanthanum; the 15 w/o lanthanum alloys could not be fabricated. Typical structures of the alloys tested are shown in Figure 44. Although no binary Th-La phase diagram has been developed, it has been reported that extensive solid solubility exists.

a. Th-5 w/o U-2 w/o Y Alloy

b. Th-5 w/o U-5 w/o Y Alloy

c. Th-5 w/o U-10 w/o Y Alloy

d. Th-5 w/o U-15 w/o Y Alloy

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
FOR Th-5 w/o U-Y ALLOYS

FIGURE 42

EFFECT OF FABRICATION TECHNIQUE ON
ULTIMATE TENSILE STRENGTH OF Th-5 w/o U-2 w/o Y ALLOY
(Typical of all Th-5 w/o U-Y Alloys)

FIGURE 43

a. 2 w/o La.

b. 5 w/o La.

c. 10 w/o La.

MICROSTRUCTURES OF Th-5 w/o U-La ALLOYS
(500 $\times$)

FIGURE 44

This was observed to be true of the alloys studied. Tensile-testing results obtained showed only slight improvement in the elevated-temperature properties as a result of adding lanthanum to the Th-5 w/o U base alloy (Figures 45, 46, 47, and 48).

The results of stress-rupture testing indicated that 2 w/o lanthanum was the most effective concentration for improving the properties of this alloy system. Some improvement was gained at all concentrations studied for the 600°C, 8000-psi, stress-rupture tests; but at 800°C-test conditions, only the 2 w/o lanthanum alloy showed properties equal to the Th-5 w/o U base alloy (Figure 49). The improvements gained by the addition of lanthanum to the base alloy are only slight; this was the least effective of all the alloy additions studied in this investigation.

A comparison of the physical properties of the thorium and lanthanum atoms shows some differences which should produce some hardening effects. However, the melting point of lanthanum is 826°C and of thorium is 1790°C. The low melting point of lanthanum is probably responsible for its ineffectiveness in improving the elevated-temperature properties of the Th-5 w/o U base alloy. This explanation was supported in part by the fact that only at the lower test temperature of 400°C was there a noticeable improvement in tensile properties as the lanthanum content increased (Figures 47 and 48). This same result might be expected from the addition of other elements having a low melting point (e. g. indium).

F. Thorium-Uranium Base Alloy with Carbon Additions

The two most effective ways of developing elevated-temperature strength are solid-solution hardening and dispersion hardening. From the limited solubility indicated by the Th-C phase diagram (Figure 50), an alloy which would include both of these mechanisms might be developed using carbon. Previous work done on thorium has shown that at room temperature a noticeable hardening of thorium results when interstitial carbon is present,¹¹ and that the formation of thorium carbide by aging heat treatment is possible.¹² Since the effects of these phenomena on the elevated-temperature mechanical properties were not known, carbon additions (of 0.2 to 0.25 w/o) in the base alloy were studied.

The tensile properties of the rolled and annealed 0.2 w/o carbon alloys are shown in Figures 51 and 52. At 400°C, a very large increase in tensile strength resulted from the carbon addition. Above 400°C, the strength fell rapidly until at 800°C there was

ULTIMATE TENSILE STRENGTH VS. TEMPERATURE FOR
Th-5 w/o U-La ALLOYS

FIGURE 45

YIELD STRENGTH VS. TEMPERATURE FOR
Th-5 w/o U-La ALLOYS

FIGURE 46

EFFECT OF LANTHANUM CONTENT ON
ULTIMATE TENSILE STRENGTH OF Th-5 w/o U-La ALLOYS

FIGURE 47

EFFECT OF LANTHANUM CONTENT ON
YIELD STRENGTH OF Th-5 w/o U-La ALLOYS

FIGURE 48

a. Th-5 w/o U-2 w/o La Alloy

b. Th-5 w/o U-5 w/o La Alloy

c. Th-5 w/o U-10 w/o La Alloy

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
FOR Th-5 w/o U-La ALLOYS

FIGURE 49

THORIUM-CARBON PHASE DIAGRAM⁹

FIGURE 50

ULTIMATE TENSILE STRENGTH VS. TEMPERATURE FOR THORIUM-BASED ALLOYS CONTAINING CARBON ADDITIONS

FIGURE 51

YIELD STRENGTH VS. TEMPERATURE FOR
THORIUM-BASED ALLOYS CONTAINING CARBON ADDITIONS

FIGURE 52

virtually no improvement over that of the base alloy. At elevated temperatures, this result is not unusual for many metals that are sensitive to the presence of interstitial impurities.

A comparison of the Larson-Miller curves for the carbon-containing alloy and for the base alloy is presented in Figure 53. The addition of carbon improved the stress-rupture properties of the base alloy. The most logical explanation for this improvement is the presence of a thorium carbide dispersion. Thorium carbide in the proper size and distribution could be expected to improve the resistance of the grain boundaries to flow.

In order to observe the effects of the dispersion on microstructure, a series of heat-treating and aging studies were performed. Figure 54 shows the effects of aging temperature on microstructure and hardness.

In general, the improvement in the elevated-temperature mechanical properties was very pronounced at 400°C, but was only slight at temperatures above 600°C. At 400°C and below, the alloys could be appreciably hardened by aging.

G. Thorium-Uranium-Zirconium Alloys with Carbon Additions

Since strengthening by both solid-solution and dispersion hardening was only partially successful with the base alloy plus carbon additions, another attempt was made using zirconium in solid solution. An initial zirconium content of 5 w/o in the base alloy was chosen, and additions of 0.1 to 0.5 w/o carbon were made.

Typical structures for the various carbon contents are shown in Figure 55. The size of the major zirconium carbide particles formed increased as the carbon content increased. Also, a noticeable change in the appearance of the grain boundaries occurred at the various levels of carbon concentration.

A decrease in microhardness was observed with increasing carbon contents. This indicated that the carbon probably reduced the amount of zirconium that had been in solid solution by the formation of a coarse dispersion of zirconium carbide. Likewise, tensile testing of the Th-5 w/o U-5 w/o Zr-0.2 w/o C alloy showed a decrease in strength when compared to the Th-5 w/o U-0.2 w/o C ternary alloy, a result of the removal of the interstitial carbon. Figures 51 and 52 show that the zirconium and carbon additions together produced an alloy weaker than if either were added separately, except at high temperatures where the Th-U-C alloy suffers a drastic reduction in strength. Aging tests were not successful. On the basis of the early results, work on this alloy system

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
FOR Th-5 w/o U BASE ALLOY WITH CARBON ADDITIONS

FIGURE 53

- a. Air Cooled from 800°C;
Aged 2 Hours at 400°C; DPH = 192.

- b. Air Cooled from 800°C;
Aged 2 Hours at 600°C; DPH = 149.

- c. Air Cooled from 800°C;
Aged 2 Hours at 800°C; DPH = 175.

EFFECT OF AGING TEMPERATURE ON MICROSTRUCTURE
AND HARDNESS OF Th-5 w/o U-0.2 w/o C ALLOY

(500X)

FIGURE 54

a. 0.1 w/o C.

b. 0.2 w/o C.

c. 0.3 w/o C.

d. 0.5 w/o C.

MICROSTRUCTURES OF Th-5 w/o U-5 w/o Zr ALLOY
WITH CARBON ADDITIONS
(500 $\times$)

FIGURE 55

was discontinued. The coarse size and distribution of the zirconium carbide prevented it from acting as a dispersion hardener. The affinity of zirconium for carbon removed their beneficial solution-hardening effects.

H. Thorium-Uranium Base Alloy with Aluminum Additions

1. Rolled and Annealed Structure

As an alternative method to dispersion hardening by carbides, a series of alloys containing aluminum additions were prepared. Aluminum forms intermetallic compounds with both thorium and uranium. Both binary phase diagrams (thorium-aluminum and uranium-aluminum) indicate that there is a very small range of solid solubility for aluminum in thorium or uranium (less than 0.5 w/o Al). Through solution-heat-treatment and aging processes, development of an age-hardened dispersion-structure alloy was attempted. Aluminum additions of 0.1 to 1.0 w/o were made to the Th-5 w/o U base alloy. The 1 w/o aluminum alloy formed an extensive network of intermetallic compound throughout the ingot. As a result, the ingot could not be rolled. The alloy containing 0.1 w/o aluminum was easily fabricated, having fabrication properties similar to those of the Th-5 w/o U base alloy. The 0.2 and 0.3 w/o aluminum alloys were easily fabricated. In the rolled and annealed condition, these alloys showed increasing strength as the aluminum content increased. The 0.5 w/o aluminum alloy was not easily fabricated and could not be rolled to the size required for physical testing even at 750°C. Metallographic examination showed that this alloy contained a large amount of intermetallic compound as shown in Figure 56. Apparently 0.5 w/o was the upper limit for aluminum additions that would produce alloys capable of being fabricated with ease.

The tensile properties of the alloys are shown in Figures 57 and 58 and the stress-rupture properties in Figure 59. As shown, no appreciable gains in the elevated-temperature properties resulted from the aluminum additions.

2. Solution-Heat-Treated Structure

Since these alloys are a dispersion type, heat-treatment and aging studies were performed. The alloys were solution heat treated for 1 hour at 1000°C, water quenched, and subsequently aged up to 4 hours at temperatures of 400, 600, and 800°C. Results for all of the alloys studied indicated that there was only a slight hardness increase,

a. Th-5 w/o U-0.1 w/o Al.

b. Th-5 w/o U-0.5 w/o Al.

c. Th-5 w/o U-1 w/o Al.

