WADC TECHNICAL REPORT

INTERMEDIATE TEMPERATURE CREEP AND RUPTURE BEHAVIOR OF TITANIUM AND TITANIUM-BASE ALLOY

PART II: INFLUENCE OF MICROSTRUCTURE IN CREEP RUPTURE PROPERTIES

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FOREWORD

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ABSTRACT

An investigation was carried out to establish the relationships between types of microstructure and creep-rupture properties of titanium alloys at 600°F to 1000°F. Chemical composition and the influence of heat treatment was investigated to a limited extent for each type of alloy. The data are fairly complete for some alloys and consist of survey test data for others.

Alpha titanium as represented by commercially pure titanium had the lowest strength. Stable alpha alloys (6% Al and 6% Al - 0.5% Si) and stable beta alloys (30% Mo and 50% V) had similar strengths. These alloys had the highest strength at 800°F and 1000°F where creep was a major factor. The alpha-beta alloys (Ti 150A and Ti 155AX) and the meta-stable beta alloys had high strengths through 600°F and for short time periods at 800°F by virtue of strengthening from transition structures from the beta to alpha transformation.

Compositional differences between alloys had rather small effects in most cases. Properties of the alpha-beta and meta-stable beta alloys were markedly influenced by heat treatment except at 1000°F. The other structures were little influenced by heat treatment. Ductility varied considerably between alloys and with heat treatment and test conditions.

None of the alloys were subject to appreciable creep at 600°F and strength was governed by tensile properties for time periods up to 1000 hours. Increasing tensile strength by cold work or by heat treatment permitted application of high stresses which usually caused immediate failure at 600°F when the stress was above the yield strength. This seemed to be a delayed tensile fracture rather than true creep to rupture. Creep and structural stability were the major factors in strength at 800°F and 1000°F.

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INTRODUCTION

The report presents the results of a survey of the relative creep-rupture properties of titanium alloys in the temperature range of 600° to 1000°F. The major emphasis was placed on the role of the crystal structure developed in the matrix by alloy additions. Alpha titanium, stabilized alpha, alpha plus beta, meta-stable beta and stable beta structures were investigated. Because similar types of crystalline structure can be developed with different alloying elements, two alloys differing in chemical composition were used as examples of each type of structure. This enabled the development of an indication of the relative importance of crystal structure and the alloying elements used to form the structure. Each alloy was further investigated in several conditions of heat treatment so that microstructural variations within each basic structural type could be evaluated.

A previous report (Ref. 1) presented the results of an initial survey limited to one alloy of each type. The data from that report is used in developing the conclusions of this report. The limited survey type tests of Reference 1 were, however, extended for the alloys covered in that investigation for selected heat treatments to cover a wider range of creep-rupture conditions: the object being to delineate properties better. Limited structural studies were made to obtain indications for causes for variations in properties.

ALLOYS AND MICROSTRUCTURES

The specific alloys chosen to represent the various structural types were determined by consultation with representatives of the Materials Laboratory, WADC. The alloys chosen were the following:

1. Alpha titanium: commerciailly pure titanium (Ti 75A)

2. Stabilized alpha: 6% aluminum (6Al); and 6% aluminum + 0.5% silicon (6Al + 0.5 Si) alloys.

3. Alpha + beta: 2.6% Cr - 1.5% Fe (Ti 150A); and 1.3% Cr - 1.5% Fe - 1% Mo - 5% Al (Ti 155AX) alloys.

4. Meta stable beta: 10% molybdenum (10 Mo); and 10% chromium (10 Cr) alloys.

5. Stabilized beta: 30% molybdenum (30 Mo); and 50% vanadium (50 V) alloys.

The Ti75A, 6Al, Ti150A and 30 Mo alloys were previously surveyed for creep-rupture characteristics for Reference 1.

The chemical compositions of the alloys are given in Table I.

The values for carbon, nitrogen and the major alloying elements were obtained from the producers. Analyses of three alloys for oxygen and hydrogen were obtained from the National Research Corporation. The remaining hydrogen values were provided by the Materials Laboratory, WADC.

The Ti75A, Ti150A and Ti155AX alloys were purchased from the Titanium Metals Corporation as bar stock. The other alloys were purchased from the Armour Research Foundation of the Illinois Institute of Technology. These had been melted as approximately 10-pound heats and forged to bar stock. Ingots weighing about 1000 grams were initially produced in a nonconsumable tungsten tip arc furnace and forged to 0.5-inch bar stock. After sand blasting and acid dipping, the bars of several ingots were remelted in a consumable electrode furnace to produce the 10-pound ingots. The ingots were scalped and forged to bar stock.

Because the initial stock of certain alloys was exhausted before all the tests were completed, it was necessary to purchase additional material. The additional stock came from different heats. The compositions reported for the individual heats are given in Table 1.

GENERAL PROCEDURES

The major steps in the investigation were:

1. Establish the heat treatments to produce the desired types of microstructures.

2. Evaluate the tensile, rupture and creep characteristics of the structure.

3. Analyze the influence of microstructural changes during tests.

4. Correlate microstructural and compositional variables with properties.

The data of Reference 1 for Ti75A, 6Al, Ti150A and 30 Mo alloys were reviewed with representatives of the Materials Laboratory, WADC to select the heat treatments of these alloys to be used for more extensive evaluation of properties over a range of stresses. The new alloys added for the investigation covered by this report were studied for the effect of heat treatment on microstructure and a series of heat treatments was then selected for the property evaluation.

The alloys carried over from Reference 1 were subjected to tests selected to provide the following information:

1. Stress-rupture time curves from 100 to 1000 hours.

2. Curves of stress versus time for total deformation of 1-percent out to 1000 hours. Data were included for other total deformations between 0.1 and 5 percent which could be obtained from the creep curves used to establish rupture and 1-percent deformation strengths.

The properties of the new alloys in the investigation were surveyed by the following tests:

1. Tensile tests at room temperature were used to eliminate excessively brittle structures from the program.

2. Tensile tests at elevated temperatures were used to guide the selection of stresses for rupture tests as well as to provide strength data for the structures.

3. The stress for rupture in 100 hours at the elevated temperatures was defined by two rupture tests.

4. One creep test intended to give a creep rate between 10^{-5} and 10^{-6} inch per inch per hour (0.0001 to 0.00001 percent per hour) was used to indicate limited deformation strength.

Structural characteristics were evaluated mainly by microstructural studies and by hardness tests. In some cases x-ray diffraction methods were used to help identify phases in the microstructures. Microstructures were also examined after creep-rupture testing.

EXPERIMENTAL TECHNIQUES

Heat Treatments

All heat treatments were carried out in a purified argon atmosphere. A stainless-steel muffle was kept under a static argon pressure. The argon was passed through heated titanium chips before entering the muffle.

Specimens

Tensile, rupture and creep tests were conducted on standard 0.250-inch diameter specimens with a 1-inch gage length. The specimens were machined after heat treatment.

Strain Measurements

Strain measurements during tensile and creep tests were made with a modified Martens type optical extensometer. Collars were threaded on to the specimens ahead of the pulling adapters.

Extension rods were attached to pins on the collars. Mirrors attached to rollers were mounted between rods attached to the upper and lower extremities of the specimen. Extension of the specimen caused the extension rods to rotate the mirror which reflected an image of scale through an optical beam to a telescope. The readings of two extensometers attached to opposite sides of the specimens were averaged. The sensitivity of the extensometer system was approximately 5 x 10^{-6} inches per inch.

Attachment of the extensometer at the shoulders of the specimens caused them to measure extension in the fillets as well as in the gage section. Correction factors were used to properly account for this effect.

Tensile Tests

Tensile tests were conducted with a hydraulic tensile machine. The load was applied in increments of 5000 psi up to the yield strength to allow strain measurements between increments. When the yield had been exceeded the tests were completed with the load being continuously applied at a free head speed of 0.05-inches per minute.

In tests at elevated temperatures, the specimens were set up in an electric resistance furnace at the test temperatures. Temperature adjustments and the tests were completed within two hours. Temperature variation along the gage length and variation from the nominal test temperature was less than ± 3°F. Thermocouples attached to the center and both extremities of the gage length were used to measure temperatures.

Stress-strain curves were plotted and the yield strength for 0.2-percent offset determined. The usual tensile strength and elongation and reduction of area values were reported.

Rupture and Creep Tests

Rupture and creep tests were conducted in single specimen simple beam loaded units equipped with an automatically controlled electric resistance furnace. The procedure for the rupture and creep tests was the same as the tensile tests with the following additions:

1. Loads were applied through a simple beam. Strain readings were taken between increments of applied load and a stress-strain curve plotted to provide a measure of the initial deformation during loading.

2. Strain measurements were made during the rupture and creep tests with sufficient frequency to provide accurate creep curves. For the lower stress creep curves, strain was measured at least once a day during the 1000-hour tests.

Creep curves including the deformation on loading were plotted and the time for total deformations of interest read from the curves. Minimum creep rates were also measured.

3. When the specimens fractured in the rupture tests, an automatic timer was shut off to record the rupture time. Ordinary elongation and rediction of area were measured after the broken specimens had been removed from the rupture-test units.

Metallographic Examination

Specimens were cut with an Allison high speed glass cutting wheel with continuously circulating coolant. Wet grinding and polishing on wet 600-mesh silicon carbide abrasive paper was used for rough polishing. The finishing operation was usually an electrolytic polish although mechanical polishing was used in some cases.

Electropolishing was carried out with a 10-percent by volume solution of perchloric acid in glacial acetic acid. Polishing times were up to 10 seconds at a current density of about 2.5 amperes per square centimeter.

In mechanical polishing, a 4 to 8 micron diamond compound applied to the back of a sheet of photographic printing paper resting on a glass surface was used for the first step. Final finish was obtained with a water suspension Linde B powder applied to a rotating lap.

Alpha alloys were etched in a solution consisting of 2 parts HF, 2 parts HNO_3 and 100 parts H_2O. Alpha-beta, meta-stable beta and beta alloys were etched in a mixture of 1 part HF and 1 part glycerine.

X-Ray Techniques

Specimens were mechanically polished and etched with HF to remove disturbed surface layers. Diffraction traverses at 2°(2θ) per minute were made with a high angle geiger tube automatic recording spectrometer. Copper radiation at 45 kilovolts and 12 milliamperes was used. The radiation was filtered with two 0.00035-inch nickel filters. Radiation was introduced through Saller slits, the combination used being 4° - 0.006-inch - 4°.

ALPHA TITANIUM

The alloy studied was commercially pure titanium (Ti75A) in the form of 0.5 inch round bar stock from Heat M626 (Table 1). Most of data obtained extended the evaluation of properties over a wider range of stresses at 600°F than was covered by the survey tests of Reference 1. Two treatments, annealed at 1500°F and annealed plus cold rolled to a reduction of 31 percent, were selected for the extended testing. These two treatments gave the extremes of strength found in the survey tests of Reference 1. Tests were limited to 600°F due to the very low strength of the material at higher temperatures.

Survey-type tests were also used to evaluate an acicular type structure developed by quenching in iced brine from 1800°F.

Experimental Material

The stock used for the investigation was from Heat M 626 having the chemical composition given in Table 1. The stock used for the survey tests of Reference 1 from Heat L 984 had been exhausted. Half-inch diameter bars of commercially produced alloy were used.

Annealed and Cold-Worked Structures

The alpha structure of the Ti75A consisted of equiaxed grains (Fig. 1) when annealed at 1500°F. A subsequent cold reduction of 31 percent elongated the grains. There also appeared to be a small amount of a second phase believed to be beta, in the grain boundaries.

Properties of Structures

Tensile Properties

Cold work increased tensile strengths (Table 2) over those of the annealed condition at both 75° and 600°F, with the largest percentage increase at 600°F. The elongation was substantially reduced while reduction of area was changed very little. The tensile properties were similar to those of Reference 1 for Heat L 984.

Rupture Properties

In the annealed condition the rupture strengths at 600°F (Table 3 and 4 and Fig. 2) for 100 and 1000 hours were equal and at most only slightly below the tensile strengths.

When cold worked the rupture strengths at 600°F were higher than for the annealed condition (Table 3 and 4 and Fig. 2). The rupture strengths, however, did decrease with increasing time for rupture so that the 1000-hour rupture strength was only 80 percent of the tensile strength.

The data show that rupture of the cold-worked condition occurred in a very short time after loading when the stress was equal to or greater than the yield strength. This was not true for the annealed condition where rupture did not occur in 1000 hours under a stress equal to 161 percent of the yield strength. This is shown by Figure 3 which superimposes rupture strengths on stress-strain curves from tensile tests for the two conditions.

The cold work reduced elongation and reduction of area in the rupture tests in comparison to the annealed condition. They remained high enough, however, to be more than adequate as judged by ordinary standards. Moreover, they did not decrease in the usual manner with increasing time for rupture.

The stress-rupture time curve for the cold-worked condition had less slope than the curve for the similar treatment in Reference 1. As a result, the longer time rupture strengths were somewhat higher than was indicated by the survey data for the first heat studied (Heat L 984).

Total Deformation and Creep Properties

The data of Table 3 and 4 and Figures 2, 3 and 4 show the following regarding creep of the alpha titanium alloy (Ti75A) at 600°F:

1. When annealed at 1500°F creep was very slight even for stresses within 3000 psi of the tensile strength. Total deformation strengths were governed by deformation on loading. For instance the test under 33,000 psi required 250 hours to reach 3 percent deformation even though the deformation on loading was 2.55 percent.

2. When cold reduced 31 percent after annealing, creep became a major factor in total deformation strength even when stresses were well below the yield strength.

3. The stress for 1 percent total deformation in 1000 hours for the cold-worked condition was 43,000 psi with a deformation of 0.5 percent on loading. The data did not define this strength for the annealed condition other than to show that it would be the same as the stress for 1 percent deformation on loading, i.e., 25,500 psi.

4. The deformation and creep data for the cold-worked condition (Figs. 2 and 4) indicate a higher relative creep resistance for stresses above 57,000 psi than for stresses of 44,000 psi or lower. The results was that much lower 1 percent total deformation strengths were indicated for time periods of 100 to 1000 hours than would be anticipated from the creep-rupture tests. Apparently the cold work was less effective when the stresses were low.

5. The total-deformation strengths considered were substantially higher (Table 4) for the cold-worked than for the annealed condition. It can, however, be inferred from Figures 2 and 4 that the superiority of the cold-worked condition decreases with time and with a decrease in the comparative total deformation. For limited amounts of deformation in very long-time periods cold working would not be of much advantage.

6. Due to the large amounts of deformation during loading to obtain appreciable creep in the annealed condition, considerable difficulty was encountered in conducting the tests. For this reason there are numerous gaps in the reported data in Table 2. Also it was only possible to obtain stress versus time for total-deformation data at odd deformations. There were also inconsistent values between tests which were due to specimen variations.

7. The creep rates measured for both conditions indicated higher creep strengths than were observed for Heat L 984 of Reference 1 in the initial survey. It will be recalled that the tensile tests showed no significant difference. Thus it appears that Heat M 626 used for this report retained strength better under creep conditions than did Heat L 984.

Microstructural Changes During Testing

No evidence of structural changes during creep and rupture testing was found by microstructural examination of several specimens. The only changes were those due to the plastic extension on loading or during subsequent creep.

Acicular Alpha Structures

Experiments were carried out to attempt to develop other microstructures of alpha titanium in the Ti75A material than the equiaxed grains used in the work for Reference 1. These are shown by Figure 5 and may be summarized as follows:

1. Quenching in iced brine from 1700°F did not change the normal equiaxed grain structure. Either the temperature of heating or the rapid cooling nearly suppressed the excess phase which was present in the grain boundaries after slow cooling from 1500°F.

2. An acicular structure was formed by quenching from 1750°F and 1800°F.

3. Slow cooling from 1800°F produced a feathery structure within the alpha grains. Raising the temperature to 2000°F resulted in a coarse basket weave structure.

The material quenched from 1800°F was selected for further testing because it exhibited the greatest deviation in structure and tensile properties from the annealed condition.

Properties of Acicular Structures

The acicular structure developed by quenching from 1800°F had considerably higher tensile and rupture strengths than the condition of annealing from 1500°F (Tables 2 and 4 and Fig. 6). There was no stress dependency for rupture at 600°F while the structure was subject to creep at 800° and 1000°F. Rupture strengths at 800° and 1000°F were higher than for either the annealed or the annealed and cold-worked conditions. Ductility was good to very high in all tests.

The material was subject to marked alteration on properties by quenching from a high temperature. The acicular structure in many respects resembled those formed when beta stabilizing alloys are present. Apparently the transition structure resulted from the high-temperature treatment, extremely drastic quench and possibly a slightly high concentration of beta stabilizing elements such as the iron content of 0.19 percent.

Discussion of Creep-Rupture Characteristics of Alpha Titanium

The general conclusions from the original survey tests at 600°F (Ref. 1) were not changed by the more extensive tests on alpha titanium (Ti75A). The major characteristics at 600°F may be summarized as follows:

1. In the annealed condition there was very little or no indication of time-dependent creep or rupture. Deformations in time periods up to 1000 hours were governed by the deformation on loading.

2. Raising the tensile and yield strengths by cold work or by quenching from high in the beta range permitted the application of much higher stresses than for the annealed condition. This did induce some time dependency for creep and rupture. In particular, rapid failure by rupture occurred when the stresses were above the yield strength. The rupture and total-deformation strengths, however, were much higher than for the annealed condition. When stresses were lowered to the same range as the tensile strength in the annealed condition, creep practically disappeared in the cold-

worked or the quenched materials.

These results suggest that creep only occurs in this material when treatments are used which raise the tensile strength to the point where high stresses can be applied without immediate failure, In the annealed condition the maximum stresses which could be aplied were too low to activate creep at 600°F, even when yielding occurred.

Other factors than raising the tensile strength appeared, however, to be involved. Quenching from 1800°F raised the creep resistance more for stresses in the range of 50,000 psi at 600°F than did 31 percent cold work, even though the cold work raised the tensile strength more. The cold work may have been too extensive and introduced structural instability in the form of recovery from cold works. The results of Reference 1 for higher test temperatures indicated that this could be true. On the other hand, the quench from 1800°F produced a transition structure which could have been inherently more creep resistant as was evidenced by the improvement in strength over the annealed condition at 800°F and 1000°F where cold work was determined.

The Ti75A material is in fact a fairly complex alloy. It is known to contain appreciable amounts of oxygen, nitrogen, carbon, hydrogen and iron. There may be other elements in trace amounts which have not yet been identified as influencing strength. It is known that Ti75A is subject to strain aging (Ref. 2) as would be expected from the small atom alloying elements known to be present. The microstructures observed in this investigation suggest the presence of retained beta and, during quenching, transition structures between beta and alpha which would require the presence of substantial amounts of beta-stabilizing elements such as the iron. It is suggested that the strengthening imparted by strain aging was effective in stopping creep in the annealed condition and was a major factor in strengthening from cold work and from quenching from 1800°F. In both the latter cases strain hardening was probably also a contributing factor. The quenching probably increased solution of odd-sized elements as well as contributing to strength from the known strengthening by retention of beta-to-alpha transition structures.

It can be assumed that the actual performance of the material includes a complex function of the amount and types of alloying elements present, as well as possible strain aging and hardening effects. Evidently some compositional factor was sufficiently different to stabilize strength at 600°F in comparison to early creep results which showed Ti75A in the annealed condition to be subject to creep at 600°F (such as the results of Reference 2).

STABILIZED ALPHA ALLOYS

Certain alloying elements inhibit the transition of alpha titanium to beta titanium upon heating. In addition, the temperature region of co-existence of alpha and beta is narrowed and increased in temperature. Proper amounts of such alloying elements result in an alloy in which it is not possible to retain beta at ordinary temperatures by rapid cooling. Such alloys are termed alpha stabilized alloys. The outstanding element for producing this type of structure is aluminum.

Two alloys, one containing 6 percent aluminum (6Al) and the other containing 6 percent aluminum plus 0.5 percent silicon (6Al + 0.5Si) were used to evaluate the creep-rupture properties of stabilized alpha structures in titanium alloys. Complete stress-rupture time curves and sufficient additional creep tests to establish total deformation strengths were used to define the properties of the 6 Al alloy more completely than the survey tests of Reference 1. Survey type tests on the 6 Al + 0.5 Si alloy were used to check the relative effects of alloy type and composition.

Stabilized Alpha Alloy 6 Al

Tensile, rupture and total deformation characteristics were evaluated for the 6 Al stabilized alpha alloy at 600° and 1000°F. The alloy was evaluated in the as-forged, as-forged plus 12 percent cold reduction, and water quenched from 2025°F conditions. These three treatments provided a survey of the general types of microstructure for the alloy.

Because the material used for the work of Reference 1 had been exhausted, a new heat (designated 4a in Table 1) was used for the evaluation. The 3 inch diameter, 10 pound ingot was forged to 0.5 inch diameter bars from 2100°F.

Microstructures Evaluated

In the as-forged condition, the structure consisted generally of elongated grains. The cold reduction of 12 percent produced an additional slight elongation of the grains. Water quenching after 1 hour at 2025°F in the beta range produced a fine, oriented lamaller structure within the boundaries of the beta grains which existed at 2025°F. It also developed a few areas of enlarged serrated alpha grains.

Tensile Properties

The cold reduction of 12 percent raised tensile strengths and slightly reduced ductility at 75°, 600° and 1000°F in comparison to the as-forged condition (Table 6). The yield strength of the cold-worked condition was raised to a very high level at 600°F and was only slightly below the tensile strength. The quench from 2025°F raised tensile and yield strength at 75°F in comparison to the as-forged condition. There was little difference in properties at 600° and 1000°F.

Rupture Properties

The rupture-test data are given in Tables 7, 8, and 9 and by Figures 7 through 11. There was no time dependency for rupture strength at 600°F for any of the three conditions evaluated (Fig. 7). The only exception to this was the immediate rupture after loading for the material cold reduced 12 percent when the stresses were above the yield strength. As pointed out in the previous section, the cold work raised the yield strength almost to the tensile strength. The other two treatments resulted in material which would not rupture in 1000 hours at 600°F when loaded well above the yield strength and just under the tensile strength. These relationships are shown by Table 10 and Figure 11.

This behavior in rupture tests at 600°F resulted in rupture strength up to 1000 hours being practically the same as the tensile strength for the as-forged material and for the material quenched from 2025°F. For the cold-worked condition it was reduced slightly to about the yield strength.

When tested at 1000°F, the rupture strengths were time dependent for all three conditions (Figs. 8, 9 and 10). The rupture strengths were slightly higher at 100 hours for the material quenched from 2025°F than for the as-forged condition. The cold work reduced strength for 100 and 1000 hours by about 5,000 psi. Consequently, the superior tensile strength of the cold-worked condition was only maintained at 1000°F for short-time periods for rupture.

The test data indicate an unusual stress-rupture time curve (Fig. 9) at 1000°F for the cold-worked material. The test at 20,000 psi had a longer rupture time than was indicated by the higher stress tests. This may be due to variation between specimens or indicates recovery from the damage from cold work after prolonged exposure to 1000°F.

Elongation was high in all the rupture tests at 1000°F. It tended to increase with time for rupture for the as-forged and for the cold-worked conditions while it decreased slightly for the quenched material.

Creep and Total Deformation Characteristics

The deformation during loading controlled total deformation characteristics at 600°F. (See Tables 7, 8 and 9 and Figure 7). Only a slight amount of creep accumulated after the loads were applied for any of the treatments. Because the stress versus time for total deformation curves would be horizontal, tests were not conducted to completely establish such curves. Minimum creep rates were very low for tests well above the yield strength for the as-forged and water-quenched from 2025°F conditions (Fig. 12). The cold-worked condition showed higher creep rates and probably more stress-dependency for creep rate.

Creep was a controlling factor at 1000°F for all three treatments. (See Tables 7, 8 and 9 and Figures 8, 9 and 10). Consequently, stress versus time for total deformation curves were obtained showing marked time dependency for the stress for a limited amount of deformation. All three conditions had curves which tended to flattened out at the longer time periods although this was very slight in the cold-worked condition. Comparative stresses for specific deformations in 100 and 1000 hours (Table 10) show that the as-forged or 2025°F water quenched condition were significantly stronger for 0.5 or 1.0 percent total deformation in 100 or 1000 hours than was the cold-worked condition.

The time period at which third-stage creep started at 1000°F seemed to be nearly independent of stress for stresses below about the 1000-hour rupture strength. This was most marked for the cold-worked condition and least for the as-forged.

The material quenched from 2025°F had the highest strengths at 1000°F based on minimum creep rates (Fig. 12). The cold-worked material was weakest. This was consistent with the trend of the total-deformation strengths.

Structural Changes During Testing

Specimens were examined microstructurally after testing. The material as-forged or water-quenched from 2025°F showed no appreciable change after any of the tests. The as-forged and cold-worked material, however, underwent recrystallization at 1000°F as is shown by Figure 13. There was some evidence of recrystallization in 82 hours at 1000°F. It had progressed to a considerable extent in 187 hours and appeared to be quite complete in 1100 hours.

Discussion

Strengths of the 6 Al alloy at 600°F for time periods up to 1000 hours were dependent on tensile properties. There was little indication of reduction in strength with time by creep. The only unusual feature was the immediate rupture of specimens of the cold-worked condition when loaded above the yield strength. The as-forged material and the material quenched from 2025°F could be stressed far above the yield strength during loading. The 6 Al alloy was similar to the alpha titanium Ti75A in these characteristics.

The cold reduction of 12 percent raised yield strength to a considerable extent. The usual margin of safety from immediate failure derived from the deformation between tensile and yield strength was absent. Thus, unusual care would be necessary not to depend on this feature when designs are based on yield strength. The investigation does not indicate how much cold work can be absorbed by the material before there is no ability to sustain a load above the yield strength. Presumably this would be rather small inasmuch as 12 percent reduction was subject to the effect.

The immediate failure at 600°F after loading above the yield strength in the case cold-worked 6 Al and Ti 75A materials is not a creep failure in the usual sense. It appears to be an unusual phenomena in which an increasing stress can be sustained only provided the stress is increasing, as in a tensile test. If, however, the load is kept constant after yielding starts, failure occurs almost immediately. The basic mechanism involved in this unusual behavior was not identified.

At 1000°F, the load carrying ability was time dependent due to rapid and extensive creep. The creep behavior appeared to be the normal type usually encountered in any material above some limiting temperature. The rupture strengths were reduced and the minimum creep rates increased by cold work. The loss of strength in the cold worked material can be due to recrystallization. The properties of the fine lamaller structure obtained by quenching from 2025°F were suprisingly similar to those of the elongated grain structure of the as-forged condition. The shapes of the total-deformation curves and the third-stage creep characteristics do suggest a structural change at about 1000 hours, although this was not obvious in the microstructures.

The comparative data in Table 10 show that tensile and 100-hour rupture strenths indicate less of a difference between treatments than do the total-deformation strengths. The survey tests of Reference 1 did not indicate that immediate rupture could occur when the cold-worked material was stressed above the yield strength at 600°F. The increase in strength from 12 percent cold work in the present investigation was more than would be anticipated from the data in Reference 1.

Stabilized Alpha 6 Al - 0.5 Si Alloy

The general relationships developed between stabilized alpha microstructures and creep-rupture properties using the 6-percent aluminum alloy were checked using a 6-percent aluminum plus 0.5-percent silicon alloy. The same types of treatments used for the 6 Al alloy to vary the form of the stabilized alpha structure were used for the 6Al + 0.5 Si alloy. A wider difference in chemical composition between the two alloys would have been desirable. However, a suitable alloy was not available at the time.

Experimental Material

The as-forged 0.5-inch diameter bar stock used for the investigation was forged (probably from 1850°F) by the Armour Research Foundation from a 10-pound ingot. The available chemical analyses for the heat are given in Table 1. It should be noted that the titanium sponge used for the Al-Si alloy had a Brinell hardness of 124 while that used for the 6 Al alloy was 148, indicating a possible significantly lower interstitial element content for the Al-Si alloy.

Heat Treatments and Microstructures

Typical photomicrographs after several heat treatments are shown in Figure 14. The as-forged condition had an alpha structure composed of elongated plates. When cold reduced 37-percent there was evidence of elongation of the alpha plates. Quenching from 2100°F resulted in much finer alpha plates. The alpha plates were made considerably more distinct from some grain growth and by the darker etching of the boundaries when the as-forged condition was heated for 24 hours at 1100°F. A quench from 1925°F plus 24 hours at 1100°F caused considerable darkening of the fine plates. Slow cooling from 1925°F to 1100°F and holding for 24 hours developed comparatively coarse plates with a dark etching phase at the boundaries.

Tensile Properties

A major change in tensile properties was only introduced by cold working (Table 11). The pronounced increases in strength were maintained to 1000°F (Fig. 15). The cold work also reduced ductility at 75° and 600°F but not at 1000°F. The other treatments had relatively little effect on properties. The reductions in ductility from exposure to 1100°F were rather small and did not show the embrittlement reported in the literature from titanium silicide precipitation at this temperature.

The as-forged condition was selected for elevated temperature tests as the base material. The cold reduction of 19 percent was included for comparison with the 6 Al alloy. A cold-reduction of 33 percent was used to show the effect of a larger reduction than could be applied to the 6 Al alloy due to cracking of the latter material at larger reductions. The material quenched from 2100°F was used to show the influence of the largest alteration of structure obtained by heat treatment alone.

Rupture Properties

The 6 Al-0.5 Si alloy was not subject to time dependent rupture at 600°F (Table 12 and Fig. 16). The two tests on the as-forged condition which ruptured during loading probably represent variations in strength between specimens of the as-forged material. A test on material cold reduced 19 percent did not rupture in 1000 hours even though the stress was above the yield strength and only 9,500 psi below the tensile strength. The material cold reduced 33 percent fractured in 0.3 hours when the stress was below the yield strength but still only 10,500 psi below the tensile strength. The material quenched from 2100°F did not fracture when tested at a stress above the yield strength.

At 800°F there was a slight indication of time-dependent rupture (Table 12 and Fig. 16) for the as-forged condition. Tests at 800°F were limited to this treatment. The ductility apparently decreased considerably with time for rupture from that in the tensile test.

At 1000°F, the alloy was subject to pronounced decrease in rupture strength with increasing time (Table 12 and Fig. 16). The material cold reduced 19 percent was strongest at 100 hours and the as-forged was strongest at 1000 hours. The material cold reduced 33 percent was weakest for time periods beyond about 10 hours. The rupture curves are not well established. It seems evident that the quench from 2100°F did not produce the strongest material as did a quench from the beta range in the case of the 6 Al alloy. The ductility of the material cold-reduced before testing tended to increase with time for rupture. There was little change for the other materials.

Creep Characteristics

Survey creep tests at 600°F (Table 12 and Fig. 17) indicated that most deformation occurred during loading. Stresses well above the yield strength had to be used to develop creep rates in the range of 10^{-6} inches per inch per hour (except for the material cold reduced 33 percent).

At 800°F the as-forged condition could carry a stress within about 7,000 psi of the yield strength without exceeding 1-percent deformation in 1000 hours. It was, however, considerably more subject to creep than at 600°F.

The alloy was very susceptible to creep at 1000°F (Table 12 and Fig. 17). The survey tests did not show much difference in minimum creep rates between the four treatments. More tests might have shown more slope to the creep curves of Figure 17, particularly for the cold-worked conditions.

Microstructural Changes During Creep-Rupture Testing

Typical microstructures after testing are shown by the photomicrographs of Figure 18. The major change in microstructure occurred in the cold-reduced materialduring testing at 1000°F. It appeared as if recrystallization and/or growth of the alpha plates occurred accompanied by the appearance of considerable dark etching phase at the boundaries after the longer times of exposure. There was little change in the grain structures for any of the treatments after testing at 600°F; or for the as-forged condition when tested at 800°F. Little change in the as-forged or quenched conditions was visible after testing at 1000°F, except that due to specimen deformation during testing.

A dark etching material formed at the boundaries of the alpha particles during testing at all three temperatures. This material was probably titanium silicide.

Discussion

Cold work increased strength to a marked extent, except at 1000°F where the recrystallization type of structural instability introduced by the cold work was detrimental to time-dependent strength. There was little evidence of creep being a factor in strength at 600°F for any treatment considered and was not much of a factor at 800°F for the as-forged condition. In spite of the considerable difference in the form of the alpha in the as-forged and 2100°F-quenched conditions there was remarkably little difference in properties.

Alpha alloys previously studied had been subject to premature failure at 600°F when cold worked and tested above the yield strength. The 6 Al + 0.5 Si alloy seemed to be free from this when cold reduced 19 percent. The material reduced 33 percent was, however, subject to this behavior and even gave some evidence of premature failure for tests below the yield strength. Apparently there is some stress at which rupture time is indefinitely deferred even though a slightly higher stress causes immediate fracture. The cold-work was applied to as-rolled material and there was some evidence that variations in the initial properties of the as-forged material influenced results.

The load carrying ability of the 6 Al + 0.5 Si alloy would be determined by yield characteristics at 600°F. At 800°F creep would be a significant factor (probably a controlling factor for cold-worked material). Creep would be the controlling factor above 800°F for all treatments. No evidence was developed to indicate that the alloy was subject to undue embrittlement in the vicinity of 1000°F. Possibly the low interstitial element sponge used to prepare the alloy reduced this effect from cases of embrittlement for aluminum-silicon alloys reported in the literature.

Comparative Properties of Two Stablized Alpha Alloys

In general the results showed little difference in properties between the 6 Al and 6 Al - 0.5 Si alloys (Table 10 and Figure 19). Any differences between the two alloys were not as large as the differences between the two heats of 6 Al alloy except at 1000°F. The 6 Al + 0.5 Si alloy maintained higher strength in creep-rupture tests at 1000°F after cold work. This was apparently due to a greater resistance to recrystallization-type reactions at this temperature.

The 6 Al + 0.5 Si alloy could be cold worked more than the 6 Al alloy without cracking. Strengths generally increased with the greater reduction (except for rupture strength at 1000°F) so that it can be assumed that the crackingof the 6 Al alloy limited the strengthening by cold work.

A solution quench to produce a fine acicular alpha apparently was more effective in the 6 Al than in the 6 Al - 0.5 Si alloy. This may be a real compositional effect inasmuch as properties otherwise seemed to be fairly independent of prior treatment except cold work.

In addition to the difference in silicon content, the 6 Al - 0.5 Si alloy was made from a significantly lower hardness titanium sponge with a consequent lower contaminant level than the 6 Al alloy. It is not known to what degree the two factors influenced comparative results. On the basis of the data obtained in this investigation it seems best to assume that the rather close general agreement in properties was due to the major alloying element imparting strength being 6-percent aluminum in both cases. Differences such as the greater ability of the 6 Al + 0.5 Si to withstand cold work without cracking and a more stable structure were probbly due to silicon and/or the low interstitial element contents. The evidence for a strength characteristic of a stable alpha structure and independent of the alloying elements is very weak in view of the major alloying element of both alloys being 6-percent aluminum.

ALPHA-BETA ALLOYS

Alpha-Beta alloys are developed through the use of beta-stabilizing elements which effectively extend the temperature region of co-existence of the alpha and beta phases to low temperatures. Heat treatment to suppress the transformation of beta to alpha can markedly increase the amount of beta in such alloys at room temperature. The characteristic properties of such structures were surveyed in Reference 1 using the Cr-Fe alloy Ti 150A. The present investigation extends the data for Ti 150A to another heat and supplements the conclusions with similar data for the Cr-Fe-Mo-Al alloy Ti 155AX.

Alpha-Beta Alloy Ti 150A

The material studied, Heat M 739, was reported to have the chemical composition given in Table 1 and was in the form of 0.5-inch diameter bars rolled from a commercial heat. The available analyses indicate that it had higher carbon and lower nitrogen than Heat L 1006 used for Reference 1.

The investigation was limited to survey type tests for comparison with the results of Reference 1. (The decision to limit the evaluation to survey tests was based on the severe embrittlement found for specimens as a result of creep testing - a subject discussed in a subsequent report.)

Heat Treatments

Based on the results of Reference 1 three treatments were selected as being the best to establish principles for relating microstructures to properties. Material furnace cooled after one hour at 1500°F had equiaxed alpha grains in a beta matrix. Air cooling from 1500°F reduced the amount of alpha present by an appreciable amount. Heating to 1800°F for one hour in the beta range and then isothermally transforming at 1300°F for one hour produced coarse angular alpha particles generally oriented at right angles to the beta grain boundaries. Due to the quench, acicular alpha also formed in the beta matrix. Typical photomicrographs were shown in Reference 1.

Tensile Properties

The new heat (M 739) had lower strength and ductility in most cases (Table 13) than Heat L 1006 used for the studies reported in Reference 1. The lowest strengths were exhibited by the material annealed from 1500°F. It also had the highest ductility except at 1000°F. The new heat had the highest tensile strength when isothermally transformed at 1200°F after heating at 1800°F. Heat L 1006, however, had higher strengths when air cooled from 1500°F except when tested at 1000°F. Both heats had very high ductility at 1000°F.

From these results it can be seen that there were fair differences in tensile properties between the two heats. In some cases these were sufficient to change the relative order of properties for a specific heat treatment.

Rupture Properties

The data from rupture tests are presented in Table 14 together with comparative rupture strengths for Heat L 1006 from Reference 1. The stress-rupture time curves are shown as Figure 20.

There was little evidence of time dependent properties at 600°F and the rupture strengths were essentially the same as the tensile strength. The material isothermally treated at 1300°F may have had a slight slope to the stress-rupture time curve, although this is not certain due to an apparent slight variation in strength of individual specimens. Because rupture strength at 600°F was governed by tensile strength, Heat L 1006 with its somewhat higher tensile strengths was the stronger. Elongation and reduction of area were similar for both heats.

At 800°F the annealed condition was the weakest and there was little difference for the other two treatments. There was also little difference between the two heats. At 1000°F there was little difference between treatments or between heats.

Creep Properties

The results of creep tests and a comparison with tests on Heat L 1006 from Reference 1 are given in Table 15. This table was prepared to compare creep on the basis of similar individual tests insofar as possible from the data.

In general, Heat L 1006 used for Reference 1 had lower minimum creep rates at 600°F than Heat M 739 tested for the present report. There was very little difference at 800°F, Heat M 739 tending to be slightly more creep resistant at high stresses and slightly less at low stresses. At 1000°F Heat M 739 tended to have considerably lower creep rates in low stress tests with little differences at high stresses. Apparently the stress-creep rate curves at 1000°F had considerably less slope for Heat M 739 than for Heat L 1006. The annealed material had the lowest creep resistance at all three temperatures, except possibly for Heat M 739 at 1000°F. There was little difference between the other two treatments except for the isothermal treatment at 1300°F tending to be strongest at 1000°F.

The limited total deformation data (Table 15) indicate that Heat M 739 required a longer time to attain a given deformation than Heat L 1006 in tests at 800° and 1000°F. The data at 600°F are insufficient for valid comparisons.

Discussion

In general the differences between the two heats were not large. The main points of difference were the lower tensile strengths and ductilities of Heat M 739, and its higher strength in low stress creep tests and total-deformation strengths at 1000°F.

The structures with the larger amount of beta tended to have the highest strength. Thus, air cooling from 1500°F or isothermal treatment at 1300°F gave the highest strengths. The coarse angular alpha obtained by the isothermal transformation at 1300°F together with the interference to further beta transformation by water quenching from 1300°F produced the highest strengths in Heat M 739. Air cooling from 1500°F was generally more effective in Heat L 1006. It can be noted, however, that the differences between heats tended to be of the same order of magnitude as the differences between heat treatments.

The conclusion of Reference 1 that the strength increased with the amount of beta present was not changed. It did point out that differences between heats for reasons not identified can be a factor in properties. Furthermore, the relative strengths can be altered for structures as widely different as equiaxed alpha in a beta matrix and a complex angular alpha structure obtained by transformation from the beta phase.

