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

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FOREWORD

This report was prepared by the Engineering Research Institute of the University of Michigan, Ann Arbor, under USAF Contract No. AF 33(616)-244. The contract was initiated under Research and Development Order No. 615-11, "Titanium Metal and Alloys" and Research and Development Order No. 615-13, "High Temperature Alloys" and was administered under the Direction of the Materials Laboratory, Directorate of Research, Wright Air Development Center, with Dr. H. K. Adenstadt, Lt. A. G. Forrest and Lt. H. M. Burte acting as project engineers.

ABSTRACT

This report summarizes the work performed under Contract No. AF 33(616)-244 during the period from 1 July 1952 to 30 September 1953. The object of the investigation was to determine the relationship between typical structural conditions of representative titanium-base alloys and their mechanical properties in the range from 600° to 1000°F. In order to accomplish this purpose, tensile, short time rupture, and creep properties were determined. The materials studied were: commercially pure titanium, Ti 75A; a commercial alpha-beta alloy, Ti 150A; an experimental stable alpha alloy, 6% Al - 94% Ti; and an experimental stable beta alloy, 30% Mo - 70% Ti. In addition, hardness, x-ray, and metallographic studies were made.

The results indicate that the single phase α or β alloys possess the best combination of properties at 1000°F. Commercially pure titanium, Ti 75A, exhibited lower properties than the alloys at all test temperatures. Wide ranges of strength variation through heat treatment were found for Ti 150A at 600° and 800°F; however, at 1000°F no essential difference between heat treatments was evident, and the absolute level of strength was quite low. The β alloy, 30% Mo, showed little or no response to prior treatment, but had the best combination of creep and rupture properties at 1000°F of the four materials tested. Both Ti 75A and the α alloy, 6% Al, showed improvement in strengths at 600°F from small amounts of cold working. At 1000°F, however, an optimum amount of cold work, less than 10%, existed, above which properties were lowered through metallurgical instability.

PUBLICATION REVIEW

This report has been reviewed and is approved.

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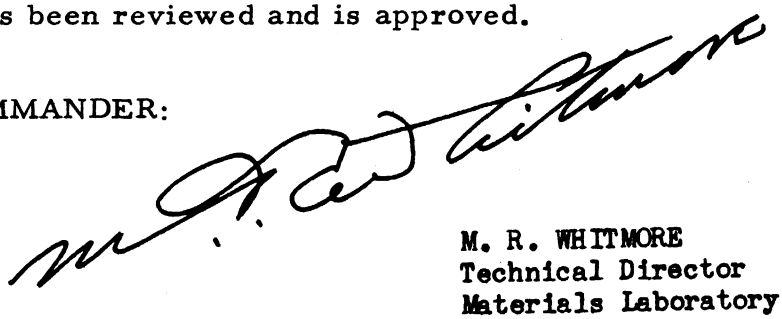

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INTRODUCTION

It was only four years ago that Adenstadt at Materials Laboratory, WADC, first reported that titanium underwent significant amounts of continued deformation by creep at room temperature under stresses well below the yield point (13). This focussed attention on the necessity of considering creep characteristics of titanium base materials in their use at all temperatures. The investigation covered by this report was undertaken to develop basic information on the creep characteristics of the various types of titanium alloys and the influence of variations of structure within any one type of alloy. The temperature range considered was 600° to 1000°F. The study was a logical outgrowth of a similar study on commercial alloys over temperatures up to 600°F which had been started previously under sponsorship of the OAR, WADC (8).

In an investigation of scope as wide as the present one, it was difficult to select any one alloy and/or condition for singular emphasis. Instead, it was desirable to survey in a systematic fashion the range of alloy types and their possible structures. Purely on the basis of metallographically observed response to heat treatments, it was possible to characterize titanium alloys as "type" alloys and selection of materials for study under this contract was made on that basis. Commercially pure titanium formed the base condition for the work.

Even commercially pure titanium can be considered an alloy, since the small amounts of impurities present have an effect on its properties. The other materials chosen for study had the following characteristics: an alloy undergoing eutectoid decomposition; an alloy in which the alpha is stabilized; a meta-stable beta alloy, one in which beta retention would be possible, yet which would permit structural variation from isothermal treatments; and a stable beta alloy, one permitting complete beta retention on slow cooling.

Thus, the materials selected for study were the following:

1. Commercially pure titanium (Ti 75A).
2. Eutectoid decomposition alloy (Ti 150A).
3. Stable alpha alloy (6% Al).
4. Meta-stable beta alloy (10% Mo).
5. Stable beta alloy (30% Mo).

The specific alloys chosen to represent the various types were determined by consultation with representatives of the Materials Laboratory, WADC, and with representatives of the Armour Research Foundation. The latter group were engaged in a program of alloy development under the auspices of the Materials Laboratory, WADC. The first two materials were available commercially, and the latter three were prepared by Armour.

A further consideration in planning the investigation were the conditions of the materials to be studied. It is well known that the effects of prior treatment often have more influence on rupture and creep behavior than do wide ranges of chemical composition. Prior treatment variables include hot and cold working, temperatures and cooling rates of solution treatments, and temperatures and times of aging treatments. Melting practice has also been shown to be of prime importance. The testing program was set up to survey systematically such treatments as appropriate to each alloy and to define general principles. The subject is covered in detail in the section on Procedure.

Finally, it was necessary to consider the type of results which might be expected. Relations between creep behavior and treatments are often very complicated. A treatment producing maximum creep resistance at a low testing temperature may have just the opposite result at a higher temperature. Treatments best for short time periods of service may not be best for prolonged times. Relationships may change, depending on the deformation criteria over the time interval under consideration. Therefore, it was necessary to set up the test program so that the maximum amount of information could be obtained regarding both short time - high stress and longer time - low stress characteristics. The subject is also covered in the section on Procedure.

TEST MATERIALS

After consultation between representatives of the Wright Air Development Center and the University of Michigan, five materials, four of which were received in the contract period, were ordered for the investigation. The test materials were selected, as mentioned previously, in order to be representative of the "type" structures possible with titanium alloy systems. Two were available from commercial sources. The remaining three were ordered in the form of small (approximately 10 pounds each) experimental ingots and were melted for the University by special arrangements with the Armour Research Foundation of Chicago, Illinois.

Material in the form of bar stock was chosen for study in order to minimize both procurement difficulties and contamination problems during treatment and testing.

All presently available data concerning the test material are summarized in Table 1.

The melting procedure followed by Armour Research Foundation for the two heats received was as follows:

6% Al alloy--"Five 1000-gram ingots were melted in the non-consumable electrode (tungsten tip) arc melting furnace in water-cooled copper spun crucibles, using the conveyor-type charging hopper. Individual 100-gram charges of the desired composition were evenly distributed over the conveyor belt and added slowly to previously molten material by revolving the belt.

"The resulting cylindrical ingots, which measured approximately 2-1/2 in. in height by 3 in. in diameter, were forged to 1/2 in. diameter round bars at a maximum temperature of 2000°F.

"After sand-blasting and acid dipping in an aqueous solution of hydrofluoric and sulfuric acids, the bars were tapped and threaded at alternate ends, and joined to give one continuous bar which was remelted in the consumable electrode furnace. The final ingot weighing about 9.5 pounds (material having been lost due to scalping), measured 3 in. in diameter by 8-1/2 in. in height. This ingot was forged at a maximum temperature of 2100°F to 1/2 in. -round and -square bars." (1)

30% Mo alloy--The above procedure was somewhat modified due to difficulties with Mo alloy preparation. "There was considerable difficulty in obtaining non-segregated homogenuous ingots of this composition using our standard practice of adding a 60% molybdenum-titanium master alloy to our arc melting furnace. We found it necessary to prepare individual 150 gram ingots of 30% molybdenum, each melted and re-melted at least five times. Each of these pancake type ingots was then rolled to approximately 1/16 in. sheet which was then cut into approximately 1/4" sections. This cut stock was then recharged to the tungsten electrode arc melting furnace in which we prepared the five 2-pound ingots then forged to rod at 2250°F" (2)

PROCEDURE

Test Procedure

In general, the experimental program was planned to study the effects on creep behavior of the variables appropriate to each alloy under consideration.

Testing was guided by experience from other research work in progress at lower temperatures. In the present program, tests were confined to 600° and 1000°F for the experimental alloys; and 600°, 800° and 1000°F, for the commercial materials. The effects of the variables were surveyed by the following testing procedure at each temperature.*

(a) A short time tensile test. The reason for running this test was to aid in selection of rupture test stresses. The tensile strength was arbitrarily used as the 0.1-hour rupture strength. In addition, a room temperature tensile test was run on each condition in order to survey relative properties and/or brittleness.

* No total deformation data are included in this report. Instead, a supplementary report containing all creep curves will be issued.

(b) Tests designed to cause fracture in 30-40 hours and 100 hours. It must be realized that it was unusual to obtain these exact times. An effort was made to cover the time period with two tests.

(c) A creep test at a stress level which would result in a creep rate of between 10^{-5} and 10^{-6} inches per inch per hour and for a length of time sufficient for the approximation of a second stage creep rate.

These tests were adequate to indicate relative rupture and creep strength properties.

In addition to creep, rupture, and tensile testing, the following general techniques were employed to delineate the observed influences of metallurgical variables.

(a) Microstructural examination both before and after creep and rupture testing.

(b) Hardness changes due both to initial treatment and to the effect of testing.

(c) Changes in x-ray diffraction characteristics induced by treatments, particularly diffraction line broadness where applicable as a measure of internal strain and lattice parameter changes as a measure of precipitation effects.

The purpose of these measurements was to establish the effects of grain size, cold work, and precipitation phenomena on creep characteristics. Where $\beta \longrightarrow \alpha$ transformation changed not only the relative amounts of the phases, but also their distribution, the influences of both effects were studied.

Variables of Investigation

As mentioned in the Introduction, the actual test program was set up to survey systematically the metallurgical treatments most appropriate to each "type" alloy.

In the case of Ti 75A and Ti 150A, investigation of which had been in progress on sheet stock for some time, there were data available as a guide (8). The objectives of that investigation were similar to those of the present investigation, except that elevated temperature testing was confined to the temperature range of 76° to 600°F. The work on the other three alloys was guided mainly by the phase diagrams developed for Materials Laboratory, WADC, by Armour Research Foundation and by the experience Armour had in producing the alloys.

The main structural variations and reasons for considering them for each alloy were:

1. Commercially pure titanium (Ti 75A). This material is not subject to the control of microstructure by regulating the cooling rate from the all

β phase region. Consequently, microstructures were governed by hot rolling or by cold working and annealing. Cold work in itself was an important consideration in controlling properties. Test stock was supplied in the commercial hot-rolled condition. Samples were prepared using two annealing temperatures and two degrees of cold work for each. The work on sheet had shown that cold work improved creep resistance remarkably up to 600°F. One result of this investigation was to define the maximum temperatures for improvement of strength by cold work.

The work on sheet had also shown that Ti 75A was subject to considerable alteration of creep resistance due to a strain aging type of reaction between 210° and 600°F. It was not expected that this would be important in the present investigation. Care was exercised to determine if other significant structural changes might occur as a result of test conditions. No changes could be found by microscopic examination.

Test Conditions--commercially pure titanium (Ti 75A). Tested at 600°, 800°, 1000°F.

1. As-produced (hot rolled and annealed).
2. Annealed at 1500°F (1 hour at temperature plus furnace cool-approximately 16 hours.)
3. Annealed at 1700°F.
4. Annealed at 1500°F + 10.9% cold work.
5. Annealed at 1500°F + 31.3% cold work.
6. Annealed at 1700°F + 10.9% cold work.
7. Annealed at 1700°F + 31.3% cold work.

2. Commercial α + β alloy (Ti 150A). This alloy is subject to considerable variation in properties due to variation in the relative proportions of α + β. In heat treatments involving simple water quenching or air cooling, it was necessary to limit heating to below the temperature at which an all β structures forms in order to avoid excessive brittleness at room temperature (8). Consequently, the heat treatments were almost entirely restricted to the α + β region. Indications were, however, that creep properties are dependent on the cooling rates as well as the temperatures of the heat treatments. Hence, it was planned to include at least two temperatures of heating, followed by water quenching, air cooling and annealing. Subsequent re-heat treatments were also projected.

Variation in temperature of heating controlled the relative amounts of α + β, the amount of β increasing with increasing temperature. Cooling rate controlled the amount of β retained in the structure--the more rapid the cooling, the more β retained at room temperature. The previous work on sheet (8) indicated that increased amounts of retained β were accompanied by greatly increased creep resistance. The retained β, however, tended to transform during testing with a decrease in volume which caused complex creep characteristics.

In addition, some experiments were designed to show the properties of the alloy resulting from full isothermal transformation. Full isothermal trans-

formation involves heating the specimen to the all β region, quenching it in a molten lead bath which is maintained at the desired transformation temperature, holding the specimen at that temperature until the β structure has transformed into the equilibrium amounts of α and β, and finally quenching it in water.

Test Conditions--commercial α + β alloy (Ti 150A). Tested at 600°, 800°, 1000°F.

1. As-produced (hot rolled and annealed).
2. Annealed at 1500°F.
3. Air Cooled from 1500°F.
4. Water Quenched from 1500°F.
5. Annealed at 1350°F.
6. Air Cooled from 1350°F.
7. Water Quenched from 1350°F.
8. Water Quenched from 1500°F + 1350°F Anneal.
9. Air Cooled from 1500°F + 900°F Anneal.
10. Water Quenched from 1350°F + 900°F Anneal.
11. Solution Treat at 1800°F + Full Isothermal Transformation at 1300°F.
12. Solution Treat at 1800°F + Full Isothermal Transformation at 1000°F.

Items 1 - 7 represented variations in the proportions of α + β; and items 8 - 10, variations of α + β proportions through a different approach. Items 11 and 12 were representative isothermal transformation structures.

3. 6% Al alpha alloy. Since the 6% Al alloy consists of a stable α structure at all testing temperatures, grain size and degree of cold work are the major variables affecting its creep and rupture properties. Thus the testing variables chosen involved cold working followed by varying degrees (including zero) of recrystallization. In addition, testing was carried out on material water quenched from the all β region because a significantly different type of alpha structure was found to result from that treatment.

Test Conditions--alpha alloy (6% Al). Tested at 600° and 1000°F.

1. As forged.
2. Cold Worked about 17% and annealed to fine grained structure.
3. Cold worked about 17% and annealed to coarse grained structure.
4. Water quenched from 2025°F (middle of β range).

5. Cold worked 10%.

6. Cold worked 17%.

4. Meta-stable beta alloy (10% Mo). As yet, the material has not been received. Attempts were made to obtain more information regarding the characteristics of this alloy. Pending more complete information, only a tentative program involving direct quenching to retain β, followed by reheating for transformation, and direct transforming by isothermal means is contemplated. It is expected that considerable experimentation to determine response to heat treatment will be necessary before samples for creep testing can be established. However, the delay in obtaining this material has since worked to advantage, for, in the interim, data have been published (3) covering the effect of structure on mechanical properties on binary titanium-molybdenum alloys in the composition range up to 11% Mo.

Planned Test Conditions--meta-stable beta alloy (10% Mo). Tested at 600° and 1000°F.

1. Water quenched from β range to retain β.

2. Quenched from β range and aged to transform β.

3. Quenched from β range to full isothermal transformation at 1100° and 1300°F.

4. Quenched from two temperatures in the $\alpha + \beta$ range in order to obtain the maximum difference in $\alpha + \beta$ properties.

The above planned procedure may yet be modified.

5. Beta alloy (30% Mo). In theory, it was felt that this alloy should respond to heat treatment. Likewise, it was also thought to be amenable to considerable variation in creep resistance by cold work. Actually, however, it was found that neither heat treatment nor cold work afforded significant control of properties.

Test Conditions--beta alloy (30% Mo). Tested at 600° and 1000°F.

1. Testing original retained β structure.

2. Quenching from 1500°F and aging β structure at 1325°F for possible transformation.

3. Quenching from 1325°F + cold work.

4. Long time isothermal holding after cooling from all β region.

EQUIPMENT

Heat treatments were carried out in a stainless steel muffle furnace under a constant static pressure of argon. The argon was first purified by passing it through heated titanium chips.

Cold working treatments were performed on a rolling mill equipped with both closed and open passes.

Tensile tests were performed with a Baldwin-Southward tensile machine having a head speed of 0.050 inches per inch per minute. Strain measurement was accomplished with a modified Martens optical extensometer system.

Creep and rupture tests were carried out in single sample creep-rupture units suitable for high precision testing. Strain in creep testing was measured using a modified Martens optical extensometer system; in rupture testing (much faster strain rates precluding resetting), a "drop of the beam" method was employed. In both cases, loading data were taken utilizing the Martens optical system.

All standard metallographic equipment was available for metallurgical examination. In addition, an electropolisher was used for specimen preparation. Hardness measurements were made using a Tukon microhardness tester.

X-ray examination was carried out using a high-angle geiger tube spectrometer equipped with a recorder.

EXPERIMENTAL TECHNIQUES

Tensile and Creep-Rupture Testing

All testing was carried out in accordance with ASTM Recommended Practices. Standard 0.250-inch diameter specimens were used for all tensile, creep and rupture tests. The specimens were 2-3/4 inches long and had a reduced gage section of 1 inch. The ends were threaded with 3/8-16 threads,

Extensometer bars were attached to pinned collars threaded to the specimen shoulder. A correction factor was used in computing the effective gage length to account for the included fillets.

Temperature control on elevated temperature tests was accomplished with three chromel-alumel thermocouples attached to the gage section. The specimen was placed in the furnace at about 50°F below the test temperature and then brought up to temperature. Temperature distribution within $\pm$ 3°F was attained over the gage section before tensile testing. Samples in rupture and creep tests were similarly brought up to temperature and distribution before the load was applied. This was usually accomplished in less than two hours.

Strain readings were taken daily on long time tests and only after temperature had been checked to $\pm$ 3°F of the nominal test temperature. Two to four readings a day were taken on rupture tests.

Rupture times were accurately obtained from individual electric timers actuated by the drop of the beam as the specimen failed.

Readings during tensile testing were taken at applied load intervals of 4,000 psi for elevated temperature tests and 8,000 psi for room temperature tests. Yield strengths were obtained as the 0.2% offset strength.

Metallographic Technique

Sections for metallographic examination were cut using a Krystolon wheel and continuously recirculating coolant. Following a rough grind, the specimens were taken unmounted through 240-, 400-, and 600-mesh wet silicon carbide paper. Finish polishing was accomplished, in the case of mechanical polishing, by taking the specimen through 4-8 micron diamond compound on a glass-backed photographic print paper, and then 4-8 micron and 0-1/2 micron diamond compounds on suitable rotating laps. However, most specimens were electropolished using a Buehler electropolisher modified so that actual unmounted creep and rupture specimens could be polished. Electropolishing was carried out on specimens taken through 600-mesh wet silicon carbide paper. The electrolyte used was 10% (volume) perchloric acid in glacial acetic acid. Polishing was most successful for time periods up to 10 seconds at a current density of about 2.5 amperes per square cm.

The etchants employed were: for Ti 75A and the 6% Al alloy, a mixture of 2 parts HF, 2 parts HNO_3, and 100 parts H_2O; for Ti 150A and the 30% Mo alloy, a mixture of 1 part HF and 1 part glycerine.

Standard techniques were employed for photomicrography.

X-Ray Techniques

Specimens for x-ray examination were mechanically polished and then etched with HF to remove disturbed surface layers. Usually 0.010 - 0.020 inches of metal were removed. Copper radiation at 45 kilovolts and 12 milli-amperes was used. Two 0.00035-inch nickel filters were used to screen out extraneous background radiation. Soller slits were employed, the combination being 4°-0.006 in.-4°. Diffraction traverses were made at a speed of 2° (2θ) per minute and recorded at maximum amplitude and low damping. The traverses covered the range from 30° to 140° (2θ).

PROPERTIES OF TI 75A

Tensile, rupture and creep tests were run at 600°, 800°, and 1000°F on the seven conditions of Ti 75A listed on page 5. Composition of the heat tested will be found in Table 1.

Tensile Properties of Ti 75A

Short time tensile tests were run at 75°, 600°, 800°, and 1000°F on the structural conditions of Ti 75A previously described (see page 5). Complete results of these tests are presented in Table 2. Figures 1 and 2 show tensile strength and elongation as functions of temperature. In addition, the effect of cold working on tensile strength at each test temperature is shown in Figure 3, a cross-plot of Figure 1.

The major points shown by the data are:

1. Raising the test temperature decreased tensile and yield strengths and the relative elastic modulus, and increased the elongation and reduction of area. Elongation and reduction of area were very high at 1000°F. Ductility characteristics of the as-received and annealed materials at 800°F were exceptions.

2. For a given test temperature, the effect of increasing prior cold working from 10% to 33% was, in general, to raise the strength properties and lower the ductility properties.

3. The increment of tensile strength added by each increment of cold work was diminished, the higher was the test temperature (Figure 3).

4. Tensile strength data at 600° and 800°F reflected a change in the slope of the strength-temperature curves. Such "strengthening" is typical of a strain aging type reaction, the occurrence of which in titanium has been confirmed by several investigations (4, 5).

5. Little difference was evident between the properties of the samples given either of the two prior annealing treatments. Differences, if any, due to the 1500° and 1700°F treatments were inconsistent.

6. The elastic modulus was computed using only one stress-strain curve at each temperature for each condition. Due to the uncertainties present in this type of modulus determination, the values shown are considered to indicate the general effects of temperature on the modulus; they do not necessarily reflect effects of prior treatment at a given temperature.

Rupture Properties of Ti 75A

Rupture tests aimed at establishing the 100-hour rupture strength at 600°, 800°, and 1000°F were run on the seven conditions of Ti 75A. The results of these tests are presented in Table 3. The short time tensile strengths were arbitrarily used as the stress for rupture in 0.1 hours.

Because the rupture testing was limited to as few tests as possible to give a reasonably good determination of the rupture strength for 100 hours, it was not possible to establish stress-rupture time curves for each treatment with reliability. For the limited number of tests made, it was found that there was very little difference in stress-rupture time relationships for materials annealed at 1500° or 1700°F. For this reason, these data have been presented graphically in Figures 4, 5, 6 and 7 without separation on the basis of heat treating temperature. This is valid for 100-hour strengths, although more complete data might show individual curves deviating from perfect agreement at shorter or longer time periods.

For the sake of clarity, Figure 7 summarizes the rupture behavior without showing the actual test points. These latter are shown in Figures 4, 5 and 6.

1. Cold working increased rupture strength substantially at 600° and 800°F, but not at 1000°F. This is shown best by Figure 3.

2. Stress-rupture time curves for material not cold worked were practically horizontal at 600°F. In fact, a testing stress only 1000 psi below the tensile strength increased rupture time by more than 1000 hours. Prior cold working slightly increased the slopes of the curves.

3. At 800°F, there was a marked dependence of rupture time on stress. Also the data for the annealed conditions indicated an increase in slope at about 60 hours. Apparently this slope would be nearly the same as for the curve at 1000°F. It is a general fact in high temperature metallurgy that when there are marked changes in the slopes of stress-rupture time curves with increasing temperature, a change in slope at the lower temperature to one

approaching the short time slope at the higher temperature usually occurs Thus, it would be expected that longer time tests at 600°F would show a change in slope to that of the shorter times at 800°F. Likewise, longer tests would probably show similar breaks in the curves for the cold worked conditions.

4. At 1000°F, the stress dependency of rupture time was much greater than at 800°F. The superiority of the material cold worked 31.3% was lost after rather short time periods.

5. While the relative positions of the stress-rupture time curves for annealed and cold worked conditions remained nearly constant between 600° and 1000°F, the increment of increase in strength decreased with temperature. This is shown by the summarized rupture strengths for 100 hours in Table 3 and by Figure 3. The increase in strength at 1000°F was slight for 10.9% cold reduction and the strength was slightly lowered by 31.3% cold reduction.

6. The ratio,

$$\frac{\text{100-hour rupture strength}}{\text{ultimate tensile strength}}$$

was nearly a straight line function of temperature between 600° and 1000°F, as is shown by Figure 8. The ratio was slightly decreased at a given temperature by cold work. This indicates that temperature had the same relative effect on both tensile and short time rupture strength for different treatments. Usually this is not the case for rupture strengths. For this reason, its use in predicting rupture strengths is probably limited to this material.

7. Figure 15 shows typical fractures and elongations obtained in rupture tests of Ti 75A at 600, 800, and 1000°F.

Creep Properties of Ti 75A

The creep properties were to be surveyed by testing under one stress expected to yield a minimum creep rate of 10^{-5} to 10^{-6} inches per inch per hour. The results of this test, together with the creep data obtained from the rupture tests, are given in Table 4 and plotted in Figure 9. The data are consequently rather limited, and the curves of Figure 9 show only general relationships.

The following generalities regarding creep resistance of Ti 75A, as measured by secondary creep rates, seem to be warranted by the data:

1. There was a very pronounced decrease in creep resistance between 600° and 800°F. Estimated stresses for a rate of 5×10^{-6} inches per inch per hour decreased from values above 36,000 psi to 5,500 to 11,000 psi. A further reduction occurred between 800° and 1000°F, so that the creep strengths at the latter temperature were very low, ranging from 100 to 3,900 psi.

2. There was little difference between material in the as-received, annealed 1500°F, and annealed 1700°F conditions.

3. Cold working increased creep resistance with a larger proportional increase resulting from 10.9% reduction than from 31.3%. At 1000°F, the ma-

terial reduced 31.3% was inferior to the annealed conditions, except at very high stresses.

4. At 600°F, considerable difficulty was encountered in attempting to establish the minimum creep rates. Although considerable primary creep occurred at stresses below the 100-hour rupture strength, little, if any, second stage creep was observed, and the minimum rates were very low (in some cases less than 10^{-8} inches per inch per hour). The slightly greater stress dependency of creep rate for cold worked materials resulted in obtaining somewhat more data than for the annealed conditions.

Metallographic Examination of Ti 75A

Metallographic examination was made of all conditions of Ti 75A prior to testing. In addition, several tensile and rupture specimens were studied after testing.

Considerable difficulty was encountered in both mechanical and electrolytic polishing of Ti 75A. With both methods, there was a pronounced tendency for pitting to occur and in mechanical polishing it was difficult to remove scratches.

However, a number of observations were possible:

1. The structure as-received (hot rolled and annealed) consists of somewhat variable sized equi-axed grains(Figure 10). Some beta phase is believed to be present in the grain boundaries.

2. Upon annealing (1 hour at temperature followed by furnace cooling over a 16 hour period), a slight coarsening effect with increased temperature was noted. Tabulation of average grain diameter is as follows:

As Received	0.0145 mm (Figure 10)
1500°F Anneal	0.0168 mm (Figure 11)
1700°F Anneal	0.0178 mm (Figure 12)

3. No noticeable change in microstructure occurred as a result of cold working other than the usual elongation of grains. The apparent decrease in grain size was calculated and found to correspond to the percentage reduction by cold work (Figure 13). The axis of the rolled bar is normal to the page in Figures 10 through 14.

4. Little change in microstructure, if any, was noted as an effect of test conditions, i.e., stress, time, and temperature. Figure 14 shows a sample originally annealed from 1700°F and then tested at 1000°F and 6,500 psi (27.8 hours). When compared with Figure 12, the same structure previous to testing, little difference is noticeable.

5. The appearance of material in the α grain boundaries of all the microstructures was not unexpected. Reference to the chemical composition (Table 1) shows that this heat of Ti 75A was slightly high in iron (0.19%). The titan-

ium-iron phase diagram (6) indicates the maximum limit of the α field to be less than 0.2%. Thus, it is entirely conceivable that some β phase is present in this heat. This contention is well supported by recent work (7) which shows the effect of small known iron additions to a commercially pure base material. Increased amount of β as islands and in the grain boundaries occurs as the iron content is raised to 0.55%. Presumably, then the grain boundary material is predominantly beta. The amount of beta present in this particular heat of the materials is, however, very small since no β lines were found on x-ray diffraction traverses (see next section).

X-Ray Studies of Ti 75A

Several x-ray diffraction traverses were made on representative conditions of Ti 75A before or after testing. The studies were confined merely to identification of diffraction lines.

The lines obtained conformed to the calculated positions for α titanium. For conditions of cold work, only the most intense lines could be observed. This is consistent with observations in another investigation (4) which showed broadening of lines and reduction of intensity as an effect of cold work. No specific lines broadness studies were made in the present investigation.

As mentioned in the preceding section, there was metallographic and phase diagram evidence to support the existence of some β phase in this material. No evidence was found to confirm this through x-ray studies; however, the spectrometer does not usually pick up lines of any phase present in less than 5% by volume.

PROPERTIES OF TI 150A

Tensile, rupture and creep properties were determined for Ti 150A at 600°, 800°, and 1000°F. The 12 structural conditions chosen for study are listed and discussed on page 6. The composition of this alloy is given in Table 1.

The conditions of Ti 150A selected for study represented conditions of heat treatment either within or brought to completion within or below the $\alpha + \beta$ phase region. Results from another investigation (8) indicates that direct cooling of strip specimens from the β region resulted in structures having either excessive brittleness or no significant improvement in properties. Consequently, the present investigation represents the study of several approaches to the variation of the relative amounts and forms of α and β phases. In addition, check tests were performed on two samples--one water quenched, the other furnace cooled from the all β region.

Tensile Properties of Ti 150A

Complete tensile data at 75°, 600°, 800°, and 1000°F for the 12 conditions selected for study are presented in Table 5 and plots of tensile strength versus temperature and elongation versus temperature are presented in Figures 16 and 17 respectively.

The following observations have been made from these data:

1. Figure 16, a plot of tensile strength versus test temperature, shows the existence of a steady decline in strength to 800°F and then a sharp decline between 800° and 1000°F. Figure 17 shows a moderate increase in elongation up to 800°F and a very large increase at 1000°F.

2. The structural conditions tended to group themsleves into three general groups as far as strength properties are concerned. Thus (referring to Figure 16), the annealed conditions and the 1500°F water quenched condition are the lower and upper strength levels to nearly 1000°F. All other treatments form a separate group in the range of moderate to fairly high strengths.

3. The effect of the various treatment temperatures and cooling rates on the tensile strength is graphically shown in Figure 18. Thus, water quenching, air cooling, and furnace cooling, in that order, gave the highest to lowest strengths at each test temperature. This effect was magnified as the heat treating temperature was increased and diminished as the test temperature was raised. Thus, at a test temperature of 1000°F there was little difference between the three cooling rates from 1500°F; however, at 600°F test temperature the water quenched condition gave about twice the tensile strength of the annealed (furnace cooled) condition. There was a considerable overlap of strength levels at the 600° and 800°F test temperatures for the several conditions of cooling. At 75° and 600°F, the faster the cooling rate and the higher the heat treating temperature, the lower was the elongation. There was little difference at 800° and 1000°F, however.

4. Two conditions of treatment from the all β range were to be tensile tested at 75°F. An 1800°F water quenched sample proved to be too brittle for machining to a test specimen by ordinary procedures. An anneal (furnace cool) from 1800°F resulted in the lowest strength at 75°F of any treatment tested (Table 5).

5. Three double treatments consisting of an original rapid cool from the $\alpha + \beta$ region, followed by reheating either in or below the $\alpha + \beta$ region, were studied. The treatments were: 1500°F water quench + 1350°F annealed, 1500°F air cooled + 900°F annealed, 1350°F water quench + 900°F annealed. With the exception of being slightly more brittle at 75°F, the 1500°F water quench + 1350°F anneal gave properties identical to a 1350°F anneal alone. However, the annealing at 900°F of material air cooled from 1500°F produced properties almost identical to 1500°F air cool alone, the only difference being a slight increase in yield strength at elevated temperatures for the doubly treated material. The 1350°F water quench + 900°F annealed material showed

increased tensile strength (at sacrifice of ductility) at 75°F and 600°F over the 1350°F water quench, together with a tendency to show less elongation at comparable tensile strengths at higher temperatures. The foregoing results indicate that reheating to the $\alpha + \beta$ region will reduce original high strength properties to those associated with the reheat temperature, while reheating 1 hour below the $\alpha + \beta$ region results in little or no change in the original properties, save some embrittlement.