AS-CAST MICROSTRUCTURES OF Th-5 w/o U-Al ALLOYS
(500 $\times$)

FIGURE 56

EFFECT OF ALUMINUM CONTENT ON
ULTIMATE TENSILE STRENGTH OF Th-5 w/o U-Al ALLOYS

FIGURE 57

EFFECT OF ALUMINUM CONTENT ON
YIELD STRENGTH OF Th-5 w/o U-Al ALLOYS

FIGURE 58

a. Th-5 w/o U-0.1 w/o Al Alloy

b. Th-5 w/o U-0.2 w/o Al Alloy

c. Th-5 w/o U-0.3 w/o Al Alloy

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
 FOR Th-5 w/o U-Al ALLOYS

FIGURE 59

this being from aging at 800°C. Metallographic examination, shown in Figure 60, illustrated no noticeable change in structure resulting from heat-treatment and aging processes. Apparently the major aluminum intermetallic particles were formed during melting and were quite stable regardless of the subsequent heat treatments.

I. Alloying Systems Using the Th-10 w/o U Base Alloy

The alloy systems discussed thus far have been based on a 5 w/o U content. Studies of these alloys resulted in the development of suitable techniques for melting, fabricating, and processing of the Th-U materials. Furthermore, these studies indicated what particular alloying constituents and principles were most effective in improving elevated-temperature mechanical properties. Based on this information, a Th-10 w/o U base alloy and ternary alloys with 10 w/o uranium were investigated. It was the purpose of this investigation to determine the effects of an increased uranium content on the properties of the alloys developed earlier in the program. Accordingly work was generally limited to the most promising alloys already developed.

1. Th-10 w/o U Base Alloy

A procedure similar to that employed in evaluating the Th-5 w/o U base alloy was followed. The results of tensile testing of the Th-10 w/o U alloy in the rolled and annealed condition are shown in Figure 61; the properties of this alloy are very similar to those of the Th-5 w/o U alloy (cf. Figure 10). Apparently the higher uranium content had little effect on the elevated-temperature tensile properties of Th-U alloys. Testing of the Th-10 w/o U alloy showed that the stress-rupture strength of this alloy was essentially the same as that of the Th-5 w/o U alloy in the rolled and annealed condition (Figure 62). Examination of the alloy microstructure showed that a noticeably greater amount of a second-phase constituent was present throughout the matrix. This second phase appeared to be uranium and should be expected due to the increase in uranium content.

In an attempt to decrease the amount of secondary phase, the alloy was subjected to the same heat-treatment process given the Th-5 w/o U alloy (3½ hours at 1350°C, followed by quenching with flowing helium gas). With the Th-5 w/o U alloy, a nearly single-phase structure resulted. However, this type of structure could not be produced with the Th-10 w/o U alloy. An extensive series of heat treatments were performed over the temperature range of 1200 to 1400°C for times varying from ½ hour

a. Water Quenched Only.

b. Aged 2 Hours at 400°C.

c. Aged 2 Hours at 600°C.

d. Aged 2 Hours at 800°C.

EFFECT OF AGING TEMPERATURE ON
MICROSTRUCTURES OF Th-5 w/o U-0.3 w/o Al ALLOY
HEAT TREATED 2 HOURS AT 1000°C AND WATER QUENCHED
(500×)

FIGURE 60

TENSILE PROPERTIES VS. TEMPERATURE FOR
Th-10 w/o U BASE ALLOY

FIGURE 61

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
 FOR Th-10 w/o U BASE ALLOY

FIGURE 62

to 15 hours. In all tests, a duplex structure resulted. Generally, the flowing-helium-gas quench was used; however, some experiments were performed using a water quench. The structures obtained were the same for both quenching media. The microstructure in Figure 63 resulted from heat treating for $3\frac{1}{2}$ hours at 1350°C and was typical of those produced over the temperature range of 1250 to 1400°C . Below 1250°C , the alloy formed some grain-boundary phase as well as the dispersed secondary phase.

As with the Th-5 w/o U base alloy (Figure 10), the Th-10 w/o U alloy showed only a slight improvement in tensile properties resulting from the heat treatment and aging (Figure 61). However, the heat-treatment and aging processes very noticeably improved the stress-rupture properties of the alloy (Figure 62).

The reasons for the beneficial effects resulting from heat treatment are associated with the change in structure. Although there was still a large amount of the secondary uranium phase dispersed throughout the matrix, it was primarily an intragranular type and, therefore, should have little effect on the nature of the grain boundaries. The boundaries appeared to be generally clean, and the grain size had been greatly increased. Both of these conditions were conducive to improving stress-rupture properties.

In summary, the Th-10 w/o U alloy showed a slightly greater improvement in stress-rupture properties when heat-treated and aged than did the Th-5 w/o U alloy (cf. Figures 10 and 61). However, the general effect of heat treatment and aging was the same for both alloys.

2. Columbium and Molybdenum Additions

The Th-10 w/o U base alloy studies indicated that a considerable amount of secondary uranium phase existed throughout the alloy structure. Attempts at putting this phase into solid solution by heat treatment were only partially successful. After heat treatment, there was still a noticeable amount of large uranium particles present throughout the matrix. It appeared desirable, therefore, to strengthen this uranium phase by alloy additions. The metals columbium and molybdenum were selected, because they have no solubility in thorium and are well known for their improvement of the radiation stability of uranium at high temperatures. Both alloys were evaluated primarily for tensile and stress-rupture properties over the temperature range of 600 to 800°C .

a. Annealed.

b. Heat Treated.

c. Heat Treated and Aged.

EFFECT OF HEAT TREATMENT AND AGING ON
THE MICROSTRUCTURE OF THE Th-10 w/o U ALLOY
(250X)

FIGURE 63

a. Th-10 w/o U-2 w/o Cb

Figure 64 shows the results of tensile tests on a Th-10 w/o U-2 w/o Cb alloy in the rolled and annealed condition. At 600°C, the columbium addition noticeably improved the properties of the Th-10 w/o base alloy as shown in Figure 64. There was also some improvement in the ultimate tensile strength up to 800°C. This alloy was also tested after undergoing solution heat treatment and aging; the only significant improvement was at 800°C.

The columbium additions were quite effective in improving stress-rupture properties as shown in Figure 65. It was apparent that both the 600 and 800°C stress-rupture properties were much improved by the columbium additions. Unlike the tensile tests, improvement resulted from heat treatment and aging at both 600 and 800°C.

The microstructure of the Th-10 w/o U-2 w/o Cb alloy is shown in Figure 66. In the rolled and annealed condition, it was very similar to the Th-10 w/o U alloy. In the heat-treated condition, however, a large portion of the matrix had some intergranular U-Cb phase present. This is generally an undesirable structure; however, in the columbium alloy the intergranular phase appeared to have sufficient strength during stress-rupture testing to resist the deformation forces that sometimes lead to a rapid intergranular failure.

b. Th-10 w/o U-1.5 w/o Mo

The Th-10 w/o U-1.5 w/o Mo alloy displayed properties generally similar to those of the Th-10 w/o U-2 w/o Cb alloy in the rolled and annealed condition (cf. Figures 67 and 64). No large increase in ultimate tensile strength was noted at 600°C, but improvements in tensile properties at 800°C were significant. The stress-rupture properties of the rolled and annealed alloy with the molybdenum addition are shown in Figure 68. The improvement in stress-rupture life was similar to that for the Th-10 w/o base alloy with columbium additions (cf. Figure 65). Attempts at solution heat treatment were unsuccessful because of a partial melting of the alloy; therefore, no comparative data for this condition were obtained.

The structure of this alloy in the annealed condition and the structure resulting from attempted solution heat treatment are shown in Figure 69. Since this Th-10 w/o U-1.5 w/o Mo alloy exhibited properties similar to the Th-10 w/o U-2 w/o Cb alloy in the annealed condition, and since it cannot be solution heat treated, it was considered inferior to the columbium alloy for the purposes of this investigation.

TENSILE PROPERTIES VS. TEMPERATURE FOR
Th-10 w/o U-2 w/o Cb ALLOY

FIGURE 64

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
 FOR Th-10 w/o U-2 w/o Cb ALLOY

FIGURE 65

a. Annealed.

b. Heat Treated.

c. Heat Treated and Aged.

EFFECT OF HEAT TREATMENT AND AGING ON
THE MICROSTRUCTURE OF THE Th-10 w/o U-2 w/o Cb ALLOY
(500×)

FIGURE 66

TENSILE PROPERTIES VS. TEMPERATURE FOR
Th-10 w/o U-1.5 w/o Mo ALLOY

FIGURE 67

LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
 FOR Th-10 w/o U-1.5 w/o Mo ALLOY

FIGURE 68

a. Annealed.

b. Heat Treated.

MICROSTRUCTURE OF Th-10 w/o U-1.5 w/o Mo ALLOY
(250 $\times$)

FIGURE 69

3. Aluminum and Beryllium Additions

The effects due to dispersed intermetallic phases in the Th-10 w/o U alloy were briefly investigated. Aluminum and beryllium are known to form intermetallic compounds with uranium and thorium. Since the Th-5 w/o U base alloy had shown some improved properties with the addition of 0.3 w/o aluminum, the Th-10 w/o U alloy with a 0.3 w/o aluminum addition was investigated. The tensile and stress-rupture tests results obtained show that little difference exists between the two base alloys with a 0.3 w/o aluminum addition.

The addition of beryllium to the Th-10 w/o U alloy had a slightly more beneficial effect than did the addition of aluminum. Alloys with additions of 0.1 w/o and 0.3 w/o beryllium were investigated. Increasing the beryllium content beyond the 0.1 w/o level had little effect. The most significant improvements were in stress-rupture properties, although the tensile properties were also somewhat improved. However, the improvements in properties of the alloys with dispersed secondary phases were only slight compared to the improvements from solid-solution hardening.

4. Zirconium Additions

In the investigation of alloy systems based on a 5 w/o uranium content, it was shown that zirconium additions were the most effective hardener, and 2 and 5 w/o zirconium were the most effective levels of alloy-metal concentration. Therefore, zirconium additions of these concentrations were used in the investigation of alloys containing 10 w/o uranium. The same general methods developed earlier in the program to produce and evaluate alloys were employed. The investigation of tensile and stress-rupture properties was carried out over the temperature range of 600 to 800°C.

a. Rolled and Annealed Structure

The effects of zirconium additions on the elevated-temperature tensile properties of Th-10 w/o U followed a slightly different trend than did zirconium additions to Th-5 w/o U. Alloys in the rolled and annealed (i.e., not heat treated) condition of the Th-10 w/o U alloy did not show the added improvement in strength when the zirconium content was increased from 2 w/o to 5 w/o. This is shown clearly by Figures 70 and 71.

