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MATERIALS LABORATORY

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ON

STUDIES OF HEAT-RESISTANT ALLOYS

Phase A

Influence of Hot Working on Structure
and Creep-Rupture Properties

Phase B

Relationship Between Strain Aging Phenomena
and High-Temperature Strength

by

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SUMMARY

Progress is reported for research carried out under Air Force Contract No. AF 33(616)-5466 from September 30, 1958 to December 31, 1958.

The over-all objective of the investigation is to establish basic relationships between structures and creep-rupture properties of heat-resistant alloys. The current work is divided into two areas of study: (1) the influence of hot worl structure and creep-rupture properties of "A" Nickel, A-286 alloy, and "17-22-A"V steel, and (2) the relationship between strain aging phenomena and high-temperature strength in 1020 carbon steel and A-286 alloy.

Photomicrographs of some rolled and creep tested samples of "A" Nickel are presented. Fine substructures are clearly revealed in these micrographs, and they are presumed to be due to creep-accelerated recovery and polygonization of the less clearly defined as-rolled structures. Unfortunately, the number of creep tested samples exhibiting clear substructures was too small to determine whether a correlation exists between substructures and creep-rupture properties for rolled "A" Nickel.

The structural study of A-286 alloy has advanced to the point where differences in age-hardening precipitate spacing, grain boundary conditions, and the extent of solution of excess phases can be observed through electron microscopy. However, no obvious correlations between structures and properties were found in the few representative samples examined thus far; a study of samples of all rolling conditions must be carried out before any conclusions may be drawn.

Creep-rupture testing of hot-rolled "17-22-A"V ferritic steel was recently completed, and the results are presented herein. Controlled hot rolling was used to produce three different austenite structures from which 100 percent bainite structures were formed by anisothermal transformation during

cooling from rolling. The three different austenite structures were produced by adjusting rolling temperatures to give complete, partial and no simultaneous recrystallization of the austenite. Rupture tests at 55,000 psi and 1100°F revealed a sharp improvement in ductility and a modest increase in rupture life when the austenite was worked without recrystallization, as compared to the case where the austenite was completely recrystallized during rolling. Tests at 40,000 and 29,000 psi at 1100°F showed that working the austenite without recrystallization produced only a minor ductility improvement but a 150-160 percent increase in rupture time over the recrystallized austenite. Compared to a martensitic structure produced by heat treatment in the same heat of "17-22-A"V, the bainitic structure formed from austenite worked without recrystallization had a ten-fold greater rupture life.

In the strain aging study, work is being done to identify the mechanism for high strength in the creep range exhibited by carbon steels susceptible to strain aging. A program has been initiated to follow-up a hypothesis involving the precipitation of silicon or aluminum nitride during constant strain rate testing using the electron microscope for identification.

The A-286 alloy being investigated regarding the effect of susceptibility to strain aging on its high temperature properties has been given a 1650°F solution treatment in an attempt to alter the strain aging characteristics observed previously for this alloy solution treated at 1800°F. Strain aging as evidenced by serrated stress strain curves on tensile testing was found to be less evident after solution treating at 1650°F. Results of tensile testing and testing at a constant strain rate of 0.004 percent per hour indicate that when the alloy is more susceptible to strain aging the high temperature strength is improved.

INTRODUCTION

This report, the third quarterly progress report issued under Air Force Contract No. AF 33(616)-5466, covers work done from September 30, 1958 to December 31, 1958.

The over-all objective of the present investigation is to determine what basic relationships exist between processing variables and high-temperature properties of heat-resistant alloys so that such alloys can be produced with consistently high strengths. This work is a continuation of earlier investigations sponsored at the University of Michigan by the Wright Air Development Center on the effect of heat treatment on microstructure and creep-rupture properties of low-alloy steels. The present research has been broadened to include:

Phase A - Influence of Hot Working on Structure and Creep-Rupture Properties

Phase B - Relationship between Strain Aging Phenomena and High-Temperature Strength

The materials being studied under Phase A are: (1) a commercially pure metal ("A" Nickel), (2) a precipitation-strengthened, austenitic alloy (A-286), and (3) a ferritic alloy of low strategic-element content ("17-22-A" V steel). Under Phase B strain aging studies are being carried out on three alloys: (1) a silicon-deoxidized 1020 steel (steel C) which is susceptible to strain aging in all conditions, (2) an aluminum-deoxidized 1020 steel (steel F) which may or may not be susceptible to strain aging, depending on the heat treatment, and (3) a high-strength, precipitation-hardened, austenitic alloy (A-286) which is susceptible to strain aging in various degrees, depending on the solution-treating temperature.

TEST MATERIALS

The materials for the hot working study were supplied gratis by the following organizations: "17-22-A" V steel from the Timken Roller Bearing Company, A-286 alloy (Heat 21030) from the Allegheny-Ludlum Steel Corporation, and "A" Nickel from the International Nickel Company.

The plain carbon steels (Steel C and Steel F) used in the strain aging study were obtained from the Chemical and Petroleum Panel of the ASTM-ASME Joint Committee on the Effect of Temperature on the Properties of Metals. The A-286 alloy (Heat 82073) used in the strain aging experiments was supplied by the Materials Laboratory, Wright Air Development Center from the stock used by Captain Domian to study strain aging by hot hardness tests.

The chemical compositions were reported by the manufacturers as follows:

Alloy	C	Mn	Si	Cr	Ni	Mo	V	Fe'	Other
"A" Nickel (Ht. N9500A)	0.06	0.27	0.06	-----	99.46 (Ni+Co)	----	----	0.09	0.03Cu;0.008S
A-286 (Ht. 21030)	0.06	1.35	0.47	14.58	25.3	1.38	0.21	Base	2.00Ti;0.17Al
A-286 (Ht. 82073)	0.03	1.27	0.62	14.58	25.44	----	0.59	Base	0.008N;0.12Al 0.37 Co;Nom. Ti
"17-22-A"V (Ht. 11833)	0.29	0.70	0.71	1.43	0.31	0.51	0.81	Base	--------------
1020 Steel C	0.20	0.68	0.27	-----	----	----	----	Base	0.015Al; 0.0048N; 0.028P;0.034S
1020 Steel F	0.19	0.68	0.24	-----	----	----	----	Base	0.053Al; 0.0046N; 0.026P;0.036S

PHASE A - INFLUENCE OF HOT WORKING ON STRUCTURE AND CREEP-RUPTURE PROPERTIES

The objective of this phase of the investigation is to seek an explanation of creep-rupture behavior in terms of microstructure for three types of heat-resistant alloys which were hot rolled systematically to introduce significant structural differences. Two of the materials, "A" Nickel and A-286 alloy, were processed and creep-rupture tested under a previous WADC contract (Ref. 1). The third, "17-22-A"V ferritic steel, was hot-rolled and creep-rupture tested recently. The results of the creep-rupture tests are presented for "17-22-A"V steel for the first time, and some general trends in the data are discussed.

Commercially Pure Metal ("A" Nickel)

Commercially pure nickel was selected for study with the idea that it would have a minimum number of structural variables and that they would be identifiable. The problem of identifying substructures, however, has been unexpectedly difficult.

Substructures in Nickel Rolled at Elevated Temperatures

Substructures have been observed in "A" Nickel rolled 5.8 percent reduction at 1800°F. The observations were made metallographically (e.g., Figures 17 and 18 in Ref. 1) and by a back-reflection x-ray diffraction technique (e.g., Figure 3 in Ref. 3). As the rolling temperature is lowered and/or the percent reduction is increased, the substructure rapidly becomes very difficult if not impossible to see in etched specimens. However, data in the literature support the idea that all of the nickel samples rolled at elevated temperatures must possess some sort of substructure, even though it may not be revealed by etching or by x-ray techniques. On the basis of work by Hirsch (Ref. 4) and Gay and Kelley (Ref. 5) it seems likely that the structure of hot-rolled (but not recrystallized) nickel consists of minute regions of near-perfect lattice (subgrains) separated by diffuse regions in which the lattice is distorted or "cold worked". The failure of both metallographic and x-ray diffraction techniques to reveal a substructure in samples with low rolling temperatures and/or high reductions can be explained on the basis of this structure model as follows: (1) The absence of a sharp boundary between the distorted lattice or "cold-worked" regions and the near-perfect-lattice regions rules out the possibility of delineating the structure by etching. (2) The presence of the distorted lattice regions prevents analysis by x-ray diffraction techniques because within any given grain diffraction spots from distorted-lattice regions are broad and diffuse and they overlap and obscure the otherwise sharp diffraction spots of the individual near-perfect subgrains.

Microscopic Examination of "A" Nickel after Creep at 1100°F and 11,000 psi

The process of creep at relatively high temperatures and/or relatively low strain rates is known to produce a substructure containing sharp subgrain boundaries, i.e., a structure free from the diffuse, "cold-worked" regions described above in the model of warm-worked nickel. If the structure contains regions of "cold-worked" lattice prior to creep, these regions are replaced by a well-defined substructure containing sharp subgrain boundaries. This process is sometimes called stress-recovery.

Representative rolled samples of "A" Nickel were examined microscopically after creep testing at 1100°F and 11,000 psi in the hope that stress-recovery or stress-accelerated polygonization would "sharpen" the as-rolled substructure so that it could be revealed by etching. The results were somewhat encouraging; however, the number of samples exhibiting well-defined substructures was still too small to permit a complete analysis of substructures over a wide range of rolling conditions. An attempt will be made to correlate substructures with creep properties and rolling conditions for the few samples in which substructures can be seen. Figure 1 shows examples of substructures found in rolled specimens after creep testing. The sample rolled 5.8 reduction at 1800°F showed exceptionally clear substructures after creep testing 702 hours at 1100°F and 11,000 psi (Fig. 1a - 1d).

Precipitation-Strengthened, Austenitic Alloy (A-286)

Creep-rupture properties of rolled and heat-treated A-286 alloy were surveyed under a previous contract (Ref. 1). The rolling temperatures were 80°, 1700°, 1950°, and 2200°F, and the reductions ranged from 0 to 40 percent.