Alpha-beta Alloy Ti 155AX

The alloy Ti 155AX was selected to check the role of microstructure in creep-rupture properties of alpha-beta alloys indicated by the results of the studies of Ti 150 A alloy. The Ti 155AX alloy differed considerably in composition (Table 1) from the Ti 150A alloy. It contained 5-percent aluminum, an alpha stabilizing alloying element. This, however, was counteracted by the beta stabilizing effects of 1.3-percent molybdenum, 1.4 percent chromium, and 1.3 percent iron.

Material Investigated

The chemical composition of the heat is given in Table 1. The stock was received in the form of 1/2-inch diameter rolled bar stock produced commercially. It was presumably rolled below 1650°F and finally heat tested at 1200°F.

Survey of Response to Heat Treatment

In the as-produced condition, the microstructure (Fig. 21a) consisted of equiaxed alpha particles in a beta matrix. Heating to temperatures between 1400° and 1600°F had little effect on microstructure or hardness. One hour at 1700°F (Fig. 21b) considerably reduced the amount of alpha and increased hardness. One hour at 1775°F practically eliminated alpha and increased hardness still further (Fig. 21c). A sample treated at 1787°F showed no residual alpha. When temperatures were increased to 1800° and 1900°F, transformation was not suppressed and martensitic alpha prime resulted with an increase in the original beta grain size (Figs. 21d and 21c). There was no further increase in hardness above that produced by the treatment at 1775°F. The hardness values indicate that transformation was not suppressed from 1775°F even though the microstructures showed little evidence of such transformation.

Air cooling from 1700° or 1800°F did not appreciably change the appearance of the microstructure from that resulting from water quenching (Fig. 22). Hardness values were, however, lower. Furnace cooling from 1700°F (Fig. 22) produced relatively large equi-axed alpha grains in a beta matrix. Furnace cooling from 1800°F produced a "basket weave" pattern of alpha plates in a beta matrix and resulted in still lower hardness values.

Reheating at temperatures between 445° to 1000°F for 1 hour after quenching from 1700°F did not change the relative amount, size and shape of the alpha grains (Fig. 23). Hardness was increased. A faint acicular structure was evident, particularly after heating at 1000°F. It is presumed that the acicular structure existed in the as-quenched structure and was rendered more visible by the heating.

The basic structure was also changed little by heating for 15 minutes at 800° to 1600°F after quenching from 1800°F (Fig. 23). Additional transformation occurred in the form of a fine elongated product growing from the beta grain boundaries. The higher temperatures reduced hardness somewhat.

Exposure to temperatures between 800° and 1600°F for 15 minutes for isothermal transformation from 1800°F resulted in similar but somewhat softer structures than were obtained by quenching and reheating (Fig. 24). The structures were, however, coarser the higher the temperature of isothermal transformation. The material exposed to 1600°F approached a basket weave type structure.

The hardness values as a function of the final treatment temperature are shown by Figure 25. This figure includes hardness values not shown with the photomicrographs. Exposure to 800°F gave maximum hardness although this is not certain for the as-produced material. Figure 25 also shows the high hardness from quenching from 1775° to 1900°F, and the reduction in hardness from air cooling or furnace cooling.

Tensile Properties

Reference to the microstructural studies and table 16 indicate that the following generalities can be made regarding structural variations and tensile properties at 75°F:

1. When the structure consisted of spheroidal alpha in a beta matrix the alloy had a good combination of strength and ductility. This applied to the as-produced, and the as-produced material heated at 1200° or 1700°F. Quenching from 1700°F increased strength and reduced ductility. Annealing had the opposite effect.

2. Reheating to 800°F after quenching from 1700°F greatly increased strength and embrittled the material. This indicates that the additional transformation from subcritical treatment of non-equilibrium beta caused pronounced changes in properties.

3. The hard alpha prime structures resulting from rapid cooling from 1800°F with or without subsequent exposure to 800°F were strong and brittle.

4. The complex structures developed by treatment at 1800°F followed by 1200°F isothermal transformation increased in ductility other than that developed by a 1200°F reheat following an 1800°F quench.

5. Apparently better combinations of strength and ductility were obtained when there was spheroidal alpha present than when the alpha prime structure was present, even when strength was at a high level.

All structures except the most brittle were tested at 600° and 1000°F with the results given in Table 16. The as-produced material was also tested at 800°F. The strength was reduced and ductility increased by increasing temperature of testing (Fig. 26), especially at 1000°F. The material quenched from 1700°F underwent little change on testing at 600°F, presumably because transformation of the beta increased strength relative to the quenched condition. It maintained a high level of strength at 1000°F as did the two treatments at 1200°F after exposure to 1800°F.

It is quite certain that most of the structures would have shown little decrease and even increases in strength if they had been tested at 800°F. Apparently the further transformation of beta at this temperature would have caused this to occur.

Rupture Properties

The heat treatments which did not show embrittlement at 75°F in tensile tests were also subjected to rupture tests with the results given in Table 17.

Rupture strengths at 600°F up to 1000 hours were practically equal to the tensile strengths with two exceptions (Fig. 27). These two exceptions were the water quench from 1700°F and the quench from 1800°F with a 15-minute reheat to 1200°F where there was a slight dependency of rupture time on stress. Both of these conditions had high tensile strengths and consequently had high rupture strengths in spite of the slight stress dependency of rupture time. Elongations and reduction of area decreased from the tensile-test values when rupture did occur.

There was considerably more creep and stress dependency for rupture at 800°F (Fig. 28) and a pronounced increase in ductility.

At 1000°F, creep was a major factor and the rupture strengths were only a fraction of the tensile strength (Table 17). The as-produced material was weakest (Fig. 28) and the material treated at 1800°F appeared to be strongest although the differences were slight.

The one test conducted on the as-produced material heated 24 hours at 1200°F before testing had similar properties to the as-produced condition at 1000°F.

Elongation and reduction of area were very high in all tests at 1000°F.

Apparently the rupture strengths of Ti 155AX are increased by either increasing the amount of beta phase originally present or by heat treating to develop an acicular transformation structure.

Creep Properties

The creep data obtained were limited to the rupture tests. The tests at 600°F showed very low rates for tests at stresses just under the tensile strength (Table 17 and Fig. 29). The as-produced material was considerably more subject to creep at 800°F. Creep was a predominating factor at 1000°F. Although the data were too sparse to establish creep strengths for the various treatments at 1000°F, it can be seen (Fig. 29) that the materials treated at 1800° and 1200°F had slightly higher creep strengths based on minimum creep rates.

The limited total deformation data in Table 17 indicate that:

1. At 600°F, most deformation occurred during loading. In the tests on the high tensile strength materials there was some creep after loading.

2. At 1000°F, most deformation accumulated by creep. Apparently minimum creep rates occurred early in the tests.

Structural Changes During Creep Testing

The as-produced structure was changed little by testing (Fig. 30) even at 1000°F. A considerable amount of dark etching transformation product was found in the beta during testing at 1000°F after the quench from 1700°F even in a short time period. No noticeable change occurred during testing at 600°F in the coarse alpha structure fromed by furnace cooling from 1700°F. A dark etching phase appeared in the beta phase of the material during testing at 1000°F. The material treated at 1800°F and then at 1200°F showed little change in structure even when tested at 1000°F.

Discussion

The strength and ductility of alpha-beta alloy Ti 155AX appears to have the following relationships to structures.

1. Increasing the amount of beta phase at the time of cooling generally increases strength and reduces ductility. The major exception to this was the high ductility at 1000°F.

2. Better combinations of strength and ductility are obtained by keeping the heating temperature below the all-beta range with some residual alpha in spheroidal form. This seems to reduce the degree of transformation of the beta.

3. When heated in the all-beta range and quenched, acicular alpha forms. Unless reheated to temperatures of 1200°F (or higher) or exposed isothermally to 1200°F (or higher) the material is quite brittle at room temperature. High creep-rupture strengths, however, were obtained when the final heating temperature was 1200°F.

4. Any treatment which provides increasing amounts of beta and is then followed by exposure below 1200°F subjects the material to severe embrittlement by the transformation of the beta. This effect is a maximum at about 800°F for normal heat-treating times. It apparently can occur at lower temperatures and longer time periods.

5. There is reason to expect that increasing strength from increasing beta is primarily due to the partial transformation of the beta during testing.

6. The alloy was not subject to creep at 600°F unless the tensile strength was increased to such high values by the strengthing from intermediate transformation product that high stresses could be applied. Creep was a larger factor at 800°F and a predominating factor at 1000°F. Differences due to initial structure variation were greatly reduced at 1000°F by structural instability.

Relative Properties of Alpha-Beta Alloys Ti 150A and Ti 155 AX

The general conslusions derived from the initial survey of properties of alpha-beta alloy Ti 150A were not appreciably changed by the check tests on a second heat of Ti 150 A or by the survey of the properties of Ti 155AX. Comparative properties of these alloys are summarized in Table 18. Certain generalities can be stated which apply to both alloys:

1. Good combinations of strength and ductility are obtained by heat treating within the temperature region of co-existence of alpha and beta. It is for this reason that commercial treatments are of this type.

2. Increasing the heat treating temperature within the alpha-beta range to reduce the amount of alpha and then using rapid cooling rates to retard transformation of the beta increases strength and reduces ductility up to at least 800°F. Rather pronounced increases in creep-rupture strength are possible by this procedure.

3. Heating to the all-beta region results in an increase in grain size and the development of an acicular transformation structure. Such structures have fairly high strength at room temperature but are brittle. The brittleness can be overcome to some degree by either reheating to temperatures of 1200°F or higher or by isothermal transformation in this temperature range. Exposure to temperatures between 700° and 1200°F intensifies embrittlement with a maximum effect at about 800°F. Embrittlement appears to be due to partial transformation of the beta. This phenomenonalso occurs in materials heated high in the alpha-beta range and rapidly cooled.

4. Heating to high temperatures and slow cooling to coarsen the structure does not improve properties over material consisting of small spheroidal alpha grains in a fine grained beta matrix obtained by hot-working and annealing low in the alpha-beta temperature range. Reheating to the alpha-beta range after rapid cooling from a high temperature did not give better properties.

5. There was very little creep at 600°F for any treatment. Rupture strengths out to 1000 hours were essentially equal to tensile strengths. A slight time dependency on rupture resulted when treatments gave high tensile strengths. This permitted the application of high stresses and rupture in less than 1000 hours under stresses slightly below the tensile strength. In general the load carrying ability of alpha-beta alloys at 600°F is governed by the stress-strain characteristics during loading. Treatments which increase the amount of nonequilibrium beta increase the tensile strength. Ductility falls to rather low values when these materials rupture due to transformation of the beta during testing.

6. The data are sparse at 800°F. The indications are that transformation of beta during testing at 800°F helped to maintain strength even in the materials annealed in the alpha-beta range. Those materials treated to increase the amount of beta would probably become very brittle at 800°F. Creep does limit load carrying ability at 800°F.

7. At 1000°F, there is relatively little difference in rupture properties between treatments. The Ti 150 A had very nearly the same properties regardless of treatment. In the Ti 155 AX there was also little difference, although some retention of strengthening from treatments at high temperatures was carried to 100 hours. All structures have very high ductility. Creep controls time dependent load carrying ability.

8. Both Ti 150A and Ti 155AX exhibited about the same response to heat treatment and a similar effect of temperature on properties. Absolute strengths were higher for Ti 155AX, although strengths relative to initial properties were of the same order of magnitude for both materials. Thus, it appears that the differences were a function of the added alloying elements present in Ti 155AX.

META-STABLE BETA ALLOYS

Certain alloying elements inhibit the beta-to-alpha transformation of titanium to the extent that an all-beta structure can be retained at room temperature by quenching. When the beta then transforms to alpha in a time and temperature dependent manner during reheating below the beta temperature range (or during slow cooling from the beta range), the alloys can be designated as a meta-stable beta type. Two binary alloys of this type, 10 percent molybdenum and 10 percent chromium, were investigated. Properties were evaluated for initial all-beta structures and for varying degrees of subsequent transformation to alpha.

The 10 Mo alloy was not delivered in time for inclusion in the work done for Reference 1. The 10 Cr alloy was purchased for the investigation covered by this report. Both alloys were evaluated by survey type tests.

Meta-Stable Beta Alloy 10 Mo

The structures resulting from various heat treatments were studied initially to check the transformation characteristics reported in the literature. This was supplemented by further study of reactions at 800° and 1000°F to explain test results. Selected structures were surveyed for properties at 600°, 800° and 1000°F.

Test Material

Test stock was melted by the Armour Research Foundation and received in the form of 1/2-inch diameter bar stock in the as-forged condition. The chemical composition is given in Table 1.

Survey of Microstructures

Microstructures and hardness values for treatments considered are given in Figure 31 and Table 19. In the as-forged condition the microstructure consisted of nearly equiaxed beta grains with a small amount of visible transformation product. A clear beta structure was obtained by quenching from 1800°F with a significant drop in hardness from the as-forged condition.

Heating the as-forged structure at 1350° to 1450°F in the alpha + beta region resulted in the appearance of alpha and a decrease in hardness. Prolonged exposure at 1425°F agglomerated the alpha mainly in the grain boundaries. At 1350°F, extensive alpha appeared within the beta grains. Hardness decreased from the R_c 32 of the as-forged condition to R_c 27-28 after heating in this range.

Hardness was increased by reheating below the alpha-beta temperature range with the highest hardness of the conditions studied being obtained at 800°F in one hour. (Table 19). The peak hardness was maintained for 5 hours and had fallen off slightly in 10 hours. A lower peak hardness developed in 30 minutes at 1000°F and dropped off slightly in 5 hours. These very short times to attain maximum hardness indicate the rapid rate of transformation. The slow rate of decrease during subsequent heating shows the slow rate of tempering. A finely divided transformation product, probably omega phase, was observed in the beta grains after heating in this range (Fig. 31).

Heating at 1100° and 1300°F did not increase hardness (Table 19) as much. During metallographic preparation an adherent blue-black coating formed on the specimen surface with a number of etchants. This coating obscured the microstructures and consequently satisfactory photomicrographs could not be obtained.

Isothermal transformation at 1200°F for one hour after heating to 1800°F produced acicular alpha prime which grew from the beta grain boundaries and was accompanied by a slight increase in hardness. A 15-minute isothermal transformation at 950°F caused a fine transformation product to form within the beta grains with an increase in hardness.

Heating at 1800°F increased grain size moderately. Heating at the lower temperatures did not change the grain size.

Tensile Properties

The structures subjected to tensile tests and the test results are given in Table 20 and Figure 32. The results show:

1. All-beta or nearly all-beta structures (as-forged or quenched from 1800°F) had relatively low strength and high ductility at 75°F. These properties were not changed appreciably by heating at 1300° or 1425°F to produce agglomerated alpha. This was also true for the beta structure with a small amount of transformation at the grain boundaries from the isothermal transformation at 1200°F for 1 hour. Strengths were increased and ductility sharply reduced by heating for 15-minutes at 1300°F or 950°F after treatment at 1800°F to produce fine generally distributed transformation products.

2. At 600°F the essentially all-beta structures did not change much in tensile strength and increased in yield strength with a sharp decrease in ductility in comparison to 75°F properties. The structures previously treated to form a fine generally distributed transformation product decreased in strength and increased in ductility from properties at 75°F.

3. The nearly all-beta as-forged structure increased in tensile and yield strength at 800°F to values higher than existed at 75°F. Two tests were conducted, with a sample which was held at test temperature a somewhat longer time before testing having the higher strength. The increase in strength at 800°F was accompanied by a sharp decrease in ductility. Presumably, all treatments considered would have shown increases in strength and reduced ductility at 800°F, although structures initially heated to form finely divided transformation products may not have shown as much increase.

4. The data at 1000°F indicates that those structures which were largely untransformed beta maintained strength better than the structures which had been exposed to transformation at higher temperatures before testing. Except for the as-forged condition, ductilities were no higher or lower than at 600°F.

These tensile properties suggest that beta transformed during heating for testing and possibly during testing itself with a consequent increase in strength and a loss in ductility. The effect was a maximum at 800°F. Those structures transformed prior to testing fell off in strength with increasing temperature in a normal way. Evidently the strengthening from transformation was not very stable at 1000°F.

Rupture Properties

The rupture strengths of the 10-Mo alloy at 600°F (Table 21 and Fig. 33) were practically the same as thetensile strengths to 1000 hours due to the absence of time-dependent rupture. The exception to this were the premature fractures with low ductility of the material isothermally treated at 950°F after treatment at 1800°F. This material had a high tensile strength in comparison to the other treatments tested. Fracture occurred in the tests at stresses 4,100-6,000 psi below the tensile strength but nevertheless much higher in stress than for the other conditions tested.

At 800°F, the alloy in the as-forged condition was subject to time dependent rupture strength. The tensile strength at 800°F was much higher than at 600°F. The susceptibility to time dependent rupture at 800°F reduced strength below that at 600°F in about 25 hours. Elongation appeared to be increasing with rupture time.

At 1000°F there was very little difference in 100-hour rupture strength between the treatments. Elongations increased markedly with time for rupture in most cases. Any differences in rupture strength imparted by initial treatment apparently disappeared in less than 50 hours. The curves of Figure 33 are not well established, however, and more tests might have indicated significant differences due to treatment at longer time periods.

There seem to be two mechanisms operating in these tests. Transformation of beta probably strengthened most of the structures as time increased at 600°F. The exception was the material transformed at 950°F where the initial high stresses probably introduced a susceptibility to premature fracture similar to the effects of cold work on alpha alloys. At 800°F overaging with time at temperature probably reduced strength in addition to the appearance of some susceptibility to normal creep. Apparently structural changes occurred so fast at 1000°F that all initial structures soon reached the same state and the strength became a function of a fairly uniformly creep resistance.

Creep Properties

Very little creep occurred in tests at 600°F (Table 21). The material isothermally transformed at 950°F may have been an exception. However, it is not certain if this represented true creep or an overaging effect which permitted delayed yielding and fracture. Total deformations at 600°F were governed by the deformation on loading with the exception of the material treated at 950°F and tested at high stresses.

The creep rates at 800°F (Table 21) from the tests on the as-forged material were quite high and in accordance with the short rupture times. Again it is uncertain to what degree true creep and overaging were involved.

Creep was a major factor in the tests at 1000°F (Table 21 and Fig. 34) and was a governing factor in fixing total deformation as a function of time. The stresses to limit total deformation to 1 percent in 100 and 1000 hours would apparently be considerably less than half the corresponding rupture strengths. The minimum creep rates showed a possible slight superiority for the material transformed at 1200°F and for the material quenched from 1800°F.

Microstructural Changes During Testing

Photomicrographs of specimens after testing are shown in Figure 35. The major points shown by these microstructures were:

1. Testing at 600°F had little effect on microstructure in most cases. The as-forged material stained during etching and the structure shown probably suggests more transformation than actually occurred. The material isothermally transformed at 1200°F underwent growth of the transformation product completely through the grains while it was restricted to the vicinity of the grain boundaries in the heat treated condition.

2. When tested at 800°F the as-forged material showed more evidence of transformation.

3. All structures showed considerably more transformation after testing at 1000°F. In all samples the initially clear beta contained a finely divided general transformation product. Even those specimens initially transformed before testing to give a fine general transformation product, showed additional transformation and possibly some agglomeration.

Discussion

Creep was a minor factor affecting properties of the 10 Mo alloy at 600°F; probably a major factor at 800°F, and was the controlling factor at 1000°F. The role of creep was, however, complicated by the transformation of beta during testing. This probably increased strength with testing time at 600°F. Over-aging at 800°F in combination with creep served to cause strength to fall off with time, even though rapid transformation could raise tensile strengths above those at 600°F. At 1000°F structural changes occurred so rapidly that there was little effect of prior treatment on creep rupture properties.

Treatments to develop finely dispersed transformation products prior to testing raised tensile strengths to the extent that such high creep test loads could be applied at 600°F that early failure occurred. It is not certain if this represents true creep or a delayed yield and fracture effect. Such structures had low ductility at 75°F.

The rapid transformation of beta at 800°F raised tensile strength and reduced ductility to a marked extent.

Meta-Stable Beta 10 Cr Alloy

Heat treatments for structure studies were selected from information available in the literature. Selected structures were then subjected to tensile tests on the basis of which three conditions were selected for rupture and creep tests.

Experimental Material

The composition of the heat is given in Table 1. The stock was received in the form of 0.5-inch diameter bars in the as-forged condition. Forging was carried out at about 1800 to 1900°F.

Survey of Microstructures

Photomicrographs illustrating the microstructural variations with heat treatment are shown in Figures 36 and 37.

The as-forged condition consisted of equiaxed alpha grains with small darkened areas indicating some transformation. Quenching from 1800° or 1900°F produced a clear beta structure. The hardness was, however, rather high for an apparently untransformed structure.

Examination of the structures and hardness values (Table 22) indicates that heating at temperatures of 600° to 1000°F produced increased hardness and a fine general transformation product, particularly at the higher temperatures. Hardness reached a maximum between 5 and 10 hours at 800°F and in only a few minutes at 1000°F. The same effects were obtained on reheating to 800°F after prior exposure in the range of 1265° to 1335°F (in the two-phase region) to produce an alpha-beta structure containing fairly massive alpha. Heating at 800°F gave the highest hardness.

Quenching from 1800°F and reheating produced the same general effects as in the as-forged condition. The shorter times of reheating resulted in finer and more generally dispersed transformation products. The angular needle-like structure which formed in the short times at 1200°F and 1300°F was presumably the source of the more massive alpha found in the as-forged material heated at these temperatures for longer time periods. There was a fine general transformation product visible in the material heated at 1600°F.

Isothermal treatment at 1470°F caused a dense transformation product to form at the grain boundaries.

Tensile Properties

The data in Table 23 and Figure 38 show the following tensile characteristics:

1. All beta structures had reasonably good ductility at room temperature. Isothermal treatment at 1470°F did not improve properties. Reheating either the as-forged or materials quenched from 1800°F reduced

strength and ductility at room temperature with severe embrittlement resulting from heating at 800°F. In general, materials exposed to 800°F were so brittle that full tensile strength could not be developed. In one case heating at 1000°F did raise the tensile strength. The relatively massive alpha structures developed by heating at 1265° to 1335°F were lower in ductility and strength than the as-forged condition. These latter structures were severely embrittled by reheating to 800°F. Only those structures which showed reasonable ductility at 75°F were subjected to tensile tests at elevated temperatures.

2. When tested at 600°F the structures which had reasonable ductility at room temperature were quite brittle and underwent some loss in strength.

3. The one test conducted at 800°F, on as-forged material, gave higher strength and ductility than when tested at 600°F. The properties were nearly as high as those found at room temperature.

4. All structures tested at 1000°F had high ductility and retained only about 30 to 40 percent of the room temperature tensile strength.

Rupture Properties

Due to embrittlement imparted by heating at temperatures below 1300°F, only three structures were subjected to rupture tests. The data obtained (Table 24 and Fig. 39) indicate the following:

1. When tested at 600°F under about 99 percent of the tensile strength, the as-forged and as-forged + 1335°F treated materials had very prolonged rupture times. The as-forged structure had considerably higher elongation than in the tensile test. The all-beta structure obtained by quenching from 1900°F broke immediately in a brittle manner. The latter material probably would not have ruptured in a slightly lower stress test.

2. The as-forged structure had a strong stress dependency for rupture time at 800°F. The ductility was good.

3. The rupture strengths at 1000°F were similar and quite low for the three treatments. The ductility fell off with increasing rupture time from the high values in the tensile tests.

Creep and Total Deformation Properties

The total deformations and creep rates at 600°F were very small (Table 24 and Fig. 40) considering that the test stresses were 99 percent of the tensile strength.

Creep was extremely rapid at 800° and 1000°F as soon as the loads were applied in the rupture tests. It was so rapid that the total deformation could not be measured with the extensometer used. Only creep rate values were obtained, even when tests were run with relatively low stresses and prolonged times for rupture. Sufficient data were not obtained to establish stress-creep rate curves. The limited data indicate little difference between treatments at 1000°F.

Microstructures after Creep-Rupture Testing

After testing at 600° and 1000°F, all specimens showed the presence of dense, dark-etching transformation products (Fig. 41). Staining during etching complicated the operation of the structures after testing at 600°F. An unidentified and unusual angular white pattern developed in samples tested at 1000°F.

Discussion

The largely all-beta structures had good strength and ductility at 75°F. Prior transformation by heat treatment up to 1200°F caused severe embrittlement. Treatment at higher temperatures to produce relatively massive alpha reduced strength and ductility from the largely all-beta form.

The all-beta structures became brittle during tensile testing at 600° and 800°F due to transformation during heating. The rapid and extensive transformation at 1000°F resulted in low strength and ductility.

In rupture tests at 600°F on essentially all-beta structures, the transformation caused embrittlement. One test on as-forged material suggested that overaging would reduce strength and allow rupture to occur with some increase in ductility in tests just under the tensile strength. At 800° and 1000°F the loss in strength due to transformation allowed rapid creep and low rupture and creep strengths.

Comparison of Meta-Stable Beta Alloys 10 Mo and 10 Cr

The 10 Mo alloy appeared to be considerably more resistant to transformation than the 10 Cr alloy. The higher tensile properties (Table 25) of the 10 Cr alloy at 75° and 600°F seem to be best accounted for on this basis. The high hardness of the 10 Cr alloy when quenched to an apparently all-beta structure suggests that transformation was not suppressed. The 10 Mo alloy only had equal strength at 75° and 600°F when it had been exposed to 950°F to induce transformation.

The lower rupture strength of the 10 Cr alloy at 800°F (Table 25) apparently was due to greater instability of structure. More rapid and extensive transformation allowed strength to fall off with time.

When tested at 1000°F, the lower tensile strength of the 10 Cr alloy (Table 25) apparently was due to rapid "overaging"during heating for testing. The near equality of rupture strengths for the two alloys suggests that when there was opportunity for transformation to attain equilibrium at 1000°F in the more sluggish Mo alloy, there was little difference in strength.

The results indicate that the properties of both alloys were largely controlled by the reaction kinetics of the transformation of beta and the consequent stability of the structures. Apparently both alloys had similar properties for the same degree of transformation. The major difference was the greater instability of the beta in the 10 Cr alloy. At 1000°F when creep was a major factor there was no great effect evident from compositional differences.

STABLE BETA ALLOYS

Certain alloying elements lower the temperature range at which beta is stable. In the proper amounts the structure is maintained as beta to low temperatures. The 30 percent molybdenum and a 50 percent vanadium alloy were used as the two examples of this type of alloy.

The 30 Mo alloy had been surveyed for properties for Reference 1. These data were supplemented by more extensive and longer time tests to verify the trends of the survey data on the relative properties of a stable beta structure. The 50V alloy was subjected to survey tests to check the properties indicated for the stable beta structure by the 30 Mo alloy.

Stable Beta Alloy 30 Mo

The extended evaluation of the 30 Mo alloy was restricted to the as-forged condition. The limited studies of Reference 1 did not disclose a significant variation in structure or properties from heat treatment. In this condition the microstructure of the alloy consisted of clear equiaxed beta grains.

Experimental Material

The composition of the material used for the experiments is given in Table 1. The material was the same as that used for Reference 1. The stock was in the form of 0.5-inch diameter bars in the as-forged condition with a forging temperature of 2250°F. The producer reported that difficulty in obtaining a homogeneous alloy was encountered during melting. Individual 1000 gram ingots made from a 60 percent molybdenum master alloy had to be melted and remelted five times. Each of these pancake-type ingots were rolled to approximately 1/16-inch sheet, cut into 1/4-inch sections and recharged into the tungsten non-consumable electrode furnace to produce five heats weighing two pounds. These were then forged to 0.5-inch round bars.

Creep-Rupture Properties

The additional tests, Table 26 and Figure 42, indicated that the stable beta structure was practically free from time-dependent rupture and creep at 600°F. Practically all the deformation in the tests was introduced during loading. Tests well above the yield strength and practically equal to the tensile strength were extended to time periods of 1000 hours or longer with no indication of rupture or appreciable creep.

At 1000°F the stable beta structure was subject to time dependent rupture and creep (Table 26 and Fig. 43). The tensile strength was 83,000 psi whereas the stresses for rupture in 100 and 1000 hours were 37,500 and 23,000 psi. Elongation and reduction of area were high in the rupture tests.

Creep was a predominant factor in the stress-time for total deformation curves at 1000°F (Table 26 and Fig. 43). Third-stage creep occurred at rather low total deformations except in the lower stress tests. The lower stress tests consequently required longer times to reach a given total deformation than was indicated by the higher stress tests. It is uncertain, however, whether this was a real effect or reflected variation in properties between specimens. The erratic nature of the stress-creep rate curves (Fig. 44) suggests that variation in properties between specimens was mainly responsible. Such variation between specimens from as-forged stock from fine separate small heats would not be unusual.

Microstructural Changes During Creep Testing

The specimens exposed to creep for 2000 hours showed the presence of a transformation product, Figure 45, concentrated towards the centers of the beta grains. Comparison with the photomicrographs in Reference 1 shows that the longer time of 2000 hours at 1000°F resulted in the development of more extensive transformation. The transformation product was previously identified as alpha (Reference 1). No change in microstructure was evident in specimens tested at 600°F.

Discussion

The more extensive testing of the all-beta structure generally tended to confirm the results of the survey tests of Reference 1. The initial absence of creep at 600°F was confirmed with the consequent total deformation in 1000 hours being mainly dependent on the deformation during loading was demonstrated. It seems evident that yield strengths would control design stresses at 600°F for a stable beta alloy, at least for time periods up to 1000 hours.

At 1000°F the creep characteristics were well delineated by the more extensive tests. The occurrence of third stage creep at relatively small total deformations was demonstrated. The somewhat erratic nature of the creep rate data suggests that the properties of the stable beta structure are not as free from prior history effects as seemed indicated by the survey tests of Reference 1. Because compositional variations between the five small ingots of the heat could be responsible for variation in properties, the prior history effect would require check experiments for verification. The survey tests of Reference 1 showed little variation with heat-treatment and suggested little influence from prior history.

Strengthening from transformation of beta also seems to be a possible cause of retained strength with increasing temperature. The sluggish nature of the transformation could make it effective at 1000°F.

Stable Beta Alloy 50 V

An alloy containing 50-percent vanadium was selected for checking the creep-rupture properties of stable beta alloys indicated by the 30 Mo alloy. This alloy had the proper microstructure according to information in the literature and was sufficiently different in composition to be satisfactory.

A brief survey of the effects of heat treatment on microstructures was used to select structures for tensile and creep-rupture at 600°F and 1000°F.

Experimental Material

Eight pounds of 5/8-inch diameter bar stock was furnished in the as-forged condition. It was produced from three small ingots having the chemical composition given in Table 1.

Considerable difficulty was encountered by the producer in preparing the stock. In the first attempt the material was difficult to prepare because long columnar grains, formed during melting, resulted in hot tears and other forging difficulties. In addition the high forging temperature of 2400°F required for the alloy resulted in nitrogen pick-up. Finally the stock was produced in three small heats to alleviate the problems resulting from columnar grains. Nitrogen pick-up was minimized by eliminating one forging operation, reducing furnace heating time, and forging oversize to give extra stock for removal of the high nitrogen surface. As a result of this procedure, the material finally received had only 25 to 40 percent of the nitrogen content obtained in the first attempt at producing this alloy.

Even with these precautions, the stock received at the University contained numerous forging cracks.

Microstructures and heat treatment

A brief survey of the response of the 50 V alloy to structural changes by heat treatment was carried out with the following results:

1. The as-forged structure consisted of large variable size equi-axed beta grains (Fig. 46). A light precipitate was scattered through the grains. The large and variable grain size probably reflects forging at 2400°F. The precipitate was not identified. In an investigation of a 40-percent V alloy Adenstadt (Ref. 3) encountered an impurity phase of similar appearance.

2. Heating the as-forged structure for 4 hours at 1800°F apparently coalesced and partially dissolved the precipitate particles (Fig. 46). This temperature is well in the region of all beta phase. An x-ray diffraction determination gave 4 beta lines and 12 alpha lines. In addition there were six faint lines at positions near those indicated in the literature for omega phase. It may be that breakdown of beta to omega and alpha took place or had taken place prior to the 1800°F treatment. Adenstadt did not find extra lines in the impurity rich samples nor was he able to correlate its presence with metallic contaminents. He also reported that prolonged heating at 1200°F after cold work did not cause diffraction lines for any extraneous phase to appear.

3. The average size of the beta grains increased by heating at temperatures of 1800°F and higher. The following average grain diameters were measured for the indicated heat treatments:

Heat Treatment	Average Grain Diameter (mm)
As forged	1.50
1 hour at 1800°F, water quench	1.57
4 hours at 1800°F water quench	1.72
1 hour at 1925°F water quench	1.84
2 hours at 1925°F, water quench	2.24

The increased grain size after heating for 2 hours at 1925°F is shown in Figure 46.

Tensile and Creep Rupture Properties

Conditions of the 50 V alloy selected for testing were the as-forged condition and quenched after 2 hours at 1925°F. These conditions represented the extremes of average grain size encountered in the preliminary surveys. Cold working was not feasible with this alloy due to the presence of many forging cracks in the bar stock.

The room temperature tensile properties of the 50 V alloy were moderate (Table 27 and Fig. 47). The reduction in strength at 600°and 1000°F was less than was the case for most of the alloys considered. However, the elongation was only about 14 percent at room temperature and decreased to 10 percent at 1000°F.

Heat treating the as-forged material at 1925°F increased tensile strength about 10,000 psi with no effect on ductility.

There was no indication of time dependency for rupture at 600°F (Table 27 and Fig. 48). Tests at stresses just under the tensile strength did not fracture in 1000 hours. Most of the deformation occurred during loading of the specimens. As-forged specimens which fractured in a short time period probably represented the effect of normal variation in properties. The material treated at 1925°F retained the superiority in strength over the as-forged condition observed in the tensile tests.

The alloy was very subject to creep at 1000°F. The tests conducted indicate rupture strengths for 100 hours which were slightly less than one-half the tensile strength. The data indicated a marked increase in elongation and reduction of area with increasing time for rupture. The material treated at 1925°F maintained superior strength at 1000°F to the as-forged condition. The superiority, however, was not as large as in tensile tests. This superiority was also reflected in the rather limited total deformation data from the rupture and creep tests. There was, however, suprisingly little difference in minimum creep rates (Fig. 49) for the two treatments.

Microstructural Changes During Testing

Testing at 600°F did not alter the microstructure appreciably (Fig. 50). When tested at 1000°F, however, a new phase formed (Fig. 50) that was reasonably well distributed in the as-forged specimens but appeared to outline a network of subgrains in the specimens quenched from 1925°F. The new phase obscured the original unidentified second phase and it is not certain whether or not it developed from the beta phase or from the original second phase. It is probable, however, that the new phase represents progress in the breakdown of beta towards the alpha phase.

Discussion

The properties of the 50 V alloy were apparently slightly more subject to alterations by heat treatment than the comparative 30 Mo stable beta alloy although the heat treatments investigated were very limited. It was considered, however, that the two conditions studied would serve to reasonably well define the properties of the alloy.

The microstructures showed the presence of an excess phase for which there was no adequate explanation even after quenching from a high temperature. Other investigators had observed similar phases although they had not found evidence of omega or alpha phases as the limited work done in this investigation suggested. There was evidence that further breakdown of beta occurred during testing at 1000°F.

This alloy was so slightly subject to creep at 600°F that yield strengths would control working stresses. At 1000°F, however, the properties were almost wholly dependent on creep characteristics.

Relative Compositional and Structural Effects for Stable Beta Alloys

The 50-percent vanadium alloy was included in the investigation to obtain an indication of the relative effects of the type of microstructure and alloying elements in controlling properties by comparison with the properties of the 30-percent molybdenum alloy. The main available comparison is for the as-forged condition. The Mo alloy when heat treated in the range of 1500° and 1300°F did not have significant variation in properties from the as-forged condition (Ref. 1). There was not sufficient stock available to subsequently establish properties for the Mo alloy after a high temperature solution treatment as was done for the vanadium alloy. The vanadium alloy was not investigated with the lower temperature treatments because the Mo alloy showed so little effect.

The tensile strengths of the 30-Mo alloy were above those for the 50-V alloy at room temperature but fell off more rapidly with temperature to lower values than the 50-V alloy (Table 28 and Fig. 47). The 30-Mo alloy also underwent a substantial increase in ductility at 1000°F whereas the 50-V alloy did not. There was little difference at 600°F.

Neither alloy showed time dependent rupture or appreciable creep at 600°F. Consequently yield strengths would govern working stresses at this temperature. The 30-Mo alloy retained yield strength at temperatures somewhat better than it did tensile strength (Table 28 and Fig. 47).

At 1000°F the properties of both alloys were governed by creep characteristics. The comparative creep-rupture properties for the two stable beta alloys summarized in Table 29 show relatively little difference between the two alloys. The largest difference was in the 50-V alloy having about half the elongation and reduction of area that the 30-Mo alloy exhibited in the rupture tests. The 50-V alloy had slightly higher rupture and creep strengths and somewhat lower total deformation strength in the as-forged condition. When treated at 1925°F, the 50-V alloy compared more favorably except possibly in creep strength.

The available data indicate remarkable similarity in properties for the two beta alloys differing so widely in chemical composition. Both alloys showed evidence of structural changes during testing. Microstructurally this was considerably more evident in the 50 V alloy than in the 30 Mo. In view of the very strong effects of beta instability on creep-rupture properties of other alloys, it is suggested then the role of this instability should be better understood before it is concluded that the beta structure was the predominant factor and not the chemical composition.

CONCLUSIONS

CORRELATION OF MICROSTRUCTURES AND CREEP-RUPTURE PROPERTIES

The comparative properties for the alloys investigated are summarized by Figures 51, 52, 53, and 54. These figures are arranged to show properties as a function of the type of microstructure, individual alloys and the ranges of properties observed for each alloy. The figures are restricted to those properties for which there were reasonably complete data for all alloys. A considerable number of other properties are given in the preceding sections for some of the alloys including total deformation and rupture strengths to 1,000 hours in a number of cases. It is particularly important to recognize that properties for only a few treatments of each alloy were established and that many treatments which would have had inferior properties were omitted.

These figures and the detailed results in preceding sections show that maximum strengths shifted from the alpha-beta and meta-stable beta alloys at room temperature and 600°F to the stable alpha and stable beta alloys at 1000°F. The temperature of shift in strength was lower the longer the time period at which strength was evaluated. The un-alloyed alphatitanium Ti 75A was weakest under all conditions of testing.

Creep was not a factor in strength at 600°F for any of the alloys and deformation up to 1000 hours was governed by the deformation from the application of the stress. It was a substantial factor in strength at 800°F and a controlling factor at 1000°F.

Rupture strengths were equal to the tensile strength at 600°F except when the tensile strength had been increased to a marked extent by cold working of alpha alloys or by transformation of alpha-beta and meta-stable beta structures at temperatures below 1000° to 1200°F. In these latter cases rupture usually occurred immediately after loading when the stresses were higher than the yield strength. This appeared to be a delayed tensile failure under the high stresses permitted by the increased tensile strength rather than true creep to rupture.

Within certain limitations, there was surprisingly little effect from variations in chemical composition. The major influences of chemical composition were:

1. Unalloyed commercially purealpha titanium (Ti 75A) had low strength under all conditions considered.

2. Stable alpha and stable beta alloys had similar properties even though 6-percent aluminum was the major alloying element in the former and 30-percent molybdenum and 50-percent vanadium in the latter type. The 50V alloy tended to have lower ductility at 1000°F as the largest difference.

3. Alpha-beta and meta-stable beta alloys had similar properties mainly because their strength was governed by strengthening from transition structures between beta and alpha. The reaction kinetics as influenced by chemical composition rather than the composition itself appeared to be the controlling factor.

4. The alpha-beta and meta-stable beta alloys had low strength at 1000°F in relation to their properties at lower temperatures. It appeared as if the rapid rates of structural change during testing removed differences arising from the initial state of transformation of the beta and properties leveled off at values characteristic of the alloy content. The Ti 150A alloy had a lower level of strength than the more complex alpha-betaalloy, Ti 155AX, or the 10Mo and 10Cr meta-stable beta alloys.