EFFECT OF HEAT TREATMENT AND AGING ON
ULTIMATE TENSILE STRENGTH OF Th-10 w/o U-Zr ALLOYS

FIGURE 70

EFFECT OF HEAT TREATMENT AND AGING ON
YIELD STRENGTH OF Th-10 w/o U-Zr ALLOYS

FIGURE 71

b. Solution-Heat-Treated Structure

As with the 5 w/o alloys, heat treating and aging improved the tensile properties of the Th-U-Zr alloys containing 10 w/o uranium (Figures 70, 71, 72, and 73). Although the general effects of solution heat treatment and aging on the tensile properties and structure were much the same as they were for the similar alloys with 5 w/o uranium, the specific properties displayed by these alloys were somewhat lower than the optimum properties displayed by the 5 w/o uranium alloys.

The microstructures of the Th-10 w/o U-Zr alloys in various conditions are shown in Figures 74 and 75. It should be noted that when solution heat treated, these ternary alloys did not possess as many of the large uranium particles as the binary Th-10 w/o U alloy did. The structures of these zirconium alloys resembled those of the Th-5 w/o U alloys with Zr additions with the same heat treatments (cf. Figures 74 and 75 with Figures 25 and 26, respectively). With the addition of zirconium, the solubility of uranium in thorium at the 1350°C heat-treating temperature appeared to have been increased, thus allowing the alloy to form a more uniformly fine dispersion of the secondary uranium phase throughout the matrix of the material.

The stress-rupture properties of the alloys containing 10 w/o uranium followed generally the same pattern as did the alloys with 5 w/o uranium. In the rolled and annealed condition, the zirconium additions improved rupture life at 600°C but not 800°C. This loss in rupture properties at 800°C for the Th-10 w/o U-Zr alloys was much less than for the Th-5 w/o U-Zr alloys (cf. Figure 76 and Figure 17), but the trend of behavior was the same. Heat treating the alloys greatly improved the stress-rupture life of both alloys at all test temperatures, as can be seen in Figure 76. As with the base alloys containing 5 w/o uranium, the best stress-rupture life was obtained in the alloy containing 2 w/o zirconium. However, the actual rupture times for the Th-10 w/o U alloys with zirconium additions were slightly less than those achieved by the similar Th-5 w/o U alloys when tested under identical conditions.

In general, the properties of the heat-treated Th-10 w/o U-Zr alloys developed during this investigation were slightly lower than exhibited by the Th-5 w/o U-Zr alloys. This result was attributed to the limited solubility of uranium in thorium.

TENSILE PROPERTIES VS. TEMPERATURE FOR
Th-10 w/o U-2 w/o Zr ALLOY

FIGURE 72

TENSILE PROPERTIES VS. TEMPERATURE FOR
Th-10 w/o U-5 w/o Zr ALLOY

FIGURE 73

a. Annealed.

b. Heat Treated.

c. Heat Treated and Aged.

EFFECT OF HEAT TREATMENT AND AGING ON
THE MICROSTRUCTURE OF Th-10 w/o U-2 w/o Zr ALLOY
(500X)

FIGURE 74

a. Annealed.

b. Heat Treated.

c. Heat Treated and Aged.

EFFECT OF HEAT TREATMENT AND AGING ON
THE MICROSTRUCTURE OF Th-10 w/o U-5 w/o Zr ALLOY
(500x)

FIGURE 75

a. Th-10 w/o U-2 w/o Zr Alloy

b. Th-10 w/o U-5 w/o Zr Alloy

**LARSON-MILLER PARAMETER VS. MINIMUM-CREEP-RATE STRESS
FOR Th-10 w/o U-Zr ALLOYS**

FIGURE 76

As the concentration of uranium increased, the amount and size of secondary uranium or uranium-rich phases were also expected to increase. Nevertheless Th-10 w/o U-Zr alloys developed during this investigation demonstrated a very marked improvement in elevated-temperature mechanical properties over the Th-10 w/o U binary alloy.

c. Aging Studies

The structural stability of these alloys was investigated to determine the effects of aging treatments. The alloys were solution heat treated, given various aging treatments, and then tensile tested. The results of this study are presented in Table V. Although the 800°C tensile properties of the alloys varied somewhat depending on the aging treatment, test specimens appeared to retain the initial strength resulting from the standard heat-treatment and aging procedures. The variation in results tabulated in Table V is attributed to experimental error resulting from process variables. An examination of microstructures showed no significant alterations resulting from the long-term heat treatments.

TABLE V
EFFECTS OF AGING TREATMENTS
ON THE THERMAL STRUCTURAL STABILITY
OF HEAT-TREATED Th-10 w/o U-Zr ALLOYS

Aging Conditions	Ultimate Tensile Strength *	Yield Strength *
	(psi)	(psi)
<u>Th-10 w/o U-2 w/o Zr Alloy</u>		
Reference: 2 hours at 850°C	9,015	7,870
100 hours at 800°C	10,500	8,900
170 hours at 850°C	8,450	7,725
2 hours at 850°C plus 200 hours at 800°C	10,550	9,230
<u>Th-10 w/o U-5 w/o Zr Alloy</u>		
Reference: 2 hours at 850°C	7,900	7,350
100 hours at 800°C	9,250	7,975
170 hours at 850°C	11,150	7,330
2 hours at 850°C plus 200 hours at 800°C	8,000	6,650

* Testing temperature was 800°C.

SUMMARY OF RESULTS

1. The elevated-temperature mechanical properties of thorium were improved to essentially the same extent with the addition of 5 w/o uranium as with the addition of 10 w/o uranium.

2. All alloys studied in this investigation exhibited some structure sensitivity with respect to size and distribution of the uranium dispersion. A solution heat treatment and aging was developed to control the dispersion of uranium and other phases to obtain optimum mechanical properties. Even though the rolled and annealed structures of most of the alloys tested exhibited mechanical properties superior to those of the binary Th-U base alloys, the improvement was further enhanced by the solution heat treatment and aging.

3. Zirconium as an Alloying Element

a. Zirconium was found to be the most effective additive for improving the elevated-temperature mechanical properties of both the Th-5 w/o U and the Th-10 w/o U base alloys.

b. A solution heat treatment for $3\frac{1}{2}$ hours at 1350°C with a quench in flowing helium followed by an aging of 2 hours at 850°C substantially improved the elevated-temperature mechanical properties of the Th-U-Zr alloys at all temperatures tested.

c. In a heat-treated condition, optimum 800°C tensile properties were obtained with a 5 w/o zirconium addition.

d. In the heat-treated condition, the optimum 800°C stress-rupture properties were obtained with a 2 w/o zirconium addition.

e. The solution heat treatment effectively eliminated an acute structure sensitivity at 800°C found in the Th-5 w/o-Zr alloys. This type of structure sensitivity was not found in the Th-10 w/o U-Zr alloys.

f. The standard aging treatment of 2 hours at 850°C produced a thermally stable structure in the Th-10 w/o U-Zr alloys tested. The Th-5 w/o U-Zr alloys tested required a longer aging time for a completely stable structure.

g. The optimum mechanical properties of the Th-5 w/o U-Zr alloys were slightly higher than the optimum mechanical properties of the Th-10 w/o U-Zr alloys.

4. Columbium Additions

- a. Addition of 2 w/o columbium to the Th-10 w/o U base alloy improved mechanical properties in both the rolled and annealed condition and in the heat-treated and aged condition.
- b. Addition of 1 w/o columbium to the Th-5 w/o U-Zr alloys eliminated their 800°C structure sensitivity in the rolled and annealed condition. In the heat-treated and aged condition, the addition of 1 w/o columbium to the Th-5 w/o U-5 w/o Zr alloy was of little consequence.

5. Molybdenum Additions

Addition of 1.5 w/o molybdenum to the Th-10 w/o U alloy improved the mechanical properties in the rolled and annealed condition.

6. Carbon Additions

- a. Addition of 0.2 and 0.25 w/o carbon to the Th-5 w/o U base alloy only slightly improved the elevated-temperature mechanical properties above 600°C. Below 600°C it was very effective.
- b. Heat treatment and aging of these alloys containing carbon resulted in a hardened structure at 300 and 400°C; but at higher temperatures, aging was either deleterious or ineffective as a hardening treatment.
- c. Addition of carbon to the Th-5 w/o U-5 w/o Zr alloy resulted in the formation of a coarse, ineffective, zirconium carbide dispersion. The tensile strength of this alloy was lowered through the apparent removal of some zirconium from solid solution.

7. Aluminum Additions

- a. Addition of 0.3 w/o aluminum to the Th-5 w/o U base alloy resulted in some slight improvement in tensile properties.
- b. All aluminum additions up to 0.3 w/o improved the 600°C stress-rupture properties slightly but did not appreciably change the 800°C properties of the base alloy.
- c. Heat-treatment and aging processes had no significant effects on the hardness of the base alloy with aluminum additions.

8. Yttrium as an Alloying Element

Addition of yttrium to the Th-5 w/o U base alloy improved the elevated-temperature mechanical properties.

a. Tensile properties were improved only slightly by the addition of 2 to 10 w/o yttrium but were improved significantly by addition of 15 w/o yttrium.

b. Stress-rupture properties were improved significantly by addition of 2 w/o yttrium but were affected only slightly by additions of 5 to 15 w/o yttrium.

9. Lanthanum as an Alloying Element

Additions of up to 10 w/o lanthanum to the Th-5 w/o U base alloy were only slightly effective in improving tensile properties and were of no significance in improving stress-rupture properties.

RECOMMENDATIONS

Irradiation tests should be conducted immediately to assess the full potential of the best alloys developed in this program. The most promising alloys are the heat-treated Th-10 w/o U-2 w/o Zr and Th-10 w/o U-5 w/o Zr, respectively. These alloys exhibit the same tensile strengths at 800°C as the binary Th-10 w/o U alloys exhibit at 650°C.

