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Part II

NOTCH SENSITIVITY OF AIRCRAFT STRUCTURAL AND ENGINE ALLOYS

Part II. Further Studies With A-286 Alloy

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FOREWORD

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ABSTRACT

Sufficient data on a single lot of A-286 alloy were obtained at 1200°F to permit future evaluation of an analysis for notched-bar rupture life in terms of normal creep-rupture properties.

Circumferential notches with theoretical stress concentration factors (K_t) of 3.0 or less increased rupture time of materials with a 2200°F solution temperature, despite smooth-bar elongations as low as 0.5% or below.

Notch behavior varied widely with notch geometry in tests on specimens with 1800°F solution temperature. Transition to notch sensitivity at longer time periods was, however, positively demonstrated for several notch geometries.

The creep rate of a smooth bar at any time in a variable-stress test appeared to be a unique function of the existing stress level and the cumulative portion of rupture life already expended. No exact criterion was found for rupture of A-286 alloy after creep under variable simple tension; for engineering purposes the portion of rupture life consumed in a time interval (Δt) during which creep deformation (ΔC) occurred could be evaluated by

$\sqrt{\frac{(\Delta t)(\Delta C)}{(R)\,(D)}}$, where $\underline{R}$ and $\underline{D}$ are the rupture time and creep deformation at rupture in a normal test at the existing stress.

Firm criteria for creep rupture under a general state of variable complex stress are still lacking for the desired evaluation and extension of the proposed calculation methods.

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[a] INTRODUCTION

The Materials Laboratory, Wright Air Development Center, has supported a continuing study at the University of Michigan aimed at clarifying the basic properties of aircraft alloys which determine their creep-rupture response to the presence of a stress concentration. The present report covers experimental results obtained through the end of March 1958 under Contract AF 33(616)-3380.

If quantitative methods can be developed to permit prediction of creep rupture properties of notched tension specimens from properties determined with conventional unnotched bars, attempts are to be made to generalize the analysis to include any and all arbitrary patterns of stress distribution with time.

Early experiments under a prior study (Contract AF 18(600)-62) surveyed notch-rupture behavior of seven materials ranging from an age-hardening aluminum alloy to heat-resistant turbine alloys. (See Ref. 1) A variety of heat treatments was used with two of the alloys and five separate heats of the material were studied in the case of Waspaloy at 1350°F.

For each material and condition tested, properties of unnotched (smooth) specimens were determined and studied for factors influencing creep-rupture response to the presence of a notch. All experimental trends appeared to be capable of qualitative explanation in terms of three general factors:

(1.) The initial stress pattern around the notch, taking into account any yielding when the load is applied.

(2.) The rate of creep relaxation of the remaining stress concentration, compared to the rate at which rupture life is used up.

(3.) Alteration of original creep and rupture properties as a result of any plastic strain at loading or of time-dependent metallurgical changes during the subsequent period of variable-stress creep.

A step-wise calculation method to predict notched-bar rupture life was developed (Ref. 1), based on the assumptions that rupture life under creep conditions is controlled by an effective stress derived from the shear stress invariant theory and that fractions of rupture life are quantitatively additive for creep under variable effective stress. Predicted life agreed well with experimental determinations on notch specimens for materials tested under conditions where they are metallurgically stable. In some cases where agreement was less favorable, observed deviations could be explained with the aid of fragmentary data on changes in creep-rupture behavior of smooth specimens subjected to a history of variable stress designed to approximate conditions at different positions of a notched specimen. However, for none of the original alloys studied was sufficient stock available to permit complete evaluation of the material's smooth-bar behavior under all types of stress history to be expected in the notched rupture specimens studied.

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[a] Manuscript released by the authors on 31 July 1958 for Publication as a WADC Technical Report.

The present investigation sought to overcome this deficiency by permitting extensive work on a single alloy. For this particular research, A-286 alloy was chosen in the belief that both notch strengthening and notch weakening should be obtainable with the same lot of material by merely altering the solution temperature used in the heat treatment. Stock produced by the vacuum consumable electrode process was selected for the major study since it represented the most advanced form of the alloy available and should typify materials produced for future severe-service applications.

Survey tests at 1200°F included specimens solution treated at temperatures covering the range from 1650° to 2300°F. For stresses of 60,000-70,000 psi, smooth and notched specimens alike exhibited a steady increase in rupture times for increasing solution temperatures up to about 2000-2200°F and then the strength fell for both types of specimens. Quite unexpected was the finding that for a moderate notch acuity (K_t = 1.8) the same order of notch strengthening was obtained for all solution temperatures, despite a drop in smooth-bar rupture ductility from 8-10% for the 1650°F treatment to 1-2% for solution at 2150-2200°F. The curves of Figure 1 bear out these observations for a test stress of 70,000 psi.

Major effort during the period covered by this report has been directed toward more intensive investigation of the original lot of A-286 (Heat 21,030) for solution temperatures of 1800° and 2200°F, which give roughly comparable smooth-specimens rupture lives but widely-different rupture ductilities. The higher solution temperature resulted in a coarse grain size but was selected to study notch behavior for conditions where elongation at fracture would be 1% or less in long-time tests on unnotched specimens.

Earlier experimental coverage was broadened to include new test temperatures (1100° and 1300°F), a wider range of stress concentration factors (plus additional intermediate values of K_t) and considerably lower stress levels with corresponding prolonged test periods.

Original plans under the current contract called for extensive tests with a single lot of A-286 to define the basic properties and conditions for the different degrees of notch sensitivity expected to be introduced by altering the heat treatment. When initial tests all showed notch strengthening for typical stress levels and specimen geometries, the study was broadened to include material from other lots of A-286 alloy for which limited tests by others suggested these lots might be more prone to notch sensitivity than was Heat 21,030.

EXPERIMENTAL INVESTIGATIONS WITH A-286 ALLOY

Experimental work under Contract AF 33(616)-3380 has centered around three phases of research into notch-bar rupture behavior of A-286 alloy:

(1.) More-complete evaluation of Heat 21,030 to define material and test conditions for which notch weakening might be encountered.

(2.) Survey of other available lots of A-286 alloy to determine whether one of these might be more prone to notch sensitivity.

(3.) Evaluation of creep-rupture behavior of smooth specimens of Heat 21,030 under variable simple tension, to provide necessary data for proposed correlation attempts to predict notched-bar rupture behavior from smooth-bar properties.

Test Materials and Specimens

Available chemical analyses and processing conditions for the four lots of A-286 alloy investigated are summarized in Table 1.

Specimen blanks for Heats 21,030 and 43,297 were sampled from the round stock supplied by the producer. Material obtained from the other two lots of A-286 alloy was in the form of billets which were re-rolled at the University of Michigan before sampling, except for one group of blanks cut from the as-supplied billet of Heat K-65-X.

Before machining, specimen blanks were solution treated for one hour at the selected temperature, oil quenched, and then aged at 1325°F for 16 hours, air cooled. The gauge section of unnotched specimens was hand polished after turning. Notches were finished by form lapping after preliminary rough-turning and grinding operations. Details of the notching operation have been published in part 3 of Reference 1.

Unnotched specimens had a cylindrical gauge section about two inches long and 0.350-inch diameter. Notched bars had a single circumferential groove with 60° included angle and a circular root contour. Dimensions were chosen so that the cross section of the specimen in the plane of the notch equalled roughly half the shank cross section and was approximately the same as the cross section of the smooth bars.

All testing was carried out in University of Michigan creep-rupture frames employing dead-weight loading through a beam. A modified Martens-type optical extensometer system measured creep of unnotched specimens.

Effect of Solution Temperature on Notched-Bar Rupture Life of A-286, Heat 21,030.

For convenient comparison, all the rupture-test results at 1200°F have been assembled in Table 2, including data previously reported in Reference 2. Separate plots in Figure 2 compare rupture properties at K_t = 1.0 (unnotched), 1.8 and 3.0 for each solution temperature investigated.

Rupture life with moderate notch acuity (K_t = 1.8) exceeded that for a sharper notch (K_t = 3.0) for all test conditions. A theoretical stress concentration factor of 1.8 produced roughly the same ratio of notch-bar life/smooth-bar life for all solution temperatures. For K_t = 3.0 the degree of notch strengthening at 70,000 psi stress tended to decline as solution temperatures were raised above about 2000°F. In several instances the scattered data suggested probable notch weakening at prolonged test periods for both of the notch acuities surveyed.

After early tests at 1200°F suggested little change in notched-bar rupture characteristics of this lot of alloy for different solution temperatures, a few experiments were run at 1100° and 1300°F to determine whether a larger effect of heat treatment might be present at these test temperatures. The test results (Table 3 and Figure 3) indicate a lesser degree of notch strengthening for K_t = 3.0 at the lower test temperature than at 1300°F. Plastic prestrain at 1100°F may also have reduced subsequent smooth-bar life at that temperature to a greater extent than occurs at 1200° or 1300°F, although too few tests with initial momentary overload have been conducted to permit quantitative conclusions. No outstanding difference at 1100° or at 1300°F was noted in properties for 1800° versus 2200°F solution temperature.

Influence of Notch Acuity and Stress Level on Rupture Life at 1200°F for A-286 Heat 21,030

Two solution temperatures (1800° and 2200°F) were selected for more extensive investigation of notch behavior for a variety of notch acuities and for rupture times extending to 1000 hours or longer. Pertinent test points were selected from Table 2 to be plotted in Figure 4. The plot for 2200°F includes a few points for specimens treated at 2225°F.

Despite the number of tests conducted, scatter of results makes detailed conclusion difficult, especially for the material solution treated at 2200°F. However, some definite differences in behavior resulted with the two different heat treatments.

The curves for unnotched specimens exhibited a noticeable influence of solution temperature. The more-conventional treatment (1800°F solution) produced higher short-time strength than did a 2200°F solution, but the rupture strengths were about the same at 70 hours. For longer test durations the rupture strength of unnotched specimens solution treated at 1800°F fell increasingly below that for corresponding specimens with the higher solution temperature.

Results for notched specimens in the two conditions of heat treatment showed equal or greater differences in rupture behavior. For both treatments notched-bar strength first increased with the theoretical stress concentration factor and then decreased, in accordance with past observations by the present authors and by others (Ref. 3). But the notch acuity producing maximum notch strength appears not to be the same for the two heat treatments studied in this research.

Curves of the test points for K_t = 1.27, 1.54 and 1.82 respectively show a progressive increase in rupture strength for specimens with the 1800°F solution temperature, and the single point at K_t = 2.39 is not far below the peak strength. An undesirable degree of scatter in the coarse-grained material with 2200°F solution temperature allows much less certain establishment of the curves of Figure 4 for that treatment, but all test points for K_t = 1.82 fell below the range of data for K_t = 1.37 and K_t = 1.54. These effects are more easily discernible in Figure 5 which shows a cross plot of the 500-hr rupture strengths from Figure 4 as a function of theoretical stress concentration factor. The curve for 1800°F solution temperature rises to a maximum strength at about K_t = 2.0, with but moderate change from K_t = 1.5 to K_t = 2.5. In contrast, the 2200°F treatment resulted in an abrupt peak at low stress concentration factor (K_t about 1.4) with large change in strength for duller or sharper notches.

Where test durations for a given notch geometry included times on both sides of 100 hours, the data for both heat treatments uniformly suggested a change in slope similar to the "break" in the smooth-bar curves of Figure 4. However, a disparity is to be noted in the slopes of the long-time portions of the curves for 1800° versus 2200°F solution temperature.

The material with higher solution temperature resulted in curves roughly parallel to that for unnotched specimens. The trends established for all notch acuities between K_t = 1.27 and K_t = 3.0 suggest freedom from notch weakening for tests lasting to at least 5000-10,000 hours. Similar indications were found for K_t = 1.54 and 1.82 in the specimens solution treated at 1800°F. On the other hand, transition to notch weakening at test times beyond 400 hours (nominal stress 50,000 psi or below) was positively established for relatively sharp notches with K_t of 4.1 and 5.7. A similar transition from notch strengthening to notch weakening is strongly indicated for K_t = 3.0 at about 1000 hours. Even for a dull notch with K_t = 1.27 notch weakening should be expected at nominal stresses below about 35,000 psi.

Transition from notch strengthening to notch weakening thus appears to have been experimentally established at 1200°F for A-286 Heat 21,030 with an 1800°F solution temperature. Successful prediction of this transition and of the stress level at which it occurs should be a stern test of any computation method such as that proposed in Reference 1. A second test is whether calculations can confirm or explain the apparent freedom from a similar shift to notch weakening at long times and the apparent pronounced effect of small changes in K_t on rupture life for the material solution treated at 2200°F.

Heat-to-Heat Variation in Notched-Bar Behavior of A-286 Alloy at 1200°F

Limited research was carried out on A-286 alloy from three heats besides that used for the major study. In all three cases, prior testing by others had suggested the material might be more prone to notch sensitivity than was Heat 21,030.

Heat 43,297: Only a small amount of 7/8-inch diameter bar stock was available from this air-melted heat, one of several used by the Allegheny-Ludlum Steel Corporation in a research study of the alloy. Limited test results are listed in Table 4 and plotted in Figure 6. Transition from pronounced notch strengthening to mild notch strengthening and then to notch weakening occurred with $K_t = 3.0$ as the solution temperature was increased from 1650° to 1800° to 1985°F.

The probable greater tendency toward notch sensitivity of Heat 43,297 compared to Heat 21,030 may be associated with its corresponding greater susceptibility to weakening from strain damage evidenced by the few prestrain tests conducted on these materials. (Compare Tables 2 and 4). At the start of these prestrain tests a momentary overload was applied to produce the desired amount of plastic yielding and the excess weight was immediately removed. Extensometer readings taken during the load application and during its reduction permitted calculation of the plastic strain at the start of the test and of the correct nominal stress present for the subsequent rupture test, assuming the plastic strain to be uniform along the gauge section of the specimens.

Material remaining from this lot of alloy would conveniently serve to make only from 8 to 12 test specimens, wherefore extensive studies are not possible.

Heat 52,853: Originally this lot of alloy was one of four materials in a research into effects of processing variables on fracture of notched disks at elevated temperatures. That research by the General Electric Company under Contract No. AF 33(616)-2778 included tests at 1100°F with notched and unnotched specimens of A-286 in two conditions of heat treatment. At the 1100°F test temperature, borderline notch properties were obtained after the following treatment: 2150°F, 2 hr, O.Q. + 1650°F, 2 hr, O.Q. + 1300°F, 16 hr + 1200°F, 16 hrs. Specimen blanks were cut from disks forged from 2000°F. (See Ref. 4)

Portions of the 4-1/2-inch diameter billet transferred from General Electric were re-rolled to 1/2-inch square bars at The University of Michigan prior to sampling and heat treating. Three different rolling temperatures (2100°, 1950° and 1800°F) were surveyed for specimens solution treated at 1800°F. Limited data indicated mild notch weakening at 20-60 hours for $K_t = 4.1$ in specimens rolled from 1950°F (See Table 5 and Figure 7). Material rolled at the more conventional temperature of 2100°F produced test results quite comparable to those found with Heat 21,030. Solution temperatures of 1800°, 2000° and 2200°F all failed to produce notch weakening at 1200°F in fragmentary studies on Heat 52,853 material rolled from 2100°F. (See Figs. 7 and 8). From these preliminary tests, no evidence was found favoring Heat 52,853 in the study of conditions of notch sensitivity in A-286.

Heat K-65-X : In the course of an investigation relating melting variables to properties of certain alloys, research personnel at Firth Sterling, Inc. procured an experimental heat of A-286 produced by the vacuum consumable-electrode method. This material, designated by them as "K-65-X", gave indications of notch embrittlement at 1200°F and 65,000 psi stress for bars sampled from an 8.5-inch square billet and then solution treated at 1800°F, oil quenched, aged at 1300°F for 16 hours and air cooled to room temperature.

Knowing of the attempts to locate notch-sensitive A-286 for the present program, Firth Sterling representatives arranged to supply enough of the experimental billet to permit comparison against Heat 21,030.

Property scatter in notched specimens sampled directly from the billet was very wide. Re-rolling from 2100°F before sampling appeared to eliminate this scatter. (See Table 6 and Figure 9) For both conditions tested, specimens with the 1800°F solution treatment and subsequent 1325°F aging for 16 hours developed notch weakening in test times of the order of one hundred hours. Corresponding specimens from other heats examined to date in this program produced notch strengthening. Heat K-65-X thus seems to offer the best opportunity found to study a lot of A-286 exhibiting notch weakening in reasonable test periods.

Comparison of Notch Behavior for the Several Lots of A-286 Alloy

Results from this and other research programs at the University of Michigan (Refs. 5 and 6) point out the important changes in creep-rupture properties that can be brought about by small variations in trace-element content and by different working conditions prior to the final heat treatment. Accurate delineation of the influence from one of these variables demands the others be held fixed but this desirable aim is difficult to fulfill. Quantities of boron, oxygen and nitrogen known to have measurable effect on rupture life of heat-resistant alloys are near the limits of present quantitative analysis for these elements. What other trace elements might produce helpful or harmful effects of like magnitude and how these different elements interact in effect remains largely unknown.

Similarly, the ways in which prior processing conditions carry through to influence later mechanical properties are little understood. Even for the same nominal rolling conditions, point-to-point variations in the history of temperature and strain must influence mechanical properties and stability of the resultant metal structure.

Despite these complications, the four heats of A-286 studied can profitably be compared for the common conditions of rolling from 2050-2100°F and subsequent 1800°F solution temperature. At rupture times of the order of 100 hours, Heats 21,030, 43,297 and 52,853 all exhibited notched-specimen rupture strengths 10-20% above the smooth-bar values for K_t's of 2 to 4. Heat K-65-X, on the other hand, showed mild notch weakening under roughly comparable conditions. The most plausible explanation lies in the boron contents for the four lots of material. The single analysis available for boron in Heat K-65-X was below the average boron content indicated by the pairs of analyses obtained for the other three heats. Unfortunately, the analysis for boron in the low amounts present is difficult and agreement between laboratories was unsatisfactory for a successful correlation attempt. Re-evaluation of these four heats would be desirable on the basis of boron contents all determined by the same laboratory with the same set of standards.

Although the amount of boron or other trace elements in the alloy seems to alter notch sensitivity, the effect is probably not one of either permitting or eliminating notch sensitivity, but rather one of shifting the test time at which notch weakening begins for a given notch geometry. For short test times, the bulk of the test data suggest notched-bar rupture plots have the same slope as do those for unnotched specimens. The rupture-test results for Heat 21,030 with the 1800°F solution temperature produced three good examples of definite and abrupt change in slope for notched-specimen curves at times beyond a few hundred hours. More extensive testing would verify this finding for other notch geometries, treatment conditions, or lots of the alloy.

PERTINENT SMOOTH-BAR PROPERTIES OF A-286 HEAT 21,030 FOR USE IN PROPOSED CORRELATION ATTEMPTS

The analysis proposed for computing rupture life of notched specimens from smooth-specimen data requires knowledge of the material's behavior under the types of stress-strain history experienced by the metal in different portions of a notched specimen. The past year's research included a number of tests devised to obtain the needed data at 1200°F for A-286 alloy from Heat 21,030.

Short-Time Tensile Properties; For many notch geometries and levels of nominal stress employed in notched-bar rupture tests, local yielding occurs in the vicinity of the notch root when the load is applied. Estimation of the extent of this yielding requires that the stress-strain properties for the material be known at the test temperature. Figure 10 shows average curves obtained during the loading of groups of smooth-bar creep-rupture tests run at 1200°F on specimens with 1800°F and with 2200°F solution temperatures.

The lower solution temperature permitted loading to noticeably-higher stress levels before extensive yielding took place. As a consequence, identical specimen dimensions and test loads would result in higher initial peak stress and lower initial plastic strain for the 1800°F treatment than for solution at 2200°F. Single short-time tensile tests at 1200°F showed a similar difference in tensile strength:

Solution Temp. (°F)	Tensile Strength (psi)	Elongation (%/4D)	Reduction of Area (%)
1800	112,000	18.	19.
2200	101,500	10.5	12.5

Conventional Creep-Rupture Properties: Conventional constant-load tests at 1200°F on smooth bars covered a range of rupture lives from less than one hour to more than 1000 hours for both the 1800° and 2200°F (or 2225°F) solution temperature. For each of the two heat treatments the rupture data plot well as a pair of intersecting straight-line segments on a graph of log stress versus log rupture life (See Figure 4). For the 1800°F solution temperature, the high-stress curve can be taken through (100,000 psi; 0.3 hr) and (62,000 psi; 80 hr). The lower-stress segment passes through the latter point and (36,000 psi; 3000 hr). The corresponding equations for rupture time (R) in hours as a function of stress (S) in pounds per square inch are:

$$S \geq 62{,}000 \text{ psi:} \quad \log R = 57.837208 - 11.672010 \log S \qquad (1)$$

$$S \leq 62{,}000 \text{ psi:} \quad \log R = 33.854465 - 6.6671042 \log S \qquad (2)$$

With the 2200°F solution treatment, the point of inflection is about (63,000 psi; 200 hr) and the high and low segments pass, respectively, through (80,000 psi; 0.9 hr) and (50,000 psi; 2000 hr). The two plotted segments may be expressed:

$$S \geq 63,000 \text{ psi:} \quad \log R = 110.86090 - 22.619748 \log S \tag{3}$$

$$S \leq 63,000 \text{ psi:} \quad \log R = 50.117276 - 9.9630868 \log S \tag{4}$$

The constants in equations 1 through 4 are accurate to many less significant figures than the eight digits stated, but this eight-digit form is easily handled by most desk calculators and electronic computers and minimizes cumulative "round-off" errors in repetitive calculations using the equations.