5. The compositional difference between the two stable alpha alloys, 0.5-percent silicon, was probably not large enough to produce a marked effect. The silicon bearing alloy, however, could be cold reduced by rolling more than the 6Al alloy. Differences between the two alloys were generally less than those observed between two heats of 6Al alloy.

The influence of treatment within a given type of alloy appeared to be as follows:

1. The commercial purity alpha titanium was considerably improved in strength by quenching from the beta range to produce an acicular structure. Cold work also improved properties at 600°F.

2. Stable alpha alloys did not appreciably change in creep-rupture properties with heat treatment alone. Improved strength could be obtained at the lower temperatures and shorter time periods by cold working. Strength improvement was maintained quite well at 600°F and 800°F. At 1000°F, however, the cold worked conditions had lower strength than the heat treated conditions for time periods longer than about 100 hours. It appeared as if recrystallization-recovery type reactions were responsible for the loss in strength.

3. In Reference 1, it was concluded that the strength of alpha-betaalloys up to about 100 hours at 1000°F increased with the amount of beta originally present. Production of these structures involved rapid cooling from temperatures high in the alpha-beta or beta temperature ranges. The conclusion was not changed by the results of this investigation. The best all around properties were obtained when the material was worked and heat treated in the alpha-beta range to produce equi-axed alpha grains in a beta matrix. Higher strengths can be developed by

heat treatment in the beta range; but at the expense of embrittlement as heat treated or during testing at 600°and 800°F.

4. The strength of meta-stable beta alloy increased with the degree of subcritical transformation of beta at temperatures up to 1000° to 1200°F. The maximum effect occurred at 800°F. It, however, caused undue brittleness at low temperatures and higher transformation temperatures were required to obtain ductility.

5. The delineation of the influence of treatment on both the alpha-beta and meta-stable beta alloys at 600°F and 800°F was complicated by transformation of beta during testing. Structures that were initially predominantly beta could become very strong and brittle, particularly at 800°F. The 10Cr alloy transformed more readily than the 10 Mo alloy.

6. The properties of the stable beta alloys were not changed appreciably by the heat treatments applied. The 30 Mo alloy could not be cold worked without cracking. A high temperature quench slightly increased the strength of the 50 V alloy but had no effect on the 30 Mo.

There are a number of limitations to the conclusions indicated by the data. The temperatures of testing were rather widely spaced. The temperatures at which creep became a major factor were therefore not well defined. In a number of cases it now appears that creep-rupture tests with other heat treatments would have been desirable to better define effects.

The alloys were mainly from small experimental heats where the melting and hot working practice may not have resulted in typical properties. In particular, the results do not reflect possible major effects from variations in the normal interstitial elements, carbon, nitrogen and oxygen, and the possible influence of hydrogen.

The tendency for creep-rupture properties to be so independent of chemical composition and, to a degree, of type of microstructure is contrary to a great deal of experience in other alloy systems. For this reason, care should be exercised in accepting the indicated conclusions until the metallurgy of titanium alloys is better understood.

The metallurgy of the alloys appears to be fairly complicated. Opportunity for more thorough study of the mechanisms involved in the various structural effects noted would be desirable. The data suggest that solid solution strengthening was the major factor in the stable alpha and stable beta alloys. It was also probably a major factor in the strength of alpha-beta and meta-stable beta alloys at 1000°F. Better proof of this would be helpful. The same would be true for the instability effects in the other alloys. There is a strong possibility that strain aging reactions of the type identified in Reference 2 were influential although not recognized.

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TABLE 1

CHEMICAL COMPOSITION OF EXPERIMENTAL ALLOYS

Alloy	Heat No.	Supplier	Sponge Hardness BHN	Alloying Constituents (Weight %) Major				Minor			
				Fe	Cr	Mo	Al	C	N	H_2	O_2
Alpha Titanium											
Ti 75A	L984	TMC	-	.19	-	-	-	.025	.061	.0087	-
	M626	TMC	-	.14	-	-	-	.042	.026	.0150	-
Stabilized Alpha Alloys											
6Al	4*	ARF	146	-	-	-	6.21	-	-	.017	.21
	4a*	ARF	124	-	-	-	5.89	.075	.055	.015	.097
				Si							
6Al - 0.5Si	7011	ARF	124	.48	-	-	6.51	-	.017	.0170	-
Alpha + Beta Alloys											
Ti 150A	L1006	TMC	-	1.52	2.68	-	-	.046	.124	-	-
	M739	TMC	-	1.47	2.76	-	-	.08	.026	.031	-
Ti 155AX	M1400R	TMC	-	1.44	1.28	1.16	5.11	.037	.017	.0180	-
Meta-Stable Beta Alloys											
10 Mo	5*	ARF	146	-	-	10.50	-	.056	.022	.0130	-
10 Cr	6824	ARF	124	-	9.95	-	-	.059	.022	.0210	-
Stable Beta Alloys											
30 Mo	6*	ARF	-	-	-	29.58	-	-	-	.030	.17
				V							
50 V	8414	ARF	124	50.0	-	-	-	.062	.069	.0177	-
	8415	ARF	124	49.7	-	-	-	.064	.073	.0205	-
	8416	ARF	124	50.5	-	-	-	.051	.122	.0203	-

*	U. of M. Designation
TMC	Titanium Metals Corporation
ARF	Armour Research Foundation

All Analyses Except H_2 and O_2 by Producers
H_2 from Materials Laboratory, WADC
O_2 from National Research Corporation

TABLE 2

TENSILE DATA FOR ALPHA TITANIUM (Ti 75A)

Test Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	Elongation (% in 1 in)	Reduction of Area (%)
		1500°F for 1 Hour + Furnace Cool		
75	88,600	73,300	26.5	44.3
600	35,000	21,400	42.0	67.5
		1500°F Anneal + 31-Percent Cold Work		
75	130,500	118,200	15.3	40.8
600	71,700	67,300	14.6	51.1

TABLE 3

CREEP-RUPTURE DATA FOR ALPHA TITANIUM (Ti 75A) AT 600°F

Stress (psi)	Rupture Time (hours)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Time to Reach Specified Total Deformation (hours) 0.5%	1.0%	5%	Special (c)	Minimum Creep Rate (in/in/hr)
1500°F for 1 Hour + Furnace Cool									
35,000	Tensile Test	42.0	67.5	--	--	--	--	--	--
34,000	0.2	--	--	6.50	a	a	a	--	--
33,000	>996.7	--	--	3.34	a	a	b	--	--
32,000	>834.8	--	--	2.55	a	a	b	(3%) 250	5.6×10^{-7}
30,000	> 1	--	--	1.94	--	--	--	--	--
25,000	> 1	--	--	1.17	--	--	--	--	--
1500°F Anneal + 31-Percent Cold Reduction									
71,700	Tensile Test	14.6	51.1	--	--	--	--	--	--
65,000	0.5	13.6	45.0	.93	a	--	--	--	--
60,000	153.0	15.5	58.8	.68	a	4.5	--	--	2.2×10^{-4}
58,500	472.2	16.7	57.4	.77	a	7.0	360	(4%) 280	7.7×10^{-5}
57,000	1390.6	12.0	59.8	.67	a	20.0	b	(4%) 710	2.2×10^{-5}
44,000	>1252.5	--	--	.49	1.0	295.0	b	(.72%) 20	1.1×10^{-6}
42,000	>1168	--	--	.42	10.0	b	b	(.72%) 200	7.1×10^{-7}
40,000	>1151	--	--	.36	20.0	b	b	(.72%) 950	6.2×10^{-7}

\> Test stopped without rupture
a Exceeded deformation on loading
b Stopped before reaching deformation
c Special deformation in parentheses

TABLE 4

SUMMARIZED PROPERTIES OF ALPHA TITANIUM (Ti 75A)

	Temp. (°F)	Annealed 1500°F	Annealed 1500°F 31% Cold Reduction	Acicular Structure Quenched from 1800°F
Tensile Strength (psi)	600	35,000	71,700	54,800
Yield Strength for 0.2% Offset (psi)	600	21,100	67,300	37,700
Rupture Strength	600			
100-hour (psi)		33,900	60,500	54,000
1000-hour (psi)		33,900	57,500	--
1% Total Deformation Strength	600			
100-hour (psi)		(~ 25,000)	(50,000)	--
1000-hour (psi)		(~ 25,000)	(43,000)	--
100-Hour Rupture Strength (psi)	800	17,500	20,500	23,000
100-Hour Rupture Strength (psi)	1000	4,900	4,600	9,000

Note 1: Data at 800° and 1000°F for annealed and cold-worked conditions taken from Reference 1 for Heat L-984.

Note 2: There was practically no creep exhibited for the annealed condition at 600°F. Rupture strengths are similar to tensile strength. Total deformation strengths were based on the stress to give a deformation of 1 percent during loading.

TABLE 5

TENSILE AND CREEP-RUPTURE DATA FOR ACICULAR ALPHA TITANIUM (Ti 75A)

Iced Brine Quenched After 1 hour at 1800°F

Tensile-Test Data

Test Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	Elongation (% in 1 in)	Reduction of Area (%)
75	104,800	88,300	14.3	11.0
600	54,800	37,700	25.0	66.0
800	45,500	36,400	21.0	71.3
1000	37,200	27,900	46.0	89.0

Creep-Rupture Data

Test Temp. (°F)	Stress (psi)	Rupture Time (hours)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Time to Reach Indicated Total Deformation: Deformation (%)	Time to Reach Indicated Total Deformation: Hours	Minimum Creep Rate (in/in/hr)	100-Hour Rupture Strength (psi)
600	53,000	>1348.0	-	-	1.24	1.65	1010	2.2×10^{-6}	54,000
	50,000	>1153.2	-	-	1.02	1.15	1150	1.3×10^{-7}	
800	20,000	304.1	47.6	89.6	.17	0.2	20	2.0×10^{-7}	23,000
						0.25	125		
1000	8,000	133.4	45.6	94.2	.16	1.0	13	6.2×10^{-4}	9,000

> Test stopped without Rupture.

TABLE 6

TENSILE DATA FOR STABILIZED ALPHA 6 Al ALLOY (HEAT 4a)

Treatment	Test Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	Elongation (% in 1 in)	Reduction of Area (%)
As-Forged	75	119,800	109,000	16.8	26.6
12% Cold Reduction	75	154,000	137,900	11.0	30.1
2025°F W. Q.	75	127,800	115,000	17.0	34.5
As-Forged	600	80,100	65,300	24.0	53.7
12% Cold Reduction	600	102,000	97,300	11.6	46.8
2025°F W. Q.	600	77,700	60,700	17.0	43.5
As-Forged	1000	73,800	59,200	13.0	54.0
12% Cold Reduction	1000	80,600	58,400	15.7	44.8
2025°F W. Q.	1000	72,500	58,000	16.0	37.8

Note: 2025°F W. Q. = 2025°F (1 hr) + Water Quench

TABLE 7

CREEP-RUPTURE DATA FOR STABILIZED ALPHA 6 Al ALLOY IN THE AS-FORGED CONDITION (HEAT 4a)

Stress (psi)	Rupture Time (hours)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Time to Reach Specified Total Deformation (hours) 0.5%	1.0%	5%	Special (c)	3rd Stage* (hrs)	Minimum Creep Rate (in/in/hr)
					600°F					
80,100	Tensile Test	24.0	53.7	-	-	-	-	-	-	-
79,000	> 984.8	-	-	1.75	a	a	b	(1.94%) 700	-	3.1×10^{-7}
77,000	on load	15.3	45.2	-	-	-	-	-	-	-
76,000	>1252.9	-	-	2.36	a	a	b	(2.47%) 1100	-	1.5×10^{-7}
74,000	>1251.9	-	-	.63	a	b	b	(0.69%) 1200	-	1.1×10^{-7}
					1000°F					
73,800	Tensile Test	13.0	54.0	-	-	-	-	-	-	-
37,000	48.0	18.1	11.0	.29	<1	5	20	-	-	-
33,000	192.2	26.8	47.0	.28	5	12	85	(10%) 140	20	6.2×10^{-4}
28,000	516.2	44.1	71.3	.27	4	12	150	(10%) 300	160	2.5×10^{-4}
26,000	440 ± 10	82.0	73.8	.24	16	21	135	(10%) 250	200	3.7×10^{-4}
25,000	1057.8	40.0	57.4	.20	20	55	370	(10%) 705	450	1.2×10^{-4}
22,000	>1871	-	-	.18	50	140	1300	-	-	3.1×10^{-5}
20,000	> 956	-	-	.23	18	50	480	(10%) 950	600	1.4×10^{-5}
18,500	>1871	-	-	.15	60	260	b	(2%) 800	b	1.6×10^{-5}
16,000	>1270	-	-	.20	45	180	b	(3%) 1175	b	1.5×10^{-6}
15,000	>1078	-	-	.16	325	b	b	(0.75%) 1080	b	2.8×10^{-6}

* Transition to 3rd Stage (hours)
\> Test stopped without rupture
< Less than
a Exceeded deformation on loading
b Stopped before reaching deformation
c Specified deformation in parentheses

TABLE 8

CREEP-RUPTURE DATA FOR STABILIZED ALPHA 6 Al ALLOY IN THE COLD-WORKED CONDITION

Heat 4a - As-Forged plus 12-Percent Cold Reduction by Rolling

Stress (psi)	Rupture Time (hours)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Time to Reach Specified Total Deformation (hours) 0.5%	1.0%	5%	Special (c)	3rd Stage* (hrs)	Minimum Creep Rate (in/in/hr)
					600°F					
102,000	Tensile Test	11.6	46.8	-	-	-	-	-	-	-
100,000	on load	11.6	37.3	-	-	-	-	-	-	-
98,000	on load	10.9	34.3	-	-	-	-	-	-	-
90,000	> 940.9	-	-	.86	a	18	1.1% / 140	(1.2%) 430	(1.28%) 950	1.8×10^{-6}
85,000	>1196.4	-	-	.92	a	20	260	(1.2%) 1150	-	1.0×10^{-6}
36,600	>1297.0	-	-	.33	.35% / 100	.38% / 1200	-	-	-	-
					1000°F					
80,600	Tensile Test	15.7	49.8	-	-	-	-	-	-	-
30,000	82.3	58.1	77.3	.30	2	6	41	-	22	-
25,000	187.3	54.5	81.4	.29	2	7	58	-	45	6.0×10^{-4}
22,500	352.0	43.6	85.8	.23	3	8	47	-	70	7.2×10^{-4}
20,000	1301.7	68.2	85.0	.19	10	30	266	-	300	1.5×10^{-4}
17,000	> 912.9	-	-	.15	20	50	415	-	575	9.4×10^{-5}
12,500	>1025	-	-	.12	40	150	b	-	850	6.1×10^{-6}
10,000	>1102.4	-	-	.16	60	280	b	(2%) 985	900	1.3×10^{-5}
7,000	> 979.8	-	-	.13	110	750	b	-	>950	5.1×10^{-6}

* Transition to 3rd Stage (hours)
> Test stopped without rupture
a Exceeded deformation on loading
b Stopped before reaching deformation
c Specified deformation in parentheses

TABLE 9

CREEP-RUPTURE DATA FOR STABILIZED ALPHA 6 Al ALLOY IN THE SOLUTION-TREATED CONDITION

Heat 4a - Water Quenched after 1 Hour at 2025°F

Stress (psi)	Rupture Time (hours)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Time to Reach Specified Total Deformation (hours) 0.5%	1.0%	5%	Special (c)	3rd Stage* (hrs)	Minimum Creep Rate (in/in/hr)
					600°F					
77,700	Tensile Test	17.0	43.5	-	-	-	-	-	-	-
75,000	>1008.7	-	-	1.85	a	a	b	(1.88%) 600	-	3.0×10^{-7}
70,000	> 940.8	-	-	1.10	a	a	b	(1.12%) 800	-	1.2×10^{-9}
67,500	>1075.2	-	-	.89	-	-	-	(0.90%) 1000	-	nil
					1000°F					
72,500	Tensile Test	16.0	37.8	-	-	-	-	-	-	-
35,000	145.7	34.3	60.8	.45	1	10	73	(10%) 116	70	5.6×10^{-4}
32,500	276.2	48.6	61.5	.42	12	22	132	-	150	2.3×10^{-4}
30,000	476.0	26.9	34.5	.27	15	50	265	(10%) 395	175	1.4×10^{-4}
26,500	681.8	24.5	56.3	.28	10	60	385	(10%) 590	250	9.2×10^{-5}
25,000	1185.6	22.6	36.9	.26	25	95	645	(10%) 1040	350	6.1×10^{-5}
18,000	>1006.1	-	-	.22	85	380	b	-	550	1.2×10^{-5}
17,000	>1196.7	-	-	.20	115	775	b	-	900	4.8×10^{-6}
16,000	>1131.4	-	-	.16	310	b	(.55%)375	(.74%) 1000	850	2.2×10^{-6}
14,500	>1025.0	-	-	.13	650	b	(.55%)1000	-	-	1.2×10^{-6}

* Transition to 3rd Stage (hours)
> Test stopped without rupture
a Exceeded deformation on loading
b Stopped before reaching deformation
c Specified deformation in parantheses

TABLE 10

SUMMARIZED PROPERTIES FOR STABILIZED ALPHA ALLOYS

Alloy	Heat No.	Treatment	Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	Rupture Strengths 100-hours (psi)	Rupture Strengths 1000-hours (psi)	Total Deformation Strengths (psi) 100-hours 0.5%	100-hours 1%	100-hours 5%	1000-hours 1%	1000-hours 5%
6Al	4a	As-Forged	600	80,100	65,300	(79,000)	(79,000)	Total deformations at 600°F governed by loading deformation				
6Al	4a	12% Cold Reduction	600	102,000	97,300	(95,000)	(95,000)					
6Al	4a	Water Quench from 2025°F	600	77,700	60,700	(77,000)	(77,000)					
6Al	4	As-Forged	600	89,000	74,000	88,000	88,000					
6Al	4	10% Cold Reduction	600	97,800	94,000	96,000	96,000					
6Al	4	Water Quench from 2025°F	600	97,900	78,500	96,500	96,500					
6Al-0.5Si	7011	As-Forged	600	97,000	86,200	91,500	90,000					
6Al-0.5Si	7011	19% Cold Reduction	600	115,500	104,000	110,000	110,000					
6Al-0.5Si	7011	33% Cold Reduction	600	132,500	124,000	(~120,000)	(~120,000)	Insufficient data to derive values				
6Al-0.5Si	7011	Ice Brine Quench from 2100°F	600	99,000	81,800	95,000	95,000					
6Al	4a	As-Forged	1000	73,800	59,200	34,000	25,000	17,000	21,000	29,000	16,000	22,500
6Al	4a	As-Forged + 12% Cold Work	1000	102,000	58,400	29,000	20,000	7,000	14,000	22,000	6,000	--
6Al	4a	Water Quench from 2025°F	1000	72,500	58,000	37,000	25,000	17,000	25,000	34,000	17,000	23,000
6Al	4	As-Forged	1000	77,500	60,000	37,500	--	--	--	--	--	--
6Al	4	As-Forged + 10% Cold Work	1000	83,300	68,500	30,000	--	--	--	--	--	--
6Al	4	Water Quench from 2025°F	1000	80,300	64,800	40,000	--	--	--	--	--	--
6Al-0.5Si	7011	As-Forged	1000	74,800	60,700	44,000	32,000					
6Al-0.5Si	7011	19% Cold Reduction	1000	92,400	82,300	46,000	--	Insufficient data to derive values				
6Al-0.5Si	7011	33% Cold Reduction	1000	104,000	84,400	34,000	--					
6Al-0.5Si	7011	Ice Brine Quench from 2100°F	1000	91,600	75,300	41,000	(26,000)					

Heat 4 data taken from Reference 1.

TABLE 11

TENSILE DATA FOR STABILIZED ALPHA 6 Al-0.5 Si ALLOY

Condition	Test Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	Elongation (% in 1 in)	Reduction of Area (%)
As Forged	75	140,000	133,500	19.4	32.1
	600	97,000	86,200	15.7	21.8
	800	90,000	76,800	16.7	52.6
	1000	74,800	60,700	11.7	22.4
2100°F (1 hr) + Ice Brine Quench	75	148,400	132,000	16.0	36.0
	600	99,000	81,800	17.2	49.0
	1000	91,600	75,300	14.8	39.8
As Forged + 19% Cold Work	75	178,800	162,700	8.2	24.2
	600	115,000	104,000	7.6	27.2
	1000	92,400	82,300	11.7	34.9
As Forged + 33% Cold Work	75	193,100	170,000	6.0	19.1
	600	132,500	124,000	10.2	36.0
	1000	104,000	84,400	17.4	32.8
1925°F (1 hr) - Furnace Cool to 1100°F - hold 24 hr + Air Cool	75	138,000	131,800	14.0	21.3
1925°F (1 hr) + Water Quench + 1100°F (24 hr) + Air Cool	75	150,000	140,700	12.0	23.2
1100°F (24 hr) + Air Cool	75	143,400	133,200	14.3	26.7

TABLE 12

CREEP-RUPTURE DATA FOR STABILIZED ALPHA 6 Al - 0.5 Si ALLOY

Condition	Test Temp. (°F)	Stress (psi)	Time (hrs)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Time to Reach Specified Total Deformation (hours) 0.5%	1.0%	Special	Minimum Creep Rate (in/in/hr)	100-hr Rupture Strength (psi)
As Forged	600	97,000T	86,200Y	15.7	21.8	-	-	-	-	-	-
		95,000	0.1	17.4	36.6	1.42	-	-	-	-	-
		92,000	0.1	18.0	42.6	-	-	-	-	-	-
		89,000	>1131.3	-	-	-	-	-	-	9.1×10^{-8}	90,000
	800	90,000T	76,800Y	16.7	52.6	-	-	-	-	-	-
		85,000	853.3	9.3	14.6	2.50	-	-	(5%) 675	2.3×10^{-5}	-
		70,000	>1128.7	L	-	.58	-	900	-	3.8×10^{-6}	87,000
	1000	74,800T	60,700Y	11.7	22.4	-	-	-	-	-	-
		50,000	41.1	14.3	31.4	.47	0.2	4	-	1.4×10^{-3}	-
		36,000	473.9	15.0	26.9	.34	4	15	(4.5%) 290	1.2×10^{-4}	-
		30,000	> 986.1	-	-	.27	15	105	(4.5%) 970	2.8×10^{-5}	44,000
2100°F (1 hr) + Ice Brine Quench	600	99,000T	81,800Y	17.2	49.0	-	-	-	-	-	-
		93,000	>1123.5	-	-	2.60	-	-	(2.75%) 300	1.4×10^{-6}	-
		90,000	> 939.1	-	-	1.75	-	-	(1.77%) 940	3.1×10^{-7}	95,000
	1000	91,600T	75,300Y	14.8	39.8	-	-	-	-	-	-
		45,000	59.4	16.0	26.0	.44	-	-	-	-	-
		39,000	141.2	17.5	23.2	.34	4	15	(5%) 98	4.2×10^{-4}	-
		17,000	>1150	-	-	.13	380	-	-	3.6×10^{-6}	41,000
As Forged + 19% Cold Work	600	115,500T	104,000Y	7.6	27.2	-	-	-	-	-	-
		106,000	> 961.7	-	-	1.02	-	-	(1.2%) 800	4.2×10^{-7}	110,000
	1000	92,400T	82,300Y	11.7	34.9	-	-	-	-	-	-
		52,000	48.4	29.1	49.5	.55	-	14	-	3.3×10^{-4}	-
		44,000	126.0	35.6	53.6	.48	-	5	(5%) 46	7.7×10^{-4}	46,000
As Forged + 33% Cold Work	600	132,500T	124,000Y	10.2	36.0	-	-	-	-	-	-
		122,000	0.3	8.0	40.1	1.41 +	-	-	-	-	(~120,000)
	1000	104,000T	84,400Y	17.4	32.8	-	-	-	-	-	-
		63,000	4.7	31.4	35.8	.70	-	-	-	-	-
		40,000	41.9	15.7	13.9	.47	-	12	-	4.2×10^{-4}	34,000

T Tensile Strength
Y 0.2% Offset Yield Strength
> Test stopped without rupture

TABLE 13

COMPARATIVE TENSILE PROPERTIES FOR TWO HEATS OF

ALPHA-BETA ALLOY Ti 150A

Treatment	Heat No.	Test Temp (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	Elongation (% in 1 in)	Reduction of Area (%)
1500°F (1 hr) + Furnace Cool	L1006	75	153,800	147,600	26.7	35.2
	M739	75	138,500	130,600	25.5	42.8
	L1006	600	85,300	61,600	29.5	61.0
	M739	600	76,800	68,700	30.4	63.5
	L1006	800	68,300	49,500	41.7	74.3
	M739	800	66,900	51,500	37.0	80.2
	L1006	1000	37,400	28,500	68.0	93.5
	M739	1000	35,800	32,600	72.5	94.4
1500°F (1 hr) + Air Cool	L1006	75	177,200	162,500	16.7	32.0
	M739	75	153,000	135,600	11.0	13.2
	L1006	600	115,100	75,800	19.6	70.4
	M739	600	102,100	69,500	21.8	63.8
	L1006	800	96,100	56,000	36.5	80.9
	M739	800	82,400	54,400	33.3	79.5
	L1006	1000	46,000	25,200	91.5	97.0
	M739	1000	39,900	31,400	99.0	98.4
1800°F (1 hr) + Isothermal Trans-formation at 1300°F (1 hr) + Water Quench	L1006	75	167,800	164,200	19.8	40.0
	M739	75	156,500	153,500	16.0	21.2
	L1006	600	113,600	70,000	15.0	35.5
	M739	600	107,100	69,600	7.5	13.5
	L1006	800	92,700	59,000	26.0	33.3
	M739	800	85,700	56,000	10.3	19.7
	L1006	1000	48,500	35,500	110.7	99.0
	M739	1000	48,700	31,800	65.7	95.0

L1006 data taken from Reference 1.

TABLE 14

CREEP-RUPTURE DATA FOR ALPHA-BETA ALLOY Ti 150A (HEAT M739)

Treatment	Test Temp. (°F)	Stress (psi)	Time (hours)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Minimum Creep Rate (in/in/hr)	100-hours Rupture Strength (psi) Heat M739	Heat L1006
1500°F (1 hr) + Furnace Cool	600	76,800	Tensile Test	30.4	63.5	-	-	-	-
		72,000	>1172.7	-	-	5.64	7.3×10^{-6}	75,000	81,000
	800	66,900	Tensile Test	37.0	80.2	-	-	-	-
		45,000	72±22	59.8	75.5	.75	1.7×10^{-3}	-	-
		44,000	109.1	50.0	75.2	.59	-	44,000	44,000
	1000	35,800	Tensile Test	72.5	94.4	-	-	-	-
		11,500	130±10	64.0	93.6	-	-	-	-
		11,000	127.9	67.0	95.0	.20	1.2×10^{-3}	14,000	14,000
1500°F (1 hr) + Air Cool	600	102,100	Tensile Test	21.8	63.8	-	-	-	-
		95,260	>1134.0	-	-	3.9	3.3×10^{-5}	101,000	112,000
	800	82,400	Tensile Test	33.3	79.5	-	-	-	-
		52,000	86.0	29.0	77.0	.70	1.0×10^{-3}	-	-
		50,000	167.9	43.0	79.9	.35	6.6×10^{-4}	50,000	50,000
	1000	39,900	Tensile Test	99.0	98.4	-	-	-	-
		12,000	169.1	53.0	82.5	.18	4.2×10^{-4}	-	-
		11,000	268.4	73.3	96.2	.12	6.4×10^{-4}	14,000	14,000
1800°F (1 hr) Isothermal Trans. at 1300°F (1 hr) + Water Quench	600	107,100	Tensile Test	7.5	13.5	-	-	-	-
		110,000	2.5	9.0	9.3	2.28	-	-	-
		105,000	13.1	5.0	8.6	2.58	-	-	-
		100,000	129.3	4.9	8.7	2.26	1×10^{-4}	100,000	110,500
	800	85,700	Tensile Test	10.3	19.7	-	-	-	-
		55,000	99.3	29.0	70.8	.70	4.0×10^{-4}	-	-
		52,000	112.0	34.0	74.5	.55	6.3×10^{-4}	50,000	50,000
	1000	48,700	Tensile Test	65.7	95.0	-	-	-	-
		14,000	87±7.0	63.0	96.2	.25	3.3×10^{-3}	-	-
		11,000	628.8	52.1	94.8	.19	1.2×10^{-4}	14,000	14,000

\> Test stopped without rupture

TABLE 15

COMPARISON OF CREEP PROPERTIES FOR TWO HEATS OF ALPHA-BETA ALLOY Ti 150A

Treatment	Heat No.	Temp. (°F)	Stress (psi)	Time (hours)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Time for Specified Total Deformation (hours)			Minimum Creep Rate (in/in/hr)
1500°F (1 hr) + Furnace Cool	M739	600	71,870	1172.7	8.8	10.1	5.64	(7%)225	(8%)1000	-	7.3×10^{-6}
	L1006	600	80,000	791.6	11.6	-	Off scale on loading			-	2.0×10^{-5}
	M739	800	23,000	1146.0	1.9	2.4	.25	(0.5%)40	(.75%)120	(1%)235	1.0×10^{-5}
	L1006	800	29,000	1534.7	8.7	9.4	.30	(0.5%)5	(.75%)12	(1%)40	3.9×10^{-5}
	M739	1000	5,000	1028.0	1.5	0.8	.07	(0.5%)250	(.75%)~1050	-	1.8×10^{-6}
	L1006	1000	5,000	860.4	1.0	1.0	.19	(0.5%)35	(.75%)115	(1%)255	4.8×10^{-5}
1500°F (1 hr) + Air Cool	M739	600	95,260	1134.0	-	-	3.9	(5%)18	(6%)80	(7%)165	3.3×10^{-5}
	L1006	600	108,000	1992.9	-	-	2.2	(5%)90	(6%)165	(7%)280	4.3×10^{-5}
	M739	800	35,000	1028	4.0	7.2	.45	(0.5%)10	(.75%)13	(1%)18	1.3×10^{-5}
	L1006	800	35,000	1190.0	3.8	2.4	.35	(0.5%)6	(.75%)10	(1%)15	5.9×10^{-6}
	M739	1000	5,000	1028.0	1.5	0.8	.08	(0.5%)125	(.75%)475	(1%)1005	4.6×10^{-6}
	L1006	1000	5,000	979.7	4.8	1.0	.08	(0.5%)23	(.75%)63	(1%)205	8.9×10^{-6}
1800°F (1 hr) Isothermal Trans. at 1300°F (1 hr) + Water Quench	M739	600	100,000	129.3*	-	-	-	-	-	-	-
	L1006	600	105,000	2230	8.8	0	1.7	(2%)20	(4%)300	(5%)500	6.7×10^{-5}
	M739	800	35,000	1028	4.0	3.2	.44	(0.5%)6	(.75%)10	(1%)14	2.1×10^{-5}
	L1006	800	35,000	1188	4.9	3.2	.51	-	(1%)10	(2%)15	7.6×10^{-6}
	M739	1000	5,000	1028	-	-	.12	(0.5%)30	(.75%)235	(1%)>1028	1.4×10^{-6}
	L1006	1000	5,000	2639	1.0	1.0	.18	(0.5%)31	(.75%)90	(1%)380	4.4×10^{-6}

* Test failed

TABLE 16

TENSILE DATA FOR ALPHA-BETA ALLOY Ti 155AX

Condition	Test Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	Elongation (% in 1 in)	Reduction of Area (%)
As Produced	75	157,500	150,000	20.4	49.2
	600	112,200	95,400	22.7	57.4
	800	102,800	85,800	25.8	53.8
	1000	74,400	55,000	49.6	93.5
1200°F (24 hr) + AC	75	153,700	152,000	19.6	48.8
1700°F (1 hr) + FC	75	149,000	146,000	22.4	43.4
	600	103,000	77,000	23.0	51.8
	1000	71,500	59,000	41.0	77.5
1700°F (1 hr) + WQ	75	207,000	151,000	13.0	16.8
	600	195,000	142,000	10.3	10.7
	1000	97,500	63,600	55.4	97.0
1700°F (1 hr) + WQ + 800°F (1 hr) + WQ	75	262,000	226,000	3.0	3.2
1800°F (1 hr) + WQ	75	182,200	182,200	2.0	1.6
1800°F (1 hr) + WQ + 1200°F (15 min.)+WQ	75	186,200	185,300	4.0	1.6
	600	159,000	140,000	6.0	6.3
	1000	106,000	68,200	55.0	95.0
1800°F (1 hr) + WQ + 800°F (20 min.)+WQ	75	Broke in threads at gage section stress of 166,000 psi			
1800°F (1 hr) - Isothermal Trans. 1200°F (15 min.) + WQ	75	172,000	161,000	10.8	17.5
	600	126,000	108,000	10.8	25.2
	1000	85,500	69,800	51.0	95.0
1800°F (1 hr) - Isothermal Trans. 800°F (20 min.) + WQ	75	194,000	180,000	1.0	2.4

AC - Air Cool
FC - Furnace Cool
WQ - Water Quench

TABLE 17

CREEP-RUPTURE DATA FOR ALPHA-BETA ALLOY Ti 155AX

Condition	Test Temp. (°F)	Stress (psi)	Time (hrs)	Elongation (% in 1 in.)	Reduction of Area (%)	Loading Deformation (%)	Time to Reach Specified Total Deformation (hrs) 0.5%	1.0%	Special	Min. Creep Rate (in/in/hr)	100-hr Rupture Strength (psi)
As Produced	600	112,200 T	95,400 Y	22.7	57.4	-	-	-	-	-	-
		109,000	>1512.8	-	-	-	-	-	-	1.0×10^{-5}	-
		105,000	>1199.0	-	-	2.41	-	-	(3.2%)1050	2.6×10^{-6}	-
		100,000	>1053.6	-	-	1.62	-	-	(2.25%)800	$<10^{-8}$	111,000
	800	102,800 T	85,800 Y	25.8	53.8	-	-	-	-	-	-
		77,000	117.5	43.1	75.5	.39	-	-	-	-	-
		76,000	121.9	37.0	72.3	.70	-	-	(5%)27	1.2×10^{-3}	-
		55,000	>1151.0	-	-	.47	-	30	(3%)635	1.8×10^{-5}	78,000
	1000	74,400 T	55,000 Y	49.6	93.5	-	-	-	-	-	-
		26,000	33 ± 14	81.4	90.4	.34	-	-	-	-	-
		23,000	49.3	69.3	92.5	.27	-	-	-	-	-
		15,000	157.4±7	95.1	95.0	.16	-	-	-	-	18,000
1200°F (24 hr) + Air Cool	1000	18,000	107.7	97.0	92.7	.40	-	-	(5%)23	2.0×10^{-3}	18,000
1700°F (1 hr) + Furnace Cool	600	103,000 T	77,000 Y	23.0	51.8	-	-	-	-	-	-
		100,000	>982.7	-	-	5.40	-	-	(6.5%)80	6.2×10^{-6}	102,000
	1000	71,500 T	59,000 Y	41.0	77.5	-	-	-	-	-	-
		27,000	59±7	53.4	91.9	.32	-	-	-	-	-
		21,000	390.5	71.0	83.7	.28	5	15	(5%)160	2.5×10^{-4}	-
		12,000	>382	-	-	.11	13	45	(5%)320	1.3×10^{-4}	26,000
1700°F (1 hr) + Water Quench	600	195,000 T	142,000 Y	10.3	10.7	-	-	-	-	-	-
		190,000	<0.1	7.7	12.4	>2.25	-	-	-	-	-
		185,000	558.6	6.0	7.1	2.25	-	-	(5%)148	3.3×10^{-5}	187,000
	1000	97,500 T	63,600 Y	55.4	97.0	-	-	-	-	-	-
		30,000	28.3	66.0	94.2	.80	-	-	-	-	-
		21,000	121.8	106.0	95.6	.28	1	4	(5%)23	1.7×10^{-3}	-
		10,000	>1101.2	-	-	.12	15	40	(5%)500	6.7×10^{-5}	22,000
1800°F (1 hr) + Water Quench + 1200°F (15 min) + WQ	600	159,000 T	140,000 Y	6.0	6.3	-	-	-	-	-	-
		153,000	0.1	2.0	7.9	1.41	-	-	-	-	-
		150,000	821.0	3.9	9.5	2.44	-	-	(3.6)800	5.0×10^{-6}	152,000
	1000	106,000 T	68,000 Y	55.0	95.0	-	-	-	-	-	-
		26,000	109 ±8	65.0	89.4	.33	2	14	(5%)56	4.8×10^{-4}	-
		10,000	>1150	-	-	.15	20	200	-	1.8×10^{-5}	26,000
1800°F (1 hr) + Iso. Trans. 1200°F (15 min) + WQ	600	126,000 T	108,000 Y	10.8	25.2	-	-	-	-	-	-
		121,000	>1436.7	-	-	4.24	-	-	(5%)900	3.5×10^{-6}	123,000
	1000	85,500 T	69,800 Y	51.0	95.0	-	-	-	-	-	-
		26,000	100.0	58.0	93.4	.30	-	-	-	-	-
		24,000	120.2	92.0	94.7	.31	2	7	(5%)50	4.2×10^{-4}	26,000

T Tensile Test
Y 0.2% Yield Strength
> Stopped without rupture

TABLE 18

SUMMARIZED PROPERTIES FOR ALPHA-BETA ALLOYS

Alloy	Heat No.	Treatment	Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	100-hr Rupture Strength (psi)	Estimated Stress for Min. Creep Rate of 5 x 10^{-6} (in/in/hr) (psi)
Ti 150A	L1006	As Produced	600	90,600	63,000	87,000	79,500
Ti 155AX	M1400R	As Produced	600	112,200	95,400	111,000	105,500
Ti 150A	L1006	As Produced	800	69,900	50,000	39,000	23,000
Ti 155AX	M1400R	As Produced	800	102,800	85,300	78,000	50,000
Ti 150A	L1006	As Produced	1000	37,800	29,600	12,500	5,000
Ti 155AX	M1400R	As Produced	1000	74,400	55,000	18,000	5,000
Ti 150A	M739	1500°F (1 hr) + FC	600	76,800	68,700	75,000	71,750
Ti 150A	L1006	1500°F (1 hr) + FC	600	85,300	61,600	81,000	78,000
Ti 155AX	M1400R	1700°F (1 hr) + FC	600	103,000	77,000	102,000	100,000
Ti 150A	M739	1500°F (1 hr) + FC	1000	35,800	32,600	14,000	-
Ti 150A	L1006	1500°F (1 hr) + FC	1000	37,400	28,500	14,000	5,000
Ti 155AX	M1400R	1700°F (1 hr) + FC	1000	71,500	59,000	26,000	-
Ti 150A	L1006	1500°F (1 hr) + WQ	600	162,300	152,800	155,000	120,000
Ti 155AX	M1400R	1700°F (1 hr) + WQ	600	195,000	142,000	187,000	-
Ti 150A	L1006	1500°F (1 hr) + WQ	1000	46,700	31,800	11,000	5,000
Ti 155AX	M1400R	1700°F (1 hr) + WQ	1000	97,500	63,600	22,000	-
Ti 150A	M739	1800°F (1 hr) + Iso. Trans. 1300°F (1 hr)	600	107,100	69,600	100,000	-
Ti 150A	L1006	1800°F (1 hr) + Iso. Trans. 1300°F (1 hr)	600	113,600	70,000	110,500	100,000
Ti 155AX	M1400R	1800°F (1 hr) + Iso. Trans. 1200°F (15 min)	600	126,000	108,000	123,000	121,000
Ti 150A	M739	1300°F Iso. Trans.	1000	48,700	31,800	14,000	5,500
Ti 150A	L1006	1300°F Iso. Trans.	1000	48,500	35,500	14,000	5,000
Ti 155AX	M1400R	1200°F Iso. Trans.	1000	85,500	69,800	26,000	-
Ti 150A	L1006	1500°F WQ (1 hr) + 1350°F (1 hr)	600	86,500	65,800	84,000	78,000
Ti 150A	L1006	1500°F AC (1 hr) + 900°F (1 hr)	600	126,500	84,500	110,000	100,000
Ti 155AX	M1400R	1800°F WQ (1 hr) + 1200°F (15 min)	600	159,000	140,000	152,000	150,000
Ti 150A	L1006	1500°F WQ (1 hr) + 1350°F (1 hr)	1000	39,500	32,300	11,500	5,000
Ti 150A	L1006	1500°F AC (1 hr) + 900°F (1 hr)	1000	43,500	25,800	13,500	5,000
Ti 155AX	M1400R	1800°F WQ (1 hr) + 1200°F (15 min)	1000	106,000	68,200	26,000	6,000