6. Isothermal heat treatments resulted in intermediate strengths at lower test temperatures, but at 1000°F produced the two highest strengths obtained from Ti 150A--59,000 psi for the 1000°F transformation and 48,500 psi for the 1300°F transformation. This increase over the other treatments may be accounted for by a difference in the transformation products, i.e., shape and distribution.

7. Some evidence of a strain aging type behavior was noted during tensile testing. These observations included definite yield points and discontinuous yielding.

Summarizing, the optimum combination of strength and ductility up to a brittle condition in Ti 150A is produced by direct, accelerated cooling from the $\alpha + \beta$ region. The higher the solution temperature and the more drastic the cooling treatment, the higher will be the strength and the lower the ductility at 75°F and 600°F. At 1000°F this generalization is not strictly true, since differences in strength between treatments become small and in some cases negligible. Isothermal transformation from the β range showed useful properties, but not substantially better than direct transformation. All treatments showed very high ductility at 1000°F.

Rupture Properties of Ti 150A

The rupture data are tabulated in Table 6 and plotted on Figure 19 and 20. The main objective of the rupture tests was to survey the influence of treatment on rupture properties by establishing the 100-hour rupture strengths with the minimum number of tests. Consequently, the rupture data for any one treatment are very limited. The data obtained led to the following generalizations:

1. Over the rupture times covered, the data grouped into three strength levels at 600° and 800°F and into one strength level at 1000°F. At 600° and 800°F, the lowest strength group encompassed the as-received material and those treatments which involved a final anneal in the $\alpha + \beta$ region. The upper strength level was the 1500°F water quench. The middle group included all other treatments. At 1000°F, the effect of the different heat treatments essentially disappeared.

In view of this finding, the data have been broken down to the average for each group as a convenient way to present the results from the limited tests now available. It should be recognized that shorter or longer time tests might disclose significant differences in the slopes of the stress-rupture time curves for the individual treatments. This procedure, however, appears reasonably valid for establishing 100-hour strengths. Furthermore, the

best estimates of exact 100-hour strength for each treatment have been included in Table 6.

2. As was noted in the case of Ti 75A, the rupture curves at 600°F for Ti 150A were horizontal. Small decreases in stress may extend rupture time by over 1000 hours. (See Table 6.) However, appreciable deformation by creep did occur at this temperature.

3. At 800° and 1000°F the rupture time became increasingly dependent on stress. No tendency exists in the 800°F data to indicate the possibility of a break in the rupture curve as was noted in the case of Ti 75A.

4. From individual log-log stress-rupture curves for each condition at each temperature, the stress for rupture in 100 hours was determined. These values are given in Table 10 and plotted in bar graph form in Figure 21. A dashed horizontal line indicates the 100-hour rupture strength for the "grouped" conditions discussed previously. This shows how well the groupings reflect the detailed rupture strengths.

5. In Figure 22 these grouped 100-hour rupture strengths are plotted as a function of temperature. This plot illustrates how the difference in strength level between groups became less as test temperature increased. Similarly plotted in Figure 22 are the averaged tensile strengths for each group at each temperature. This illustrates the changing relation between tensile strength and 100-hour rupture as temperature was increased.

6. The significance of the above mentioned change is graphically brought out in Figure 23. Here the quotient,

$$\frac{\text{100-hour rupture strength}}{\text{ultimate tensile strength}},$$

was computed for each group and plotted as a function of temperature and grouping. As noted before in the discussion of Ti 75A, the linear decrease with increasing temperature holds over the range from 600° to 1000°F. At a given temperature, the less drastic cooling treatments possess the slightly higher quotient. These curves illustrate how the relationship between tensile strength and rupture strength was dependent on temperature and prior treatment.

Creep Properties of Ti 150A

The creep data taken in tests on Ti 150A are tabulated in Table 7 and plotted on Figure 24. Although incomplete and sparse for any one treatment, the data tend to confirm the behavior pattern of Ti 150A at elevated temperature portrayed by the rupture test results, i.e., at 600° and 800°F, the treatments were separable into three groups, depending on cooling rate and/or heat treating temperature. At 1000°F, little difference in properties existed between prior treatments.

1. At 600°F the data fall into three distinct groupings. The points for 1500°F water quench are well above the other conditions. (For purposes of

showing the data on as large a set of coordinates as possible, the 1500°F water quenched points have been displaced vertically by 20,000 psi.) The points for the as-received material and those conditions of annealing conducted in the $\alpha + \beta$ region, the limits of which are approximately 1050° to 1575°F, fall into the lowest grouping. In the middle fall the points for all other heat treatments. However, considerable scatter of points was obtained for the lower two groupings. It was thus difficult to draw the stress-creep rate curves, and those presented must be regarded as very tentative.

2. Comparison of 600°F creep data taken in this investigation was made with the results of creep tests of similarly treated strip specimens (8). The highest stress at 600°F that creep tests were successfully performed on strip specimens were below the lower test stresses studied in this investigation, i.e., no overlap of stresses. The reason for this was the high notch sensitivity and contamination of the strip material (almost 0.5% oxygen) which resulted in premature failure at higher test stresses.

The bar stock specimens used in this study were not subject to these defects. Therefore, direct comparison was not possible. It does, however, appear, from studying lower temperature data, that Ti 150A showed a break in the creep curve at stresses about 80,000 psi. Extrapolation of curves for strip material tested at 600°F up to the level of the creep curves presented in this report shows that such a break may be the case. The break would occur at a creep rate of about 10^{-5} inches per inch per hour and would, therefore, agree well with both sets of data. In order to confirm this, it would be necessary to run low stress creep tests at 600°F on bar stock.

3. The stress to produce a second stage creep rate of 5×10^{-6} inches per inch per hour was estimated and is tabulated in Table 7. Reference to this table and Figure 24 shows that a considerable drop in the creep strength of Ti 150A was encountered in raising the temperature from 600° to 800°F. In addition, the three group distinction became much less distinct at 800°F since the 1500°F water quenched material showed minimum creep rates close to the other heat treatments. There is, however, a distinct difference between the properties of annealed material and that given an accelerated cooling treatment and not reheated in the $\alpha + \beta$ phase region.

4. At 1000°F the level of creep strength was very low and there was very little difference in properties between the various heat treatments. The dependency of creep rate on stress was very high and very slight changes in stress could raise the creep rate by factors of ten to 100.

Relation of Microstructure of Ti 150A to the 600° to 1000°F Mechanical Properties

The conditions of heat treatment studied were selected to represent several approaches to the variation of the relative amounts of α and β phase and their form of occurrence. Similar work at lower temperatures (8) indicated that significant differences in creep behavior could be obtained by such variation in heat treatment. In the present investigation, however, it was discovered that higher temperature behavior (above 600°F) could be categorized in three groupings: annealed, 1500°F water quench, and all other treatments.

At 1000°F, even these groupings broke down so that there was equivalencey of creep and rupture properties for all prior heat treatments. Metallographic studies were made on samples of heat treated Ti 150A both before and after testing.

1. As produced, Ti 150A was hot rolled in the $\alpha + \beta$ region, quenched, and annealed in the $\alpha + \beta$ region. The resulting structure (as received) is shown in Figure 25 and consists of small, partially elongated α grains in a β matrix. Although the equilibrium diagram for the Ti-Cr-Fe system indicates the formation of a Ti Cr_2 + Ti Fe + αTi ternary eutectoid, the approach to equilibrium was so slow that, for all practical purposes, there were only α and β phases present. At the testing conditions, the non-equilibrium β broke down. This will be discussed later.

2. Major microstructural variations of Ti 150A studied in this investigation are shown on page 85. The photomicrographs, Figures 26 through 33, are of the as treated conditions. These structures represent several approaches to phase variation. Two heat treating temperatures in the $\alpha + \beta$ range were employed, 1350° and 1500°F, and three cooling rates--furnace cooling, air cooling and water quenching--were used from each temperature. Figures 26, 27, and 28 are of the 1350°F treatments, and Figures 29, 30, and 31 are of 1500°F treatments.

Little noticeable microstructural variation over the as-received condition was found for treatments at 1350°F. A slightly increased amount of β phase was noticeable as cooling rate became more drastic.

Treatments at 1500°F, on the other hand, produced striking changes. The annealed structure, Figure 29, showed greatly enlarged α grains. Figure 30 shows that after air cooling from 1500°F the structure consists of enlarged α grains (fewer than in the annealed condition) in a slightly transformed β matrix. Figure 31 shows that water quenching from 1500°F produced a structure consisting of α, undissolved at 1500°F, in a retained β matrix. For temperatures up to 800°F, this structure represented the strongest material tested in this investigation. The relative amounts of α and β produced from a 1500°F water quench compare quite well with those predicted from the phase diagram for this material.

Taken together, Figures 29 and 31 represent the practical extremes of α - β ratio variation possible with Ti 150A. Lineal analysis shows the 1500°F annealed material contains 70 - 75% α and the 1500°F water quenched material contains 17 - 20% α. Both these values check quite well with equilibrium α - β proportions (11).

Figures 32 and 33 show the results of a full isothermal transformation after a 1 hour solution treatment well within the all β region, 1800°F. The structures are of interest because of the different transformation products obtained, yet for mechanical testing in the range from 600° to 1000°F, their properties fell well within the "medium strength" group of direct cooling treatments. Both photomicrographs show evidence of the original β grain boundary in existence at 1800°F.

Transformation at 1300°F, Figure 32, resulted in the growth of coarse, oriented α particles generally at right angles to the β grain boundaries in an acicular α matrix. The dark etching product at the boundaries may be Ti Cr_2, although x-rays could not confirm this.

Transformation at 1000°F, Figure 26, which is below the $\alpha + \beta$ region, resulted in the formation of finely dispersed acicular α, generally referred to as α', within the grains. In addition, dark etching areas at the boundaries and occasionally within the grains are most probably an α + Ti Cr_2 eutectoid. The third component in this structure--clear, equi-axed particles at grain boundaries--is believed to be α undissolved even after 1 hour at 1800°F. Since transformation at 1300°F consisted of nucleation and growth of α on existing particles in an acicular α matrix, undissolved particles as such would not be visible in Figure 32. In Figure 33, however, formation of only an acicular transformation product below the $\alpha + \beta$ region gave structural emphasis to the undissolved α.

Temperatures of isothermal transformation were selected so as to result in reasonable mechanical properties. Published data (9) indicated that isothermal transformation temperatures above 1400°F and below 900°F resulted in excessively brittle structures.

The structure obtained by a 1500°F water quench + 1350°F anneal was almost the same as from a 1350°F anneal alone, since the annealing was done in the $\alpha + \beta$ region. In the cases where annealing was below the $\alpha + \beta$ region, 900°F, the structures were little changed from the as cooled structures, 1500°F air cool and 1350°F water quench. No as treated photomicrographs are presented for these conditions.

3. Examination of microstructural changes occurring in heat treated samples of Ti 150A after testing was directed toward obtaining the explanation for the drop in creep, rupture and tensile properties in going to 1000°F and the fact that there was no essential difference between creep and rupture properties at 1000°F for all treatments.

The breakdown of non-equilibrium β and/or eutectoid decomposition reactions occurred in short times at 1000°F. Such effects had been previously noted (8) for the conditions of 1500°F air cooling and 1500°F water quenching after long times under test conditions at 400° and 600°F. In these cases, a dark etching product appearing either as precipitate particles or rods was observed in the β matrix.

In the present investigation, a large number of specimens were examined metallographically after testing at either 600°, 800°, or 1000°F.

For the specimens examined, representing relatively short test periods at 600° or 800°F, there was little change, if any, of microstructure as a result of test conditions. Unfortunately, metallographic specimens are not presently available for long time creep tests at the lower temperatures. Otherwise, this effect could be explored further. Examples of structures unchanged after testing are shown in Figures 34 and 44.

However, even short periods of exposure at 1000°F were sufficient to result in striking structural changes. Figures 35 through 43 bring out these effects.

Figure 35 shows 1500°F annealed material tested 116 hours at 11,000 psi and 1000°F. A dark etching product appeared around the α grains. (Compare with Figure 34.)

Figures 36 and 38 show, respectively, 1500°F air cooled and 1500°F water quenched material after a 1000°F tensile test. In both cases, a dark etching product has appeared in the matrix (β) of the material. (Compare with Figures 30 and 31.) Figures 37 and 39 indicated that these same structures underwent further formation of a product in the matrix during the longer time periods of rupture testing.

Other heat treated conditions underwent similar structural changes during testing at 1000°F. Figure 40 shows 1350°F annealed material; Figure 41, 1350°F air cooled material; and Figure 42, 1350°F water quenched materials. In comparison with Figures 26, 27 and 28, these photomicrographs show appearance of the dark product in the matrix. In addition, Figure 43 shows the 1500°F water quenched +1350°F annealed material after 43 hours at 1000°F. This structure is identical to that in Figure 35, 1350°F annealed, after testing at almost identical conditions. This result emphasizes the similarity in properties of the two treatments. Study of isothermally transformed structures tends to confirm these results, although microstructural evidence was not so clear cut.

In all these cases, the effects of testing at 1000°F were the same. The α grains formed on the original cooling were unaffected. All changes occurred in the β matrix. The product formed was a dark etching one. The possibility was considered that the dark product might be a hydride formed in electro-polishing (10). In order to check this, several samples strongly showing this effect were rough ground and repolished solely by mechanical means. When etched, they showed structures identical to those obtained by electro-polishing. Therefore, the possibility of this effect being due to a hydride was minimized.

The dark product is the result of decomposition of non-equilibrium β. All the β phase observed in this material is technically non-equilibrium since the equilibrium diagram (11) fixes the lower limit of the α+β region at about 1050°F. Below this the equilibrium phases are α, Ti Cr_2, and Ti Fe plus the ternary eutectoid. Since the composition of Ti 150A is on the chromium-rich side (2.7 Cr to 1.3 Fe), it is most likely that the formation of Ti Cr_2 is favored. The appearance of Ti Cr_2 has been tentatively identified by x-rays in the case of long time tests of 600°F of 1500°F water quenched strip material (8). Thus, the process occurring at 1000°F in short time periods is the breaking down of non-equilibrium β to α and Ti Cr_2 (and possibly, Ti Fe).

Following this line of reasoning, it is then reasonable to assume that the β matrix may be mainly responsible for the strength properties of Ti 150A above 600°F. Furthermore, it is the β matrix as affected by cooling rate that governs the strength groupings. Thus, the two extremes of treatment, 1500°F water quench (to produce retained β) and annealing treatments (representing an approach to equilibrium), resulted in the two extremes of strength. Intermediate treatments resulted in a β having intermediate strength. At 1000°F, however, the breakdown of all forms of β to α and compounds resulted in lowering all strengths, creep and rupture to the same level. The effect was not quite as marked in the case of tensile testing, since exposure time to temperature was not as long and the breakdown of β appears to have some time dependency. The result of isothermal transformation was to change the form and distribution of the transformation products from the basic structures obtained in direct cooling treatments, i.e., α grains in a β matrix. However, not much change in properties was obtained, with the exception that the isothermally transformed structures appeared to be more stable at 800° and 1000°F.

4. In an effort to following the breakdown effect mentioned before, hardness data were taken for a large number of completed test specimens. The material as treated showed fairly wide variations in hardness as a result of solution temperature and cooling treatment. Hardnesses are presented in Table 15 and ranged from a Knoop Hardness Number of 271 for 1500°F annealed to 398 for 1500°F water quenched. All as treated hardnesses compared very well with similar studies on strip material (8). As a result of test conditions, it is noted that there is a general tendency for hardness to increase after testing at 600° and 800°F. This could be due to work hardening effects as a result of testing strains, or to precipitation effects acting to harden the β phase.

For testing at 1000°F, however, there was a decrease of hardness over that at 600° and 800°F. This was probably due to the breakdown of β to softer product or stress relief. The hardness after testing at 1000°F ranged from about 320 (Knoop Hardness Number) to about 375. In general, it began to approach the as-treated hardness, and in the case of 1500°F water quenched material tested 348.8 hours at 1000°F, it dropped to 80 points lower than the as-treated hardness.

5. Typical appearance of fractured specimens of Ti 150A is shown in Figure 45. As the photograph shows, there was a tendency for necking to become more pronounced as test temperature was raised. The fractures show a definite cup and cone appearance.

X-Ray Studies of Ti 150A

X-ray diffraction traverses were run on a limited number of representative conditions of Ti 150A after testing. Changes in the α and β phases were believed to occur as a result of test conditions. The results are briefly summarized in Table 8. Because the actual test specimens were used (representing different reductions of area) and there was a consequent difference in sample size, it was impossible to obtain a correlation with diffraction line intensity. Instead, the number of discrete α and β lines observed have been recorded. Therefore, some anomolies exist.

The data suggest, together with the metallographic data, that a breakdown in non-equilibrium β occurred under the influence of stress, time and temperature.

Thus, conditions of accelerated cooling have a fairly high number of β lines. This was especially true for heat treating temperatures well up in the $\alpha + \beta$ region. These lines tended to disappear to diminish under conditions of testing. Since this is a crude measure of the relative intensity, it is also a crude measure of changing phase proportions. This result is only reasonable since the material should seek equilibrium and is aided by exposure to test conditions. Hence, disappearance of the weaker β lines is to be expected.

Of particular interest is the full isothermal transformation at 1000°F (below the $\alpha + \beta$ region). This result shows that the transformation product consisted of modified α and the breakdown of β to α and $TiCr_2$ may have

been under way. No lines, however, were found to suggest the presence of Ti Cr_2. Perhaps the time period was too short, or the amount present was less than the 5% by volume of a phase which must be present before its lines are detected by the high-angle spectrometer. For 1300°F isothermal treatment, it will be noticed that some β lines were present.

PROPERTIES OF EXPERIMENTAL ALPHA ALLOY: 6% AL - 94% TI

The experimental alpha alloy was prepared by the Armour Research Foundation. Reported composition of the alloy is given in Table 1. Tensile, rupture, and creep tests were run at 600° and 1000°F.

The maximum amount of cold work that the as forged structure could tolerate was found to be 17%. Above this amount, splitting and cracking occurred.

The six structural conditions tested are discussed on pages 6 and 7.

Tensile Properties of Experimental Alpha Alloy - 6% Al

Short time tensile tests at both room and elevated temperatures were run on samples of the 6% Al alloy representing various conditions of cold work and recovery. Complete data are tabulated in Table 9 and tensile strength and elongation are plotted against test temperature in Figure 46. The following observations have been made from the data:

1. Cold working the 6% Al alloy resulted in raising the tensile strength at all temperatures; however, the relative effect of cold working was somewhat different at different temperatures. Figure 47, a cross plot of the data showing the effect of cold work, indicates an almost linear increase in tensile strength at 75° and 600°F with cold work up to the maximum of 17% the material would tolerate. At 1000°F, there appeared to be little improvement in strength above 10% cold work.

2. Elongation data, however, are puzzling. Figure 47 shows that there was little difference in absolute elongation between 75° and 600°F and little change in elongation between 10% and 17% cold work, at these two temperatures. At 1000°F there was a linear decrease in elongation with increased cold work. The anomoly comes from the fact that elongation as a function of prior cold work and test temperature behaved the opposite of tensile strength as a function of the same two variables.

3. The effect of a high temperature anneal on cold worked material was to reduce strength while improving ductility. The annealing temperatures of 1500° to 1700°F, chosen to be certain that recrystallization would occur,

resulted in a lowering of tensile strength to below the as forged condition. The 1700°F treatment produced the lower strength of the two annealing treatments over the whole range of test temperatures.

4. Another treatment was also studied over the range from 600° to 1000°F. This treatment consisted of 1 hour solution treatment of the as forged material in the all β region (2025°F), followed by rapid cooling. In this manner, a different form of α was obtained (see page 26, item 2). Tensile strength was lowered at 75°F, but raised slightly at 600° and 1000°F. As a matter of interest, a sample was also air cooled from 2025°F and, while showing tensile and yield strengths almost identical to the water quenched treatment, elongation and reduction of area were reduced by almost two-thirds.

5. For all conditions it will be noted that the tensile strength at 75°F of the 6% Al alloy is only moderate compared to that attained at 75°F for Ti 150A. However, the drop in strength with increased temperature was much less severe for the 6% Al, and at 600°F the two alloys showed strengths of the same order; while at 1000°F, the 6% Al alloy exhibited strengths in the order of twice those for Ti 150A. In addition, the elastic modulus of the 6% Al alloy appeared to be less affected by temperature than that of either Ti 75A or Ti 150A.

Rupture Properties of Experimental Alpha Alloy - 6% Al

The rupture properties of the 6% Al α alloy were determined at 600° and 1000°F. Quite different behavior existed at each temperature. The data are summarized in Table 10 and plotted on Figure 48.

1. The rupture curves at 600°F for the 6% Al alloy show that there was no decrease in strength with time. Tests at stresses up to 98% of the 600°F tensile strength showed no tendency towards rupture for time periods out to almost 1500 hours.

2. At 1000°F the rupture curves were somewhat more dependent on stress; however, the slope of the curves at 1000°F was much less than those for Ti 75A and Ti 150A at the same temperature.

3. The absolute level of rupture strength at 600°F for the 6% Al alloy was only midway between the lower two groupings of Ti 150A at this temperature. At 1000°F, however, the rupture strength for all conditions of the 6% Al alloy was two to three times that for Ti 150A.

4. The effect of prior cold work on rupture properties is brought out by examination of Figure 47 which is a plot of cold work versus 100-hour rupture strength. The 100-hour rupture strengths were determined from Figure 48. At 600°F it is noted that the rupture strength was very close to the tensile strength and increased with increased cold work. At 1000°F, however, the 100-hour rupture strength was well below the tensile strength and became lower for greater amounts of cold work.

5. The 1500° and 1700°F anneals of material cold worked 17% had the effect of reducing the slope of the 1000°F rupture curve and thus raising

the 100-hour rupture strength. At 600°F these treatments, however, were consistent with the reduction in tensile properties and resulted in lowered rupture curves.

6. The condition representing the highest 100-hour rupture (40,000 psi) strength at 1000°F was the water quench from 2025°F. The 2025°F water quenched samples also showed higher rupture properties at 600°F than other conditions, save 17% cold work.

7. The elongation on rupture for the various conditions of the 6% Al alloy covered a much smaller range of values than it did for either Ti 75A or Ti 150A. The cold worked conditions did show a low value of rupture elongation at 600°F. However, for increased times and temperatures the rupture elongation increased to an upper limit of about 60%.

Creep Properties of Experimental Alpha Alloy - 6% Al

Creep data for the 6% Al α alloy are presented in Table 11 and plotted in Figure 49.

1. The data are limited at 600°F due to the difficulty of selecting stresses that would give measurable creep rates ($>10^{-8}$ inches per inch per hour) and yet not cause rupture prematurely. The creep strength followed the tensile and rupture strengths fairly well. Therefore, the conditions having the highest strengths at 600°F--10% and 17% cold work and 2025°F water quench--similarly showed the highest creep strength. Similarly, the as forged and 17% cold work plus recyrstallization conditions showed the lowest creep strength.

2. At 1000°F, indications are that the conditions having highest 100-hour rupture strength, as forged and 2025°F water quench, had the lower creep strengths. Conversely, cold worked material and material cold worked and recrystallized prior to creep testing appeared to show better creep properties at 1000°F, although their rupture properties were poorer than non-cold worked material.

3. The conditions having the best long time minimum creep rate at 1000°F were those of recrystallization to large grains. From a metallurgical standpoint, this was not unexpected.

The metallurgical aspects of these results are discussed in the next section.

Effect of Testing on Microstructure of Experimental Alpha Alloy - 6% Al

Metallographic examination of the 6% Al alloy revealed several significant changes in microstructure as a result of prior treatment and subsequent test conditions.

1. The as forged structure is shown in Figure 50. The material consisted of a single α phase and was composed of somewhat variable-sized, elongated grains. There is a slight suggestion that the as-cast structure was not completely broken up by forging at 2100°F.

2. A 1 hour solution treatment at 2025°F, followed by an agitated water quench, resulted in the structure shown in Figure 51. The solution temperature of 2025°F was well within the all β region (12). Even upon rapid cooling there was no tendency for any retention of β; therefore, the quenched material had an all α structure. X-ray data confirm this result. The resulting structure was quite different from the original as forged structure. The nature and form of the product may account for the increased strength and hardness. In addition, there was an effect upon cooling as a result of the phase equilibria of this system. For a nominal composition of 6% Al, at the $\beta/\alpha + \beta$ transus line, the equilibrium composition of α contains 8% Al. As the temperature drops through the $\alpha + \beta$ region, the composition of α approached 6% Al. However, this is a nucleation and growth process depending upon the diffusion of Al, and the cooling rate is too fast to permit completion of the diffusion. Thus, the as quenched structure contains areas of Al-enriched α. As Figure 51 indicates, the transformation occurs around the β grain boundaries existent at 2025°F. The clear material at the grain boundary is enriched α. The remainder of the α formed within the grains as crystallographically oriented elongated particles. Microhardness tests indicated the original β grain boundaries of this material to be somewhat harder than the transformed material within the grains. The overall hardness for this condition was greater than for the as forged. (See Tabe 15.)

The existence of "hardspots" within the material helps explain somewhat superior strength properties this treatment exhibited over the as forged condition. An 8% Al alloy has a tensile strength 5,000 to 15,000 psi higher and a DPH hardness 25 to 50 points higher at all temperatures over a 6% Al alloy. These hard spots, thus, acted to impede continuous deformation processes, i.e., creep and rupture.

Test conditions had the effect of breaking up this non-homogeneous structure. Figure 52 shows the structure after a 1000°F tensile test. Here, the structure is less well oriented than the untested material. Holding at an elevated temperature would probably allow continued diffusion and the elimination of the as quenched compositional variations.

3. Increased amounts of cold work resulted in the break up of the as forged structure. Figure 53 shows the effect of 10% cold work; and Figure 59, the effect of 17% cold work. In addition, the material cold worked 17% was recrystallized, using appropriate time-temperature annealing conditions. Holding 2 hours at 1500°F resulted in the structure shown in Figure 60. Medium sized equi-axed α grains were obtained. Figure 61 shows the effect of holding 3 hours at 1700°F. In this case, very large α grains were formed as a result of the increased temperature and time.

4. The effect of certain time-temperature-stress conditions in creep and rupture testing up to 1000°F was to promote recrystallization. Efforts to recrystallize metallographic samples of cold worked materials in the absence of stress were unsuccessful at temperatures up to 1500°F. Hence, the use of high annealing temperatures in order to obtain grain growth. (See preceding section.)

For both 10% and 17% prior cold work, there was little evidence of recrystallization during testing at 600°F. Figures 54 and 55 show the effect of a tensile test and a 1216 hour creep test, respectively on 10% cold worked material. These structures, representing different stress-time combinations, showed little change, if any, over the as cold worked conditions. Similarly, the 17% cold worked material when tested 1172 hours at 600°F (Figure 62) showed little change over the as cold worked condition. (See Figure 59.) Perhaps the very beginning of recrystallization was occurring.

At 1000°F, the effects were much more striking. For 10% cold work, there is evidence (Figure 56) that recrystallization occurred during tensile testing. The effect appears to be somewhat stress and time dependent, since a specimen tested 20.5 hours at 40,000 psi (about half the tensile strength) showed only the beginning of recrystallization (Figure 57). Lengthening the time (176.4 hours) at a lower stress level (26,800 psi) resulted in more complete recrystallization with suggestion of grain growth (Figure 58).

For 17% cold work, when tested at 1000°F, the recrystallization took place much more rapidly. Figure 63 (13.4 hours at 37,000 psi) shows the reaction to be complete. Small equi-axed grains have replaced the elongated forged and cold worked structure (Figure 59).

To summarize:

Condition	Temp (°F)	Type Test, Stress Time	Progress of Recrystallization
10% Cold Work	600	Tensile - 97,800 psi	none
	600	1216 hrs - 96,500 psi	none
	1000	Tensile - 83,300 psi	fairly complete
	1000	20.5 hrs - 40,000 psi	in progress
	1000	176.5 hrs - 26,800 psi	fairly complete, some grain growth
17% Cold Work	600	1172 hrs - 101,000 psi	none
	1000	13.4 hrs - 37,000 psi	complete

The effect of this metallurgical instability at 1000°F showed up very strongly in the case of rupture testing. Previous discussion (see page 24) brought out the fact that increased cold work resulted in poorer rupture properties at 1000°F. Figures 47 and 48 show this graphically.

If recrystallization and grain growth are brought to completion before testing, then rupture properties are somewhat improved. (See Table 10.)

5. Hardness data (see Table 15) confirm the effects mentioned in the preceding section. Increased cold working increased hardness over the as forged or 2025°F water quenched material.

Elevated temperature testing resulted in an immediate decrease in hardness, thus indicating the relaxation of cold working stresses. There is a tendency for material tested at 1000°F to show slightly lower hardness

than that tested at 600°F. This was the case for both 10% and 17% cold work.

6. Typical fractures obtained from rupture tests are shown in Figure 64. At 600°F the fracture appeared to be transgranular. At 1000°F there was, of course, some oxidation darkening the fracture. Examination of the specimen indicated intergranular fracture to have taken place.

X-Ray Studies of Experimental Stable Alpha Alloy - 6% Al

Several x-ray diffraction traverses were run on representative conditions of the 6% Al alloy before or after testing.

The lines obtained were α phase lines and showed good agreement with their calculated positions. There was no evidence, x-ray, metallographic or otherwise, that indicated any β phase in this material.

PROPERTIES OF EXPERIMENTAL BETA ALLOY: 30% MO - 70% TI

The four structural conditions of the β alloy that were studied are listed and discussed on page 7. The material was prepared by the Armour Research Foundation, and the chemical analysis is given in Table 1.

Tensile, creep, and rupture tests were run on the β alloy at 600° and 1000°F.

Difficulty was encountered in cold working this material. Four bars were solution treated 1 hour at 1325°F. Following this, one attempt was made to cold reduce the bars. Two bars were reduced 3.3%; however, the other bars split on attempts to obtain reductions only slightly greater than 3.5%.

Tensile Properties of Experimental Beta Alloy - 30% Mo

Little variation in tensile properties with heat treatment was expected for the 30% Mo alloy and such was the case. The data are summarized in Table 12, and plots of tensile strength and elongation versus temperature are presented in Figure 65. The following observations are made:

1. Variations of tensile strength were small from treatment to treatment and on the order of 5,000 psi at any test temperature.

Elongation and reduction of area properties were moderately good and show little variation.

2. There was little, if any, effect of raising the solution temperature from 1325° to 1500°F and changing solution time from 1 hour to 2 hours. The 24 hour anneal at 1375°F, following a 1500°F solution treatment, resulted in a slight drop in strength and a loss of ductility.

3. The small amount of cold work (3.3%) that the material would tolerate resulted in a slight increase in strength at 600° and 1000°F and a slight decrease in ductility.

4. As in the case of the 6% Al alloy, the tensile strength of the 30% Mo alloy was only moderate at 75°F, but showed less decrease with increasing temperature than did the tensile strengths of Ti 75A and Ti 150A. This is also the case for the modulus of elasticity. The level of strength for the 30% Mo alloy was about the same as that for the 6% Al at the three test temperatures.

In summary, it appears that the 30% Mo alloy is relatively insensitive to heat treatment and not amenable to cold work.

Rupture Properties of Experimental Beta Alloy - 30% Mo

The results of rupture tests at 600° and 1000°F on the 30% Mo alloy are presented in Table 13. The relation of stress to rupture time is plotted on the conventional log-log plot in Figure 66.

1. Reference to Figure 66 shows rupture curves similar to those for the other titanium alloys. At 600°F, the relationship is very flat. Tests run at stresses from 95 to 98% of the 600°F tensile strength showed no tendency or inclination to rupture in time periods out to almost 1500 hours. Thus, it appears that at 600°F loads to within 5,000 psi of the tensile strength may be tolerated indefinitely. Time-elongation data shows no further deformation following a short period of first-stage creep. This will be discussed later.