Structural Studies

Until recently, attempts to observe the age-hardening precipitate in A-286 alloy were unsuccessful because of etching problems. Now, it has been found that a modified form of glyceregia, 4 HCl - 1 HNO_3 - 5 glycerine, reveals the precipitate adequately in most samples. In a few samples the precipitate size and spacing is so small that a clear delineation of the structure is apparently not possible with this etchant or, perhaps, any etchant. Representative electron micrographs are presented in Figures 2 and 3. Figure 2 shows the effect of heating alone, that is, the zero reduction series of specimens. This series is shown because a large difference in creep-rupture properties was found for these specimens. The 80° and 1700°F specimens were considerably weaker than the 1950° and 2200°F specimens (Fig. 27, Ref. 1). Figure 3 shows the effect of percent reduction for rolling started at 2200°F. As yet, no obvious correlations between structures and creep-rupture properties have been found. There are observable differences in particle spacing, grain boundary conditions, and degrees of solution of excess phases. Not all of the specimens have been examined yet, and neither has the full extent of structure variability within individual samples been determined. Therefore, a complete correlation between properties and structures is not ready at this time.

Ferritic Alloy ("17-22-A"V Steel)

The study of the effect of hot working on the creep-rupture properties of "17-22-A"V steel has been based on the assumption that the structural factor which probably controls the properties is the condition of the austenite just prior to the transformation to bainite. Accordingly, the influence of rolling conditions on the austenitic structure was determined (See Figs. 5 through 8, Ref. 3) so that specimens with controlled differences in prior austenite structure could be prepared or rupture tests.

Control of Structures

Specimens containing 100-percent bainite were produced from three prior austenite structures which differed as a result of controlled hot rolling. The procedure was to heat to a very high austenitizing temperature, air cool to a predetermined lower temperature and reduce the bar a large amount by two passes n rapid succession, and finally allow the rolled bar to cool in air. The initial eating produced a controlled degree of solution of excess phases and subsequent rain growth. The amount of reduction and the number of passes were held constant. The three rolling temperatures were adjusted so that in one case the austenite was ompletely recrystallized, in another case the austenite was partially recrystallized, nd in the third case the austenite was not at all recrystallized but simply worked. Microscopic examination confirmed that in every case complete transformation to ainite occurred during air cooling from the rolling operation.

The details of the treatments are presented below along with the resulting ustenite grain sizes and tempered hardnesses.

Treatment	Description	Mean Austenite Grain Size (ASTM No.)	Tempered Hardness (BHN)
(1)	1 Hr. 2200°F, air cooled to 2100°F and rolled 50 percent reduction in 2 passes, air cooled + tempered 6 hrs. 1250°F (Complete recrystallization)	7.5	355
(2)	1 Hr. 2200°F, air cooled to 2000°F and rolled 50 percent reduction in 2 passes, air cooled + tempered 6 hrs. 1250°F. (Partial recrystallization)	Mixed 7.5 and 4.5	349
(3)	1 Hr. 2200°F, air cooled to 1900°F and rolled 50 percent reduction in 2 passes, air cooled + tempered 6 hrs. 1250°F (No recrystallization)	4.5	326

Rupture Properties at 1100°F

Rupture tests on "17-22-A"V steel samples were run at stress levels at

1100°F to establish the approximate relative rupture strengths of the three structures described above and to compare their properties with those obtained by the usual heat treatments on the same heat of "17-22-A"V. The data from hot-rolled specimens are presented in Table I and in Figure 4, these data are compared with previous data on heat-treated "17-22-A"V steel.

Under testing conditions involving high stresses, 55,000 psi, and high strain rates, working the austenite without recrystallization just prior to the bainite transformation brought about a sharp improvement in ductility and a slight increase in rupture life (Table I) over the case where the austenite was completely recrystallized during rolling. At lower stresses, 40,000 and 29,000 psi, and lower strain rates the bainite formed from strained austenite had a 150-160 percent longer rupture life than for bainite transformed from recrystallized austenite. Unfortunately, the ductility in the low-stress tests was only slightly improved by working the austenite without recrystallization. Reduction of area appears (Table I) to be a more reliable measure of ductility when such small values are concerned.

Consideration of these hot-rolling data together with previous data on heat-treated "17-22-A"V steel from the same heat (Fig. 4) suggests the following general trends:

1. The final austenitizing temperature has a pronounced effect on rupture ductility at 1100°F. In every case where the steel was heated to 2200°F and where bainite was allowed to transform from strain-free austenite (cross and open circles) the elongation was only about 1 percent. However, similar bainitic structures produced from austenite heated below the grain coasening range, to 1850°F, (half-solid circles) had elongations in the range of 12 to 20 percent.

2. Differences in ductility appear to result from some factor other than grain size. An excellent demonstration of this point is the fact that two specimens (cross and open circle at 40,000 psi) with the same austenitizing temperature,

2200°F, had the same ductility despite the fact that in one specimen considerable grain refinement (average grain diameter reduced from 0.086 to 0.030 mm) had been effected by hot working. The close correlation between ductility and austenitizing temperature suggests that the amount of excess phases in solution in the austenite is the factor controlling the ductility of "17-22-A"V at 1100°F and 55,000 psi.

3. As mentioned earlier, the effect of plastically straining the austenite just prior to the bainite transformation was to improve ductility under conditions of high stress and high strain rate at 1100°F. If the idea suggested in Item 2 above is correct, that the amount of excess phases dissolved in the austenite is the factor controlling ductility at 1100°F and 55,000 psi, then a likely explanation would seem to be that the numerous lattice defects introduced by the plastic deformation provides new sites for reprecipitation of the excess phases so that the average concentration of these phases at prior austenite boundaries is reduced and the ductility is improved. Unfortunately, this beneficial effect is lost at lower stresses and strain rates. The reason for this is not clear.

4. For material austenitized at 2200°F and tested at 1100°F and 40,000 or 29,000 psi the effect of working the austenite without recrystallization was to improve the rupture life considerably, but the ductility was essentially unaltered.

5. Considering the over-all range of rupture strength for this one heat of "17-22-A"V steel at 1100°F, heating to 2200°F and working the austenite without recrystallization produced a very striking tenfold increase in rupture time at 40,000 and 29,000 psi over the weak, martensitic structure (solid, inverted triangles in Fig. 4).

FUTURE WORK

"A" Nickel

In another attempt to observe substructures in severely worked samples the transmission type of electron microscopy will be tried. Some preliminary work has indicated that the major difficulty to be resolved is that of reducing the specimens from the bulk form down to a foil thin enough to allow an electron beam to pass through it.

Some preliminary work has also been carried out to establish a melting practice suitable for producing nickel of considerably higher purity than "A" Nickel to be used in a study of the influence of trace elements on the response of nickel to hot working. It is planned to produce the nickel from cobalt-free, carbonyl nickel powder in such a way that a purity of the order of 99.95+ percent will be obtained.

A-286 Alloy

Now that a reasonably effective etchant has been found for revealing the age-hardening precipitate in A-286 alloy, a detailed study of the samples will be carried out to ascertain what correlations there may be between structures and creep-rupture properties. Samples examined so far have not shown any obvious correlations.

"17-22-A"V Steel

The electron microscope will be used to determine whether significant structural differences exist between bainitic structures formed from strain-free austenite and bainitic structures formed from austenite worked without recrystallization.

PHASE B - RELATIONSHIP BETWEEN STRAIN-AGING PHENOMENA AND HIGH-TEMPERATURE STRENGTH

During the previous periods steels susceptible to strain aging were found to exhibit high strengths at higher temperatures for a strain rate of 0.004 percent per hour. In attempt to follow-up a precipitation strengthening hypothesis a program of microstructural studies using the electron microscope has been initiated on the carbon steels.

A-286 alloy is concurrently being studied as to the effect of strain aging on its high temperature properties. Specimens were given a 1650°F solution treatment to determine if contributions to strength from strain aging might be reduced in comparison to that from the treatment at 1800°F previously investigated. Comparison of results after tensile testing and testing at a strain rate of 0.004 percent per hour with those obtained for a solution temperature of 1800°F have indicated what effect strain aging has on this alloy. Constant strain rate tests and tensile tests were also conducted on this alloy after the 1800°F solution treatment to widen the scope of the data obtained during the previous period.

PROCEDURE

Processing of Materials

For the electron microscope investigation the carbon steels were given various aging treatments under no load to try to precipitate any nitride which might be responsible for the increase in strength at about 800° or 900°F. Because the aluminum-deoxidized steel has been made essentially immune to strain aging by furnace cooling after one hour at 1650°F it is apparent that the nitrogen atmospheres have been made inactive possibly due to combination with aluminum as an agglomerated precipitate. In an attempt to bring out this precipitate in the steels susceptible to strain aging, a long time treatment of 240 hours at 1300°F was used. To determine the effect of time alone at temperature on any structural changes or precipitate formation a stepwise aging program at times of 72, 144, 288 and 500 hours at 1200°F has been initiated on these same steels. This aging program is nearly complete at the present time. Since plastic strain is also believed to be instrumental in activating or accelerating any precipitation which might occur samples of the aging steels C and FA tested at a strain rate of 0.004 percent per hour out to 2 percent total strain at various temperatures are also being examined.

The A-286 alloy (Heat 82073) has been tested under two different conditions of heat treatment. All testing previously was conducted on the alloy after a solution treatment of one hour at 1800°F then oil quenched plus a 16 hour aging treatment at 1325°F. Much of the testing during the period has been done on the alloy after one hour at 1650°F and oil quenched plus 16 hours at 1325°F.

Testing

During this period constant strain rate tests were run on the A-286 alloy at a strain rate of 0.004 percent per hour over the temperature range from 1000°F to 1400°F for the 1650°F solution treatment and at 900°, 1300°, and 1400°F for the 1800°F solution treatment. The tests were carried out by manual adjustment of the load to a total deformation of two percent or fracture which ever occurred first. Tensile tests were also conducted on A-286 in both conditions of solution treatment to obtain a complete set of comparative tensile properties from room temperature to 1300°F. Strain was measured with an inductance strain gauge extensometer system out to a total deformation of two percent where ever possible.