TABLE 19

HARDNESS OF META-STABLE BETA ALLOY 10 Mo

AFTER HEAT TREATMENT

(Rockwell C Hardness)

As-Forged = 32.1

As-Forged + Reheated to 800°F for Indicated Time + Air Cool	30 min.	1 hr.	5 hr.	10 hr.
	34.8	49	48	43.1

As-Forged + Reheated to 1000°F for Indicated Time + Air Cool	10 min.	20 min.	30 min.	5 hr.	24 hr.
	41	41	41.7	38.7	36

As-Forged + Reheated as Indicated:

30 minutes at 1100°F, + Water Quench	33
24 hours at 1350°F + Water Quench	28
25 hours at 1425°F + Water Quench	28
30 minutes at 1450°F + Water Quench	27
30 minutes at 1800°F + Water Quench	23.7

1800°F - 30 minutes plus Indicated Treatment:

Water Quench + Reheat to 1300°F for 15 minutes, Water Quench	33.3
Isothermal Transformation at 1025°F for 15 seconds, Water Quench	27

1800°F - 1 hour plus Isothermal Transformation:

1200°F - 1 hour + Water Quench	26.2
950°F - 15 minutes + Water Quench	38.8

TABLE 20

TENSILE DATA FOR META-STABLE BETA ALLOY 10 Mo

Treatment	Test Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	Elongation (% in 1 in)	Reduction of Area (%)
As Forged	75	114,100	106,800	40.3	65.3
	600	107,100	-	11.8	45.6
	800	151,000	143,000	5.1	12.8
	800	132,800	131,800	6.0	28.9
	1000	96,000	48,400	20.2	32.8
1800°F (1 hr) + WQ (Water Quench)	75	107,200	81,500	48.0	80.6
	600	101,200	99,800	9.0	47.0
	1000	88,700	61,000	9.8	13.9
1800°F (1 hr) + WQ + 1300°F (15 min.)+WQ	75	148,500	141,800	7.0	18.2
1425°F (24 hr) + WQ	75	112,000	101,300	39.0	66.7
1300°F (24 hr) + WQ	75	131,200	125,500	18.0	50.1
	600	91,800	83,500	16.5	39.1
	1000	61,000	49,000	11.8	32.8
1800°F (1 hr) - Isothermal Trans. 1200°F (1 hr) + WQ	75	109,600	89,200	34.0	44.0
	600	90,100	65,900	15.7	46.3
	1000	71,300	57,400	11.1	13.8
1800°F (1 hr) - Isothermal Trans. 950°F (15 min.) + WQ	75	178,300	156,800	4.0	10.2
	600	144,000	124,500	10.8	12.3
	1000	92,700	71,300	6.9	7.9

TABLE 21

CREEP-RUPTURE DATA FOR META-STABLE BETA ALLOY 10 Mo

Condition	Test Temp. (°F)	Stress (psi)	Time (hrs)	Elongation (% in .1 in.)	Reduction of Area (%)	Load Deformation (%)	Time to Reach Specified Total Deformation (hrs) 0.5%	1.0%	Special	Min. Creep Rate (in/in/hr)	100-hr Rupture Strength (psi)
As Forged	600	107,100 T	-	11.8	45.6	-	-	-	-	-	-
		105,000	>1442.5	-	-	1.25	-	-	(1.75%)700	2.4×10^{-6}	106,000
	800	151,000 T	143,000 Y	5.1	12.8	-	-	-	-	-	-
		132,800 T	131,800 Y	6.0	28.9	-	-	-	-	-	-
		125,000	1.0	7.1	19.7	1.43	-	-	-	-	-
		100,000	68.1	14.0	14.7	1.59	-	-	(2%)14	2.9×10^{-4}	98,000
	1000	96,000 T	48,400Y	20.2	32.8	-	-	-	-	-	-
		57,000	1.6	23.5	30.8	1.23	-	-	-	-	-
		40,000	13.9	29.1	36.6	.55	-	-	-	-	-
		26,500	45.3	38.6	71.5	.37	-	-	-	-	-
		14,000	502.4	44.8	77.2	.19	-	12	(5%)128	1.9×10^{-4}	22,000
1800°F (1 hr) + Water Quench	600	101,200 T	99,800 Y	9.0	47.0	-	-	-	-	-	-
		99,000	>1077.0	-	-	.80	-	-	(.9%)60	nil	-
		98,000	> 962.5	-	-	.70	-	-	-	nil	100,000
	1000	88,700 T	61,000 Y	9.8	13.9	-	-	-	-	-	-
		25,000	71.8	25.7	38.6	.40	-	4	(5%)30	1.5×10^{-3}	-
		21,000	162.8	47.0	45.7	.35	-	6	(5%)48	8.3×10^{-4}	-
		7,500	>1120	-	-	.13	50	330	-	1.0×10^{-5}	23,500
1300°F (24 hr) + Water Quench	600	91,800 T	83,500 Y	16.5	39.1	-	-	-	-	-	-
		90,500	694.5	9.0	30.5	.99	-	-	(.99%)645	nil	-
		89,500	> 963.6	-	-	.52	-	5	(3%)25	nil	91,000
	1000	61,000 T	49,000 Y	11.8	32.8	-	-	-	-	-	-
		21,000	92.8	37.0	64.2	.30	-	6	(5%)35	1.1×10^{-3}	-
		18,000	187.6	46.6	70.3	.17	-	15	(5%)93	4.4×10^{-4}	-
		6,000	>1009.8	-	-	.09	90	380	-	1.1×10^{-5}	20,200
1800°F (1 hr) - Iso. Trans. 1200°F (1hr) + WQ	600	90,100 T	65,900Y	15.7	46.3	-	-	-	-	-	-
		89,500	>1170.6	-	-	1.15	-	-	-	nil	-
		88,500	> 938.1	-	-	3.20	-	-	-	nil	89,800
	1000	71,300 T	57,400Y	11.1	13.8	-	-	-	-	-	-
		24,500	81.5	32.4	36.1	.32	2	7	(5%)45	1×10^{-3}	-
		22,000	122.4	25.7	35.5	.24	9	-	-	2.5×10^{-4}	-
		8,000	>1120	-	-	.09	67	215	(2%)525	1.0×10^{-5}	23,500
1800°F (1 hr) - Iso. Trans. 950°F (15 min) + WQ	600	144,000 T	124,500 Y	10.8	12.3	-	-	-	-	-	-
		140,000	118.2	2.9	4.7	2.03	-	-	(2.3%)20	nil	140,000
		138,000	166.1	2.0	5.5	1.87	-	-	(2.3%)25	1.5×10^{-5}	-
	1000	92,700 T	71,300 Y	6.9	7.9	-	-	-	-	-	-
		25,000	54.2	40.0	36.0	.32	-	3	(5%)16	2.8×10^{-3}	-
		20,000	200.8 •	33.7	41.7	.20	-	26	(5%)77	7.7×10^{-4}	-
		7,000	>1120	-	-	.11	13	60	(2%)300	2.4×10^{-5}	22,500

T Tensile Strength
Y 0.2% Offset Yield Strength
> Stopped without rupture

TABLE 22

INFLUENCE OF HEAT TREATMENT ON HARDNESS OF

META-STABLE BETA ALLOY 10 Cr

As Forged			
Temp. (°F)	Time (hours)	Cooling	Hardness Rockwell "C"
As Forged			39.0
1900	1	IBQ	41.2
1800	0.5	WQ	35.0
1335	4	WQ	27.3
1335	24	WQ	34.0
1300	4	WQ	30.7
1265	4	WQ	42.3
1000	0.17	AC	41.7
1000	0.33	AC	41.0
1000	0.5	AC	40.3
1000	5	AC	38.3
1000	24	AC	34.7
800	0.5	AC	50.2
800	1	AC	49.9
800	5	AC	50.0
800	10	AC	46.9
600	0.5	AC	44.8
600	1	AC	45.5
600	5	AC	45.6
600	10	AC	45.7

1800°F - 1 Hour - Water Quench		
Temp. (°F)	Time (minutes)	Hardness Rockwell "C"
1600	15	39.0
1400	15	37.7
1300	10	33.3
1200	15	34.7
1000	15	40.7
800	15	49.7
800	(24 hours)	47.3

Heated 30 minutes at 1800°F plus isothermal treatment at 1470°F for 15 minutes and water quenched = 37.3 R_C

TABLE 23

TENSILE DATA FOR META-STABLE BETA ALLOY 10 Cr

Condition	Test Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% offset (psi)	Elongation (% in 1 in)	Red. of Area (%)
As Forged	75	166,500	156,500	11.5	42.8
	600	138,600	--	1.0	1.6
	800	156,500	132,700	17.8	23.9
	1000	58,500	36,400	72.8	90.2
1900°F (1 hr) + Ice Brine Quench	75	>156,000	>156,000	10.8	39.4
	600	140,000	135,000	4.0	8.6
	1000	49,300	39,000	77.0	95.7
1800°F (30 min) + Water Quench	75	160,000	156,500	12.2	33.6
1800°F (1 hr)-Iso. Trans. 1470°F (15 min)+WQ	75	151,300	145,000	13.3	33.9
1800°F (30 min)+WQ + 1300°F (10 min) + WQ	75	158,000	156,200	9.5	17.5
	600	122,000	122,000	2.0	0
	1000	62,000	41,000	59.0	96.2
1800°F (1 hr) + WQ + 1200°F (15 min) + WQ	75	145,900	139,500	8.8	21.8
1800°F (1 hr) + WQ + 1000°F (15 min) + WQ	75	180,000	174,500	2.0	3.3
1800°F (1 hr) + WQ + 800°F (15 min) + WQ	75	Failed in threads at gage section stress of 127,000 psi			
1335°F (4 hrs) + WQ	75	158,000	151,000	5.1	9.6
1335°F (24 hrs) + WQ	75	153,800	149,000	6.9	11.6
	600	121,000	--	2.0	1.6
	1000	49,100	41,100	120.0	99.0
1265°F (4 hr) + WQ	75	143,500	138,300	9.4	27.5
1265°F (24 hrs) + WQ	75	154,000	151,000	2.9	10.9
800°F (24 hrs) + WQ	75	164,000	164,000	1.0	1.0
1335°F (4 hrs) + WQ + 800°F (24 hrs) + WQ	75	Failed in threads at gage section stress of 112,000 psi			
1265°F (4 hrs) + WQ + 800°F (24 hrs) + WQ	75	71,200	71,200	< 1.0	1.6

TABLE 24

CREEP-RUPTURE DATA FOR META-STABLE BETA ALLOY 10 Cr

Condition	Test Temp. (°F)	Stress (psi)	Time (hrs)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Time to Reach Specified Total Deformation (hrs)	Minimum Creep Rate (in/in/hr)	100-hr Rupture Strength (psi)
As Forged	600	138,600T	-	1.0	1.6	-	-	-	-
		137,000	801.7	6.2	7.9	1.31	(2%)330	6.7×10^{-6}	138,000
	800	156,500T	132,700Y	17.8	23.9	-	-	-	-
		120,000	0.8	18.0	44.0	1.18	-	-	-
		77,000	206.0	22.5	37.8	.78	-	-	-
		66,000	982.5	20.0	32.7	-	-	1.9×10^{-5}	80,000
	1000	58,500T	36,400Y	72.8	90.2	-	-	-	-
		24,000	23.7	44.0	88.5	.46	-	-	-
		20,000	164.6	41.0	92.8	-	-	1.3×10^{-4}	-
		14,000	1102±5	39.1	88.[illegible]	-	-	4.1×10^{-5}	21,000
1900°F (1 hr) + Ice Brine Quench	600	140,000T	135,000Y	4.0	8.6	-	-	-	-
		138,500	<0.1	1.0	0.8	>.89	-	-	-
	1000	49,300T	39,000Y	77.0	95.7	-	-	-	-
		22,500	13.9±7	85.0	96.6	.40	-	-	-
		19,000	302.2	35.0	86.9	.28	-	5.0×10^{-5}	20,000
1335°F (24 hr) + Water Quench	600	121,000T	-	2.0	1.6	-	-	-	-
		119,500	>1003.7	-	-	.85	(.9%)310 (1.0%)775 (1.05%)1000	1.4×10^{-6}	120,000
	1000	49,100T	41,100Y	120.0	99.0	-	-	-	-
		21,000	8.1	51.0	90.3	.58	-	-	-
		19,000	93.5	37.0	89.0	.46	-	-	-
		18,000	204.2	48.0	92.8	.34	-	-	19,000

T Tensile Strength
Y 0.2% Offset Yield Strength
> Test stopped without rupture

TABLE 25

COMPARATIVE PROPERTIES FOR META-STABLE BETA ALLOYS

Alloy	Treatment	Temp. (°F)	Tensile Strength (psi)	Yield Strength (psi)	Rupture Strength 100-hour (psi)	Rupture Strength 1000-hour (psi)
10 Mo	As Forged	75	114,100	106,800	-	-
10 Cr	As Forged	75	166,500	156,500	-	-
10 Mo	1800°F - WQ	75	107,200	81,500	-	-
10 Cr	1900°F - IBQ	75	>156,000	>156,000	-	-
10 Mo	1300°F - WQ	75	131,200	125,500	-	-
10 Cr	1325°F - 24 hr	75	153,800	149,000	-	-
10 Mo	1800°F 1200°F - IT	75	109,600	89,200	-	-
10 Mo	1800°F 950°F - IT	75	178,300	156,800	-	-
10 Mo	As Forged	600	107,100	-	106,000	106,000
10 Cr	As Forged	600	138,600	-	138,000	136,000
10 Mo	1800°F - WQ	600	101,200	99,800	100,000	100,000
10 Cr	1900°F - IBQ	600	140,000	135,000	(138,000 est)	-
10 Mo	1300°F - WQ	600	91,800	83,500	90,000	90,000
10 Cr	1325°F - 24 hr	600	121,000	-	120,000	120,000
10 Mo	1800°F 1200°F - IT	600	90,100	65,900	89,800	89,800
10 Mo	1800°F 950°F - IT	600	144,000	124,500	140,000	(137,000)
10 Mo	As Forged	800	151,100	143,000	98,000	-
10 Cr	As Forged	800	156,500	132,700	80,000	65,000
10 Mo	As Forged	1000	96,000	48,400	22,000	(~15,000)
10 Cr	As Forged	1000	58,500	36,400	21,000	15,000
10 Mo	1800°F - WQ	1000	88,700	61,000	23,500	(~15,000)
10 Cr	1900°F - IBQ	1000	49,300	39,000	20,000	18,000
10 Mo	1300°F - 24 hr	1000	61,000	49,000	20,200	(~15,000)
10 Cr	1325°F - 24 hr	1000	49,100	41,100	19,000	16,000
10 Mo	1800°F 1200°F - IT	1000	71,300	57,400	23,500	(~15,000)
10 Mo	1800°F 950°F - IT	1000	92,700	71,300	22,500	(~15,000)

TABLE 26

CREEP-RUPTURE AND TENSILE DATA FOR STABLE BETA ALLOY 30 Mo IN THE AS-FORGED CONDITION

Test Temp. (°F)	Stress (psi)	Rupture Time (hours)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	Time to Reach Specified Total Deformation (hours) 0.5%	1.0%	5%	Special (c)	Transition to 3rd Stage (hours)	Minimum Creep Rate (in/in/hr)
75	144,700T	142,000Y	14.0	27.4	-	-	-	-	-	-	-
600	*106,000T	96,000Y	13.3	39.7	-	-	-	-	-	-	-
	105,000	>1199.4	-	-	3.03	b	b	-	(3.07%)1150	b	5.0×10^{-7}
	104,500	>1509.1	-	-	3.57	b	b	-	-	b	-
	100,000	>1151	-	-	.91	a	b	-	(.98%)1100	b	4.0×10^{-7}
	98,000	>1555	-	-	.78	a	b	-	(.79%)1000	b	3.3×10^{-7}
	* 90,000	>1158	-	-	.36	b	b	-	-	b	$<10^{-8}$
1000	* 83,200T	74,500Y	25.5	59.4	-	-	-	-	-	-	-
	* 40,000	77±2	54.8	49.4	.36	1	7	38	-	-	6.7×10^{-4}
	* 35,000	133.0	62.8	70.0	.34	6	20	60	-	-	2.0×10^{-4}
	29,500	327.5	72.0	64.0	.32	20	60	160	-	20	8.3×10^{-5}
	25,250	650.9	91.3	64.4	.25	30	80	250	-	60	1.0×10^{-4}
	22,500	1261.4	57.3	61.2	.18	150	215	565	-	90	1.7×10^{-5}
	21,000	1351.7	59.6	68.2	.24	200	290	625	-	110	5.6×10^{-6}
	* 20,000	> 912.5	-	-	.15	238	360	b	-	145	7.1×10^{-6}
	18,000	> 740	-	-	.14	265	500	b	-	190	7.4×10^{-6}
	16,500	>1122	-	-	.14	205	585	b	-	1100+	1.2×10^{-5}
	14,000	>2000	-	-	.17	630	1975	b	-	2100+	3.3×10^{-6}

* Data from Reference 1
\> Test stopped without rupture
a Exceeded deformation on loading
b Stopped before reaching deformation
c Reported deformation in parentheses
T Tensile Strength
Y Yield Strength

TABLE 27

TENSILE AND CREEP-RUPTURE DATA FOR STABLE BETA ALLOY 50 V

Condition	Test Temp. (°F)	Stress (psi)	Time (hrs)	Elongation (% in 1 in)	Reduction of Area (%)	Loading Deformation (%)	1.0%	2.0%	Special	Minimum Creep Rate (in/in/hr)	100-hr Rupture Strength (psi)
As Forged	75	118,800T	115,000Y	12.8	19.6	-	-	-	-	-	-
	600	107,300T	86,300Y	14.3	25.1	-	-	-	-	-	-
		105,000	0.1	12.4	29.3	-	-	-	-	-	-
		100,000	>1107.3	-	-	2.15	-	-	(2.22%)600	2.7×10^{-7}	103,000
	1000	93,600T	79,000Y	9.8	26.1	-	-	-	-	-	-
		40,000	110.6	31.0	27.2	.41	9	25	-	6.2×10^{-3}	-
		30,000	346.6	39.8	42.6	.30	49	70	(3%)92	3.6×10^{-5}	-
		21,000	> 620.0	-	-	.22	105	200	(3%)300	1.2×10^{-5}	40,000
1925°F (2 hr) + Water Quench	75	127,900T	123,000Y	14.0	28.7	-	-	-	-	-	-
	600	116,500T	98,000Y	11.0	17.2	-	-	-	-	-	-
		110,000	>1663.3	-	-	5.91	-	-	(6%)50	nil	112,000
	1000	102,600T	90,000Y	10.7	33.6	-	-	-	-	-	-
		45,000	100.6	34.0	26.0	.52	-	-	-	-	-
		40,000	173.6	49.0	36.1	.40	27	34	-	7.7×10^{-4}	-
		15,000	>1150	-	-	.15	350	750	-	2.8×10^{-6}	45,000

T Tensile Strength
Y 0.2% Offset Yield Strength
> Stopped without rupture

TABLE 28

COMPARATIVE TENSILE PROPERTIES FOR STABLE BETA ALLOYS

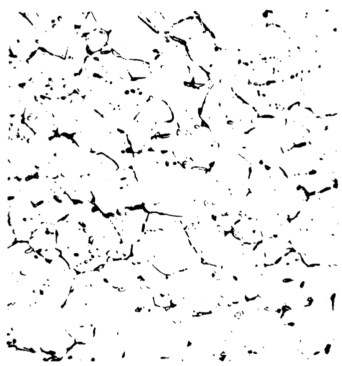
Alloy	Test Temp. (°F)	Tensile Strength (psi)	Yield Strength 0.2% Offset (psi)	Elongation (% in 1 in)	Reduction of Area (%)
			As Forged		
30 Mo	75	144,700	142,000	14.0	27.4
50 V	75	118,800	115,000	12.8	19.6
30 Mo	600	106,000	96,000	13.3	39.7
50 V	600	107,300	86,300	14.3	25.1
30 Mo	1000	83,200	74,500	25.5	59.4
50 V	1000	93,600	79,000	9.8	26.1
			Water Quenched after 2 Hours at 1925°F		
50 V	75	127,900	123,000	14.0	28.7
50 V	600	116,500	98,000	11.0	17.2
50 V	1000	102,600	90,000	10.7	33.6

Note: The data for 30 Mo alloy were taken from Reference 1. This reference gives tensile data for the 30 Mo alloy after the following heat treatments: 1325°F, 1 hour, W.Q.; 1500°F, 2 hours, W.Q.; 1500°F, 2 hours, W.Q. + 1375°F, 24 hours, W.Q.; 1325°F, 1 hour + W.Q. + 3.3% cold reduction. The tensile properties after these treatments did not vary signficantly from those given above for the as-forged condition.

TABLE 29

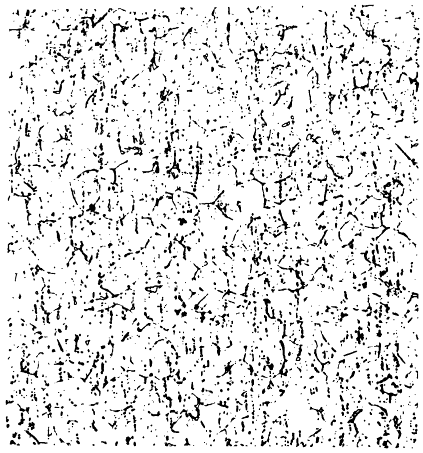
COMPARATIVE RUPTURE AND CREEP PROPERTIES OF STABLE BETA ALLOYS AT 1000°F

		Rupture Properties				Stress for a	Total Deformation Strengths							
		100-hour		1000-hour		Creep Rate	100-hour				1000-hour			
Alloy	Treatment	Strength (psi)	Elongation (%)	Strength (psi)	Elongation (%)	of 10^{-5} in/in/hr (psi)	0.5% (psi)	1% (psi)	2% (psi)	5% (psi)	0.5% (psi)	1% (psi)	2% (psi)	5% (psi)
30 Mo	As forged	37,500	60	23,000	60	19,000	23,500	26,000	29,500	33,000	12,000	15,000	19,000	20,500
50 V	As forged	40,000	30	--	--	21,000	(18,000)	21,000	26,000	(30,000)	--	--	--	--
30 Mo	1500°F, W.Q.	37,000	--	--	--	--	--	--	--	--	--	--	--	--
50 V	1925°F, W. Q.	45,000	34	--	--	21,000	(21,000)	26,000	32,000	--	--	--	-	--



X250

Original



X100

Tested at 600°F for 984.6 Hours under 31,500 psi

(a) Annealed at 1500°F for 1 Hour and Furnace Cooled

X100

Original

X100

Tested at 600°F for 1252 Hours under 44,000 psi

(b) Annealed at 1500°F + 31-Percent Reduction by Rolling

Figure 1. - Microstructures of Alpha Titanium (Ti 75A).

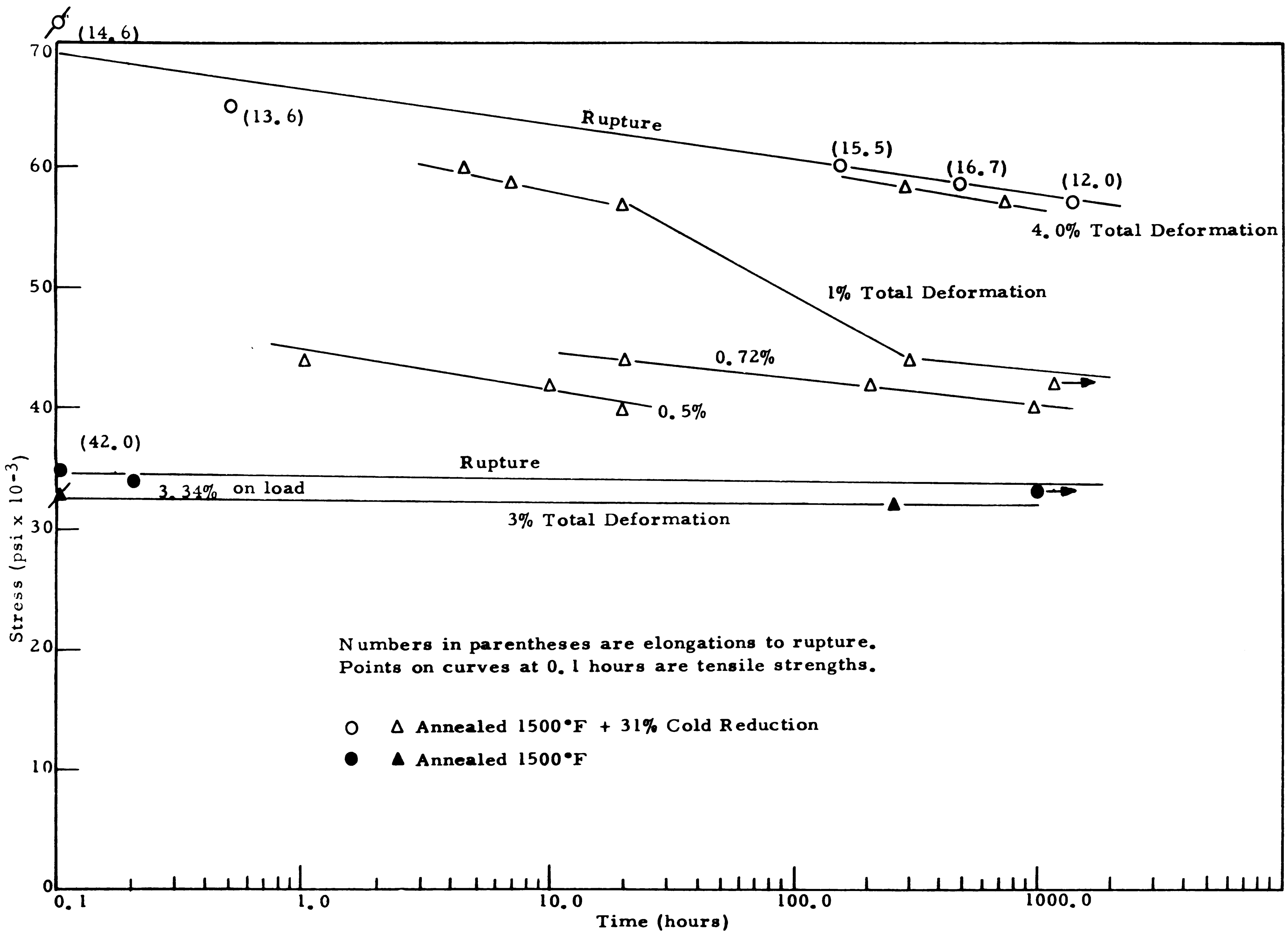
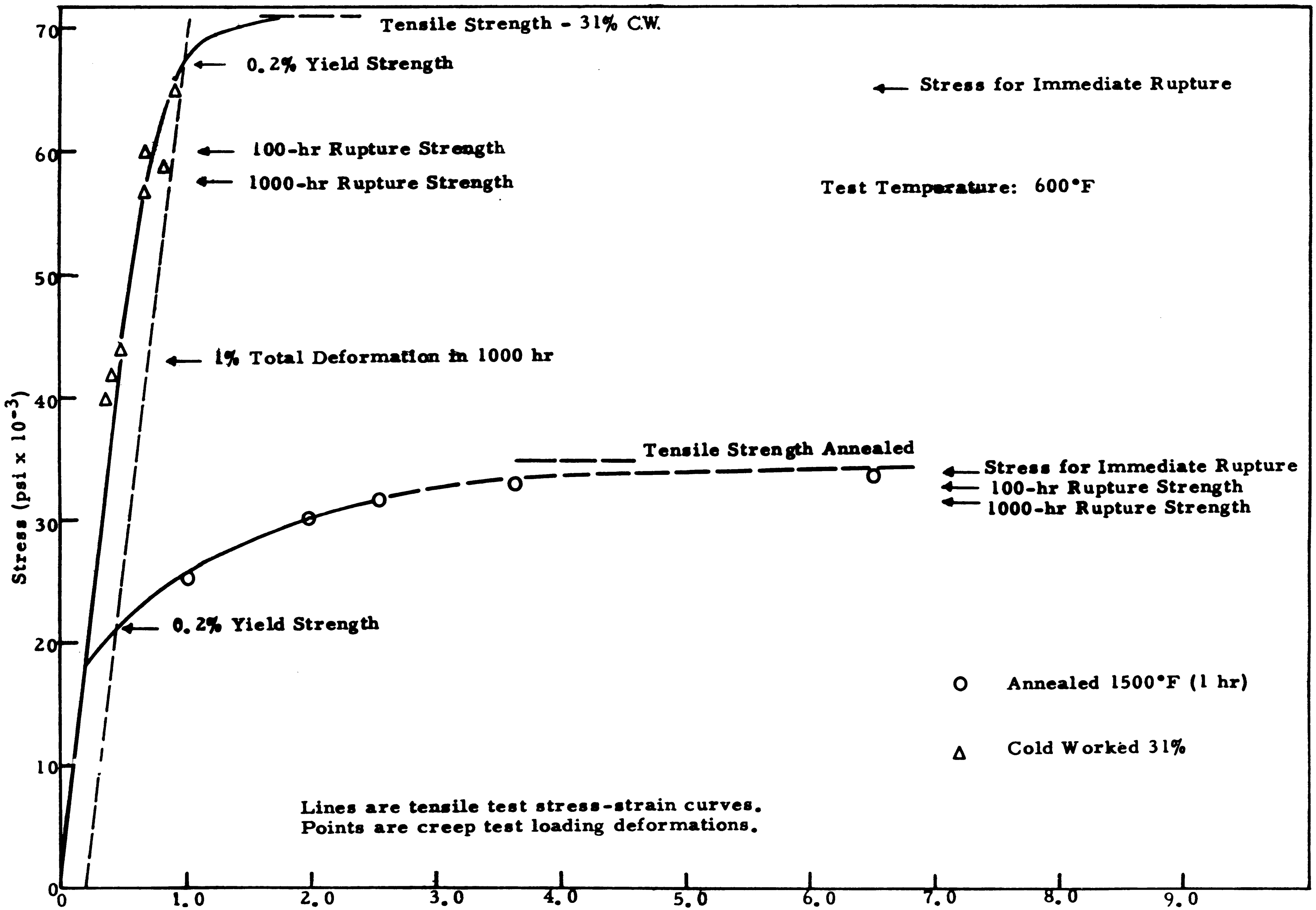
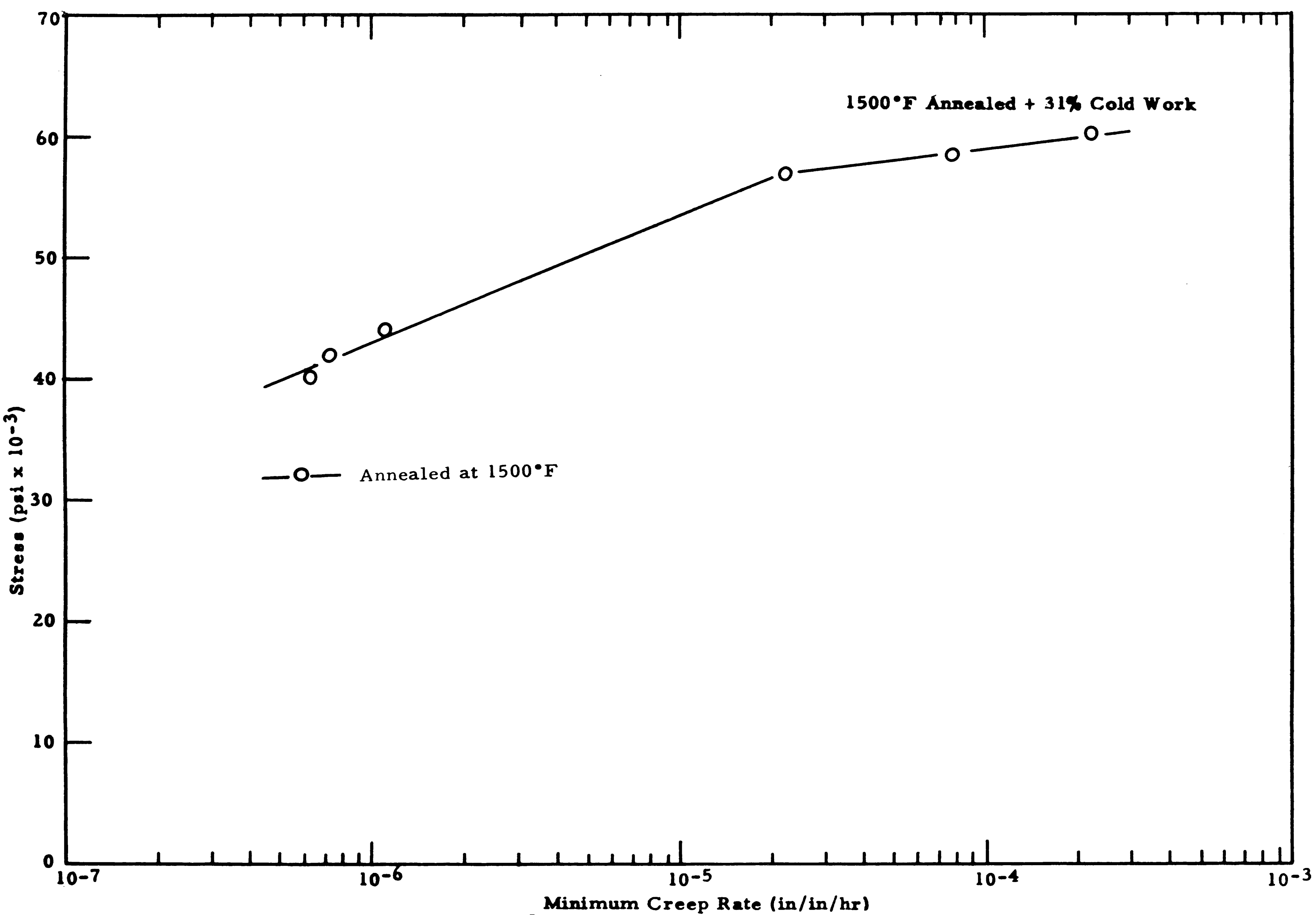


Figure 2. - Stress-Rupture Time and Stress-Time for Total Deformation Curves at 600°F for Alpha Titanium (Ti 75A)



Stress (psi x 10-3)
0
10
20
30
40
50
60
70
1.0
2.0
3.0
4.0
5.0
6.0
7.0
8.0
9.0
Tensile Strength - 31% C.W.
0.2% Yield Strength
100-hr Rupture Strength
1000-hr Rupture Strength
1% Total Deformation in 1000 hr
Stress for Immediate Rupture
Test Temperature: 600°F
Tensile Strength Annealed
Stress for Immediate Rupture
100-hr Rupture Strength
1000-hr Rupture Strength
0.2% Yield Strength
Annealed 1500°F (1 hr)
Cold Worked 31%
Lines are tensile test stress-strain curves.
Points are creep test loading deformations.

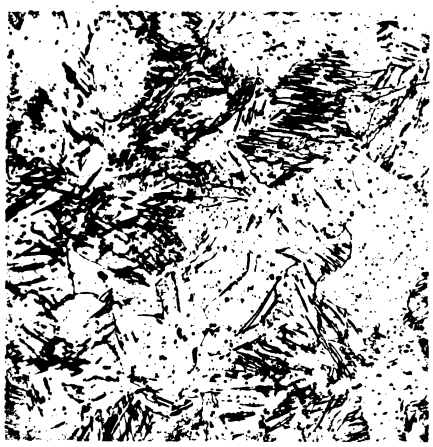


Stress (psi x 10-3)
70
60
50
40
30
20
10
0
1500°F Annealed + 31% Cold Work
Annealed at 1500°F
10-7
10-6
10-5
10-4
10-3
Minimum Creep Rate (in/in/hr)

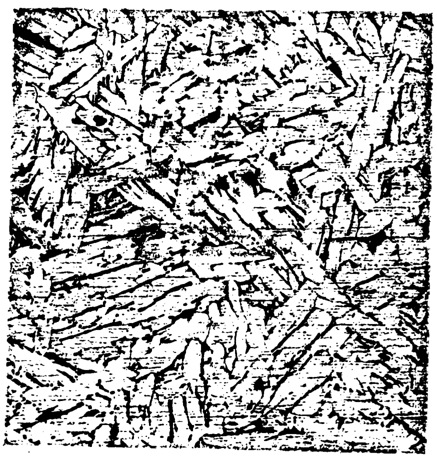
1700°F-2 hours -Furnace Cool

1700°F-2 hours -Iced Brine Quench

1800°F-1 hour - Furnace Cool



1750°F-1 hour -Iced Brine Quench



2000°F- 1 hour - Furnace Cool

1800°F-1 hour -Iced Brine Quench

Figure 5. - Influence of Heat Treatment at 1700° to 2000°F on the Microstructure of Alpha Titanium (Ti 75A).