Creep curves for this alloy were distinguished by a small proportion of primary creep and an extended period of tertiary creep, especially for the material solution treated at 1800°F. (See Fig. 11). For each of the two solution temperatures studied extensively, a plot of stress versus log minimum creep rate (C_m) could be represented by a single straight line. (Fig. 12). These lines may be expressed by the following equations:

$$1800°\text{F Solution Temp.:} \quad \log C_m = 0.0000849485\, S - 9.3055710 \tag{5}$$

$$2200°\text{F Solution Temp.:} \quad \log C_m = 0.00015688382\, S - 14.236938 \tag{6}$$

The existing creep rate at any time and at any location in a notched specimen may be presumed to depend on the local stress level and on the stage of creep in which the material exists after its prior stress-strain history. For the proposed calculation method, creep rates would be conveniently expressed as the simple product $C_m \times f(F)$, where $\underline{C_m}$ is, the minimum creep rate for the stress level acting at the moment and $\underline{f(F)}$ is a factor uniquely related to the accumulated fraction of rupture life expended to date in the portion of material under consideration.

The data of Table 7 have been assembled to examine the feasibility of using such factors. Creep rates at various fractions (F) of experimental rupture time were determined from the creep curves and were compared to the minimum creep rates for the same test. The rupture-time fraction at which the minimum rate occurred is also listed.

Specimens solution treated at 1800°F reached their slowest creep rate at a rather early stage of the test (at about 9% of the rupture time). Initial creep was relatively slow (about 4.0 and 2.0 times the minimum rate, respectively, at 0.005 and 0.02 of the rupture life). After about half the total life time of a specimen had been expended the rate of creep rose rapidly. For tests at 40,000-70,000 psi stress the creep rate was approximately fifty times the minimum value when 90% of the respective rupture times had been reached.

Findings for the 2200°F (or 2225°F) solution temperature varied in considerable detail from those just stated. During the first few percent of the rupture life the ratio of existing creep rate to minimum creep rate was at least double that for the 1800°F treatment, the minimum rate occurred later (at a life fraction of about 0.2) and creep showed only limited acceleration in later portions of the test. Even at 0.95 of the rupture time, creep took place at only about triple the minimum rate.

Except at large fractions of the rupture time for two high-stress tests on material with the 1800°F solution temperature, the relative creep-rate factor at any given life-time fraction was quite uniform for like specimens run at different stress levels. Results from these two deviant tests were ignored when the average factors, $f(F)$, were computed since the stresses used for these particular tests greatly exceed those of interest during the latter stages of the proposed calculations for notch geometries and loads of interest to the present program.

The creep-rate factors determined from the experimental data on constant-load tests with smooth specimens have been expressed in equation form to expedite calculations. No accepted theoretical basis was known which would determine the correct form of such an equation. A smoothed curve of $f(F)$ versus F was drawn through the average values from Table 7. Slopes of this curve were used to construct a curve of the first derivative $d\,f(F)/dF$, which was used in turn to prepare a plot of the second derivative $d^2\,f(F)/dF^2$. Close geometrical similarity between this second derivative and the original function ruled out simple polynominals in F and suggested use of a sum of terms of the form $A\,e^{nF}$, with both positive and negative values of n needed to conform to the minimum in the curve to be fitted.

The following expressions with five terms fit the data within the experimental deviations among the several tests:

1800°F Solution Temperature:

$$f(F) = 3.0\,e^{-300F} + 3.0\,e^{-70F} + 0.80\,e^{-7F} + 0.40\,e^{5F} + 0.000\,000\,20\,e^{20F} \qquad (7)$$

2200°F Solution Temperature:

$$f(F) = 9.0\,e^{-84F} + 0.74\,e^{-22F} + 1.3\,e^{-5.3F} + 0.35\,e^{2.0F} + 0.020\,e^{4.0F} \qquad (8)$$

Agreement between the curves of these equations and the average test points is shown in Figure 13.

Creep-Rupture Under Variable Load

Conditions in a notched tension specimen vary in at least three important ways from those of the usual creep-rupture test:

(1.) Stresses in two or three principal directions act simultaneously in a notched specimen.

(2.) The stress level and degree of triaxiality at any point change with time in the notched bar, even under a steady axial load.

(3.) Material near the notch root may be subjected to an initial plastic strain, with the amount of strain depending on the specimen geometry, the applied load and material properties. This plastic strain may alter subsequent creep and rupture characteristics.

Our knowledge of creep-rupture under multi-axial loading is very sparse. Even in the few experimental results now at hand, points of disagreement exist. These may best be discussed after the following survey of experiments involving variable stress.

Review of Past Tests for Variable Simple Tension: As one step toward understanding the more-complicated conditions in a notched specimen, attention has been directed to creep-rupture behavior of smooth specimens subjected to variable axial load. In particular, evidence has been sought to substantiate or refute the quantitative addibility of rupture-time fractions suggested by prior studies in this research.

Published data from other laboratories are confined largely to a 1954 symposium sponsored by the American Society for Testing Materials. (See Ref. 7).

Many of the experiments reported therein involved periodic unloading to zero stress. Such a stress history differs considerably from that of fibers in notched creep-rupture specimens, but does furnish a base for evaluating tests in which the stress is cycled between finite levels of stress.

In their paper at the symposium, Smith and Houston cited two creep-rupture tests on 18 Cr, 8 Ni (0.03C) steel in which the load was removed for 24 hours once each week, the temperature remaining constant. At 1100°F the periodic-load operation had no significant effect on creep under the 15,000 psi stress, but the total time under stress before fracture was about 60% longer than the normal rupture time for 15,000 psi stress. In a test at 1500°F the periodic load removal resulted in a 70% reduction in minimum creep rate at the 3000 psi stress used, while the total time under stress until fracture was very near that for a test at constant stress of 3000 psi.

Extensive studies on aircraft sheet alloys reported by Guarnieri indicate that periodic removal of the load during a test at constant elevated temperature may either extend or shorten the duration of on-load time before fracture. In Guarnieri's research the load was automatically applied and removed for alternate equal periods, of either one hour or eight hours duration according to the particular tests. His results for rupture time are summarized in the following table:

Alloy and Test Temperature	Ratio: (Accumulated time at load / Rupture time for constant load) for a stress giving the indicated normal rupture life: 10-hr	20-hr	100-hr
321 stainless steel; 1200°F	--	0.80	0.55
321 stainless steel; 1350°F	0.85	--	1.10
N-155 alloy; 1350°F	--	a) 0.70	a) 0.88
N-155 alloy; 1350°F	--	a) 0.50	a) 0.80
N-155 alloy; 1500°F	--	a) 1.05	a) 0.90
N-155 alloy; 1500°F	--	0.95	0.90
Inconel X; 1350°F	1.20	--	0.80
Inconel X; 1500°F	0.72	--	0.97
24S-O Aluminum, Alclad; 600°F	1.38	--	1.92
24S-T3 Aluminum, Alclad; 600°F	0.30	--	0.30
RC130A Titanium alloy; 800°F	1.00	--	b) 3.57
FS-1H Magnesium alloy; 300°F	2.10	--	1.75
FS-1H Magnesium alloy; 450°F	1.20	--	1.70

a) Cycle = load on 8 hrs, off 8 hrs. All other tests: cycle = load on 1 hr, off 1 hr.
b) At 70-hour rupture stress.

With attention confined to iron- or nickel-base materials, rupture-time fractions offer a fair criterion for the time to failure in nearly all cases. Deviations caused by such factors as creep recovery and continued aging during the no-stress periods appear to vary with material and test conditions, but the average total rupture-time fraction was less than 1.0 when fracture took place.

Such time- and temperature-dependent effects should be lessened by lowering the temperature while the load is not acting. Mr. J. R. McDowell's discussion to the Smith-Houston paper mentioned tests on forged 19-9DL alloy which tend to confirm these expectations. When the temperature was held constant at 1200°F, but the load applied for only 16 hours each day, the rupture time was half the standard life of the material except for one lot of specimens with higher hardness and strength level. (In the latter case the interrupted-load life was reduced only about 20%). In a second group of tests the temperature was lowered during most of the off-load period. The accumulated time under load when fracture occurred was now 55%, 70% and 115% of normal rupture time, respectively, for groups of specimens with normal rupture lives of 90, 110 and 300 hours. An inference drawn by McDowell from these last results was that effects of load cycling for a given material should be most pronounced at high stress levels where the normal rupture life is quite short, and should be less at lower stress levels corresponding to longer life.

A paper by Caughey and Hoyt reported on tests in which Inconel specimens were at all times under load at a constant test temperature of 1800°F. Either a single period of increased stress (at 3200 or 2500 psi) or a series of overstress periods was applied--in each case after second-stage creep had become established at the base stress of 1700 psi.

Figure 14 (reproduced from their Figure 10) typifies Caughey and Hoyt's findings for overstress effects. In four separate tests, a single overload for 15-75% of the normal life time at the overload stress had no damaging effect on rupture time. Test number 3 shown in Figure 14 lasted a total of 122% of the normal life time at the base condition, plus the sojourn at 3200 psi for 40% of the normal rupture time at that stress. The other three tests with single overload were not continued all the way to fracture, but cumulative rupture-time fractions were 1.10, 0.88 and 1.24 respectively when the tests were stopped for specimens which had been subjected to 15, 50 and 75% of the normal rupture time at the overload stress.

Another series of three tests attempted to apply the same total duration of overstress, but in ten equal periods instead of all at once. In all three tests, fracture occurred before the ten overload cycles could be completed. Elongation at rupture was only 7.5, 8.3 and 11.5% respectively for corresponding total rupture-time fractions at failure of 0.31, 0.52 and 0.68. The very short life for the first of these tests may have been due in part to reported difficulties with temperature control for this particular experiment.

One could conclude from these findings that stress-rupture performance is adversely affected by increasing the number of abrupt changes in the applied stress level. Examination of Figure 14 suggests that the portion of the test where the overstress is applied may be a more pertinent factor. The base condition of 1700 psi exhibited a moderate and rather-uniform rate of creep for the first 40% of its life time. The creep rate then rose rather abruptly, so that about 95% of the total rupture elongation occurred in the last 60% of the time until rupture. For the overstress condition of 3200 psi the creep curve shown is nearly linear for its entire life. In all variable-stress tests the overload was first applied soon after the creep approached a constant rate. With a single overload, much or even all of the time at higher stress was during the time when the specimen material was at its state of maximum resistance to creep. In contrast, some of the overload periods in the cyclic-overload tests were not applied until after the transition to rapid third-stage creep.

The particular base-test data included in Figure 14 suggest a change in rupture elongation with stress level. However, a constant rupture elongation of about 30% is indicated by the complete listing of constant-load tests, reported in Table II of the paper cited.

Multiple-stress creep-rupture data have been obtained at the University of Michigan both under WADC sponsorship and as part of unsponsored doctoral research on the rupture life of heavy-walled pressure vessels. These results from parts 1 and 3 of Reference 1 and from Reference 8 are combined in Table 8 for convenient comparison. A few additions and corrections have been made to the original referenced tables.

In these tests, creep specimens were run under one load for part of the test time and then changed to other load levels for other portions of the total rupture life. So that conditions near the notch root of a notched specimen might be simulated, several of the later tests on A-286 alloy involved high initial stresses which produced measurable plastic strain at loading. The stress level was lowered for each succeeding portions of these particular tests, giving a step-wise approximation to the continuous creep-relaxation of high initial stresses in material near a notch root.

For Inconel X-550 alloy, the sum of rupture-time fractions deviated as much as 25-30% from the average value of 1.03. Scatter seemed to be random, with no apparent influence of whether the stress level was raised or lowered from the initial value. In each of the eight experiments a single change in stress was made after the material was in the prolonged stage of increasing creep rate which typifies this material.

Scatter of results was slightly less for Waspaloy at 1500°F and for S-816 at 1350°F, with the average sum of rupture-time fractions equal to 1.06 and 0.91, respectively, for these materials. The first three tests listed for S-816 suggested that stress changes during a test might shorten the rupture life, but a new experiment conducted during the past year refutes this conclusion and counters the findings of Caughey and Hoyt for Inconel. The S-816 specimen in the recent test was subjected to sixteen cycles of 11 hours at 30,000 psi alternating with 1 hour at 45,000 psi. The specimen then ran another 159.3 hours at the 30,000 psi stress before rupture at an elongation of about 46%. Total rupture-time fraction at rupture was 1.06.

Rupture-time fractions at rupture averaged about 0.92 for seven tests at 900° or 1050°F with annealed carbon steel, in substantial agreement with the simple time-fraction rule. As might have been anticipated, this rule proved to be considerably in error for 2024-T4 aluminum alloy at temperatures where the material exhibits pronounced over-aging. All four tests deviated in the expected direction for a time-dependent loss in inherent strength at service temperature; i.e., long exposure to a low stress shortened the subsequent time at higher stress by a far greater amount than would the short time required at the high stress to cover a like fraction of rupture time. In step-down tests, the initial period at high stress permitted less over-aging than would the corresponding long time to use up the same rupture-time fraction in a test at low stress throughout the experiment.

Multiple-stress rupture results at 1100°F are available for two lots of 17-22A(S) steel, but not at corresponding stress levels in the two cases. Six tests on Heat 26,157 of this alloy (Ref. 8) lend confirmation to addibility of rupture-time fractions, whereas five tests on Heat 31,158 material used for notch studies (Ref. 1) were in poor agreement with the simple time-fraction rule.

(The time-fraction sum of 0.57 listed for a test started at 60,000 psi and reduced to 20,000 psi in three steps should probably not be considered on a par with the other tests for Heat 31,158 since for some unexplained reason the creep at the initial stress was five times faster than for another test run to rupture at a constant 60,000 psi load. In 1.1 hours at 60,000 psi the creep deformation was nearly 3%, suggesting that the specimen would only have lasted 2-3 hours at this stress instead of the 9.8 hours rupture life determined from the average curve of stress versus rupture life.)

A probable factor in the apparent disparity in results for the two lots of 17-22A(S) is the larger range of stresses covered in a given test in the experiments on Heat 31,158. The plot of log stress versus log rupture life for both lots of steel exhibited a decided "break" in slope at about 50,000 psi stress. Except for one brief period during one test, the entire stress history for all specimens from Heat 26,157 was at or below 50,000 psi. In contrast, every multiple-stress experiment performed on material from Heat 31,158 included time at 60,000 or 65,000 psi and also at 25,000 psi.

Re-examination of the rest of Table 8 shows that the experiments which correlate well with rupture-time fractions were characterized by having all stress levels on the same portion of the stress-rupture curve or else the span of stress levels around the "break" in the curve was quite moderate. The two cases for which the correlation broke down the most (step-up tests on 2024-T4 alloy and on 17-22A(S) steel from Heat 31,158) were marked by an initial period at a stress well below that at which the slope of the rupture curve changes, followed by a period at a stress considerably above the "break" in the rupture curve. Such behavior might simply indicate a change in the mode of failure between low-stress and high-stress conditions, probably with a different criterion for the rate at which rupture life is consumed in different stress ranges. More probably, time-controlled metallurgical alterations are the major factor involved. This last conclusion is based primarily on higher-than-expected rates of creep for step-up tests following the stress increase. For the 17-22A(S) steel specimens run at 25,000 psi for one-third or one-half of the normal life at this stress, raising the stress to 65,000 and 60,000 psi respectively in the two tests, produced creep rates approximately ten and fifty times the corresponding creep rate in a conventional test run at the high stress throughout.

Stress Direction Changed During Tests: A few unpublished exploratory tests performed in connection with the research of Reference 8 may be pertinent. These tests employed the 17-22A(S) steel from Heat 26,157 at a test temperature of 1100°F. A stubby specimen of this material was subjected to creep in compression at 40,000 psi for 40 hours, then cooled under load, remachined to a miniature tension specimen (0.160-inch diameter) and run to rupture in tension at 40,000 psi. Fracture occurred after 23.8 hours of tensile creep. Normal rupture life at 40,000 psi is about 84 hours. In a second test at 30,000 psi (330 hours normal rupture life), 119 hours of creep in compression and 163.3 hours in tension elapsed before fracture.

Addibility of rupture-time fractions probably has little meaning when pure compression loads are involved but these two tests did show that a period of creep in compression "used up" an amount of rupture life roughly equal to the time of the creep divided by the normal rupture life under constant tension.

Other experiments with this material sought to explore creep-rupture life for the changing stress pattern of a thick-walled vessel under internal pressure. In principle, analysis of creep-rupture behavior for such a vessel is like that of a notched specimen creeping under axial tension, although the details differ considerably for the two cases. Initial stress concentrations at the bore should tend to level out by action of non-uniform creep across the vessel wall.

A steady-state stress pattern equivalent to the changing stress level at any point in a thick-walled pressure vessel at elevated temperature is difficult to define. However, the approximate magnitude of such an equivalent stress at the mid-thickness should be given by the average values defined by equilibrium between the pressure forces and the resistance offered by the walls. For a wall 1/2-inch thick in a vessel with inside diameter of 1.0 inch, the average tangential stress is numerically equal to the pressure, p , and the average axial stress would be 1/3 times the pressure. The radial stress is a compression which varies with time at interior points but can probably be approximated closely enough as the mean of the steady boundary values of zero and -p at the outer and inner surfaces.

Creep for 370 hours at a pressure of 20,045 psi caused failure in such a vessel. Two samples were then cut from the vessel wall a short distance from the break, one oriented in the longitudinal direction and the other tangentially. Specimens machined from these samples ruptured in 7.9 and 4.3 hours respectively with 50,000 psi stress. This compares with a normal life of 25 hours in steady tension. Another vessel failed in 223 hours at 24,925 psi internal pressure. A specimen sampled in the longitudinal direction from this failed vessel ran 12.5 hours at 50,000 psi before rupture, while a tangential specimen ruptured after 7.6 hours at the same stress.

Creep curves for all four specimens were quite similar and had a shape rather as would be expected for the latter stages of a test run at 50,000 psi nominal stress under constant axial load.

The measured residual life-time in the tangential direction (equal to 17-30% of the original value) seems reasonable for material which has been under conditions similar to those at the critical point which failed in the pressure tests. These results can probably be interpreted as confirming addibility of rupture-time fractions for the case where the major principal stress is always applied in the same direction.

Greater difficulty arises in interpreting the findings for the two specimens taken from the longitudinal direction of the pressure tubes. Residual rupture life was nearly double that for the tangential direction, as might be anticipated. The more important observation seems to be that creep under complex stressing lowered the subsequent rupture strength in a direction other than that of the largest principal stress. At least, the axial stress should have been of much too low a magnitude to account for the observed "consumption" of 50-70% of a total rupture life. Time-dependent metallurgical alterations may be of greater importance than the stress pattern for very long-time creep, but nearly identical rupture lives have reportedly been found for specimens sampled from the longitudinal and transverse directions of tubes removed from extended pressure service at elevated temperature. (Ref. 9).

These observations cast serious doubt on the generality of published findings (Ref. 10) that the largest principal tension stress gave the best correlation for creep-rupture of thin-walled tubes subjected to combined tension and torsion. Reasons for this apparent disparity are not understood. Part of the difference may derive from static versus changing pattern of applied stresses. Conceivably, the mechanism of fracture under the pure shear of a torsion test could differ from that in tension. Or the lack of general agreement may result from data scatter present in the few tests performed to date. In any case, need exists for additional study before one can state with certainty the criteria for creep rupture under variable complex stresses. Such further study should preferably employ specimens of sufficient thickness to minimize surface effects, while at the same time attempting to eliminate stress gradients in the gauge section. To best meet the needs of the analysis for notched rupture bars, the changing tension-tension stress pattern in the vicinity of a notch root should be approximated in the test specimen.

Multiple-Stress Tests with A-286 Heat 21,030: Since addibility of rupture-time fractions proved to be valid for the overwhelming majority of tests performed for many materials and conditions, results for A-286 Heat 21,030 have been examined first on this same basis. (Table 9).

In tests with initial plastic strain less than 0.1% (0.0010 in./in.) and in which the stress levels covered a range of 25,000 psi or less, the cumulative rupture-time fraction at the experimentally-observed time of fracture was approximately equal to unity. This was true whether the stress history consisted of a single reduction, two successive reductions, an increase followed by a reduction or a reduction followed by an increase. The specimens heat treated at 2200°F exhibited considerable scatter in results, but so did the conventional constant-load rupture tests for this treatment.

Deviation from simple addibility of rupture-time fractions was prominent for six specimens with the 1800°F solution temperature, all started at an initial stress of 90,000 or 100,000 psi. These specimens consistently broke at about half the expected time-fraction sum whether the stress reduction was made in one or in several steps, and apparently independent of whether the final stress level was 50,000, 60,000 or 70,000 psi. Under comparable conditions the life of material with 2200°F solution temperature was not reduced and may actually have been longer than predicted by the summation of rupture-time fractions.

A reasonable impulse is to attribute these disparate findings for the two heat treatments to a difference in response to the plastic strains experienced by the specimens upon initial load application. Available data on effects of plastic prestrains hardly predict the magnitude of effects required to explain the experimental results of Table 9.

Modification of Rupture Times by Prior Plastic Strains

Table 2 included rupture data for smooth bars prestrained by momentary overloading at the test temperature. Ductilities at fracture agreed substantially with values obtained in normal tests.

Experimental rupture times for 1800° and 2200°F solution temperatures are compared below against values calculated by equations 1 through 4 of the previous section. Comparative results for other solution temperatures were determined from the curves of Figure 2.

EFFECT OF PLASTIC PRESTRAIN ON RUPTURE LIFE OF A-286 AT 1200°F

Solution Temp. (°F)	Plastic Prestrain (%)	Stress (psi)	Rupture Life, R (hrs)	Corresponding Life, R_0, with no Plastic Prestrain (hrs)	Ratio, R/R_0
1800	0.80	70,560	17.9	17.6	1.02
1800	0.90	60,540	59.8	93.8	0.64
1950	0.72	70,525	31.3	43.2	0.72
1950	1.33	65,870	90.0	100.0	0.90
2050	2.03	61,220	183.6	395	0.46
2200	1.31	60,750	99.8	290	0.34
2200	0.90	60,240	307.0	315	0.97
2200	0.97	60,000	[a]226.6 +	330	0.69 +
2200	0.37	60,000	[a]328 +	315	1.04 +
2300	1.56	60,940	17.6	42.0	0.42

a) Test discontinued by equipment difficulty.