2. At 1000°F the rupture curves became considerably more stress dependent. It will be remembered that the effect of prior treatment on tensile properties at 1000°F was very slight. The results of rupture tests at 1000°F suggest that even these small differences between treatments were eliminated, at least for time periods near 100 hours. Some differences might be found for shorter or longer time periods if more extensive testing were carried out. One long time point (1143.9 hours) was available because of a creep test running to rupture. This point helped confirm the impression that a slight break occurred in the curves at some time less than 100 hours.

3. Elongation and reduction of area values were high at 1000°F and increased with time for fracture.

Consequently, for heat treatments tested to date, no essential difference in rupture properties has been found.

Creep Properties of Experimental Beta Alloy - 30% Mo

Creep data for the 30% Mo alloy are presented in Table 14 and plotted in Figure 67. These results are rather sparse, but do suggest that creep rates of this material are somewhat less stress dependent than for alloys tested previously.

1. The stress-creep rate curve at 600°F is almost horizontal and was constructed on the basis of the following reasoning: The tensile strength of this material at 600°F is approximately 106,000 psi and does not seem to be too affected by heat treatment. Only one creep test, at 102,500 psi, resulted in a readily measurable creep rate. Tests at 100,000 and 90,000 psi showed very low minimum creep rates, rates that could only be estimated as less than 10^{-8} inches per inch per hour.

2. At 1000°F the rate was somewhat less stress dependent, and the data are more numerous. As in the case of tensile and rupture data, there was little difference in properties between heat treatments, and so one line has been used to correlate the data. The extrapolated stress for a creep rate of 5×10^{-6} inches per inch per hour was found to be about 12,000 psi.

3. At 1000°F the estimated stress for a creep rate of 5×10^{-6} inches per inch per hour was high and indicates a higher ratio of creep to rupture strength than has been obtained for other alloys tested. This may be of importance because creep strength in other alloys has appeared to be low in relation to rupture strength. This would imply that excessive deformation by creep at stresses well below the rupture strength would severely limit load carrying ability in these alloys. In the 30% Mo alloy, this condition seemed somewhat improved.

4. However, lest the observance of low minimum creep rates be misinterpreted, it should be noted that at both temperatures, 600° and 1000°F, substantial loading and first stage deformation were observed, although less than those obtained with the other alloys.

Metallographic Examination of Experimental Beta Alloy - 30% Mo

Metallographic examination of the 30% Mo alloy was made on specimens both before and after testing in order to ascertain what changes, if any, occurred as a result of test conditions.

These studies indicate that, although the alloy is essentially all β material, it did break down, and a precipitate formed under certain test conditions. Presence, at least, of α phase has been identified under these conditions.

1. The as forged material, Figure 68, consists of large equi-axed β grains. The "speckled" appearance within the grains is believed to be an

etching effect. The spots may, however, be residual α, although no confirmation could be made of this.

2. Heating the material to 1325°F and water quenching resulted in no significant change in microstructure (and tensile and rupture properties). The speckles (Figures 69) have disappeared and a few line markings remain within the grains. Again, this is probably an etching effect.

3. Raising the solution temperature to 1500°F, holding 2 hours, and then water quenching had no effect on microstructure (Figure 74).

4. No effect on microstructure was found for tests at 600°F and long time periods.

5. At 1000°F test temperature there was very conclusive evidence of precipitation reaction occurring in time periods of the order of 100 hours.

Figure 70 shows the as forged structure after testing for 77 hours at 1000°F and 40,000 psi. No essential change could be noted over the untested structure (Figure 68). However, the same condition after 133 hours at 1000°F and 35,000 psi (Figure 71) showed a fine, dark precipitate within the β grains.

The heat treated conditions also displayed this behavior. Figure 72 showed a 1325°F water quench after 836 hours at 600°F and 100,000 psi. Slight precipitation, if any, is visible. However, Figure 73 showed the same condition after 1412 hours at 1000°F and 35,000 psi. Here precipitation was visible, although finer and more widely dispersed than the precipitate occurring in the as forged condition (Figure 71).

Figure 74 shows the microstructure of the 1500°F water quenched material after tensile testing at 1000°F. No evidence of precipitation appeared. However, Figure 75 shows the effect of 83.4 hours at 1000°F and 37,500 psi. Again precipitation occurred, this time as lines of dispersed particles along crystallographic planes.

The nature of the precipitate is not clear; however, evidence exists from x-ray data (see next section) to suggest α phase as a component. The formation of the precipitate has thus far been found only at 1000°F after time periods of about 100 hours. In addition, the stress condition (35,000 to 45,000 psi) may have some effect.

6. Hardness data (Table 15) suggest that a slight decrease in as treated hardness may be experienced as a result of test conditions. No effect on hardness was detectable as a result of precipitation occurring at 1000°F.

7. Typical appearance of rupture specimens is shown in Figure 76. At 600°F there was slight elongation and failure occurred through necking down. The fractured surface had a crystalline appearance. At 1000°F there was considerable elongation; however, the specimen did not neck as much as the one tested at 600°F. The fracture was jagged and evidence was visible of cracking all along the gage section, as though failure were occurring simultaneously at many points. These observations suggest that fracture at 600°F was transgranular and at 1000°F was intergranular.

8. The rupture data indicate that the effect of the metallurgical instability (see page 31, item 5) at 1000°F is slight. However, Figure 66 shows that a slight break in the curve did occur at about 100 hours and this weakening effect may be due to the precipitation reaction.

X-Ray Examination of Experimental Beta Alloy - 30% Mo

X-ray techniques were employed in order to ascertain, if possible, the nature of the precipitate formed in the 30% Mo alloy during rupture tests at 1000°F.

The specimen whose microstructure appears in Figure 73 was subjected to a complete diffraction traverse using the high-angle geiger tube spectrometer. The specimen had ruptured after 141.2 hours at 1000°F and 35,000 psi. The microstructure shows large amounts of precipitate within the β grains.

The results of the diffraction traverse showed positive identification of the α 100 and 102 lines, generally the strongest α lines, together with almost all the possible β lines. In addition, the appearance of the α 002 and 102 lines was also suspected, although they were not strong enough for a positive identification.

The influence of time periods in the order of 100 hours at 1000°F was to break down the original all β structure of this alloy.

It is, therefore, very likely that the precipitate contained at least some α phase, although it may be part of a more complex mixture.

CORRELATION OF STRUCTURAL VARIATION OF TITANIUM ALLOYS WITH PROPERTIES IN THE RANGE FROM 600° TO 1000°F

Of the four materials thus far tested, three showed variation of tensile and creep-rupture properties in the range from 600° to 1000°F as a result of prior treatment, i.e., either heat treatments or cold working. The fourth material, the 30% Mo - 70% Ti beta stabilized alloy, showed little or no response to prior treatment.

The alloys investigated represented structural variability associated with commercially pure titanium, an alpha-beta alloy undergoing eutectoid decomposition reactions, and stable alpha and beta alloys. A meta-stable beta alloy remains to be tested in order to complete the study of structural variations.

Considerable overlap of properties occurred between the alloys at several test temperatures. Comparisons between structures were made on the

basis of two criteria: stress for rupture in 100 hours, and stress resulting in a minimum creep rate of 5×10^{-6} inches per inch per hour. Ranges of variation of these properties with test temperature are presented in Table 16 and plotted in Figures 77 (rupture strength) and 78 (creep strength). In the case of rupture strengths, the tensile strength had been previously shown to be the controlling factor over the range from 75° to 600°F.

Reference to these figures shows that both the rupture and creep strengths of these materials decreased as the test temperature increased. This decrease was very marked above 600°F, although, where data were available, the creep strength appeared to level off somewhat from 800° to 1000°F. In addition, the range of possible strength variation for a given alloy due to prior treatment became less as the test temperature was increased.

Commercially pure titanium (Ti 75A) was appreciably weaker than the alloys at all test temperatures. In the alloys, overlap in strength levels was found in the range from 600° to 800°F. In addition, the two phase alloy, Ti 150A, exhibited capcity for the highest rupture strengths of all the materials up to 750°F. However, above 800°F, the single phase alloys became considerably stronger.

At 600°F all the materials exhibited the capacity to tolerate large amounts of total deformations without seriously affecting the rupture life. The 30% Mo - 70% Ti binary alloy showed the best creep resistance of the materials tested over almost all the range of temperature above 600°F.

All the materials that failed during testing showed transgranular fractures at 600°F. At 1000°F the stable binary alloys showed intergranular fracture, while the commercial alloys continued to show transgranular fractures up to 1000°F.

From a metallurgical standpoint, the principle effect of alloying titanium was to increase its strength properties up to 1000°F. Of the alloys tested, two were essentially single phase; and the third, a two phase alloy. Overlapping of properties among these three materials occurred at 600°F; however, at 1000°F, the single phase materials were superior. A qualitative comparison of the metallurgical aspects of the differences between these materials has been made in Table 17. The synthesis compares for the structural types a number of criteria not conveniently expressable in numerical form. Examination of this tabulation in relation to the discussions on the individual materials brings out the fact that at the higher test temperatures all alloys showed some evidences of matallurgical instability. Thus, at 800°F and particularly at 1000°F, Ti 150A experienced a breakdown of non-equilibrium β to α, Ti Cr_2 and/or a eutectoid. At 1000°F the stable α alloy, 6% Al, exhibited a tendency for recrystallization of cold-worked conditions, while the β alloys, 30% Mo, showed formation of a precipitate containing α after a suitable time period at temperature.

Of the strength variations induced by prior treatment, only Ti 150A showed significant improvement through heat treatment. The other materials, single phase or essentially so, showed significant improvement only by cold working. Yet, at 1000°F, it was evident that an optimum amount of cold work existed, above which the properties were little improved and possibly even lowered. The limited data on the effect of cold working suggested that the optimum amount of cold work might be 10% or less.

In the case of Ti 150A, the improvement in strength at 75°F appeared to be related to the amount and/or form of non-equilibrium β present in the structure. Heat treatments resulting in an improvement of properties in the range from 600° to 1000°F were separated into three groupings. These were: annealing treatments, water quench from high in the α-β region (1500°F), and intermediate solution temperatures and/or cooling rates. At temperatures below 600°F, it has been shown (8) that tensile and creep strength may be a direct, linear function of β content. However, at the higher temperatures studied in this investigation, the effect of exposure to test conditions for even a short time was sufficient to eliminate the relative effects of all but the extremes of heat treatment.

In the case of the single phase alloys, however, the effect of temperature on the metallurgically stable structures (relatively speaking) was much less. Thus, the higher temperature properties of titanium alloys depended not so much on the ability of the material to be strengthened by prior treatment, but rather on the inherently better stability of the single phase structure. Expressed another way, alloys of a meta-stable nature are not satisfactory for service at temperatures sufficiently high enough to "trigger" the material to seek its equilibrium state in the time period considered for application. Thus, although certain structural conditions of Ti 150A showed eutectoid decomposition at 210° and 400°F in another investigation (8), the time of exposure was on the order of 1000 hours. The present investigation showed that above 600°F these reactions took place at shorter and shorter times as test temperature was raised. The properties became poorer and difference between treatments became less at shorter times. Single phase materials, on the other hand, showed time-temperature effects of a more subtle and/or less deleterious nature. Structural stability appears to be the controlling factor for elevated temperature properties.

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APPENDIX

Shape of the Time-Elongation Curve

It is not the purpose of this report to present the actual time-elongation curves from the creep and rupture tests run in this investigation. In the interests of clarity, all time-elongation data are being correlated and reproduced in a supplement to this report. These data will be filed with the Materials Laboratory, Wright Air Development Center.

However, the nature of these curves makes it important that attention be called to their shapes as a function of temperature. Three representative time-elongation curves are plotted in Figure 79. The testing temperatures were 600°, 800° and 1000°F.

One curve shows the data for Ti 150A as-received tested at 600°F and 80,000 psi. Most of the deformation occurred within a relatively short time after loading. After 200 to 300 hours the curve levelled off and the actual minimum creep rate was very small. After about 950 hours the test was discontinued. The permanent deformation of this specimen after cooling was found to be 5.7%. This test was a good example of the type of deformation at 600°F showing little or no effect upon rupture life, although substantial first stage deformation could occur.

An example of the type of deformation obtained at 800°F is shown by the curve for Ti 75A annealed at 1500°F and cold worked 10.9%. The test stress was 14,000 psi. In this case, the curve showed the three classical elements of the creep curve, i.e., a short first stage, a period of steady state creep (second stage), and the gradual occurence of an increasing rate (third stage). This particular test was stopped after 1097 hours and showed a permanent deformation upon cooling of 11.3%.

At 1000°F many of the low stress curves for Ti 75A and Ti 150A were of the nature shown by the 600°F curve in Figure 79. The initial deformation was very small and a steady state rate of about 5×10^{-6} inches per inch per hour was attained after 500 hours. Tests showing this creep rate exhibited permanent deformations on cooling of the order of 1 - 5%.

The binary alloys (6% Al or 30% Mo) exhibited somewhat different creep curves at 1000°F than the commercial alloys. The curve for 1000°F in Figure 79 represents an extreme case. The test was of the 30% Mo alloy at 22,000 psi. After a small initial deformation, a short period of steady state creep was obtained. After only 100 hours, the creep rate continually increased until failure occurred after 1143 hours. At lower stresses these materials exhibited the type of creep curve shown by the 800°F curve in Figure 79, i.e., third stage was not attained until 600 or 700 hours.

The important points brought out by examination of these curves are that substantial first stage deformation could occur at 600°F without seemingly affecting rupture properties. At 800° and 1000°F, tests showing a minimum rate above about 10^{-5} inches/inch/hour had a tendency to go into third stage well before 1000 hours were reached. However, third stage could continue for a considerable length of time (up to 90% of the total test time) before failure occurred.

TABLE I

Test Materials

Designation and Supplier	Chemical Composition (%)						Form and Date of Receipt
		O_2	N_2	C	Fe	Ti	
1. Ti 75A (Commercially Pure Ti) (Titanium Metals Corp.)	Nominal	trace	.02	--	.10	bal.	50 ft. 1/2" round bar stock October 15, 1952
	Actual	--	.061	.025	.19	bal.	
		O_2	N_2	C	Fe	Cr	Ti
2. Ti 150A (Commercial $\alpha + \beta$ alloy (Titanium Metals Corp.)	Nominal	.25	.02	.02	1.3	2.7	bal.	50 ft. 1/2" round bar stock December 10, 1952
	Actual	--	.124	.046	1.52	2.68	bal.	
3. Experimental Alpha Alloy (Armour Res. Foundation)	Nominal	6.0% Al - bal. Ti						20 ft. mixed 1/2" rounds and squares, October 15, 1952
	Actual	6.21% Al - bal. Ti						
4. Experimental Stable Beta Alloy (Armour Res. Foundation)	Nominal	30% Mo - bal. Ti						20 ft. 1/2" rounds January 15, 1953.
	Actual	29.58% Mo - bal. Ti						
5. Experimental Meta-Stable Beta Alloy (Armour Res. Foundation)	Nominal	10% Mo - bal. Ti						Not yet received (September 30, 1953)

TABLE 2

Tensile Tests for Ti 75A

Treatment	Test Temp (°F)	Ultimate Strength (psi)	Yield Strength (0.2% Offset) (psi)	Elongation (% in 1 in.)	Reduction of Area (%)	Elastic Modulus x 10^{-6} psi
As Received	75	90,700	67,500	30.2	50.4	13.8
	600	37,600	19,100	34.9	68.6	11.2
	800	32,700	19,500	22.6	69.5	8.1
	1000	22,000	12,700	72.0	83.4	6.2
1500°F Annealed (1 hr at temp + furnace cool)	75	87,900	68,400	30.4	48.6	12.9
	600	34,900	16,900	41.5	67.0	8.7
	800	28,300	14,700	34.3	70.4	8.4
	1000	17,790	10,550	114.0	92.0	4.5
1700°F Annealed	75	87,600	65,600	34.0	44.6	14.1
	600	33,600	17,100	40.7	69.6	11.7
	800	28,500	14,800	36.6	69.2	5.9
	1000	19,750	11,200	93.4	87.0	5.6
1500°F Annealed + 10.9% Cold Work	75	115,000	102,000	12.7	43.5	14.2
	600	53,300	50,600	14.6	57.5	10.2
	800	41,300	36,600	17.3	67.3	9.0
	1000	23,800	14,700	103.0	85.4	5.5
1500°F Annealed + 31.3% Cold Work	75	129,000	116,000	12.3	39.8	14.0
	600	67,000	63,000	13.3	51.1	10.2
	800	50,800	44,100	21.4	66.7	8.9
	1000	26,800	17,000	70.5	87.4	3.3
1700°F Annealed + 10.9% Cold Work	75	109,000	99,400	10.7	32.5	12.3
	600	53,700	50,700	14.8	55.2	9.8
	800	42,400	37,000	15.6	61.5	9.3
	1000	20,900	14,000	144.5	94.6	4.3
1700°F Annealed + 31.3% Cold Work	75	127,000	114,000	11.5	34.1	14.0
	600	66,500	62,200	14.0	57.5	10.2
	800	48,900	39,800	18.6	66.2	7.8
	1000	23,700	15,000	91.4	91.8	5.1

TABLE 3

Rupture Results for Ti 75A

Treatment	Test Temp (°F)	Stress (psi)	Time (hrs)	Elongation (% in 1 in.)	Reduction of Area (%)	100-Hour Rupture Strength (psi)
As Received (hot rolled and annealed)	600	37,600	0.1(T)	--	--	
		37,000	0.15	45.0	65.5	
		35,000	>911.0	(5.5)*	--	
		33,000	>1313.6	(2.8)*	--	36,000
	800	32,700	0.1(T)	--	--	
		20,000	38.5	47.6	82.7	
		16,000	155.9	86.1	76.9	17,000
	1000	22,000	0.1(T)	--	--	
		9,500	4.9	109.2	89.4	
		5,000	54.4	119.4	92.4	4,400
1500°F Annealed	600	34,900	0.1(T)	--	--	
		33,000	0.3	43.5	71.5	
		32,000	>977.7	(7.4)*	--	
		28,000	>1512.0	(3.7)*	--	32,500
	800	28,300	0.1(T)	--	--	
		20,000	54.1	59.6	72.0	
		16,000	126.6	66.5	74.3	17,500
	1000	17,970	0.1(T)	--	--	
		6,500	25.0	--	--	
		5,000	92.5	122.0	88.3	4,900
1700°F Annealed	600	33,600	0.1(T)	--	--	
		31,500	0.2	34.0	55.4	
		30,000	>957.1	(7.8)*	--	
		25,000	>1002.4	(2.0)*	--	31,000
	800	28,500	0.1(T)	--	--	
		19,000	79.0	47.0	70.0	
		15,000	184.3	69.5	76.8	18,000
	1000	19,750	0.1(T)	--	--	
		6,500	27.8	102.0	89.0	
		5,500	65.2	107.5	89.8	
		2,500	766.0			5,000

TABLE 3 , Continued

Treatment	Test Temp (°F)	Stress (psi)	Time (hours)	Elongation (% in 1 in.)	Reduction of Area (%)	100-Hour Rupture Strength (psi)
1500°F Annealed + 10.9% Cold Work	600	53,300	0.1(T)	--	--	
		51,000	0.3	14.7	45.0	
		48,500	1.0	14.7	57.0	45,000+
	800	41,300	0.1(T)	--	--	
		26,000	13.6	48.0	77.6	
		19,000	159.4	54.0	76.8	20,000
	1000	23,800	0.1(T)	--	--	
		6,500	88.4	90.2	88.4	
		5,000	231±7	77.5	90.3	
		4,000	259.8	129.5	94.0	6,200
1500°F Annealed + 31.3% Cold Work	600	67,000	0.1(T)	--	--	
		64,000	<0.2	16.5	52.0	
		61,000	0.2	15.4	46.6	
		50,000	237.9	18.3	58.1	
		47,000	720.9	66.9	66.5	55,000
	800	50,800	0.1(T)	--	--	
		31,000	7.3	34.0	63.0	
		22,000	64.9	49.5	75.4	20,500
	1000	26,800	0.1(T)	--	--	
		5,800	30.5	111.5	91.5	
		4,200	170.1	159.0	91.9	4,600
1700°F Annealed + 10.9% Cold Work	600	53,700	0.1(T)	--	--	
		51,000	<0.1	13.1	53.5	
		45,000	54.4	15.2	61.6	44,000
	800	42,400	0.1(T)	--	--	
		26,000	20.4	49.5	70.0	
		22,000	63 ± 3	67.5	74.0	20,500
	1000	20,900	0.1(T)	--	--	
		7,000	60.3±3	108.5	90.5	
		5,500	220.2	106.5	90.4	
		4,200	610.1	176.5	94.0	6,200

TABLE 3 , Continued

Treatment	Test Temp (°F)	Stress (psi)	Time (hours)	Elongation (% in 1 in.)	Reduction of Area (%)	100-Hour Rupture Strength (psi)
1700°F Annealed + 31.3% Cold Work	600	66,500	0.1(T)	--	--	
		62,000	<0.1	15.5	56.2	
		59,000	0.3	14.4	59.5	
		53,000	26.9	23.2	62.0	
		50,000	523.5	19.6	55.4	52,000
	800	48,900	0.1(T)	--	--	
		29,000	13.8	39.8	54.4	
		20,000	180.6	64.0	69.0	21,500
	1000	23,700	0.1(T)	--	--	
		6,500	25.8	103.0	98.9	
		5,000	80.5	87.6	90.8	
		3,000	765.3	171.0	92.0	4,800

NOTE: > Greater than, i.e., test stopped without rupture.
* Permanent deformation on cooling.
(T) Tensile test used as 0.1-hour rupture time.

All Heat Treatment Times 1 Hour.

TABLE 4

Creep Data for Ti 75A

Treatment	Test Temp (°F)	Stress (psi)	Minimum Creep Rate (in/in/hr)	Test Time (hrs)	Est. Stress for Rate of 5×10^{-6} in/in/hr-(psi)
As Received	600	35,000	4.7×10^{-7}	911.0	
		33,000	$<10^{-8}$	1313.6	36,500
	800	16,000	8.6×10^{-4}	155.9	
		7,000	1.8×10^{-5}	736.0	5,500
	1000	5,000	3.3×10^{-3}	54.4	
		2,000	5.0×10^{-5}	575.0	1,500
1500°F Anneal (1 hr at temp. + furnace cool)	600	32,000	$<10^{-8}$	977.7	36,500
	800	20,000	1.9×10^{-3}	54.1	
		16,000	8.3×10^{-4}	126.6	
		6,000	1.8×10^{-5}	779.5	5,500
	1000	5,000	3.4×10^{-3}	92.5	
		2,500	1.6×10^{-4}	1126.4	1,500
1700°F Anneal	600	30,000	$<10^{-8}$	957.1	36,500
	800	15,000	8.0×10^{-4}	184.3	
		6,000	3.5×10^{-6}	1177.5	5,500
	1000	5,500	6.0×10^{-3}	65.2	
		2,500	1.4×10^{-5}	766.0	1,500
1500°F Anneal + 10.9% Cold Work	600	40,000	3.1×10^{-6}	835.6	40,100
	800	19,000	2.9×10^{-4}	159.4	
		12,000	4.6×10^{-5}	1052.4	9,500
	1000	6,500	1.9×10^{-3}	88.4	
		4,000	8.0×10^{-4}	259.8	3,900
1500°F Anneal + 31.3% Cold Work	600	50,000	3.7×10^{-4}	237.9	
		47,000	9.5×10^{-5}	720.9	42,000
	800	22,000	4.2×10^{-4}	64.9	
		15,000	5.2×10^{-5}	1217.0	11,100
	1000	4,200	1.6×10^{-3}	170.1	
		2,500	1.3×10^{-4}	1200.0	100
1700°F Anneal + 10.9% Cold Work	600	45,000	4.0×10^{-4}	54.4	
		40,000	1.7×10^{-6}	1026.6	40,200
	800	22,000	2.2×10^{-3}	63.0	
		14,000	7.8×10^{-5}	1097.0	9,500
	1000	7,000	3.6×10^{-3}	60.3	
		5,500	6.7×10^{-5}	220.2	
		4,200	2.1×10^{-5}	610.0	3,900
1700°F Anneal + 31.3% Cold Work	600	50,000	9.1×10^{-5}	523.5	42,000
	800	20,000	7.6×10^{-4}	180.6	
		15,000	2.4×10^{-5}	1188.0	11,100
	1000	6,500	8.0×10^{-3}	25.8	
		5,000	1.6×10^{-3}	80.5	
		3,000	3.8×10^{-4}	765.3	100

TABLE 5

Tensile Data for Ti 150A

Treatment	Test Temp (°F)	Ultimate Strength (psi)	Yield Strength (0.2% Offset) (psi)	Elongation (% in 1 in.)	Reduction of Area (%)	Modulus of Elasticity x 10^{-6} psi
As Received	75	159,000	152,000	17.1	23.8	14.5
	600	90,600	63,000	30.4	60.6	10.1
	800	69,900	50,000	36.2	75.5	9.4
	1000	37,800	29,600	93.0	98.0	6.0
1500°F Annealed	75	153,800	147,600	26.7	35.2	15.5
	600	85,300	61,600	29.5	61.0	12.9
	800	68,300	49,500	41.7	74.3	10.7
	1000	37,400	28,500	68.0	93.5	8.9
1500°F Air Cooled	75	177,200	162,500	16.7	32.0	14.5
	600	115,100	75,800	19.6	70.4	13.5
	800	96,100	56,000	36.5	80.9	9.0
	1000	46,000	25,200	91.5	97.0	7.0
1500°F Water Quenched	75	222,000	220,000	3.9	3.5	15.6
	600	162,300	152,800	4.8	11.6	13.4
	800	138,700	105,200	36.9	76.9	11.4
	1000	46,700	31,800	99.2	98.0	4.9
1350°F Annealed	75	156,200	154,100	25.7	43.5	14.6
	600	87,000	62,000	28.4	56.0	10.7
	800	70,200	53,600	41.1	76.5	10.8
	1000	39,500	32,800	105.6	84.0	6.2
1350°F Air Cooled	75	165,900	164,500	22.3	44.2	14.3
	600	107,300	82,800	16.4	54.3	12.5
	800	88,500	59,000	40.4	77.0	10.7
	1000	41,700	28,500	115.4	98.0	6.3

TABLE 5 , Continued

Treatment	Test Temp (°F)	Ultimate Strength (psi)	Yield Strength (0.2% Offset) (psi)	Elongation (% in 1 in.)	Reduction of Area (%)	Elastic Modulus x 10^{-6} psi
1350°F Water Quenched	75	170,100	159,900	18.2	39.2	14.3
	600	119,000	98,400	12.9	47.0	13.4
	800	106,200	75,000	34.3	69.2	10.6
	1000	41,500	24,300	98.0	98.0	4.9
1500°F Water Quenched + 1350°F Annealed	75	156,900	152,800	14.6	15.2	16.6
	600	86,500	65,800	30.5	57.9	12.0
	800	69,800	51,500	40.0	77.3	11.1
	1000	39,500	32,300	123.0	98.4	8.2
1500°F Air Cooled + 900°F Annealed	75	184,900	160,000	<1.0	0	13.0
	600	115,900	78,800	21.9	53.7	10.8
	800	93,800	64,900	36.8	78.5	7.7
	1000	44,700	31,400	95.2	98.0	5.1
1350°F Water Quenched + 900°F Annealed	75	>193,000	190,000	*	*	14.7
	600	126,500	84,500	19.4	27.0	12.0
	800	98,800	59,000	31.8	76.0	9.5
	1000	43,500	25,800	91.5	99.0	6.8
1800°F+Isothermal 1300°F + Water Quenched	75	167,800	164,200	19.8	40.0	13.3
	600	113,600	70,000	15.0	35.5	12.7
	800	92,700	59,000	26.0	33.3	15.1(?)
	1000	48,500	35,500	110.7	99.0	6.5
1800°F + Isothermal 1000°F + Water Quenched	75	197,500	189,000	4.0	5.6	15.3
	600	98,300	76,400	17.6	51.4	11.6
	800	98,900	79,400	20.8	65.5	11.0
	1000	59,100	41,700	79.3	99.0	6.1
1800°F Annealed	75	150,800	132,000	11.0	14.7	14.7
1800°F Water Quenched		Too brittle to machine by ordinary methods.				

* Broke in threads.

NOTE: All heat treatment times at temperature 1 hour. Elastic modulus computed from one stress-strain curve.

TABLE 6

Rupture Data for Ti 150A

Treatment	Test Temp (°F)	Stress (psi)	Time (hours)	Elongation (% in 1 in.)	Reduction of Area (%)	100-Hour Rupture Strength (psi)
As Received (hot rolled and annealed)	600	90,600	0.1(T)	--	--	
		89,000	~0.1	30.6	60.0	
		85,000	1036.1	34.0	56.3	
		80,000	>979.3	(5.7)*	--	87,000
	800	69,900	0.1(T)	--	--	
		50,000	5.5	44.2	76.6	
		38,000	142.7	51.5	82.0	
		35,000	412.0	52.5	82.9	39,000
	1000	37,800	0.1(T)	--	--	
		20,000	3.2	65.5	87.8	
		12,000	150.2	55.3	92.6	12,500
1500°F Annealed	600	85,300	0.1(T)	--	--	
		82,500	0.2	32.7	44.5	
		80,000	>791.6	(11.6)*	--	82,000
	800	68,300	0.1(T)	--	--	
		45,000	94.6	53.4	78.4	
		42,000	118.1	44.8	76.0	43,000
	1000	37,400	0.1(T)	--	--	
		14,000	24.7	87.0	88.4	
		11,000	116.0	82.0	92.0	11,500
1500°F Air Cooled	600	115,100	0.1(T)	--	--	
		112,000	128.8	22.8	50.4	
		110,000	>985.0	(12.4)*	--	111,500
	800	96,100	0.1(T)	--	--	
		60,000	31.8	30.0	70.0	
		52,000	100.5	48.0	81.0	52,000
	1000	46,000	0.1(T)	--	--	
		16,000	16.1	90.0	92.0	
		11,000	188.8	42.7	92.0	12,000

TABLE 6 , Continued

Treatment	Test Temp (°F)	Stress (psi)	Time (hours)	Elongation (% in 1 in.)	Reduction of Area (%)	100-Hour Rupture Strength (psi)
1500°F Water Quenched	600	162,300	0.1(T)	--	--	
		157,000	146.0	19.6	28.6	
		155,000	125.4	11.8	30.8	~155,000
	800	138,700	0.1(T)	--	--	
		85,000	2.3	37.0	75.5	
		48,000	659.2	27.0	84.0	60,000
	1000	46,700	0.1(T)	--	--	
		16,000	7 ± 1	67.0	94.0	
		9,000	348.8	60.0	95.0	11,000
1350°F Annealed	600	87,000	0.1(T)	--	--	
		84,000	4.7	28.0	58.4	
		83,000	>1172.0	--	--	
		82,000	>974.1	(19.4)*	--	83,500
	800	70,200	0.1(T)	--	--	
		44,000	84.0	40.2	77.4	
		40,000	166.4	42.5	76.4	43,000
	1000	39,500	0.1(T)	--	--	
		14,000	34.0	59.5	92.0	
		12,000	157 ± 4	62.0	96.0	13,500
1350°F Air Cooled	600	107,300	0.1(T)	--	--	
		104,000	>1824	--	--	
		102,000	>1033	(7.8)*	--	~106,000
	800	88,500	0.1(T)	--	--	
		55,000	43.6	42.0	76.5	
		50,000	101.6	52.0	79.4	50,000
	1000	41,700	0.1(T)	--	--	
		14,000	33.6	57.2	71.8	
		11,000	194.3	55.0	95.0	12,500

TABLE 6 , Continued

Treatment	Test Temp (°F)	Stress (psi)	Time (hours)	Elongation (% in 1 in.)	Reduction of Area (%)	100-Hour Rupture Strength (psi)
1350°F Water Quenched	600	119,000	0.1(T)	--	--	
		115,000	>764.2	(7.5)*	--	
		114,000	368.8	7.6	1.6	~115,000
	800	106,200	0.1(T)	--	--	
		65,000	26.4	41.0	--	
		55,000	61.8	39.2	77.0	
		37,500	1912.0	28.4	83.0	54,000
	1000	41,500	0.1(T)	--	--	
		15,000	16.7 ± 2.5	69.2	89.4	
		10,000	160.6	44.4	90.6	10,500
1500°F Water Quenched + 1350°F Annealed	600	86,500	0.1(T)	--	--	
		83,000	>1296	in progress		~84,000
	800	69,800	0.1(T)	--	--	
		45,000	21.8	40.0	74.1	
		39,000	191.5	41.6	78.8	41,000
	1000	39,500	0.1(T)	--	--	
		13,000	43.0	71.5	96.0	
		10,500	247.7	172.0	94.0	11,500
1500°F Air Cooled + 900°F Annealed	600	115,900	0.1(T)	--	--	
		112,000	4.4	15.2	55.4	
		110,000	92.3	13.6	58.3	110,000
	800	93,800	0.1(T)	--	--	
		60,000	50.0	32.6	77.9	
		56,000	103.4	30.2	78.7	56,000
	1000	44,700	0.1(T)	--	--	
		14,000	74 ± 3	40.0	95.6	
		12,500	145.1	40.5	91.5	13,500
1350°F Water Quenched + 900°F Annealed	600	126,500	0.1(T)	--	--	
		123,000	3.5	17.4	45.4	
		122,000	310.6	19.8	53.3	122,500

TABLE 6 , Continued

Treatment	Test Temp (°F)	Stress (psi)	Time (hours)	Elongation (% in 1 in.)	Reduction of Area (%)	100-Hour Rupture Strength (psi)
1350°F Water Quenched + 900°F Annealed	800	98,800	0.1(T)	--	--	
		58,000	37 ± 5	37.4	76.0	
		51,000	107.5	30.8	71.0	
		40,000	1530.0	42.0	77.2	51,000
	1000	43,500	0.1(T)	--	--	
		14,000	48.9	57.5	93.5	
		11,500	127.0	67.0	96.5	12,500
1800°F Solution + 1300°F Isothermal + Water Quenched	600	113,600	0.1(T)	--	--	
		110,000	37.6	6.8	8.6	
		108,000	>1146.2	(13.6)*	--	110,500
	800	92,700	0.1(T)	--	--	
		55,000	55 ± 4	42.0	78.0	
		45,000	320.4	26.4	76.0	52,000
	1000	48,500	0.1(T)	--	--	
		17,500	9 ± 7	51.5	91.0	
		11,000	582.0	45.5	93.0	14,000
1800°F Solution + 1000°F Isothermal + Water Quenched	600	126,000(?)	0.1(T)	--	--	
		120,000	0.4	11.8	32.2	
		110,000	787.8	13.5	40.0	
		95,000	>1291.0	(1.9)*	--	~117,000
	800	98,900	0.1(T)	--	--	
		60,000	8.4	30.4	73.7	
		55,000	95.1	30.5	80.6	55,000
	1000	59,100	0.1(T)	--	--	
		20,000	7 ± 2.5	36.0	93.5	
		11,000	203.7	47.0	93.5	12,000

NOTE: (T) Tensile test used as 0.1-hour rupture strength.
> Test stopped, did not rupture.
(*) Permanent deformation on cooling.
~ Approximately
All Heat treatment times 1 hour at temperature - annealing followed by furnace cool.