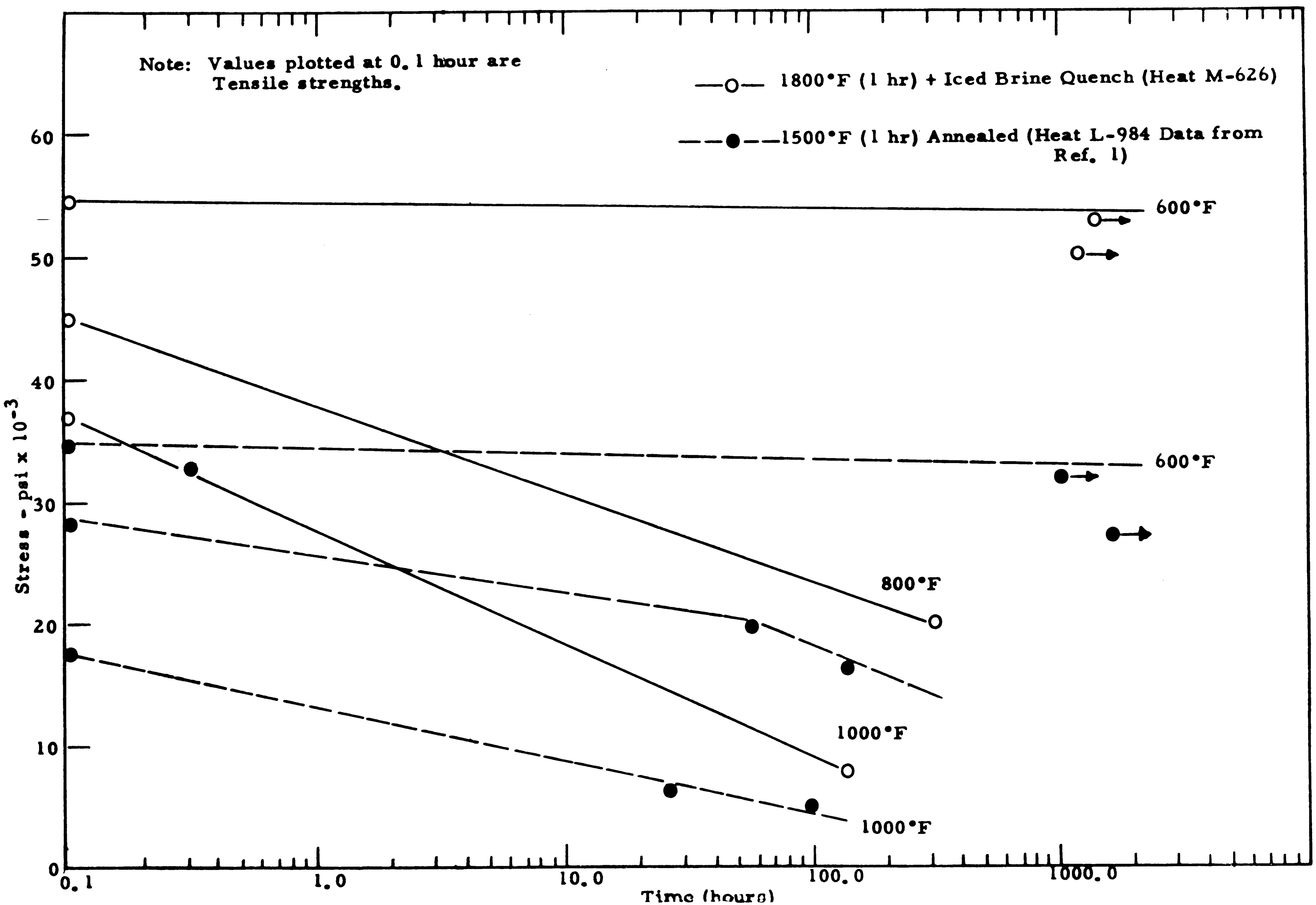
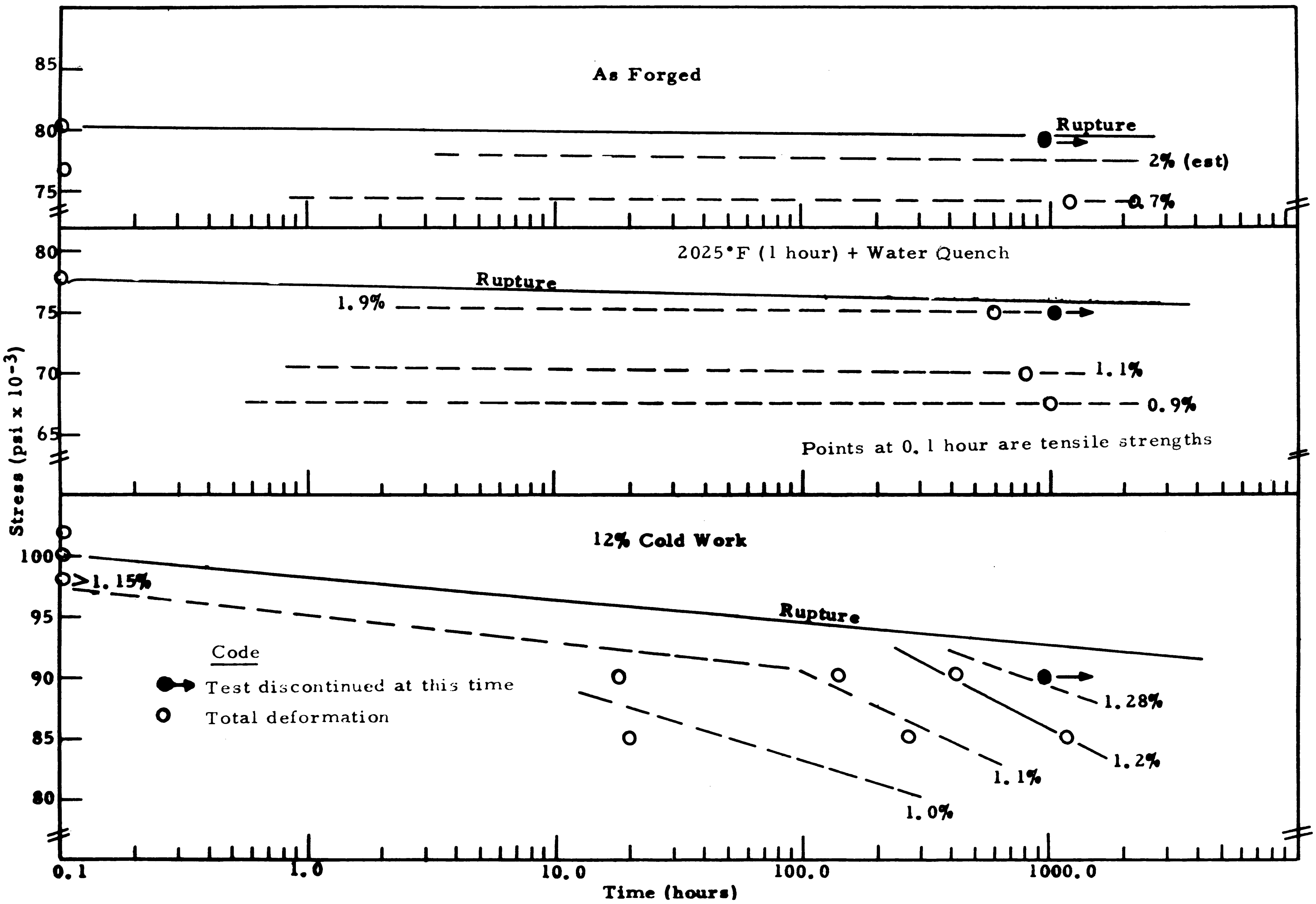


Figure 6. - Comparative Stress-Rupture Time Characteristics for Acicular and Annealed Alpha Titanium (Ti 75A)



As Forged
Rupture
2% (est)
0.7%
2025°F (1 hour) + Water Quench
Rupture
1.9%
1.1%
0.9%
Points at 0.1 hour are tensile strengths
12% Cold Work
≥1.15%
Rupture
Code
Test discontinued at this time
Total deformation
1.28%
1.2%
1.1%
1.0%
Stress (psi x 10-3)
Time (hours)
0.1
1.0
10.0
100.0
1000.0
85
80
75
70
65
100
95
90

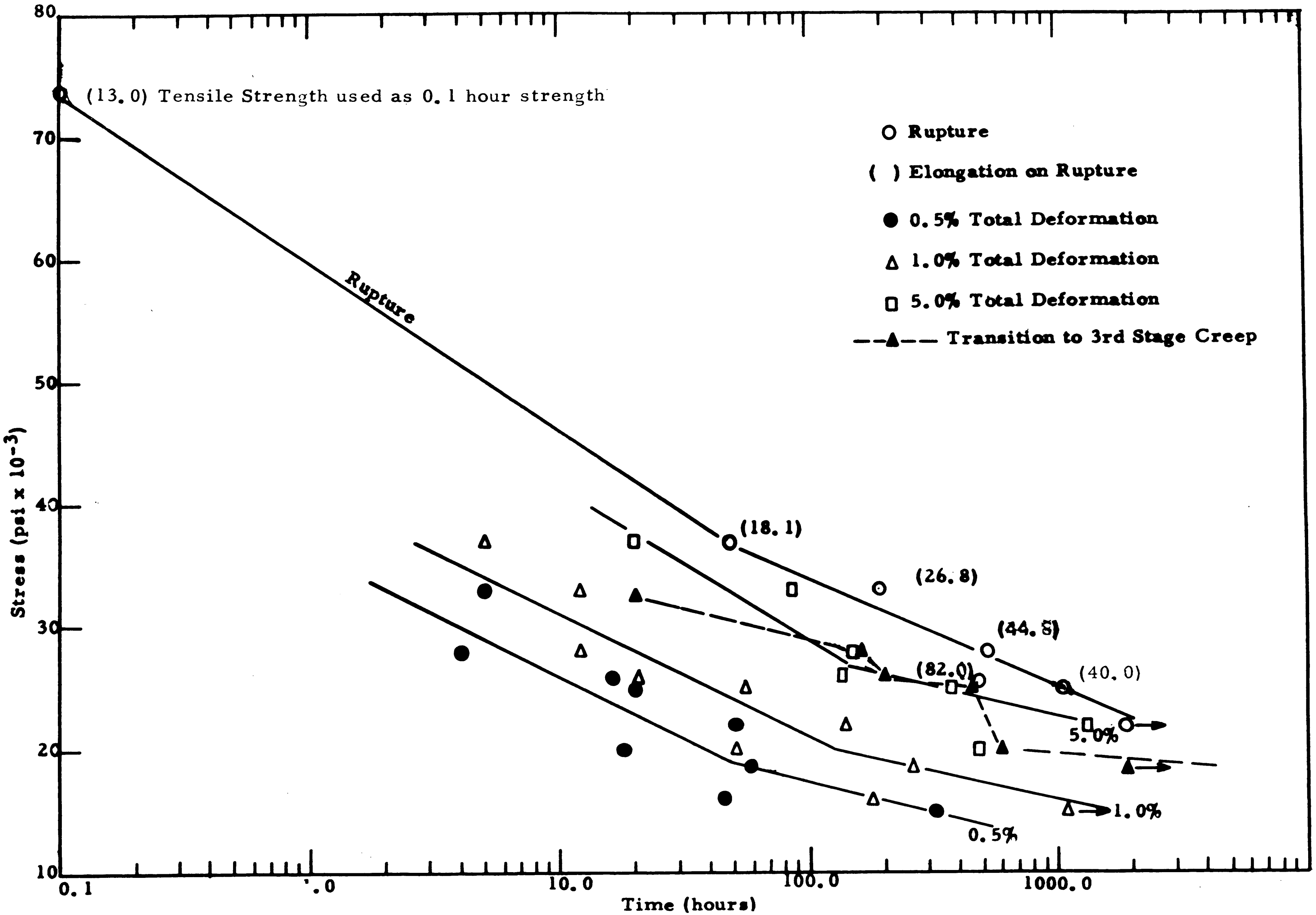


Figure 8. - Stress-Rupture Time and Stress-Time for Total Deformation Curves at 1000°F for Stabilized Alpha

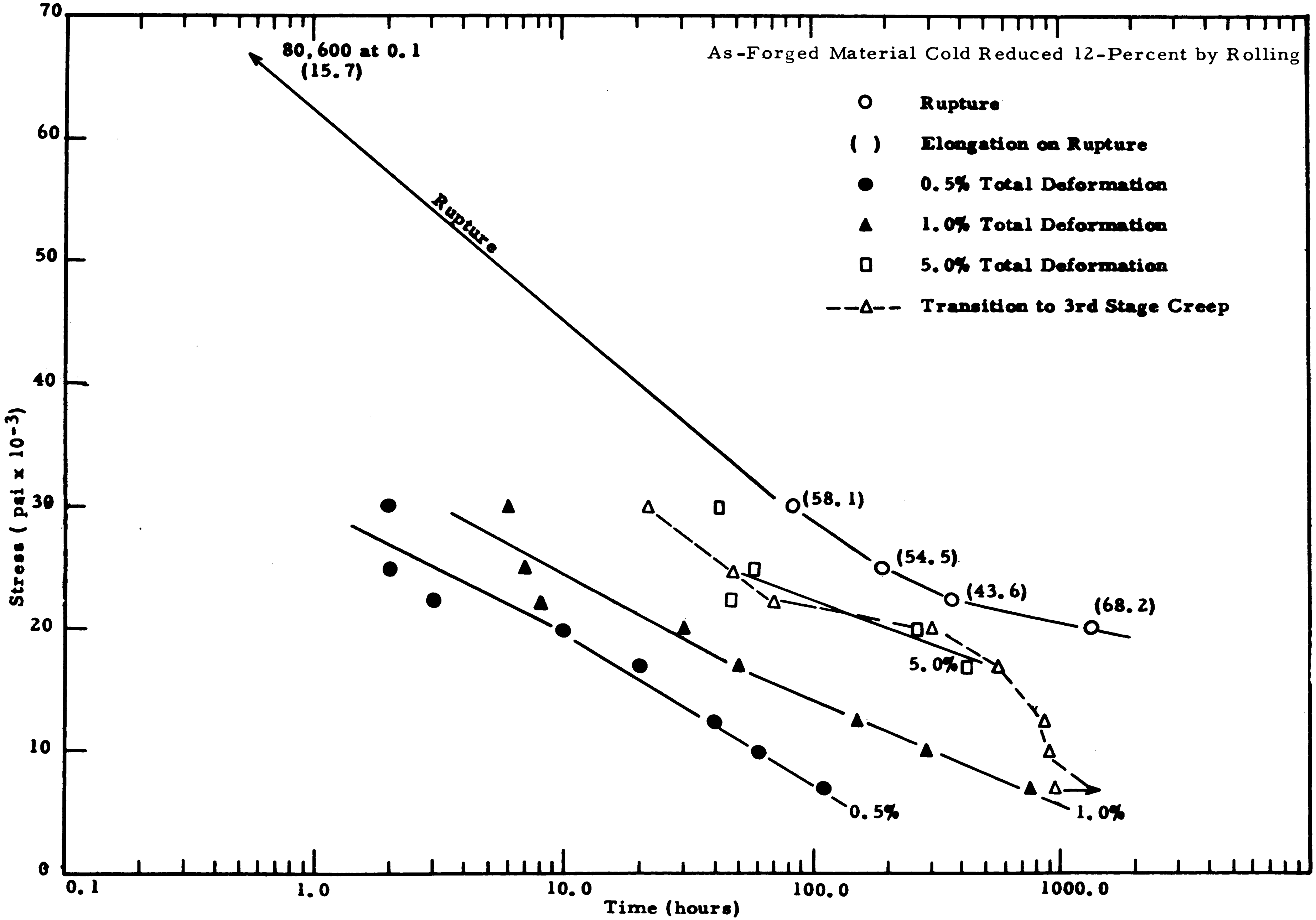
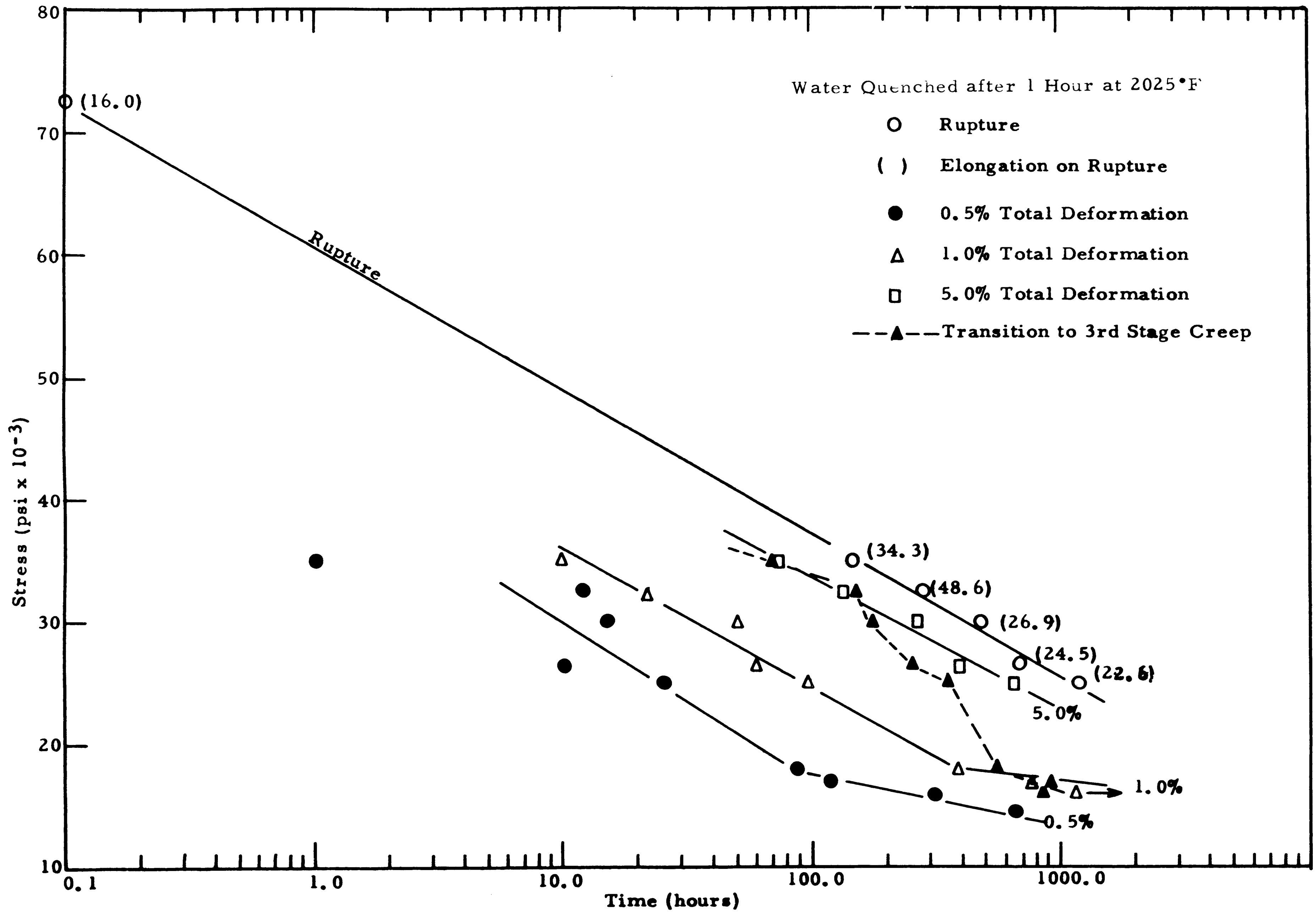
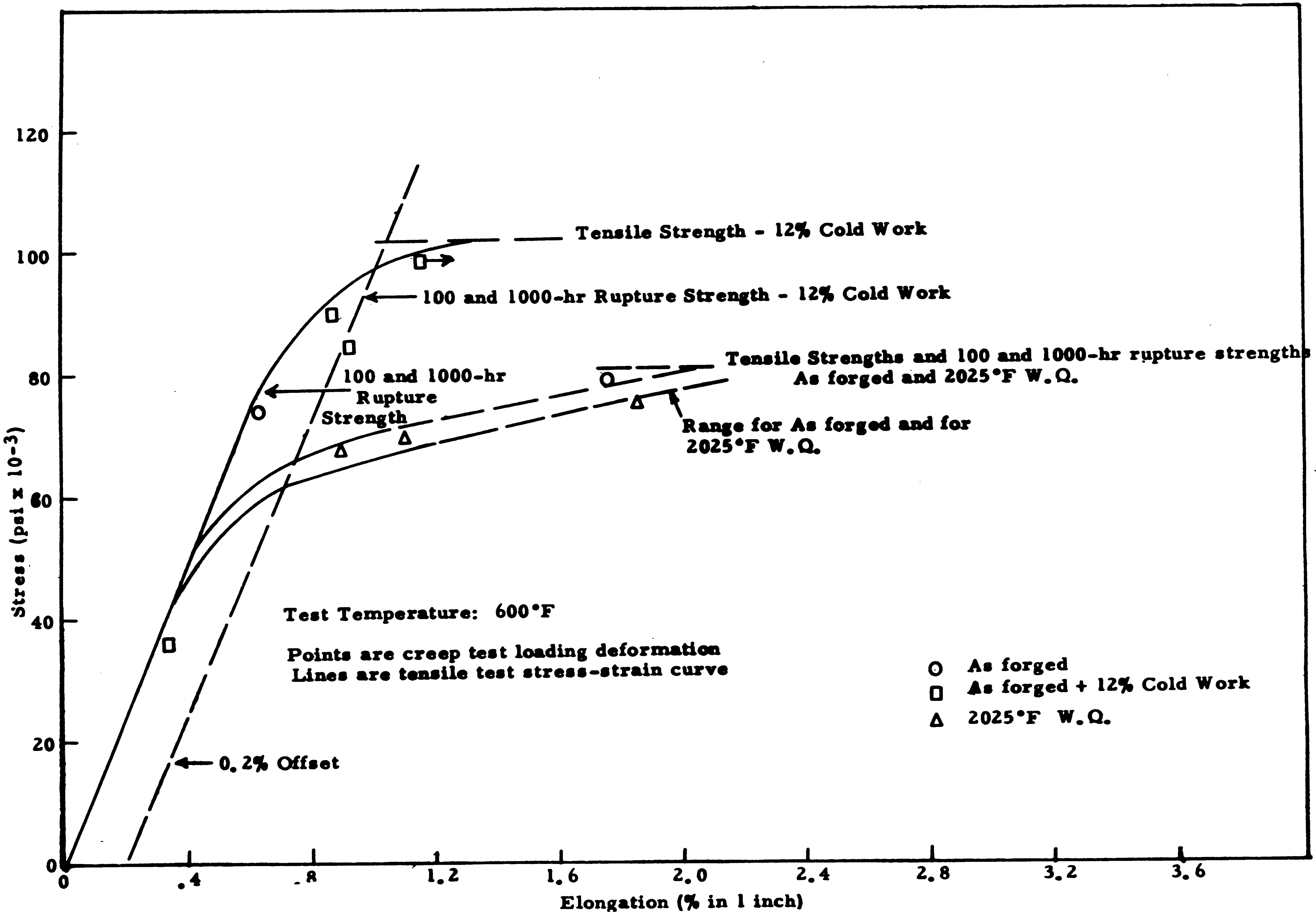


Figure 9. - Stress-Rupture Time and Stress-Times for Total Deformation Curves at 1000°F for Stabilized


Water Quenched after 1 Hour at 2025°F
Rupture
( ) Elongation on Rupture
0.5% Total Deformation
1.0% Total Deformation
5.0% Total Deformation
Transition to 3rd Stage Creep
Rupture
(16.0)
(34.3)
(48.6)
(26.9)
(24.5)
(22.6)
5.0%
1.0%
0.5%
Stress (psi x 10^{-3})
80
70
60
50
40
30
20
10
0.1
1.0
10.0
100.0
1000.0
Time (hours)



Stress (psi x 10-3)
0
20
40
60
80
100
120
0
.4
.8
1.2
1.6
2.0
2.4
2.8
3.2
3.6
Elongation (% in 1 inch)
Tensile Strength - 12% Cold Work
100 and 1000-hr Rupture Strength - 12% Cold Work
Tensile Strengths and 100 and 1000-hr rupture strengths
As forged and 2025°F W.Q.
100 and 1000-hr
Rupture
Strength
Range for As forged and for
2025°F W.Q.
Test Temperature: 600°F
Points are creep test loading deformation
Lines are tensile test stress-strain curve
0.2% Offset
As forged
As forged + 12% Cold Work
2025°F W.Q.

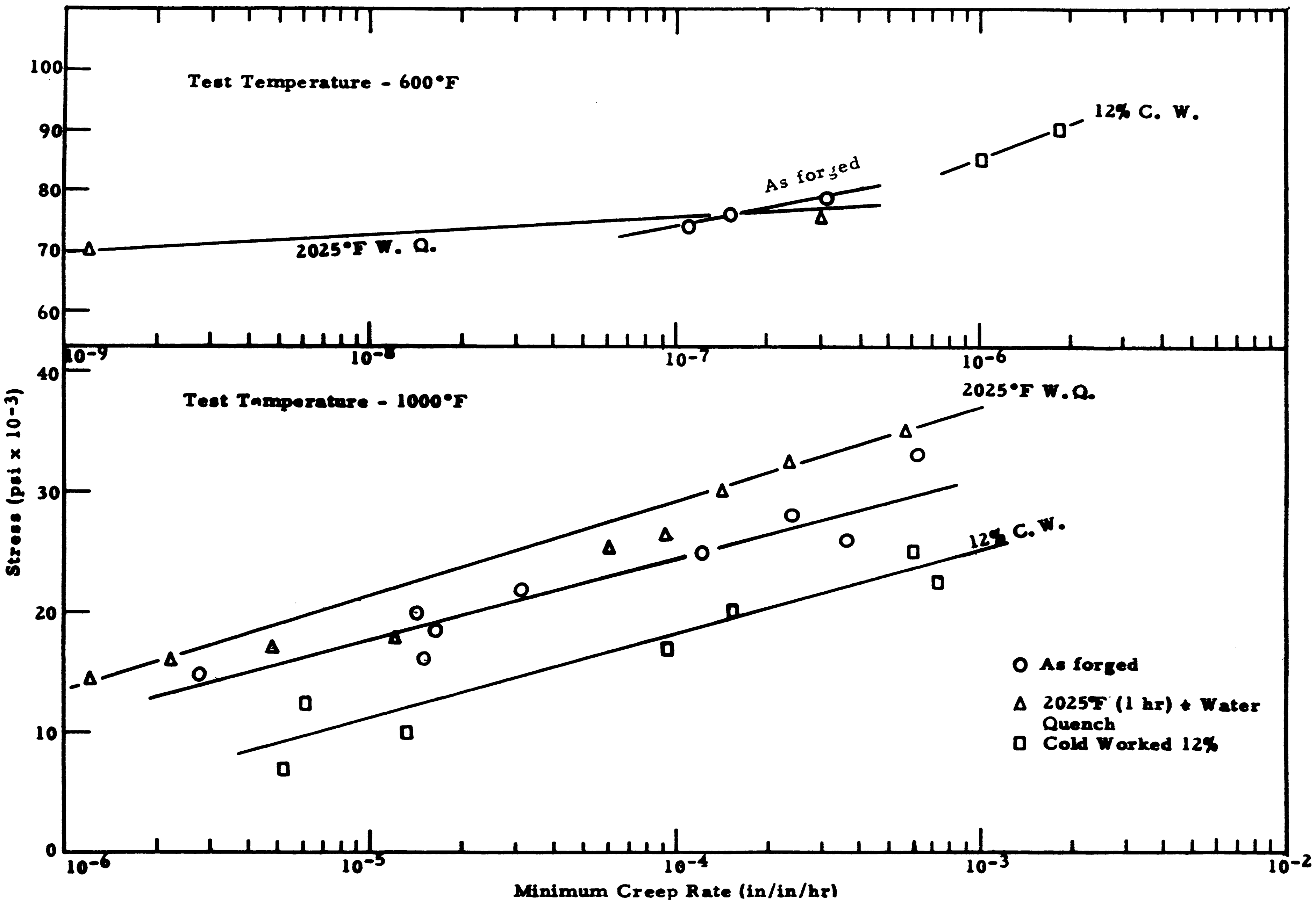


Figure 12. - Stress-Minimum Creep Rate Curves at 600° and 1000°F for Stabilized Alpha 6 Al Alloy.

X250

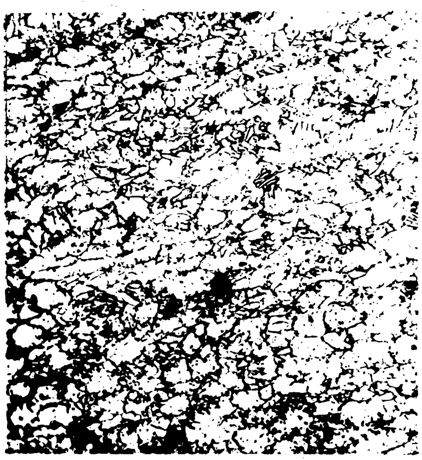
75°F Tensile Test

X250

Tested at 600°F for 941 hours under 90,000 psi

X250

Tested at 1000°F for 82.3 hours under 30,000 psi



X250

Tested at 1000°F for 187.3 hours under 25,000 psi

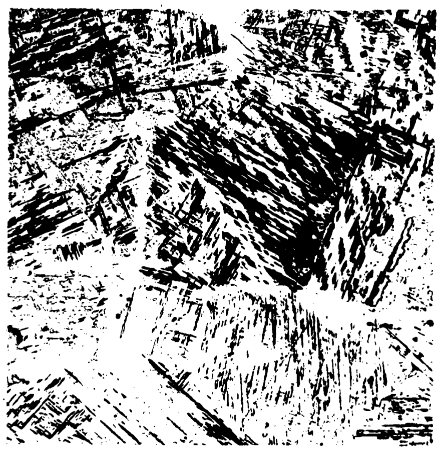
X250

Tested at 1000°F for 1102 hours under 10,000 psi

Figure 13. - Influence of Testing Conditions on the Microstructure of Stabilized Alpha 6 Al Alloy in the Cold-Worked Condition (Original Treatment was 12-Percent Cold Reduction by Rolling on As Forged Material).

X75
As Forged

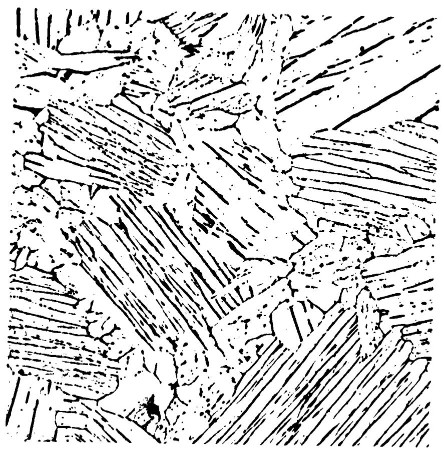
X100
As Forged + 37-Percent Cold Reductio



X100
1 hour at 2100°F + Iced Brine Quench

X100
As Forged + 24 hours at 1100°F

X100
1 hour at 1925°F, Water Quenched +
24 hours at 1100°F



X100
1 hour at 1925°F + Furnace Cool to
1100°F and held 24 hours

Etchant: 2HF + $2HNO_3$ + 100 H_2O

Figure 14. - Influence of Heat Treatment on the Microstructure of Stabilized Alpha 6 Al - 0.5 Si Alloy.

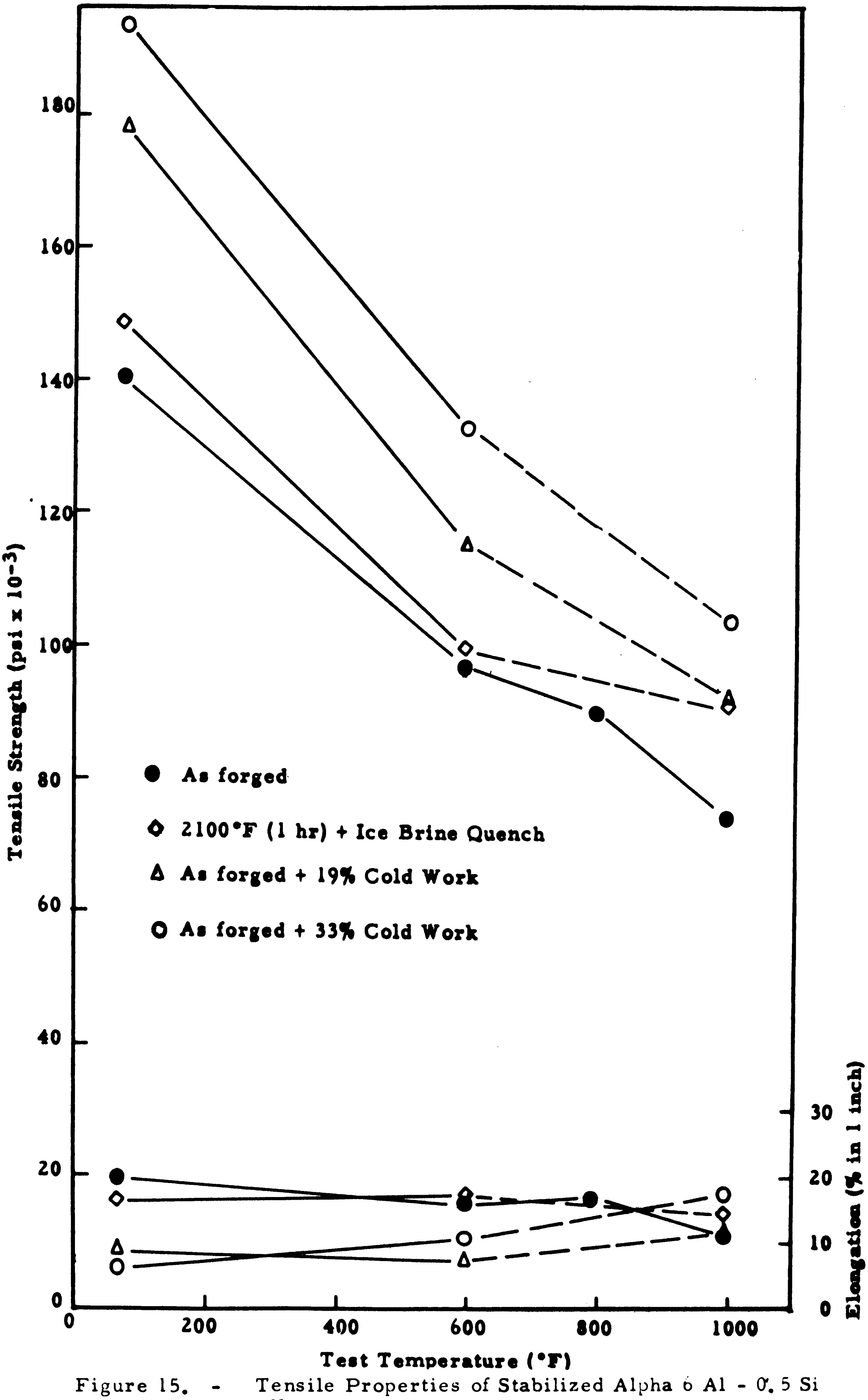


Figure 15. - Tensile Properties of Stabilized Alpha 6 Al - 0.5 Si Alloy.

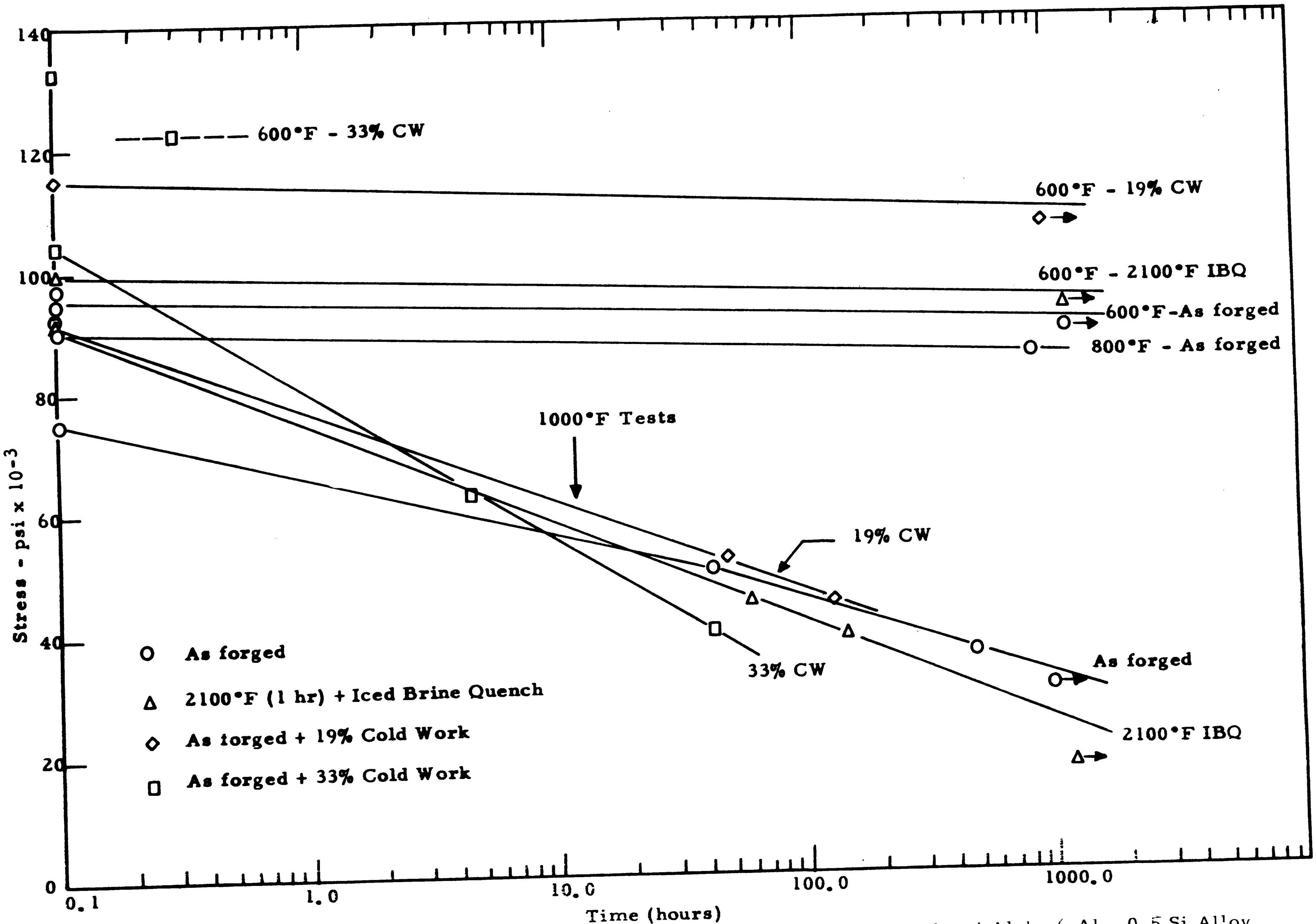


Figure 16. - Stress Rupture Time Curves at 600°, 800° and 1000°F for Stabilized Alpha 6 Al - 0.5 Si Alloy.

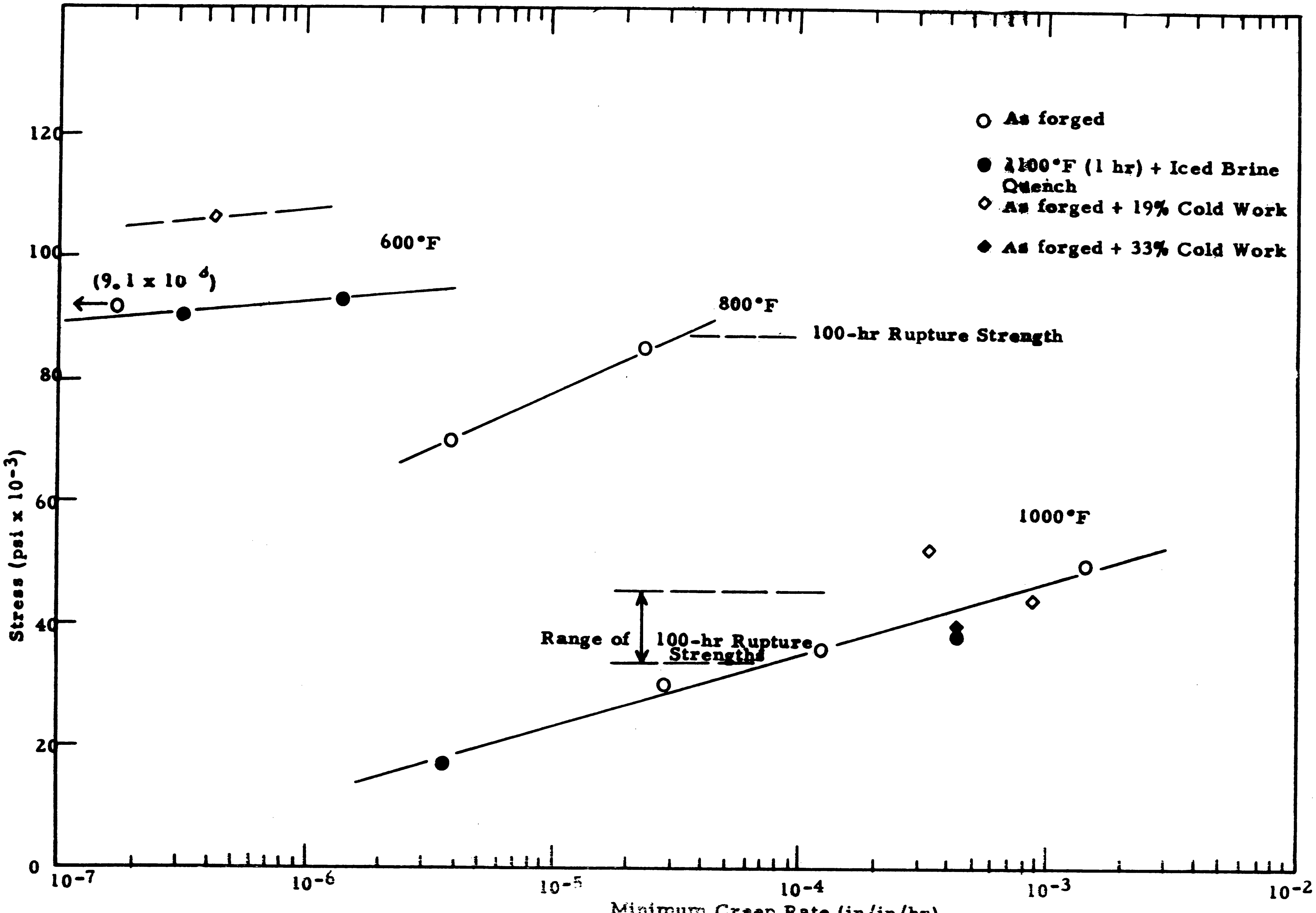
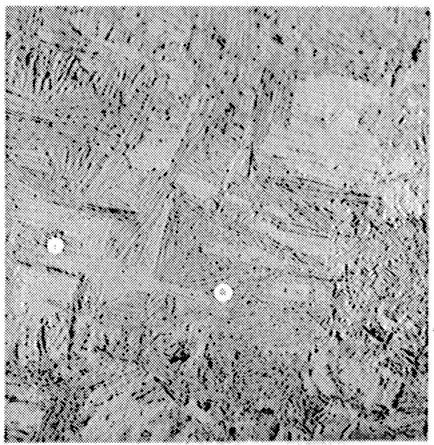
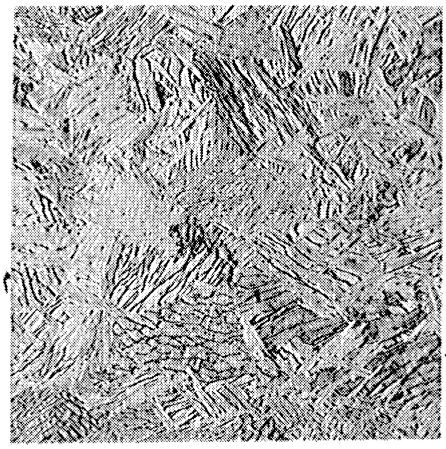


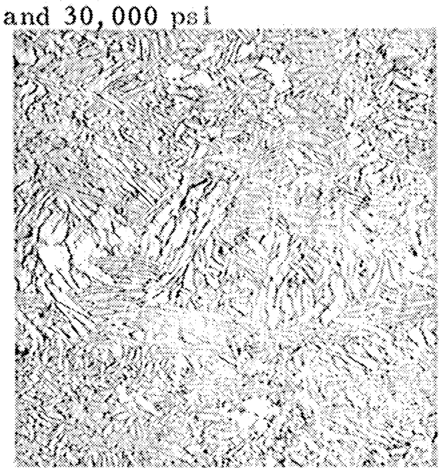
Figure 17. - Stress Minimum Creep Rate Curves at 600°, 800° and 1000°F for Stabilized Alpha + Al - 0.5 Si Alloy.