Results for any one solution temperature are too meager to permit quantitative definition of effects of plastic prestrain. Treating all the prestrain results as one set of data, as is indicated by the solid curve of Figure 15, plastic strains of 0.5% or more should produce a measurable loss in subsequent rupture strength. Such a trend would agree with more-extensive prior studies on Waspaloy tested at 1350°F. (See Table 10 and Fig. 16). In these prior studies, effects of plastic prestrain were pronounced. Initial strains of 0.5% or less cut the residual rupture time to half the value for a like stress in a normal test. For a given stress level, the logarithm of the rupture time dropped roughly in proportion to the amount of plastic strain imposed at the start of the test.

Should such an interpretation be applied to the data of Figure 15, the correlation with time-fraction addibility becomes slightly better for the six deviant specimens solution treated at 1800°F. But the correlation for material solution treated at 2200°F would be made even worse.

The dashed curve of Figure 15 indicates a possible alternate interpretation for the four specimens with 2200°F treatment, whereby plastic strains of the magnitude of one-half percent would cause an increase in rupture strength. Prestrains of 1% or more would again be damaging. Considered in this light, the prestrain data could partially explain the large rupture-time fraction determined in the test with 0.73% initial plastic strain and the smaller fraction calculated for tests with 0.41 and 1.62% plastic strain at the start of the respective tests. Failure of the 1.62% initial strain to reduce the cumulative rupture-time fraction to less than 1.0 might be attributable to poor definition of the normal rupture properties for this condition or to data scatter.

However, no reasonable interpretation of Figure 15 seems capable of bringing into line the time-fraction results listed in Table 9 for the last six specimens with 1800°F solution temperature. A range of initial plastic strains between 0.1 and 0.5% produced no discernible difference in the computed sum of rupture-time fractions. One might argue that six multiple-stress tests with initial plastic strain are not a reasonable sample or that the prestrain tests of Figure 15 contain only two valid comparison points, but the fact that all six of the multiple-stress results would fall outside the maximum envelope for all points of Figure 15 does seem a significant demonstration that the deviant values in Table 9 cannot be explained by prestrain effects alone.

The internal consistency of the six results in question attests that the findings are probably real. Errors may exist due to faulty selection of rupture properties for conventional tests. However, the properties chosen gave satisfactory correlation between experiment and prediction when all stresses were 75,000 psi or below, and so should be near the correct values. Furthermore, if one examines the tabulated values for the multiple-stress tests in question, a large error from poor selection of rupture times for 80,000-100,000 psi stress seems unlikely. The rupture-time fraction computed for any one period in this stress range is 0.10 or less in nine out of eleven cases. To bring the computed rupture-time fraction up to unity would require a five-fold or greater reduction in the rupture life selected for 100,000 psi. The rupture-test data of Figure 4 and a tensile strength in excess of 110,000 psi warrant no such selection.

Even allowing for possible prestrain effects and possible errors in the curves of Figure 4, addibility of rupture-time fractions seems of doubtful quantitative validity under the conditions of the step-down tests from 90,000-100,000 psi with specimens solution treated at 1800°F.

The unexpected multiple-stress results for A-286 solution treated at 1800°F are not attributable to the use of stress levels on both sides of the change in slope of log stress versus log rupture life for normal tests. In two of the tests in question the stress range never fell below the "break" in the rupture curve and for three of the remaining four tests the lowest stress was only slightly below the stress at which the change in slope is indicated. Moreover, time-dependent deterioration of inherent rupture strength would be expected to result in low sums of rupture time fractions only for step-up tests, not for experiments where the stress was lowered as the test proceeded.

Alternate Criteria for Correlating Rupture Life Under Variable Stress

Any criterion of rupture useful to the proposed analysis of notched bars should be capable of expression in terms of time, stress and deformation. These variables may or may not be pertinent to the actual behavior of the material in a notched specimen, but they are the terms in which available creep-rupture properties are given.

Until some general phenomenology of variable-stress fracture is established, theoretical consideration can offer little aid toward determining a priori the form in which these experimental variables enter, or even whether the same variable or combination of variables will hold in correlations for different materials or for different test conditions.

A helpful guide is that the parameter used to define expenditure of rupture life must be of the same order as the ratio of the actual time at the given condition to the time until rupture in a test run entirely at that condition of stress. Otherwise the generally-good correlation already demonstrated for many materials and stress histories should never have been statistically possible. A parameter is therefore sought which is essentially linear in time for those creep conditions where the cumulative sum of rupture-time fractions reaches unity at the time rupture occurs, but which deviates from time-wise linearity for conditions where the time-fraction rule fails.

Strain seems to be a logical alternate to time as a measure of life consumption during creep, particularly since creep deformation is often rather linear in time for an extended portion of a test. Under some conditions, however, a sharp curvature may exist in the deformation-time relationship, or the character of the creep curves may change for different stress levels. Careful examination of such creep curves may provide insight into the deviant factor which renders the time-fraction rule invalid for certain materials and test conditions.

The shapes of creep curves may be more readily compared if they are plotted on "reduced" coordinates; i.e., showing creep strain and time as fractions of their values at rupture, plotted on scales running from zero to unity. Most of the materials for which multiple-stress rupture data are available exhibit a rather narrow band of "reduced" creep curves. The chief exceptions have been the Inconel alloy used by Caughey and Hoyt (Ref. 7), the Inconel X-550 of Reference 1 and the A-286 from Heat 21,030 with 1800°F solution temperature. In each of these cases, a plot of creep strain versus time was rather close to a straight line for high-stress tests, but in low-stress tests the creep at early times was relatively slow and the bulk of the creep deformation at fracture occurred during the last 20-50% of the test duration. (Figures 14, 17 and 11 show typical creep curves for these three alloys. "Reduced" creep curves for the A-286 material are also shown in Figure 18).

Pre-strain tests discussed in the previous section suggest that short-time plastic strain does not "use up" creep-rupture life in the same manner as does time-dependent creep. Initial plastic strains of 1% or more had little noticeable effect on subsequent rupture time, or on the shape of creep curves or on creep deformation at fracture. This conclusion is supported by Clauss (Ref. 11) who showed that subsequent stress-rupture life and creep ductility of S-816 and Inconel X-550 alloys were but little affected by thermal cycling to a large fraction of the cycles which would result in failure under continued thermal fatigue. The prior thermal cycling actually increased subsequent rupture strength of S-816, while the life of the Inconel X-550 was lowered somewhat, but the observed change in rupture properties for either material was relatively small considering the rather extensive plastic strains involved in the thermal cycles.

Even when attention is confined to time-dependent creep deformation, the particular manner of combining creep strains for different stress levels is subject to a wide choice. The simplest reasonable hypothesis seems to be that fractions of rupture deformation are additive; i.e., the portion of rupture life expended by a period of creep at a given stress would equal the ratio:

$$\frac{\text{(Actual creep strain during the time at the stress)}}{\text{(Creep strain to rupture in a normal test at that stress.)}}$$

This criterion presupposes a life-damage effect linear with creep strain. A given amount of creep deformation is assumed to correspond to expenditure of the same amount of rupture life whether it occurs at the start, near the middle or at the end of a test. A time interval of given duration would "consume" a larger portion of the total life during the primary and tertiary stages of rapid creep. Expenditure of life would be essentially linear in time for materials exhibiting a prolonged period of nearly-constant creep rate.

Table 3 of Reference 1 has previously indicated a definite superiority for addibility of rupture-time fractions over addibility of creep-strain fractions as the criterion of rupture life for Inconel X-550 alloy under variable stress. Table 9 extends the comparison between these criteria to the A-286 alloy at 1200°F. Casual inspection might suggest that the criterion based on deformation fractions is as good as or better than that for rupture-time fractions. Closer examination shows that the deformation criterion gave good correlation mainly for those tests in which the stress levels employed all had about the same "reduced" creep curves. Creep-deformation fractions consistently added to less than 1.0 for step-up tests involving a large stress increase, and added to more than unity for most of the tests with a step-down stress history.

Tabulated calculations for both rupture-life criteria reflect usual data scatter for rupture time and rupture ductility. (Comparison ductilities for normal tests were determined from average curves shown in Figure 19.) If due allowance is made for exceptions due to this scatter, errors in the two methods seem to be in opposite directions; i.e., where addibility of rupture-time fractions yields a low answer, addibility of rupture-deformation fractions gives a high result and vice versa. Conditions giving good correlation by one of the criteria seem also to be favorable to use of the second. On the basis of these limited observations, a proper criterion for consumption of rupture life under variable stress seems to be somewhere between the two considered and perhaps closer to the time-fraction rule.

A reasonably-close answer should be obtained from the geometric mean of these two factors. The portion (ΔL) of total rupture life consumed in a time interval (ΔT) during which the creep deformation was ΔC would then equal

$$\Delta L = \sqrt{\frac{(\Delta T)}{(R)} \frac{(\Delta C)}{(D)}} , \tag{9}$$

where R and D are the rupture time and creep deformation at rupture in a normal test at the existing stress. For a very short time interval dT, the creep strain may be set equal to the product of the momentary creep rate $\dot{C}$ and the duration of the interval, wherefore the differential portion of total life used up could be expressed by:

$$dL = \sqrt{\frac{(dT)}{(R)} \frac{(dT)(\dot{C})}{(D)}} = dT \sqrt{\frac{\dot{C}}{RD}} \tag{10}$$

For the three conditions examined in Table 9, correlation by Equation 9 was more consistent than was that by either of the other criteria. In the three cases examined, the average summation of ΔL calculated by Equation 9 was slightly below 1.0, but the deviation from the mean value never exceeded a total range of 0.66. Deviations were twice this range for each of the other criteria in two out of the three conditions listed in the table.

Despite the apparent satisfactory correlation of variable-stress rupture life by the geometric mean of the time fraction and creep-strain fraction for each stress period, no valid basis exists for Equation 9. Any general rule for behavior under variable stress level must apply exactly to the limiting cases of constant high stress or constant low stress throughout the test. Such is not true for the geometric mean of Equation 9.

Fractions of the total time or of the total creep strain are, by definition, strictly additive to 1.0 for a test at constant stress. So are portions of absolute total time and absolute creep deformation, but neither of these latter criteria is plausible. (Otherwise one would experience a shortening of rupture time by reducing the stress on a specimen of Inconel X-550 or A-286 run for part of its normal life at an initial high stress level.) If a convenient expression were available for the total arc length of a creep curve, fractions of arc length might be considered since they would also add to unity for a constant-stress test. In most instances the geometric mean of Equation 9 should furnish a reasonable approximation to this arc-length criterion.

Many more-complicated combinations of stress time and creep deformation could be investigated. However, the present multiple-stress results are probably not sufficiently general to permit definitive statement of a law for rupture following a variable-stress creep history. The rule proposed in Equation 9 should satisfy engineering needs for the A-286 under the conditions investigated. More detailed examination of alternate rupture-life criteria will probably be justified only when insight into the basic factors involved is gained through tests in which the level and direction of stress application both vary during the experiment.

Effects of Stress History on Creep Rates

Results for the several alloys studied earlier (Refs. 1 and 2) suggested that for metallurgically-stable alloys not subjected to loads causing short-time plastic strain the creep rate at any instant depended uniquely on the existing stress level and the cumulative fraction of rupture-time which had been expended to date. (In view of the usual scatter in creep-rate data, the same result should be obtained if another measure of expended life fractions is found to replace time fractions.) The previous studies also showed a measurable increase in creep rates, at least for the early portions of the test, when specimens were subjected to initial plastic strain by a momentary overload applied at test temperature.

Experiments now completed with A-286 seem to confirm both these earlier observations.

Influence of Plastic Prestrains: Four prestrain tests, two each for the 1800° and for the 2200°F solution temperature, permitted calculation of creep-rate factors for comparison against results on specimens not prestrained. These values have been added at the end of Table 7 along with the minimum creep rates for these and two other tests interrupted by equipment difficulties. The patterns of creep-rate variation as the test progressed in time agreed reasonably well with those determined from specimens not subjected to prior plastic strain, although the tertiary creep rates in prestrained specimens solution treated at 1800°F failed to show the extreme rise observed in the last half of the conventional tests for this condition.

For both solution temperatures studied extensively, prior plastic strains of around one percent seemed to result in a general increase in creep rates over the values given by Equation 5 and 6. Individual tests showed considerable variation, but the following tabulation does demonstrate this trend:

EFFECT OF PLASTIC PRESTRAIN ON CREEP RATE OF A-286 AT 1200°F

Solution Temp. (°F)	Plastic Prestrain (%)	Stress (psi)	Minimum Creep Rate, C (in./in./hr)	Corresponding Creep Rate with no Plastic Prestrain, C_m (in./in./hr)	Ratio C/C_m
1800	0.80	70,560	0.00125	0.00049	2.6
1800	0.90	60,540	0.00023	0.000069	3.3
2200	1.31	60,790	0.00027	0.000020	1.35
2200	0.90	60,240	0.0000087	0.000016	0.55
2200	0.97	60,000	0.000036	0.000015	2.4
2200	0.37	59,920	0.000058	0.000015	2.3

Creep Rates in Multiple-Stress Tests: Normal creep rates at any rupture-time fraction for A-286 alloy are readily obtained by combined use of Figures 12 and 13. A comparison between the rates so determined and those found experimentally in the multiple-stress tests is included on Table 9 for specimens not subjected to initial plastic prestrain. The experimental creep rates represent the approximate slope of the creep curve determined from the first few test points after a change in stress level.

Discrepancies between individual measured creep rates and the average rates for normal tests are as much as five-fold in one instance and are nearly two-fold on the average. But this amount of scatter is also present in the data of Figure 12 for duplicate tests at identical constant-load conditions. At least for the purposes of the proposed analysis of rupture life in notched specimens, a prior history of variable stress below the yield range seems to produce negligible modification of the creep rate for a given stress level.

DISCUSSION AND CONCLUSIONS

The major aim of the current phase of research was to obtain both notch strengthening and notch weakening in the same lot of material by use of different heat treatments, and then to study in detail how other properties were altered simultaneously, in the hope of thereby ascertaining which properties were associated with the observed difference in notch response.

These ends have in large measure been reached. In some respects the results actually surpassed original hopes--Heat 21,030 material with a single heat treatment (1800°F solution plus aging at 1325°F) has been found to exhibit transition from pronounced notch strengthening to notch weakening by lowering the nominal stress and, moreover, this transition appears to vary with notch geometry. The observed occurrence of notch weakening in A-286 alloy at times as short as 400 hours at $K_t = 4.1$, and for rupture times of a few thousand hours with duller notches should be of interest to users of this material. An important practical result is the demonstration that the particular geometry of a notch (or other stress raiser) can have an important effect on the resulting rupture strength.

The number of experiments has been sufficient and the scatter small enough to establish with considerable certainty the trends for Heat 21,030 with 1800°F solution temperature, affording a good set of experimental properties against which to check behavior patterns computed from usual smooth-bar data.

Some questions remain concerning rupture behavior under a general state of stress or even for variable simple tension, but creep-rupture of A-286 at 1200°F appears to be but little affected by prior plastic strains of about 1% and normal creep and rupture properties for this alloy seem to permit ready statement in mathematical forms which lend themselves to machine computations.

Such calculations along the lines suggested in earlier reports may now be profitably pursued, at least for the A-286 alloy solution treated at 1800°F. Present information suggests that the portion of total rupture life "used up" during any calculation interval of assumed constant stress can be most adequately measured by the geometric mean of the fraction of rupture life and the fraction of creep ductility passed through during that interval. Successful prediction of the experimentally-observed trends for rupture of notched bars would supply indirect confirmation for the shear-stress invariant theory or other theory assumed in the computations to correlate rupture under multi-axial stressing. However, should no adequate correlation result between predictions and actual test results on notched bars, one might have difficulty determining whether lack of agreement was due to a fundamental flaw in the overall analysis, improper definition of the criterion for rupture under complex stresses or some combination of these and/or other factors. Independent direct determination of the factors defining rupture time under changing compound stresses would permit much more certain acceptance and generalization of the results from the proposed computations.

A secondary goal of the current research was to determine whether notch strengthening could be obtained in the absence of localized time-independent plastic strains introduced during load application. This has very definitely been demonstrated for specimens of A-286 with the 1800°F solution temperature. Two specimens for K_t = 1.27 at nominal stresses of 47,500 and 50,000 psi never exceeded the proportional limit even at the notch root and yet the notched-bar rupture time for both was double that of smooth bars at corresponding nominal stress level. (Compare Figs. 4 and 10). The lowest-stress tests for K_t = 1.54 and 1.82 with this same treatment had local peak stresses above the proportional limit, but the amount of plastic strain on loading should have been negligible. Decided notch strengthening was obtained for these tests and the general trend of data strongly indicates the same result if the nominal stress level had been somewhat lower.

For the 2200°F solution treatment the uniformly-high degree of notch strengthening precluded tests in which the proportional limit was not exceeded. Again, however, the trend of data seems to assure that definite notch strengthening should still be obtained if the nominal stress were lowered the small amount to keep initial loading stresses completely in the elastic region.

With short-time plastic strains ruled out as the key factor permitting notch strengthening, the conclusion seems inescapable that notch behavior must be determined by something associated with the triaxial state of stress near a notch--either of itself or as modified by the creep process during a test. One striking feature of the findings of this research was the small degree of triaxiality (K_t = 1.2 - 1.5) which produced peak notched-bar rupture strength for A-286 Heat 21,030 with a 2200°F solution treatment. Any acceptable theory for notch behavior must be able to explain why this result was obtained and also why the peak strength with the 1800°F solution temperature was found for somewhat sharper notch; i.e., for a higher degree of triaxiality.

Even without positive present knowledge of the laws governing stress rupture under either variable simple stress or steady complex stress, much insight into notch effects can probably be gained by attempting to explain these behaviors with calculations founded on alternate reasonable assumptions for the unknown laws.

Experimental properties found at 1200°F for other lots of A-286, or for Heat 21,030 at other test temperatures, differed only in detail from those obtained in the major study. The added insight into notch behavior to be gained by extensive research on any of these alternate materials or test temperatures would probably be small in value compared to results from the same amount of effort on studies designed to clarify specific issues raised by the current research.

From both practical and theoretical standpoints, particular value should derive from further investigation of the role of ductility in notch sensitivity. The present study has shown that pronounced notch strengthening is possible in materials exhibiting 1% or less elongation in unnotched specimens at like nominal stress levels. Planned extensive metallographic examination of the fractured specimens for A-286 in the low-ductility condition (2200°F solution treatment) may provide some clues, but new studies would be desirable with an alloy showing an even lower order of elongation in the normal rupture test.

In such studies attention may profitably be directed to overall elongation versus localized deformations in or between individual grains. Indeed, the entire field of microstructure and its influence on fracture behavior merits careful study. A better knowledge of the basic mechanisms of fracture under creep conditions should clear up many of the present mysteries of trace-element effects and the varied effects of structural alternations during test or service.

Without proper allowance for these effects, the search for a general criterion--or even several alternate criteria--for rupture under variable complex stresses may be impossible. In any case, careful investigation of the relationship of microstructure to failure should further an explanation of how plastic prestrain, variable load during test, varied rolling and heat treating operations or small alternations in composition can produce the profound differences which have been noted in a material's response to a stress concentration.

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TABLES

TABLE 1

CHEMICAL ANALYSES AND PROCESSING CONDITIONS FOR THE VARIOUS LOTS OF A-286 ALLOY INVESTIGATED

Heat No.	Type of Melt	C	Mn	Si	Cr	Ni	Mo	Ti	V	Al	S	P	Fe	B
		Chemical Composition, percent by weight												
21,030	Vacuum consumable electrode	0.06	1.35	0.47	14.58	25.30	1.38	2.00	0.21	0.17	0.014	0.018	Bal	*(0.004) (0.0019)
43297	Two-ton air	0.025	1.21	0.86	15.12	26.16	1.32	1.94	0.23	0.40	0.017	0.016	Bal	*(0.002) (0.0008)
52853	Not Stated	*(0.06 (0.064	1.21 1.37	0.66 0.62	15.12 15.55	26.06 26.50	1.38 1.27	2.12 2.12	0.16 0.10	0.20 0.10	0.012 0.010	0.022 0.023	Bal) Bal)	*(0.0016) (<0.001)
K-65X	Vacuum consumable electrode	0.053	1.32	0.54	14.75	26.60	1.38	2.00	0.25	0.146	0.012	0.025	Bal	0.0007

Other:	W	Co	Cu	Pb	O	H	N
	0.13	0.11	0.06	0.0009	0.08	0.0008	0.04

Heat No.	Processing Procedure Prior to Receipt at the University of Michigan
21,030	20" round ingot pressed and cogged to 4-1/4" square from 2150°F. Recogged to 2-7/8" square from 2100°F; Rolled to 3/4" diameter from 2100°F. Bars neither straightened nor annealed before shipping.
43297	Hot rolled from 2050°F; As-rolled hardness BHN 163-179. (Finishing temperature approximately 1650°F). Supplied as 7/8" diameter rounds.
52853	5-1/2" round-corner square billets ground all over; soaked at 2025-2050°F; rolled as a hand round 4-3/4" diameter billet with finishing temperature approximately 1850°F, air cooled. Rough turned to 4-1/2" diameter.
K-65X	19" diameter ingot reduced to 8.5" square from 2075-2100°F.

* Check analyses by different Laboratories.