TABLE 7

Creep Data for Ti 150A

Treatment (See Note)	Test Temp (°F)	Stress (psi)	Minimum Creep Rate (in/in/hr)	Test Time (hrs)	Est. Stress for Rate of 5×10^{-6} in/in/hr - (psi)
As Received (Hot Rolled and Annealed)	600	85,000		1036.1	
		80,000	2.0×10^{-5}	974.3	79,500
	800	38,000	1.1×10^{-4}	142.7	
		35,000	8.5×10^{-4}	412.0	23,000
	1000	12,000	6.5×10^{-4}	150.2	
		6,000	1.1×10^{-5}	768.6	5,000
1500°F Anneal	600	80,000	2.0×10^{-5}	791.6	79,500
	800	45,000	2.5×10^{-3}	24.6	
		42,000	1.2×10^{-3}	118.1	
		29,000	3.9×10^{-5}	1534.7	23,000
	1000	14,000	8.0×10^{-3}	24.7	
		11,000	1.4×10^{-3}	116.0	
		5,000	4.8×10^{-6}	860.4	5,000
1500°F Air Cool	600	112,000	5.0×10^{-4}	128.8	
		110,000	1.1×10^{-4}	985	
		108.000	4.3×10^{-5}		100,000
	800	52,000	1.4×10^{-3}	100.5	
		35,000	5.9×10^{-6}		33,000
	1000	11,000	4.0×10^{-4}	188.8	
		5,000	8.9×10^{-6}	979.7	5,000
1500°F Water Quench	600	157,000	4.0×10^{-4}	146.0	
		155,000	3.2×10^{-4}	125.4	
		150,000			120,000
	800	48,000	7.4×10^{-5}	659.2	
		41,000	1.5×10^{-5}	810.9	37,000
	1000	9,000	3.3×10^{-4}	348.8	
		5,000	1.5×10^{-5}	1034.9	5,000
1350°F Anneal	600	83,000	1.7×10^{-5}	1172.9	
		82,000	3.5×10^{-6}	979.1	79,500
	800	44,000	1.6×10^{-3}	84.0	
		40,000	8.0×10^{-4}	166.4	
		15,000	2.8×10^{-7}	1626.5	23,000
	1000	14,000	6.2×10^{-4}	34.0	
		12,000	5.3×10^{-4}	157±4	
		5,000	5.6×10^{-6}		5,000
1350°F Air Cool	600	104,000	1.4×10^{-4}	1824.0	
		102,000	1.6×10^{-5}	1033.0	100,000
	800	55,000	8.0×10^{-3}	43.6	
		50,000	1.5×10^{-3}	101.6	
		35,000	2.5×10^{-5}		33,000
	1000	14,000	4.7×10^{-3}	33.6	
		11,000	4.6×10^{-4}	194.3	
		6,000	1.6×10^{-5}	1033.9	5,000

TABLE 7, Continued

Treatment (See Note)	Test Temp (°F)	Stress (psi)	Minimum Creep Rate (in/in/hr)	Test Time (hrs)	Est. Stress for Rate of 5×10^{-6} in/in/hr - (psi)
1350°F Water Quench	600	116,000	7.2×10^{-5}		
		115,500	3.3×10^{-5}	764.2	100,000
	800	55,000	2.0×10^{-3}	61.8	
		37,500	1.6×10^{-5}	1912.0	33,000
	1000	10,000	1.4×10^{-3}	160.6	
		5,000	9.1×10^{-6}	1337.0	5,000
1500°F Water Quench + 1350°F Anneal	600	83,000	1.1×10^{-5}	1296.0	
		79,000	1.0×10^{-5}		78,000
	800	39,000	8.7×10^{-4}	191.5	
		18,000			23,000
	1000	10,500	4.6×10^{-4}	247.7	
		5,000	3.6×10^{-6}		5,000
1500°F Air Cool + 900°F Anneal	600	110,000	4.0×10^{-4}	92.3	100,000
	800	60,000	3.6×10^{-3}	50.0	
		56,000	1.1×10^{-3}	103.4	33,000
		35,000	1.4×10^{-3}		
	1000	14,000	5.3×10^{-3}	74.3	
		12,500		145.1	
		5,000	7.0×10^{-6}		5,000
1350°F Water Quench + 900°F Anneal	600	122,000	6.7×10^{-4}	310.6	100,000
	800	51,000	1.0×10^{-3}	107.5	
		40,000	3.6×10^{-5}	1530.0	33,000
	1000	14,000	3.6×10^{-3}	48.9	
		11,000	1.1×10^{-3}	127.0	
		5,000	6.7×10^{-6}	1367.5	5,000
1800°F (1 hr) + Isothermal (1 hr) 1300°F + Water Quench	600	108,000	1.4×10^{-4}	1146.2	
		105,000	6.7×10^{-5}		100,000
	800	45,000	2.9×10^{-4}	320.4	
		35,000	7.6×10^{-6}		35,000
	1000	11,000	8.3×10^{-5}	582.0	
		5,000	4.4×10^{-6}		5,000
1800°F (1 hr) + Isothermal (1 hr) 1000°F + Water Quench	600	110,000	7.2×10^{-5}	787.8	
		95,000	9.1×10^{-6}	1291.1	100,000
	800	55,000	1.3×10^{-3}	95.1	
		45,000	1.8×10^{-5}	790.0	33,000
	1000	11,000	6.7×10^{-4}	203.7	
		5,000	1.2×10^{-5}		4,700

NOTE: All Heat Treatment Times One Hour.

TABLE 8

X-Ray Diffraction Traverse of Ti 150A Specimens after Testing

Treatment and Test Details	Number of Discrete Diffraction Lines Observed	
	α Lines	β Lines
As Received - 979 hr - 80,000 psi - 600°F	11	5
As Received - 150 hr - 12,000 psi - 1000°F	14	1
1500°F Air Cooled - 1000°F Tensile Test	10	3
1500°F Air Cooled - 16 hr - 16,000 psi - 1000°F	8	1
1500°F Water Quench - 1000°F Tensile Test	11	5
1500°F Water Quench - 7 hr - 16,000 psi - 1000°F	11	3
1500°F Water Quench - 157 hr - 12,000 psi - 1000°F	11	4
1350°F Air Cooled - 33 hr - 14,000 psi - 1000°F	12	3
1350°F Water Quench - 26 hr - 65,000 psi - 800°F	4	0
1350°F Water Quench - 160 hr - 10,000 psi - 1000°F	10	4
1350°F Annealed - 84 hr - 44,000 psi - 800°F	7	1
1300°F Isothermal - 800°F Tensile Test	12	3
1000°F Isothermal - 8 hr - 60,000 psi - 800°F	10	0
1000°F Isothermal - 10 hr - 20,000 psi - 1000°F	12	0

NOTE: All heat treatment times 1 hour.

Isothermal Treatments: 1 hr 1800°F + 1 hr at temperature (leadpot) + water quench.

TABLE 9

Tensile Test Data for Experimental Alpha Alloy - 6% Al

Treatment	Test Temp (°F)	Ultimate Strength (psi)	Yield Strength (0.2% Offset) (psi)	Elongation (% in 1 in.)	Reduction of Area (%)	Elastic Modulus x 10^{-6} psi
As Forged	75	146,000	135,000	18.8	34.1	17.1
	600	89,000	74,000	20.0	46.4	15.0
	1000	77,500	60,000	24.8	41.2	9.8
2025°F (1 hr) + Water Quench	75	142,000	122,000	18.2	38.6	15.0
	600	97,900	78,500	17.6	34.1	12.4
	1000	80,300	64,800	14.9	33.8	10.8
2025°F (1 hr) + Air Cool	75	141,200	121,000	7.1	14.7	15.4
As Forged + 10% Cold Work	75	168,000	155,000	9.7	26.6	16.2
	600	97,800	94,000	9.6	45.8	14.2
	1000	83,300	68,500	18.6	38.1	10.1
As Forged + 17% Cold Work	75	177,200	163,800	10.7	28.0	15.1
	600	105,100	100,600	10.0	34.7	13.4
	1000	83,400	67,200	14.6	32.8	11.1
As Forged + 17% Cold Work + 1500°F (2 hrs) + Air Cool	75	137,000	129,800	15.5	20.7	14.8
	600	87,600	62,800	21.5	43.8	12.3
	1000	78,600	57,000	23.0	46.1	9.4
As Forged + 17% Cold Work + 1700°F (3 hrs) + Air Cool	75	135,000	129,000	13.0	34.6	14.3
	600	82,300	61,500	25.5	44.0	11.5
	1000	72,200	56,200	18.5	31.3	11.0

NOTE: Elastic Modulus computed from one stress-strain curve.

TABLE 10

Rupture Data for Experimental Alpha Alloy - 6% Al

Treatment	Test Temp (°F)	Stress (psi)	Time (hrs)	Elongation (% in 1 in.)	Reduction of Area (%)	100-Hour Strength (psi)
As Forged	600	89,000	0.1(T)	--	--	
		87,000	>500	in progress		88,000
	1000	77,500	0.1(T)	--	--	
		37,500	102.3	28.6	66.6	37,500
2025°F (1 hr) + Water Quench	600	97,900	0.1(T)	--	--	
		96,000	>1443.9	broken accidentally		96,500
	1000	80,300	0.1(T)	--	--	
		35,000	361.0	30.6	34.2	40,000
As Forged + 10% Cold Work	600	97,800	0.1(T)	--	--	
		96,500	>1216	(1.0)*	--	
		95,000	259.8	4.9	22.4	96,000
	1000	83,300	0.1(T)	--	--	
		40,000	20.5	43.2	68.6	
		26,800	176.4	53.5	81.6	30,000
As Forged + 17% Cold Work	600	105,100	0.1(T)	--	--	
		103,500	<0.2	12.9	52.8	
		101,000	>1172.2	(1.0)*	--	102,000
	1000	83,400	0.1(T)	--	--	
		37,000	13.4	19.0	32.8	
		26,000	123.6	56.0	83.0	27,000
As Forged + 17% Cold Work + 1500°F (2 hrs) + Air Cool	600	87,600	0.1(T)	--	--	
		84,000	>500	in progress		86,000
	1000	78,600	0.1(T)	--	--	
		28,000	351.5	48.0	59.0	33,000
As Forged + 17% Cold Work + 1700°F (3 hrs) + Air Cool	600	82,800	0.1(T)	--	--	
		79,500	>500	in progress		80,500
	1000	72,200	0.1(T)	--	--	
		30,000	55.0	9.8	19.7	
		27,500	452.1	47.0	58.6	29,500

Note: (T) Tensile Test used as 0.1 hour rupture stress.
* Permanent deformation on cooling.
> More than.
< Less than.

TABLE 11

Creep Data for Experimental Stable Alpha Alloy - 6% Al

Treatment	Test Temp (°F)	Stress (psi)	Minimum Creep Rate (in/in/hr)	Test Time (hrs)	Est. Stress for Rate of 5×10^{-6} in/in/hr - (psi)
As Forged	600	87,000	8.7×10^{-7}	>1000	88,000
	1000	37,500	1.1×10^{-3}	102.3	
		18,000	1.6×10^{-4}	975.5	<3,500
2025°F (1 hr) + Water Quench	600	96,000	$<10^{-8}$	1443.9	97,000
	1000	35,000	2.3×10^{-4}	361.0	
		20,000	3.8×10^{-5}	912.0	5,000
As Forged + 10% Cold Work	600	96,500	6.6×10^{-7}	1216.0	
		95,000	3.8×10^{-6}	259.8	97,000
	1000	26,800	1.0×10^{-3}	176.4	
		15,000	5.4×10^{-5}	1198.9	5,000
As Forged + 17% Cold Work	600	102,000	5.7×10^{-6}	954.8	
		101,000	2.8×10^{-6}	1172.2	102,000
	1000	37,000	9.9×10^{-3}	13.4	
		26,000	1.6×10^{-3}	123.6	
		15,000	8.4×10^{-5}	>1000	5,000
As Forged + 17% Cold Work + 1500°F (2 hrs) + Air Cool	600	84,000	1.8×10^{-6}	>1000	85,000
	1000	28,000	4.3×10^{-4}	351.5	
		17,000	1.4×10^{-5}	>1000	12,500
As Forged + 17% Cold Work + 1700°F (3 hrs) + Air Cool	600	79,500	2.5×10^{-6}	>1000	80,000
	1000	30,000	8.3×10^{-4}	55.0	
		27,500	4.2×10^{-4}	452.1	12,500

TABLE 12

Tensile Data for Experimental Stable Beta Alloy - 30% Mo

Treatment	Test Temp (°F)	Ultimate Strength (psi)	Yield Strength (0.2% Offset) (psi)	Elongation (% in 1 in.)	Reduction of Area (%)	Elastic Modulus x 10^{-6} psi
As Forged	75	144,700	142,000	14.0	27.4	13.6
	600	106,000	96,000	13.3	39.7	11.8
	1000	83,200	74,500	25.5	59.4	11.5
1325°F (1 hr) + Water Quench	75	145,800	142,200	12.6	28.0	14.6
	600	105,400	93,600	15.8	43.4	13.6
	1000	82,500	74,400	22.1	40.5	13.0
1500°F (2 hrs) + Water Quench	75	144,000	142,400	12.2	21.6	13.1
	600	106,000	95,600	19.4	47.5	11.7
	1000	83,200	73,700	21.3	62.0	12.0
1500°F (2 hrs) + Water Quench + 1375°F (24 hrs) + Water Quench	75	140,500	136,600	11.6	22.6	14.0
	600	103,400	92,900	15.7	50.9	11.7
	1000	82,100	76,400	24.0	64.0	11.6
1325°F (1 hr) + Water Quench + 3.3% Cold Work	600	110,000	102,600	9.7	38.9	13.6
	1000	83,700	79,200	19.6	37.8	10.9

NOTE: Elastic modulus computed from one stress-strain curve.

TABLE 13

Rupture Test Results for Experimental Stable Beta Alloy - 30% Mo

Treatment	Test Temp (°F)	Stress (psi)	Time (hours)	Elongation (% in 1 in.)	Reduction of Area (%)	100-Hour Rupture Strength (psi)
As Forged	600	106,000	0.1(T)	--	--	
		90,000	>1158	overheated and damaged		~95,000
	1000	83,200	0.1(T)	--	--	
		40,000	77 ± 2	54.8	49.4	
		35,000	133.0	62.8	70.0	37,500
1325°F (1 hr) + Water Quench	600	105,400	0.1(T)	--	--	
		100,000	>836	broken accidentally		
		90,000	>1485.0	0	--	~103,000
	1000	82,500	0.1(T)	--	--	
		40,000	74.5	54.4	37.9	
		35,000	141.2	54.3	75.4	
		22,000	1143.9	84.0	82.1	37,500
1500°F (2 hrs) + Water Quench	1000	83,200	0.1(T)	--	--	
		37,500	83.4	66.6	71.2	37,000
1500°F (2 hrs) + Water Quench + 1375°F (24 hrs) + Water Quench	1000	82,100	0.1(T)	--	--	
		37,500	96.1	60.0	61.6	37,500

NOTE: > Test stopped, did not rupture.
~ Approximate.
(T) Tensile test used as 0.1-hour rupture time.

TABLE 14

Creep Data for Experimental Stable Beta Alloy - 30% Mo

Treatment	Test Temp (°F)	Stress (psi)	Minimum Creep Rate (in/in/hr)	Test Time (hrs)	Est. Stress for Rate of 5×10^{-6} in/in/hr - (psi)
As Forged	600	90,000	$<10^{-8}$	1158.0	--
	1000	40,000	8.8×10^{-4}	77±2	
		35,000	2.0×10^{-4}	133.0	
		20,000	4.6×10^{-5}	912.5	12,000
1325°F (1 hr) + Water Quench	600	102,500	8.3×10^{-6}	982.5	
		100,000	$<10^{-8}$	836.0	102,000
	1000	35,000	4.3×10^{-4}	141.2	
		22,000	2.5×10^{-5}	1143.9	12,000

TABLE 15

Hardness Data

Treatment (See Note)	Test Conditions	Knoop Hardness No. 1 Kg. Load
Ti 75A		
As Received	As Treated	201
1500°F Anneal	As Treated	214
1700°F Anneal	As Treated	197
1500°F Anneal + 10.9% Cold Work	As Treated	258
1500°F Anneal + 31.3% Cold Work	As Treated	291
1700°F Anneal + 10.9% Cold Work	As Treated	231
1700°F Anneal + 31.3% Cold Work	As Treated	267
Ti 150A		
As Received	As Received	323
	600°F-979 hrs-80,000 psi	368
	800°F-142.7 hrs-38,000 psi	375
	1000°F-150.2 hrs-12,000 psi	366
1500°F Anneal	As Treated	271
	600°F Tensile Test	361
	800°F Tensile Test	380
	800°F-118 hrs-42,000 psi	395
	800°F-94 hrs-45,000 psi	378
	1000°F Tensile Test	354
	1000°F-24.7 hrs-14,000 psi	333
	1000°F-116 hrs-11,000 psi	346
1500°F Air Cool	As Treated	363
	600°F Tensile Test	401
	800°F Tensile Test	415
	800°F-31.8 hrs-60,000 psi	401
	800°F-100.5 hrs-52,000 psi	424
	1000°F Tensile Test	375
	1000°F-16.1 hrs-16,000 psi	365
	1000°F-188.8 hrs-11,000 psi	375

TABLE 15 Continued

Treatment (See Note)	Test Conditions	Knoop Hardness No. 1 Kg. Load
1500°F Water Quench	As Treated	398
	600°F Tensile Test	521
	800°F Tensile Test	407
	800°F-2.3 hrs-85,000 psi	515
	1000°F Tensile Test	405
	1000°F-7 hrs - 16,000 psi	405
	1000°F-348.8 hrs-9,000 psi	317
1350°F Annealed	As Treated	279
	800°F Tensile Test	364
	800°F-84 hrs-44,000 psi	384
	1000°F Tensile Test	346
	1000°F-34 hrs-14,000 psi	359
1350°F Air Cooled	As Treated	345
	600°F Tensile Test	384
	800°F Tensile Test	375
	800°F-43.6 hrs-55,000 psi	370
	800°F-101.6 hrs-50,000 psi	378
	1000°F-33.6 hrs-14,000 psi	403
	1000°F-194.3 hrs-11,000 psi	365
1350°F Water Quench	As Treated	378
	600°F Tensile Test	419
	800°F Tensile Test	390
	800°F-26.4 hrs-65,000 psi	430
	800°F-61.8 hrs-55,000 psi	386
	1000°F Tensile Test	384
	1000°F-160.6 hrs-10,000 psi	384
1500°F Water Quench + 1350°F Anneal	As Treated	279
	1000°F Tensile Test	384
	1000°F-43 hrs-13,000 psi	342
1350°F WAter Quench + 900°F Anneal	As Treated	370
	75°F Tensile Test	384
	800°F Tensile Test	428
	800°F-37.5 hrs-58,000 psi	424
1300°F Isothermal	As Treated	326
	600°F Tensile Test	388
	800°F Tensile Test	493
	800°F-55 hrs-55,000 psi	392
	1000°F Tensile Test	380
	1000°F-16 hrs-17,500 psi	375
	1000°F-582 hrs-11,000 psi	343

TABLE 15, Continued

Treatment (See Note)	Test Conditions	Knoop Hardness No. 1 Kg. Load
1000°F Isothermal	As Treated	425
	600°F Tensile Test	365
	800°F Tensile Test	496
	800°F-8.4 hrs-60,000 psi	399
	1000°F Tensile Test	432
	1000°F-7.5 hrs-20,000 psi	423
	1000°F-203.7 hrs-11,000 psi	346
Experimental Stable Alpha Alloy - 6% Al		
As Forged	As Forged	317
2025°F (1 hr) + Water Quench	As Treated	352
	600°F Tensile Test	343
	600°F-1443 hrs-96,000 psi	388
	1000°F-1200 hrs-20,000 psi	375
10% Cold Work	As Treated	370
	600°F Tensile Test	317
	600°F-1216 hrs-96,500 psi	375
	1000°F-20.5 hrs-40,000 psi	343
	1000°F-1198 hrs-15,000 psi	324
17% Cold Work	As Treated	407
	600°F Tensile Test	330
	600°F-1172 hrs-101,000 psi	343
	1000°F Tensile Test	321
	1000°F-13.4 hrs-37,000 psi	334
17% Cold Work + 1500°F (2 hrs) + Air Cool		401
17% Cold Work + 1700°F (3 hrs) + Air Cool		393
Experimental Stable Beta Alloy - 30% Mo		
As Forged	As Forged	308
	1000°F-77 hrs-40,000 psi	288
1325°F (1 hr) + Water Quench	As Treated	305
	600°F-1485 hrs-90,000 psi	287
	600°F-836 hrs-100,000 psi	287
	1000°F-74.5 hrs-40,000 psi	289

TABLE 15, Continued

Treatment (See Note)	Test Conditions	Knoop Hardness No. 1 Kg. Load
1500°F (2 hrs) + Water Quench	As Treated	295
	600°F Tensile Test	289
	1000°F Tensile Test	297
	1000°F-83.4 hrs-37,500 psi	291

NOTE; All heat treatment times 1 hour unless otherwise noted.

TABLE 16

Comparison of Rupture and Creep Strengths

Property	Test Temp (°F)	Material			
		Ti 75A	Ti 150A	α-6% Al	β-30% Mo
100-Hour Rupture Strength (psi)	600	31,000-55,000	82,000-155,000	88,000-102,000	95,000-103,000
	800	17,000-21,500	39,000-60,000	--	--
	1000	4,400-6,200	10,500-14,000	27,000-40,000	37,500
Stress for Minimum Creep Rate of 5×10^{-6} in/in/hr (psi)	600	36,500-42,500	79,000-113,000	80,000-102,000	102,000
	800	5,200-11,000	23,000-36,000	--	--
	1000	100-3,500	5,000	3,500-12,500	12,000

TABLE 17

Structural Comparison of Alloy Properties

Property	Commercially Pure Ti Ti 75A	Alpha-Beta Alloy Ti 150A	Stable Alpha Alloy 6% Al	Stable Beta Alloy 30% Mo
Response to cold work	Strengthens (up to 30% in this study)	Not investigated	Strengthens. Limit 17%	Little Effect. Limit 3.5%
Response to heat treatment	Nil	Wide Range of Structures and property variations to 800°F	Very little. Grain growth above 1500°F after cold work	None (several solution temperatures and times tried)
Structural stability	Very stable	Unstable at 1000°F (or lower for very long times). Eutectoid decomposition reaction	Recrystallization possible at 1000°F under test stress for cold worked material	Precipitate w/ α phase after 100 hours at 1000°F
Strength governing agent	Internal strain aging (cold work)	Non-equilibrium β (α-β ratio)	Al content, inherent stability	Inherent stability, β strength
Fracture ductility	Excessive at 800°F and above. Max 175%	Excessive at 1000°F. Max 140%	Good up to 1000°F. Max 60%	Good up to 1000°F. Max 60%
Highest strength condition (up to stability temp)	30% cold work	1500°F water quench	2025°F water quench	as forged
Optimum temperature range for best properties	Up to 600°F	Up to 800°F	Above 800°F	Above 800°F
Relative strength-weight ratio	Good	Good	Best	Fair
Strength stability	Poor at 800°F and 1000°F	No difference at 1000°F between treatments	Good at 1000°F for annealed, cold worked and recrystallized structures	Good at 1000°F

TABLE 17, Continued

Property	Commercially Pure Ti Ti 75A	Alpha-Beta Alloy Ti 150A	Stable Alpha Alloy 6% Al	Stable Beta Alloy 30% Mo
Fracture type	600°F transgranular 1000°F transgranular	600°F transgranular 1000°F transgranular	600°F transgranular 1000°F intergranular	600°F transgranular 1000°F intergranular

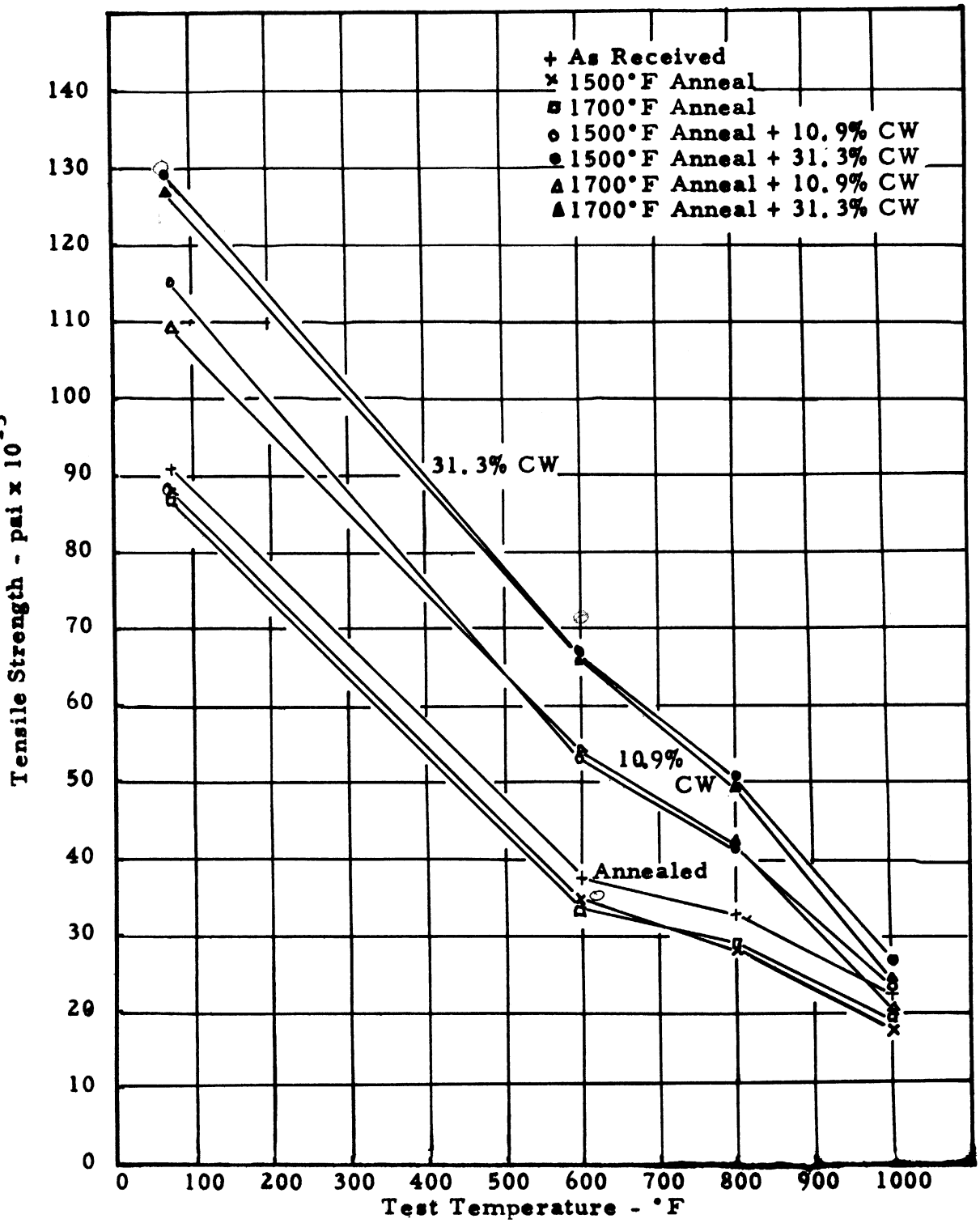


Fig. 1. Tensile Strength Vs. Test Temperature for Various Conditions of Ti 75A

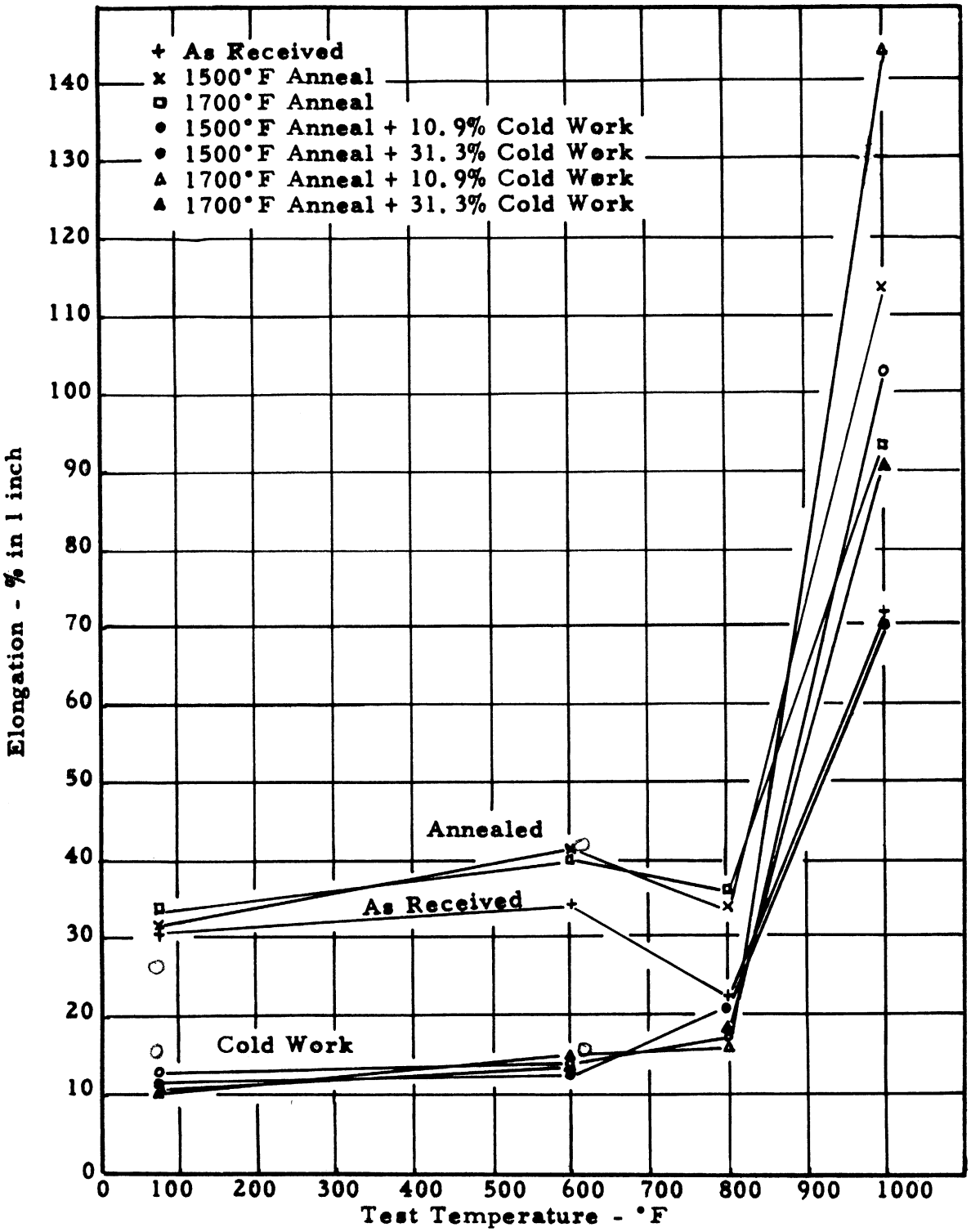


Fig. 2. Tensile Test Elongation Vs. Test Temperature for Various Conditions of Ti 75A