The irradiation tests conducted to date at 550² and 650°C¹ on the binary Th-10 w/o U alloys have demonstrated good resistance to dimensional changes during burnup as high as 30,000 mwd/tonne. Now, with a fuel alloy available that maintains the 650°C strength of the binary alloy at 800°C, radiation stability at high temperatures to high burnups should be attainable.

It is recommended that a series of capsule irradiations be subjected to burnups of 10,000, 20,000 and 30,000 mwd/tonne. The specimen temperatures should be selected to produce swelling that would clearly define the upper temperature limit of these fuels. On the basis of the tensile tests conducted during this program and the slight swelling observed at 550 and 650°C in binary-irradiation programs, it is recommended that specimen surface temperatures of 650 and 700°C should be tested with axial temperatures of 750 and 850°C.

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APPENDIX A

TENSILE TESTING RESULTS

Note: A = Fabrication Method A.

B = Fabrication Method B.

* = Solution heat treated and aged.

Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) (× 10 ⁶)
High-Purity Thorium							
1117A	A	Room	10,900	21,700	32.0	65.5	8.2
1122B	A	100	12,450	22,900	32.5	55.0	2.0
1118A	A	200	5,020	11,600	31.5	67.0	0.15
1128A	A	200	7,150	18,700	35.5	67.0	1.5
177D	A	200	2,800	14,350	31.5	37.0	0.44
1118B	A	300	4,875	16,000	27.0	60.5	0.46
1119A	A	400	3,375	11,500	30.0	59.0	0.40
111A	A	400	7,440	16,000	34.8	30.5	0.37
129A	A	400	8,250	15,500	43.4	34.8	1.7
129B	A	400	5,500	14,850	33.9	27.6	0.69
177C	A	400	4,620	14,600	26.3	22.7	0.49
132B	A	500	3,770	11,140	46.4	57.9	2.9
1119B	A	500	5,100	13,200	57.0	68.0	3.7
11B	A	600	4,650	8,750	47.2	46.5	0.69
130A	A	600	4,550	8,380	63.0	43.5	0.29
130B	A	600	4,920	8,730	70.1	48.6	0.44
177A	A	600	2,240	6,560	58.0	41.2	0.27
177B	A	600	4,870	9,250	62.2	43.0	0.38
132A	A	700	2,550	5,470	80.4	34.9	0.16
1120B	A	700	2,340	4,850	42.5	92.0	0.15
1128B	A	800	688	1,105	16.0	96.0	0.92

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
High-Purity Thorium (cont.)							
1122A	A	800	1,930	2,930	39.0	88.0	0.15
131B	A	800	1,038	2,555	55.7	84.8	0.61
131A	A	800	2,400	3,155	42.9	88.0	0.28
Commercial Grade Thorium							
134B	A	200	5,740	17,130	29.1	54.3	1.6
133A	A	400	10,950	16,250	30.6	33.0	0.27
134A	A	600	5,160	8,700	62.0	49.2	0.16
136A	A	600	5,380	8,550	63.9	42.0	0.21
135A	A	700	4,330	5,590	58.0	47.8	0.28
133B	A	800	2,140	2,885	40.2	87.0	0.52
Th-5 w/o U Alloy							
218B	B	200	9,600	20,810	14.9	53.8	1.6
223A	B	200	9,355	20,645	16.7	51.6	1.3
226B	B	200	8,957	21,275	43.4	54.4	2.9
219A	B	300	10,605	19,165	-	-	-
222B	B	300	7,420	19,760	21.7	51.2	1.5
219A	B	300	10,605	19,165	14.4	46.0	1.6
52	A	400	10,250	21,100	22.4	24.7	0.54
86	A	400	9,050	21,600	19.3	24.8	0.13
226A	B	400	6,695	17,675	21.0	52.0	1.3
494A	A	400	9,450	22,165	27.0	53.2	1.3
78	A	500	9,780	17,750	35.3	31.8	0.61

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U Alloy (cont.)							
79	A	600	7,900	12,500	41.4	28.2	0.60
225B	B	600	5,425	11,060	48.7	73.3	2.2
497	A	600	8,065	11,440	48.7	72.8	0.34
866	A	700	4,380	6,240	57.5	75.0	0.87
85	A	700	3,880	6,250	56.9	48.2	0.66
494B	A	700	3,320	5,150	77.0	70.0	1.0
90	A	800	2,850	3,930	40.2	67.8	1.1
224B	B	800	1,140	2,325	22.6	80.0	1.2
220A	B	800	1,420	2,130	51.5	65.0	0.34
496	A	800	3,325	3,930	41.0	43.4	0.12
156A	A	800	3,180	4,000	36.0	24.0	0.87
354A	*	600	8,700	13,800	33.0	40.0	0.11
355B	*	600	7,875	14,200	33.5	41.5	0.19
352B	*	700	3,420	5,850	40.0	32.0	1.0
353A	*	700	6,200	8,150	41.0	31.0	0.046
353B	*	700	5,550	7,300	52.0	28.0	0.071
156A	*	800	3,180	4,000	36.0	24.0	0.87
218A	*	800	760	1,400	22.0	7.0	0.0095
352A	*	800	835	1,950	30.0	6.0	0.042
612B	*	800	3,470	4,255	32.0	39.6	1.5
612A	*	800	2,700	3,400	15.0	39.0	0.35

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-2 w/o Zr Alloy							
182B	A	200	14,700	27,500	20.6	14.5	0.32
184B	A	300	11,800	28,600	21.5	14.2	0.59
71	A	400	-	28,200	16.8	17.2	-
237B	B	400	9,950	25,875	18.4	36.0	1.9
503A	A	400	14,250	29,660	20.0	39.3	0.81
72	A	500	11,450	27,100	23.6	22.5	0.17
516A	A	500	14,016	29,100	29.5	50.8	1.4
73	A	600	9,940	17,800	7.9	10.8	1.7
183A	A	600	11,477	19,367	38.4	63.7	0.51
333A	B	600	9,330	17,830	35.6	62.3	0.84
501B	A	600	11,420	19,578	44.7	67.7	1.7
182A	A	700	9,830	12,400	30.5	17.3	0.77
502B	A	700	10,000	12,340	30.1	61.2	1.3
185A	A	800	5,820	7,930	1.29	1.76	0.32
333B	B	800	Failed on application of load				
236B	B	800	"	"	"		
456B	B	800	"	"	"		
501A	A	800	"	"	"		
502A	A	800	"	"	"		
516B	A	800	"	"	"		
457B	B	800	"	"	"		
458B	B	800	"	"	"		
337A	*	600	17,200	23,200	16.0	55.0	0.40
633B	*	600	12,500	22,000	27.0	48.0	0.085
765A	*	600	15,500	22,800	31.5	51.5	2.0

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) 10 10 ⁶
Th-5 w/o U-2 w/o Zr Alloy (cont.)							
834A	*	600	16,000	23,100	30.5	45.0	2.7
838B	*	700	13,480	16,960	28.0	45.0	1.3
458A	*	700	9,000	12,350	21.5	36.0	0.14
613B	*	700	9,250	11,900	25.0	6.0	0.73
456B	*	700	10,400	13,300	53.0	51.0	0.049
765B	*	700	13,000	17,300	31.0	53.0	2.6
767A	*	700	12,200	15,350	23.0	60.0	1.1
459A	*	800	-	2,340	-	-	-
459B	*	800	8,500	11,300	4.0	10.0	0.59
767B	*	800	11,200	13,380	14.0	65.0	1.5
834B	*	800	11,190	12,900	29.0	39.0	1.7
336A	*	800	No yield	7,200	No deformation		
336B	*	800	"	8,825	"		
634A	*	800	10,870	11,670	23.5	27.0	0.52
838A	*	800	3,055	3,655	4.0	17.0	0.29
860A	*	800	7,530	8,200	19.5	26.5	0.67
861A	*	800	7,250	7,700	5.0	7.0	0.93
861B	*	800	5,150	6,900	4.0	9.0	0.59
862A	*	800	6,500	7,450	6.0	8.0	0.72
862B	*	800	7,585	8,350	16.5	20.0	0.77
860B	*	800	4,475	6,125	1.0	5.5	1.4
1037B	*	800	4,325	4,425	21.0	41.0	2.0
1093A	*	800	8,450	10,100	7.5	19.5	3.7
837B	*	900	6,530	7,280	17.5	36.0	0.50

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-5 w/o Zr Alloy							
181B	A	200	16,700	37,500	20.4	16.0	1.0
339A	B	200	12,675	31,200	17.2	47.5	1.5
178A	A	300	16,900	35,700	14.4	17.7	1.4
74	A	400	16,450	36,900	19.3	18.8	3.3
517B	A	400	18,016	37,085	19.0	35.6	2.1
178B	A	450	18,250	37,900	18.0	18.0	0.64
232B	B	450	12,908	30,680	22.0	32.2	1.4
76	A	500	17,500	34,300	21.5	21.5	4.2
179A	A	550	11,900	30,100	25.9	25.7	0.44
75	A	600	-	22,000	28.7	32.6	1.5
232A	B	600	11,120	18,260	43.8	60.6	0.68
267A	B	600	10,710	19,185	32.6	67.4	2.0
505B	A	600	12,730	21,200	36.9	65.0	2.0
172B	A	700	12,650	15,350	48.2	34.7	0.37
172A	A	800	9,950	10,800	63.5	41.4	0.63
339B	B	800	Failed on application of load - < 1,000 psi UTS				
338A	B	800	"	"	"	"	"
228B	B	800	"	"	"	"	"
448A	A	800	"	"	"	"	"
342A	*	600	13,900	22,000	34.0	36.0	1.4
269B	*	600	16,200	24,500	14.0	21.0	0.12
802B	*	600	14,170	20,650	38.0	35.0	0.56
755A	*	600	18,480	26,810	29.0	37.0	1.4
754B	*	600	19,225	27,230	32.0	56.8	1.6
265B	*	700	13,300	15,900	44.0	47.0	0.40