TABLE 2 - RESULTS OF TESTS AT 1200°F FOR A-286 SPECIMENS WITH DIFFERENT SOLUTION TEMPERATURES

(Heat No. 21,030. Heat Treatment: 1 hr Solution, Oil Quench + Age at 1325°F, 16 hr. Air Cool)

Smooth Bars

Solution Temp. (°F)	Stress (psi)	Plastic Strain on Loading (%)	Rupture Life (hr)	Elongation at Rupture (%)	Reduction of Area (%)
1650	70,000	0.02	15.7	9.5	10.
1650	65,000		23.8	13.	18.
1650	65,000		19.9	8.5	11.5
1650	60,000		45.4	8.5	10.5
1800	100,000	0.36	0.26	17.	18.
1800	90,000	0.075	1.45	8.	9.
1800	80,000	0.045	4.25	8.	8.5
1800	70,000		14.9	9.5	13.5
1800	70,000	0.03	20.3	6.5	8.5
1800	65,000		62.4	5.5	10.5
1800	60,000		99.1	7.	8.5
1800	60,000		79.9	5.	10.
1800	50,000		384.6	6.	8.
1800	45,000		615.5	4.	8.
1800	45,000		771.3	3.5	4.5
1800	40,000		1435.1	5.5	5.5
1900	70,000	0.045	47.4	6.	9.
1900	65,000		112.2	5.	6.5
1900	60,000		302.9	3.5	5.5
1950	65,000		103.0	5.	7.
2000	70,000	0.03	62.3	5.	8.
2000	65,000		198.9	4.	7.5
2000	60,000		493.9	5.	5.5
2050	70,000	0.045	48.8	6.5	9.5
2050	65,000		128.4	4.5	6.
2100	70,000		25.3	3.5	5.5
2100	65,000		118.2	4.	7.
2100	60,000		350.9	3.	7.5
2150	70,000	0.055	33.8	1.5	6.5
2150	65,000		291.6	1.5	4.
2150	65,000		203.4	2.	4.
2200	80,000	0.165	0.87	4.	7.
2200	75,000	0.065	3.8	2.5	5.5
2200	70,000	0.045	7.2	1.5	7.
2200	60,000		308.7	1.	2.5
2200	55,000	0.01	587.9	1.	2.
2200	50,000		2068.3	<0.5	1.
2225	70,000	0.04	15.1	1.5	5.
2225	70,000	0.045	4.8	2.	7.
2225	65,000		117.3	1.5	3.
2225	60,000		316.1	1.	2.5
2300	70,000	0.86	3.8	4.	15.
2300	65,000	0.33	15.2	4.	12.5
2300	60,000		50.8	1.5	6.5

(Prestrained by momentary overloading at test temperature)

Solution Temp. (°F)	Overload Stress (psi)	Plastic Prestrain (%)	Stress (psi)	Rupture Life (hr)	Total Elongation at Rupture (%)	Total Reduction of Area at Rupture (%)
1800	100,000 +	0.80	70,560	17.9	8.	9.5
1800	108,240	0.90	60,540	59.8	6.5	11.
1950	98,000	0.72	70,525	31.3	8.	9.5
1950	100,000	1.33	65,870	90.0	7.	9.
2050	100,000	2.03	61,220	183.3	6.	9.5
2200	84,070	1.31	60,790	99.8	2.5	4.5
2200	100,000	0.97	60,000	a)226.6 +	--	--
2200	82,000	0.37	59,920	a)328 +	--	--
2200	81,000	0.90	60,240	307.0	1.5	4.
2300	83,740	1.56	60,940	17.6	3.5	8.5

c) Notched Bars

Solution Temp. (°F)	Stress (psi)	Rupture Life (hr)
	K_t = 1.27	
1800	70,000	97.2
1800	70,000	22.6
1800	65,000	241.6
1800	60,000	319.3
1800	50,000	840.0
1800	47,500	1043.6
2200	80,000	127.4
2200	75,000	264.5
2200	70,000	996.0
2200	70,000	451.3
2200	65,000	416.4
2200	60,000	417.5
2200	60,000	1242.5
2200	55,000	783.5
	K_t = 1.37	
2200	80,000	242.8
2200	65,000	954.2
	K_t = 1.54	
1800	70,000	215.3
1800	65,000	292.3
1800	60,000	481.3
2200	80,000	225.0
2200	70,000	657.5
2200	60,000	568.7
2200	55,000	4442.6
2200	55,000	4834 + (Test in progress)
	K_t = 1.8	
1650	70,000	106.3
1650	65,000	245.6
1800	80,000	109.5
1800	70,000	239.6
1800	65,000	366.2
1800	60,000	543.4
1800	55,000	994.3
1950	70,000	716.1
1950	65,000	814.3
2050	70,000	563.7
2050	65,000	688.9 + b)
2150	70,000	565.8
2150	65,000	660.1
2200	70,000	206.3
2200	65,000	257.7
2200	60,000	1151.1
2200	55,000	727.4
2225	70,000	143.3
2225	65,000	602.9
2300	70,000	41.2
2300	65,000	62.8

Solution Temp. (°F)	Stress (psi)	Rupture Life (hr)
	K_t = 2.39	
1800	62,000	426.0
	K_t = 3.0	
1650	70,000	76.9
1800	70,000	167.5
1800	45,000	908.1
1950	70,000	320.7
2050	70,000	296.4
2150	70,000	127.6
2200	80,000	203.6
2200	65,000	219.3
2200	60,000	414.2
2200	50,000	2724.9
2225	70,000	26.3
2225	65,000	44.9
2300	70,000	22.3
2300	65,000	23.8
	K_t = 4.1	
1800	80,000	20.9
1800	70,000	112.8
1800	65,000	168.1
1800	60,000	227.6
1800	50,000	363.0
1800	45,000	490.4
1800	40,000	508.8
1800	35,000	1651.5
2200	70,000	14.2
2200	65,000	43.5
2200	60,000	288.5
2200	52,500	140.1
	K_t = 5.7	
1800	70,000	27.3
1800	65,000	106.7
d)1800	65,000	43.4
1800	60,000	119.7
1800	55,000	257.1
1800	45,000	464.8
1800	40,000	1295.7
	K_t = 8.7-10	
d)1800	70,000	6.5
d)1800	65,000	6.65
d)1800	60,000	11.1
1800	40,000	451.9

a) Overheated -- discontinued
b) Discontinued (Controller Failure)
c) Nominal Notch Geometries (Inches)

Theoretical stress concentration factor, K_t:	1.27	1.37	1.54	1.82	2.39	3.0	4.1	5.7	8.7-10
Diameter of shank, D:	0.462	0.336	0.462	0.600	0.500	0.460	0.500	0.600	0.600
Minimum diameter, at base of notch, d:	0.326	0.248	0.327	0.424	0.350	0.325	0.350	0.424	0.424
Notch root radius, r:	0.237	0.109	0.108	0.081	0.032	0.017	0.009	0.005	0.0015-0.002

d) Aged after notching operation

TABLE 3

TEST RESULTS AT 1100° AND 1300°F FOR A-286 HEAT 21,030

Heat Treatment: 1 hr Solution, Oil Quench + Age at 1325°F, 16 hr, Air Cool

Solution Temp. (°F)	Test Temp. (°F)	Stress (psi)	Rupture Life (hr)	Total Elong. at Rupture (%/2 in.)	Reduction of Area at Rupture (%)	Remarks
					SMOOTH BARS	
1800	1100	85,000	31.5	4.	8.5	
1800	1100	75,000	176.8	4.	7.5	
1800	1100	75,070	102.4	4.	7.5	1.1% Plastic prestrain by momentary overload to 110,000 psi
2200	1100	85,000	23.1	4.	6.5	1.65% Plastic strain on loading to test stress
2200	1100	75,000	642.6	1.5	3.5	
2200	1100	85,400	54.7	2.5	8.5	1.43% Plastic prestrain by momentary overload to 95,000 psi
2200	1100	76,000	20.1	4.	7.5	2.23% Plastic prestrain by momentary overload to 96,370 psi
1800	1300	45,000	44.6	14.	24.	
2200	1300	45,000	303.3	4.	6.5	
2200	1300	44,970	262.1	5.	10.	1.05% Plastic prestrain by momentary overload to 70,000 psi
					a)NOTCHED BARS	
1800	1100	85,000	43.6	--	--	
2200	1100	85,000	126.4	--	--	
1800	1300	45,000	167.2	--	--	
2200	1300	45,000	740.2	--	--	

a)Nominal Notch Geometry: Diameter of Shank, D = 0.460 inch
Min. drain, at base of notch, d = 0.325 inch
Notch root radius, r = 0.017 inch
Theoretical stress conc., K_t = 3.0

TABLE 4

RESULTS OF SURVEY TESTS AT 1200°F ON A-286 HEAT 43,297

Heat Treatment: 1 hr Solution, Oil Quench + 16 hr Age at 1325°F, Air Cool

Solution Temp. (°F)	Stress (psi)	Rupture Life (hrs)	Elongation at Rupture (%/2 in.)	Reduction of Area (%)	Remarks
			SMOOTH BARS		
1650	65,000	24.9	10.5	14.5	
1650	60,000	41.9	8.5	12.5	
1650	55,000	75.4	9.	13.	
1650	55,000	30.1	10.5	19.	1.0% plastic prestrain by momentary initial overload to 108,060 psi
1800	65,000	29.2	4.	6.5	
1800	55,000	121.4	9.	11.5	
1800	59,850	34.4	4.	7.5	0.76% plastic prestrain by momentary initial overload to 104,520 psi
1985	65,000	73.5	3.	5.5	
1985	55,000	385.5	4.	5.5	
1985	55,020	176.2	1.5	4.5	1.13% plastic prestrain by momentary initial overload to 98,000 psi
			[a]NOTCHED BARS		
1650	70,000	86.5	--	--	
1650	65,000	126.8	--	--	
1650	60,000	109.2	--	--	
1800	65,000	60.4	--	--	
1800	60,000	124.9	--	--	
1800	55,000	284.1	--	--	
1985	65,000	4.0	--	--	
1985	55,000	270.8	--	--	

[a] Nominal Geometry: Diameter of Shank, D = 0.460 inch
Min. drain, at base of notch, d = 0.325 inch
Notch root radius, r = 0.017 inch
Theoretical stress conc., K_t = 3.0

TABLE 5

RESULTS OF SURVEY TESTS AT 1200°F ON A-286 HEAT 52853

Heat Treatment: 1 hr Solution, Oil Quench + 16 hr Age at 1325°F, Air Cool

(Rolled to 1/2-inch squares at the University of Michigan prior to heat treatment)

UNNOTCHED SPECIMENS

Rolling Temp. (°F)	Stress (psi)	Rupture Life (hours)	Elongation at Rupture (%)	Reduction of Area (%)
		1800°F Solution Temperature		
1800	65,000	15.5	8.	11.
1950	65,000	16.4	6.	7.
1950	60,000	60.5	5.	11.
1950	55,000	125.2	5.5	8.5
2100	70,000	8.7	7.5	10.
2100	60,000	27.4	7.	11.
2100	50,000	145.0	8.	9.5
		2000°F Solution Temperature		
2100	70,000	11.7	3.	6.
2100	65,000	114.5	5.	6.5
2100	60,000	281.1	3.5	7.
		2200°F Solution Temperature		
2100	70,000	24.2	2.	3.
2100	60,000	120.4	2.	3.5
2100	55,000	501.4	1.5	2.

NOTCHED SPECIMENS

Rolling Temp. (°F)	Stress (psi)	Rupture Life (hours)	Rolling Temp. (°F)	Stress (psi)	Rupture Life (hours)
		1800°F Solution Temperature			
	K_t = 1.27			K_t = 4.1	
1800	65,000	81.1	1800	65,000	21.8
1800	60,000	220.7	1800	60,000	88.9
1950	70,000	55.5	1950	65,000	17.4
1950	65,000	140.5	1950	60,000	45.6
			1950	55,000	63.8
2100	70,000	41.8	2100	70,000	16.0
2100	60,000	115.5	2100	60,000	58.2
		2000°F Solution Temperature			
	K_t = 3.1				
2100	75,000	381.1			
2100	70,000	531.6			
2100	65,000	793.4			
		2200°F Solution Temperature			
	K_t = 1.72				
2100	75,000	73.1			
2100	70,000	171.0			
2100	65,000	389.6			

TABLE 6

RESULTS OF SURVEY TESTS AT 1200°F ON A-286 HEAT K65X

Heat Treatment: 1800°F, 1 hr, Oil Quench + 1325°F, 16 hr, Air Cool

UNNOTCHED SPECIMENS

Stress (psi)	Rupture Life (hours)	Elongation at Rupture (%)	Reduction of Area (%)
Specimens Sampled Lengthwise from the Billet Supplied			
70,000	49.1	2.	3.5
60,000	192.6	2.	2.5
55,000	300.4	1.	1.5
Material Rerolled from 2100°F Before Sampling			
65,000	35.6	3.5	7.
55,000	116.8	4.5	5.5
50,000	209.6	4.	5.

NOTCHED SPECIMENS

Stress (psi)	Rupture Life (hours)	Stress (psi)	Rupture Life (hours)
$K_t = 1.72$		$K_t = 4.1$	
Specimens Sampled Lengthwise from the Billet Supplied			
65,000	137.0	70,000	41.8
55,000	232.3	65,000	109.6
		60,000	90.8
		55,000	35.3
		55,000	44.9
		50,000	128.4
Material Rerolled from 2100°F Before Sampling			
		65,000	36.1
		55,000	87.8
		50,000	146.1

TABLE 7 - CREEP-RATE FACTORS AT SELECTED FRACTIONS OF RUPTURE TIME FOR A-286 HEAT 21,030 TESTED AT 1200°F

Stress (psi)	Minimum Creep Rate (in/in/hr)	Time Fraction at Minimum Creep Rate	Creep-Rate Factor = Ratio: $\frac{\text{Existing Creep Rate}}{\text{Minimum Creep Rate}}$ at Indicated Fraction of Experimental Rupture Time											
			0.005	0.01	0.02	0.05	0.10	0.20	0.40	0.60	0.80	0.90	0.95	0.98
							1800°F Solution Temperature							
100,000	0.425	0.18	--	--	1.07	1.035	1.01	1.00	1.08	1.25	1.78	2.27	3.25	--
90,000	0.013	0.04	2.69	1.54	1.19	1.07	1.22	1.91	3.17	3.94	4.51	5.14	5.84	6.30
80,000	0.0030	0.06	2.57	1.83	1.34	1.02	1.09	1.61	--	--	--	--	--	--
70,000	0.00080	0.08	--	3.00	1.88	1.21	1.04	1.74	4.56	8.12	--	--	--	--
70,000	0.00042	0.05	5.95	2.93	1.91	1.00	1.30	1.55	--	--	--	--	--	--
65,000	0.000085	0.11	--	--	3.09	1.81	1.04	1.27	2.40	6.71	12.7	21.77	--	--
60,000	0.000060	0.175	7.34	5.40	4.40	2.12	1.42	1.02	3.35	8.34	--	--	--	--
60,000	0.000061	0.14	3.87	2.84	2.07	1.34	1.03	1.07	1.71	4.92	23.8	--	--	--
50,000	0.0000079	0.08	3.39	2.23	1.52	1.12	1.00	1.30	3.29	8.48	26.4	53.20	65.81	89.9
45,000	0.0000062	0.04	2.09	1.31	1.10	1.03	1.11	1.31	2.34	6.70	18.4	33.57	42.28	--
45,000	0.000002	0.05	--	--	--	--	1.32	3.08	5.05	13.80	31.0	88.00	116.7	128.5
40,000	0.000002	0.05	--	--	--	1.25	1.50	2.15	4.45	9.50	24.0	51.0	87.0	153.5
	Average:	0.09	3.98	2.64	1.96	1.27	(1.0)	1.58	3.14	(8.32)	(22.72)	(49.50)	(77.9)	(124.0)
							2200°F Solution Temperature							
80,000	0.0185	0.65	13.5	10.8	7.3	4.0	2.76	1.68	1.24	1.0	1.03	1.22	2.52	
75,000	0.00220	0.15	--	1.74	1.55	1.25	1.09	1.0	1.13	1.30	1.48	1.62	--	
70,000		--	--	3.16	2.47	1.0	--	--	--	--	--	--	--	
60,000	0.0000089	--	--	--	--	--	--	(1.0)	1.09	1.88	2.56	2.98	--	
55,000	0.0000048	0.25	--	--	--	1.23	1.15	1.08	1.0	1.27	2.38	3.44	4.09	
50,000	0.000000925	0.30	--	--	1.26	1.23	1.15	1.04	1.0	1.12	2.62	3.24	3.92	
							2225°F Solution Temperature							
70,000	0.0017	0.20	--	11.2	5.5	2.3	1.41	1.0	--	--	--	--	--	
70,000	0.000172	--	--	--	--	--	--	--	--	--	--	--	--	
65,000	0.000025	0.15	--	--	--	--	1.00	1.00	1.08	1.36	2.32	2.74	2.80	
60,000	0.0000040	0.18	--	3.60	2.15	1.60	1.15	1.00	1.60	2.25	2.52	--	--	
	Average:	0.2 +	--	6.10	3.37	1.80	1.38	(1.00)	1.16	1.45	2.13	2.54	3.33	
							1800°F Solution Temperature (Prestrain Tests)							
70,560	0.00125	0.065	2.56	2.00	1.31	1.12	1.04	1.33	1.92	--	--	--	--	--
60,540	0.00023	0.085	4.35	3.22	2.13	1.10	1.01	1.17	2.27	4.04	6.65	8.22	9.57	--
							2200°F Solution Temperature (Prestrain Tests)							
60,790	0.000027	0.10	3.15	2.91	1.91	1.28	1.00	1.09	1.37	1.81	2.26	2.52	--	--
60,240	0.0000087	0.15	4.94	2.84	2.17	1.36	1.00	1.00	1.05	1.40	2.39	3.43	--	--
60,000	0.000058													
59,920	0.000036													

TABLE 8 - MULTIPLE-STRESS DATA FROM PAST STUDIES, EXAMINED ON THE BASIS OF RUPTURE-TIME FRACTIONS.

Inconel X-550 at 1350°F

Stress Level (psi)	Actual Time at Stress / Rupture Life for Stress	
38,060	404.6/670 =	0.60
50,000	56.3/122 =	0.46
		1.06
41,190	321.4/420 =	0.77
50,000	47.8/122 =	0.39
		1.16
44,300	200.5/280 =	0.72
50,000	13.6/122 =	0.11
		0.83
50,000	48.6/122 =	0.40
70,000	2.0/6.0 =	0.33
		0.73
47,410	166/185 =	0.90
40,000	204.3/570 =	0.36
		1.26
53,680	40.8/75 =	0.54
40,000	392.9/570 =	0.68
		1.22
56,820	20.9/50 =	0.42
40,000	265.9/570 =	0.47
		0.89
70,000	2.0/6.0 =	0.33
50,000	88.0/122 =	0.72
		1.05

Waspaloy at 1500°F

Stress Level (psi)	Actual Time at Stress / Rupture Life for Stress	
60,000	0.177/0.46 =	0.38
30,000	34.0/215 =	0.76
		1.14
40,000	3.4/8.0 =	0.42
30,000	22.4/45 =	0.50
20,000	164.5/525 =	0.31
		1.23
20,000	107.3/525 =	0.21
40,000	3.4/8.0 =	0.42
20,000	177.2/525 =	0.34
		0.97
40,000	5.1/8.0 =	0.64
20,000	287.3/525 =	0.55
		1.19
20,000	165/525 =	0.315
40,000	2.5/8.0 =	0.315
20,000	182.4/525 =	0.35
		0.98
10,000	215/33,500 =	0.01
30,000	44.2/45 =	0.99
		1.00
18,000	263.2/1000 =	0.26
30,000	22.67/45 =	0.51
18,000	172.9/1000 =	0.17
		0.94

Annealed Carbon Steel at 900°F

Stress Level (psi)	Actual Time at Stress / Rupture Life for Stress	
20,000	5.5/35 =	0.16
15,000	469.4/650 =	0.72
		0.88
15,000	241.3/650 =	0.37
20,000	17.5/35 =	0.50
		0.87

Annealed Carbon Steel at 1050°F

Stress Level (psi)	Actual Time at Stress / Rupture Life for Stress	
7,500	114.5/220 =	0.52
6,000	304/700 =	0.43
		0.95
6,000	135.3/700 =	0.19
10,000	23.7/47 =	0.51
6,000	174.2/700 =	0.25
		0.95
6,000	296.5/700 =	0.42
10,000	21.2/47 =	0.45
		0.87
12,000	6.5/15 =	0.43
10,000	25.3/47 =	0.54
		0.97
10,000	22.4/47 =	0.48
6,000	291.9/700 =	0.42
		0.90

2024-T4 at 400°F

Stress Level (psi)	Actual Time at Stress / Rupture Life for Stress	
20,000	173/460 =	0.38
30,000	0.4/18 =	0.02
		0.40
10,000	95/10,000) =	0.01
30,000	0.375/18 =	0.02
		0.03
30,000	10/18 =	0.56
20,000	277.2/460 =	0.60
		1.16

2024-T4 at 500°F

Stress Level (psi)	Actual Time at Stress / Rupture Life for Stress	
15,000	10.5/24.5 =	0.43
10,000	168.75/165 =	1.02
		1.45

S-816 at 1350°F

Stress Level (psi)	Actual Time at Stress / Rupture Life for Stress	
45,000	9.8/32 =	0.31
35,000	93.9/210 =	0.45
		0.76
55,000	1.33/4.8 =	0.28
45,000	8.42/32 =	0.26
35,000	80.05/210 =	0.38
		0.92
35,000	68/210 =	0.32
45,000	10.0/32 =	0.31
35,000	56.2/210 =	0.27
		0.91
*30,000	335.3/600 =	0.56
45,000	16/32 =	0.50
		1.06