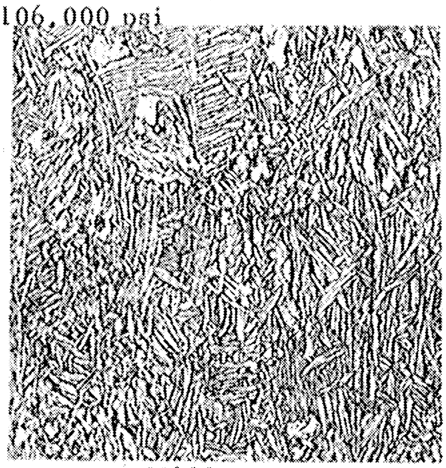
X100
As Forged
Tested for 986.1 hours at 1000°F and 30,000 psi



X100
As Forged + 37-Percent Cold Reducti
Tested for 961.7 hours at 600°F and 106,000 psi

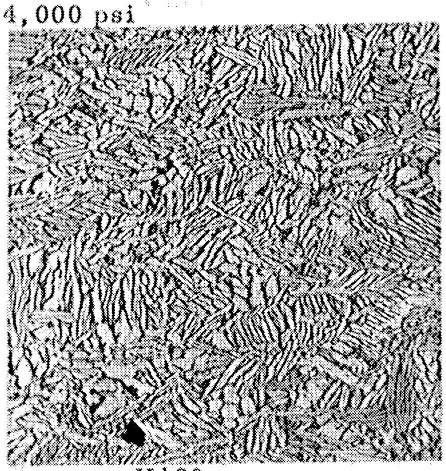


X100
As Forged + 19-Percent Cold Reduction
Tested for 48.4 hours at 1000°F and 52,000 psi



X100
As Forged + 19-Percent Cold Reducti
Tested for 126 hours at 1000°F and 44,000 psi

X100
As Forged + 33-Percent Cold Reduction
Tested for 4.7 hours at 1000°F and 63,000 psi



X100
As Forged + 33-Percent Cold Reducti
Tested for 41.9 hours at 1000°F and 40,000 psi

Etchant: 2HF, $2HNO_3$, 100 H_2O

Figure 18. - Microstructures of Stabilized Alpha 6 Al - 0.5 Si Alloy after Creep-Rupture Tests.

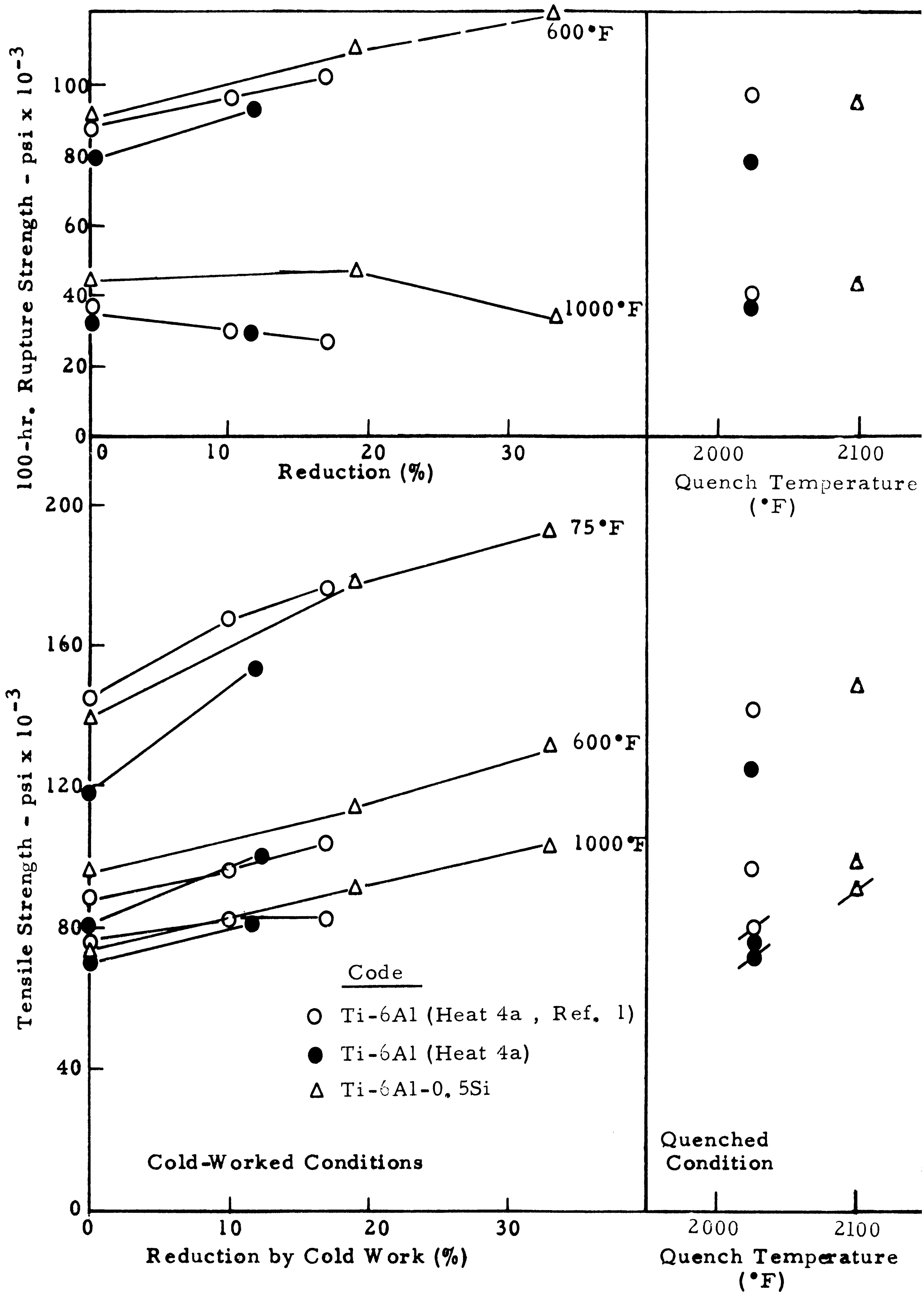
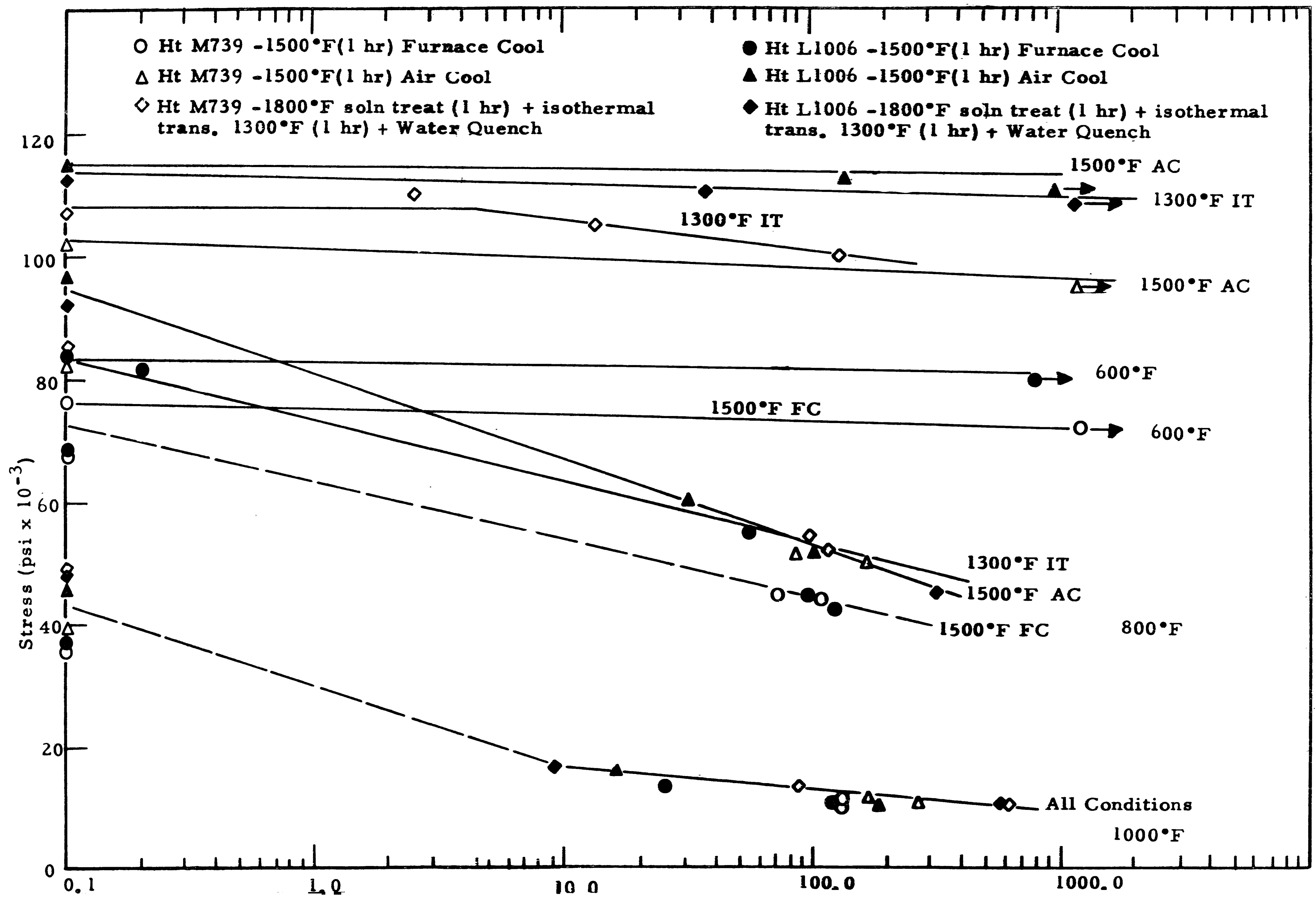
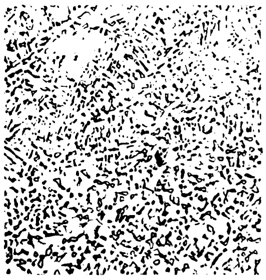


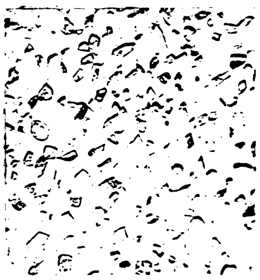
Figure 19. - Comparative Properties for Stabilized Alpha Alloys in Cold-Worked or Quenched Condition.



O Ht M739 -1500°F(1 hr) Furnace Cool
Δ Ht M739 -1500°F(1 hr) Air Cool
◇ Ht M739 -1800°F soln treat (1 hr) + isothermal trans. 1300°F (1 hr) + Water Quench
● Ht L1006 -1500°F(1 hr) Furnace Cool
▲ Ht L1006 -1500°F(1 hr) Air Cool
◆ Ht L1006 -1800°F soln treat (1 hr) + isothermal trans. 1300°F (1 hr) + Water Quench
Stress (psi x 10^{-3})
0
20
40
60
80
100
120
0.1
1.0
10.0
100.0
1000.0
1500°F AC
1300°F IT
1300°F IT
1500°F AC
600°F
1500°F FC
600°F
1300°F IT
1500°F AC
1500°F FC
800°F
All Conditions
1000°F



X500
a. As Produced - R_c 33 hardness



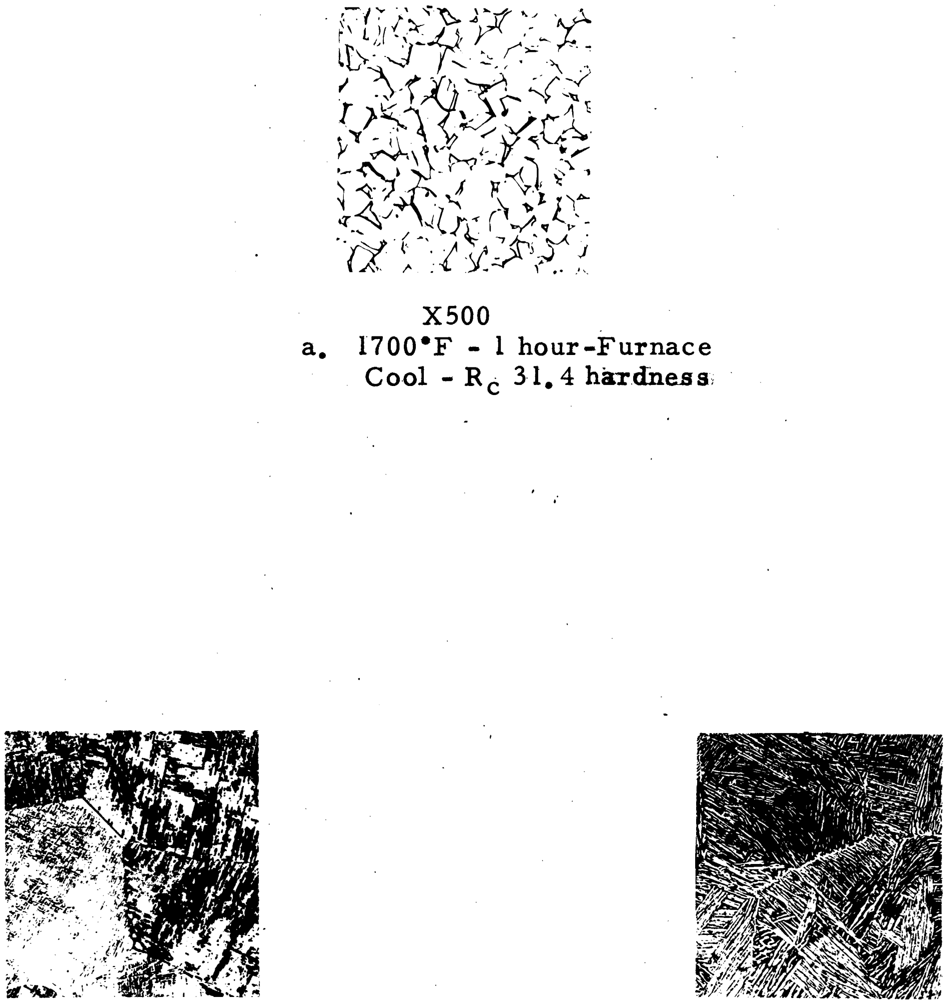
X500
b. 1700°F - 1 hour - Water Quench
R_c 38 hardness

X250
c. 1775°F - 1 hour - Water Quench
R_c 48 hardness

X100
d. 1800°F - 1 hour - Water Quench
R_c 48 hardness

X100
e. 1900°F - 1 hour - Water Quench
R_c 48 hardness

Figure 21. - Influence of Quenching Temperature on the Microstructure and Hardness of Alpha-Beta Alloy Ti 155AX.



X500
a. 1700°F - 1 hour-Furnace Cool - R_c 31.4 hardness

X100
b. 1800°F - 1 hour - Air Cool R_c 39 hardness

X100
c. 1800°F - 1 hour - Furnace Cool - R_c 30.7 hardness

Figure 22. - Microstructures and Hardness for Alpha-Beta Alloy Ti 155AX when Air and Furnace Cooled from 1700° and 1800°F.

X500

a. 1700°F - 1 hour - Water Quench + 445°F - 1 hour - Water Quench - R_C 47 hardness

X100

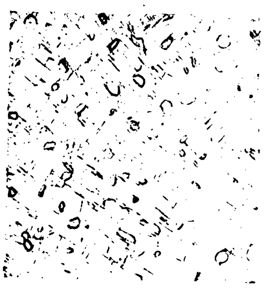
b. 1800°F - 1 hour - Water Quench + 800°F - 15 minutes - Water Quench - R_C 49 hardness

X500

c. 1700°F - 1 hour - Water Quench + 800°F - 1 hour - Water Quench - R_C 48.5 hardness

X100

d. 1800°F - 1 hour - Water Quench + 1200°F - 15 minutes - Water Quench - R_C 43.7 hardness



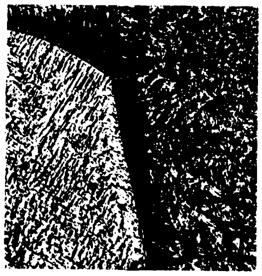
X500

e. 1700°F - 1 hour - Water Quench + 1000°F - 1 hour - Water Quench - R_C 44.7 hardness

X100

f. 1800°F - 1 hour - Water Quench 1600°F - 15 minutes - Water Quench - R_C 39 hardness

Figure 23. - Microstructures and Hardness of Alpha-Beta Alloy Ti 155AX when Reheated after Quenching from 1700° and 1800°F.



X100

a. 1800°F - 1 hour + isothermal transformation at 800°F for 15 minutes - Water Quenched R_c 47.2 hardness

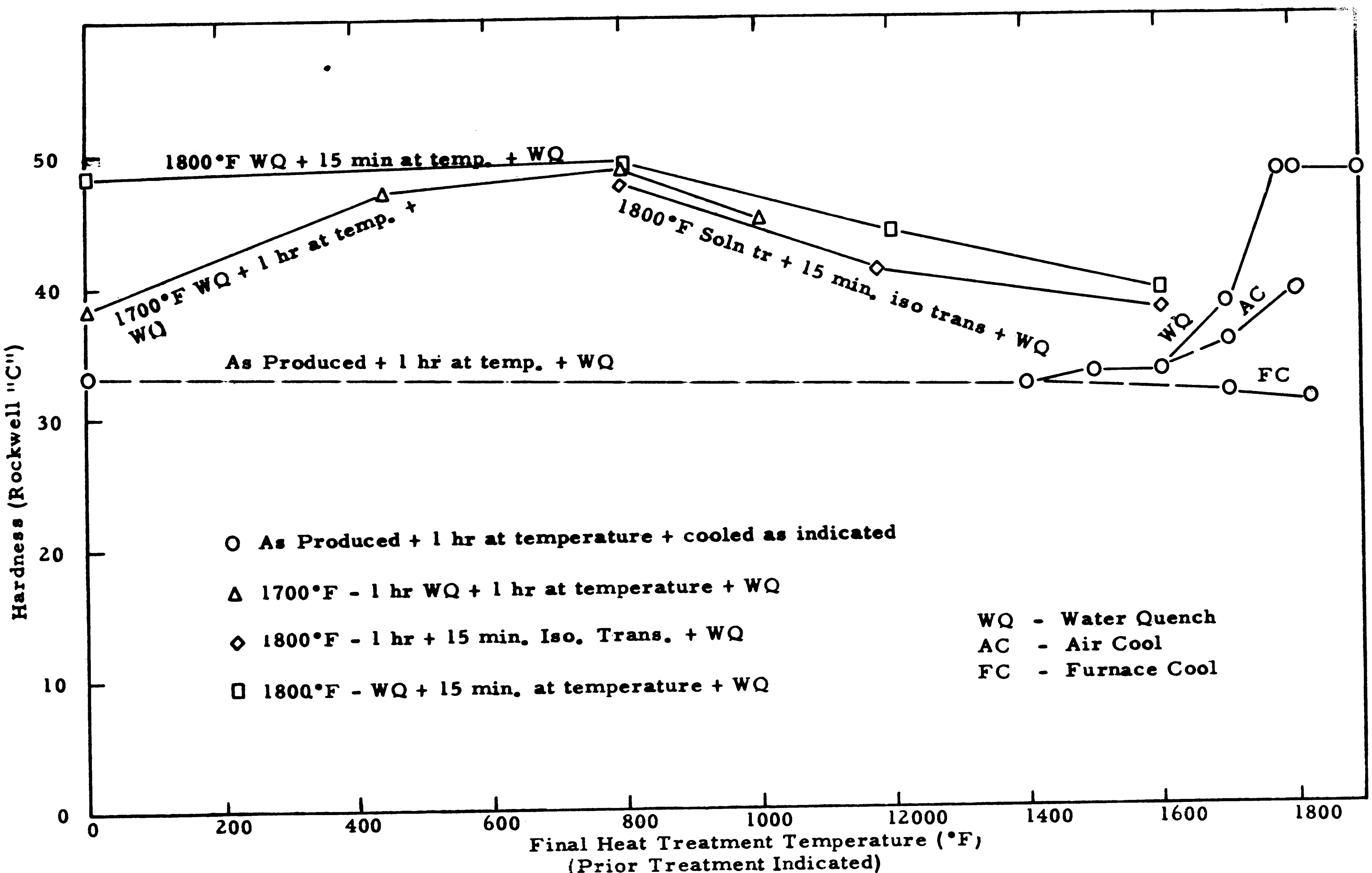
X100

b. 1800°F - 1 hour + isothermal transformation at 1180°F for 15 minutes + Water Quench - R_c 40.7 hardness

X100

c. 1800°F - 1 hour + isothermal transformation at 1600°F for 15 minutes + Water Quench R_c 37.7 hardness

Figure 24. - Microstructures and Hardness of Alpha-Beta Alloy Ti 155AX after Isothermal Transformation from 1800°F.



rature on Hardness of Alpha Beta Alloy Ti 155AX.

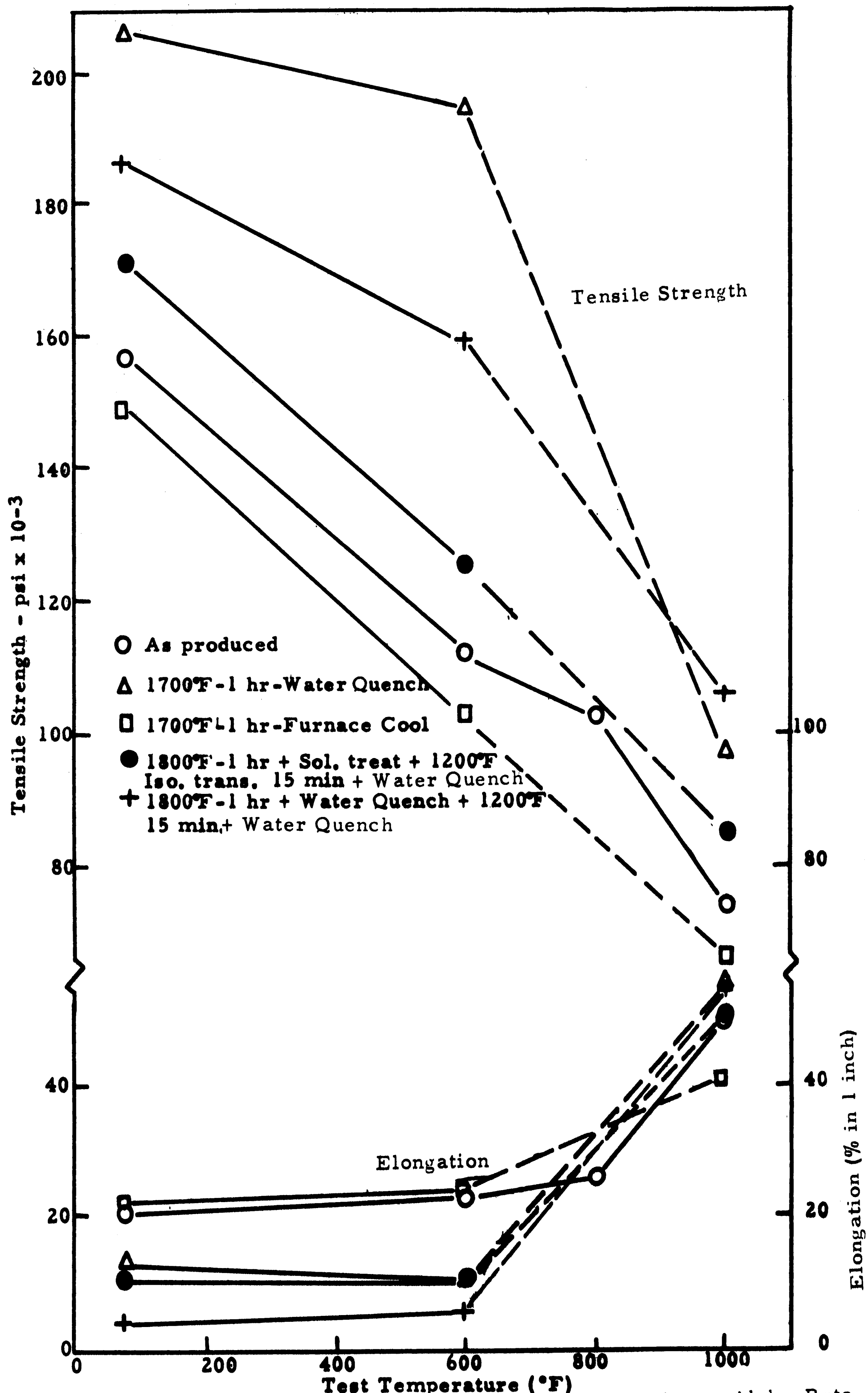
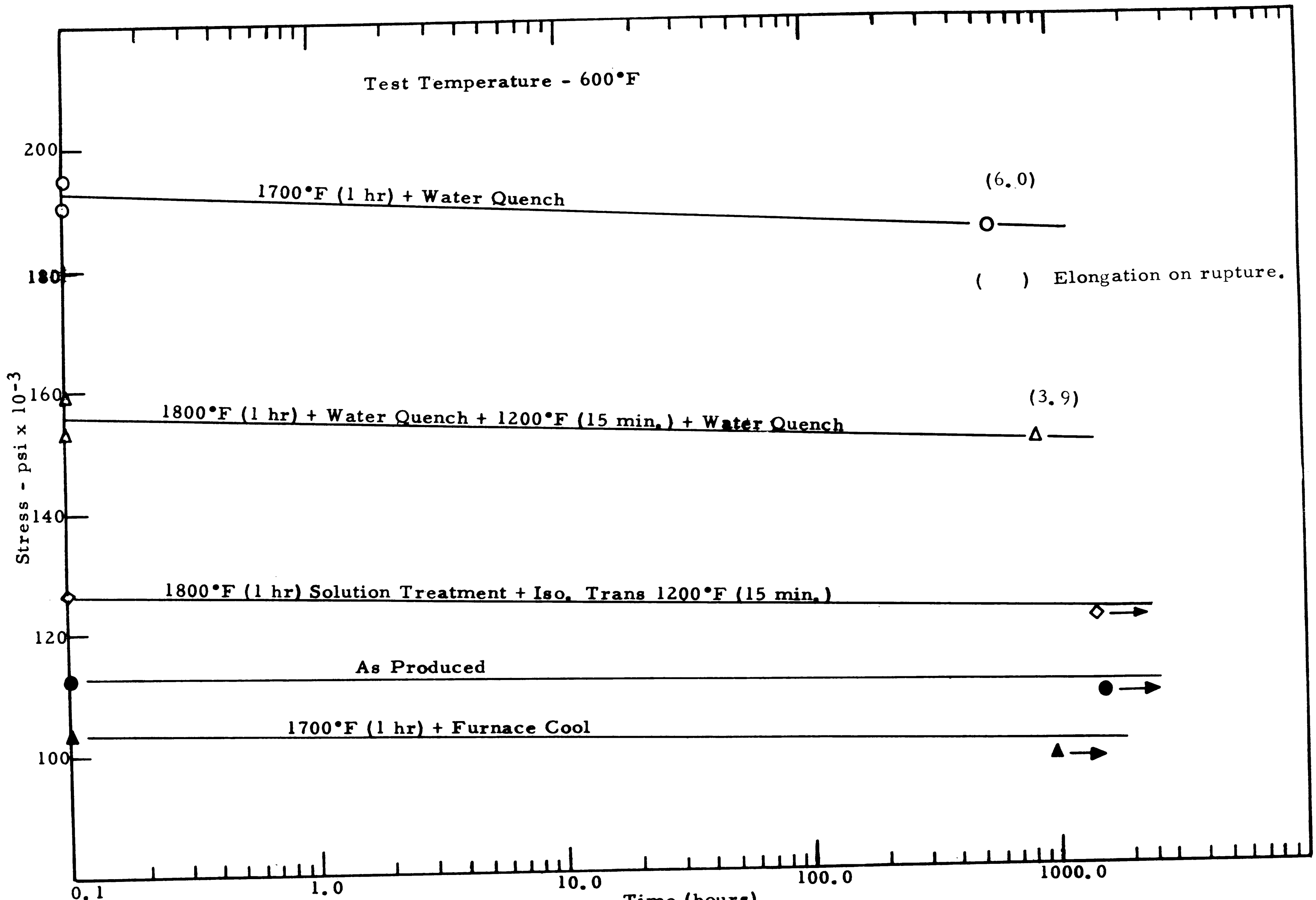


Figure 26. - Effect of Test Temperature on Tensile Properties of Alpha-Beta Alloy Ti 155AX



Test Temperature - 600°F
200
180
160
140
120
100
Stress - psi x 10^{-3}
1700°F (1 hr) + Water Quench
(6.0)
( ) Elongation on rupture.
1800°F (1 hr) + Water Quench + 1200°F (15 min.) + Water Quench
(3.9)
1800°F (1 hr) Solution Treatment + Iso. Trans 1200°F (15 min.)
As Produced
1700°F (1 hr) + Furnace Cool
0.1
1.0
10.0
100.0
1000.0

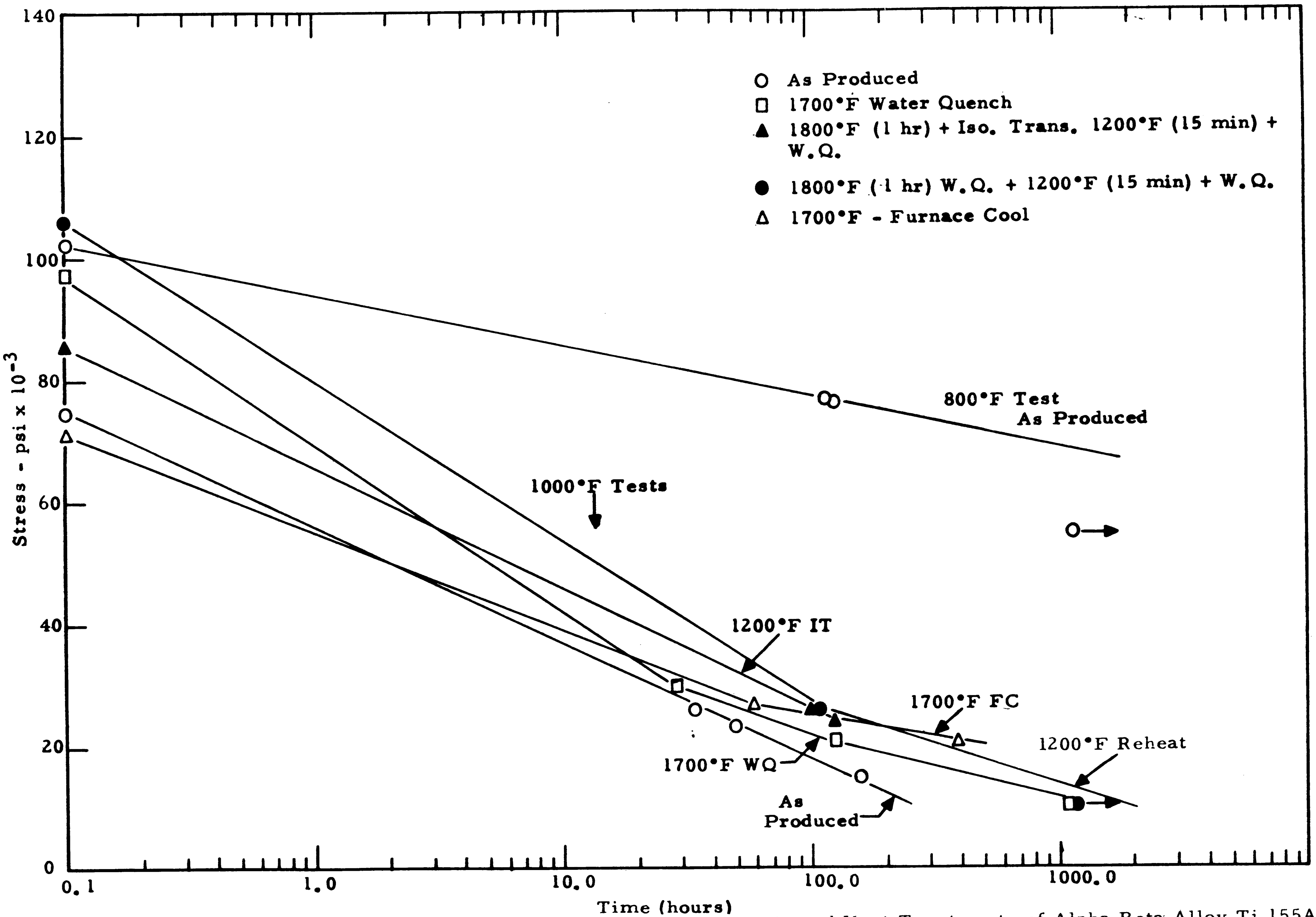


Figure 28. - Stress Rupture Time Curves at 800° and 1000°F for Indicated Heat Treatments of Alpha Beta Alloy Ti 155AX.

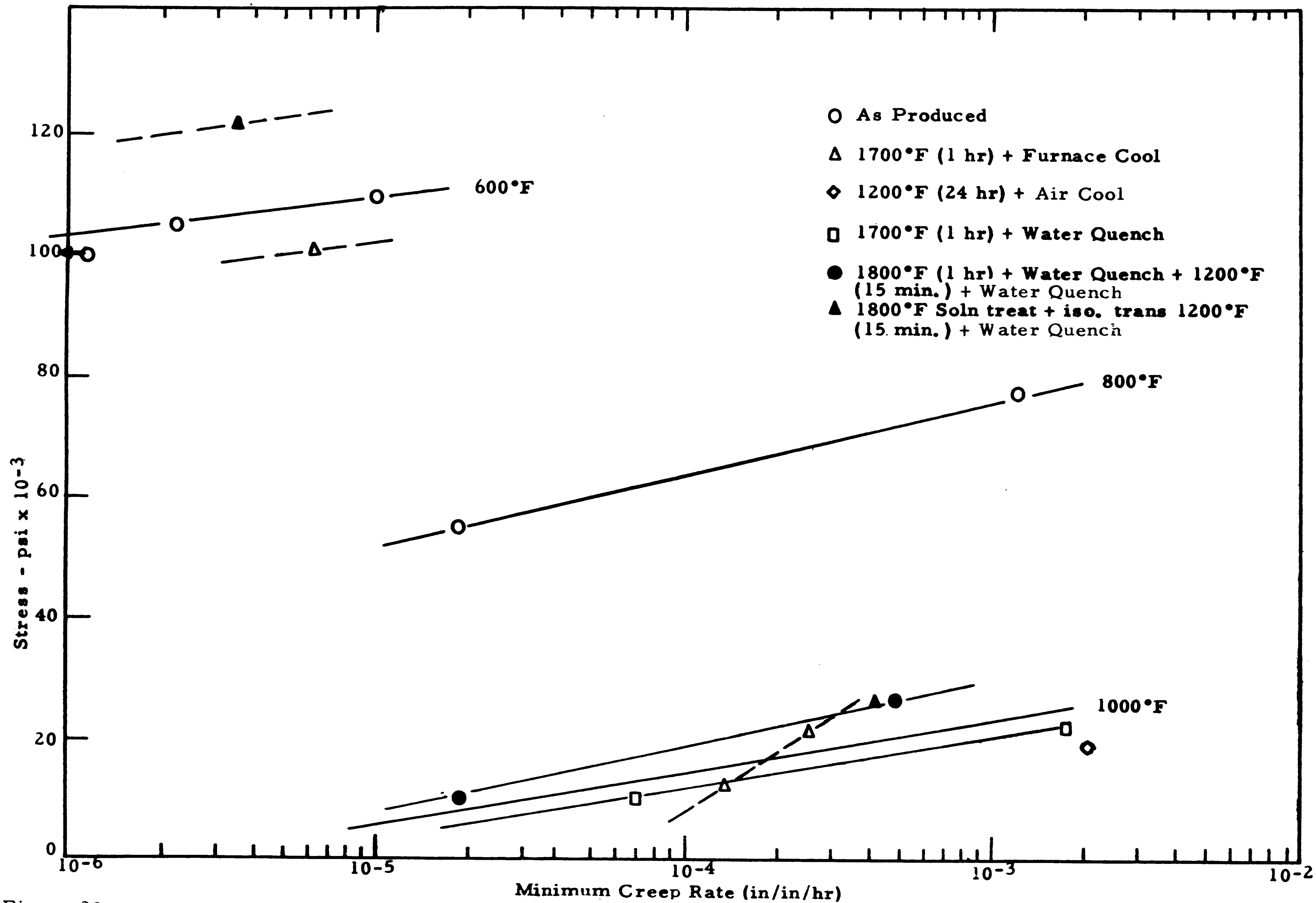
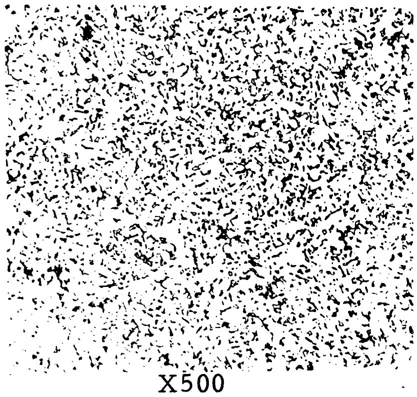
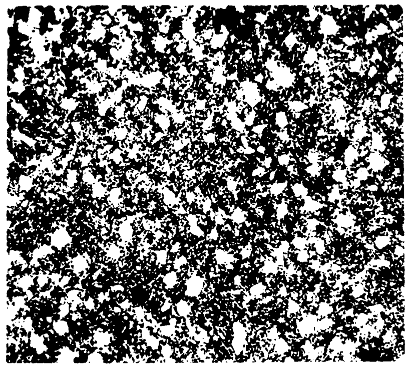


Figure 29. - Stress-Minimum Creep Rate Curves at 600°, 800° and 1000°F for Indicated Heat Treatments of Alpha-Beta Alloy Ti 155AX.