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-5 w/o Zr Alloy (cont.)							
651B	*	700	14,300	18,650	17.0	29.0	0.99
269A	*	700	12,700	17,600	19.0	53.0	0.091
805A	*	700	15,800	18,970	24.0	32.0	1.3
839A	*	700	11,135	14,040	13.0	22.5	0.71
842A	*	700	13,730	17,065	21.5	30.0	1.1
265B	*	700	13,300	15,900	44.0	47.0	0.40
841B	*	800	10,000	12,430	34.5	39.0	1.4
805B	*	800	11,730	13,230	42.5	51.5	0.90
340B	*	800	-	1,120	-	-	-
651A	*	800	8,500	12,200	19.0	7.0	1.0
341A	*	800	No yield	9,450	No deformation		0.028
614B	*	800	9,400	9,500	0.4	8.0	
461B	*	800	No yield	3,770	No deformation		0.44
462C	*	800	7,150	10,300	-	8.0	
652B	*	800	9,000	11,800	17.0	7.0	0.083
805B	*	800	11,730	13,230	42.5	51.5	0.90
859A	*	800	12,460	14,115	30.5	40.0	0.83
859B	*	800	7,445	9,400	23.0	29.0	1.5
863B	*	800	4,800	11,800	22.5	50.0	3.0
863A	*	800	6,550	7,800	32.0	54.0	0.69
864A	*	800	6,500	8,150	50.0	62.0	0.64
864B	*	800	10,400	12,050	25.0	69.0	3.5
865A	*	800	13,300	13,900	28.5	68.8	0.29
842B	*	900	6,045	7,275	6.5	25.0	0.60

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) (× 10 ⁶)
Th-5 w/o U-10 w/o Zr Alloy							
189B	A	200	36,000	53,250	11.5	7.0	4.6
344A	B	200	28,085	44,965	13.7	37.3	2.0
189A	A	300	39,400	50,600	11.5	14.0	2.7
173B	A	400	34,500	53,750	13.8	11.4	1.0
519A	A	400	30,149	48,985	14.0	38.7	2.3
346A	B	400	26,200	43,000	11.0	7.0	1.6
173A	A	500	28,000	43,250	16.5	14.9	3.8
174A	A	600	20,700	23,700	53.3	37.7	0.51
519B	A	600	19,236	21,450	61.0	90.4	1.2
344B	B	600	15,300	18,800	84.0	79.0	0.70
174B	A	700	13,050	13,850	56.2	30.6	0.23
536B	A	700	9,836	11,065	90.0	98.7	0.31
345A	B	700	9,000	10,400	59.0	94.0	0.24
190A	A	800	9,950	10,600	49.1	24.9	0.34
346B	B	800		Failed by excessive yield			
536A	A	800	4,185	4,730	70.0	41.0	0.59
343B	B	800	3,500	4,250	250.0	80.0	0.096
759	*	600	7,750	12,800	30.0	52.0	0.41
758B	*	600	11,075	14,765	34.0	42.3	1.2
760B	*	600	22,860	25,785	15.5	28.6	2.0
758A	*	600	21,010	24,115	17.0	33.0	1.7
562A	*	700	12,960	15,890	10.5	39.4	0.94
655A	*	700	12,900	15,200	19.0	6.5	0.094
207B	*	800	5,630	7,400	50.0	17.0	0.033
347B	*	800	9,200	10,900	23.0	37.0	0.084

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-10 w/o Zr Alloy (cont.)							
561B	*	800	7,445	9,000	60.5	90.5	0.50
654A	*	800	8,050	9,900	42.0	11.0	0.071
190B	*	800	7,850	9,000	48.0	77.0	2.0
347A	*	800	10,140	12,450	3.0	14.0	0.98
Th-5 w/o U-15 w/o Zr Alloy							
349A	B	200	37,275	65,745	10.3	25.5	55.0
193A	A	400	39,800	53,750	10.5	8.1	2.0
175	A	500	33,200	45,750	56.2	19.5	0.96
194B	A	500	34,455	40,340	22.4	66.4	1.6
197C	A	500	39,652	51,130	18.3	55.7	3.6
193B	A	600	20,554	21,750	54.0	43.2	0.68
348B	B	600	18,300	18,850	40.0	22.0	1.1
192A	A	700	11,350	11,800	46.2	41.9	0.39
349B	B	700	9,850	11,000	72.0	75.0	0.81
192B	A	800	6,500	7,210	132.0	63.4	0.26
466B	A	800	5,180	5,775	41.1	13.9	0.18
466A	A	800	6,940	7,650	156.0	91.2	0.37
Th-5 w/o U-2 w/o Y Alloy							
160B	A	200	14,350	27,700	17.1	8.5	0.56
261B	B	200	10,340	24,090	23.2	51.7	0.56
382A	B	200	9,790	22,380	23.0	57.1	1.2
444A	B	200	13,545	26,770	24.6	50.3	1.3
159A	A	400	15,600	24,400	22.2	27.7	0.31

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-2 w/o Y Alloy (cont.)							
161A	A	400	10,465	22,965	19.2	43.1	2.4
374A	B	400	9,150	19,875	24.7	64.7	2.0
160A	A	600	8,750	12,200	51.5	32.5	0.33
161B	A	600	7,325	11,690	46.4	67.1	1.8
374B	B	600	5,650	9,350	48.6	77.5	0.70
375A	B	700	3,830	5,235	49.1	85.9	1.5
482B	A	700	7,370	8,220	51.0	56.0	0.61
159B	A	800	3,910	4,690	24.9	23.0	0.16
261A	B	800	2,300	2,630	38.0	47.0	0.44
382B	B	800	1,440	2,365	40.9	45.5	0.48
381A	B	800	1,435	1,910	34.9	50.1	0.64
444B	B	800	2,575	3,676	39.4	46.3	0.28
Th-5 w/o U-5 w/o Y Alloy							
163A	A	200	20,084	27,426	21.2	20.3	0.49
377A	B	200	12,045	24,155	19.3	52.8	3.6
386A	B	200	9,630	23,980	26.0	49.5	2.6
445A	B	200	10,630	25,945	19.2	40.0	2.0
484A	A	200	17,000	28,850	19.2	40.0	0.92
162B	A	400	10,200	23,000	23.0	29.6	0.92
164A	A	400	-	22,880	21.7	44.0	0.35
378A	B	400	8,980	19,185	22.6	52.0	1.3
262B	B	400	6,851	18,680	20.9	48.5	1.9
262A	B	400	7,230	18,596	27.0	57.2	4.3
162A	A	600	8,150	12,350	51.5	43.6	0.28

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-5 w/o Y Alloy (cont.)							
164B	A	600	7,010	11,880	58.3	70.5	0.97
378B	B	600	6,795	10,450	69.2	80.4	1.4
379A	B	700	4,190	5,560	42.4	84.1	1.4
486B	A	700	6,250	7,900	50.0	23.0	0.022
163B	A	800	4,468	4,936	22.6	21.7	0.045
377B	B	800	1,975	2,460	31.0	34.3	0.85
386B	B	800	2,170	3,390	33.6	41.3	0.57
387A	B	800	2,585	3,240	34.0	38.3	0.86
445B	B	800	2,860	3,810	33.0	37.5	0.55
484B	A	800	4,665	5,080	24.2	33.4	0.60
Th-5 w/o U-10 w/o Y Alloy							
67	A	200	7,759	24,051	13.6	9.5	0.43
263B	B	200	8,540	21,425	26.1	49.1	2.5
438A	A	200	10,068	24,350	23.0	45.9	3.5
166A	A	400	9,260	21,100	17.4	20.2	0.15
167A	A	400	9,250	21,375	22.6	46.2	0.20
166B	A	600	6,840	12,300	43.0	40.9	0.33
167B	A	600	6,070	11,540	32.8	61.5	5.4
489B	A	700	4,800	6,900	60.0	25.0	0.043
165	A	800	3,480	5,000	18.9	16.1	0.22
263A	B	800	2,265	3,335	28.3	34.3	0.56
438B	A	800	3,320	4,590	29.0	33.2	0.59
439A	B	800	2,850	3,820	28.3	26.7	1.9

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-15 w/o Y Alloy							
169A	A	200	16,000	27,300	21.9	24.2	1.0
443A	B	200	10,140	23,855	24.9	50.2	1.8
442	B	200	9,870	23,465	23.7	56.0	1.2
190A	A	200	14,780	26,870	18.0	41.7	1.6
168B	A	400	12,650	23,600	24.1	23.5	0.70
170A	A	400	13,925	21,900	23.2	51.5	0.47
168A	A	600	8,580	14,750	41.3	39.3	0.47
170B	A	600	7,120	14,150	42.3	66.1	1.1
492B	A	700	7,160	9,670	46.0	67.0	0.99
169B	A	800	5,590	6,190	10.9	13.3	0.13
264A	B	800	2,460	3,890	17.7	30.4	1.9
264B	B	800	2,175	3,260	12.3	25.7	0.64
443B	B	800	3,692	4,740	21.0	25.8	0.98
490B	A	800	5,000	5,445	25.0	26.6	7.8
Th-5 w/o U-2 w/o La Alloy							
300A	B	200	13,145	23,065	20.3	54.8	3.4
113	A	400	6,550	18,300	23.3	23.1	0.13
664	A	400	9,000	19,900	25.0	51.5	1.1
526	A	400	15,334	22,715	18.5	43.3	0.82
637	A	600	5,650	12,200	27.0	23.0	0.28
114	A	600	8,370	12,500	49.4	32.8	1.1
527	A	700	4,965	5,390	35.0	46.5	0.32
300B	B	700	3,333	5,465	36.3	80.6	0.87
432	A	800	2,695	3,700	31.2	32.6	0.33
638	A	800	2,800	3,560	34.0	51.0	0.075