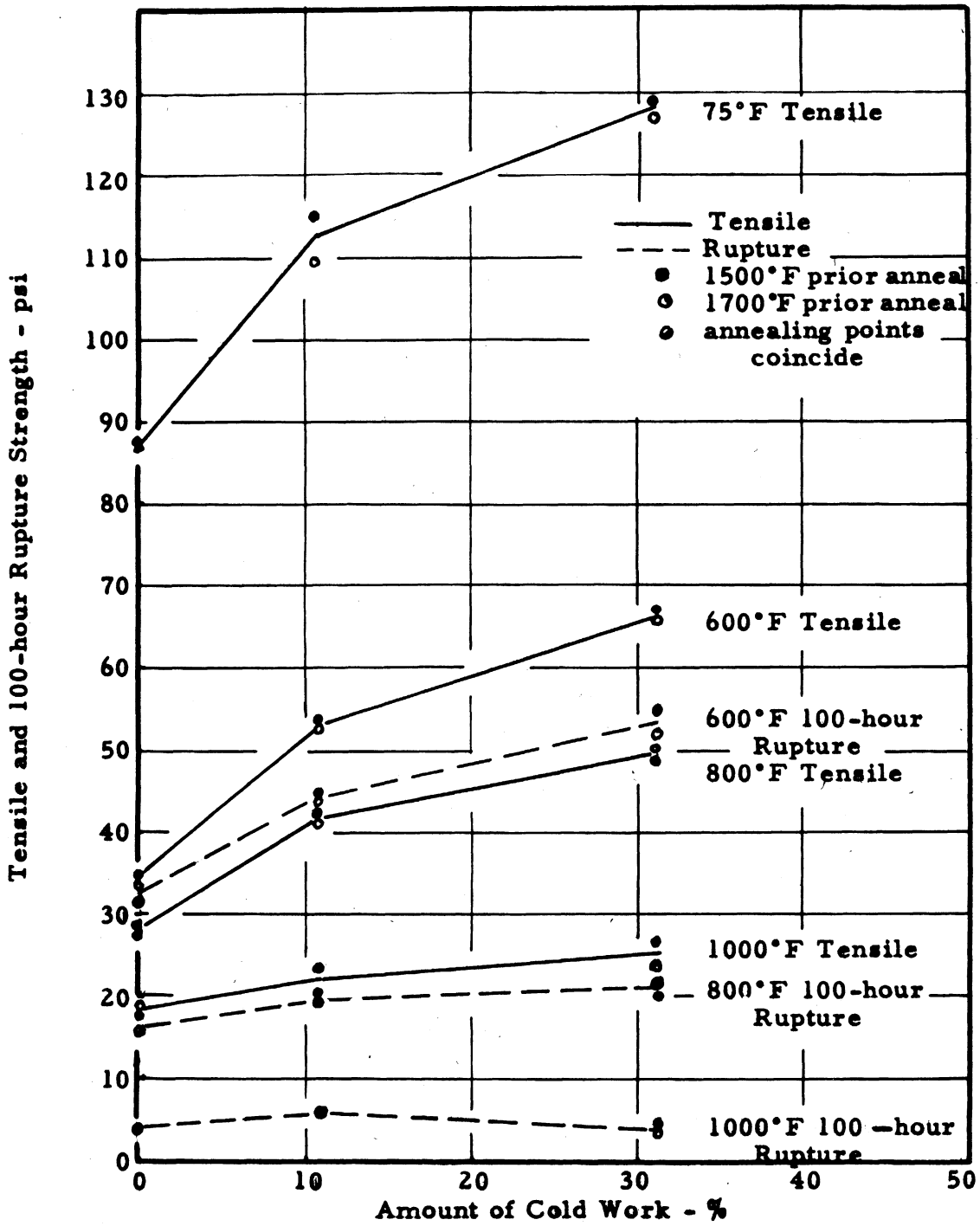


Fig. 3. Effect of Cold Work on Tensile and Rupture Strengths of Ti 75A at 75°, 600°, 800° and 1000°F

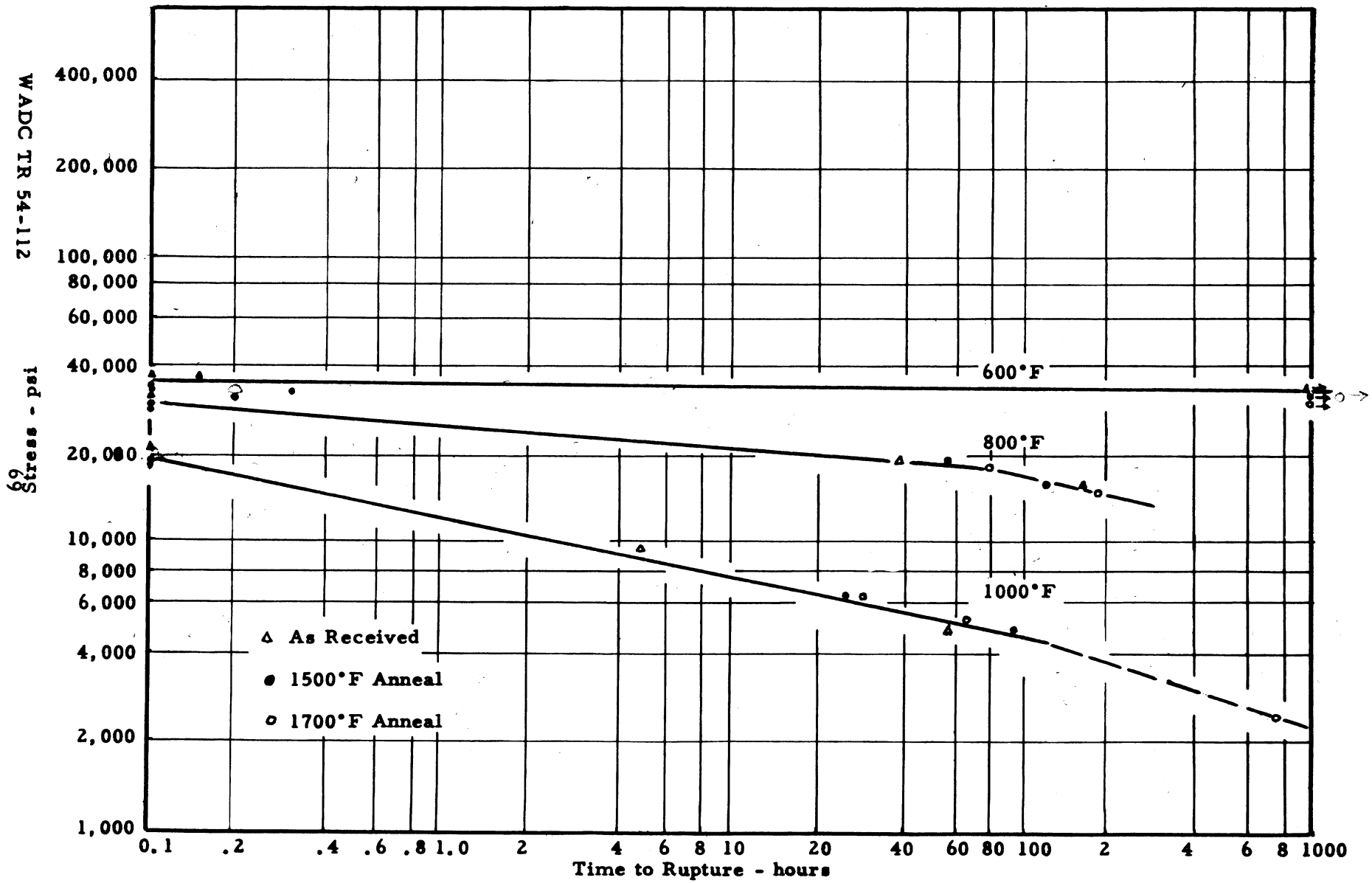


Fig. 4. Stress Vs. Time to Rupture for Annealed and As-Received Ti 75A at 600°, 800° and 1000°F.

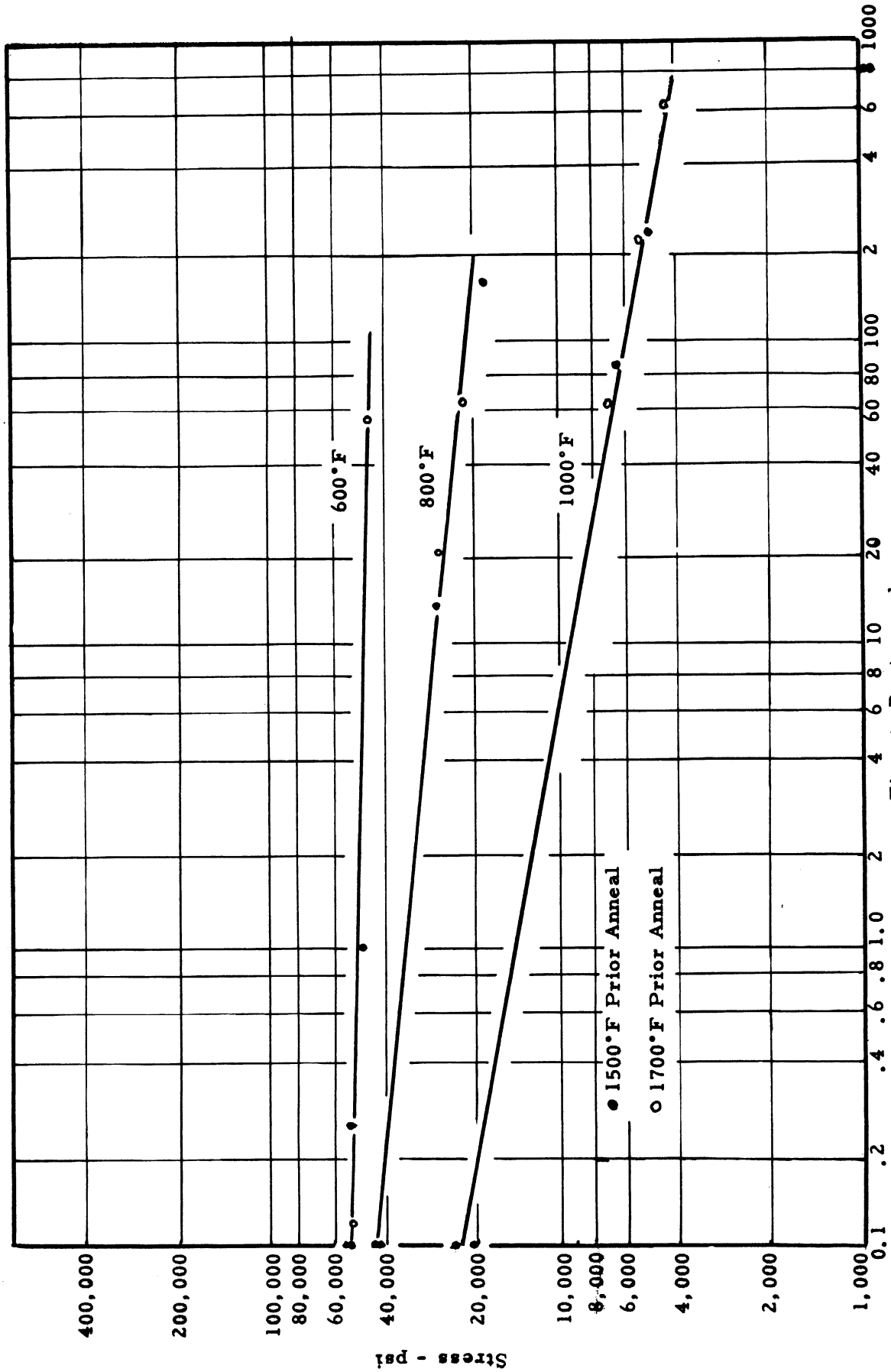


Fig. 5. Stress Vs. Time to Rupture at 600°, 800° and 1000°F for Ti 75A Cold Worked 10.9%

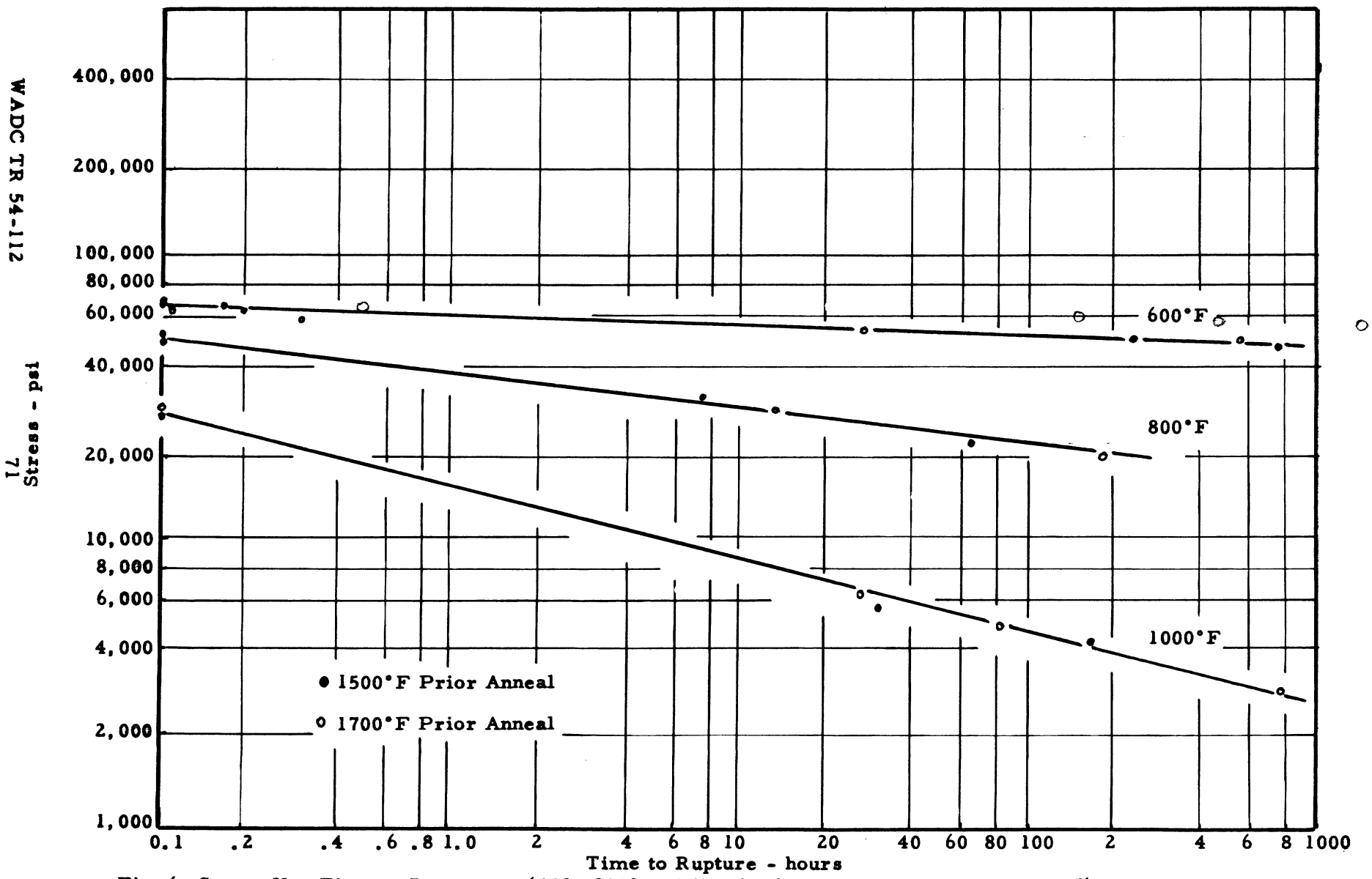


Fig. 6. Stress Vs. Time to Rupture at 600°, 800° and 1000°F for Ti 75A Cold Worked 31.3%

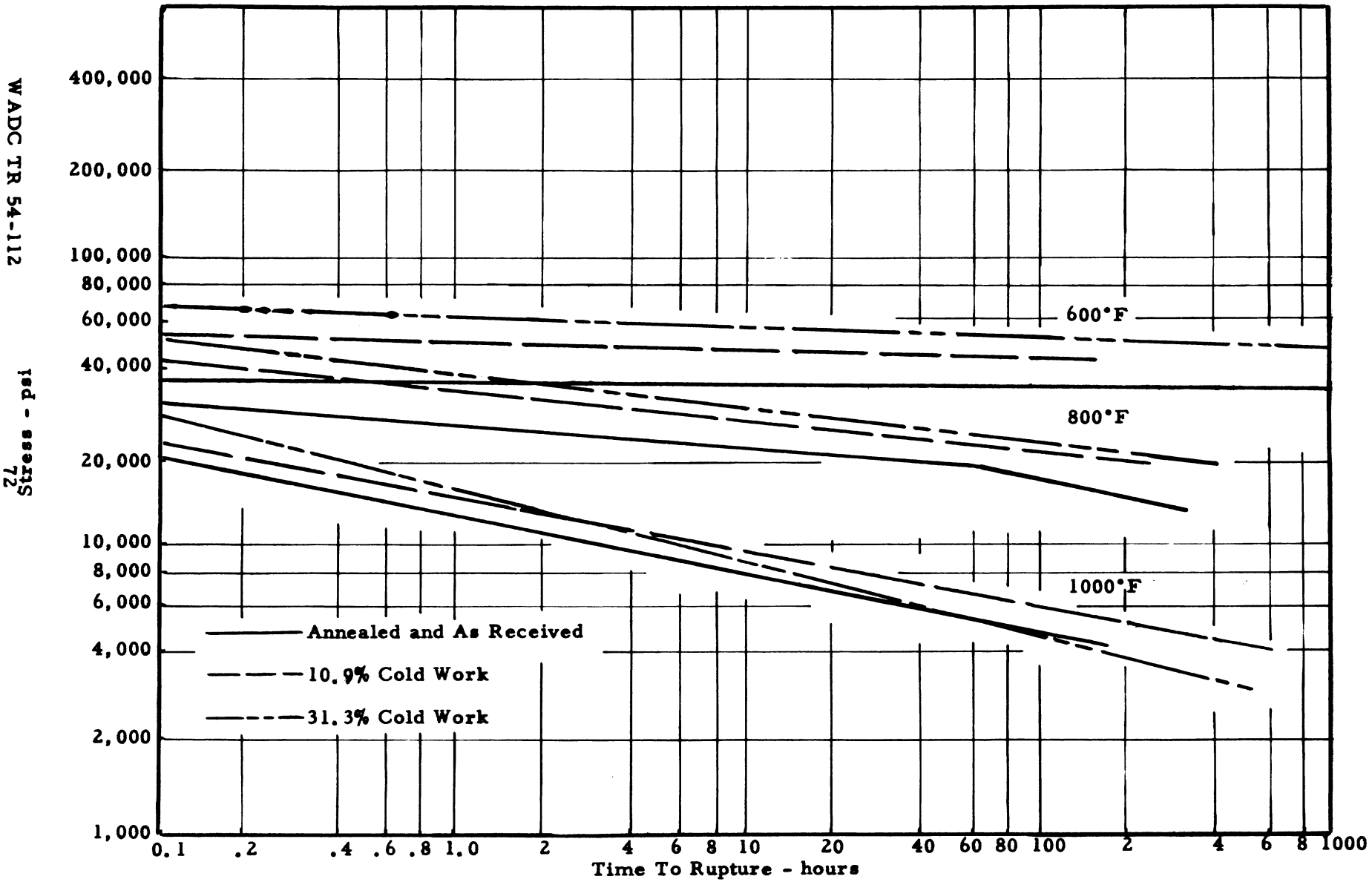


Fig. 7. Summary of Rupture Curves at 600°, 800° and 1000°F for All Conditions of Ti 75A

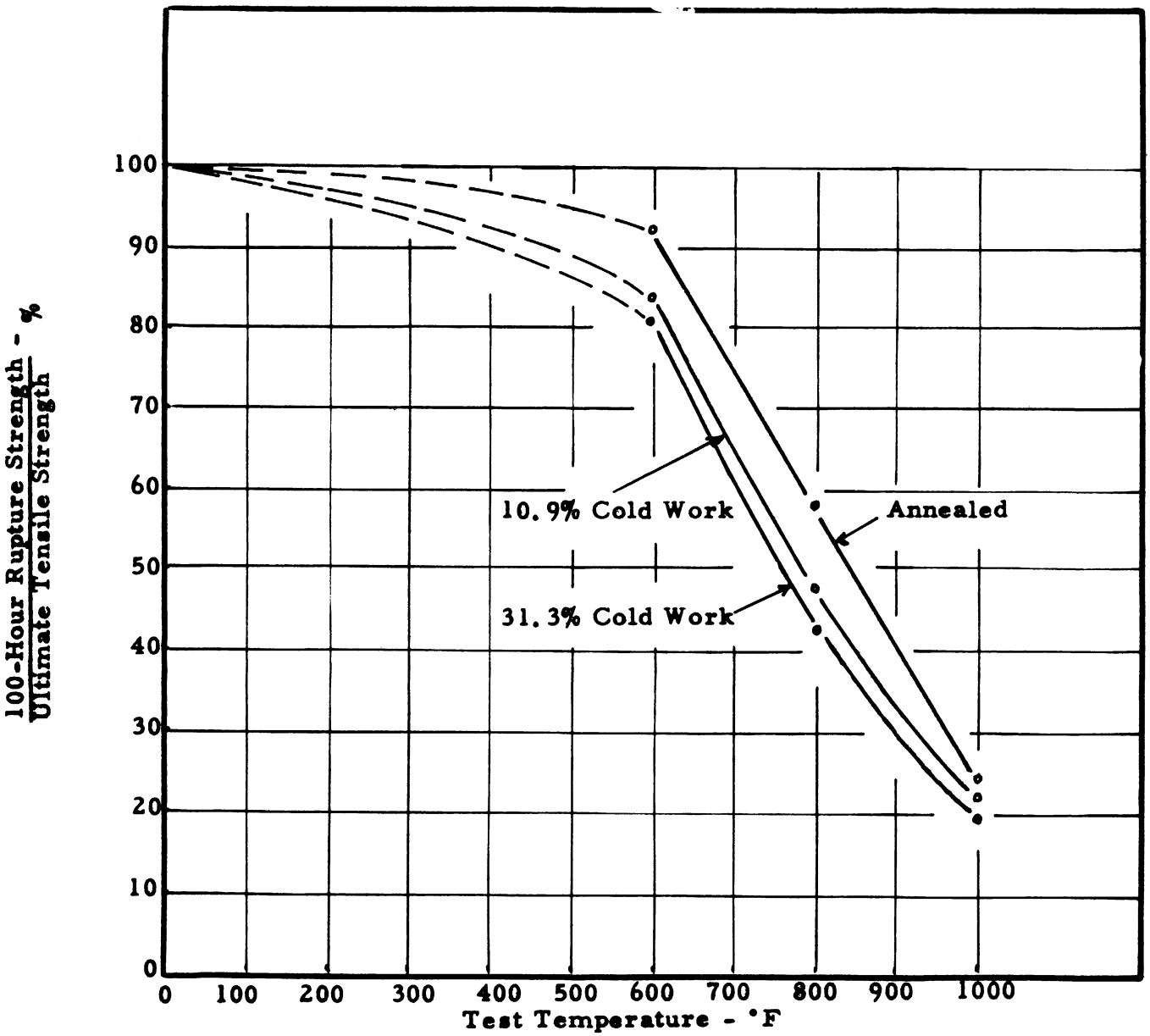


Fig. 8. Quotient 100-Hour Rupture Strength/Ultimate Tensile Strength Vs. Test Temperature for Various Conditions of Ti 75A

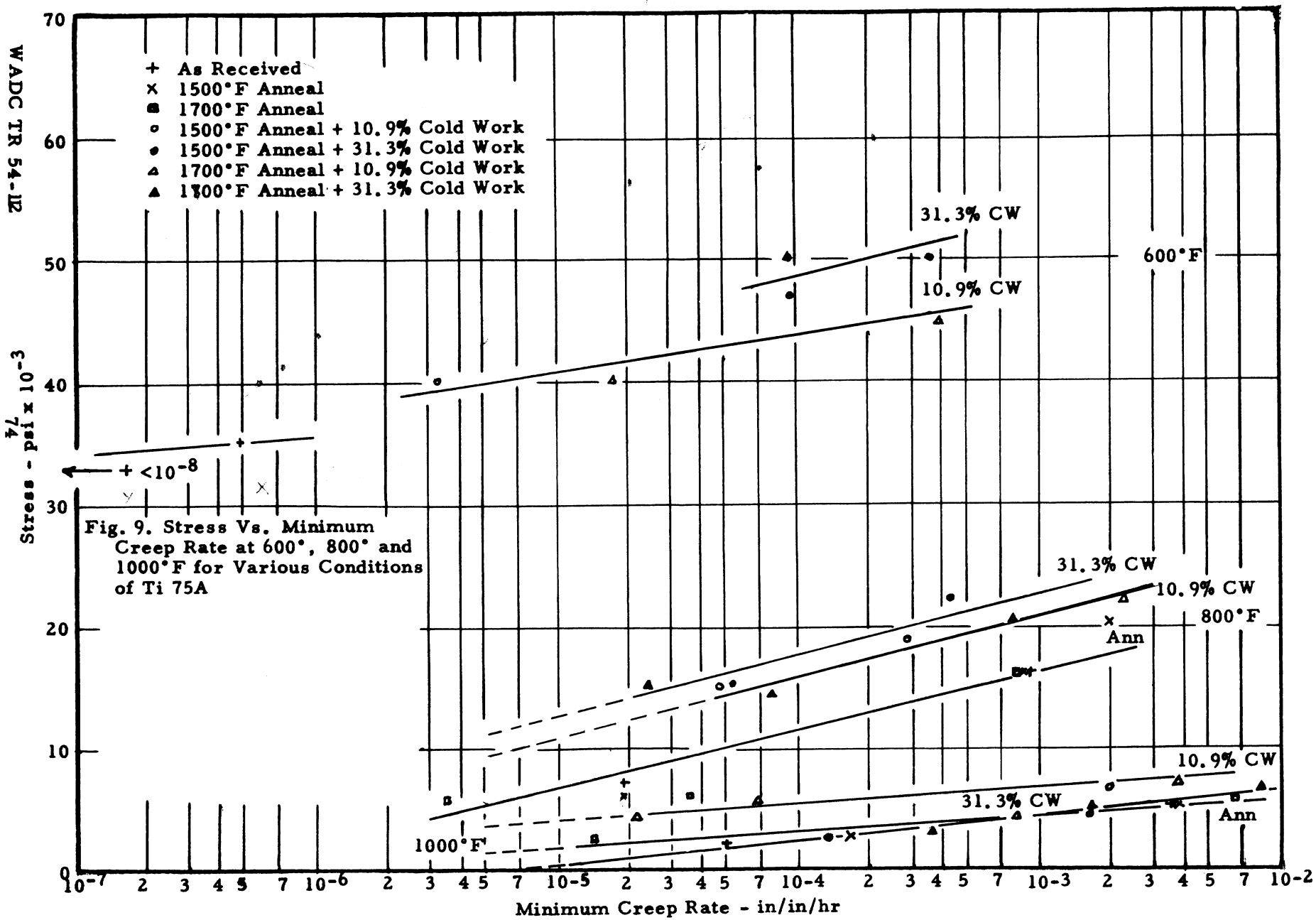


Fig. 9. Stress Vs. Minimum Creep Rate at 600°, 800° and 1000°F for Various Conditions of Ti 75A

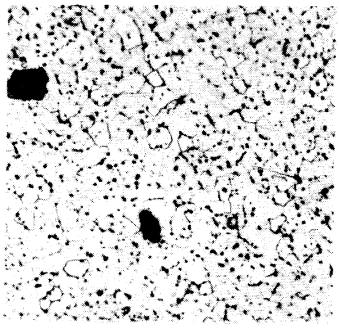


Fig. 10. Ti 75A As Received - X250D

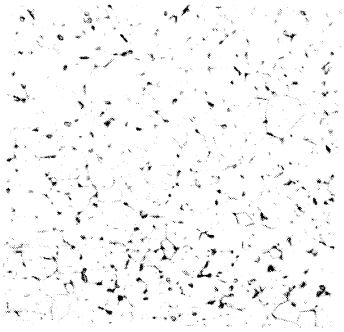


Fig. 11. Ti 75A Annealed 1 hr at 1500°F - X250D

Fig. 12. Ti 75A Annealed 1 hr at 1700°F - X250D

Fig. 13. Ti 75A Annealed 1 hr at 1700°F + 10.9% Cold Work - X250D

Fig. 14. Ti 75A Same Treatment as (12) + 27.8 hrs at 1000°F and 6,500 psi - X250D

Etchant:
2 HF, 2 HNO_3,
100 H_2O

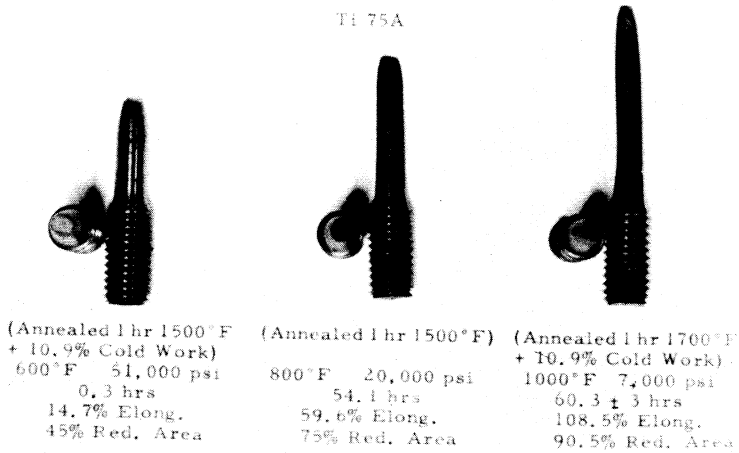


Fig. 15. Typical Rupture Fractures at 600°, 800°, and 1000°F for Ti 75A

Area reduced 21% in reproduction.