RESULTS AND DISCUSSION

Carbon Steel

In an attempt to explain the high strength observed in carbon steels which are susceptible to strain aging a study of the steel structures at high magnification has been started. The hypothesis is that a precipitation of a silicon or aluminum itride takes place during constant strain rate testing at the dislocations where 1e free nitrogen is located initially according to the Cottrell theory for steels usceptible to strain aging. This precipitation of nitrides at the dislocations, which ; accelerated by both plastic strain and increasing temperatures, would be esponsible for the high strength at 800° to 900°F by offering a higher resistance to 1e movement of dislocations than did the free nitrogen. This precipitate when ›cated at its optimum position at the dislocations very probably would not be .sable even at 20,000 magnification; however, the fact that "overaging" occurred ith increasing plastic strain and temperature indicates that agglomeration of this recipitate may have taken place rendering the precipitate visible at high magnifications. lectron micrographs have been taken of the steels in both the aging and non-aging onditions after constant strain rate testing as well as after various heat treatments. : the present time the work has not progressed to a point where definite conclusions ın be made but particles of less than 1000Å in diameter have been observed for :veral conditions which may be nitrides of silicon or aluminum. Future work will : devoted to identifying the precipitate and determining it characteristics as to e effect of time, temperature and amount of plastic strain.

-286 Alloy

The testing of A-286 alloy in the 1650°F solution-treated condition has been arly completed and several comparisons can be made between these results and e results of the testing done during the previous period on this alloy solution eated at 1800°F.

To determine the difference in the susceptibility to strain aging as a result the 1650°F and the 1800°F solution treatments, tensile tests were conducted on the oy from room temperature to 1300°F. The way that normal manifestations of :ain aging appear in this alloy is through the presence of serrated stress-strain rves over a limited temperature range. This temperature range was from 500°F to)0°F for this alloy after solution treating at 1800°F with maximum serrations in both frequency and intensity occurring at 900°F. Tensile testing on this alloy after solution treating at 1650°F showed a marked decrease in the serrated nature of the stress strain curves. The behavior was quite similar in that sharp yield points, evident in carbon steels, were not observed at any temperature and that the serrations, when they occurred, were similar in nature to those observed for the 1800°F solution treatment. However, the temperature range where serration was evident was smaller and the severity of the serration observed was markedly less than that observed for the 1800°F solution treatment. At 500°F only very minor irregularities or serrations were observed, these increased in intensity with increasing temperature to a maximum at about 800°F then decreased rapidly after 900°F to an almost completely smooth curve at 1000°F and higher. From the stress-strain curves obtained, values of stress for deformations of 0.5, 1.0 and 2.0 percent were plotted as a function of temperature and appear in Figure 5. This figure contains the corresponding data obtained from the 1800°F solution treatment during the previous period for comparison purposes. In comparing these curves it can be seen that strengths are maintained at a high level for both conditions over the range where strain aging is evidenced by the serrated curves. The strengths seem to drop off at a lower temperature for the 1650°F condition which fits in with the fact that serrated stress strain curves cease to be evident at a temperature of 100° to 200°F lower than the 1800°F condition.

Figure 6 shows the comparison of the ultimate and breaking strength for the two conditions as a function of temperature. It is evident that up to a temperature of 1000°F there is no significant difference between the two conditions. At this point the strength drops off markedly for the 1650°F conditions whereas strengths are held up to almost 1200°F for the 1800°F condition before dropping off sharply. This comparison again illustrates that there seems to be a correlation between susceptibility to strain aging as evidenced by serrated stress-strain curves and high strength in the A-286 alloy.

Figure 7, presenting the reduction of area and elongation from tensile testing shows the effect of solution temperature on these properties. The reduction of area curve indicates that the 1650°F condition exhibits higher reduction of area at low temperatures but falls below the 1800°F condition at temperatures higher than 900°F. Since the differences are not large and no significant difference is evident in elongation measurements the differences in ductility from one condition to the other appears to be negligible over the temperature range investigated.

Constant strain rate tests were run at 0.004 percent per hour up to 1400°F on the 1800°F solution treatment to see if there was any effect of strain aging on this alloy at higher temperatures. The test at 1000°F was also rerun to check a high modulus value which was found to be in error. This was responsible for a high value at 0.5 percent reported previously. A series of tests were also run on the A-286 alloy in the 1650°F solution treatment. At all temperatures between 1000°F and 1400°F a decrease in stress was necessary to maintain a c onstant strain rate which is the equivalent of third stage creep for constant load creep tests.

Figure 8 showing the comparative results of the constant strain rate tests for the two treatments illustrates that very little difference in strength occurs at deformations of 0.5, 1.0 and 2.0 percent out to a temperature of 1200°F. At 1300°F a slight inflection in the curves for the 1800°F solution treatment is evident, whereas the 1650°F curves continue to fall off. The test for the 1650°F condition being run at 1400°F is still in progress.

Tests run at 1000° and 1100°F on the 1650°F solution treated condition ruptured with brittle notch-like fractures in less than 500 hours as did the tests run at these temperatures on the 1800°F material.

FUTURE WORK

Work on trying to determine just what is responsible for the strengthening mechanism in carbon steel using the electron microscope to examine the microstructure will be intensified. The results of the testing on the A-286 alloy will be analyzed when all of the tests have been completed and future work will be carried out in accordance with these results.

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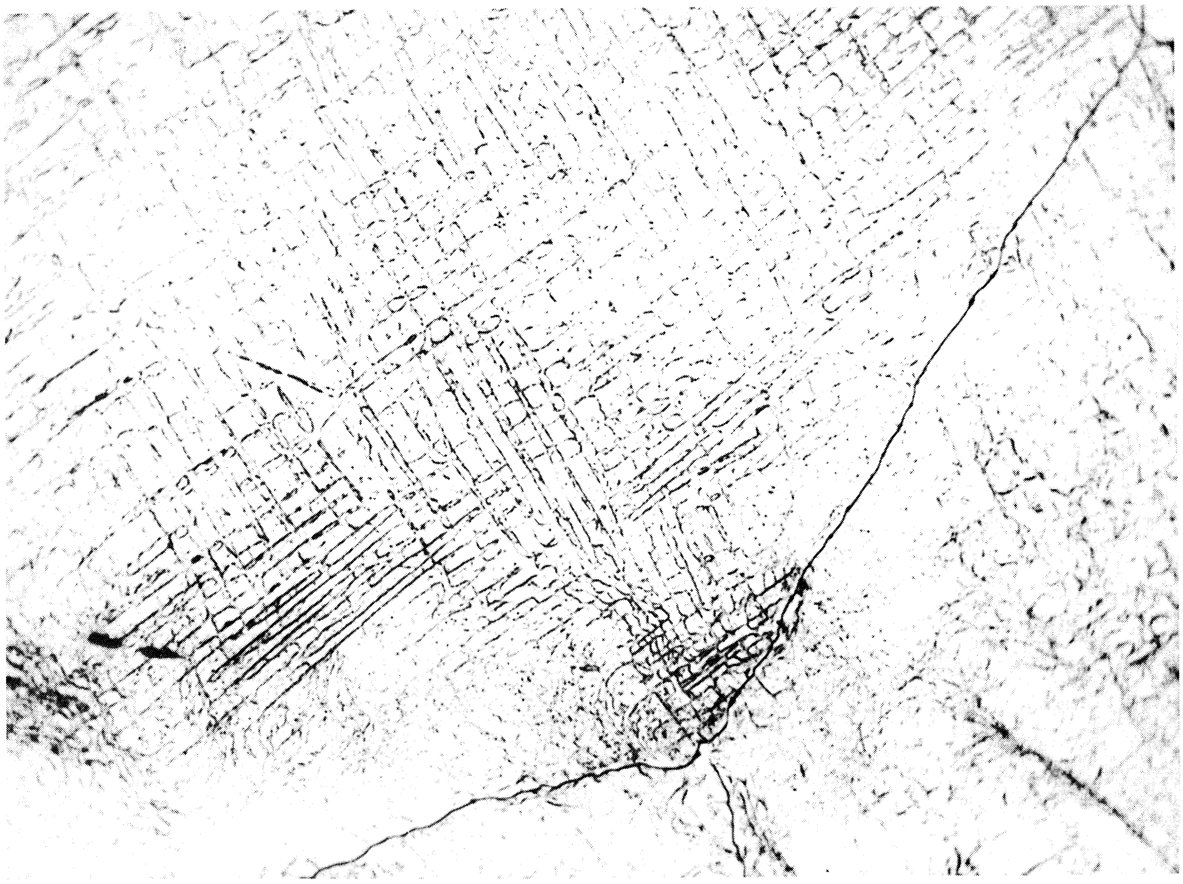
TABLE I

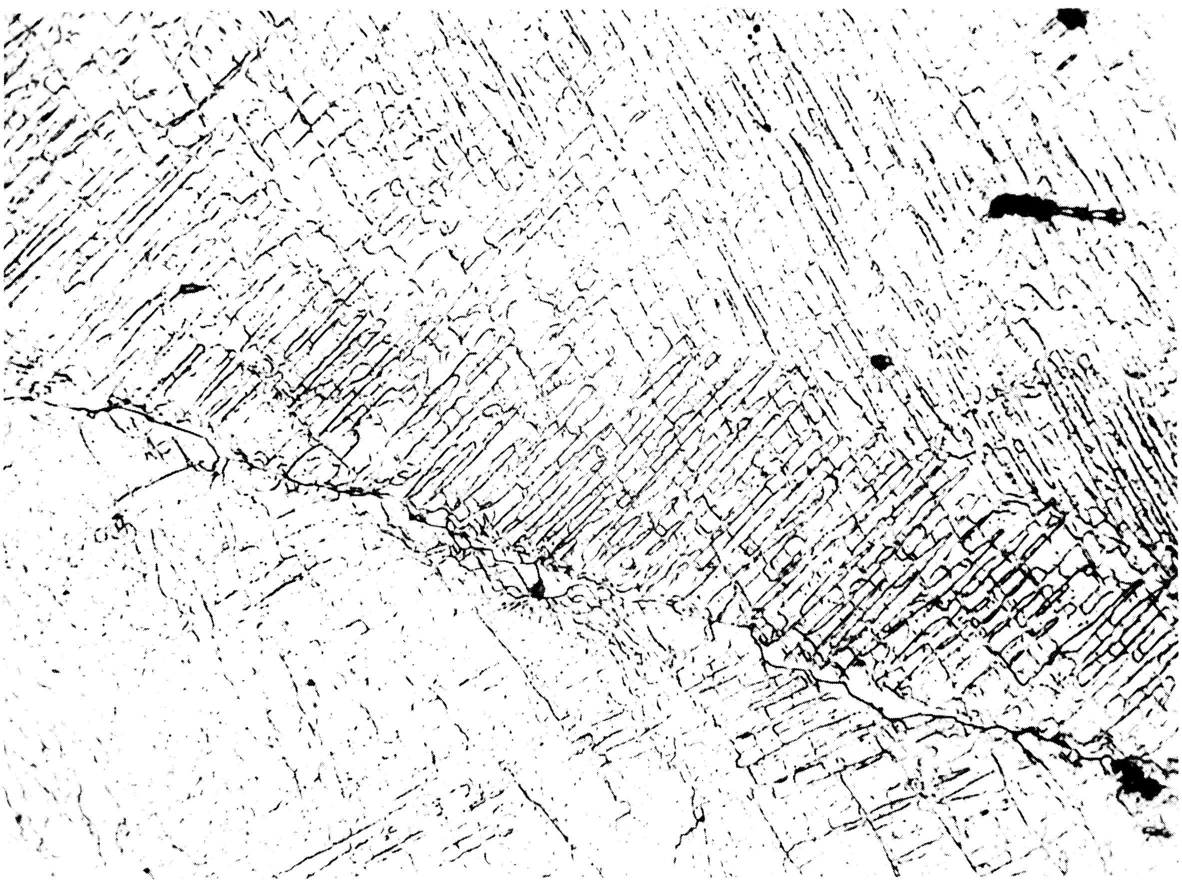
Influence of Hot Working on the Creep-Rupture Properties of "17-22-A"V Steel at 1100°F