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-5 w/o La Alloy							
274A	B	200	14,310	26,265	20.0	51.8	3.7
115B	A	400	12,550	21,100	17.0	18.4	0.65
116A	A	400	-	21,300	25.0	19.7	-
639	A	400	10,800	22,600	24.0	51.0	0.11
640	A	600	4,650	9,400	47.0	35.0	0.13
116B	A	600	9,200	12,850	41.5	36.3	0.89
117A	A	600	5,720	11,150	45.9	39.1	0.44
117B	A	600	8,600	11,890	37.9	38.3	0.20
274B	B	700	4,365	6,340	48.2	75.7	0.76
529	A	700	5,725	7,140	40.0	74.0	0.38
641	A	800	3,570	4,390	17.5	23.0	0.82
433	A	800	3,390	4,030	19.1	23.7	1.2
Th-5 w/o U-10 w/o La Alloy							
118	A	400	13,100	25,300	19.3	13.3	2.8
120B	A	400	11,000	21,900	15.0	22.9	0.88
643	A	400	12,040	23,610	19.5	33.0	2.8
644	A	600	9,050	13,420	28.0	53.0	1.2
119	A	600	9,400	13,450	73.7	24.7	0.57
120A	A	600	8,090	12,350	31.7	29.6	0.55
532	A	700	5,963	7,445	46.0	58.5	0.56
645	A	800	3,530	4,275	8.0	11.4	0.78
437	A	800	3,975	4,375	14.2	19.2	1

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-2 w/o Zr- 1 w/o Cb Alloy							
629A	A	400	13,200	31,600	18.0	49.0	0.70
556A	A	600	10,815	19,520	49.0	72.0	1.5
556B	A	700	9,360	11,800	46.0	83.0	1.3
555A	A	800	5,910	6,790	15.0	10.5	0.35
723B	*	600	10,700	21,800	30.0	45.0	1.9
806B	*	600	13,915	22,170	26.5	56.0	3.4
809A	*	600	11,430	20,990	42.5	39.0	3.2
979B	*	600	12,600	22,000	27.0	37.5	2.4
722	*	700	6,300	13,400	35.0	30.0	3.9
808A	*	700	13,885	15,700	13.5	47.0	1.4
809B	*	700	10,160	12,680	26.5	70.5	1.2
924B	*	700	11,250	14,375	52.0	63.0	2.0
724A	*	800	7,975	9,150	50.0	21.0	1.2
808B	*	800	9,445	10,080	40.5	60.0	1.3
868B	*	800	5,860	7,140	17.0	35.0	1.4
925B	*	800	7,500	8,650	38.0	62.5	0.59
Th-5 w/o U-5 w/o Zr- 1 w/o Cb Alloy							
604A	A	400	18,400	39,200	13.0	60.0	0.42
559B	A	600	16,245	23,755	52.0	70.0	1.0
558B	A	700	11,935	14,320	48.5	93.0	0.82
559A	A	800	5,880	6,425	4.0	7.0	0.23
727	*	600	19,850	27,100	27.0	59.0	1.5

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-5 w/o Zr- 1 w/o Cb Alloy (cont.)							
810A	*	600	15,230	23,095	18.0	25.5	1.1
844B	*	600	18,580	28,175	28.0	49.0	1.7
923A	*	600	13,195	24,365	18.0	22.5	1.4
619B	*	700	14,650	19,800	14.0	6.0	0.32
811B	*	700	11,550	13,725	48.0	69.0	0.44
844A	*	700	14,235	18,115	17.5	58.5	1.1
845A	*	700	16,200	20,495	29.5	45.5	0.91
630B	*	800	5,550	12,200	34.0	58.0	0.70
810B	*	800	10,975	12,925	17.5	52.0	1.2
845B	*	800	9,100	10,900	29.5	76.0	3.3
Th-5 w/o U-C Alloys							
326B - 0.2 w/o C	B	400	15,835	25,980	29.1	63.7	0.79
512B - 0.2 w/o C	A	400	23,600	37,800	38.0	47.0	0.59
327A - 0.2 w/o C	B	600	8,370	13,740	48.3	80.2	1.7
510B - 0.2 w/o C	A	600	11,045	14,515	105.0	76.0	0.36
415A - 0.2 w/o C	B	600	7,330	11,800	43.9	70.7	0.57
511A - 0.2 w/o C	A	700	6,820	7,675	46.0	74.0	0.74
327B - 0.2 w/o C	B	800	2,330	3,020	43.6	57.8	1.3
511B - 0.2 w/o C	A	800	3,690	4,155	47.0	41.0	0.096
415B - 0.2 w/o C	B	800	2,845	3,700	39.5	44.1	1.0
416A - 0.2 w/o C	*	600	6,200	11,000	44.0	33.0	5.1
417A - 0.2 w/o C	*	700	5,525	7,050	49.0	16.0	0.33
616A - 0.2 w/o C	*	700	7,050	10,600	34.0	35.0	0.54
416B - 0.2 w/o C	*	800	3,100	3,400	63.0	48.0	0.67

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-C Alloys (cont.)							
554B - 0.25 w/o C	A	600	8,900	10,930	92.0	77.6	0.84
554A - 0.25 w/o C	A	700	3,170	5,185	67.0	75.0	3.0
553A - 0.25 w/o C	A	800	3,200	4,080	51.0	46.3	0.22
Th-5 w/o U-5 w/o Zr-C Alloys							
249A - 0.1 w/o C	B	500	11,200	29,017	27.7	46.2	2.4
249B - 0.1 w/o C	B	700	7,840	11,800	63.5	88.7	0.94
548A - 0.2 w/o C	A	400	14,850	33,470	18.5	25.0	1.0
251B - 0.2 w/o C	B	500	12,930	27,450	26.8	69.4	0.85
548B - 0.2 w/o C	A	600	12,265	21,510	45.0	60.0	1.8
550A - 0.2 w/o C	A	700	9,620	12,550	39.0	46.4	0.76
251A - 0.2 w/o C	B	700	8,312	11,515	66.4	87.0	0.72
254B - 0.3 w/o C	B	500	13,620	27,830	24.8	68.5	0.96
254A - 0.3 w/o C	B	700	8,395	10,700	19.4	25.1	0.67
255B - 0.5 w/o C	B	400	10,000	21,030	18.5	30.4	0.76
255A - 0.5 w/o C	B	500	8,000	10,480	22.0	36.5	0.81
259B - 0.5 w/o C	B	700	6,820	9,365	24.6	41.8	1.0
Th-5 w/o U-Al Alloys							
707A - 0.1 w/o Al	A	400	11,800	27,170	26.0	51.0	1.9
632A - 0.1 w/o Al	A	600	7,400	11,700	37.0	19.0	0.078
709B - 0.1 w/o Al	A	600	7,450	12,900	57.5	81.5	5.3
707B - 0.1 w/o Al	A	700	3,515	5,590	63.0	83.0	0.49
710A - 0.1 w/o Al	A	700	4,650	6,125	65.0	87.0	0.93

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-5 w/o U-Al Alloys (cont.)							
632B - 0.1 w/o Al	A	800	1,920	2,500	72.0	57.0	0.14
710B - 0.1 w/o Al	A	800	3,370	3,900	37.5	65.0	1.1
712A - 0.2 w/o Al	A	600	7,957	11,756	21.5	72.0	8.6
858A - 0.2 w/o Al	A	700	6,020	8,065	39.0	79.0	0.57
1098A - 0.2 w/o Al	A	700	4,190	6,125	-	89.5	0.70
712B - 0.2 w/o Al	A	800	2,957	3,875	50.0	59.0	0.71
920A - 0.3 w/o Al	A	600	7,985	13,310	51.0	74.0	1.1
920B - 0.3 w/o Al	A	700	2,475	3,530	76.0	83.0	0.29
994A - 0.3 w/o Al	A	700	5,400	7,375	64.0	83.0	0.74
921 - 0.3 w/o Al	A	800	4,345	5,245	87.0	75.0	1.1
Th-10 w/o U Alloy							
701A	A	400	11,750	23,430	17.5	47.0	1.1
1100A	A	500	10,500	17,500	39.5	63.5	1.4
703B	A	600	8,245	12,825	54.5	70.0	0.92
701B	A	700	4,510	6,400	48.5	82.5	0.96
579	A	800	2,825	3,410	60.0	54.0	0.25
704B	*	600	9,580	13,820	24.5	48.0	1.2
846A	*	700	5,290	7,660	24.5	58.5	1.1
704A	*	800	1,765	2,370	11.0	46.5	1.7
706B	*	800	1,510	2,000	31.5	64.0	0.90
855	*	800	3,435	3,780	7.0	12.5	2.7
929A	*	800	2,050	2,480	24.5	53.0	1.5

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-10 w/o U-2 w/o Cb Alloy							
851A	A	600	3,900	10,285	37.0	68.0	0.98
993A	A	600	6,625	19,875	34.0	44.1	0.94
931A	A	700	5,510	8,200	35.0	77.0	1.3
851B	A	800	3,140	4,380	51.0	86.0	0.58
848A	*	600	8,925	15,415	11.0	24.0	0.99
992B	*	600	14,200	15,900	13.0	32.8	1.5
849A	*	700	5,840	7,080	31.0	49.0	1.0
848B	*	800	2,140	2,785	25.5	63.5	0.30
932B	*	800	4,550	5,690	30.0	63.5	0.51
Th-10 w/o U-1.5 w/o Mo Alloy							
939A	A	600	8,600	14,170	40.0	71.0	0.65
944B	A	600	7,300	14,600	39.0	64.0	0.72
939B	A	700	5,815	8,410	68.0	79.5	0.78
940A	A	800	3,575	4,450	60.0	80.0	0.63
Th-10 w/o U-2 w/o Zr Alloy							
933A	A	600	13,290	21,210	38.0	66.0	1.2
968B	A	600	14,250	23,750	31.0	21.6	0.82
987A	A	600	10,650	17,900	5.5	69.0	0.99
933B	A	700	2,040	3,470	32.5	50.0	1.8
935B	A	700	9,840	11,440	39.0	46.0	0.77
934A	A	800	5,380	6,250	38.5	38.0	0.28
988A	A	800	5,875	6,500	41.5	49.0	0.62
967A	*	600	15,830	22,650	31.5	48.5	1.6