* **Alternate 11 hr at 30,000 and 1 hr at 45,000 psi**

17-22A(S) Steel, at 1100°F

Heat 31158

Stress Level (psi)	Actual Time at Stress / Rupture Life for Stress	
25,000	131.3/387 =	0.34
65,000	0.085/4.7 =	0.02
		0.36
25,000	197/387 =	0.51
60,000	0.66/8.8 =	0.08
		0.59
60,000	5.5/8.8 =	0.62
25,000	238.3/387 =	0.62
		1.24
60,000	1.1/8.8 =	0.12
50,000	3.9/35.0 =	0.11
30,000	40.0/205 =	0.20
20,000	114.7/830 =	0.14
		0.57
60,000	3.67/8.8 =	0.42
25,000	130/387 =	0.34
60,000	3.0/8.8 =	0.34
		1.10

17-22A(S) Steel, at 1100°F

Heat 26157

Stress Level (psi)	Actual Time at Stress / Rupture Life for Stress	
60,000	0.05/1 =	0.05
50,000	1/25 =	0.04
40,000	79/82 =	0.96
		1.05
50,000	2.4/25 =	0.10
40,000	20.3/82 =	0.25
30,000	125.5/320 =	0.40
		0.75
46,000	11.3/45 =	0.25
36,000	26/135 =	0.19
26,000	274.4/520 =	0.53
		0.97
45,000	17/50 =	0.34
35,000	77/154 =	0.50
		0.84
25,000	111.1/550 =	0.20
35,000	109.7/154 =	0.71
		0.91
Relaxation from 45,200 to 36,200 lb./sq. in. in 12 min.		0.01
36,200	133.1/130 =	1.02
		1.03

TABLE 9 - RESULTS OF MULTIPLE-STRESS TESTS, COMPARED BY DIFFERENT CRITERIA FOR CREEP-RUPTURE

Initial Plastic Strain (in./in.)	Stress Level (psi)	ΔT = Time Period at Stress (hr)	R = Rupture Life at Stress (hr)	ΔT/R	ΔC = Creep during Period (in./in.)	D = Creep to Rupture in Normal Test at Stress, (in./in.)	ΔC/D	$\sqrt{\frac{(\Delta T)}{(R)}\frac{(\Delta C)}{(D)}}$	A = Experimental Creep Rate at New Stress (in./in./hr)	B = Corresponding Normal Creep Rate (in./in./hr)	A/B
					A-286 at 1200°F (2200°F Solution Temperature)						
0.000	50,000	50.0	2000.0	0.025	0.00005	0.0053	0.01	0.02			
	55,000	52.0	773.8	0.065	0.00015	0.0073	0.02	0.04	0.0000037	0.0000069	0.54
	60,000	236.6	325.2	0.73	(0.01)	0.010	1.00	0.85	0.0000067	0.000020	3.35
				0.82			1.03	0.91			
0.000	55,000	164.0	773.8	0.21	0.000068	0.0073	0.01	0.05			
	60,000	76.6	325.2	0.24	0.000414	0.010	0.04	0.10	0.000022	0.000015	1.47
	65,000	92.5	98.6	0.94	(0.0067)	0.0135	0.50	0.69	0.000043	0.000099	0.44
				1.39			0.55	0.84			
0.000	50,000	600.0	2000.0	0.30	0.00044	0.0053	0.08	0.15			
	70,000	32.7	18.4	1.78	(0.015)	0.0186	0.81	1.20	0.00059	0.00056	1.05
				2.08			0.89	1.35			
0.000	55,000	210.0	773.8	0.27	0.0012	0.0073	0.16	0.21			
	65,000	15.0	98.6	0.15	0.00061	0.0135	0.05	0.08	0.000125	0.000090	1.39
	55,000	612.3	773.8	0.79	(0.0022)	0.0073	0.30	0.49	0.0000021	0.0000023	0.91
				1.21			0.51	0.78			
0.000	65,000	15.0	98.6	0.15	0.000488	0.0135	0.04	0.08			
	55,000	210.0	773.8	0.27	0.000423	0.0073	0.06	0.13	0.0000026	0.0000025	1.04
	65,000	73.9	98.6	0.75	(0.0042)	0.0135	0.31	0.48	0.000051	0.000099	0.52
				1.17			0.41	0.69			
0.000	65,000	52.7	98.6	0.53	0.00202	0.0135	0.15	0.28			
	60,000	159.1	325.2	0.49	(0.005)	0.010	0.50	0.50	0.0000083	0.000018	0.46
				1.02			0.65	0.78			
0.0004	70,000	0.5	18.4	0.03	0.00185	0.0186	0.10	0.06			
	65,000	5.0	98.6	0.05	0.0001	0.0135	0.01	0.02			
	60,000	351.1	325.2	1.08	(0.007)	0.010	0.70	0.87	0.0000032	0.000015	0.21
				1.16			0.81	0.96			
0.0041	70,000	1.0	18.4	0.05	0.00916	0.0186	0.05	0.05			
	60,250	24.0	312.0	0.08	0.00046	0.0101	0.05	0.06			
	50,200	2232.6	1922.	1.16	(0.025)	0.0054	0.46	0.72			
				1.29			0.56	0.83			
0.0073	76,000	0.02	2.87	0.01	0.00178	0.0272	0.06	0.02			
	72,600	0.05	8.09	0.01	0.00152	0.022	0.07	0.03			
	68,500	3.45	30.1	0.11	0.00047	0.0169	0.03	0.02			
	64,500	26.0	117.5	0.22	0.00050	0.0133	0.04	0.09			
	60,400	521.7	304.4	1.71	(0.009)	0.0102	0.88	1.23			
				2.06			1.08	1.39			
0.0162	80,000	0.004	0.90	0.02	0.000179	0.035	0.01	0.01			
	73,200	0.096	6.71								
	69,100	1.1	24.7	0.04	0.000253	0.0177	0.01	0.02			
	65,000	14.0	98.6	0.14	0.000663	0.0135	0.05	0.08			
	61,000	353.8	275.8	1.28	(0.006)	0.0105	0.57	0.85			
				1.48			0.64	0.96			
		Average Values:		1.37			0.71	0.95			
		Maximum Range		+0.71			+0.37	+0.44			
		of Deviation:		-0.55			-0.30	-0.26			
					Inconel X-550 at 1350°F						
	50,000	48.6	122.	0.40	0.0006	0.010	0.06	0.15			
	70,000	2.0	6.0	0.33	0.064	0.041	1.56	0.72			
				0.73			1.62	0.87			
	70,000	2.0	6.0	0.33	0.0037	0.041	0.09	0.17			
	50,000	88.0	122.	0.72	0.016	0.010	1.60	1.07			
				1.05			1.69	1.24			
	44,300	200.5	280.	0.72	0.0017	0.0088	0.19	0.37			
	50,000	13.6	122.	0.11	0.013	0.010	1.30	0.38			
				0.83			1.49	0.75			
	53,680	40.8	75.	0.54	0.0015	0.0155	0.10	0.23			
	40,000	392.9	570.	0.68	0.0085	0.0080	1.06	0.85			
				1.22			1.16	1.08			
	47,410	166.0	185.	0.90	0.0016	0.0095	0.17	0.39			
	40,000	204.3	570.	0.36	0.0035	0.0080	0.44	0.40			
				1.26			0.61	0.79			
	41,190	321.4	420.	0.77	0.0010	0.0082	0.12	0.30			
	50,000	47.8	122.	0.39	0.019	0.010	1.90	0.86			
				1.16			2.02	1.16			
	56,820	20.9	50.	0.42	0.0025	0.020	0.12	0.23			
	40,000	265.9	570.	0.47	0.0075	0.0080	0.94	0.83			
				0.89			1.06	1.06			
	38,060	404.6	670.	0.60	0.0007	0.0075	0.09	0.24			
	50,000	56.3	122.	0.46	0.010	0.010	1.00	0.68			
				1.06			1.09	0.92			
		Average Value:		1.03			1.34	0.98			
		Maximum Range		+0.23			+0.68	+0.26			
		of Deviation:		-0.30			-0.73	-0.23			

TABLE 9 - CONTINUED

RESULTS OF MULTIPLE-STRESS TESTS, COMPARED BY DIFFERENT CRITERIA FOR CREEP-RUPTURE

Initial Plastic Strain (in./in.)	Stress Level (psi)	ΔT = Time Period at Stress (hr)	R = Rupture Life at Stress (hr)	ΔT/R	ΔC = Creep during Period (in./in.)	D = Creep to Rupture in Normal Test at Stress, (in./in.)	ΔC/D	$\sqrt{\frac{(\Delta T)}{(R)}\frac{(\Delta C)}{(D)}}$	A = Experimental Creep Rate at New Stress (in./in./hr)	B = Corresponding Normal Creep Rate (in./in./hr)	A/B
					A-286 at 1200°F (1800°F Solution Temperature)						
0.000	45,000	164.0	677.7	0.24	0.00135	0.047	0.03	0.08	0.000059	0.000035	1.69
	55,000	48.0	177.8	0.27	0.00383	0.056	0.07	0.14	0.00126	0.00078	1.62
	65,000	17.6	45.8	0.38	(0.05)	0.067	0.75	0.53			
		Average:		0.89			0.85	0.75			
0.000	50,000	40.0	335.7	0.12	0.00070	0.051	0.01	0.03	0.000064	0.000025	2.56
	55,000	27.0	177.8	0.15	0.00135	0.056	0.02	0.06	0.00013	0.000104	1.25
	60,000	56.5	99.6	0.57	(0.0575)	0.062	0.93	0.73			
		Average:		0.84			0.96	0.82			
0.000	50,000	72.0	335.7	0.21	0.00128	0.051	0.02	0.05	0.00014	0.000080	1.75
	60,000	18.0	99.6	0.18	0.0028	0.062	0.05	0.10	0.00145	0.000135	1.07
	70,000	10.2	19.3	0.53	(0.055)	0.075	0.73	0.62			
		Average:		0.92			0.80	0.78			
0.000	50,000	105.0	335.7	0.31	0.00315	0.051	0.06	0.14	0.00035	0.00031	1.13
	65,000	12.0	45.8	0.26	0.00792	0.067	0.12	0.18	0.000091	0.000063	1.44
	50,000	146.5	335.7	0.44	(0.039)	0.051	0.78	0.58			
		Average:		1.01			0.96	0.90			
0.000	65,000	12.0	45.8	0.26	0.00219	0.067	0.03	0.09	0.0000105	0.0000140	0.75
	50,000	105.0	335.7	0.31	0.00104	0.051	0.02	0.08	0.00046	0.00082	0.56
	65,000	24.8	45.8	0.54	(0.065)	0.067	0.97	0.72			
		Average:		1.11			1.02	0.89			
0.000	50,000	240.5	335.7	0.72	0.0037	0.051	0.07	0.22			
	75,000	5.0	8.6	0.58	(0.05)	0.082	0.61	0.59			
		Average:		1.30			0.68	0.81			
0.00025	75,000	3.0	8.6	0.35	0.0082	0.082	0.10	0.19	0.000036	0.000021	1.72
	50,000	202.5	335.7	0.60	(0.04)	0.051	0.78	0.68			
				0.95			0.88	0.87			
0.0011	90,000	0.1	1.03	0.10	0.00622	0.109	0.06	0.07			
	60,110	42.8	98.3	0.44	(0.063)	0.062	1.02	0.67			
				0.54			1.08	0.74			
0.0018	90,000	0.1	1.03	0.10	0.01105	0.109	0.10	0.10			
	80,140	0.9	3.98	0.23	0.0415	0.090	0.46	0.33			
	70,125	2.4	18.9	0.13	(0.03+)	0.075	0.40+	0.23			
				0.46			0.96+	0.66			
0.00135	90,000	0.1	1.03	0.10	0.00932	0.109	0.09	0.09			
	80,110	0.55	3.99	0.14	0.0233	0.090	0.26	0.19			
	70,090	3.25	19.0	0.17	0.0506	0.075	0.67	0.34			
	60,080	0.1	98.7	0.00	(0.01)	0.062	0.16	0.01			
				0.41			1.18	0.53			
0.0040	100,000	0.0167									
0.0040	100,000	0.0167	0.30	0.06	0.00718	0.132	0.05	0.05			
	60,240	38.5	96.9	0.40	(0.109)	0.062	1.76	0.84			
				0.46			1.82	0.89			
0.00505	100,000	0.0167	0.30	0.06	0.00952	0.132	0.07	0.06			
	90,500	0.1	0.96	0.10	0.0119	0.110	0.11	0.10			
	80,400	0.983	3.83	0.26	0.0362	0.090	0.40	0.32			
	70,400	2.35	18.0	0.13	(0.044)	0.076	0.58	0.27			
				0.55			1.16	0.75			
0.00295	100,000	0.004	0.30	0.01	0.00287	0.132	0.02	0.01			
	90,270	0.021	9.91	0.02	0.00091	0.110	0.01	0.01			
	80,240	0.1	3.92	0.03	0.00116	0.090	0.01	0.02			
	70,210	0.375	18.6	0.02	0.00105	0.075	0.01	0.01			
	60,180	2.5	97.6	0.03	0.0013	0.062	0.02	0.02			
	50,150	187.6	329.1	0.57	(0.071)	0.051	1.39	0.89			
				0.68			1.46	0.96			
		Average Value:		**0.78**			**1.06**	**0.80**			
		Maximum Range of Deviation:		**+0.51 -0.37**			**+0.76 -0.38**	**+0.16 -0.27**			

TABLE 10

SMOOTH-BAR RUPTURE TESTS AT 1350°F FOR THREE HEATS OF WASPALOY

Conventional H. T.

Heat Number	Momentary Overload Stress (psi)	Measured Plastic Strain from Overload (%)	Test Stress (psi)	Rupture Life (hours)	Total Elongation at Rupture (%)	Total Reduction of Area at Rupture (%)
63,559	--	--	70,000	4.55	1.	5.
63,559	--	--	50,000	90.9	1.5	3.5
63,559	--	--	40,000	832.8	1.5	3.5
63,559	98,000	0.725	50,360	14.4	1.5	4.
63,559	95,000	0.515	50,250	26.9	1.5	3.5
63,559	104,620	1.36	45,200	21.1	2.	2.
63,561	--	--	70,000	5.7	3.	5.
63,561	--	--	55,000	43.3	1.5	5.
63,561	--	--	42,000	1007.3	1.5	4.5
63,561	98,000	1.14	44,800	50.7	2.	3.5
63,561	95,000	0.59	50,300	53.2	2.5	3.
63,561	95,000	0.78	50,390	34.7	2.	4.5
63,561	99,000	1.01	41,800	120.6	2.5	2.5
63,613	--	0	75,000	1.6	3.	3.5
63,613	--	0	40,000	587.3	--	3.5
63,613	98,000	0.765	69,820	1.1	4.	6.5
63,613	98,000	0.88	64,900	2.4	2.	6.5
63,613	98,000	0.515	59,670	9.85	--	4.5
63,613	98,000	0.93	52,140	16.0	2.5	3.5
63,613	98,000	0.44	44,730	174.85	1.5	2.
63,613	100,000	1.02	44,850	32.5	1.5	4.

1550°F Age Omitted

Heat Number	Momentary Overload Stress (psi)	Measured Plastic Strain from Overload (%)	Test Stress (psi)	Rupture Life (hours)	Total Elongation at Rupture (%)	Total Reduction of Area at Rupture (%)
63,559	--	--	70,000	0.95	1.	5.5
63,559	--	--	50,000	19.2	2.	2.5
63,559	--	--	40,000	308.1 +	--	--
63,559	95,000	1.23	50,650	1.5	2.	3.5
63,559	92,000	0.74	50,370	3.6	2.5	3.
63,559	90,000	0.51	45,030	17.7	1.	1.
63,561	--	--	70,000	4.9	--	6.5
63,561	--	--	55,000	19.0	1.5	6.
63,561	--	--	42,000	614.3	3.	4.5
63,561	92,000	0.52	44,930	50.3	1.5	2.
63,561	98,000	1.44	50,720	7.4	3.	4.
63,561	92,000	0.51	50,250	20.8	2.	4.
63,561	95,000	1.25	42,100	17.3	3.	4.
63,613	--	0	75,000	2.4	2.5	8.5
63,613	--	0	45,000	579.6	5.	3.5
63,613	100,000	2.2	70,240	0.12	4.	8.
63,613	98,000	0.93	64,940	4.35	3.5	6.
63,613	98,000	1.33	60,210	3.85	2.5	4.5
63,613	98,000	1.27	52,300	5.3	3.5	3.5
63,613	98,000	1.72	45,340	9.7	2.5	3.5
63,613	95,000	0.45	44,700	42.4	1.5	3.

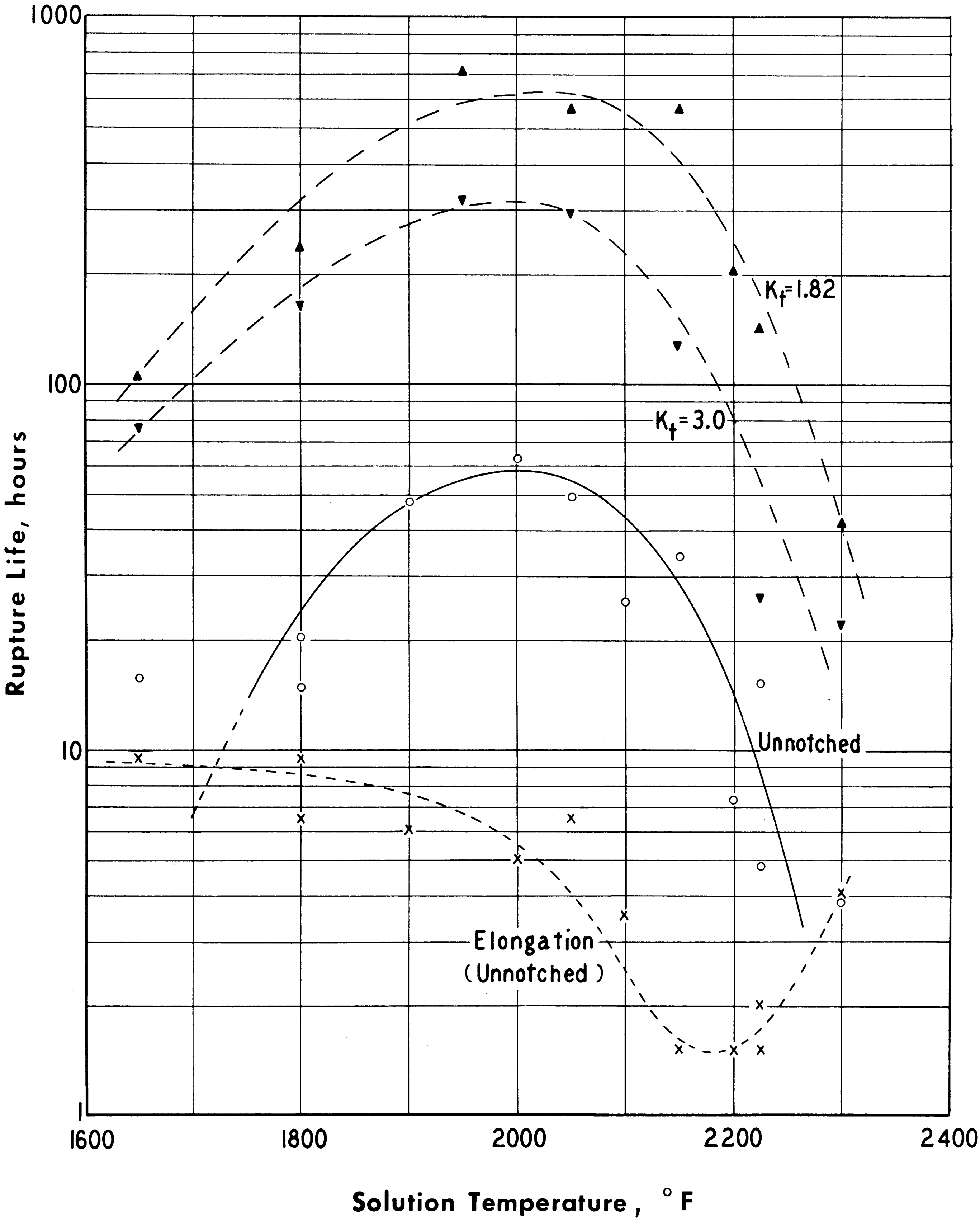


Fig. 1 - Effect of Solution Temperature on Rupture Time and on Smooth-Bar Rupture Ductility of A-286 Heat 21,030 Under 70,000 Psi Stress at 1200°F.