Processing Conditions	Stress (psi)	Rupture Life(hrs)	Elong. (%in 4D)	R. A. (%)	Min. Creep Rate(%/hr)	Def. on Load(%)	Time to Reach Indicated Total Deformation (hrs) 0.2%	0.5%	1.0%	2.0%
1 hr. 2200°F, AC* to 2100°F	55,000	23.8	0.9a	2.0	0.0253	0.36	b	<1.0	14.2	---
and rolled 50% reduction, AC	40,000	131.0	1.1a	1.6	0.00560	0.30	b	12.0	98.0	---
+ tempered 6 hrs. 1250°F, AC.	29,000	459.8	0.7a	1.2	0.00106	0.16	20.0	280.0	--	---
(Austenite 100% recrystallized during rolling)										
1 hr. 2200°F, AC to 2000°F	55,000	38.5	2.7	3.6	0.0288	0.37	b	<1.0	11.0	35.0
and rolled 50% reduction, AC	40,000	230.7	1.5a	3.2	0.00385	0.27	b	10.0	125.0	---
+ tempered 6 hrs. 1250°F, AC.	29,000	832.6	1.2a	2.4	0.00084	0.22	b	200.0	650.0	---
(Austenite partially recrystallized during rolling)										
1 hr. 2200°F, AC to 1900°F	55,000	32.2	9.1	16.8	<0.63	0.36	b	<1.0	1.5	12.0a
and rolled 50% reduction, AC	40,000	324.6	1.5a	4.8	0.00256	0.33	b	3.0	150.0	---
+ tempered 6 hrs. 1250°F, AC	29,000	1229.1	0.9a	3.6	0.00052	0.22	b	390.0	1135.0	---
(No recrystallization during rolling--100% hot-cold-worked austenite)										

*AC - air cooled.
a - Obtained through small extrapolation of elongation versus time curve.
b - Deformation exceeded on loading.
< - "Less than".



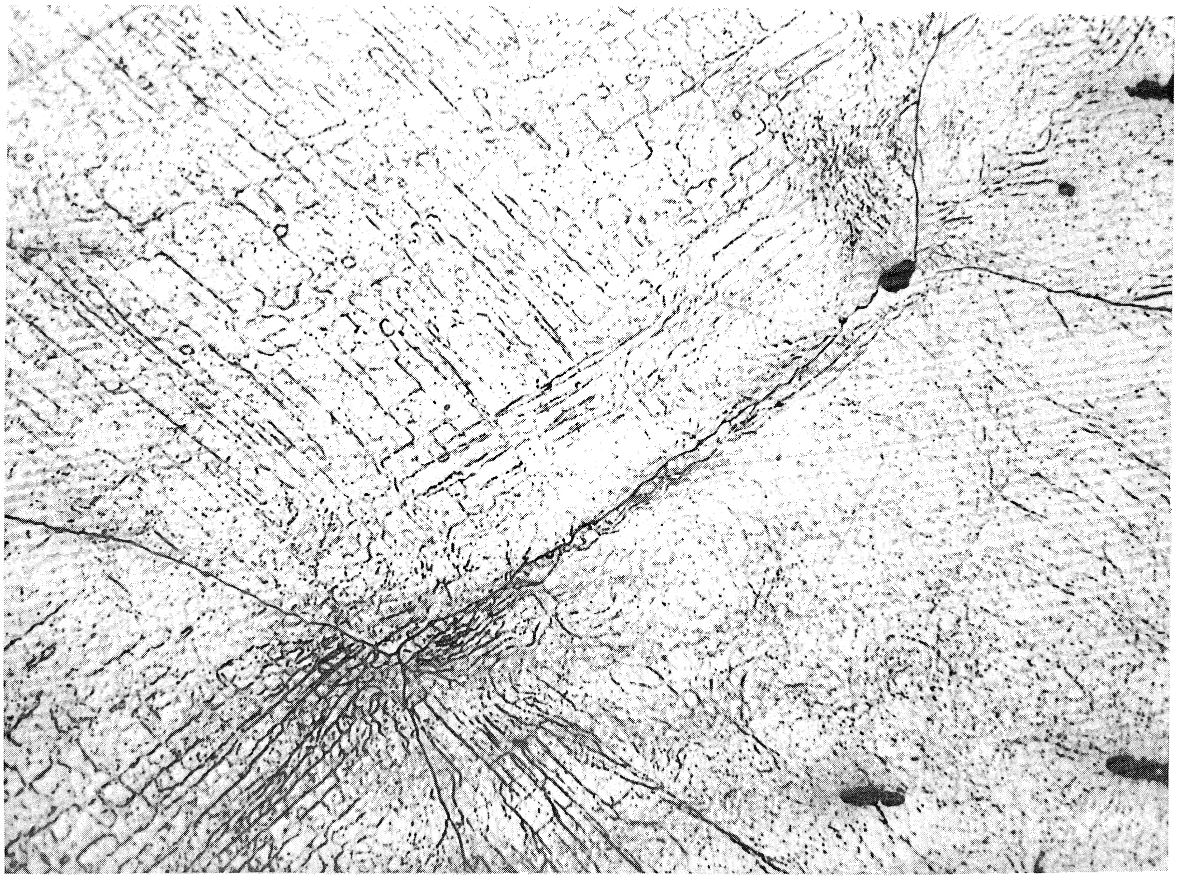
(a) Rolled 5.8 Percent at 1800°F + Tested 702 Hours at 1100°F and 11,000 psi.



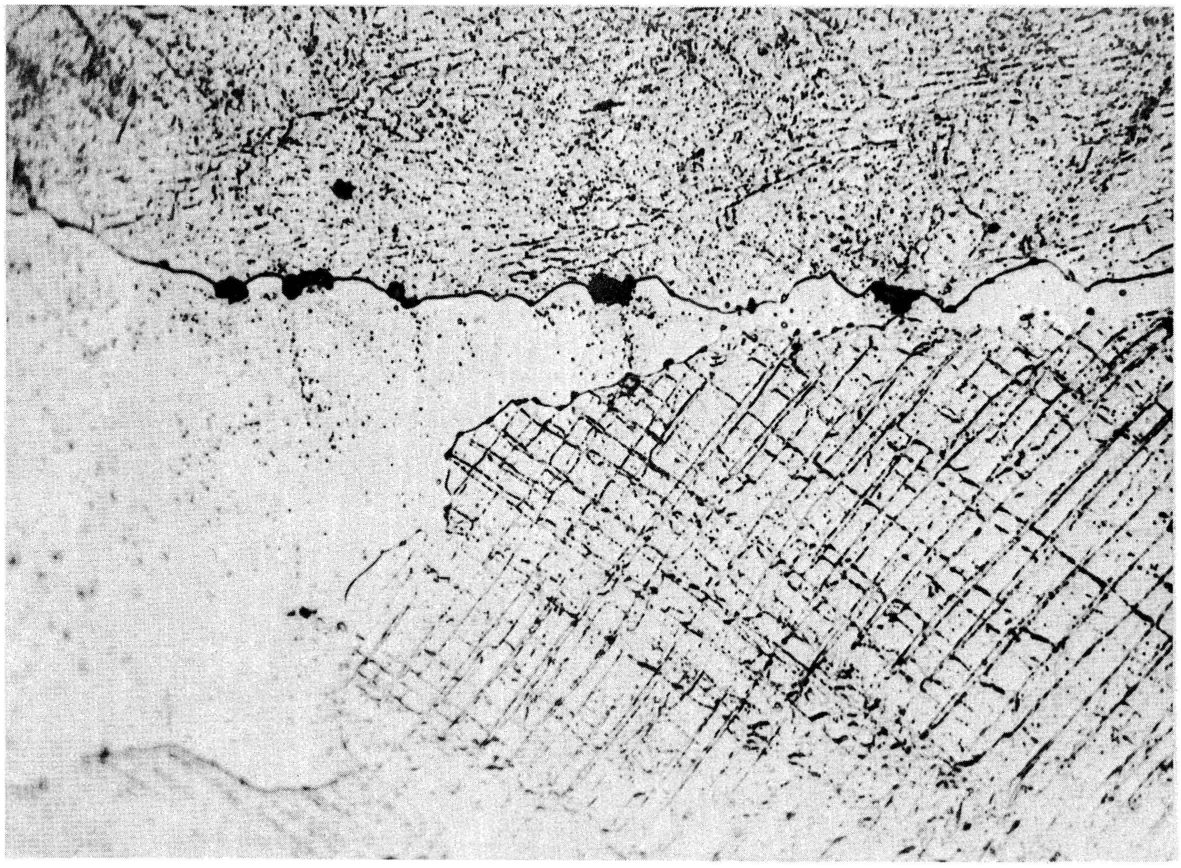
(b) Same as in (a) except different area.

Figure 1. - Substructures in Rolled "A" Nickel after Creep Testing. All Micrographs are at X1000 Magnification.