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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-10 w/o U-2 w/o Zr Alloy (cont.)							
967B	*	700	13,575	16,280	18.5	32.0	1.3
968A	*	800	7,870	9,015	17.0	32.0	1.2
986B	*	800	8,900	10,500	26.0	46.5	3.5
988B	*	800	7,725	8,450	10.5	46.0	2.6
987B	*	800	9,230	10,550	36.0	93.5	0.75
Th-10 w/o U-5 w/o Zr Alloy							
973A	A	600	12,200	18,800	42.0	69.2	1.7
974B	A	600	8,180	16,740	34.0	33.4	6.3
989B	A	600	14,400	19,800	16.0	64.5	1.1
936B	A	700	9,065	10,970	125.0	89.5	0.86
936A	A	800	5,495	6,230	99.0	92.0	0.33
937A	A	800	5,190	6,280	150.0	91.0	0.42
971B	*	600	18,050	24,365	22.5	32.0	1.6
972A	*	700	7,400	14,500	10.0	19.0	3.7
974A	*	700	13,600	16,700	5.5	11.0	0.65
972B	*	800	7,350	7,900	24.0	39.1	0.00034
989A	*	800	7,975	9,250	14.5	28.7	0.34
990	*	800	7,330	11,150	17.0	33.9	1.2
1104B	*	800	6,650	8,000	11.5	24.0	3.1

APPENDIX A
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Specimen Number	History	Test Temperature (°C)	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elongation (%)	Reduction Area (%)	Young's Modulus (psi) ($\times 10^6$)
Th-2 w/o Zr Alloy							
1048A	A	Room	14,400	29,600	23.5	55.5	9.3
1048B	A	100	12,100	20,500	23.5	55.5	1.9
1051B	A	100	15,900	29,900	21.0	57.5	2.0
1051A	A	100	15,900	29,400	22.5	50.0	1.7
1052A	A	200	-	28,000	21.0	43.5	-
1052B	A	200	13,100	28,850	25.5	45.0	1.6
1053A	A	300	14,400	29,900	25.0	52.7	1.7
1053B	A	400	13,600	29,100	20.0	48.0	5.0
1049A	A	500	13,100	27,100	29.0	52.5	1.9
1049B	A	600	12,700	20,600	41.0	69.0	1.3
1050A	A	700	11,550	14,200	49.5	78.0	6.6
1050B	A	800	7,300	8,000	8.5	19.4	0.30
Th-5 w/o Zr Alloy							
1056A	A	Room	16,800	34,700	9.0	52.0	7.2
1057A	A	100	15,600	33,400	20.0	44.2	1.8
1056B	A	200	16,000	33,300	19.0	22.8	7.6
1057B	A	300	14,700	33,200	18.0	39.0	1.6
1058A	A	400	19,700	35,000	19.0	48.0	1.7
1058B	A	500	18,000	31,200	32.0	52.0	9.6
1126B	A	500	14,150	23,900	41.0	61.0	1.0
1083A	A	600	14,500	20,250	34.0	42.0	0.75
1084A	A	700	11,100	13,400	59.0	89.0	7.0
1159A	A	800	5,500	5,600	1.0	2.0	0.35

APPENDIX B

STRESS-RUPTURE TESTING RESULTS

Note: A = Fabrication Method A

B = Fabrication Method B

* = Solution heat treated and aged

Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U Alloy						
222A	B	600	8,000	28	9,360	2.2
227B	B	600	10,000	18	12,400	0.3
495A	A	600	8,000	2.9	8,850	0.6
223B	B	700	2,000	0.019	2,225	357.1
227A	B	700	5,000	7.6	6,000	0.9
224A	B	800	1,000	0.43	1,148	164.6
220B	B	800	2,000	17	2,180	3.0
495B	A	800	1,000	0.38	1,240	90.2
158A	A	800	2,000	1.0	2,080	4.4
1091A	A	600	6,000	0.082	6,950	69.1
1091B	*	600	10,000	1.4	10,600	2.1
351A	*	600	8,000	0.041	8,160	62.9
355A	*	800	2,000	0.0050	2,020	95.1
Th-5 w/o U-2 w/o Zr Alloy						
233B	B	500	25,000	0.70	29,275	0.5
186A	A	600	8,000	9.6	8,102	113.0
335A	B	600	8,000	0.022	9,000	214.0
515A	A	600	8,000	0.16	9,100	68.0
234A	B	600	10,000	0.052	11,400	47.5
334A	B	600	15,000	5.3	16,700	0.9

APPENDIX B
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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U-2 w/o Zr Alloy (cont.)						
185B	A	600	15,000	0.48	22,634	1.1
186B	A	700	2,000	0.030	2,569	408.0
235B	B	700	8,000	0.22	8,300	0.6
233A	B	800	1,000	8.3	1,008	0.1
237A	B	800	1,000	6.6	1,005	<0.1
515B	A	800	1,000	Instant failure upon loading		
1036A	A	800	1,000	4.0	1,280	2.1
1036B	A	800	2,000	9.5	2,110	0.6
1092A	A	800	1,000	3.8	1,420	3.4
1092B	A	800	2,000	5.9	2,610	1.3
836A	*	600	16,000	0.16	16,525	118.8
635A	*	600	20,000	Elong. 0.321"; failed in 0.03 hr		
633A	*	600	16,000	0.065	16,340	2.0
766A	*	600	16,000	0.058	16,130	43.2
459B	*	800	1,000	Test stopped at 0.034" elong., 23.7 hr		
613A	*	800	4,000	44	4,337	63.2
634B	*	800	4,000	0.016	4,090	49.7
635B	*	800	6,000	0.64	6,170	2.7
766B	*	800	6,000	0.22	6,285	12.8
835B	*	800	4,000	0.014	4,145	143.4
836B	*	800	4,000	0.021	4,035	97.7
837A	*	800	6,000	0.15	6,145	14.3

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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-- Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U-5 w/o Zr Alloy						
266A	B	500	15,000	0.0022	17,900	214.0
229A	B	500	25,000	2.2	29,792	1.9
181A	A	500	32,000	8.3	49,166	0.4
228A	B	550	10,000	0.023	10,894	219.4
180B	A	600	8,000	0.076	8,721	25.5
180A	A	600	15,000	5.2	16,942	1.1
229B	B	600	6,000	0.016	6,355	164.5
266B	B	600	8,000	0.0066	8,500	42.0
267B	B	600	8,000	0.0068	8,820	51.5
230A	B	700	6,000	1.1	6,730	3.5
265A	B	700	5,000	0.52	5,380	6.7
338B	B	700	2,000	Test stopped at 0.100" elong. , 405 hr		
248A	B	800	1,000	1.6	1,030	0.6
1094A	A	800	1,000	3.1	1,460	10.9
1094B	A	800	2,000	13	2,260	1.1
840A	*	600	16,000	0.33	16,500	11.9
1038A	*	600	20,000	8.2	22,000	0.5
615B	*	600	32,000	Elong. 0.390"; failed in 0.03 hr		
652A	*	600	16,000	1.3	16,703	18.8
753A	*	600	16,000	0.16	17,045	22.6
804A	*	600	16,000	0.19	16,750	20.0
804B	*	800	6,000	2.6	6,140	1.7
839B	*	800	4,000	1.0	4,215	7.3
841A	*	800	6,000	2.8	6,140	1.6
840B	*	800	4,000	1.0	4,200	5.0

APPENDIX B
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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U-5 w/o Zr Alloy (cont.)						
341B	*	800	1,000	0.11	1,004	3.7
461A	*	800	2,000	24	2,028	0.1
463B	*	800	4,000	1.9	4,262	2.3
755B	*	800	4,000	1.0	4,330	6.7
756B	*	800	6,000	3.2	6,210	1.5
753B	*	800	4,000	-	-	3.4
757	*	800	6,000	1.6	6,350	2.3
754A	*	800	4,000	0.54	4,210	9.5
756A	*	800	4,000	0.76	4,370	7.4
865B	*	800	2,000	0.99	2,250	88.4
Th-5 w/o U-10 w/o Zr Alloy						
199B	A	400	40,500	0.048	49,700	84.0
200A	A	400	44,000	0.065	52,316	59.5
200B	A	400	48,000	-	-	<0.1
198A	A	500	20,000	0.074	21,450	51.2
198B	A	500	25,000	0.73	26,100	5.9
199A	A	500	30,000	1.7	32,100	1.8
202A	A	550	10,000	2.6	12,611	23.8
191A	A	600	3,000	0.042	3,320	98.5
188B	A	600	4,000	0.090	4,090	39.0
188A	A	600	6,000	0.28	6,110	<17.5
204A	A	600	8,000	0.073	8,450	4.8
205A	A	600	10,000	1.5	10,600	2.6
187B	A	600	12,000	1.4	12,150	1.7

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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U-10 w/o Zr Alloy (cont.)						
206A	A	600	15,000	15	16,450	0.5
187A	A	600	16,000	5.9	16,666	0.6
345B	B	600	8,000	0.91	8,820	4.6
206B	A	700	5,000	4.5	5,230	1.1
207A	A	700	8,000	14	8,300	0.2
203A	A	700	10,000	290	10,580	0.1
201B	A	800	1,000	0.59	1,039	10.5
201A	A	800	2,000	7.9	2,112	0.5
Th-5 w/o U-15 w/o Zr Alloy						
195B	A	500	25,000	0.98	28,420	5.9
194A	A	600	8,000	1.1	8,604	2.3
350A	B	700	2,000	0.39	2,250	9.3
350B	B	800	1,000	1.7	1,040	2.5†
196B	A	800	1,000	0.16	1,500	3.5‡
196A	A	800	1,000	0.063	1,100	7.3
Th-5 w/o U-2 w/o Y Alloy						
383A	B	600	8,000	2.6	10,100	1.7
481A	A	600	8,000	1.3	8,800	5.2
376A	B	600	8,400	2.8	10,630	2.6
483A	A	600	12,000	0.010	12,164	24.5

† Defective.