X500

a. As Produced - Tested at 1000°F for 157.4 hours under 15,000 psi



X500

b. 1700°F - 1 hour - Water Quenched - Tested at 1000°F for 28.3 hours and 30,000 psi

X500

c. 1700°F - 1 hour - **Furnace Cool** Tested at 600°F **for 982.7 hours** under 100,000 psi

X500

d. 1700°F - 1 hour - Furnace Cool Tested at 1000°F for 390.5 hours under 21,000 psi

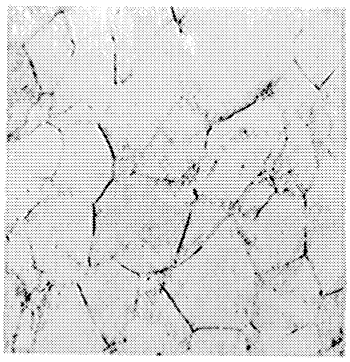
X100

e. 1800°F - 1 hour - Water Quench + 15 minutes at 1200°F, Water Quench - Tested at 1000°F for 109 hours under 26,000 psi

X100

f. 1800°F - 1 hour + isothermal transformation at 1200°F for 15 minutes - Tested at 1000°F for 100 hours under 26,000 psi

Figure 30. - Microstructures of Alpha-Beta Alloy Ti 155AX after Creep-Rupture Testing.

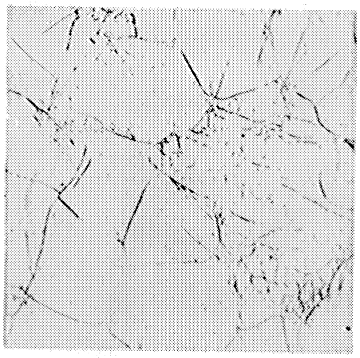


X250

a. As Forged
R_c 32.1 hardness

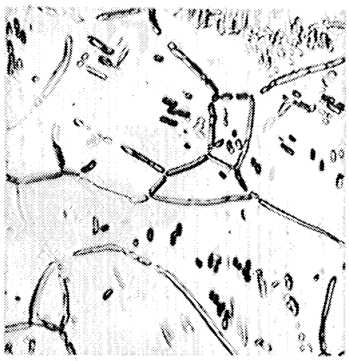
X250

b. 1800°F - 30 minutes - Water Quench - R_c 23.7 hardness



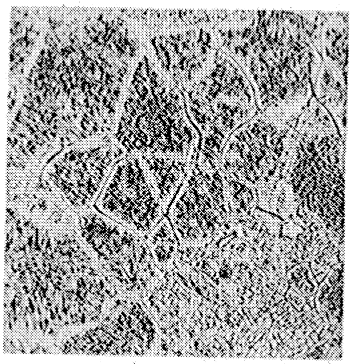
X250

c. 1450°F - 30 minutes - Water Quench - R_c 27 hardness



X500

d. 1425°F - 25 hours - Water Quench - R_c 28 hardness



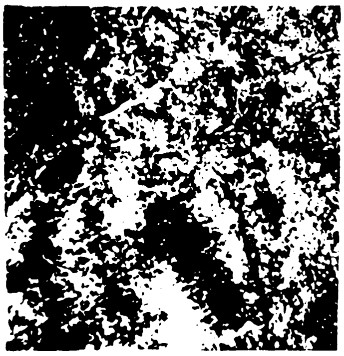
X250

e. 1350°F - 24 hour - Water Quench - R_c 28 hardness

Figure 31. - Influence of Heat Treatment on the Microstructure of Meta Stable Beta Alloy 10 Mo.

X500
f. 1000°F - 5 hours - Air Cool - R_c 38.7 hardness

X500
g. 1000°F - 24 hours - Air Cool - R_c 36 hardness



X500
h. 800°F - 1 hour - Air Cool - R_c 49 hardness

X500
i. 1800°F - 1 hour + isothermal transformation at 1200°F for 1 hour - Water Quench - R_c 26.2 hardness

X500
j. 1800°F - 1 hour + isothermal transformation at 950°F for 15 minutes - Water Quench R_c 38.8 hardness

Figure 31. (concluded) - Influence of Heat Treatment on the Microstructure of Meta Stable Beta Alloy 10 Mo.

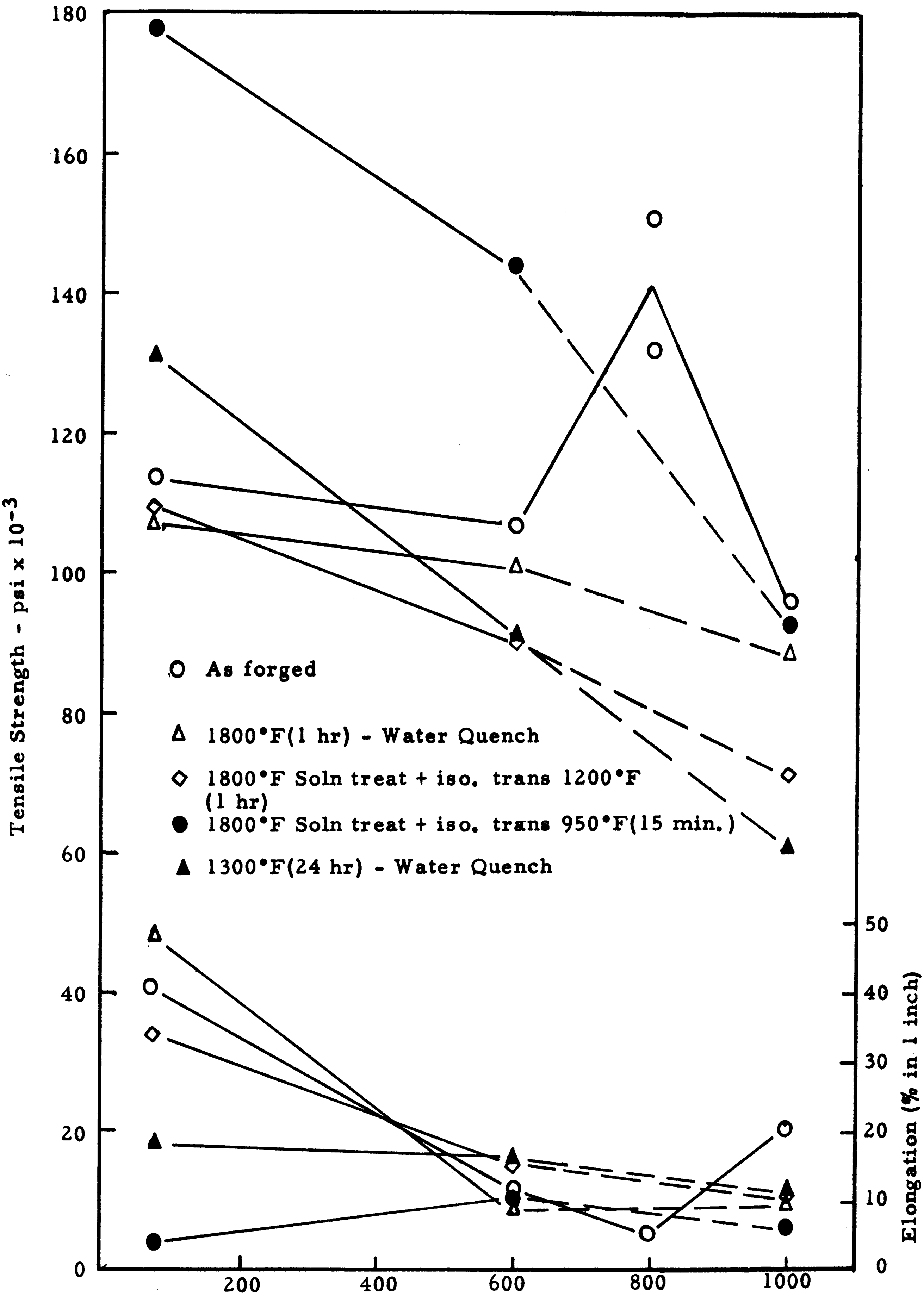


Figure 32. - Tensile Properties of Meta-Stable Beta Alloy 10 Mo.

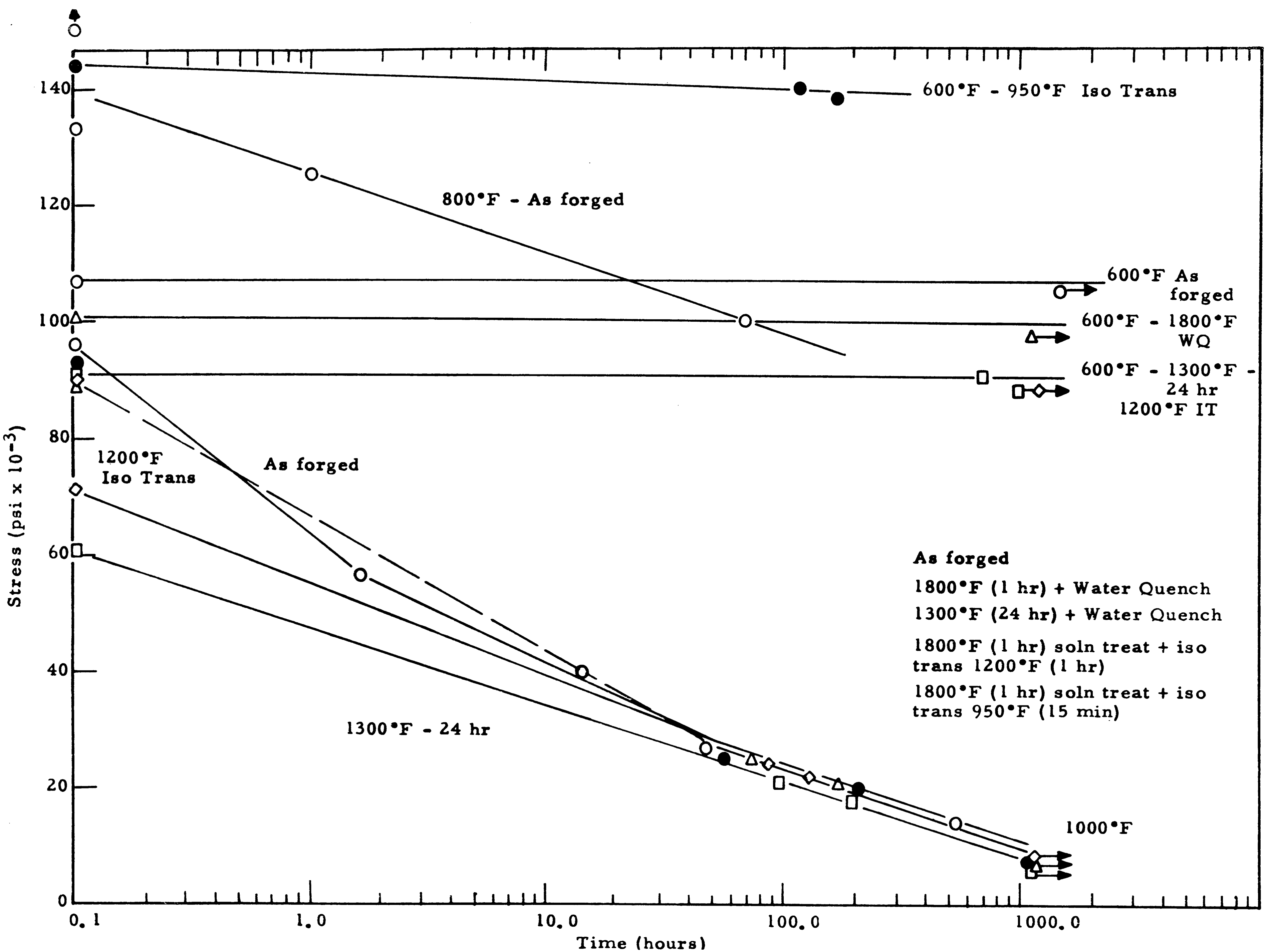


Figure 33 — Stress Rupture Time Curves for ...

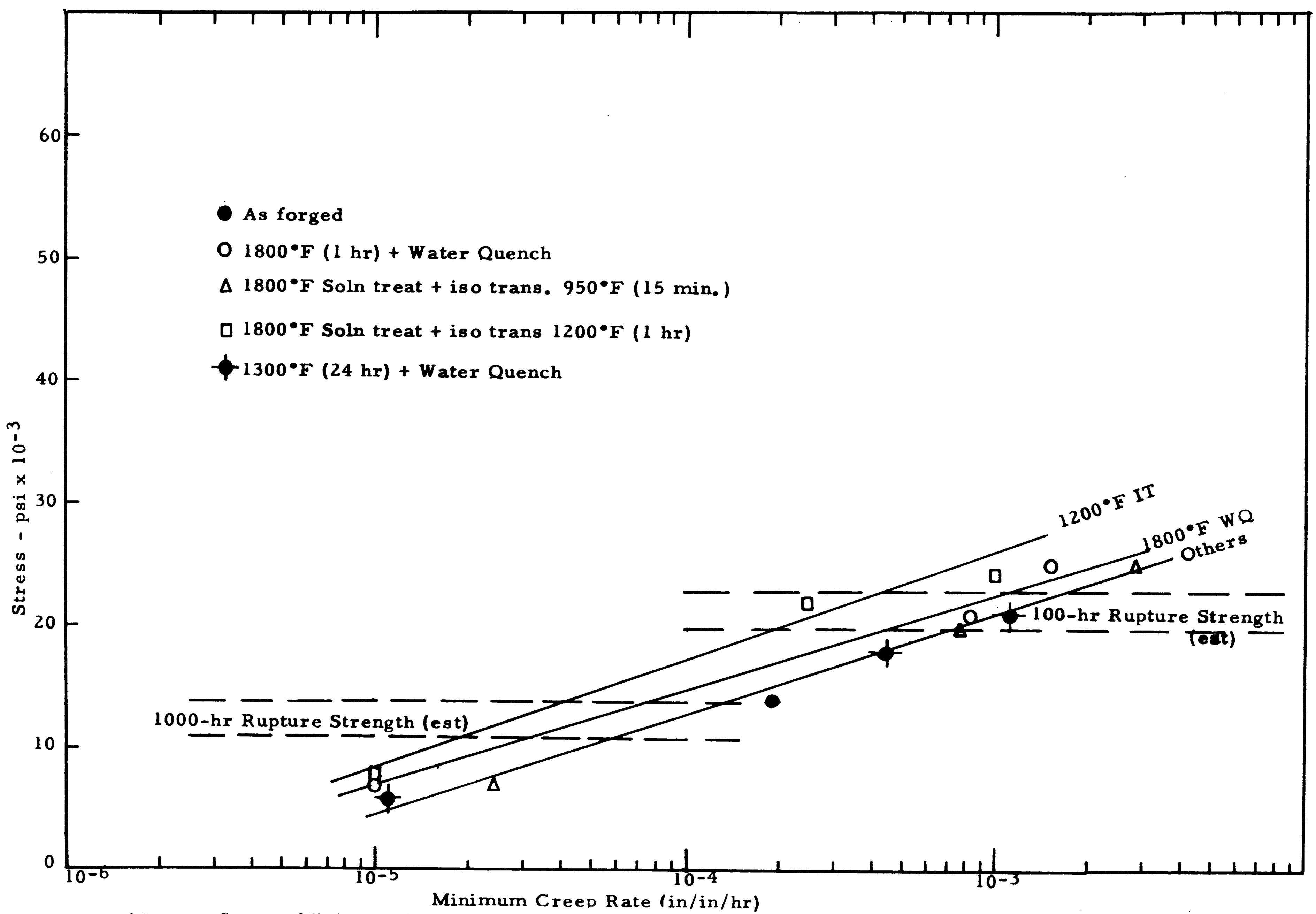


Figure 34. - Stress-Minimum Creep-Rate Curves at 1000°F for Meta-Stable Beta Alloy 10 Mo.

X1000

a. As Forged: Tested at 600°F for 1442 hours under 105,000 psi

X1000

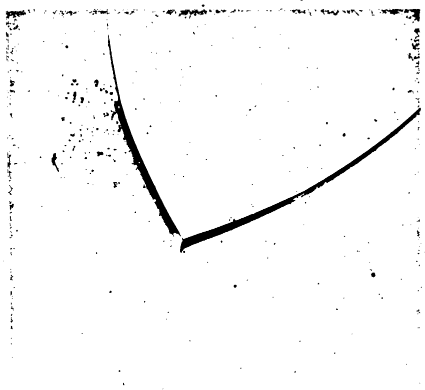
b. As Forged: Tested at 800°F for 68.1 hours under 100,000 psi

X500

c. **As Forged: Tested at 1000°F for 13.9 hours at 40,000 psi**

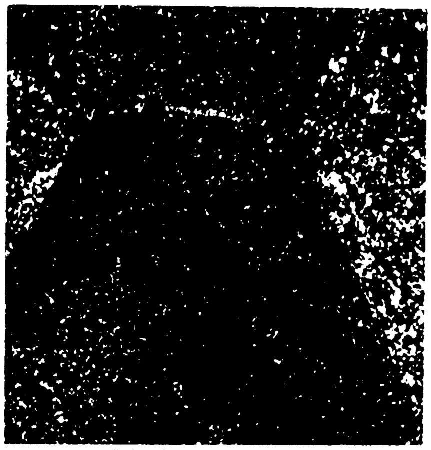
X500

d. As Forged: Tested at 1000°F for 502 hours under 14,000 psi



X1000

e. 1800°F - 1 hour - Water Quench - Tested at 600°F for 962.5 hours under 98,000 psi



X1000

f. 1800°F - 1 hour - Water Quench - Tested at 1000°F for 71.8 hours at 25,000 psi

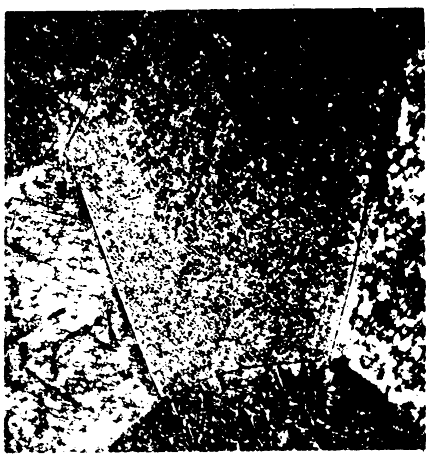
Figure 35. - Microstructures of Meta Stable Beta Alloy 10 Mo after Creep-Rupture Testing.

X500

g. 1300°F - 24 hours - Water Quench - Tested at 600°F for 963.6 hours under 89,500 psi

X1000

h. 1300°F - 24 hours - Water Quench - Tested at 1000°F for 1009 hours under 60,000 psi

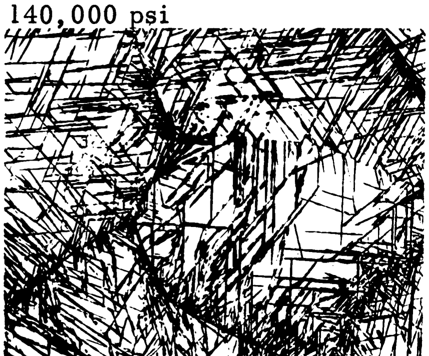


X500

i. 1800°F - 1 hour and isothermally transformed at 950°F for 15 minutes - Water Quench - Tested at 600°F for 118.2 hours under 140,000 psi

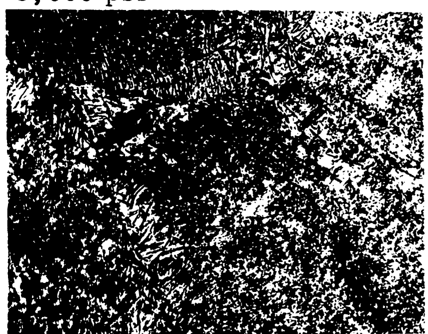
X500.

j. 1800°F - 1 hour and isothermally transformed at 950°F for 15 minutes - Water Quench - Tested at 1000°F for 54.2 hours under 25,000 psi



X500

k. 1800°F - 1 hour and isothermally transformed at 1200°F for 1 hour - Water Quench - Tested at 600°F for 938.1 hours under 88,500 psi



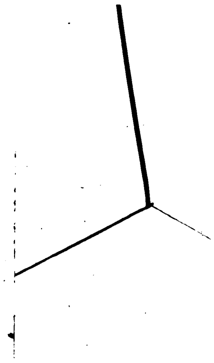
X500

l. 1800°F - 1 hour and isothermally transformed at 1200°F for 1 hour Water Quench - Tested at 1000°F for 122.4 hours under 22,000 psi

Figure 35. (concluded) - Microstructures of Meta Stable Beta Alloy 10 Mo after Creep-Rupture Testing.

X75

a. As Forged
R_C 39 hardness



X1000

b. 1800°F - 30 minutes - Water Quench - R_c 35 hardness

X1000

c. 1335°F - 4 hours - Water Quench - R_c 27.3 hardness

X1000

d. 1335°F - 24 hours - Water Quench - R_c 34 hardness

X1000

e. 1335°F - 4 hours - Water Quench - Reheated to 800°F for 24 hours - Water Quench R_c 34 hardness

X1000

f. 1300°F - 4 hours - Water Quench - R_c 30.7 hardness

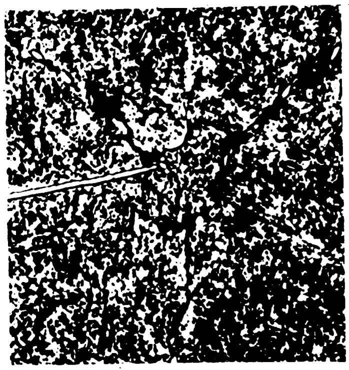
Figure 36. - Influence of Heat Treatment on Microstructure of Meta-Stable Beta Alloy 10 Cr.

X1000

g. 1265°F - 4 hours - Water Quench
R_C 42.3 hardness

X1000

h. 1265°F - 24 hours - Water Quench



X1000

i. 1265°F - 4 hours - Water Quench - Reheated to 800°F 24 hours - Water Quench
R_C 45.3 hardness

X1000

j. 1000°F - 5 hours - Air Cool
R_C 38.3 hardness

X1000

k. 1000°F - 24 hours - Air Cool
R_C 34.7 hardness

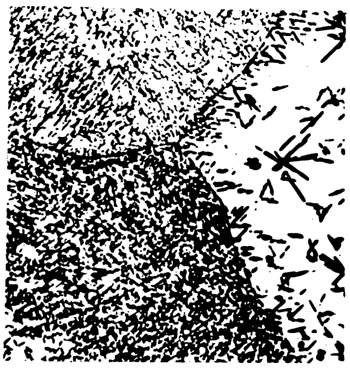
Figure 36. (concluded) - Influence of Heat Treatment on Microstructures of Meta-Stable Beta Alloy 10 Cr.

X1000

a. 1600°F - 15 minutes - Water Quench - R_C 39 hardness

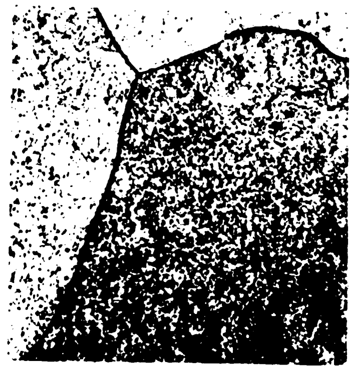
X1000

b. 1300°F - 10 minutes - Water Quench - R_C 33.3 hardness



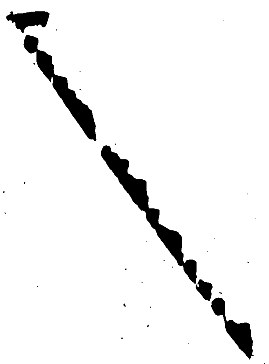
X1000

c. 1200°F - 15 minutes - Water Quench - R_C 34.7 hardness



X1000

d. 1000°F - 15 minutes - Water Quench - R_C 40.7 hardness



X1000

e. 1000°F - 30 minutes + isothermal transformation at 1470°F for 15 minutes - Water Quenched - R_C 37.3 hardness

a, b, c and d heated 1 hour at 1800°F and Water Quenched prior to indicated treatment.

Figure 37. - Influence of Reheating after Treatment at 1800°F on the Microstructure of Meta-Stable Beta Alloy 10 Cr.

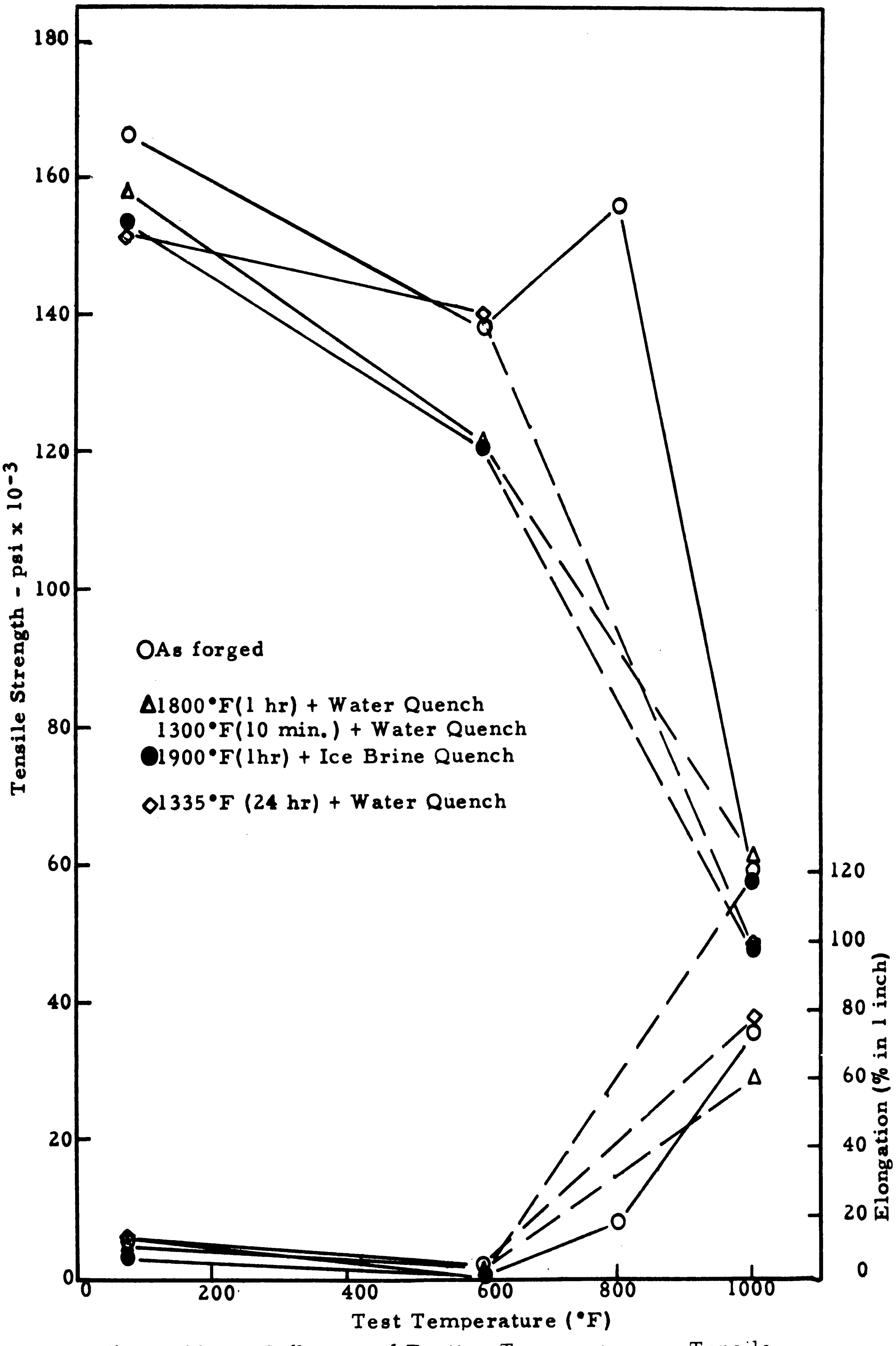


Figure 38. - Influence of Testing Temperature on Tensile Properties of Meta-Stable Beta Alloy 10 Cr.

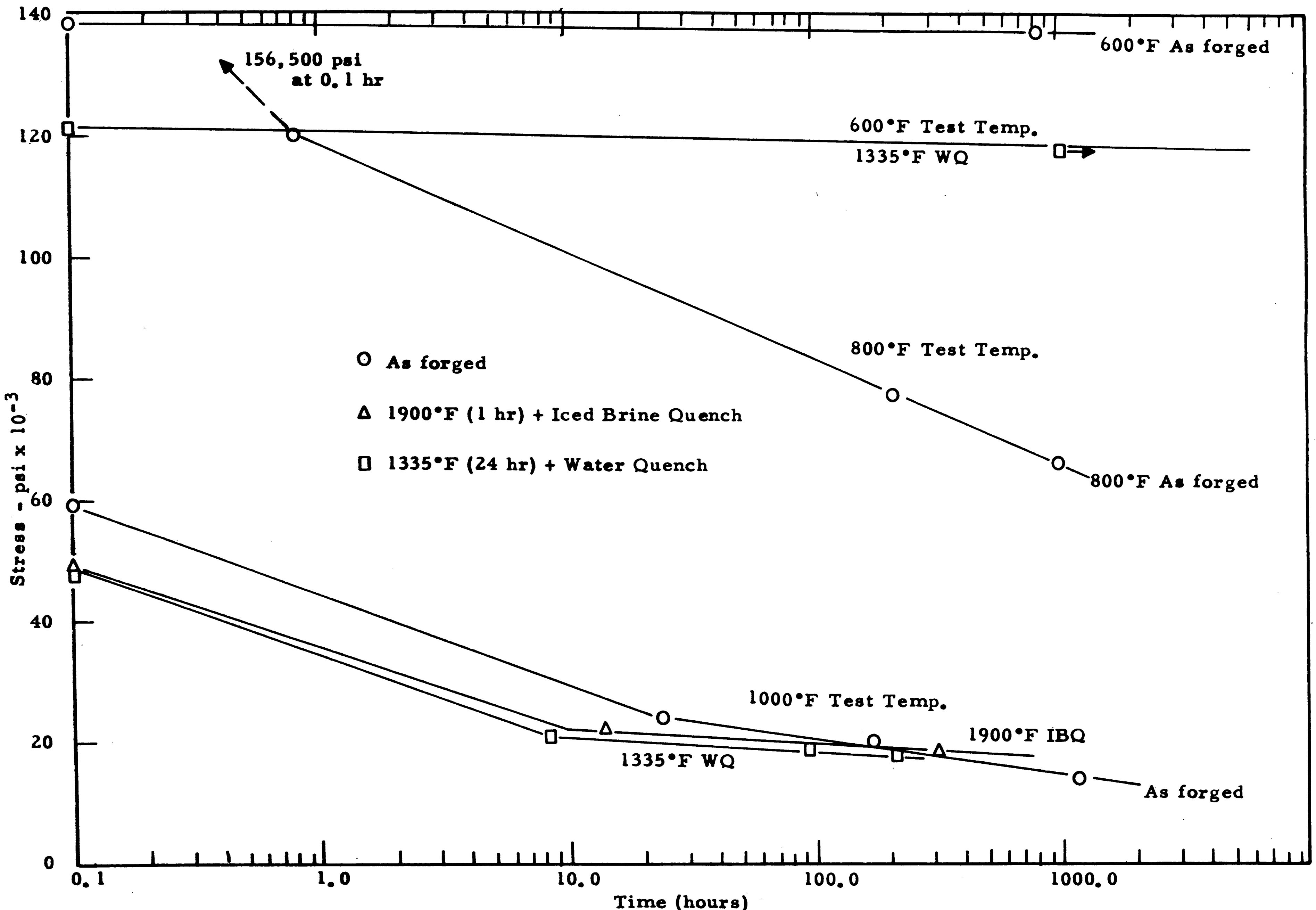
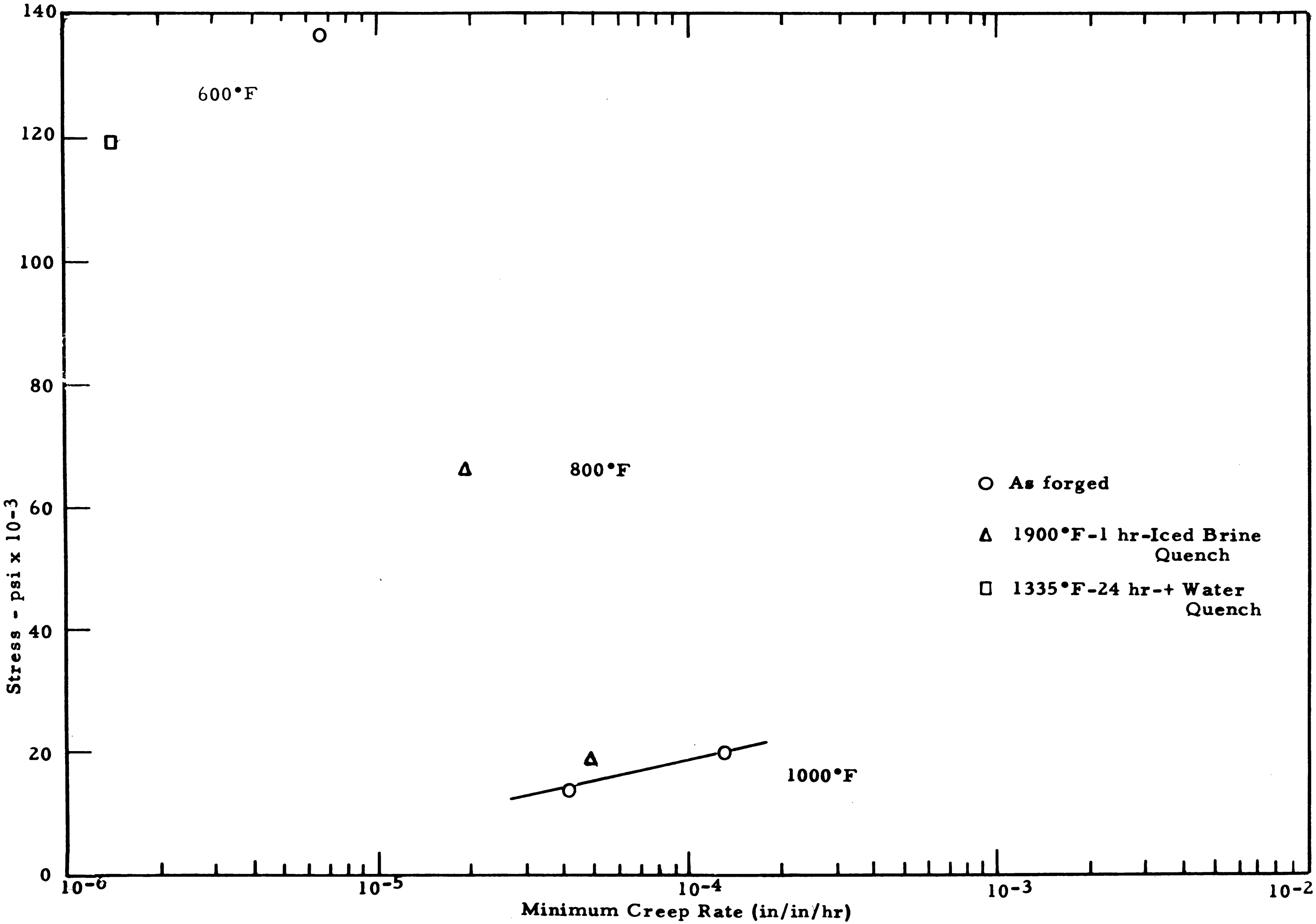
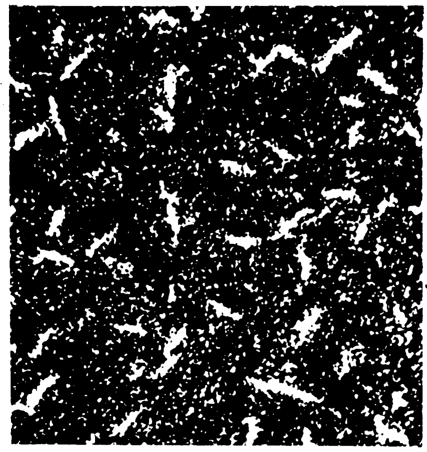


Figure 39. - Stress Rupture Time Curves at 600°, 800° and 1000°F for Meta-Stable Beta Alloy 10 Cr.



Stress - psi x 10-3
140
120
100
80
60
40
20
0
10-6
10-5
10-4
10-3
10-2
Minimum Creep Rate (in/in/hr)
600°F
800°F
1000°F
As forged
1900°F-1 hr-Iced Brine Quench
1335°F-24 hr-+ Water Quench

X1000
As Forged: Tested at 600°F for
801.7 hours under 137,000 psi



X1000
1900°F - 1 hour - Iced Brine Quench
Tested 1000°F for 13.9 hours under
22,500 psi.

X1000

1335°F - 24 hours - Water
Quench - Tested at 1000°F
for 209.2 hours under 18,000 psi

Figure 41. - Microstructures of Meta-Stable Beta Alloy 10 Cr after Creep-Rupture Testing.