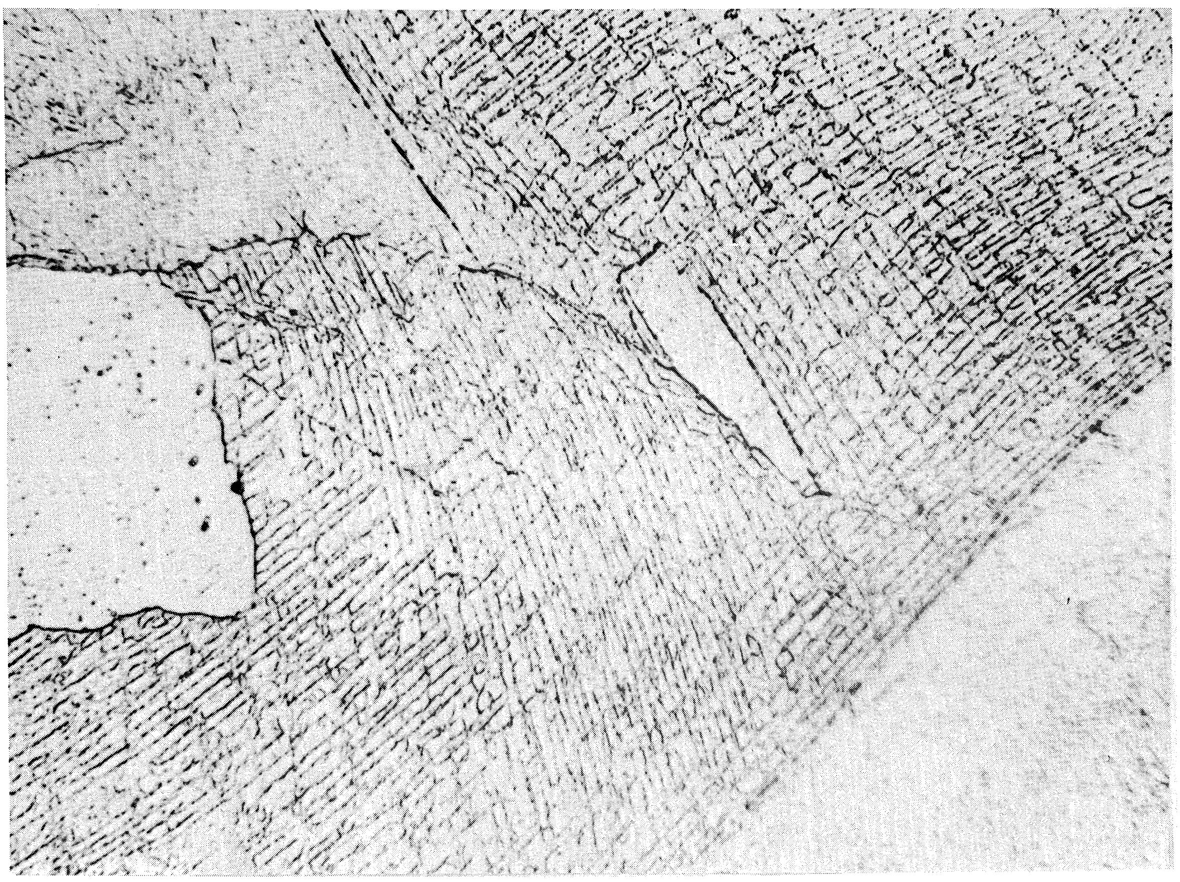
(c) Same as in (a) except different area.



(d) Same as in (a) except different area.

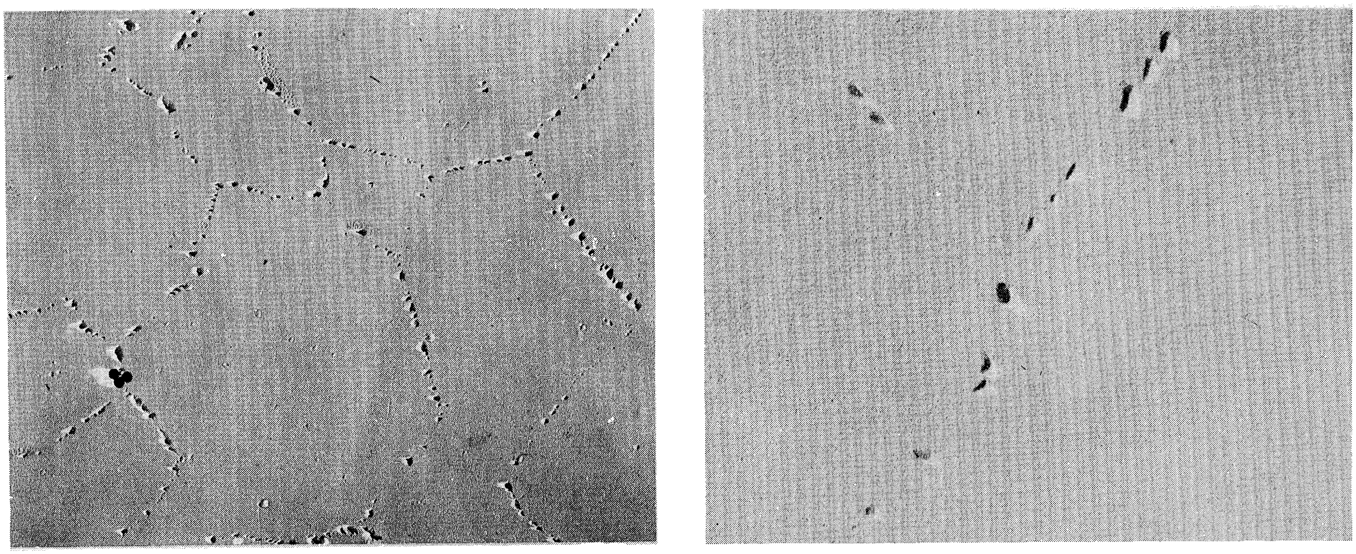


(e) Rolled 15.7 Percent at 1800°F + Tested 843 Hours at 1100°F and 11,000 psi.

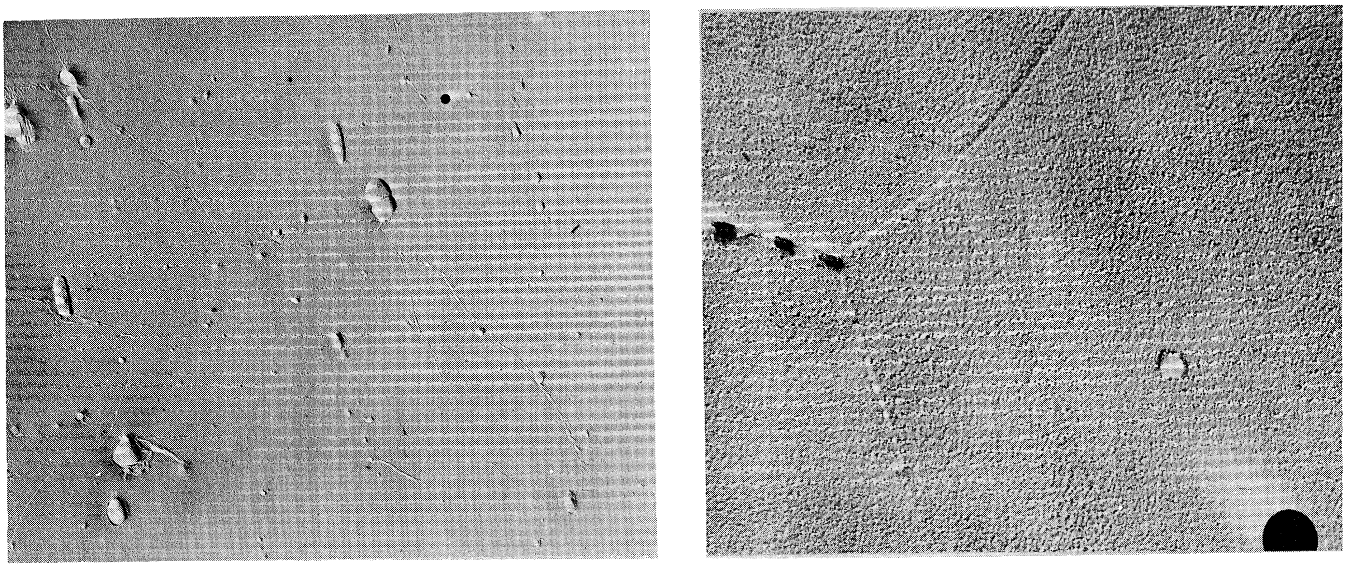


(f) Rolled 27.3 Percent at 1800°F + Tested 988 Hours at 1100°F and 11,000 psi.

X3,500 X22,000



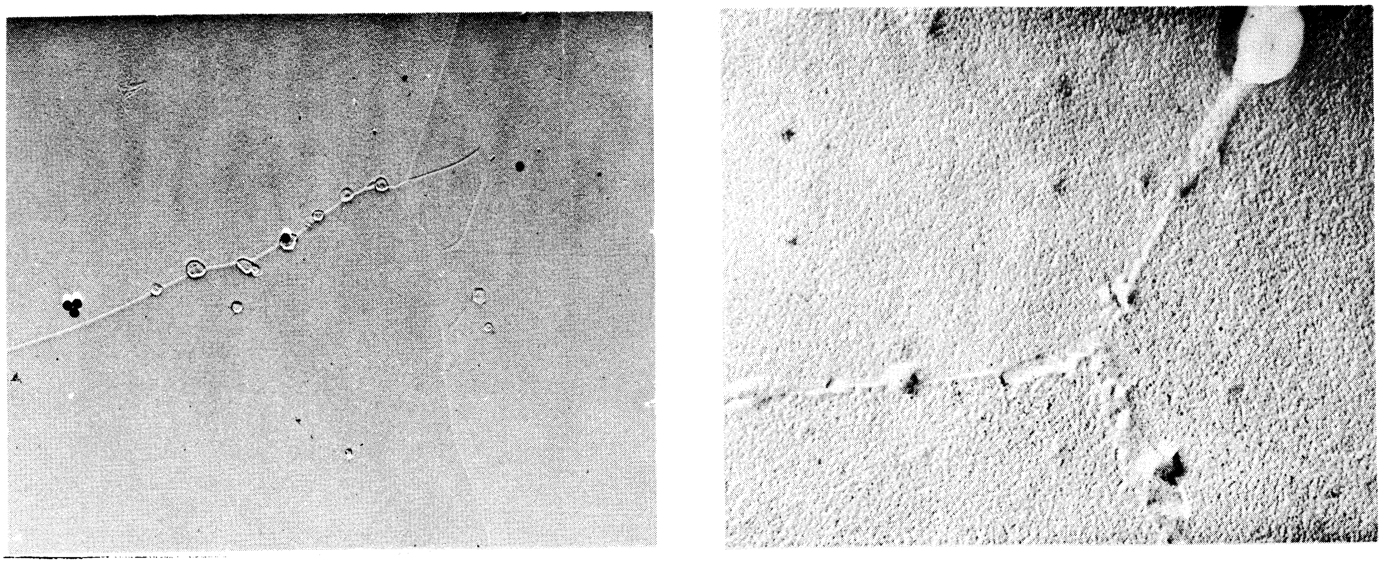
(a) Solution Treated and Aged.