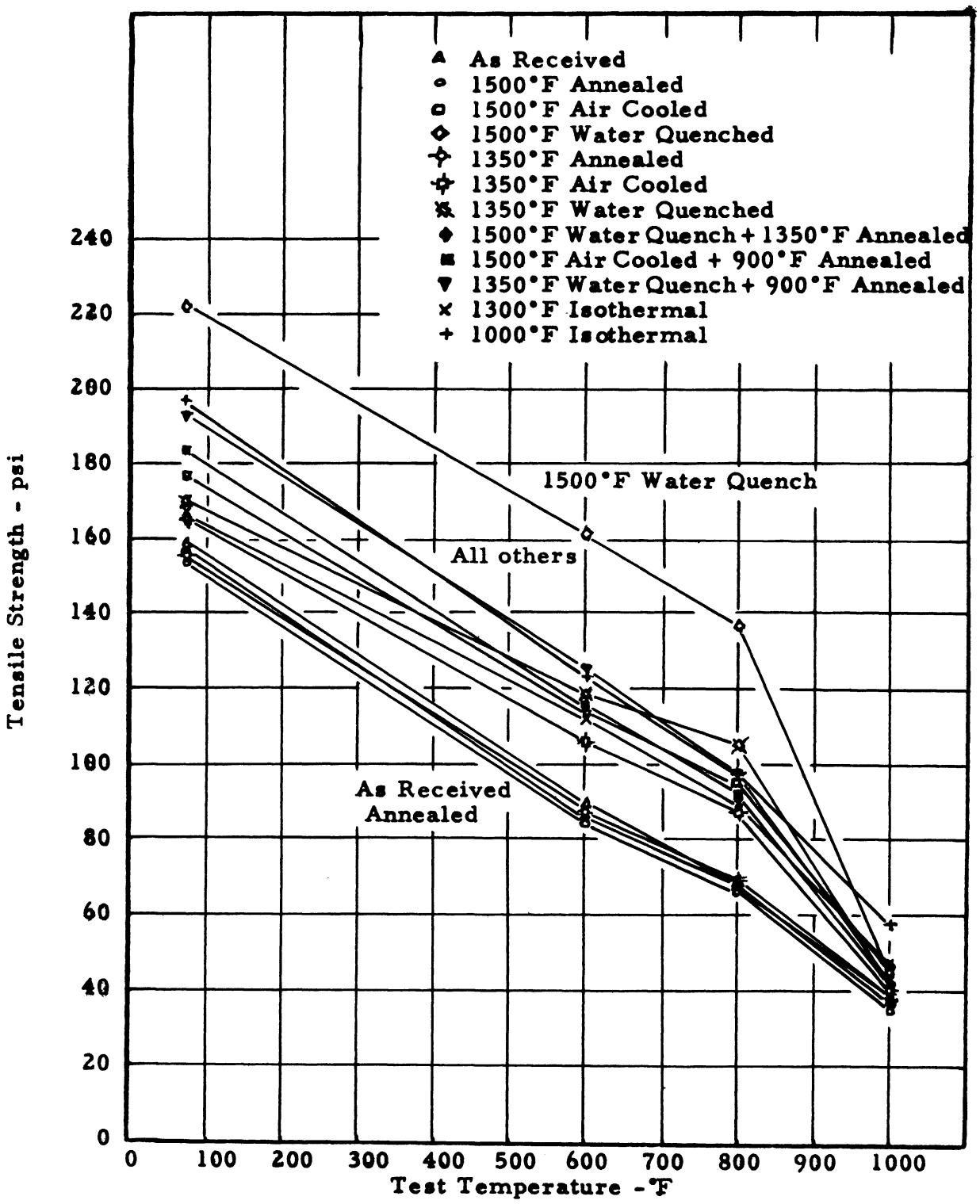


Fig. 16. Tensile Strength Vs. Test Temperature for Various Conditions of Ti 150A

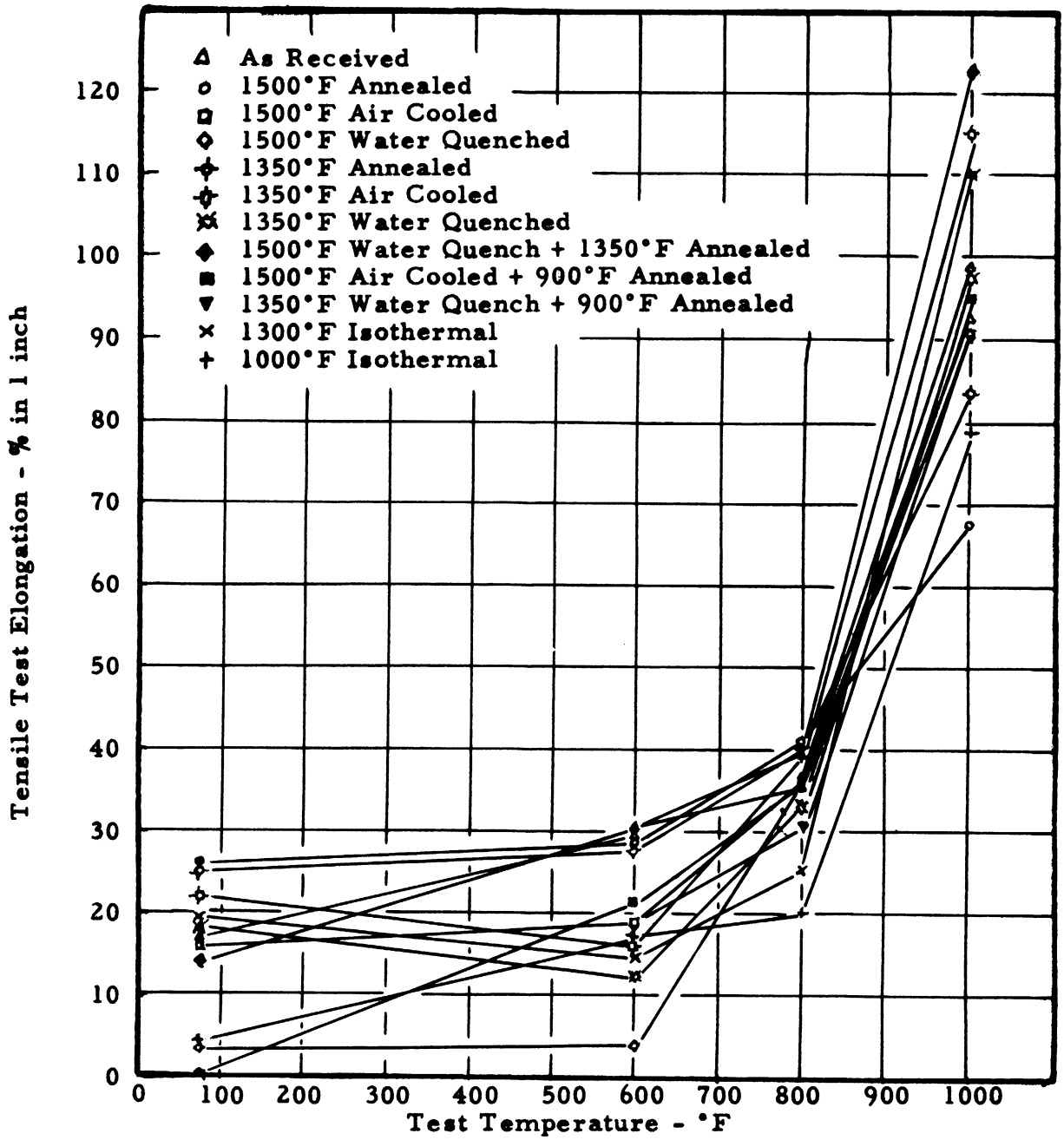


Fig. 17. Tensile Test Elongation Vs. Test Temperatures for Various Conditions of Ti 150A

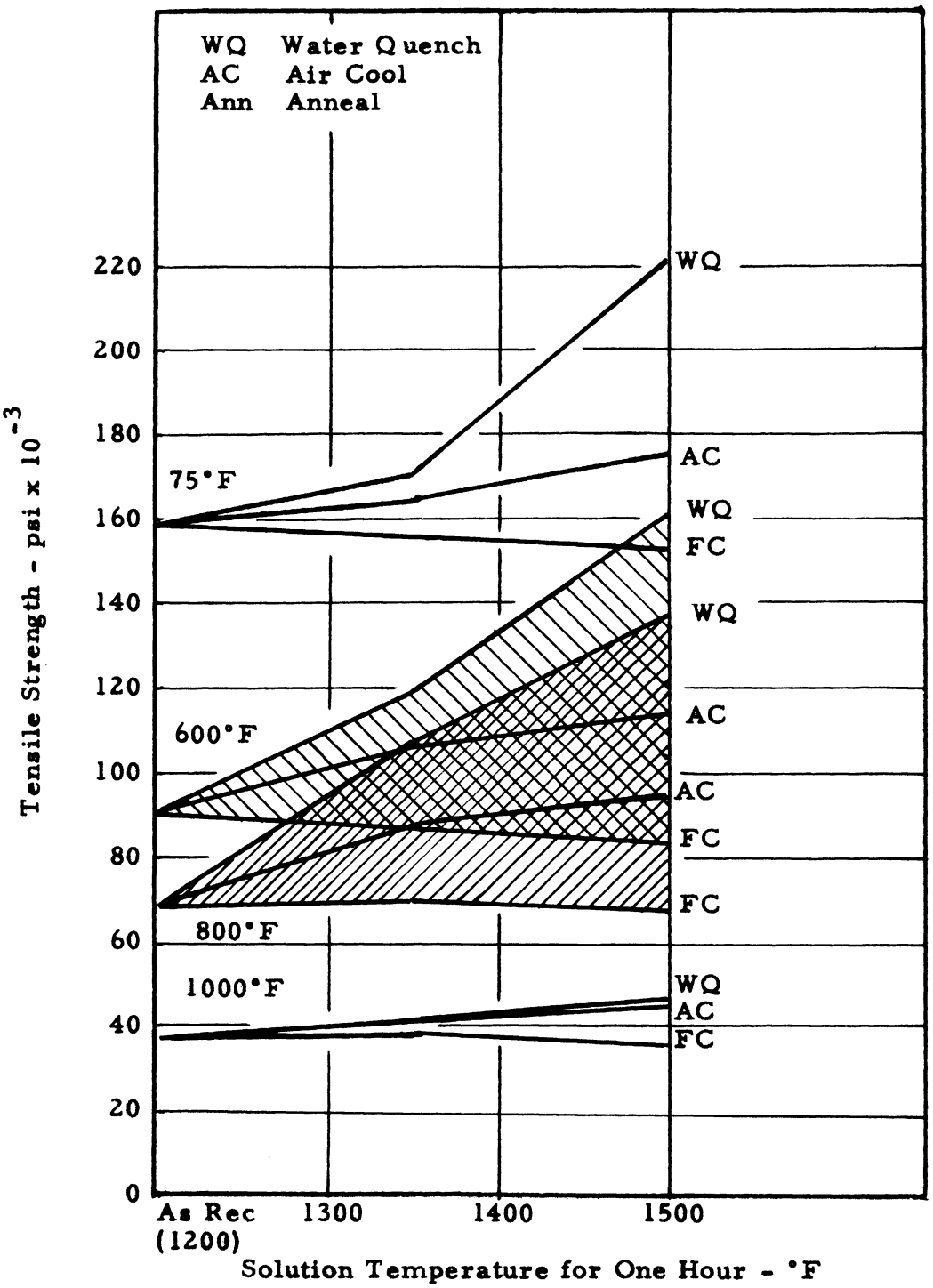


Fig. 18. Tensile Strength Vs. Solution Temperature of Ti 150A as a Function of Test Temperature and Cooling Rate

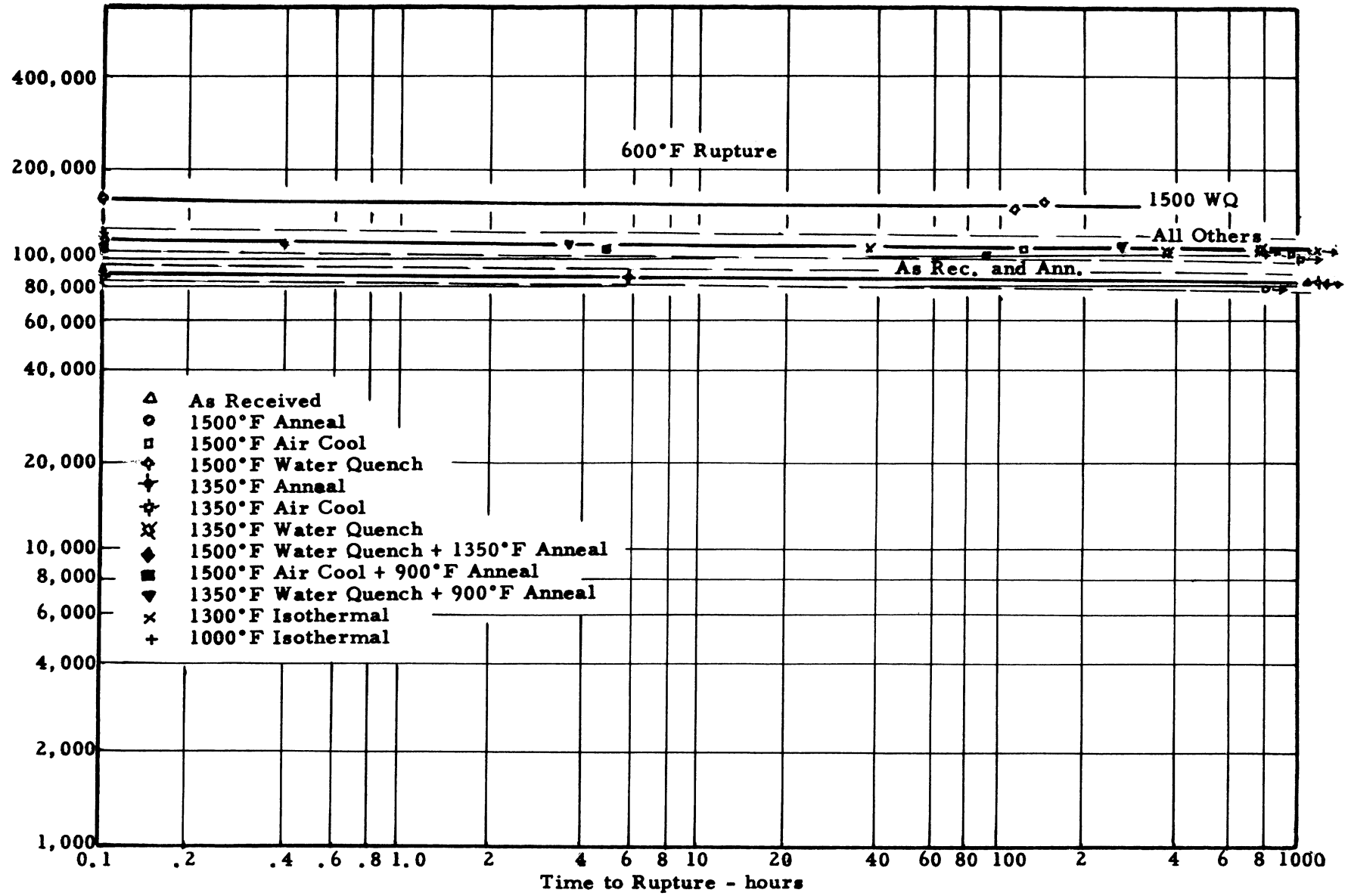


Fig. 19. Stress Vs. Time to Rupture at 600°F for Various Conditions of Ti 150A

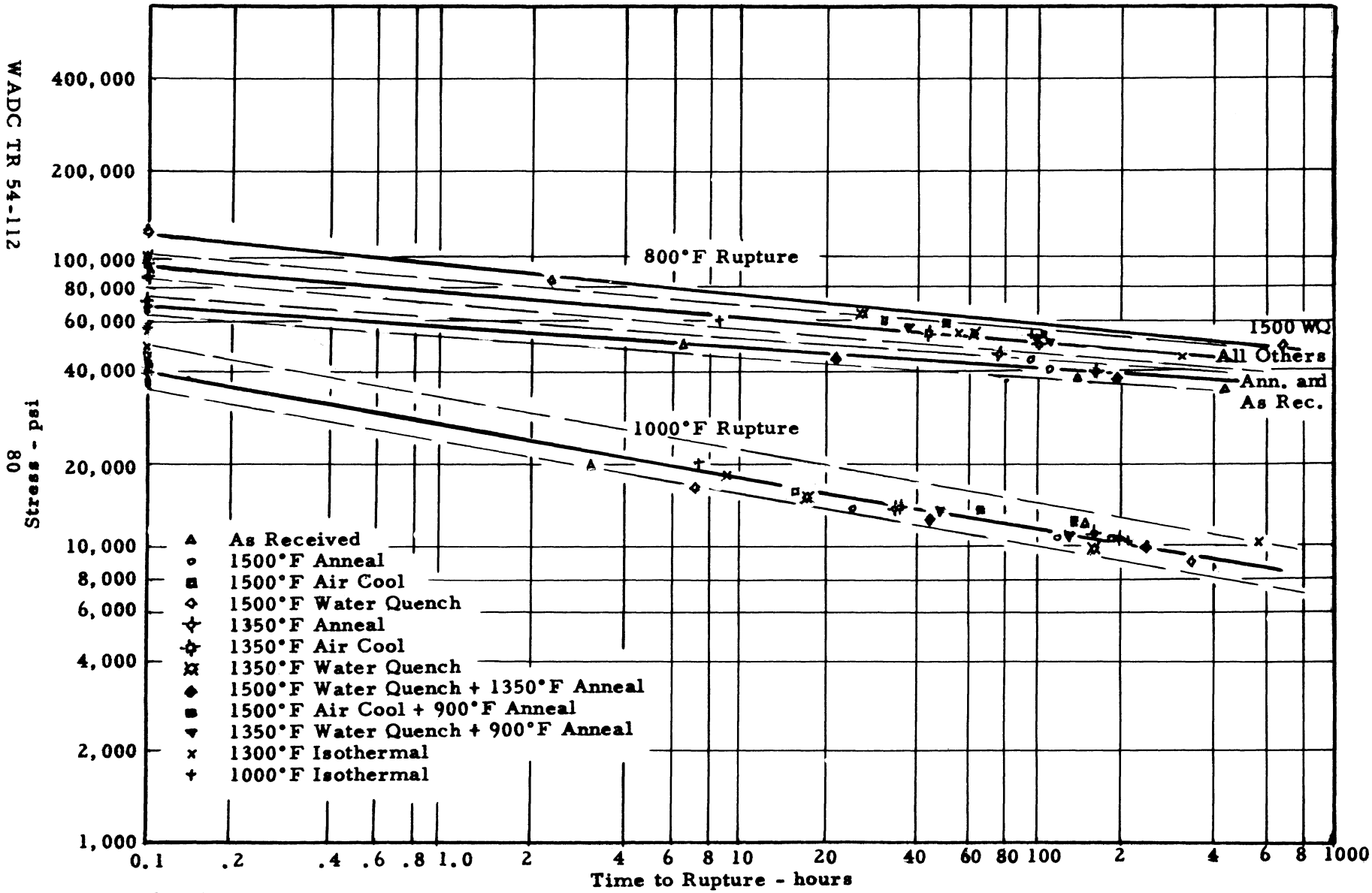


Fig. 20. Stress Vs. Time to Rupture at 800° and1000°F for Various Conditions of Ti 150A

Fig. 21. One-Hundred Hour Strength at 600°, 800° and 1000°F for All Conditions of Ti 150A

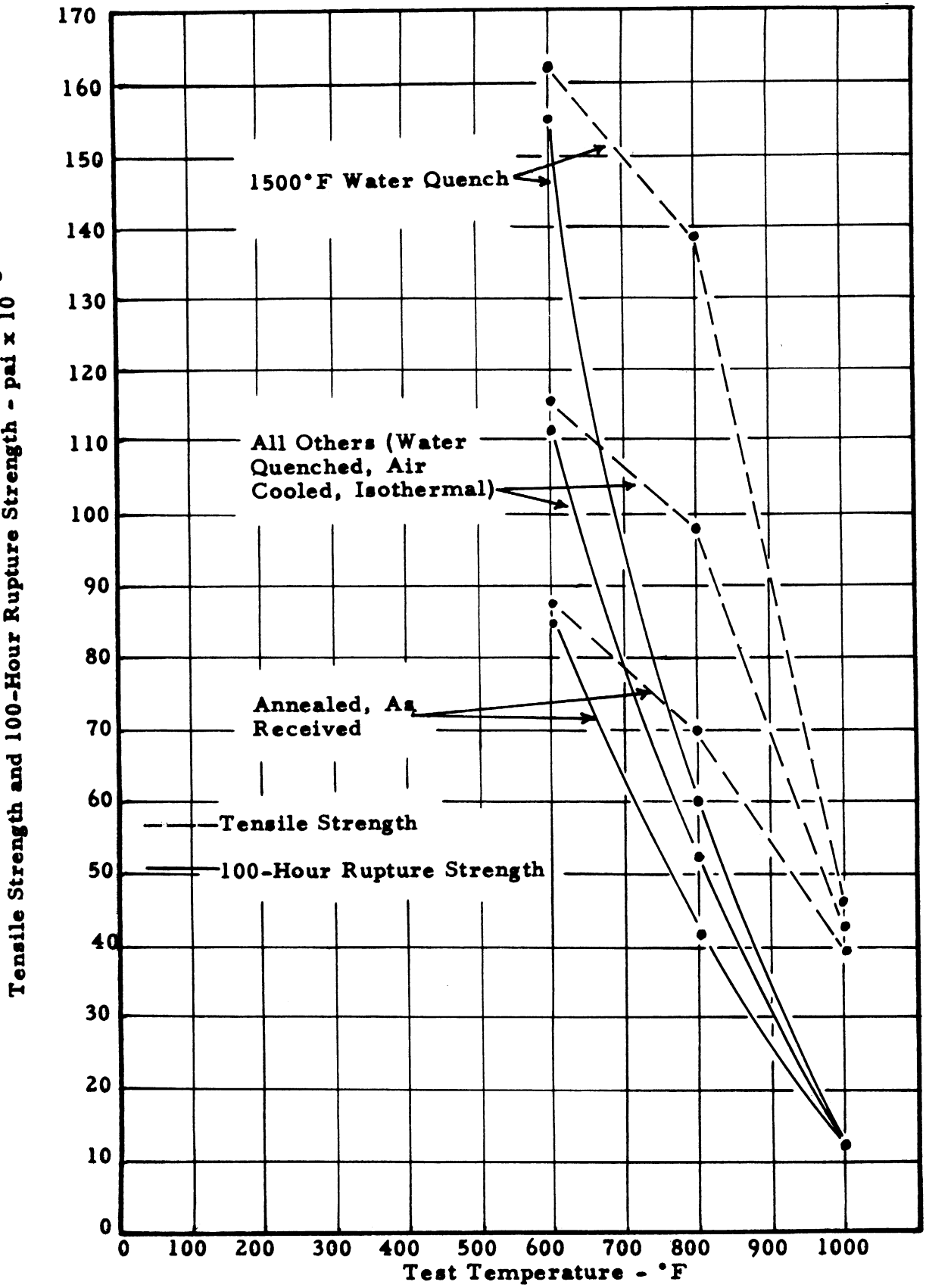


Fig. 22. Tensile and 100-Hour Rupture Strengths Vs. Temperature for Grouped Treatments of Ti 150A

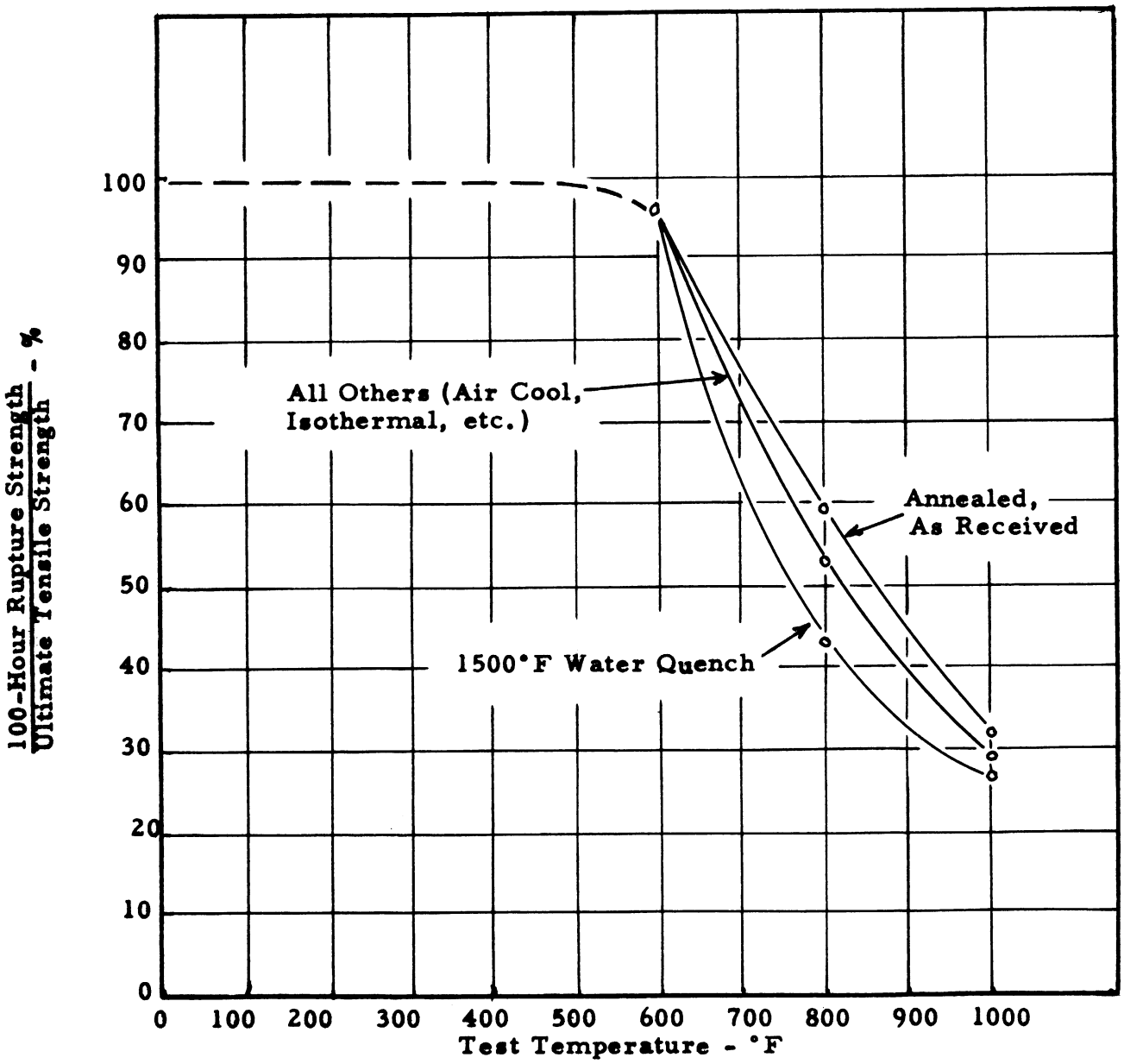


Fig. 23. Quotient 100-Hour Rupture Strength/Ultimate Tensile Strength for Various Conditions of Ti 150A

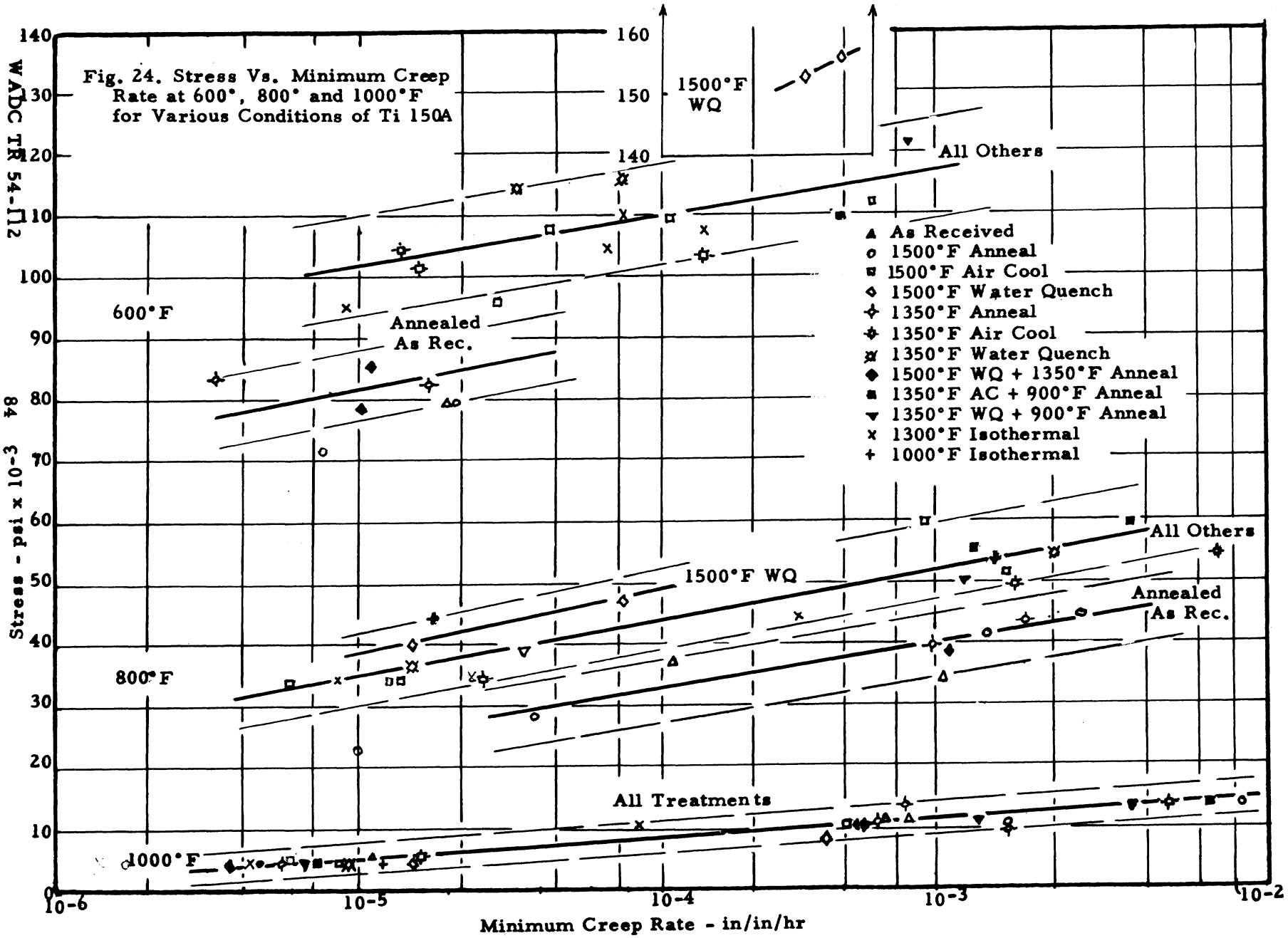


Fig. 24. Stress Vs. Minimum Creep Rate at 600°, 800° and 1000°F for Various Conditions of Ti 150A
Stress - psi x 10-3
Minimum Creep Rate - in/in/hr
600°F
800°F
1000°F
1500°F WQ
All Others
Annealed As Rec.
All Treatments
As Received
1500°F Anneal
1500°F Air Cool
1500°F Water Quench
1350°F Anneal
1350°F Air Cool
1350°F Water Quench
1500°F WQ + 1350°F Anneal
1350°F AC + 900°F Anneal
1350°F WQ + 900°F Anneal
1300°F Isothermal
1000°F Isothermal