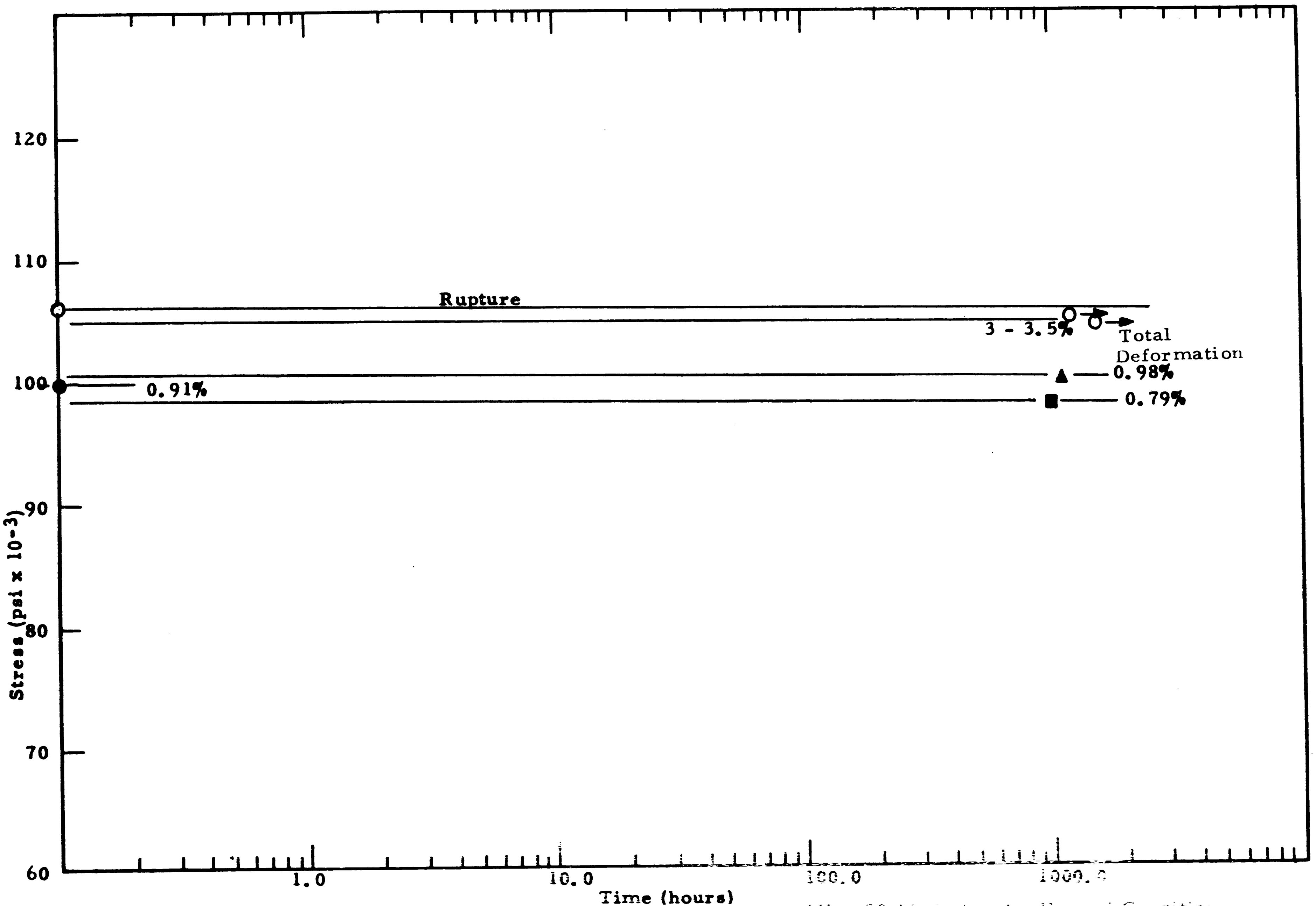


Figure 42. - Stress-Rupture Time Curves at 600°F for Stable Beta Alloy 30 Mo in the As-Forged Condition

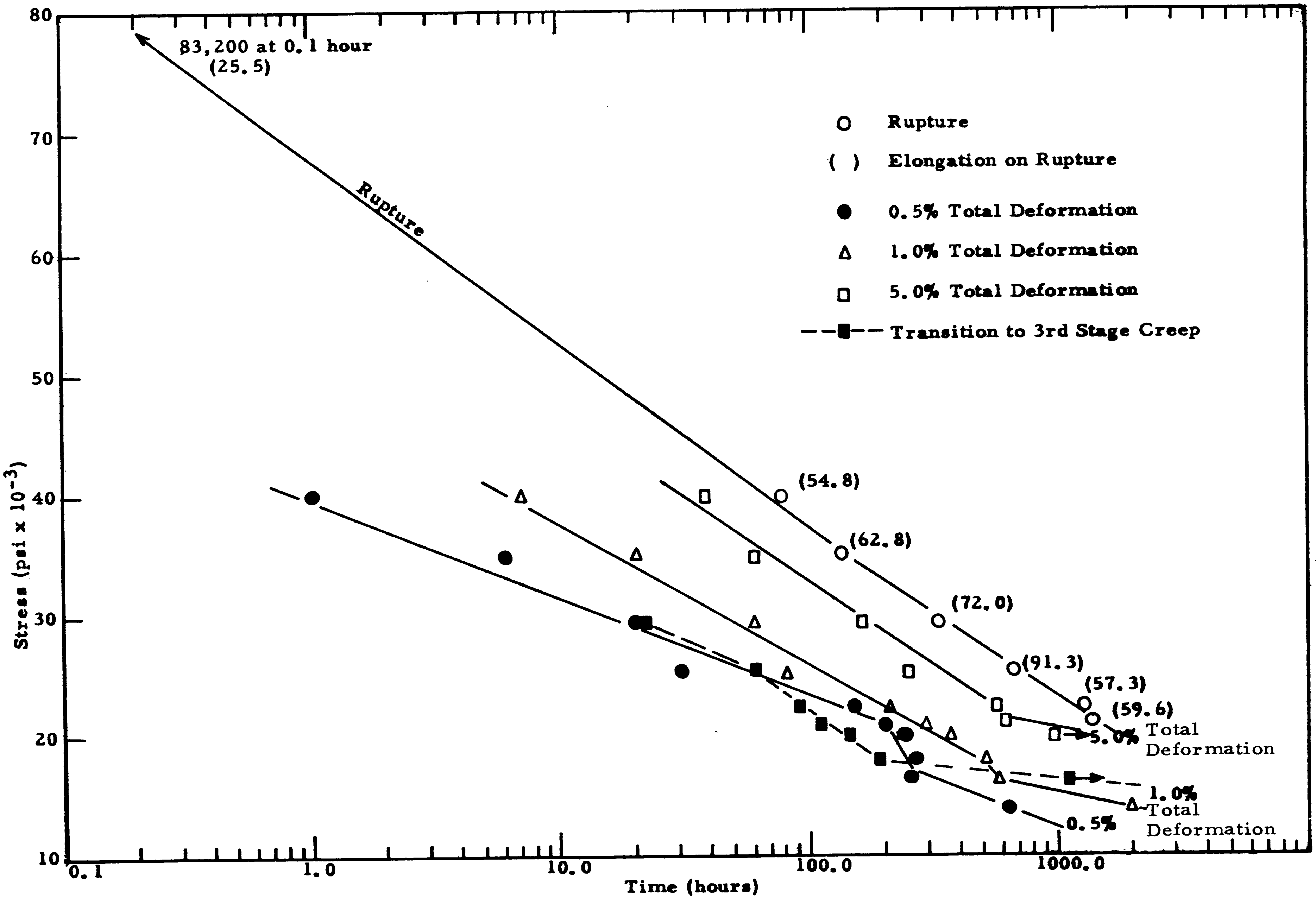


Figure 43. Stress-Rupture Time and Stress-Time for Total Deformation Curves at 1000°F for Stable Beta Alloy 30 Mo in the As-Forged C...

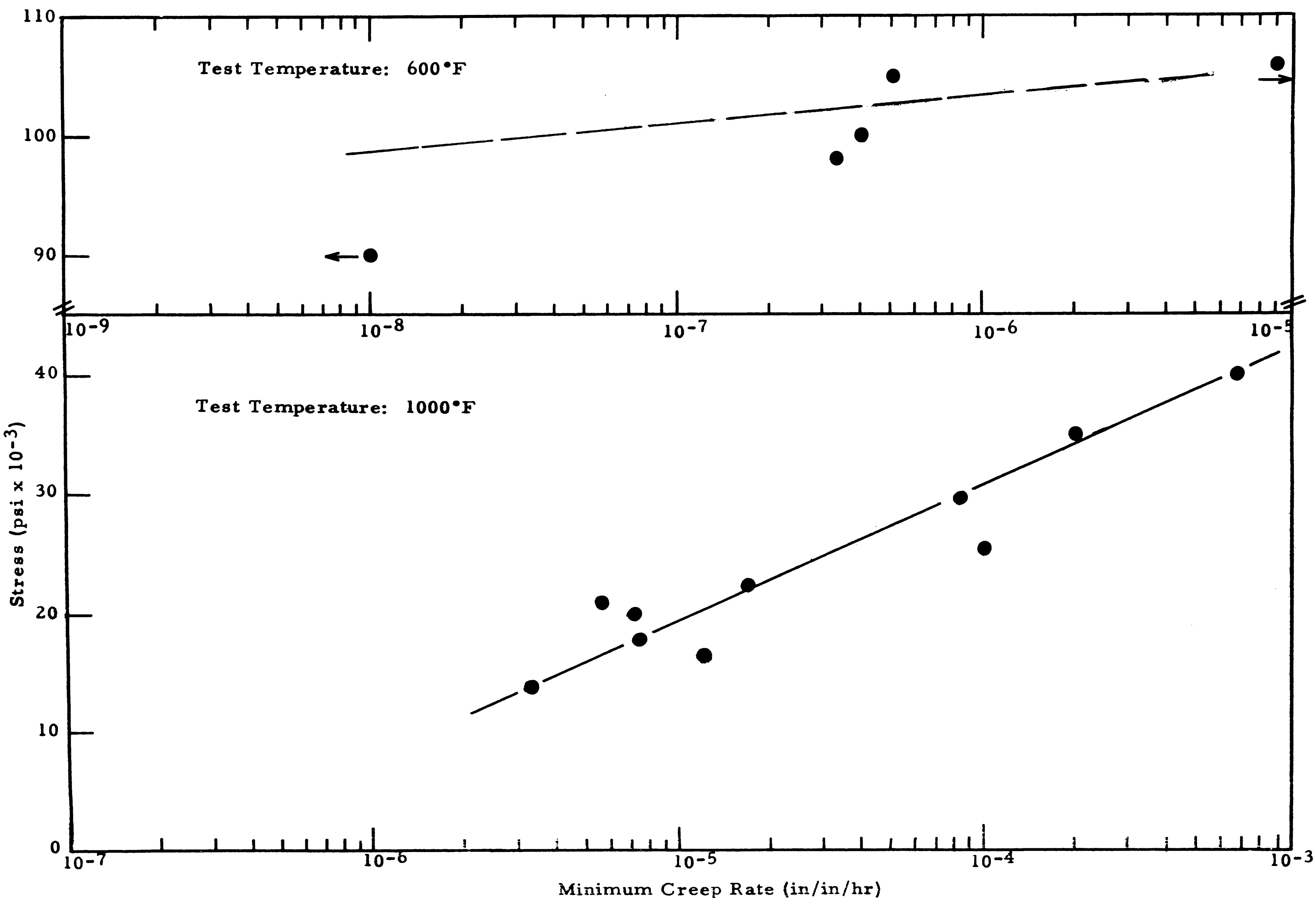


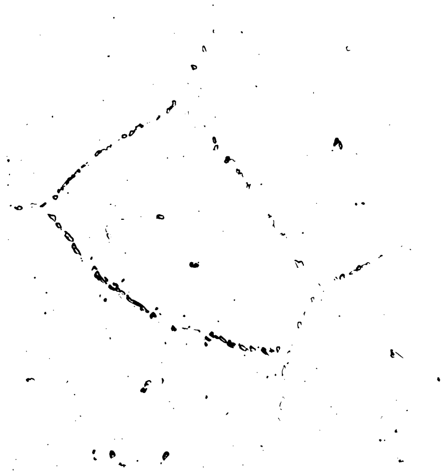
Figure 44. - Stress Minimum Creep Rate Curves at 600° and 1000°F for Stable Beta Alloy 30 Mo in the As-Forged Condition

X250

As Forged: Tested at 1000°F
for 2000 hours at 14,000 psi

Figure 45. - Microstructure of Stable Beta Alloy 30 Mo after Creep Testing at 1000°F.

X100
a. As Forged



X250
b. 1800°F - 4 hours - Water Quench

X100
c. 1925°F - 2 hours + Water Quench

Etchant: 1HF, 1 Glycerine

Figure 46. - Influence of Heat Treatment on the Microstructure of Stable Beta Alloy 50 V.

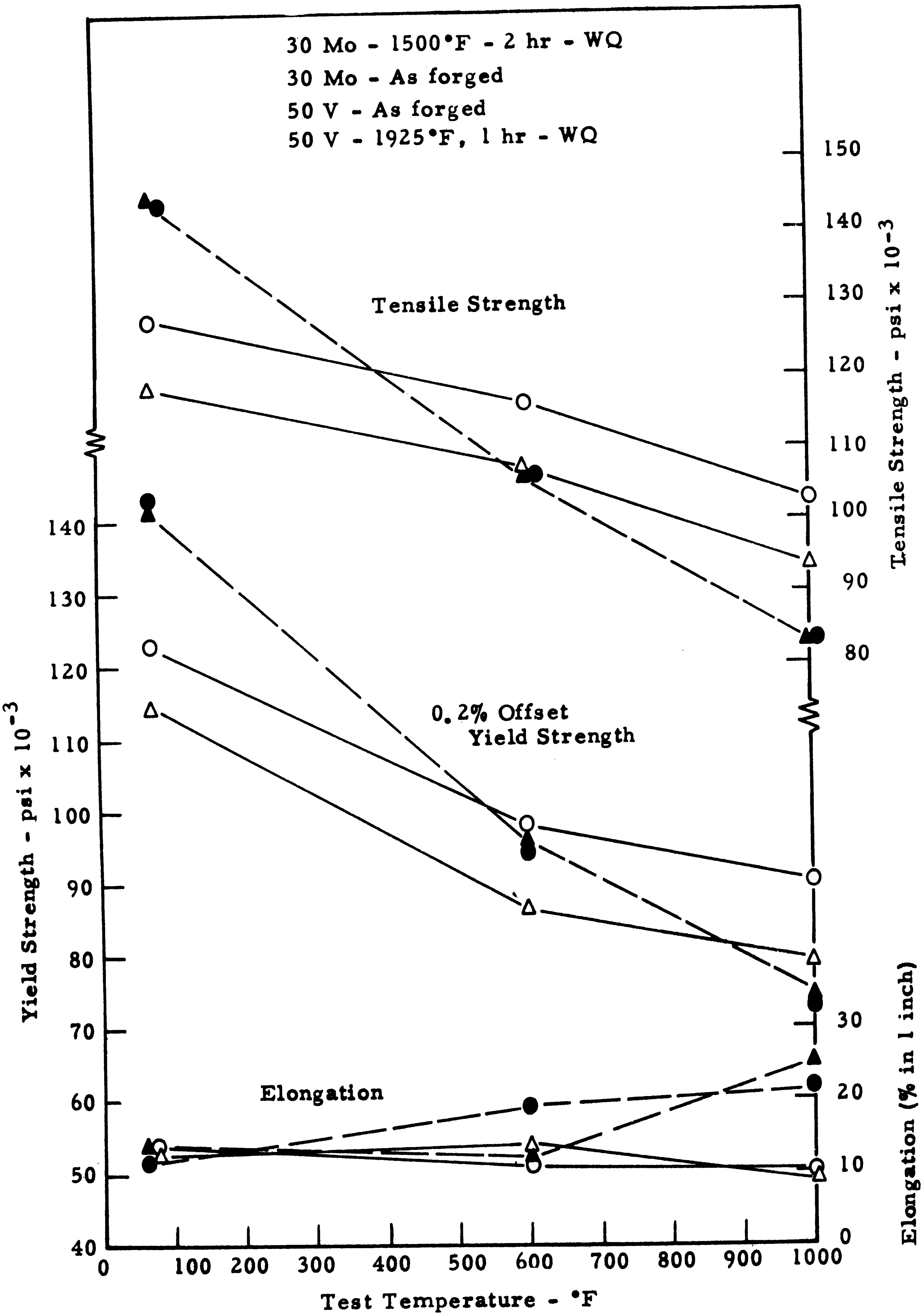


Figure 47. - Comparative Tensile Properties for Stable Beta Alloys 30 Mo and 50 V.

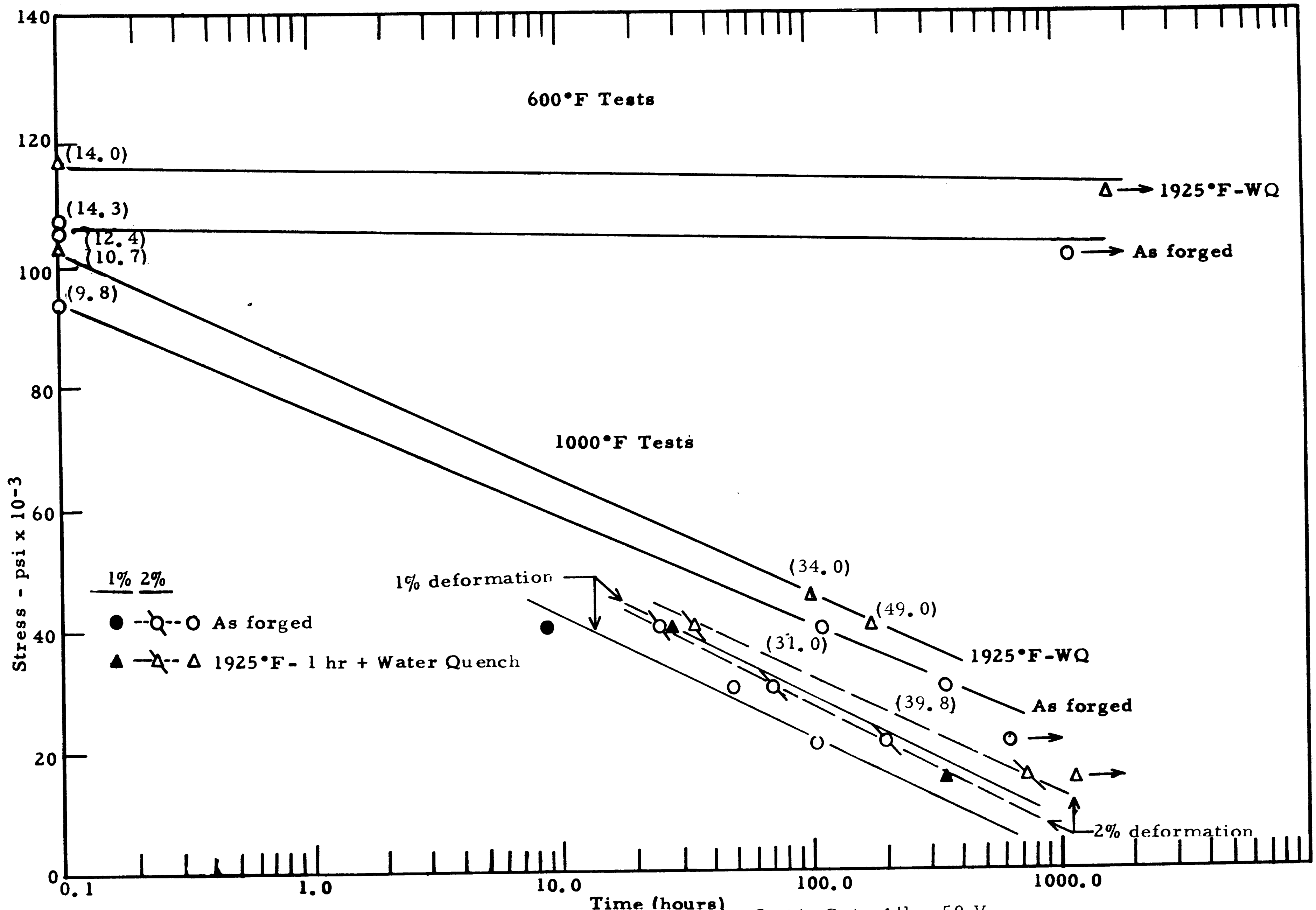


Figure 48. - Stress Rupture Time Curves at 600° and 1000°F for Stable Beta Alloy 50 V.

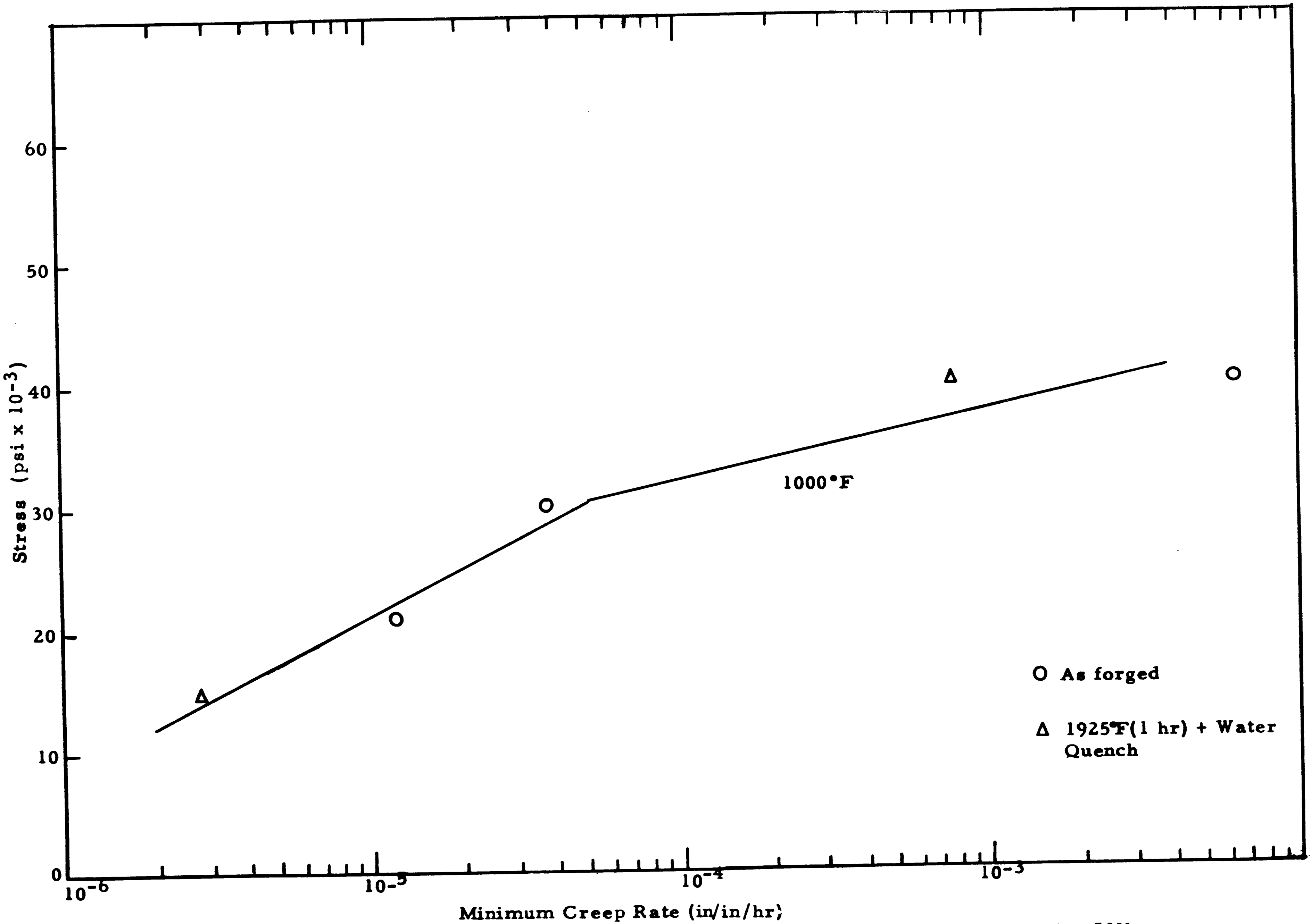
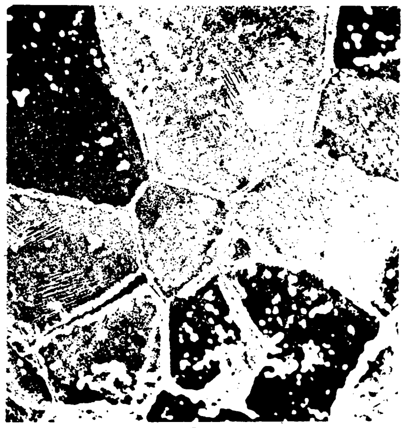


Figure 49 - Stress Minimum Creep Rate Curves at 1000°F for Stable Beta Alloy 50V.

X250

a. As Forged
Tested at 600°F for 1107 hours
under 100,000 psi

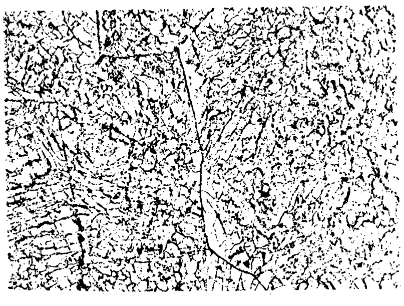


X100

b. As Forged
Tested at 1000°F for 110.6
hours under 40,000 psi

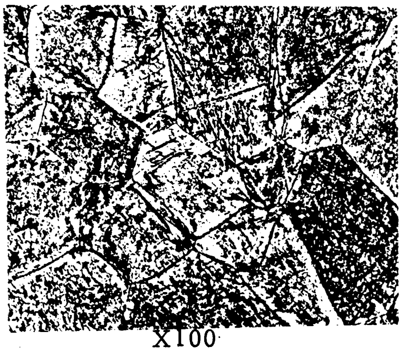
X100

c. As Forged
Tested at 1000°F for 346.6
hours under 30,000 psi



X100

d. 1925°F - 2 hours + Water Quench
Tested at 1000°F for 100.6 hours
under 45,000 psi



X100

e. 1925°F - 2 hours - Water Quench
Tested at 1000°F for 173.6 hours
under 40,000 psi

Etchant: 1HF, 1 Glycerine

Figure 50. - Microstructures of Stable Beta Alloy 50 V after Creep-Rupture Testing.

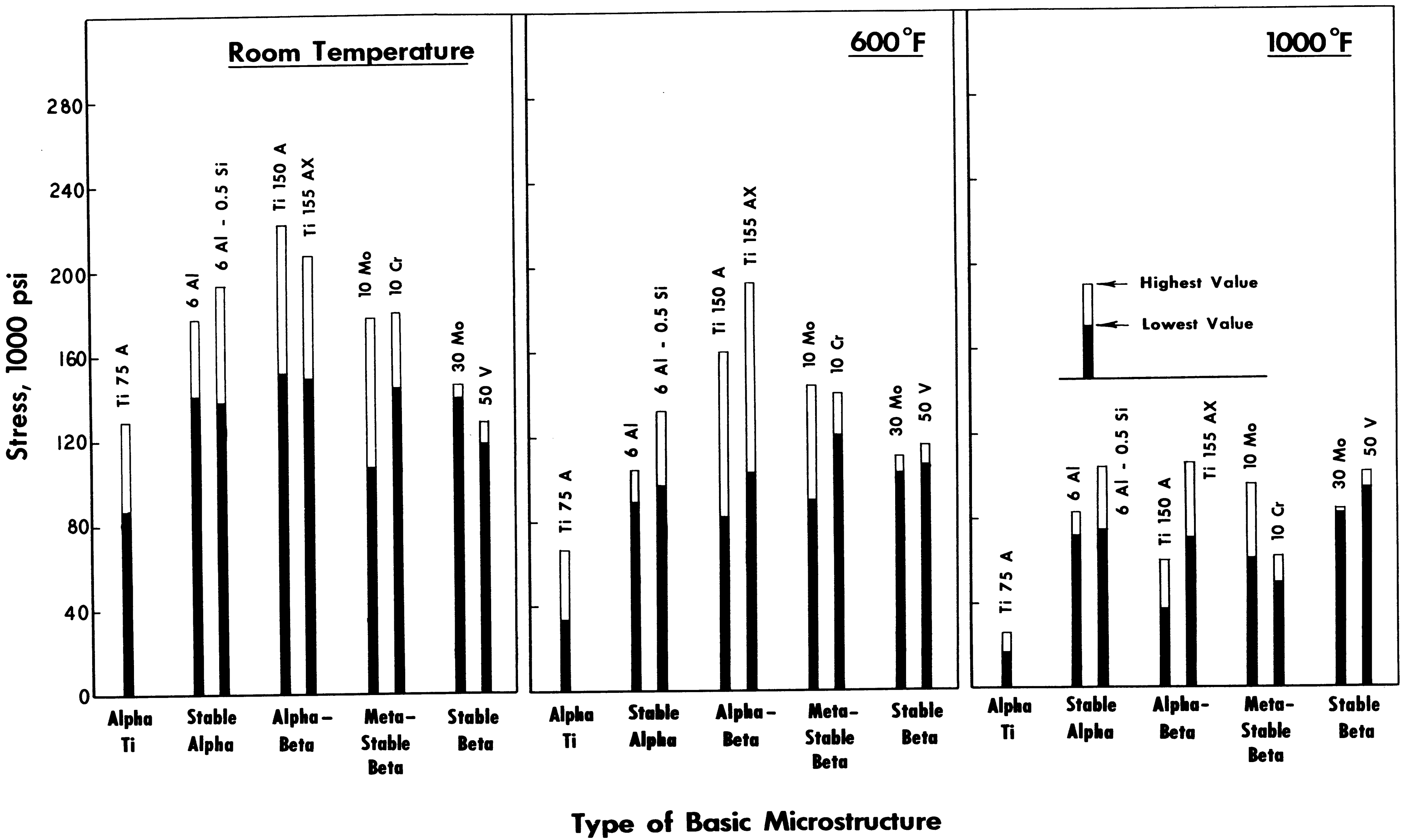


Figure 51 - Experimentally Observed Influence of Type of Microstructure and Alloy Composition on the Tensile Strength of Titanium Alloys at 75°, 600° and 1000°F

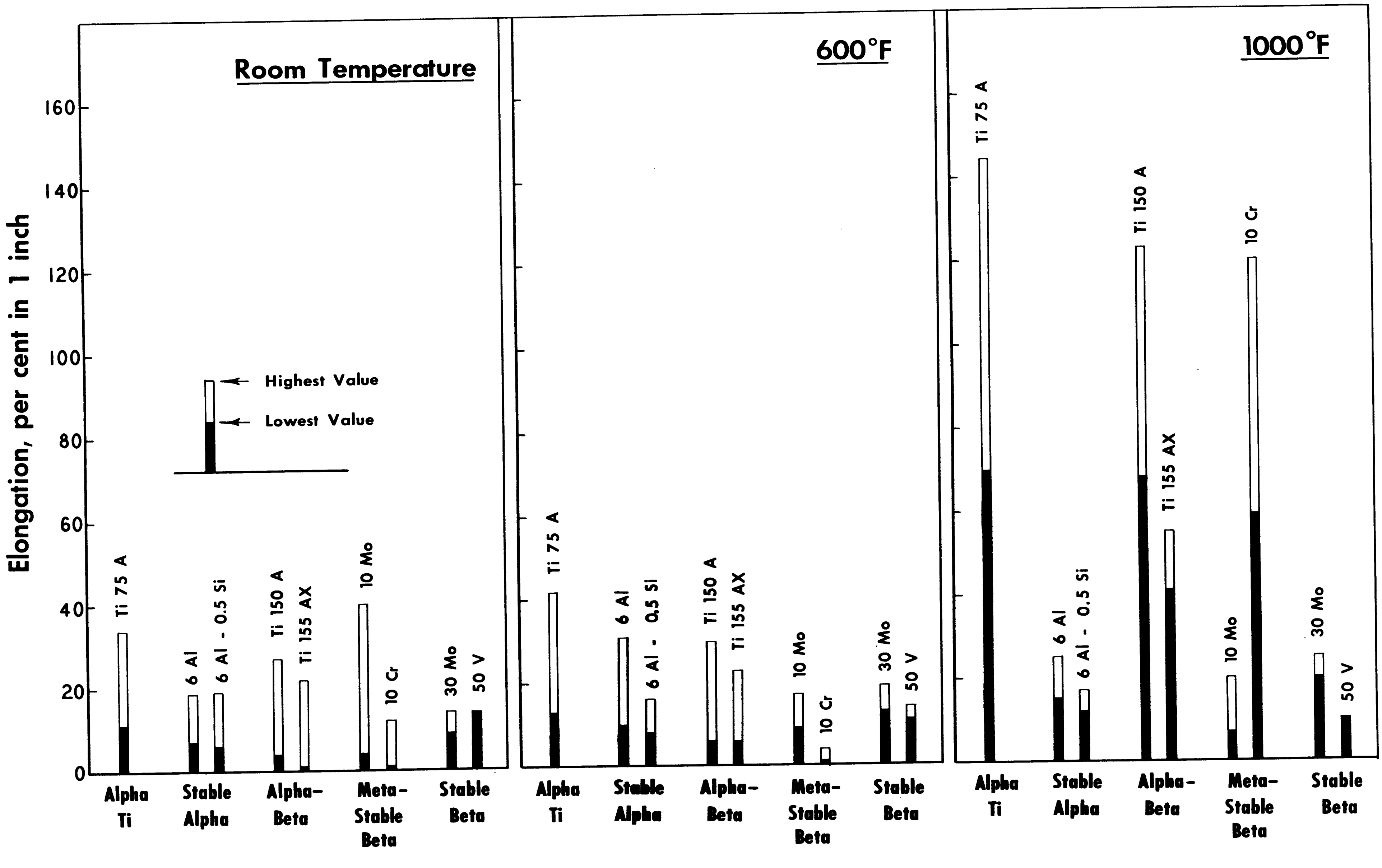


Figure 52 - Experimentally Observed Influence of Type of Microstructure and Alloy Composition on the Tensile Test Elongation of Titanium Alloys at 75°, 600° and 1000°F

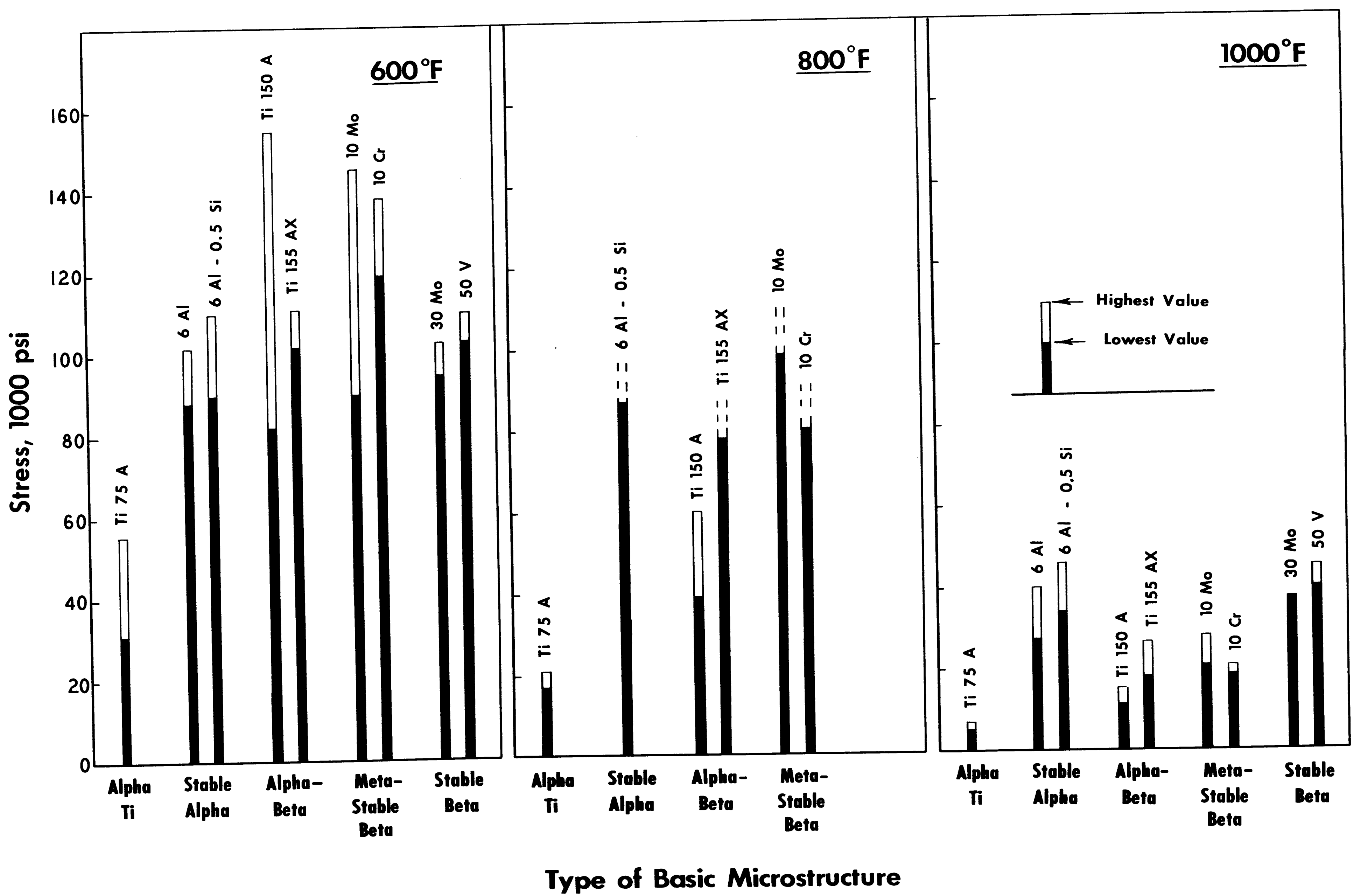


Figure 53 - Experimentally Observed Influence of Type of Microstructure and Alloy Composition on the 100-Hour Rupture

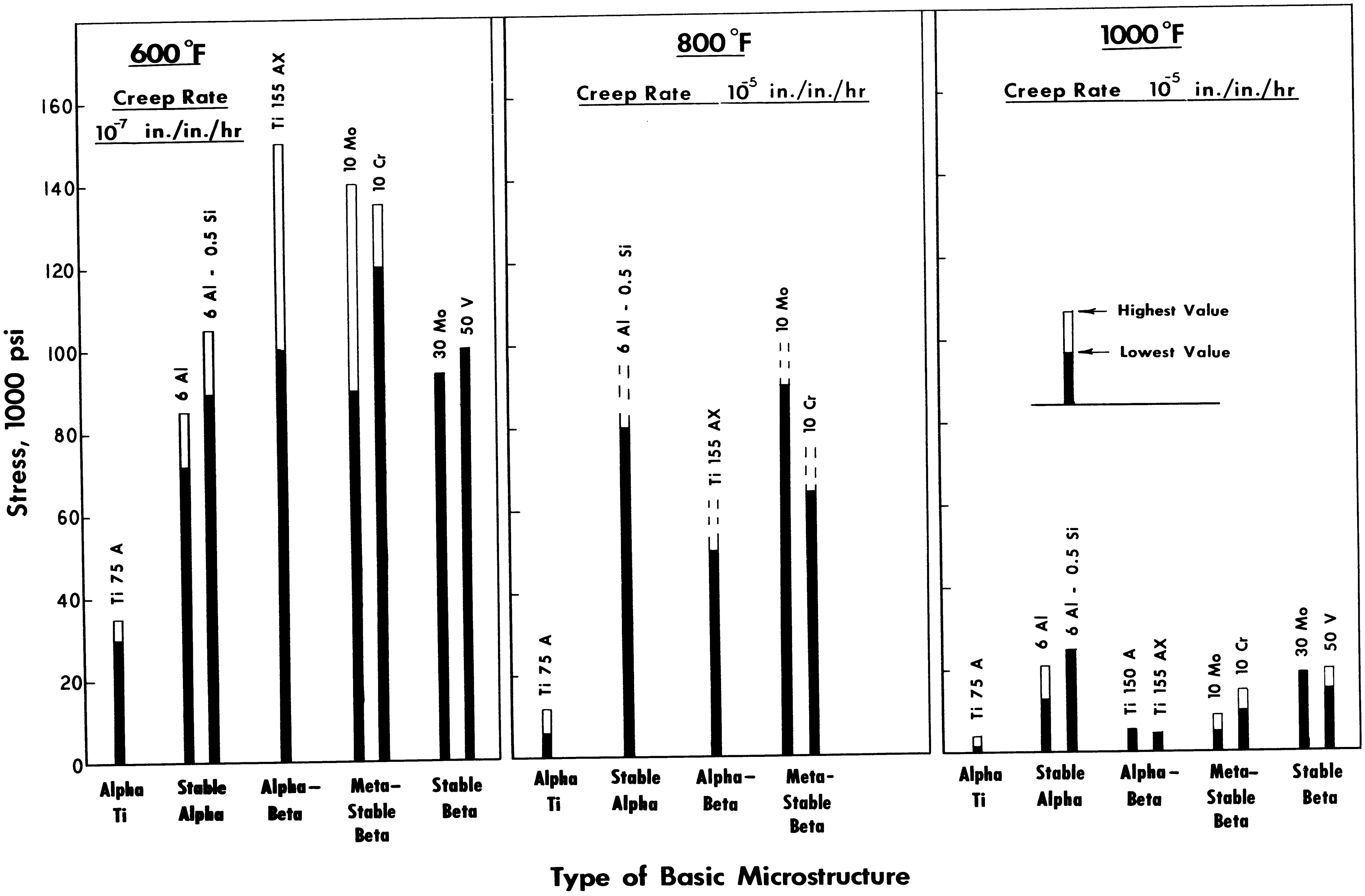


Figure 54 - Experimentally Observed Influence of Type of Microstructure and Alloy Composition on the Creep Resistance