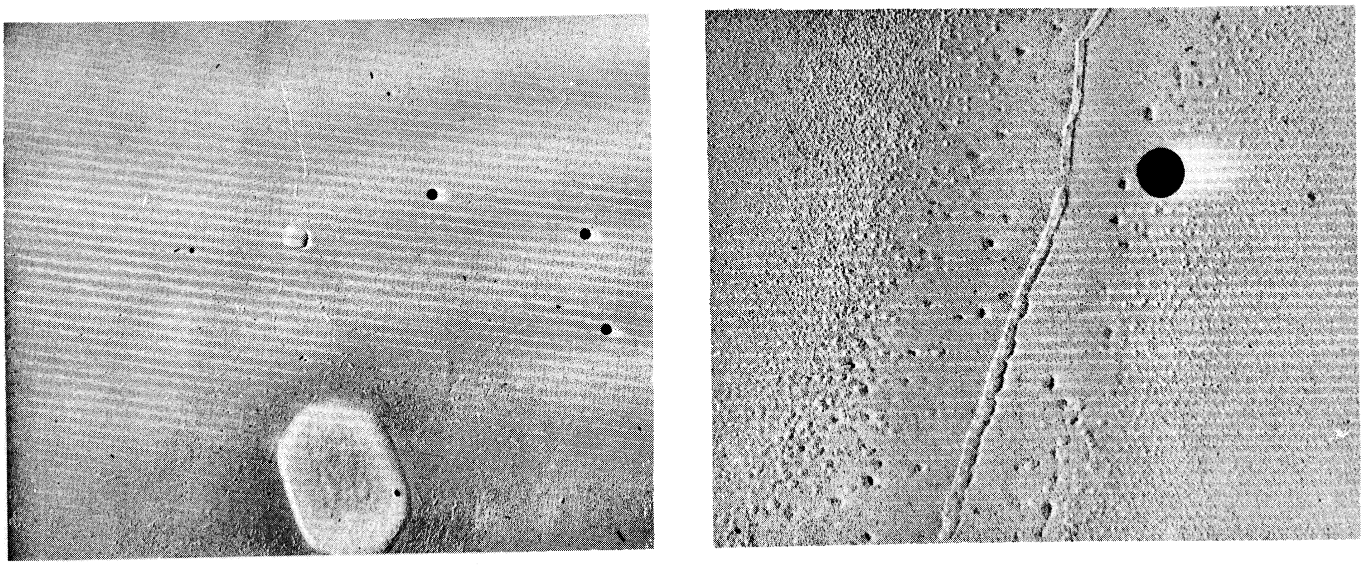
(b) 45 Minutes at 1700°F., Air Cooled + Solution Treated and Aged.

Figure 2. - Electron Micrographs of A-286 Alloy Solution Treated 1 Hour at 1650°F and Aged 16 Hours at 1325°F with (a) No Prior Heating, (b) Prior Heating to 1700°F, (c) Prior Heating to 1950°F, and (d) Prior Heating to 2200°F. Collodion replicas are shadowed with palladium. Polystyrene balls are 0.34 micron in diameter.

X3,500 X22,000



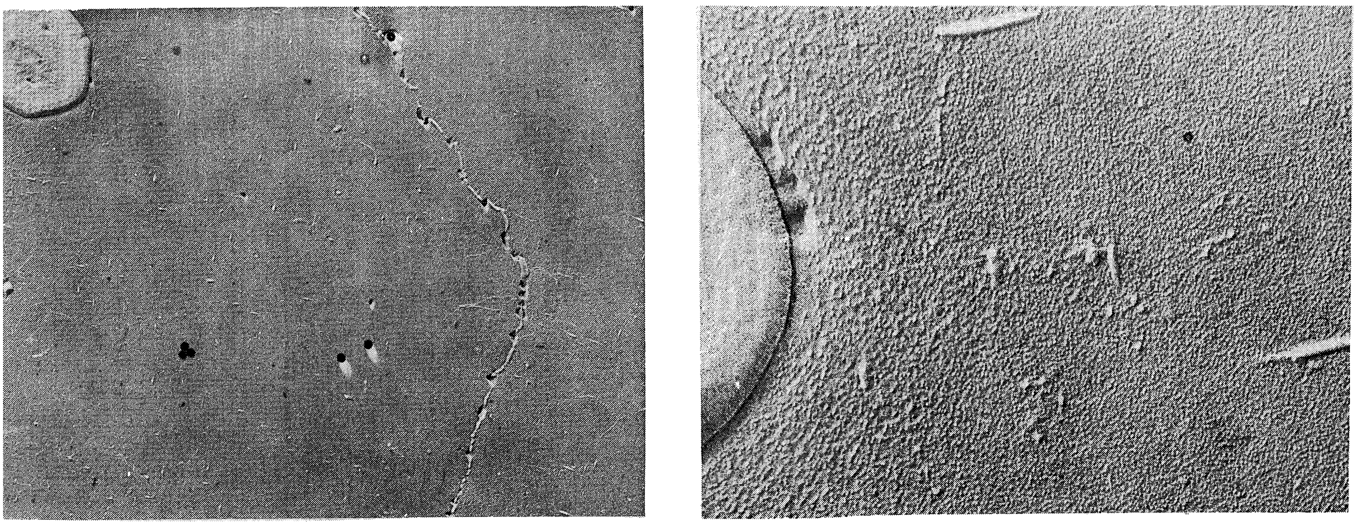
(c) 45 Minutes at 1950°F, Air Cooled + Solution Treated and Aged.



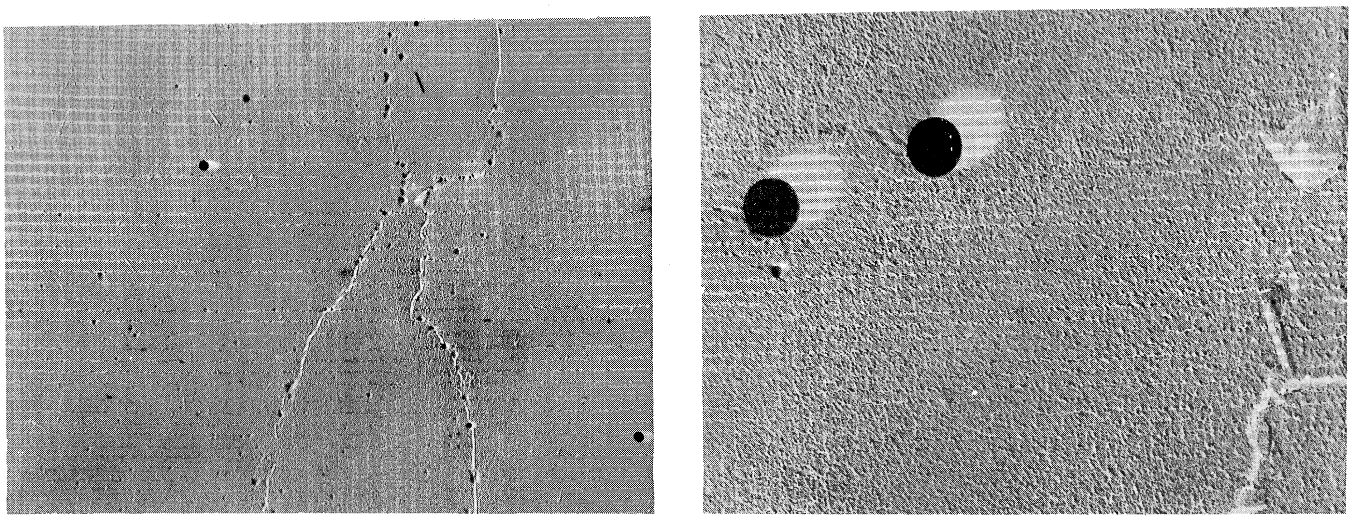
(d) 45 Minutes at 2200°F, Air Cooled + Solution Treated and Aged.

Figure 2. - Concluded.

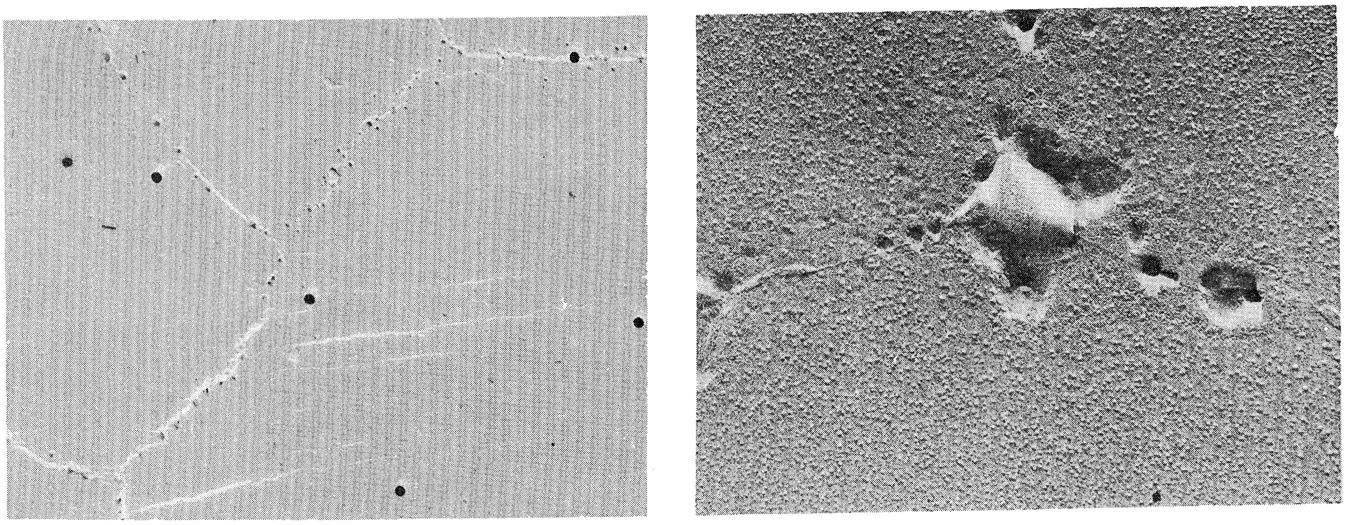
X3,500 X22,000



(a) 14.5 Percent Reduction by Rolling.