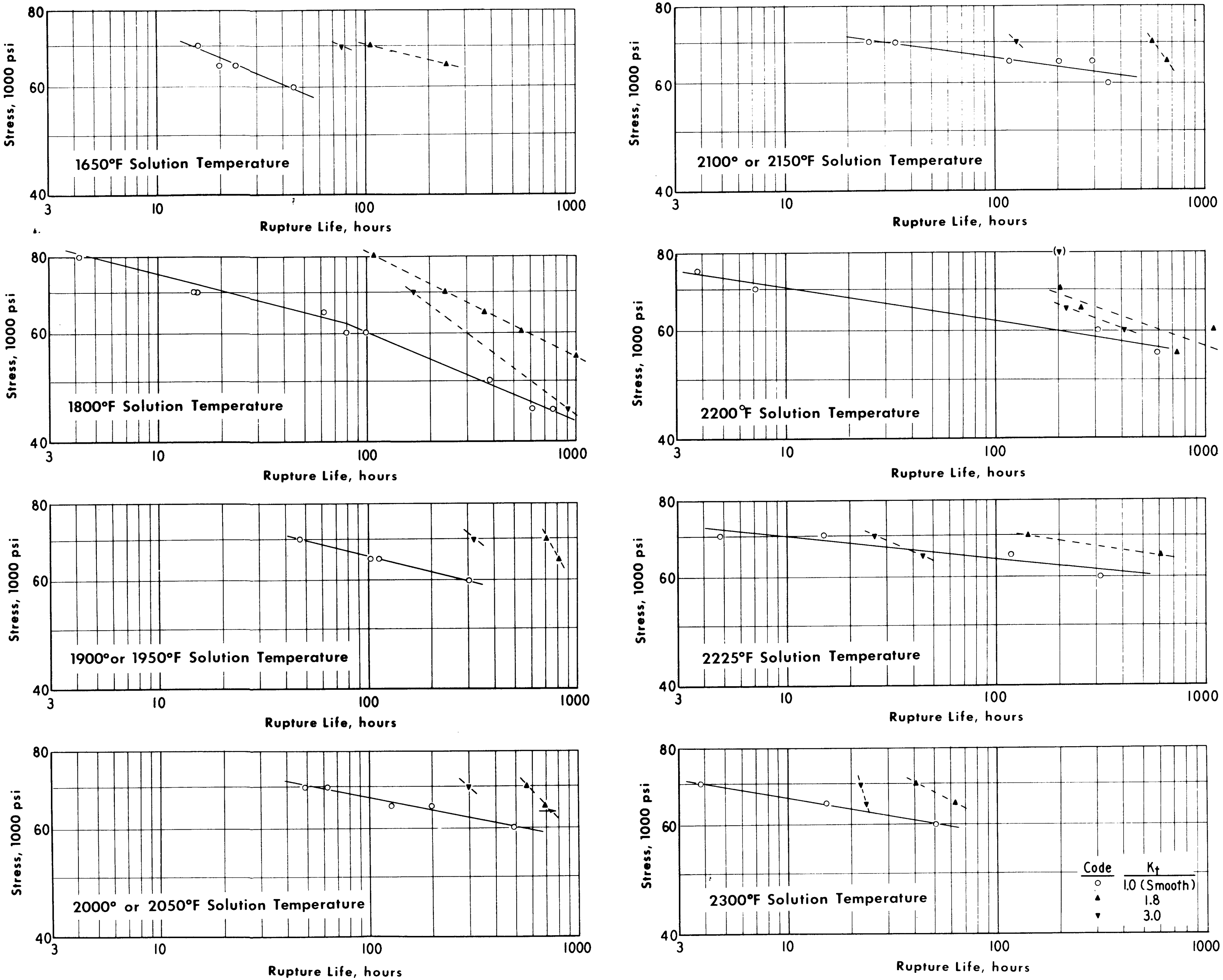


Fig. 2 - Comparative Rupture Properties at 1200°F for A-286 Specimens With Various Solution Temperatures. (Heat 21,030, K_t = 1.0, 1.8, 3.0).

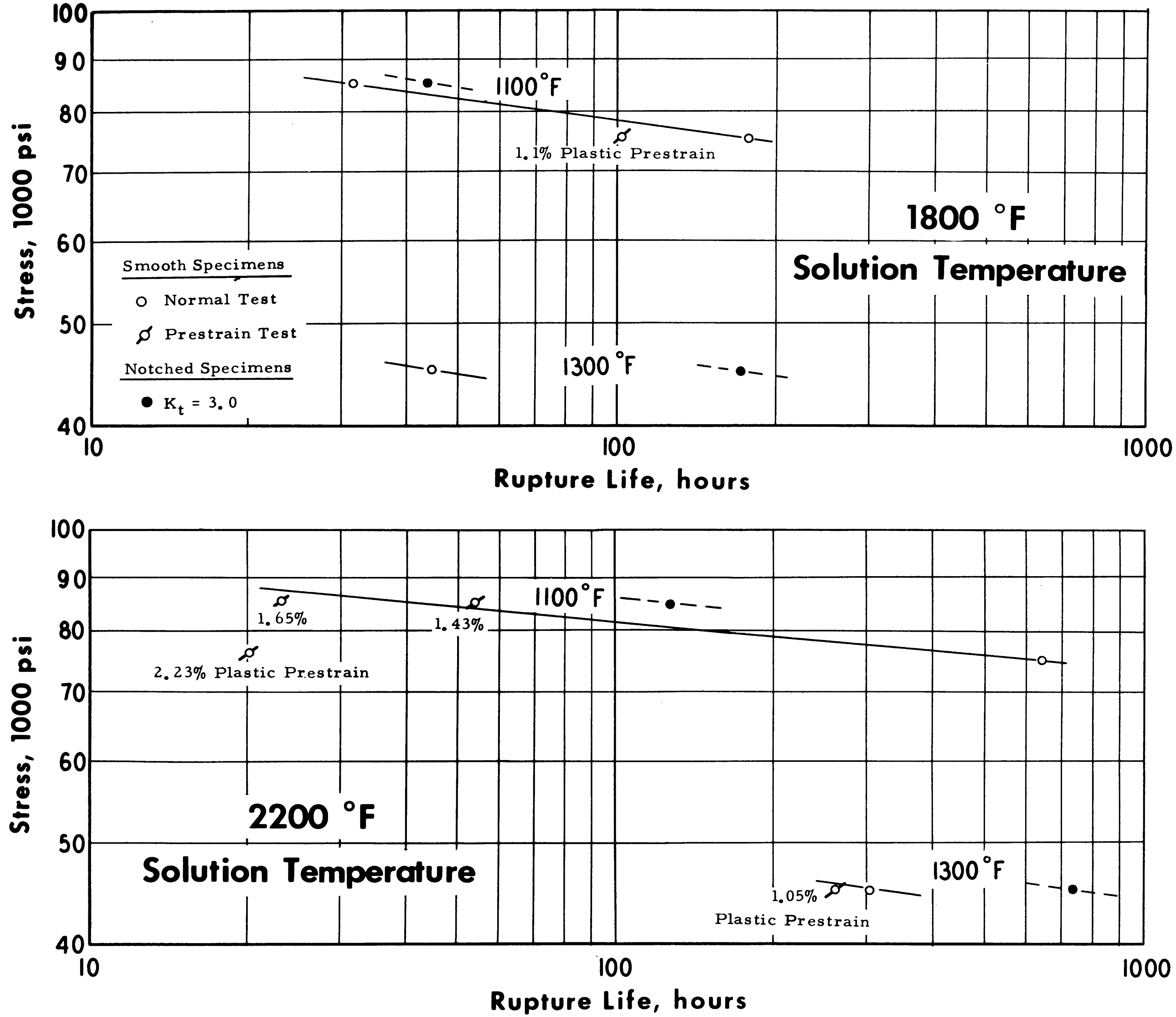


Fig. 3 - Rupture-Test Results at 1100 °F and 1300 °F for A-286 Heat 21,030

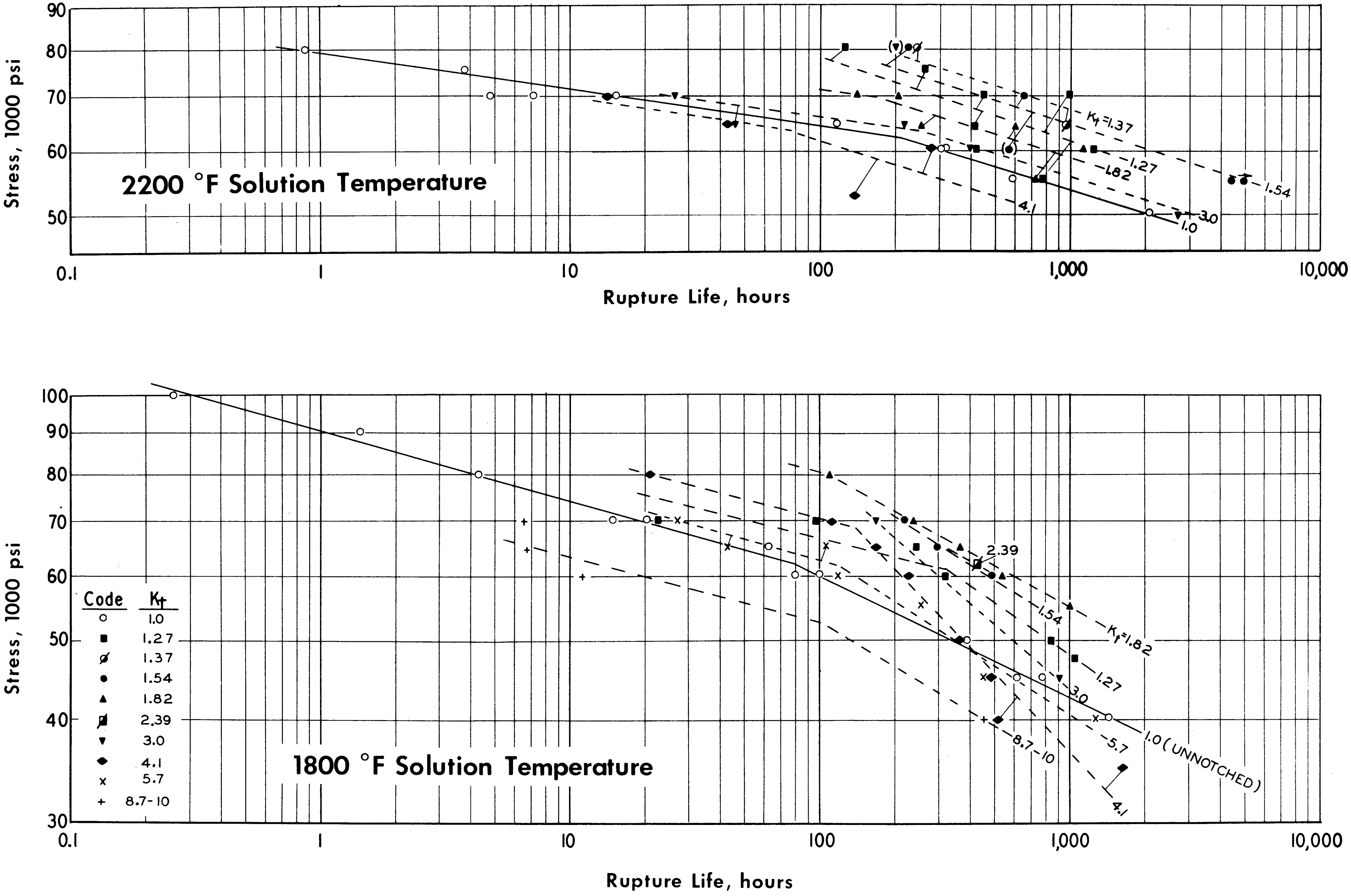


Fig. 4 - Rupture Properties at 1200 °F for Smooth and Notched Specimens of A-286 Heat 21,030.

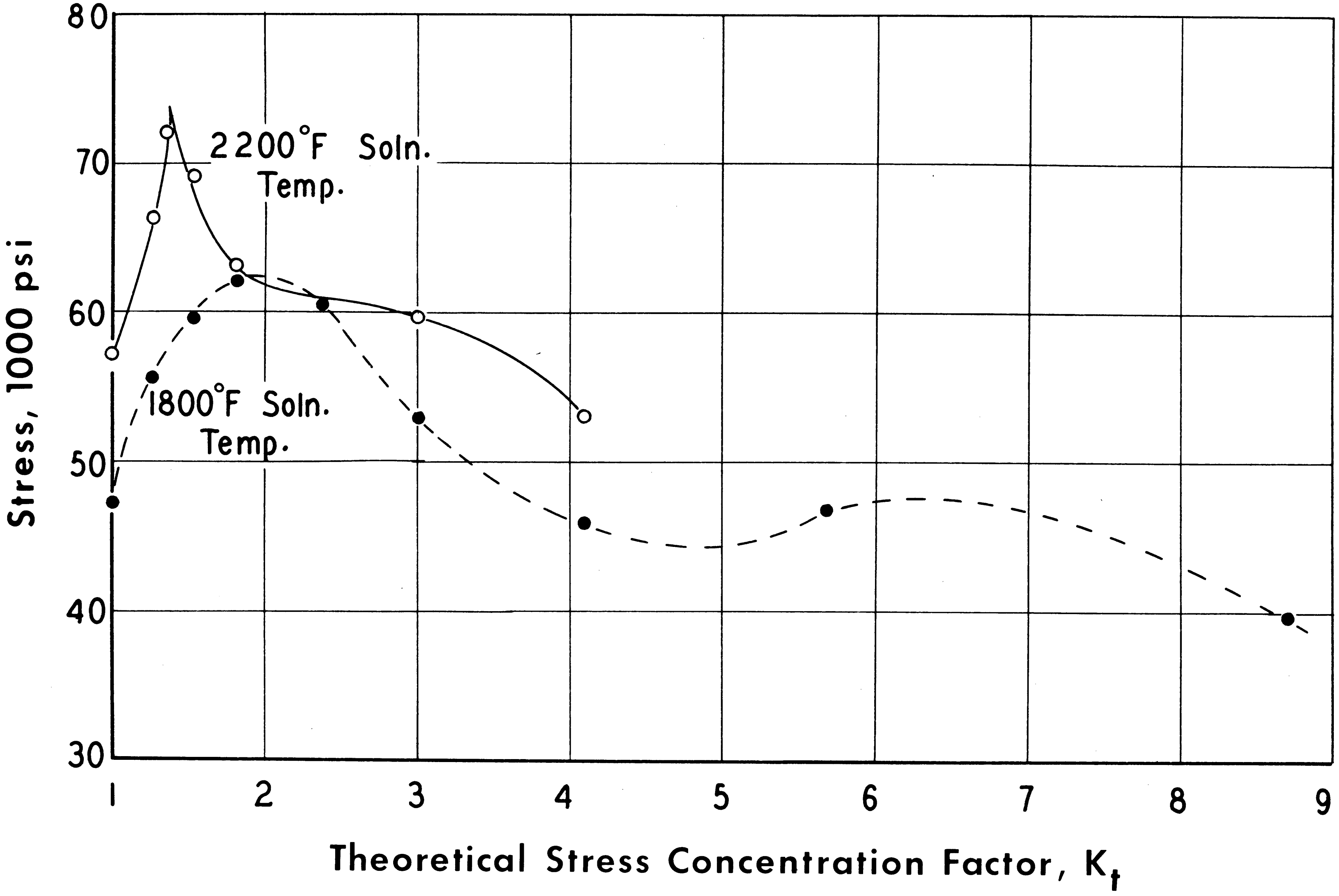


Fig. 5 - Effect of Notch Acuity on 500-Hour Rupture Strength at 1200°F for A-286 Heat 21,030.

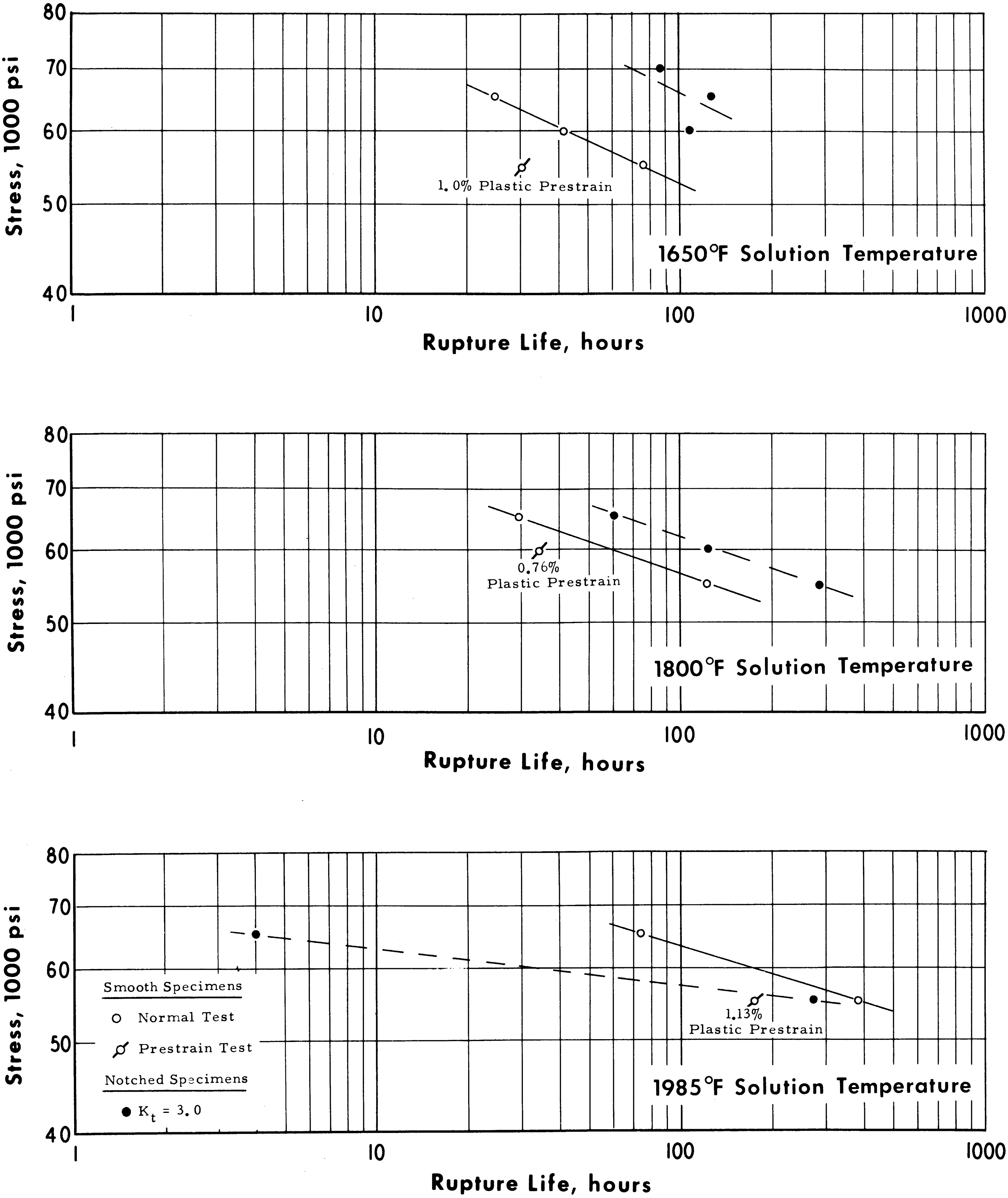


Fig. 6 - Rupture-Test Results at 1200 °F for A-286 Heat 43,297.

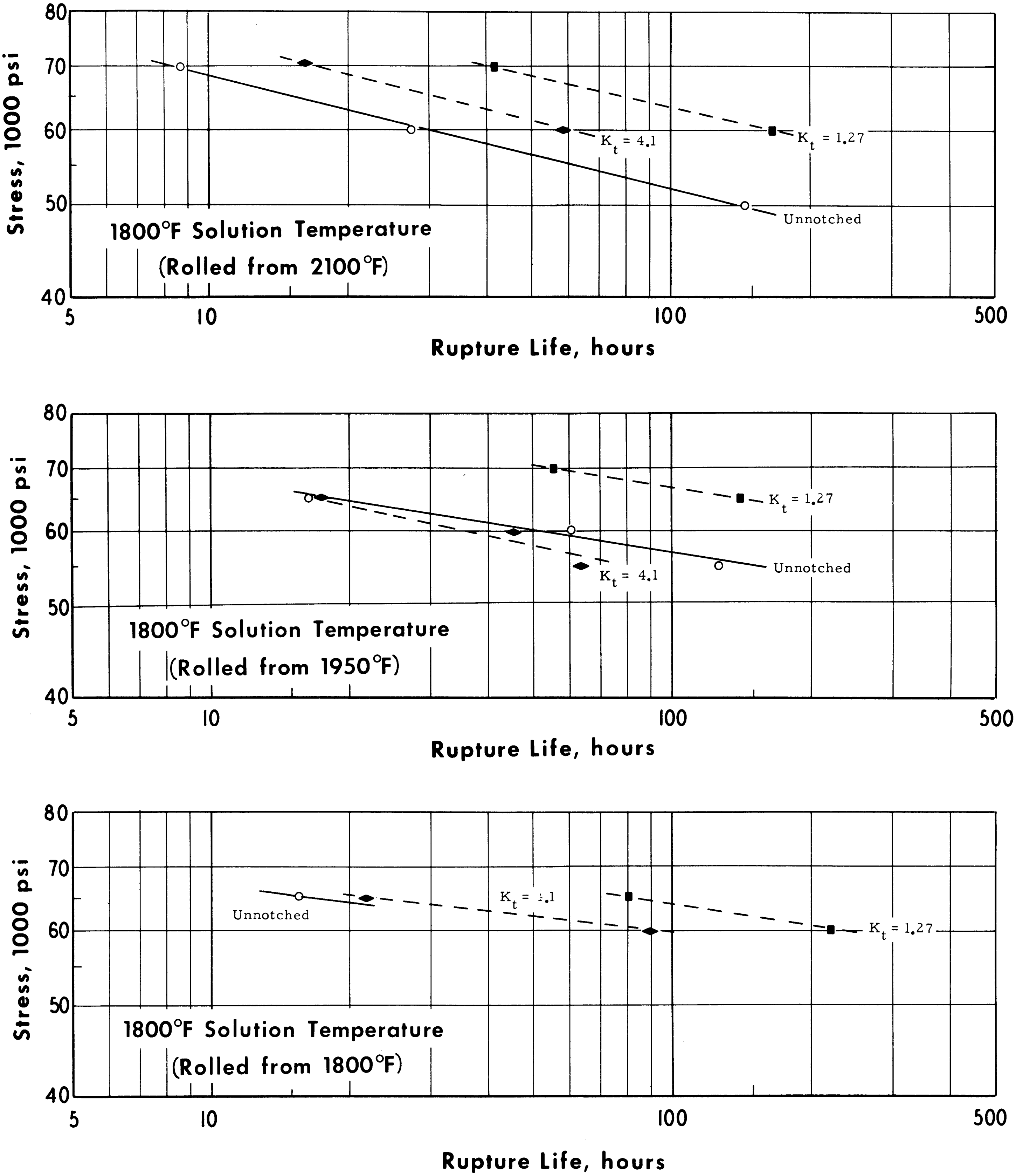


Fig. 7 - Results of Survey Tests on A-286 Heat 52,853 for 1800°F Solution Temperature.