‡ Failed by excessive elongation.

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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U-2 w/o Y Alloy (cont.)						
383B	B	800	1,000	1.8	1,390	51.2
481B	A	800	1,000	0.076	1,900	221.0
381B	B	800	2,000	1.4	2,040	2.0
482A	B	800	2,000	13	2,160	8.0
623A	A	800	2,000	1.0	2,074	6.9
375B	B	800	3,400	48	3,842	0.1
Th-5 w/o U-5 w/o Y Alloy						
624A	A	600	8,000	0.50	8,655	12.2
485A	A	600	8,000	11	8,250	9.0
388A	B	600	8,000	1.7	10,060	4.0
379B	B	600	8,400	1.9	9,750	2.8
388B	B	800	1,000	0.16	1,380	63.0
485B	A	800	1,000	2.0	1,200	111.0
389A	B	800	2,000	3.4	2,130	1.4
62	A	800	2,000	1.7	2,260	4.5
624B	A	800	2,000	1.0	2,017	4.8
Th-5 w/o U-10 w/o Y Alloy						
488A	A	600	6,000	0.15	6,765	70.7
489B	A	600	8,000	1.0	9,900	7.1
439B	B	600	8,000	1.2	10,500	6.5
487A	A	800	1,000	0.22	1,240	13.0
487B	A	800	2,000	0.23	2,120	6.2
488B	A	800	2,000	3.0	2,190	2.5
626A	A	800	1,000	0.23	1,235	55.8

APPENDIX B
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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum- Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U-15 w/o Y Alloy						
493A	A	600	6,000	0.023	6,812	330.7
491A	A	600	8,000	0.17	8,700	25.0
491B	A	800	1,000	0.24	1,250	46.5
493B	A	800	1,000	0.16	1,309	60.9
492A	A	800	2,000	2.8	2,160	2.5
628A	A	800	1,000	0.47	1,370	23.7
Th-5 w/o U-2 w/o La Alloy						
419	A	600	8,000	2.0	8,970	5.7
636	A	800	1,000	0.12	1,356	90.8
528	A	800	2,000	3.0	2,087	2.2
Th-5 w/o U-5 w/o La Alloy						
420	A	600	8,000	0.89	8,895	6.9
531	A	800	1,000	0.29	1,222	50.2
530	A	800	2,000	6.3	2,080	2.0
Th-5 w/o U-10 w/o La Alloy						
533	A	600	8,000	1.7	9,024	3.3
534	A	800	1,000	0.88	1,220	13.8
642	A	800	2,000	2.8	2,181	1.9

APPENDIX B

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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U-2 w/o Zr-1 w/o Cb Alloy						
555B	A	600	8,000	0.045	8,650	118.0
557A	A	800	1,000	0.18	1,200	33.4
557B	A	800	2,000	3.8	2,080	3.0
725B	*	600	16,000	1.8	17,450	2.8
807A	*	600	16,000	0.51	16,545	7.8
809A	*	600	16,000	2.9	16,765	0.9
721A	*	600	16,000	1.6	17,200	3.0
925A	*	600	16,000	65	16,550	0.6
979A	*	600	20,000	0.15	20,900	9.0
869B	*	800	4,000	0.60	4,150	4.3
870A	*	800	6,000	4.1	6,130	0.8
807B	*	800	4,000	0.44	4,115	8.7
806A	*	800	6,000	0.57	6,150	4.3
721B	*	800	6,000	6.4	6,240	0.7
723A	*	800	4,000	0.74	4,175	6.4
870B	*	800	4,000	0.35	4,230	13.4
978B	*	800	6,000	6.3	6,450	0.5
Th-5 w/o U-5 w/o Zr-1 w/o Cb Alloy						
980A	A	800	2,000	1.9	2,320	5.5
560A	A	800	1,000	2.0	1,160	2.9
560B	A	600	8,000	0.13	10,200	62.7
558A	A	800	1,000	3.5	1,159	2.8
603A	A	800	2,000	17	2,248	0.5
873A	*	600	16,000	0.15	16,425	13.3

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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U-5 w/o Zr-1 w/o Cb Alloy (cont.)						
726A	*	600	16,000	0.069	16,335	43.3
604B	*	600	16,000	1.2	16,315	2.5
922B	*	600	16,000	0.89	16,550	18.9
981A	*	600	20,000	0.38	20,600	5.4
630A	*	800	4,000	0.14	4,185	21.7
619A	*	800	6,000	3.4	6,245	1.4
726B	*	800	4,000	0.46	4,160	11.1
873B	*	800	6,000	2.1	6,150	2.0
728B	*	800	6,000	1.8	6,250	2.3
728A	*	800	4,000	0.43	4,220	11.5
Th-5 w/o U-C Alloys						
326A - 0.2 w/o C	B	600	8,000	0.95	9,150	7.7
512A - 0.2 w/o C	A	600	8,000	0.85	9,500	9.4
510A - 0.2 w/o C	A	800	1,000	0.69	1,445	160.4
418A - 0.2 w/o C	B	800	2,000	1.8	2,310	2.4
418B - 0.2 w/o C	B	800	2,000	2.0	2,120	3.3
713A - 0.25 w/o C	A	600	8,000	0.19	8,270	31.2
714 - 0.25 w/o C	A	800	2,000	0.22	2,327	24.8
Th-5 w/o U-5 w/o Zr-C Alloys						
258B - 0.2 w/o C	B	600	8,000	0.0076	8,500	113.1
258A - 0.2 w/o C	B	600	8,000	0.021	8,650	112.1
551A - 0.2 w/o C	A	600	8,000	0.044	9,550	105.4

APPENDIX B
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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-5 w/o U-5 w/o Zr-C Alloys (cont.)						
549B - 0.2 w/o C	*	800	2,000	Failed out of heat-treated area		
405A - 0.25 w/o C	B	600	15,000	3.5	16,200	1.6
405B - 0.25 w/o C	B	700	5,000	0.19	5,270	9.6
406A - 0.25 w/o C	B	800	2,000	Instant failure upon loading		
253 - 0.3 w/o C	B	600	8,000	0.012	8,850	212.8
Th-5 w/o U-Al Alloys						
708A - 0.1 w/o Al	A	600	8,000	1.7	8,870	3.1
708B - 0.1 w/o Al	A	800	2,000	4.4	2,190	2.3
709A - 0.1 w/o Al	A	800	1,500	0.15	1,640	17.0
857A - 0.2 w/o Al	A	800	2,000	3.6	2,050	2.9
857B - 0.2 w/o Al	A	600	8,000	1.1	8,335	5.4
858B - 0.2 w/o Al	A	800	1,500	0.99	1,690	14.6
872A - 0.3 w/o Al	A	800	2,000	3.8	2,120	4.5
872B - 0.3 w/o Al	A	600	8,000	0.93	8,585	6.7
994B - 0.3 w/o Al	A	800	1,500	2.4	1,690	6.9
1099B - 0.3 w/o Al	A	800	1,500	2.4	1,690	5.5
Th-10 w/o U Alloy						
977A	A	600	6,000	0.11	7,500	59.9
705B	A	600	8,000	4.5	8,960	1.4
706A	A	800	2,000	4.9	2,060	1.3
928A	A	800	1,000	1.5	1,055	67.3
976B	*	600	10,000	3.9	11,800	1.3
977B	*	600	10,000	1.5	11,100	2.9

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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-10 w/o U Alloy (cont.)						
975A	*	600	8,000	0.14	8,750	20.0
846B	*	600	8,000	0.060	8,440	43.4
856	*	800	4,000	1.6	4,100	0.8
928B	*	800	4,000	6.9	4,160	0.4
929B	*	800	3,000	0.046	3,080	24.7
975B	*	800	4,000	1.3	4,175	1.0
976A	*	800	3,000	0.31	3,120	8.9
Th-10 w/o U-2 w/o Cb Alloy						
930A	A	600	8,000	0.29	8,350	13.8
930B	A	800	2,000	0.21	2,065	18.7
1101A	A	800	3,000	8.9	3,775	0.7
849B	*	600	8,000	0.060	8,380	33.3
932A	*	800	4,000	4.5	4,200	1.0
992A	*	800	3,000	250	3,065	0.7
Th-10 w/o U-1.5 w/o Mo Alloy						
940B	A	600	8,000	4.1	9,350	11.0
941A	A	800	2,000	9.0	2,140	9.0
944A	A	800	3,000	67	3,240	0.7
Th-10 w/o U-2 w/o Zr Alloy						
1102B	A	600	18,000	14	20,900	0.3
934B	A	600	8,000	0.20	9,100	105.0
966B	A	800	1,000	24	1,320	9.3

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Specimen Number	History	Test Temperature (°C)	Initial Stress (psi)	Minimum Creep Rate (in./sec) ($\times 10^{-5}$)	Minimum-Creep-Rate Stress (psi)	Time to Rupture (hours)
Th-10 w/o U-2 w/o Zr Alloy (cont.)						
935A	A	800	2,000	13	2,045	0.6
1103B	*	600	18,000	6.7	19,500	0.4
965A	*	600	16,000	1.1	17,400	20.0
986A	*	600	20,000	23	23,600	0.2
966A	*	800	4,000	1.0	4,250	14.8
965B	*	800	6,000	4.1	6,200	3.4
Th-10 w/o U-5 w/o Zr Alloy						
1104A	A	600	10,000	0.48	10,400	6.2
937B	A	600	8,000	0.30	8,200	15.5
938B	A	800	1,000	22	1,390	7.5
938A	A	800	2,000	80	2,080	1.0
970A	*	600	16,000	3.4	17,400	4.7
991A	*	600	20,000	3.3	21,800	1.5
973B	*	800	6,000	44	6,350	0.5
971A	*	800	4,000	160	4,400	3.3