(b) 20.2 Percent Reduction by Rolling.



(c) 35.1 Percent Reduction by Rolling

Figure 3. - Electron Micrographs of A-286 Alloy Rolled at 2200°F, Air Cooled + Solution Treated and Aged.

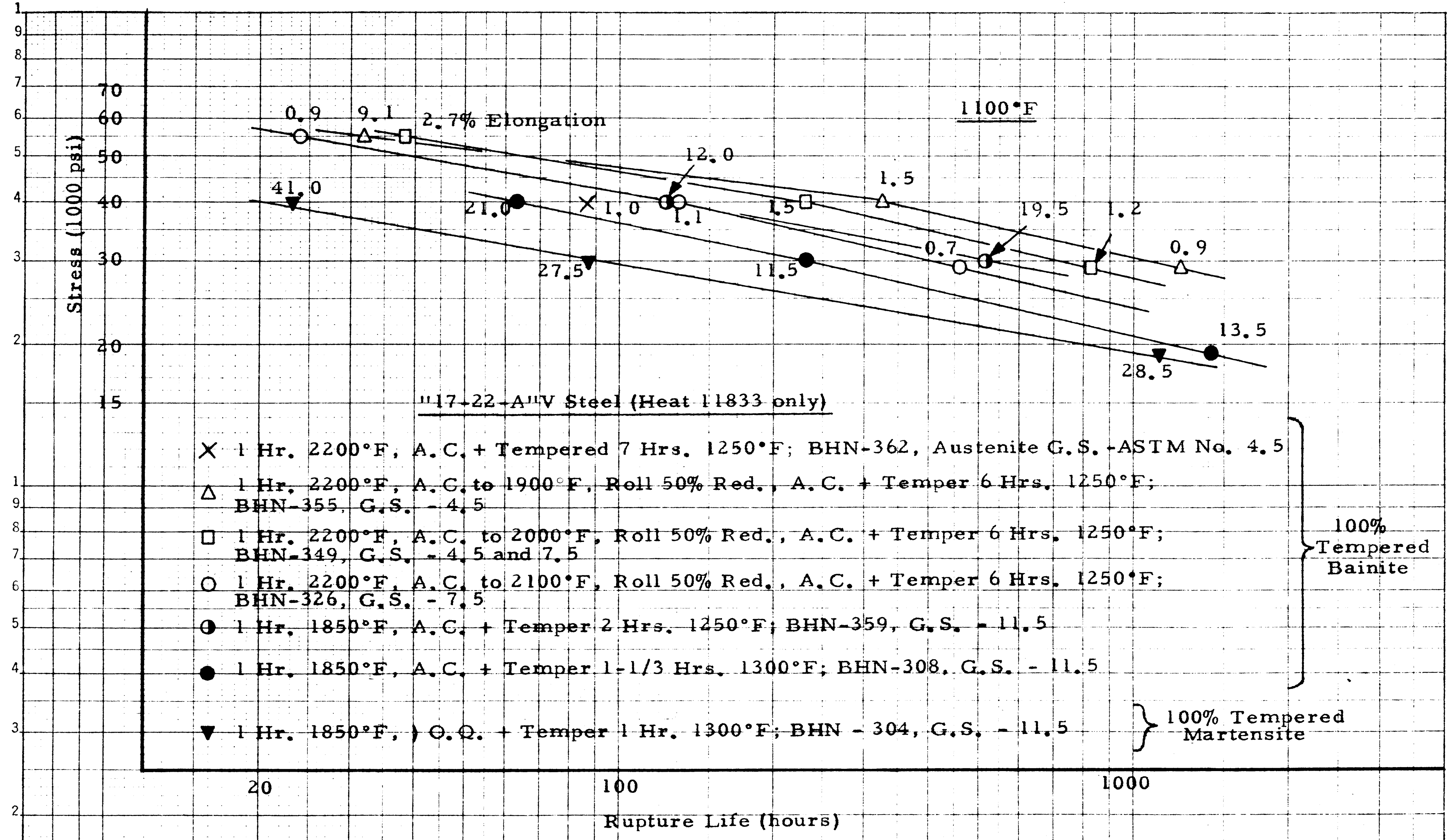


Figure 4. - Variation of Stress-Rupture Properties of "17-22-A"V Steel at 1100°F with Processing Conditions.

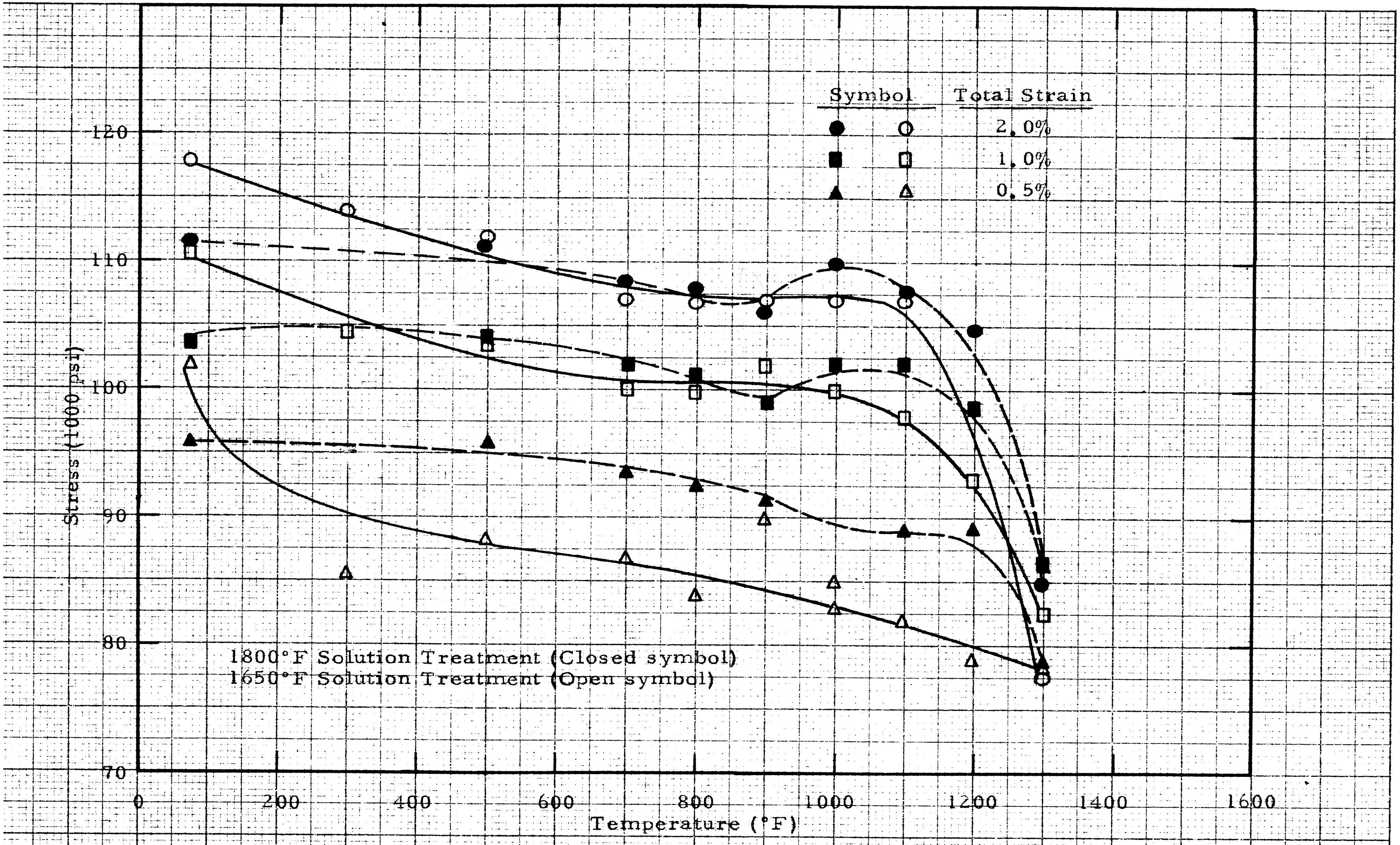


Figure 5. - Stress-Temperature Curves of A-286 Alloy for 0.5, 1.0 and 2.0 Percent Total Strain on Tensile Testing.