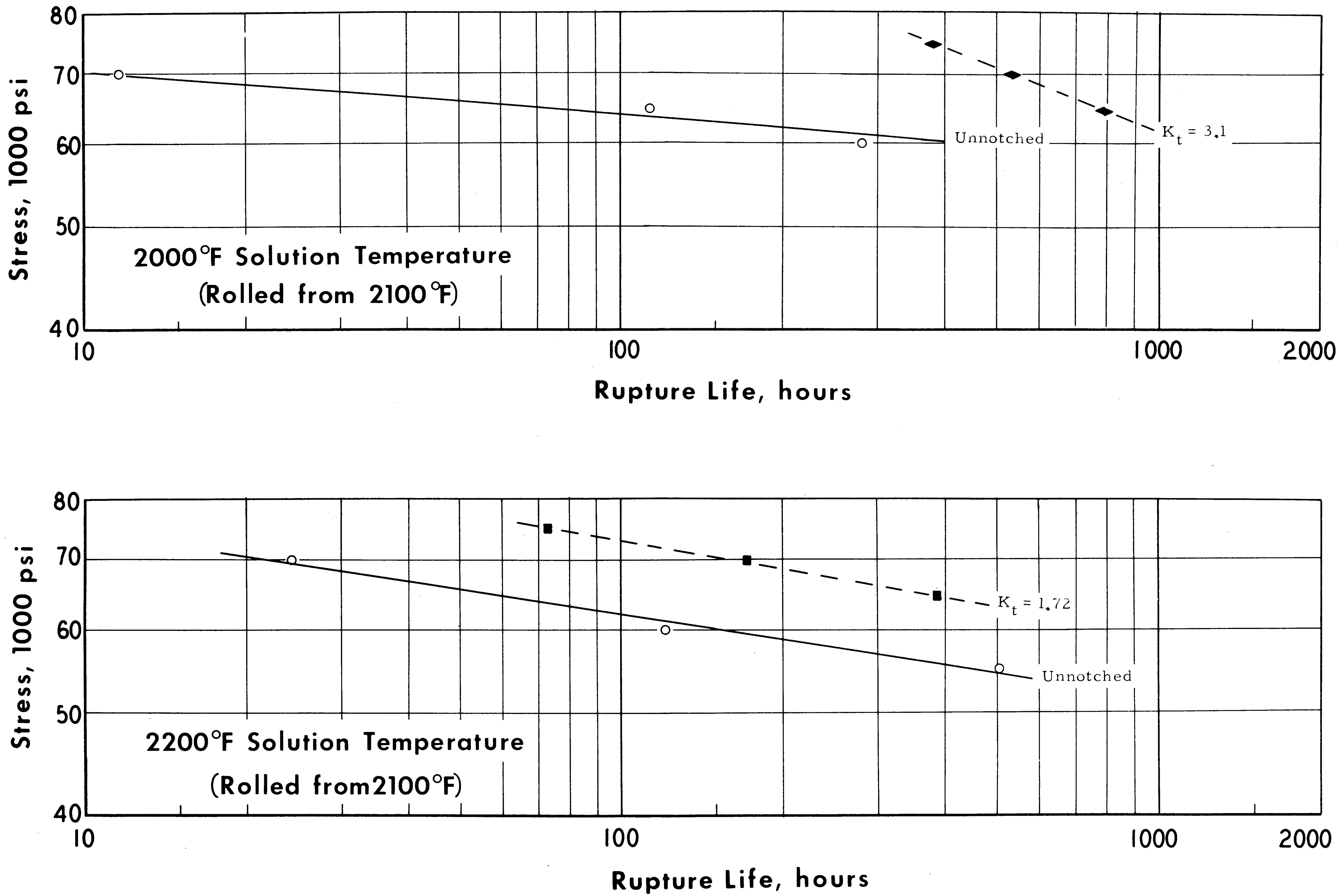


Fig. 8 - Results of Survey Tests on A-286 Heat 52,853 for 2000°and 2200°F Solution Temperatures

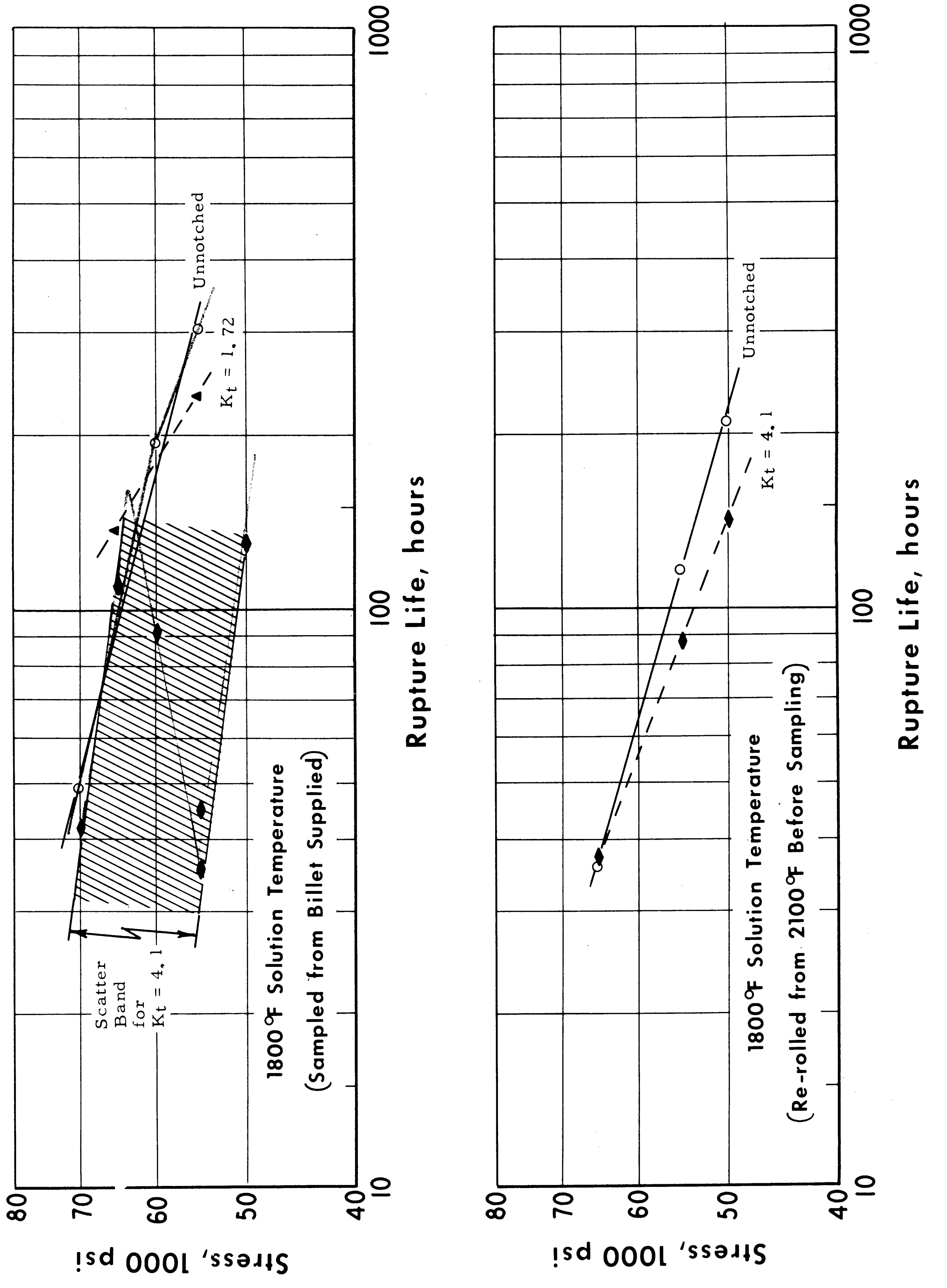


Fig. 9 - Results of Survey Tests on A-286 Heat K-65-X.

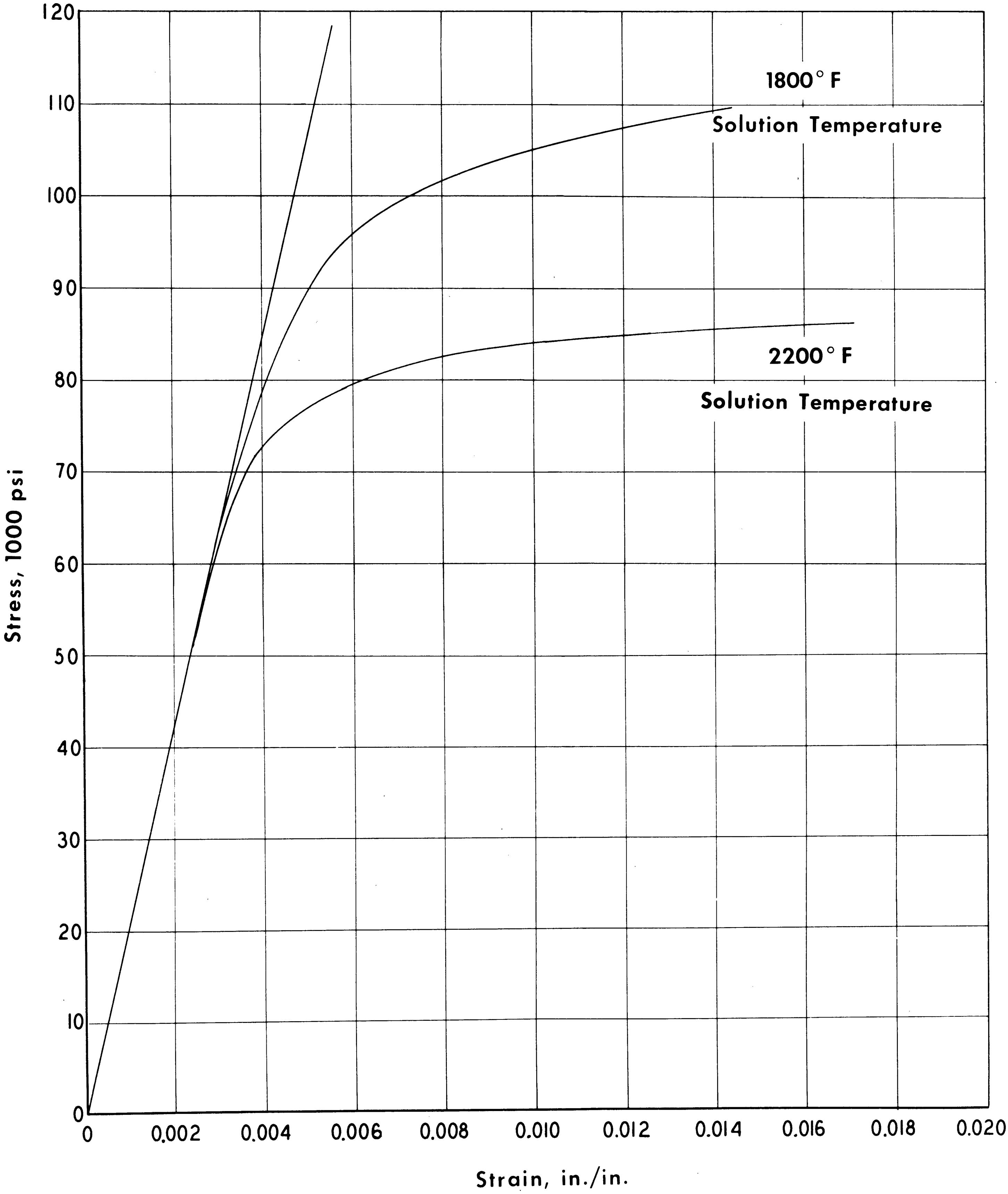


Fig. 10 - Stress-Strain Properties at 1200 °F for A-286 Heat 21,030 with 1800 ° and 2200 °F Solution Temperatures.

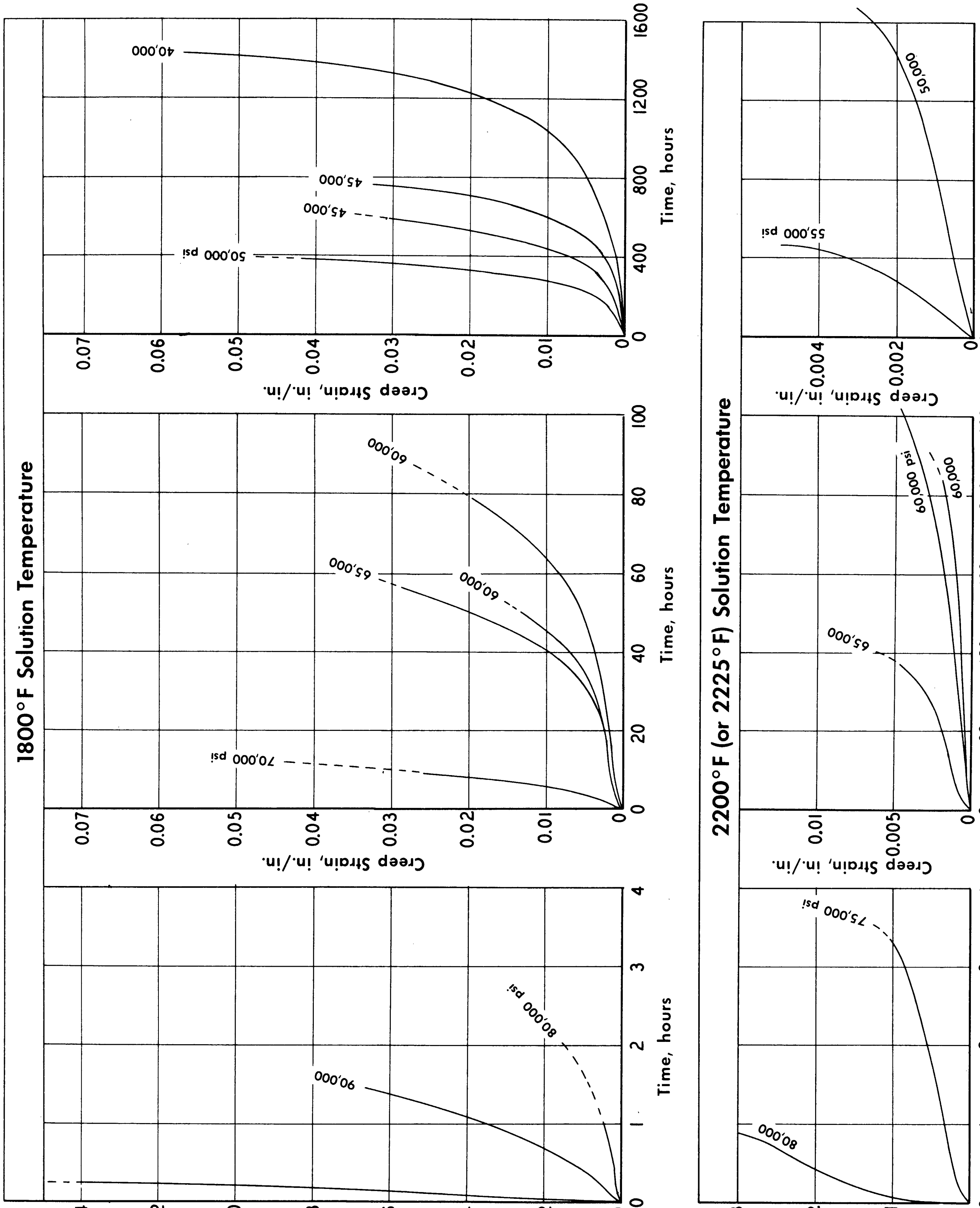


Fig. 11 - Creep Curves for A-286 Heat 21,030 Tested at 1200°F.

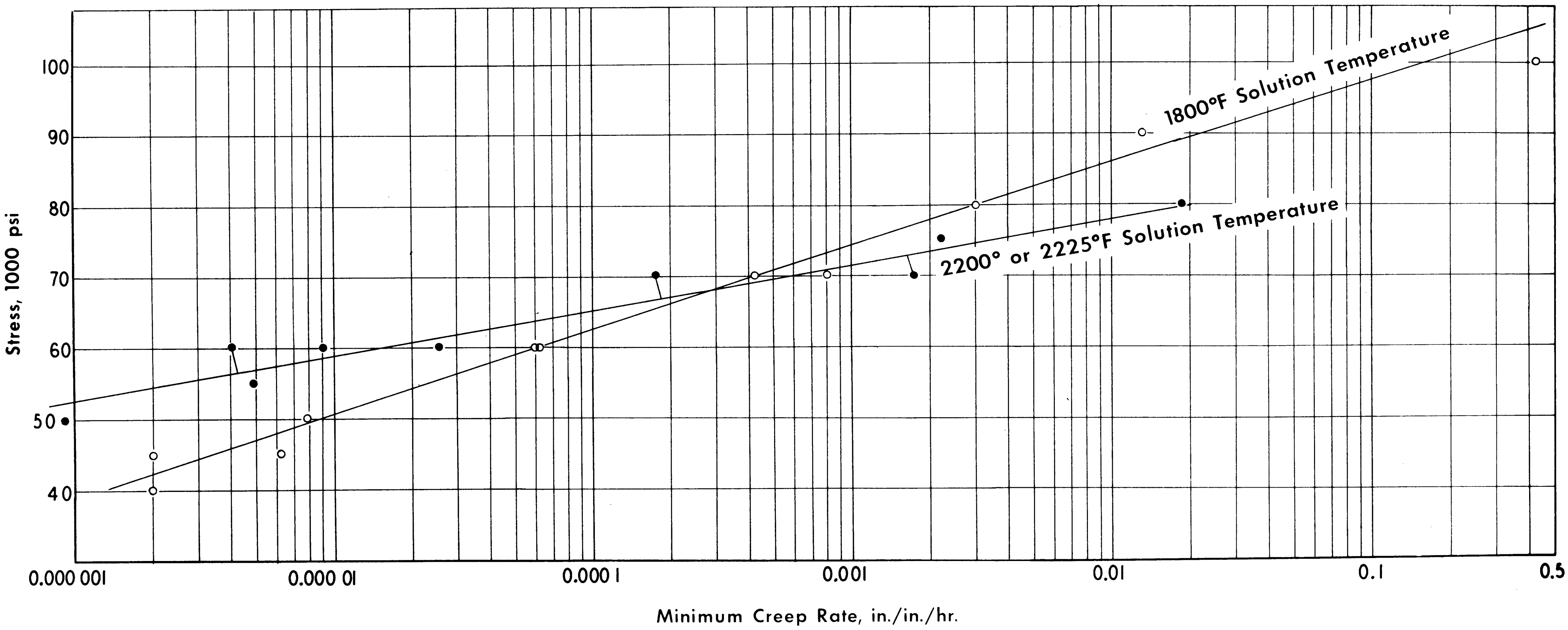


Fig. 12 - Stress Versus Minimum Creep Rate at 1200°F for A-286 Heat 21,030.