1350°F Solution Temperature

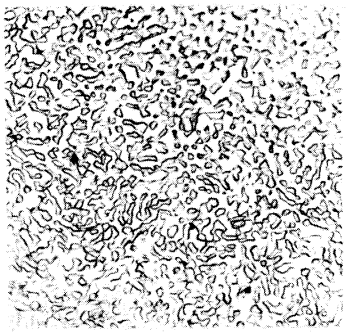


Fig. 26. Ti 150A Annealed 1 hr at 1350°F - X500D

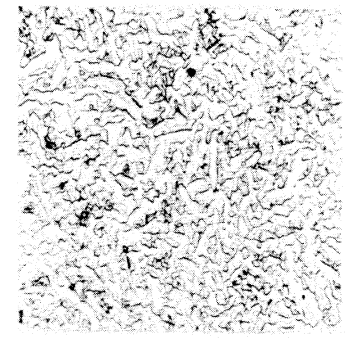


Fig. 27. Ti 150A 1 hr at 1350°F + Air Cool - X500D

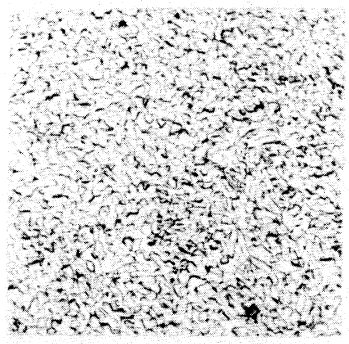


Fig. 28. Ti 150A 1 hr at 1350°F + Water Quench - X500D

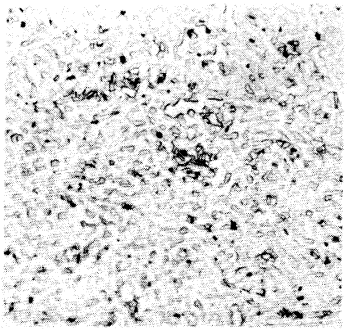


Fig. 25. Ti 150 As Received - X500D

Isothermal Treatments

Fig. 32. Ti 150A 1 hr at 1800°F + 1 hr Iso. Trans. at 1300°F + Water Quench - X500D

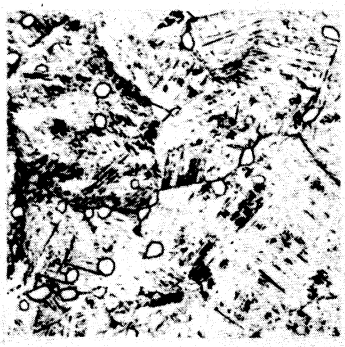


Fig. 33. Ti 150A 1 hr at 1800°F + 1 hr Iso. Trans. at 1000°F + Water Quench - X500D

Etchant: 1 HF, 1 Glycerine

1500°F Solution Temperature

Fig. 29. Ti 150A Annealed 1 hr at 1500°F - X500D

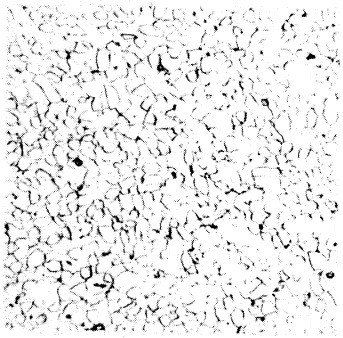


Fig. 30. Ti 150A 1 hr at 1500°F + Air Cool - X500D

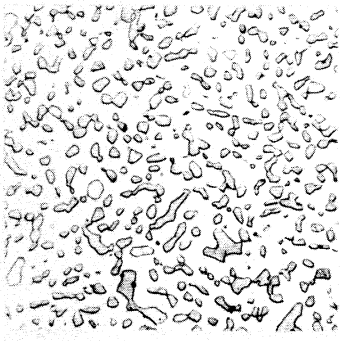


Fig. 31. Ti 150A 1 hr at 1500°F + Water Quench - X500D

Area reduced 21% in reproduction

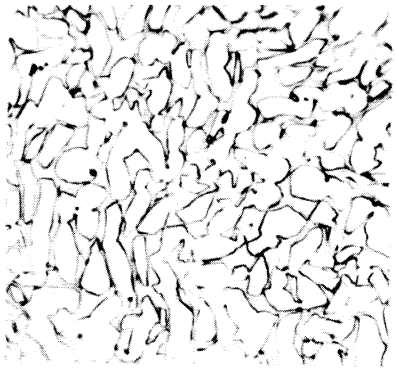


Fig. 34. Ti 150A 1500°F Anneal + Tensile Test at 800°F - X1000D

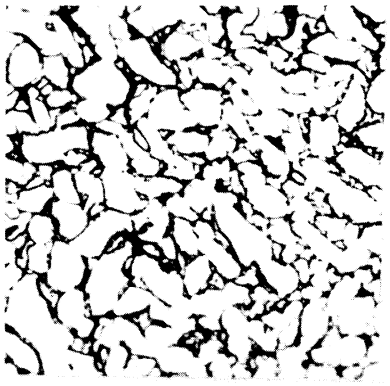


Fig. 35. Ti 150A 1500°F Anneal + 116 hrs at 1000°F and 11,000 psi - X1000D

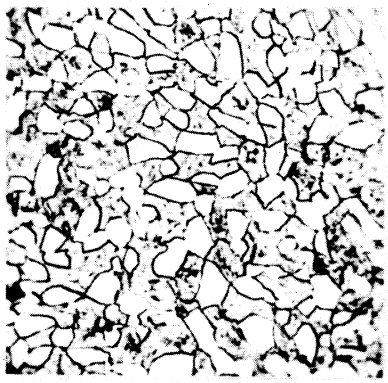


Fig. 36. Ti 150A 1500°F Air Cool + Tensile Test at 1000°F - X1000D

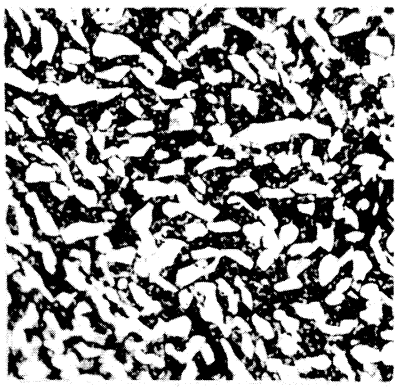


Fig. 37. Ti 150A 1500°F Air Cool + 16.1 hrs at 1000°F and 16,000 psi - X1000D

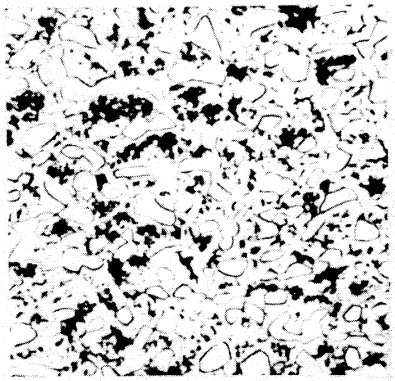


Fig. 38. Ti 150A 1500°F Water Quench + Tensile Test at 1000°F - X1000D

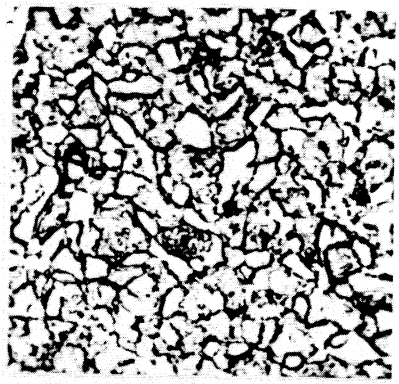


Fig. 39. Ti 150A 1500°F Water Quench + 348.8 hrs at 1000°F and 9,000 psi - X1000D

Etchant: 1 HF, 1 Glycerine

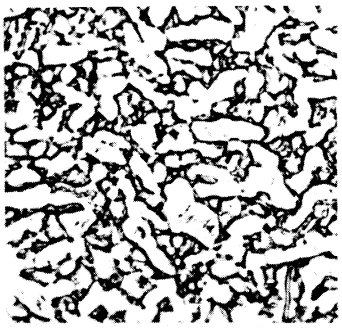


Fig. 40. Ti 150A 1350°F Anneal + 34 hrs at 1000°F and 14,000 psi - X1000D

Fig. 41. Ti 150A 1350°F Air Cool + 194.3 hrs at 1000°F and 11,000 psi - X1000D

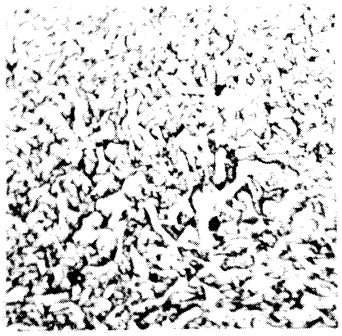


Fig. 42. Ti 150A 1350°F Water Quench + 16.7 hrs at 1000°F and 15,000 psi - X1000D

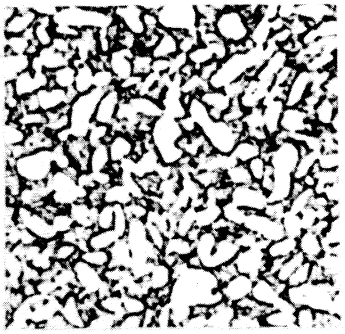
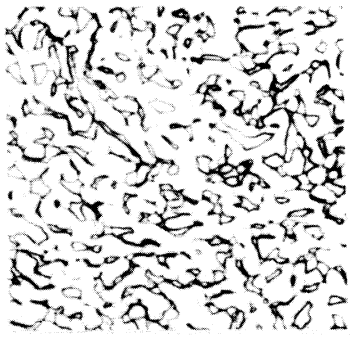


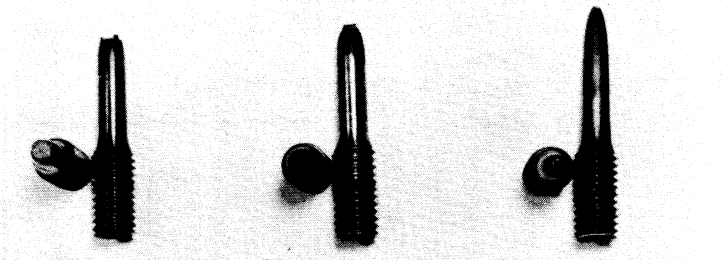
Fig. 43. Ti 150A 1500°F Water Quench + 1350°F Anneal + 43 hrs at 1000°F and 13,000 psi - X1000D



Etchant: 1 HF, 1 Glycerine

Fig. 44. Ti 150A 1350°F Water Quench + 900°F Anneal + 37 hrs at 800°F and 58,000 psi - X1000D

Ti 150A



(1500°F 1 hr Air Cooled)	(1500°F 1 hr Water Quenched)	(Annealed 1 hr 1350°F)
600°F 112,000 psi	800°F 48,000 psi	1000°F 12,000 psi
128.8 hrs	659.2 hrs	157 ± 4 hrs
22.8% Elong	27% Elong.	62% Elong.
50.4% Red. Area	84% Red. Area	96% Red. Area

Fig. 45. Typical Rupture Fractures at 600°, 800°, and 1000°F for Ti 150A

Area reduced 21% in reproduction

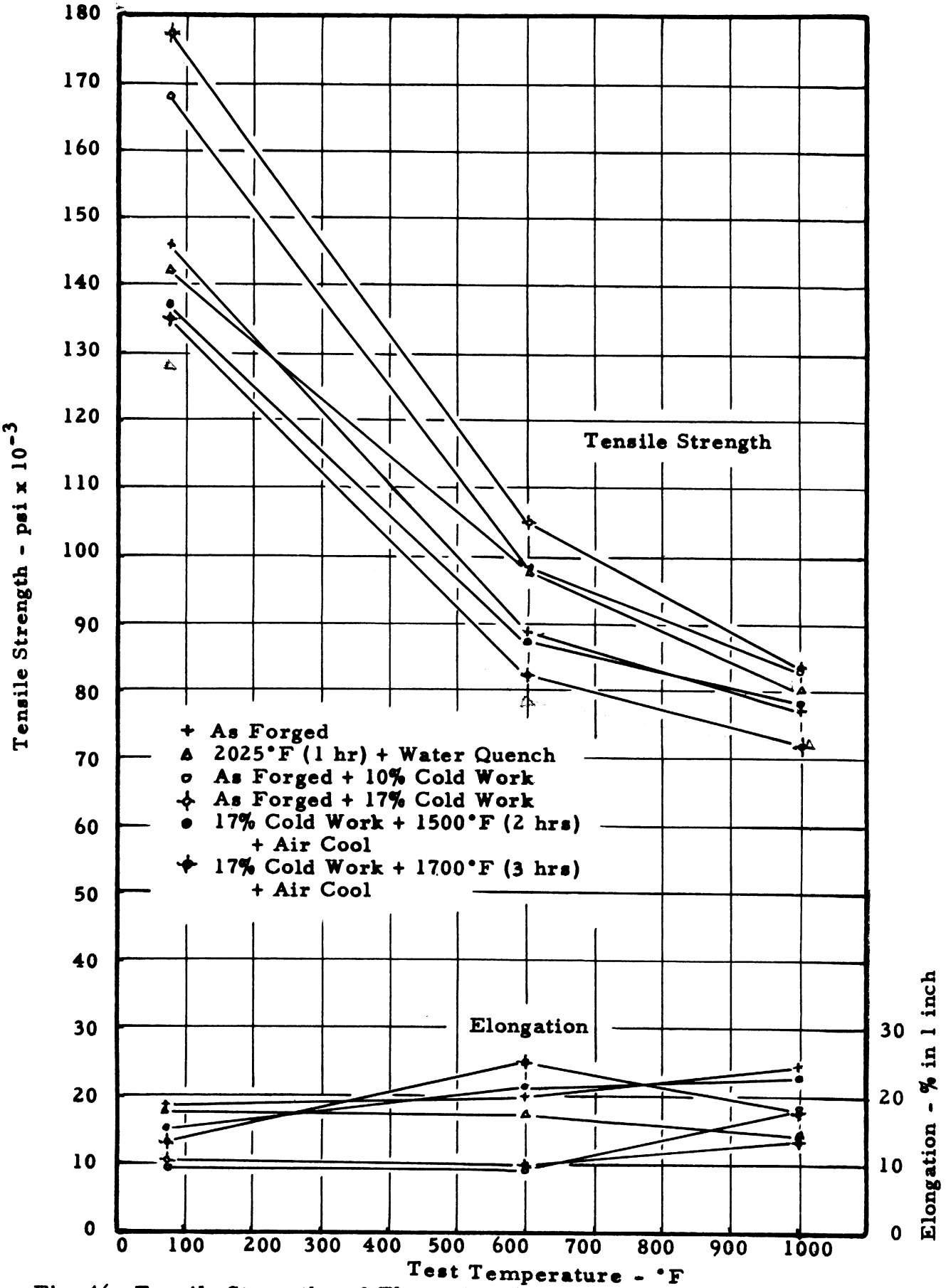


Fig. 46. Tensile Strength and Elongation Vs. Test Temperature for Various Conditions of the α Alloy - 6% Al

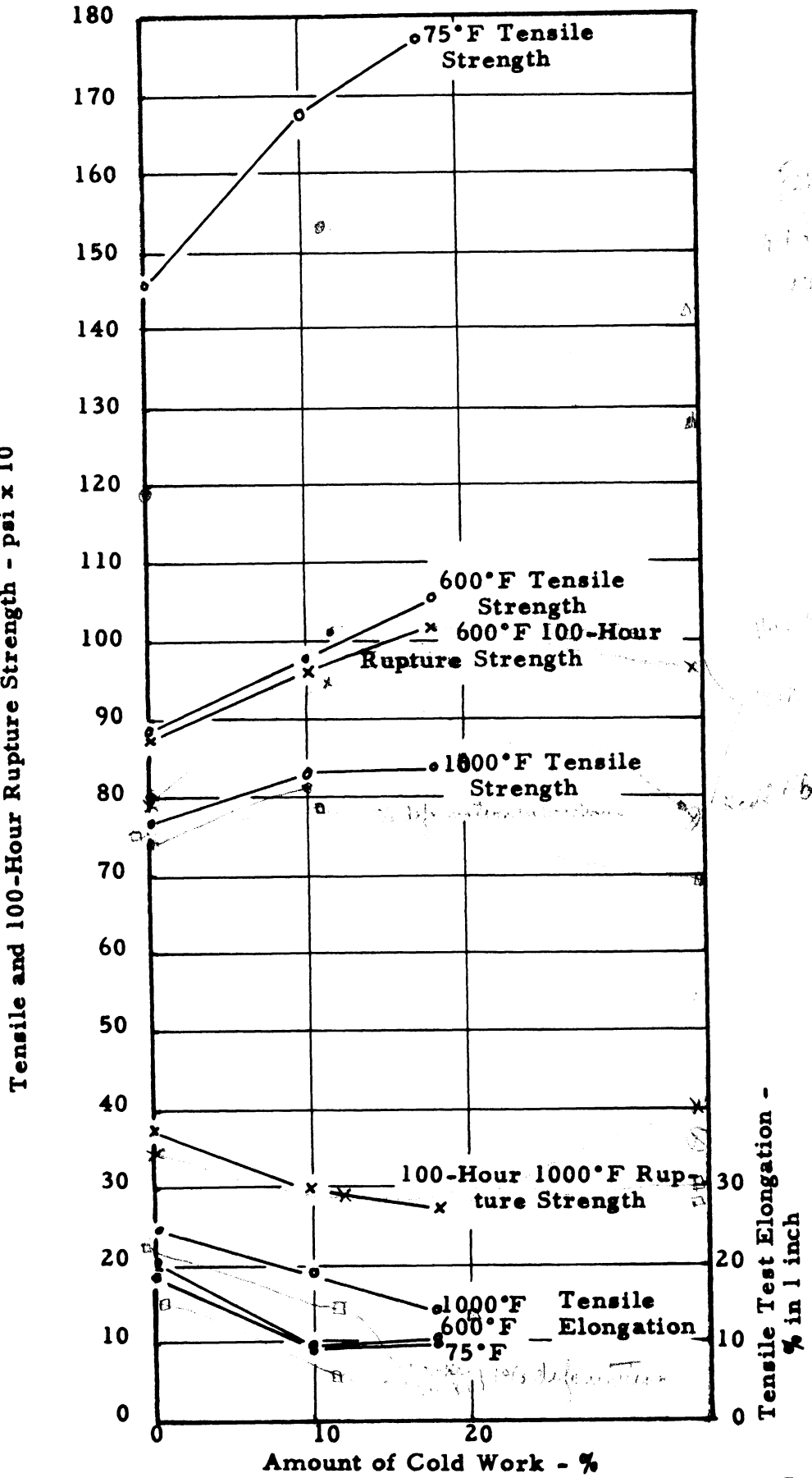


Fig. 47. Effect of Cold Work on Tensile and 100-Hour Rupture Strengths at 600° and 1000°F of the *a* Alloy - 6% Al

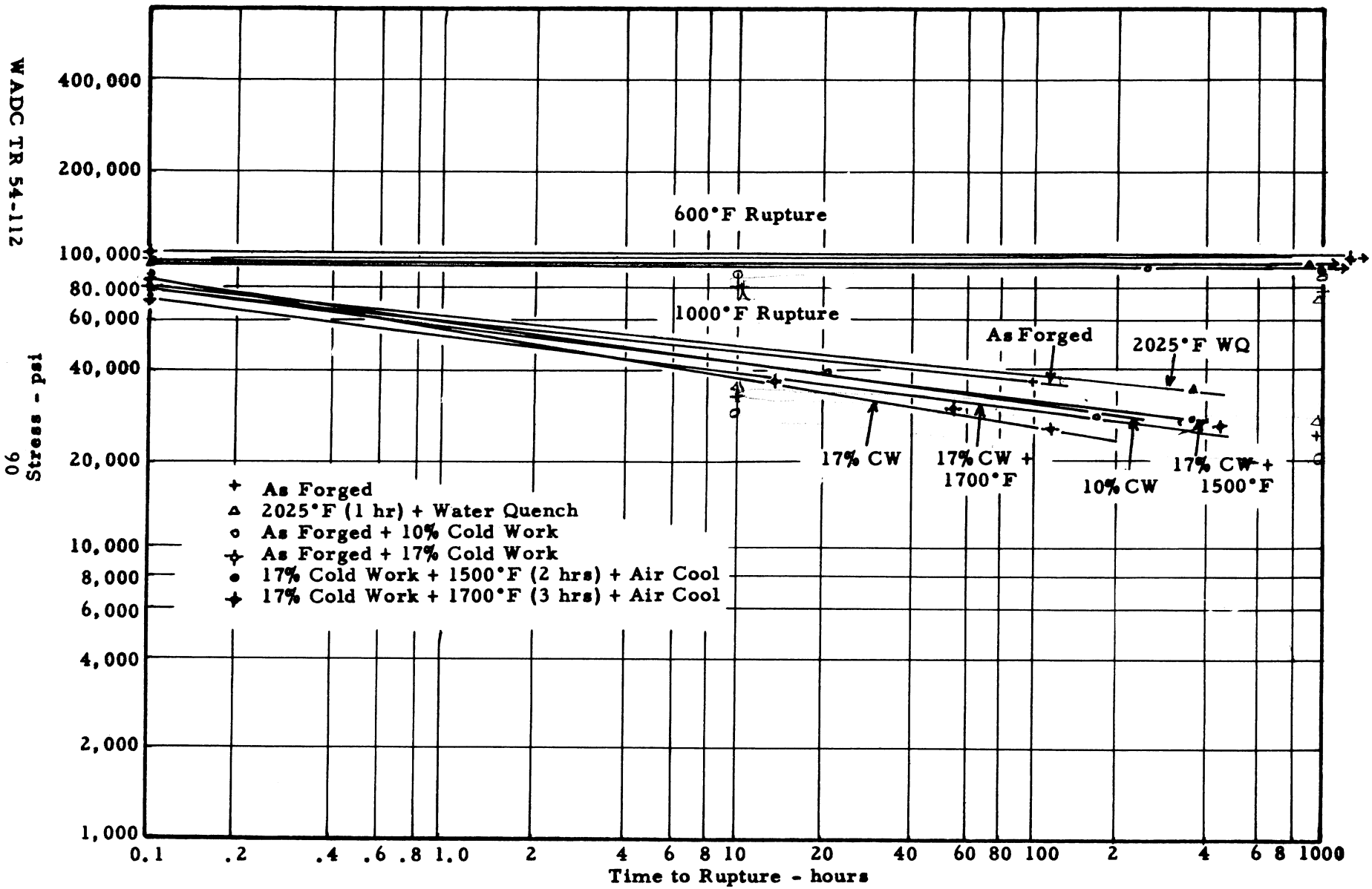


Fig. 48. Stress Vs. Time to Rupture at 600° and 1000°F for Various Conditions of the α Alloy - 6% Al

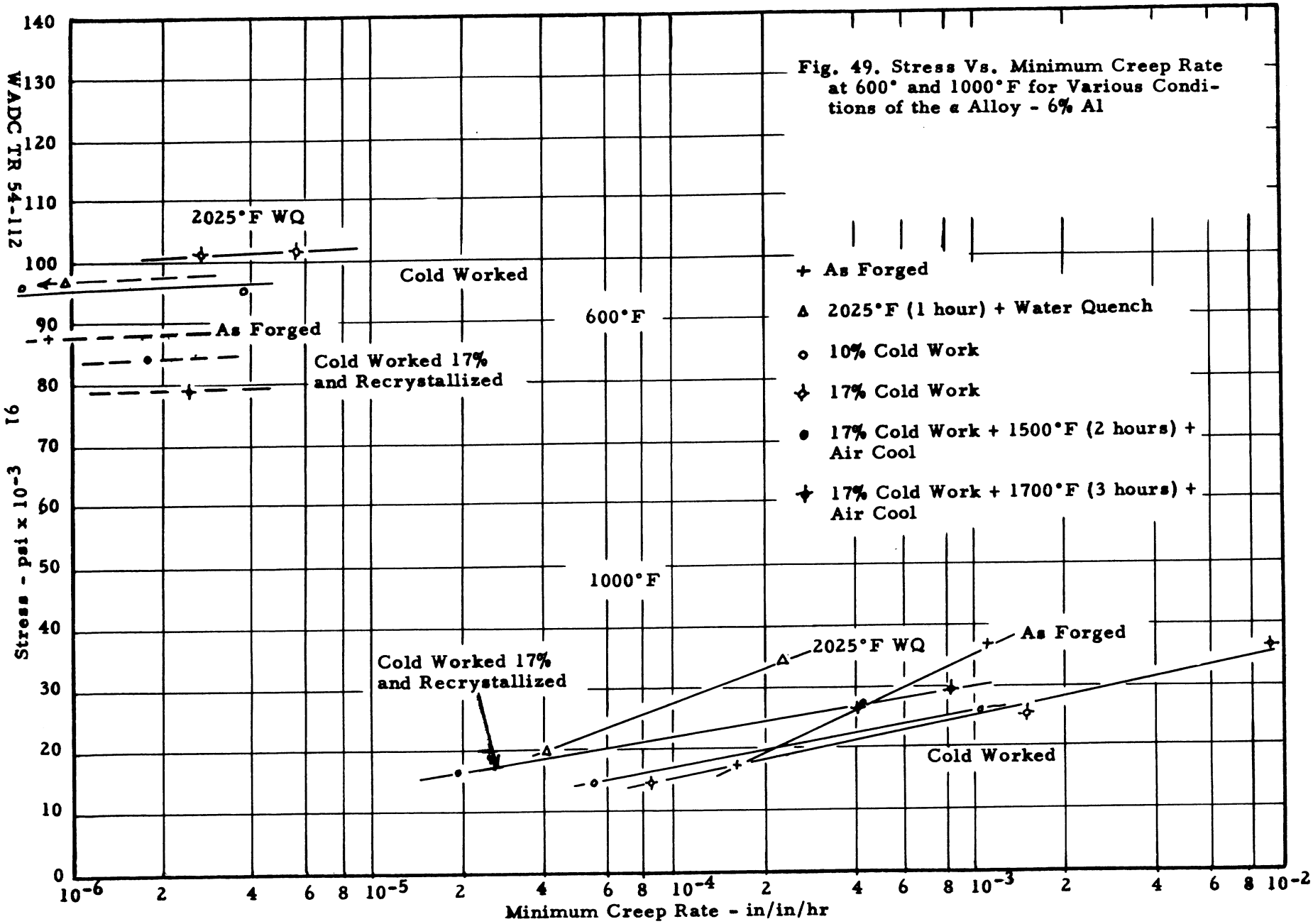


Fig. 49. Stress Vs. Minimum Creep Rate at 600° and 1000°F for Various Conditions of the α Alloy - 6% Al

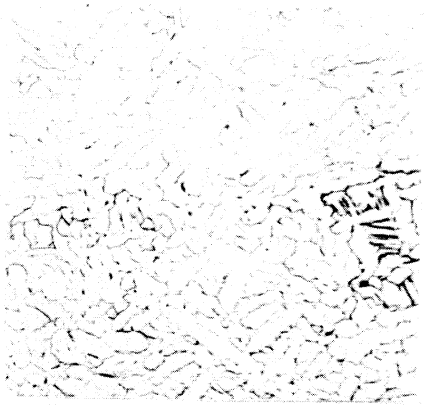


Fig. 50. 6% Al α Alloy
As Forged - X250D

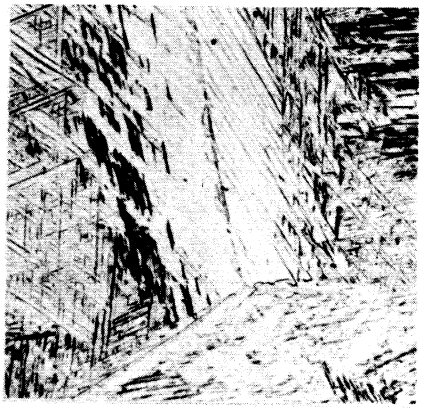


Fig. 51. 6% Al α Alloy
1 hr 2025°F + Water
Quench - X250D

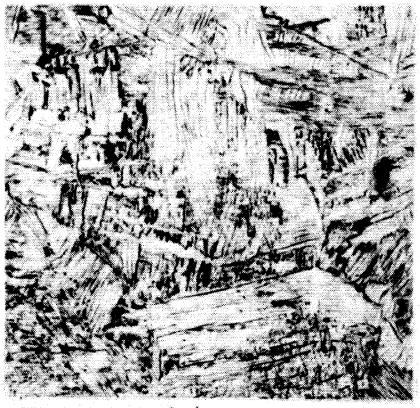


Fig. 52. 6% Al α Alloy
2025°F Water Quench +
Tensile Test at 1000°F
- X100D

Etchant: 2 HF, 2 HNO_3, 100 H_2O

Area enlarged 18% in reproduction

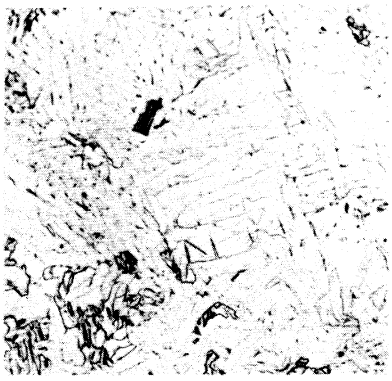


Fig. 53. 6% Al α Alloy As Forged + 10.9% Cold Work - X250D

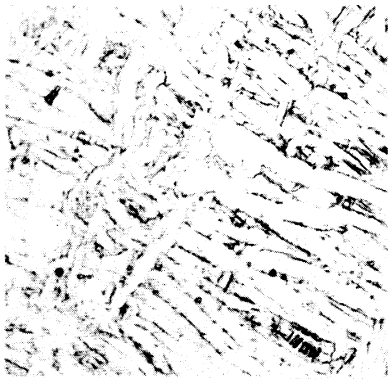


Fig. 54. 6% Al α Alloy 10% Cold Work + 600° Tensile Test - X250D

Fig. 55. 6% Al α Alloy 10% Cold Work + 1216 hrs at 600° F and 96,500 psi - X250D

Fig. 56. 6% Al α Alloy 10% Cold Work + Tensile Test at 1000° F - X250D

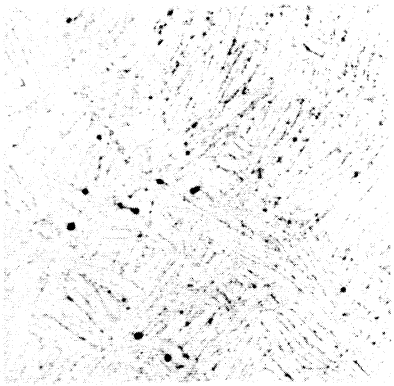


Fig. 57. 6% Al α Alloy 10% Cold Work + 20.5 hrs at 1000° F and 40,000 psi - X250D

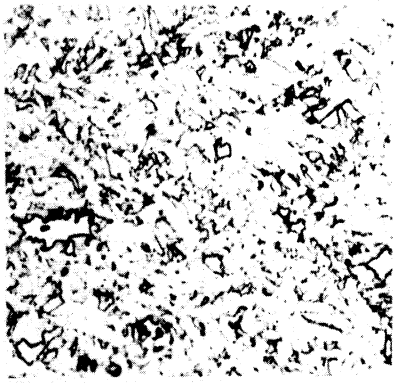


Fig. 58. 6% Al α Alloy 10% Cold Work + 176.4 hrs at 1000° F and 26,800 psi - X250D