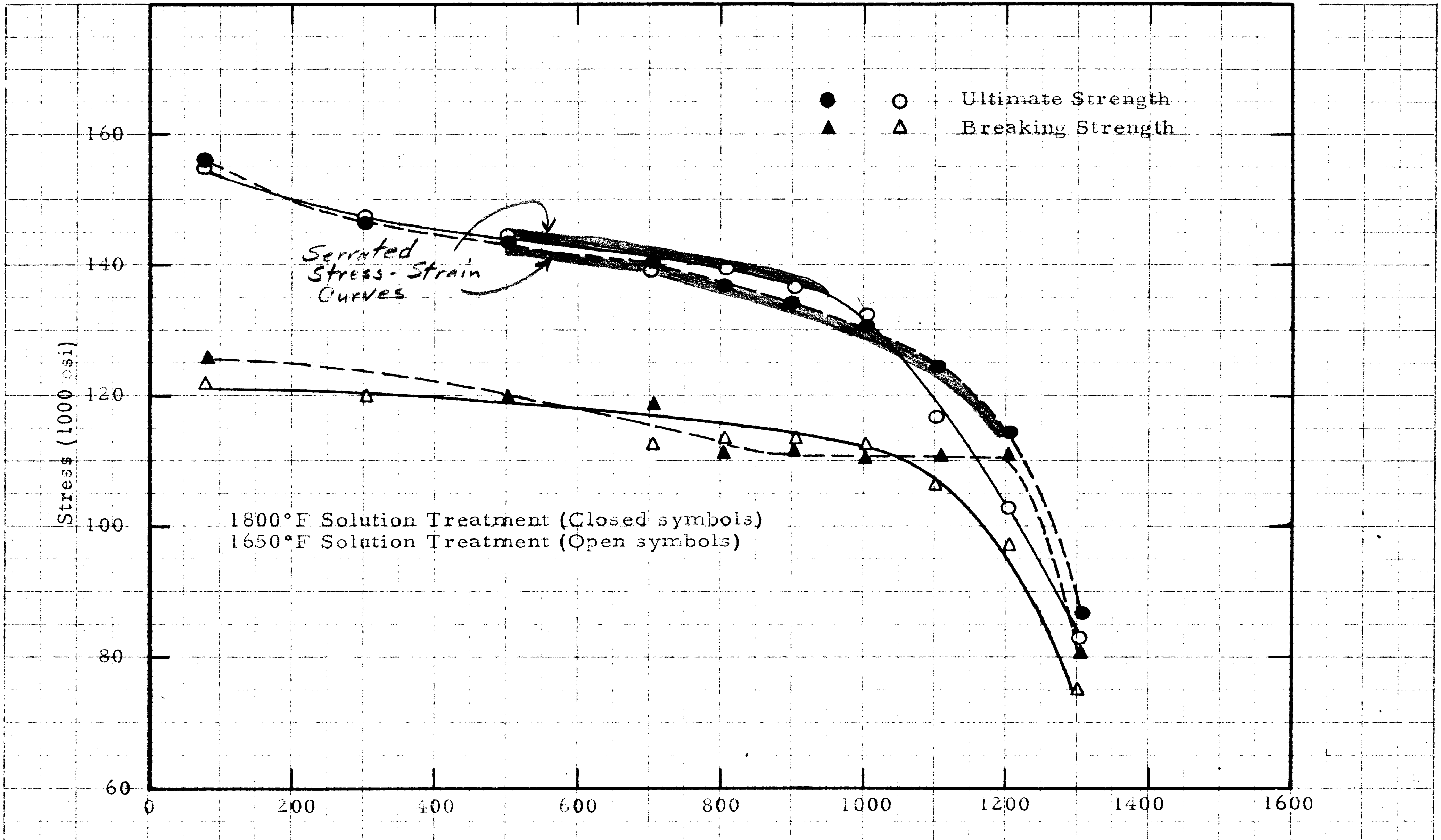


Figure 6. - Ultimate and Breaking Strength versus Temperature of A-286 Alloy on Tensile Testing.

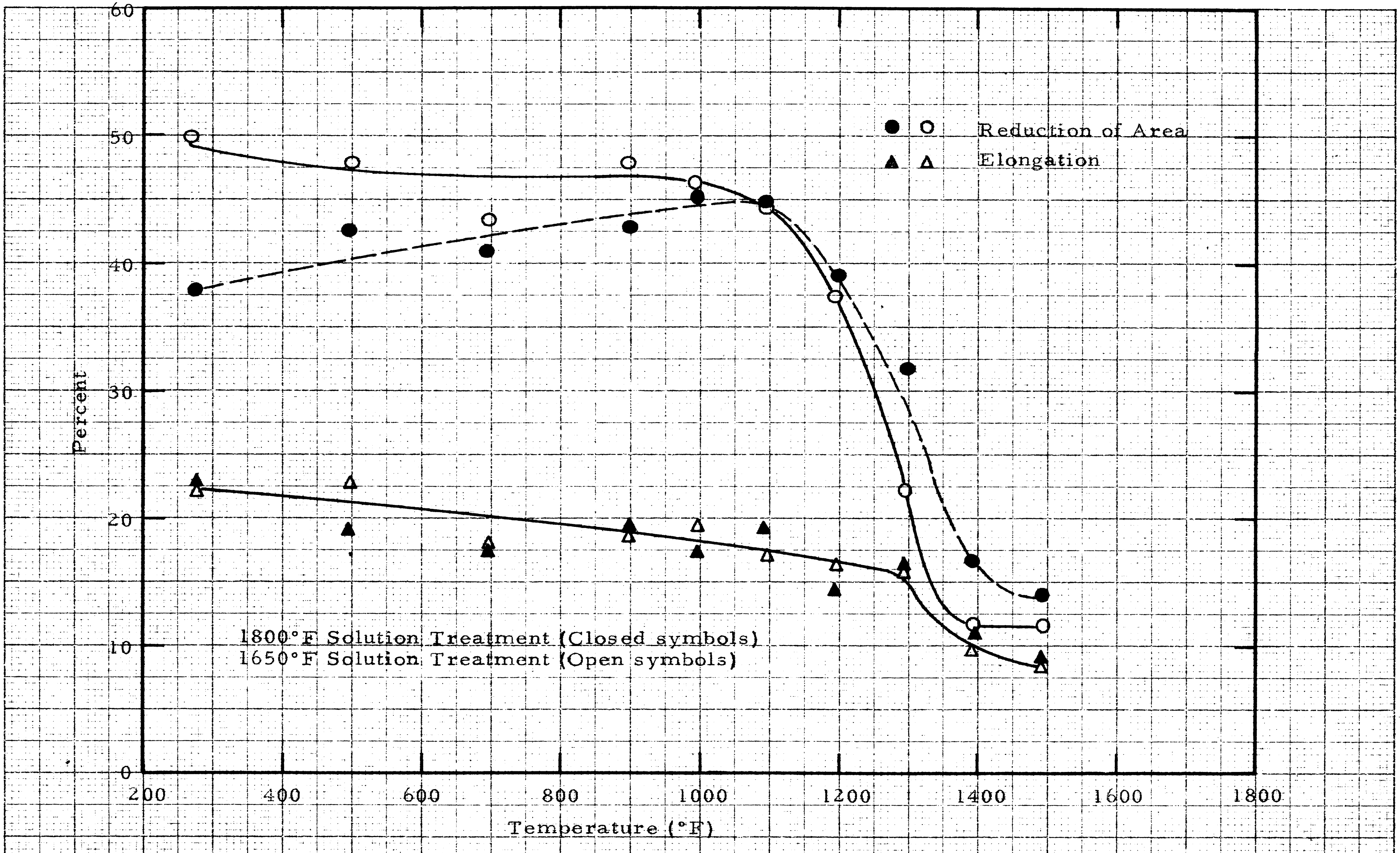


Figure 7 - Percent Elongation and Reduction of Area versus Temperature for A-286 Alloy on Tensile Testing.

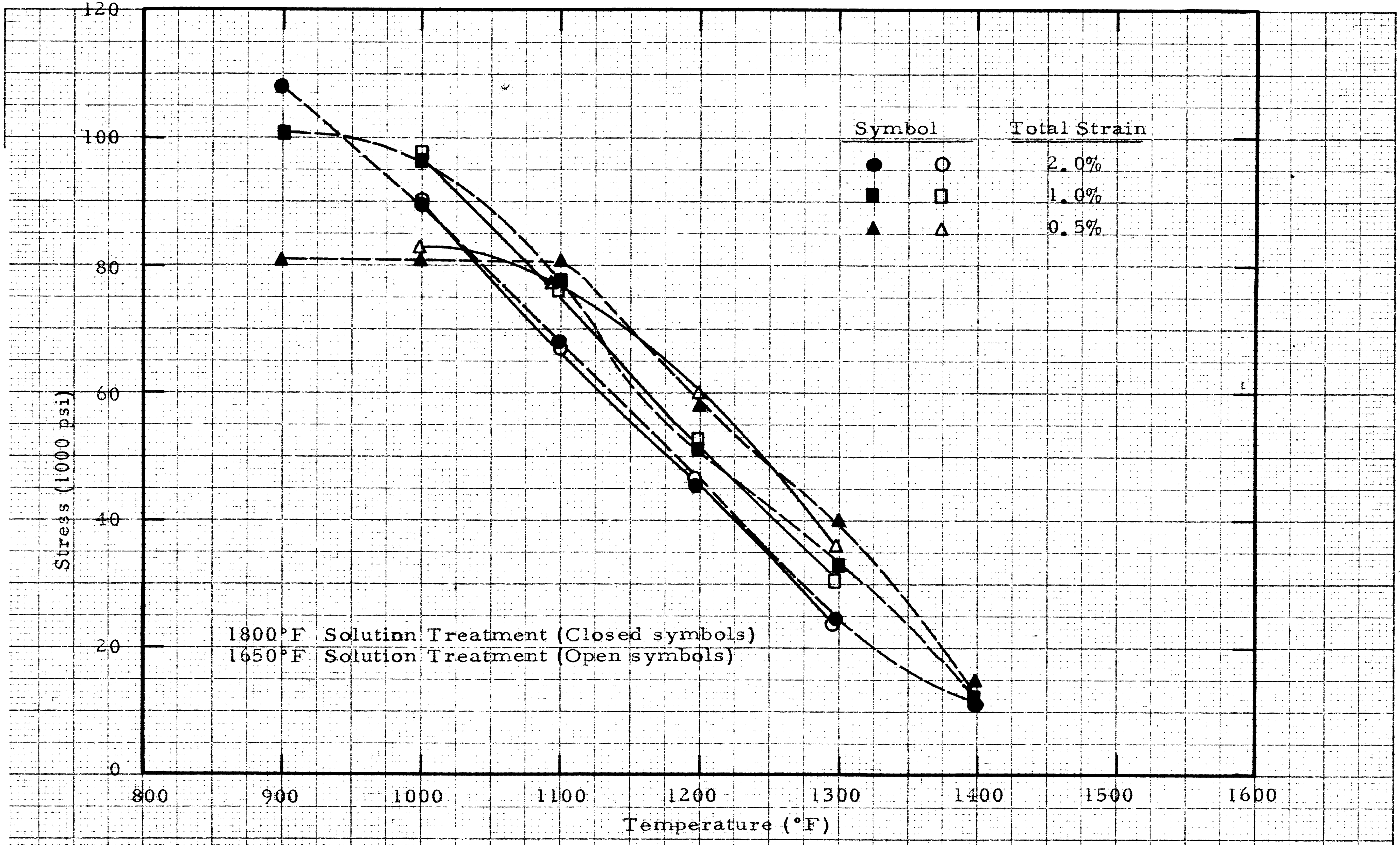


Figure 8. - Stress-Temperature Curves for A-286 Alloy at Total Strains of 0.5, 1.0 and 2.0 Percent for a Strain Rate of 0.004 Percent per Hour.

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