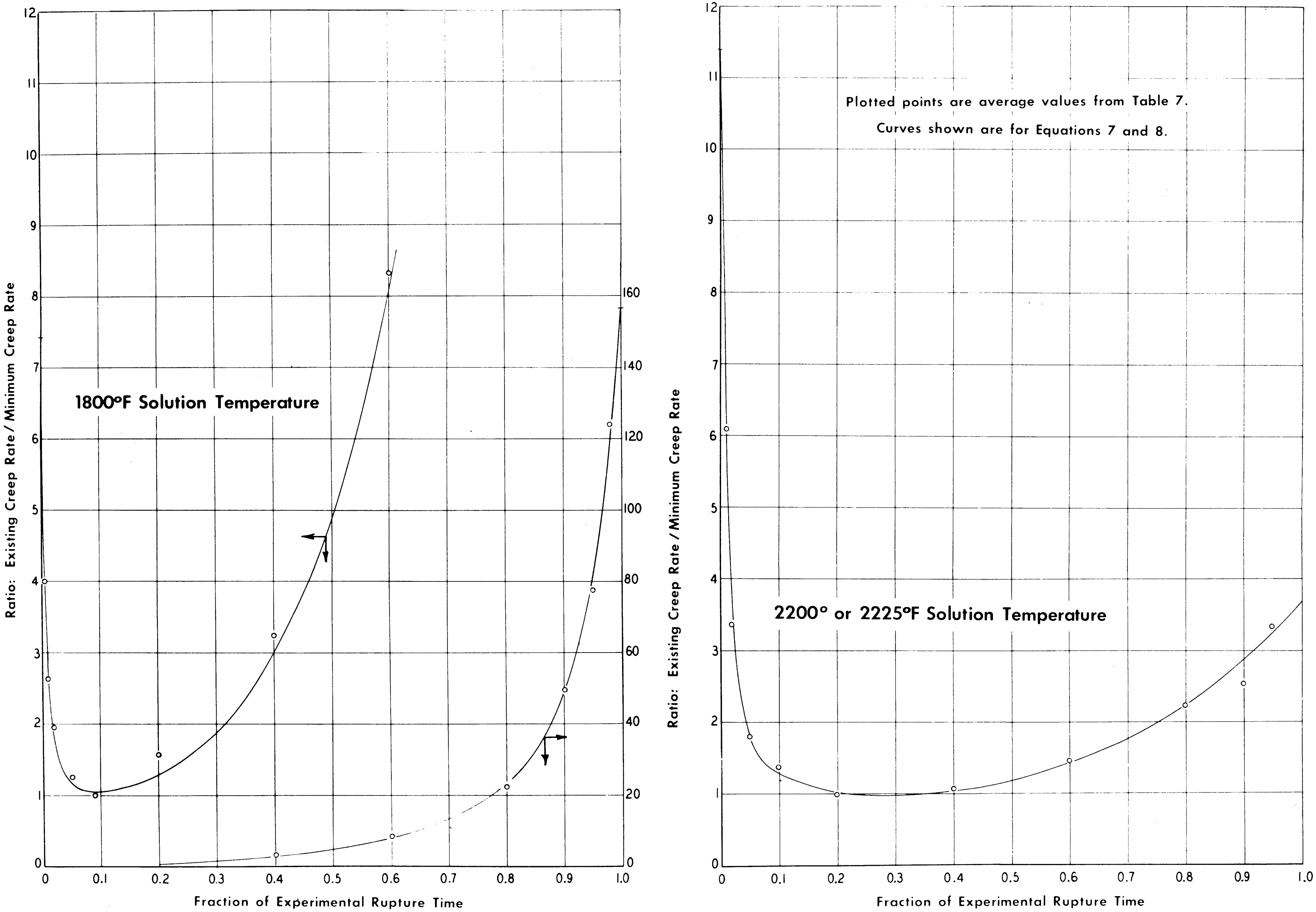
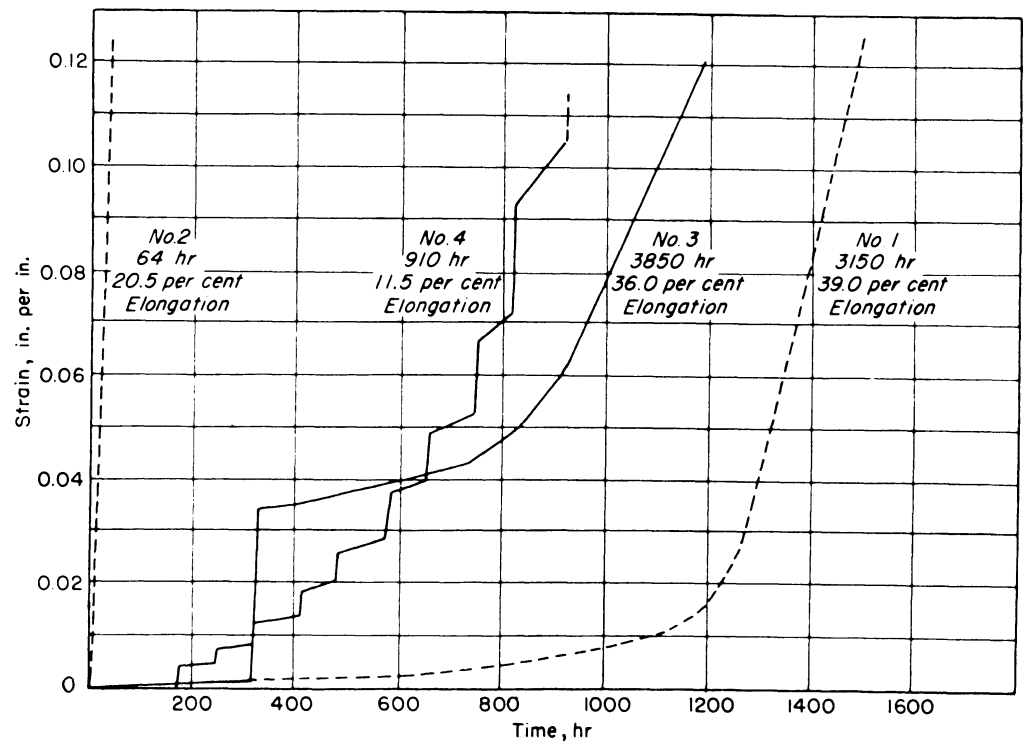


Fig. 13 - Variation in Creep Rate With Fraction of Rupture Time for Constant-Load Tests at 1200°F on A-286 Heat 21,030.



Base Conditions:
Temperature --- 1800 F
Stress -------- 1700 psi

Overstress, psi ----- 3200
Single load, hr -------- 25
Multiple load, hr ------ 2.5

Total hr --- 23

Fig. 14 - Effects of Single and Multiple Overloads on the Creep of Inconel. (Caughey and Hoyt, Ref. 7).

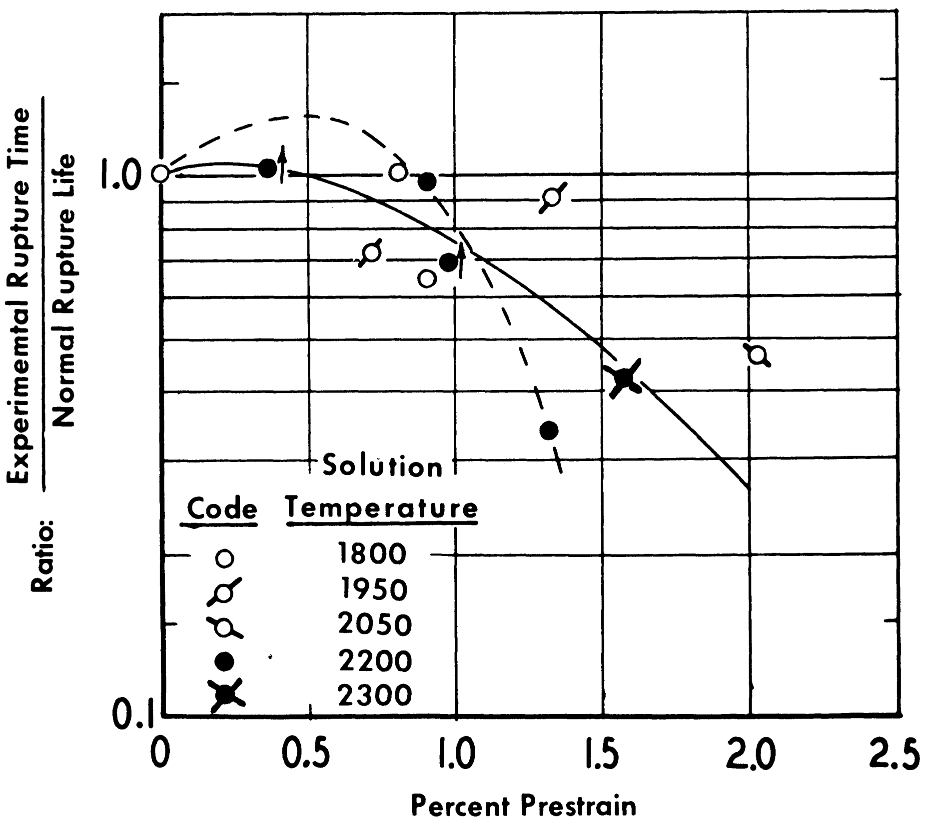


Fig. 15 - Effect of Plastic Prestrain on Rupture Life of A-286 Heat 21,030 at 1200°F.

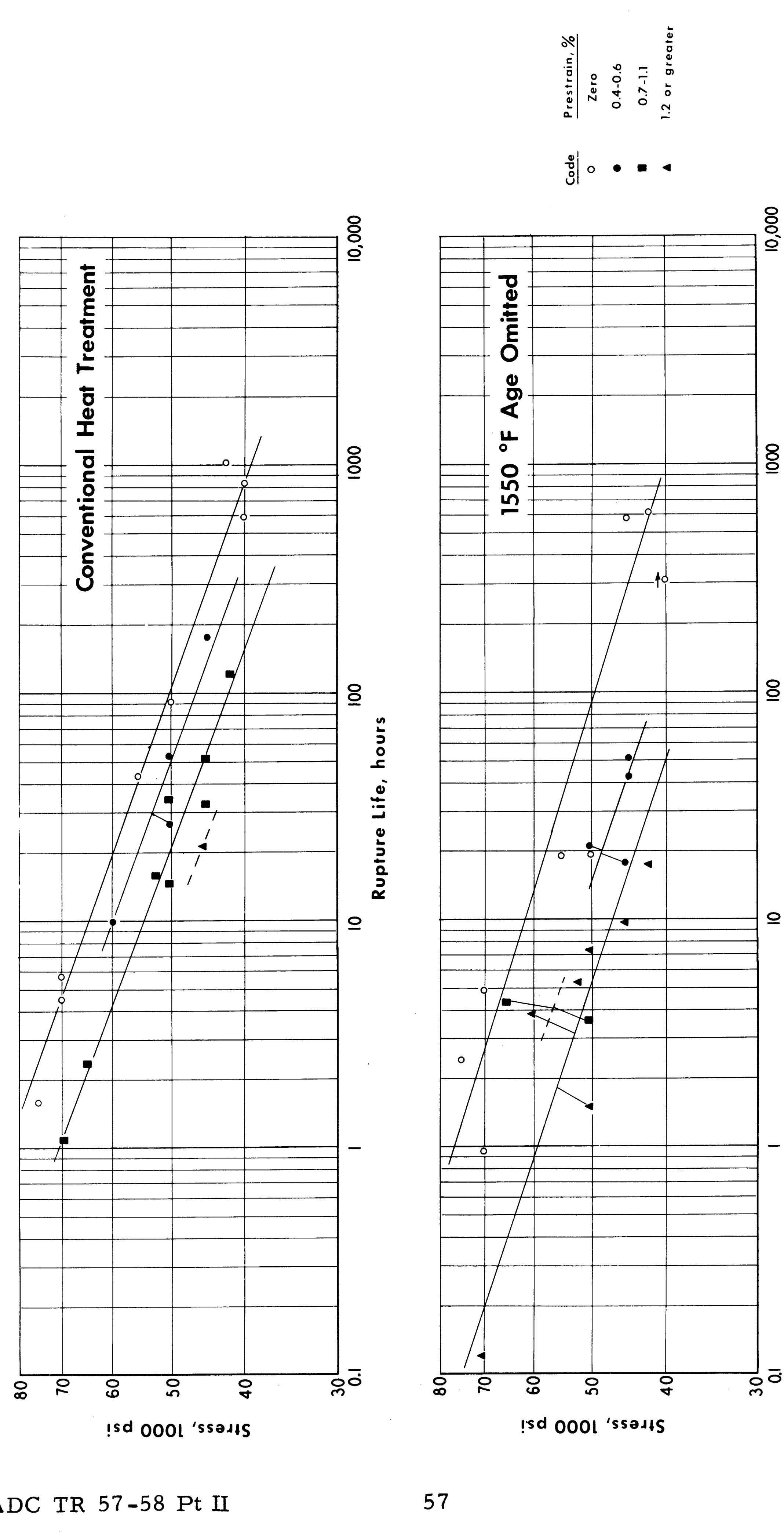


Fig. 16 - Effect of Plastic Prestrain at Test Temperature on Subsequent Rupture Life at 1350 °F for Three Heats of Waspaloy.

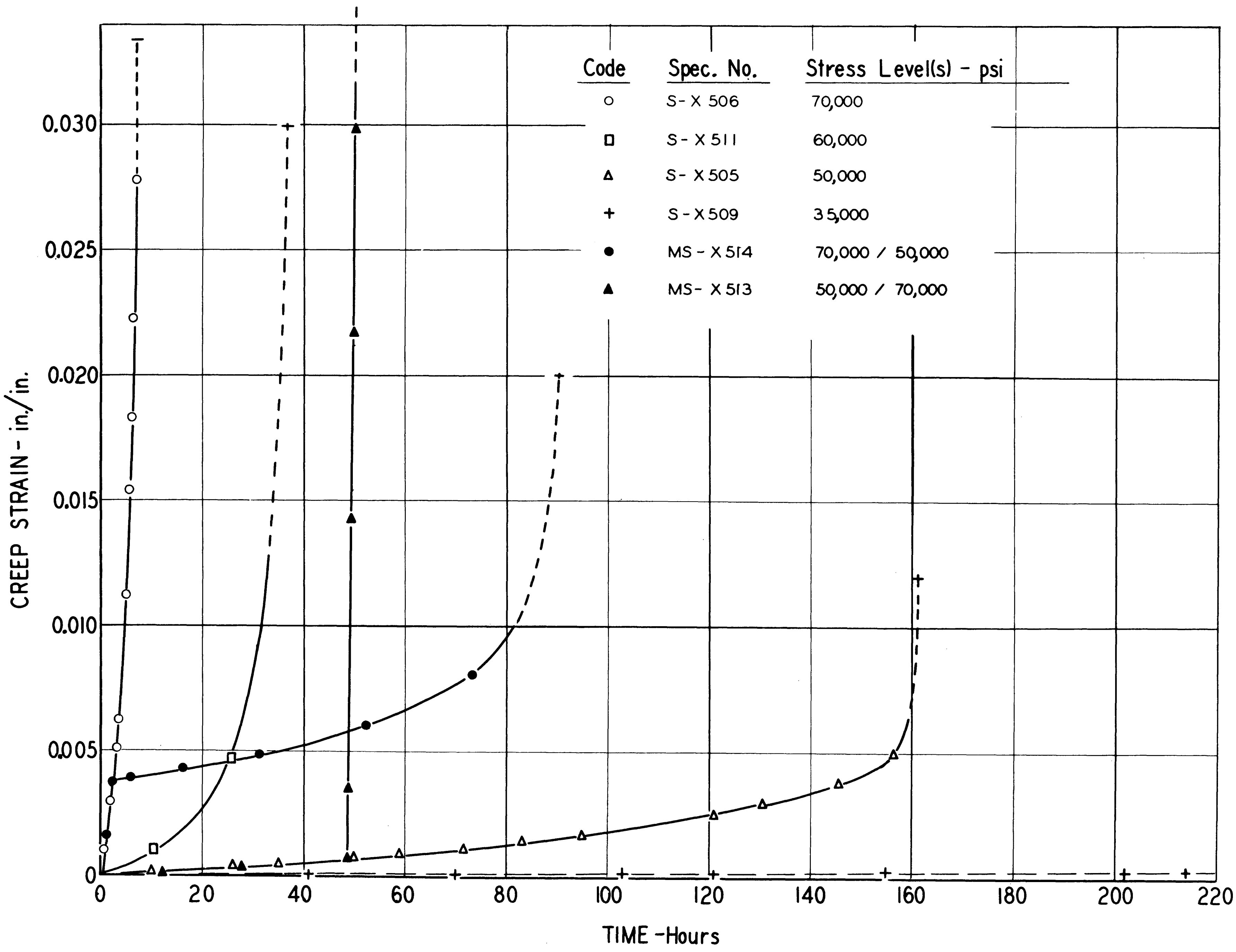


Fig. 17 - Typical Creep Curves Under Single- and Multiple- Stress Loading for Inconel X-550 Alloy at 1350°F.

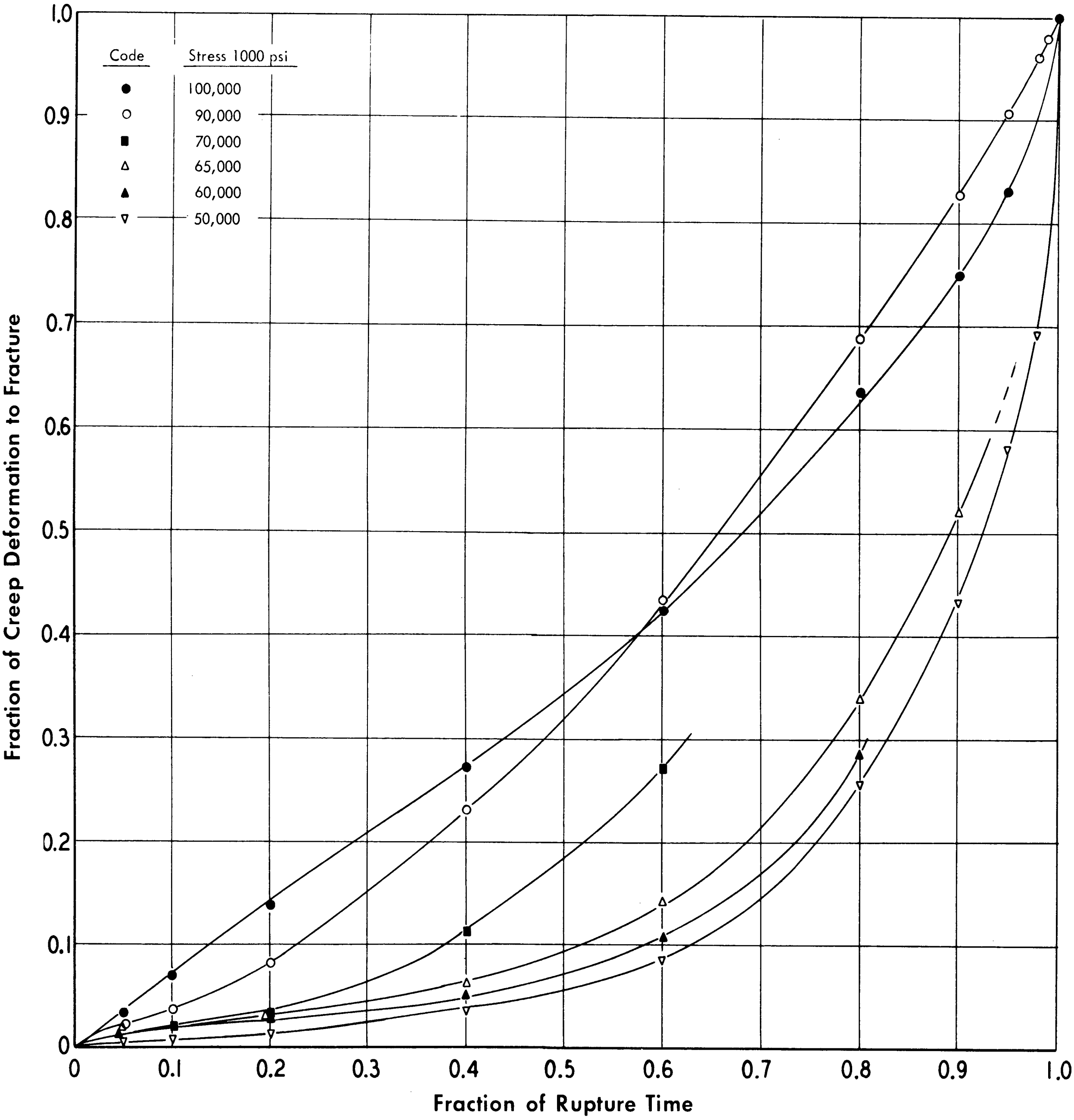


Fig. 18 - "Reduced" Creep Curves for A-286 at 1200°F. (Heat 21,030 with 1800°F Solution Temperature).

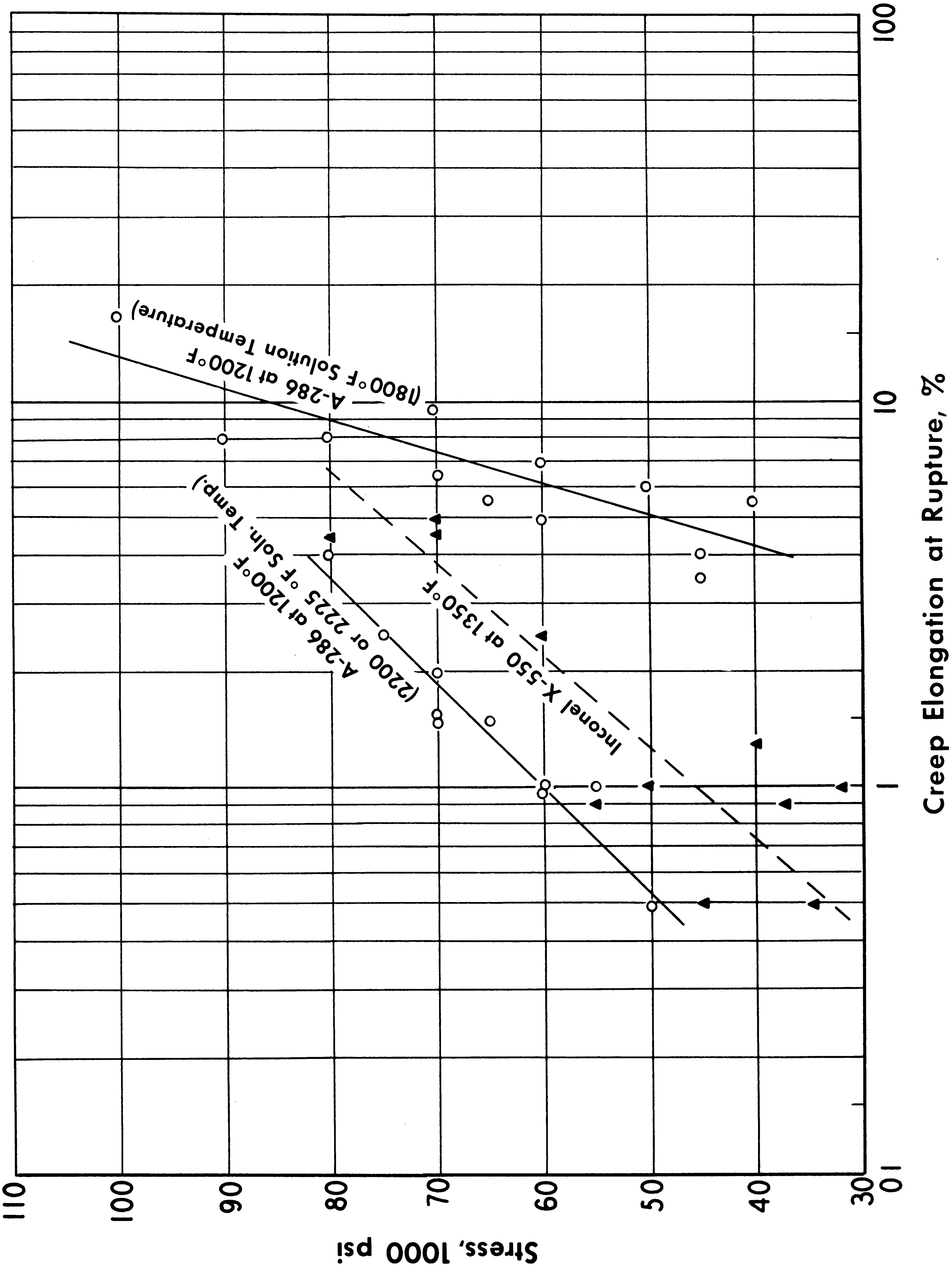


Fig. 19 - Stress Versus Creep Elongation at Rupture for Materials Considered in Table 9.