Etchant: 2 HF, 2 HNO_3, 100 H_2O

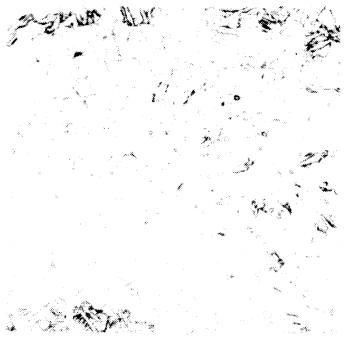


Fig. 59. 6% Al α Alloy As Forged + 17% Cold Work - X250D

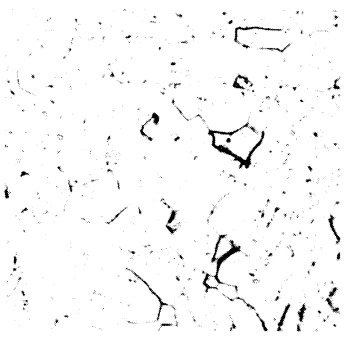


Fig. 60. 6% Al α Alloy 17% Cold Work + 2 hrs at 1500°F + Air Cool - X250D

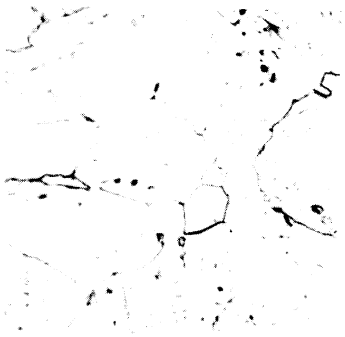


Fig. 61. 6% Al α Alloy 17% Cold Work + 3 hrs at 1700°F + Air Cool - X250D

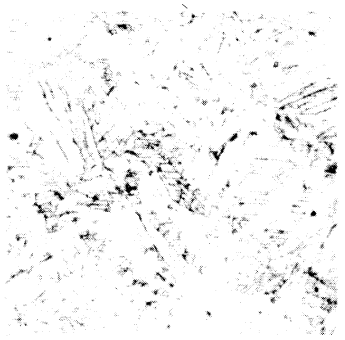


Fig. 62. 6% Al α Alloy 17% Cold Work + 1172.2 hrs at 600°F and 101,000 psi -X250D

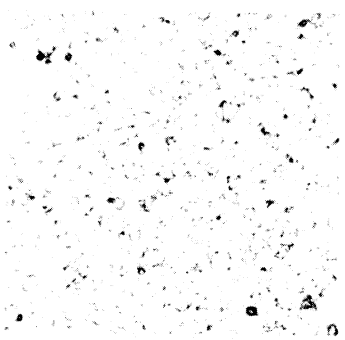


Fig. 63. 6% Al α Alloy 17% Cold Work + 13.4 hrs at 1000°F and 37,000 psi - X250D

Etchant:
2 HF, 2 HNO_3,
100 H_2O

6% Al - 94% Ti

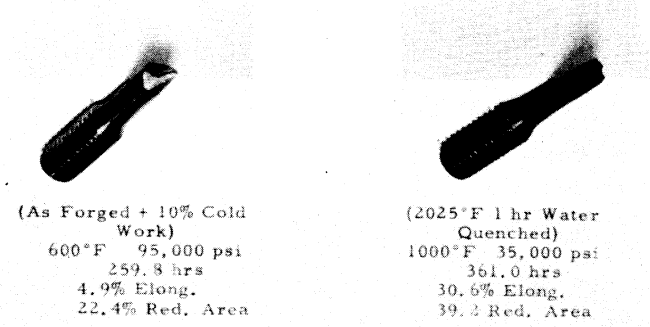


Fig. 64. Typical Rupture Fractures at 600° and 1000°F for 6% Al α Alloy

Area reduced 21% by reproduction

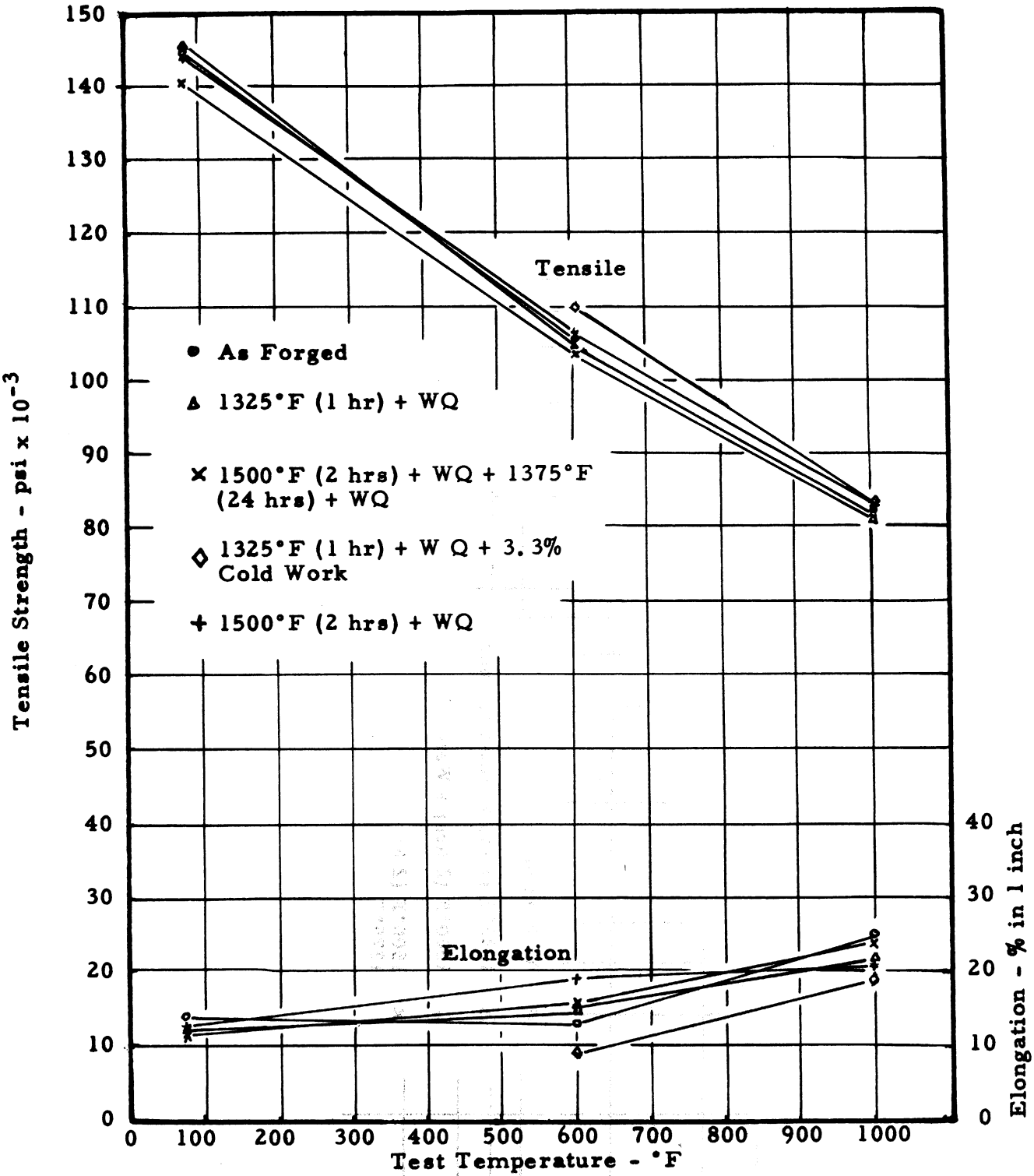


Fig. 65 Tensile Strength and Elongation Vs. Temperature for Various Conditions of the β Alloy - 30% Mo

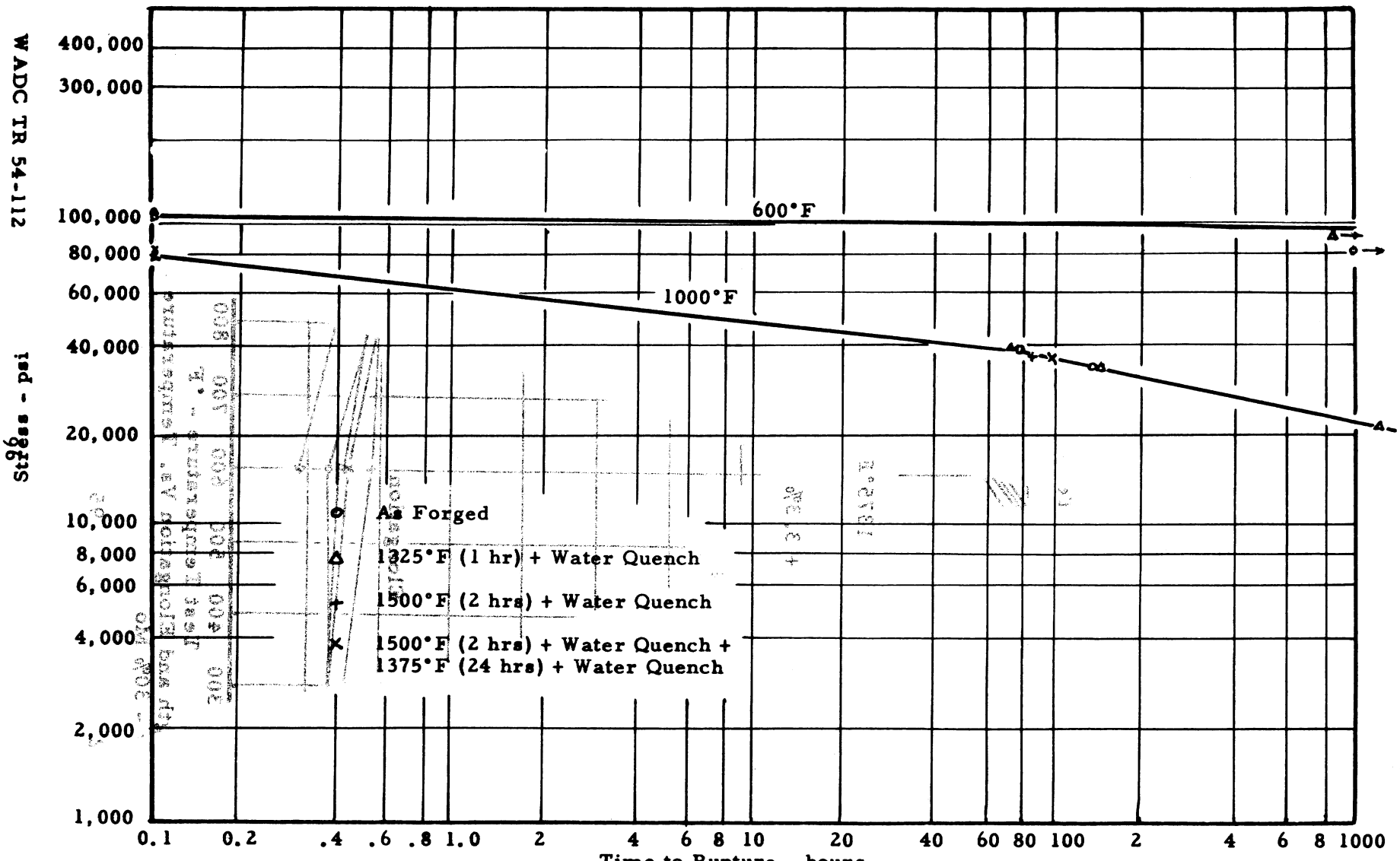


Fig. 66. Stress Vs. Time for Rupture at 600° and 1000°F for the β Alloy - 30% Mo

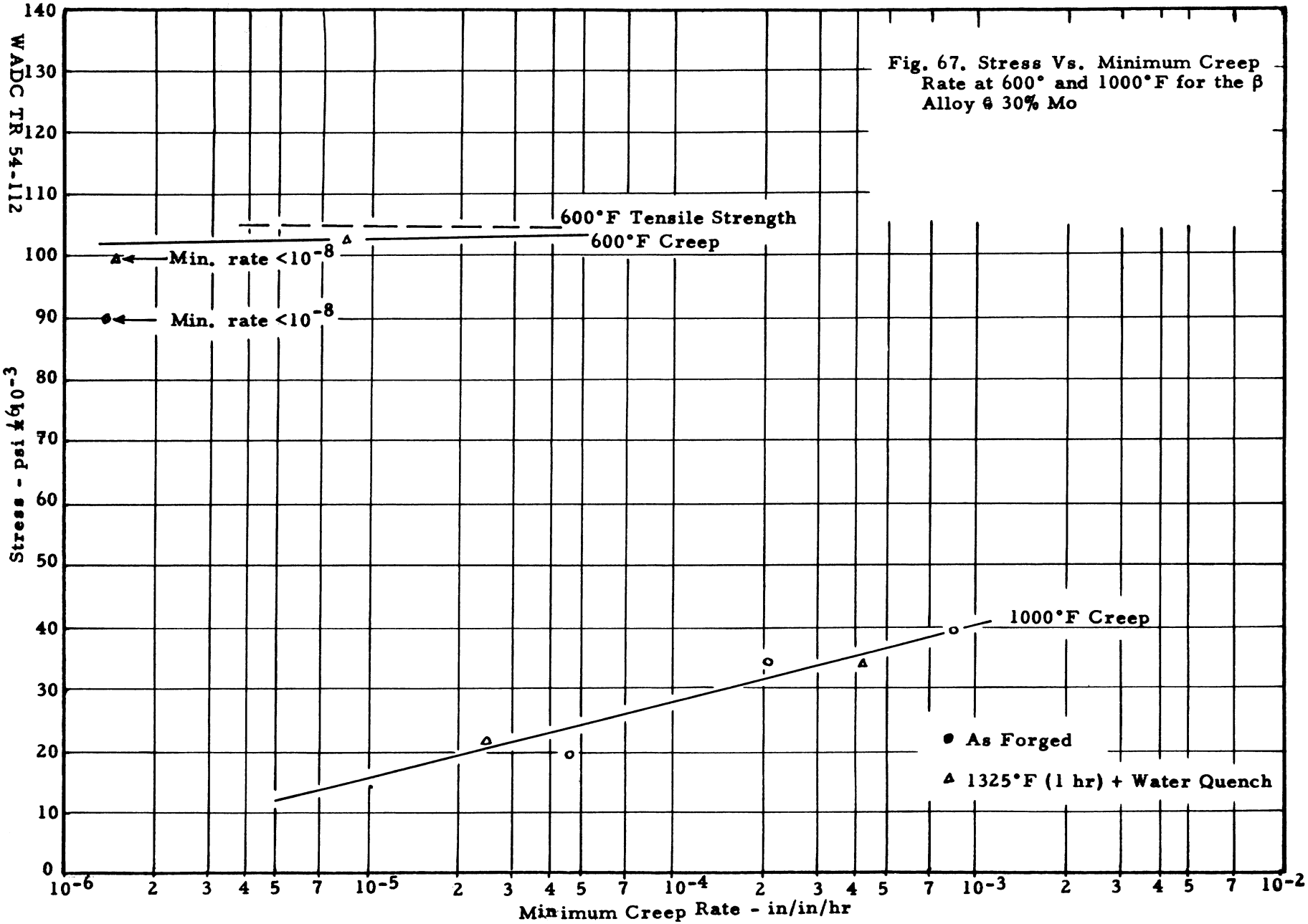


Fig. 67. Stress Vs. Minimum Creep Rate at 600° and 1000°F for the β Alloy @ 30% Mo
600°F Tensile Strength
600°F Creep
Min. rate <10^-8
Min. rate <10^-8
1000°F Creep
As Forged
1325°F (1 hr) + Water Quench
Stress - psi x 10^-3
Minimum Creep Rate - in/in/hr
140
130
120
110
100
90
80
70
60
50
40
30
20
10
0
10^-6
2
3
4
5
7
10^-5
2
3
4
5
7
10^-4
2
3
4
5
7
10^-3
2
3
4
5
7
10^-2

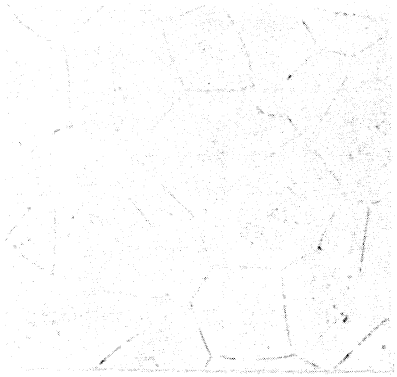


Fig. 68. 30% Mo β Alloy As Forged - X250D

Fig. 69. 30% Mo β Alloy 1 hr 1325°F + Water Quench - X250D

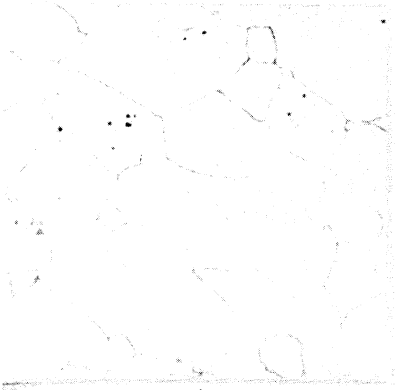


Fig. 70. 30% Mo β Alloy As Forged + 77 hrs at 1000°F and 40,000 psi - X250D

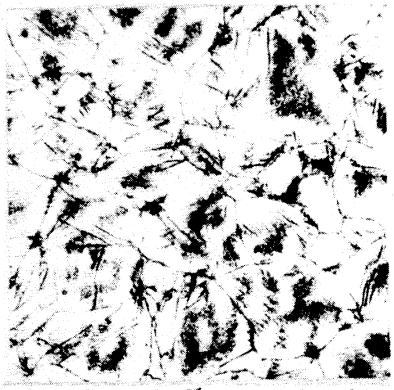


Fig. 71. 30% Mo β Alloy As Forged + 133 hrs at 1000°F and 35,000 psi - X250D

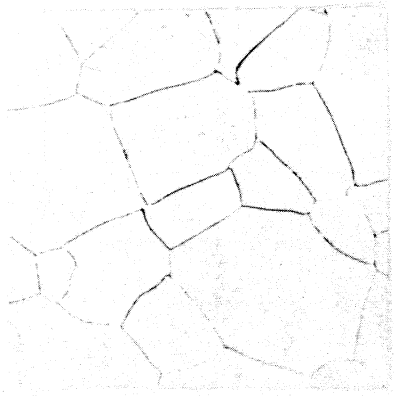


Fig. 72. 30% Mo β Alloy 1325°F Water Quench + 836 hrs at 600°F and 100,000 psi - X250D

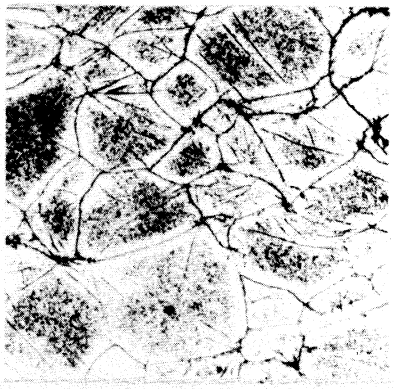


Fig. 73. 30% Mo β Alloy 1325°F Water Quench + 141.2 hrs at 1000°F and 35,000 psi - X250D

Etchant: 1 HF, 1 Glycerine

Fig. 74. 30% Mo β Alloy
2 hrs 1500°F + Water
Quench + Tensile Test
at 1000°F - X250D

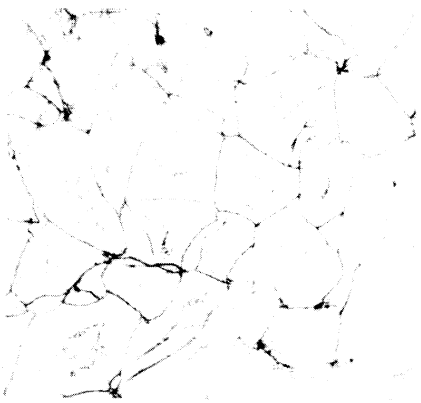
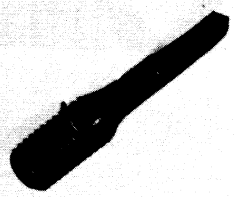


Fig. 75. 30% Mo β Alloy
1500°F Water Quench +
83.4 hrs at 1000°F and
37,500 psi - X250D

Etchant: 1 HF, 1 Glycerine

30% Mo - 70% Ti

(1 hr 1325°F + Water
Quench)
600°F 100,000 psi
>836.0 hrs
15.5% Elong.
43.0% Red. Area
(broken accidentally)



(1 hr 1325°F + Water
Quench)
1000°F 40,000 psi
74.5 hrs
54.4% Elong.
37.9% Red. Area

Fig. 76. Typical Rupture Fracture at 600° and 1000°F
for the 30% Mo β Alloy

Area enlarged 18% in reproduction

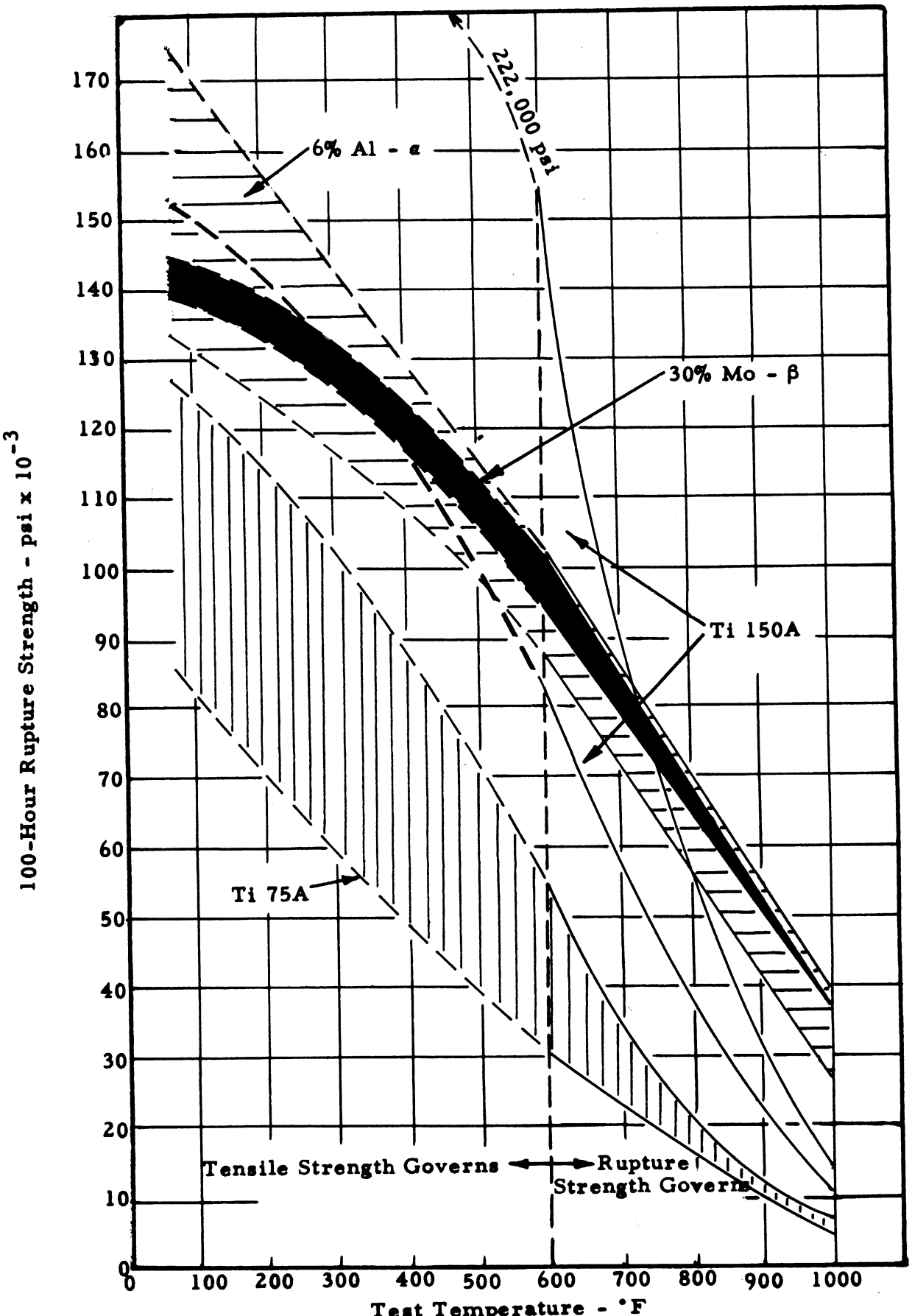


Fig. 77. Ranges of 100-Hour Rupture Strength Vs. Test Temperatures for Ti 75A, Ti 150A, 6% α Alloy and 30% Mo β Alloy.

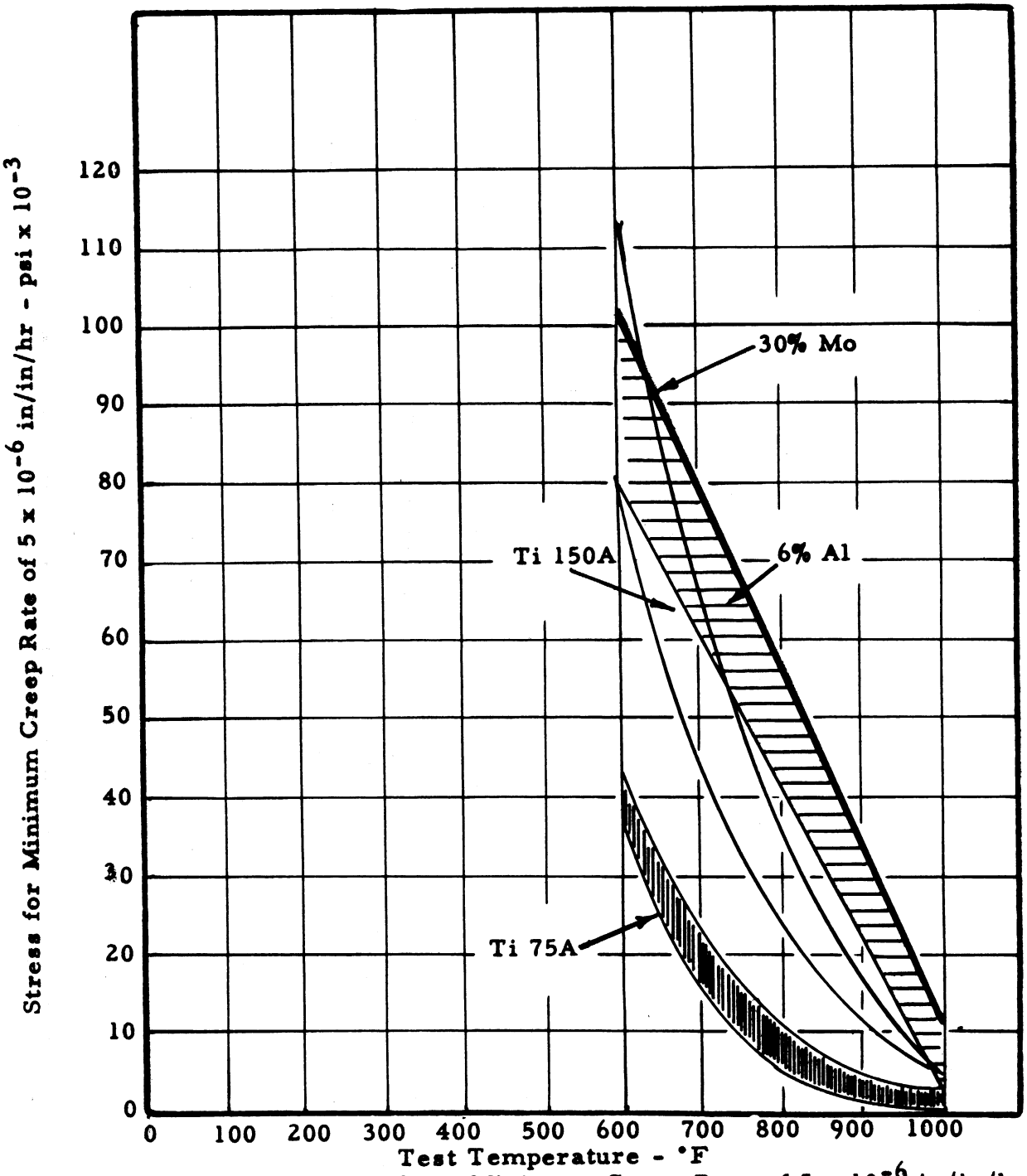


Fig. 78. Range of Stress for a Minimum Creep Rate of 5 x 10^{-6} in/in/hr Vs. Test Temperature for Ti 75A, Ti 150A, 6% Al α Alloy and 30% Mo β Alloy

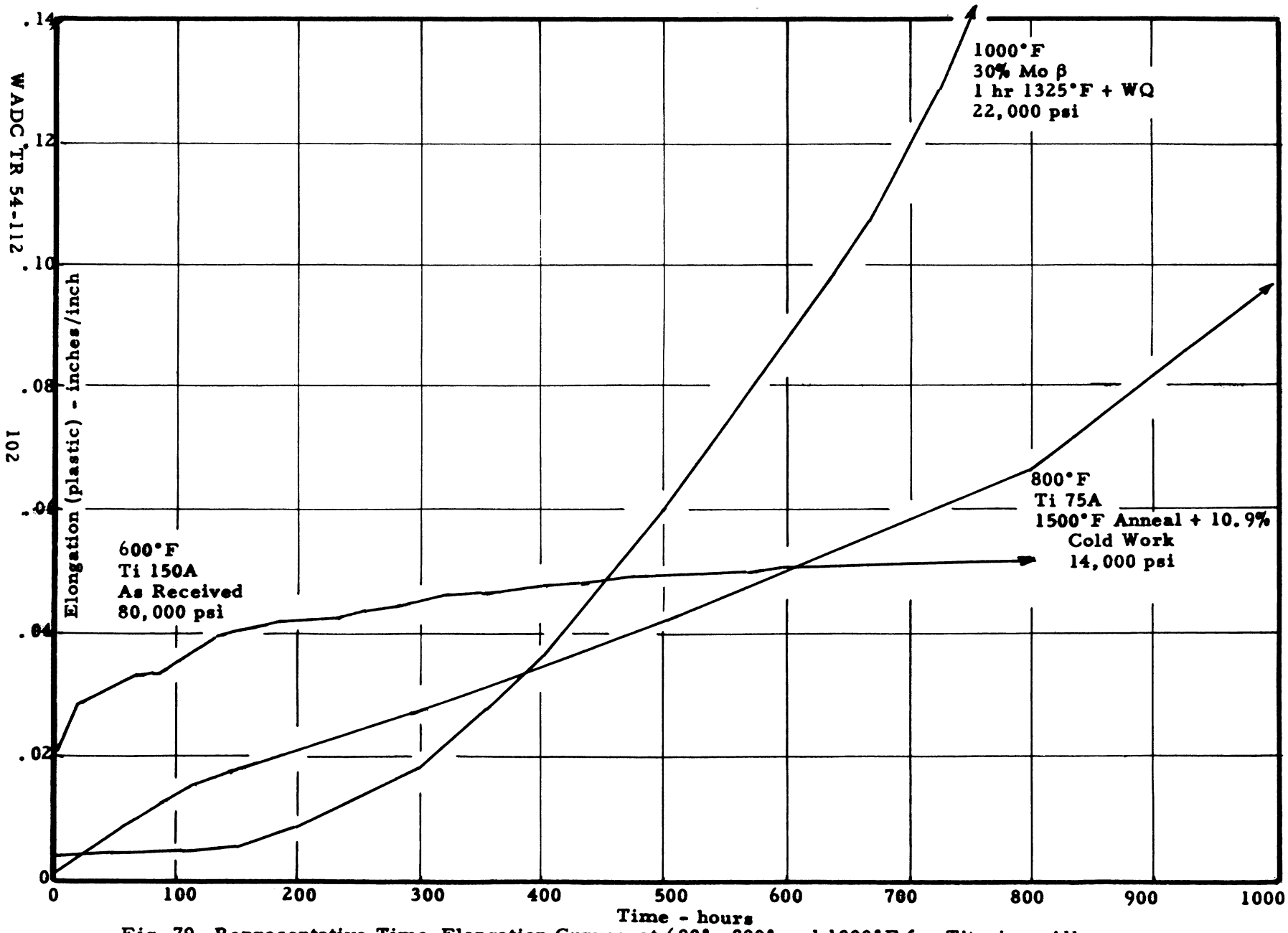


Fig. 79. Representative Time-Elongation Curves at 600°, 800° and 1000°F for Titanium Alloys