THE UNIVERSITY OF MICHIGAN RESEARCH INSTITUTE
ANN ARBOR, MICH.

SECOND PROGRESS REPORT

TO

MATERIALS LABORATORY

WRIGHT AIR DEVELOPMENT CENTER

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

A. P. Coldren
J. E. White
J. W. Freeman

Project 2760

Air Force Contract
AF 33(616)-5466
Task No. 73512

September 30, 1958

SUMMARY

Progress is reported for research carried out under Air Force Contract No. AF 33(616)-5466 covering the work period of June 30, 1958 to September 30, 1958. The present research is related to previous investigations carried out under the sponsorship of the Materials Laboratory, Wright Air Development Center, concerning basic factors which affect the creep-rupture properties of heat-resistant alloys. The current work is divided into two general areas of study: (1) the influence of hot working on structure and creep-rupture properties, and (2) the relationship between strain aging phenomena and high-temperature strength.

A microstructural study of "A" Nickel specimens rolled at 80°, 1400°, and 1600°F and rupture tested at 1100°F and 20,000 psi indicated that the leveling-off and final falling-off of the log rupture life versus percent reduction by rolling curves was a result of extensive recrystallization during the test.

Attempts to observe fine substructures in samples of as-rolled "A" Nickel samples by means of X-ray diffraction techniques have been unsuccessful to date. A coarse type of substructure was observed in Berg-Barrett X-ray micrographs taken of annealed samples and in back-reflection photograms of samples rolled small amounts at 1600° and 1800°F. It is planned to explore the possibility of heat treating or creep testing rolled samples in such a way that the fine substructure in all of the samples can be revealed by etching. In the as-rolled condition, fine substructures have thus far been revealed by etching only in the samples rolled 5 percent at 1600° and 1800°F.

The simultaneous recrystallization temperature range for "17-22-A" V steel was determined to be 1950° to 2100°F for a 50-percent reduction of area by rolling. On the basis of these data samples were prepared in which complete, partial, and no recrystallization of the austenite occurred prior to the transformation to 100 percent bainite during cooling from rolling.

In the strain aging study, constant strain rate tests were carried out at a strain rate of 0.004 percent per hour on 1020 steels to determine the way strain aging manifests itself in this material under creep conditions at elevated temperatures. The result of these tests show clearly that steels susceptible to strain aging exhibit higher strengths for small deformation at about 800°F.

Concurrently, testing at a constant strain rate of 0.004 percent per hour was initiated on the high-temperature alloy A-286. A series of tensile tests were also conducted on the A-286 alloy not only to acquire tensile data but to see in what manner strain aging manifests itself in this alloy at higher strain rates.

INTRODUCTION

This report, the second quarterly progress report issued under Air Force contract No. AF 33(616)-5466, covers work done from June 30, 1958 to September 30, 1958.

The background of the present research was presented in the First Progress Report (Ref. 1). Briefly, the purpose of the current work is to establish basic relationships between processing variables and high-temperature properties of heat-resistant alloys so that these alloys can be produced with consistently high strengths. Under earlier contracts with the Wright Air Development Center the effect of heat treatment on the structure and creep-rupture properties of low-alloy steels was studied. Since that time the research has been broadened. There are currently two main phases of research:

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

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

The hot working studies are being carried out on three types of materials, namely, (a) a commercially pure metal ("A" Nickel), (b) a precipitation-strengthened, austenitic alloy (A-286), and (c) a ferritic alloy of low strategic-element content ("17-22-A" V steel). The strain aging studies are at the present time being carried out on the following three materials: (a) a silicon-deoxidized 1020 steel (steel C) which is susceptible to strain aging in all conditions (b) an aluminum-deoxidized 1020 steel (steel F) which may or may not be susceptible to strain aging depending on the heat treatments, and (c) a high-strength, austenitic alloy A-286.

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 Pretroleum 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

****************************** BRC SEARCH FORM ******************************

OCLC RLIN SERIES LOCATION: ____________

1. Access point searched: A P COLDREN	2. Access point searched: J E WHITE James Edward ?

1.

	yes	no		yes	no
MIR. Aut.	✓	✓	****	***	***
MIR. Bib.	___	✓	DCF=a	___	___
OCLC Aut.	___	✓	****	***	***

Sample call number needed? yes / no
Sample call number found? yes / no

2.

	yes	no		yes	no
MIR. Aut.	?	___	****	***	***
MIR. Bib.	___	✓	DCF=a	___	___
OCLC Aut.	___	___	****	***	***

Sample call number needed? yes / no
Sample call number found? yes / no

3. Access point searched: J W FREEMAN	4. Access point searched:

3.

	yes	no		yes	no
MIR. Aut.	___	___	****	***	***
MIR. Bib.	✗	___	DCF=a	___	✓
OCLC Aut.	___	___	****	***	***

Sample call number needed? yes / no
Sample call number found? yes / no

4.

	yes	no		yes	no
MIR. Aut.	___	___	****	***	***
MIR. Bib.	___	___	DCF=a	___	___
OCLC Aut.	___	___	****	***	***

Sample call number needed? yes / no
Sample call number found? yes / no

****

SERIES

1.	2.
NUMBERED : UNNUMBERED	NUMBERED : UNNUMBERED
USED : NOT USED	USED : NOT USED
CLASS SET : CLASS SEP	CLASS SET : CLASS SEP
RECORDED : NOT RECORDED	RECORDED : NOT RECORDED
NEW : SEND TO CSR	NEW : SEND TO CSR
MIRLYN Aut. record found YES / NO	MIRLYN Aut. record found YES / NO
with decisions YES / NO	with decision YES / NO
AF Card found YES / NO	AF Card found YES / NO
OCLC Aut. record found YES / NO	OCLC Aut. record found YES / NO

CLS 10/14/92 Searcher: CV DATE: 11 / 12 / 92

```
LTUL DONE                                                              AAP5098
                                                   NOTIS CATALOGING              ADA8
UL- AAP5098 FMT A RT z DT 05/03/90 R/DT none      STAT mn E/L n
SRC   LANG ??? ROM n MOD   UNIQUE a GOVT ? S/SYS a D/I n SUBD   NUM ? S/TYP ?
NAME c SUBJ a SER b KIND a H/ESTAB ? T/EVAL a IP ? RULES c

010:  : |a n  83006892
035/1:  : |a (UTLAS)17945865
040:  : |a DLC |c DLC |d ICU
100:10: |a White, J. E., |d 1918-
400/1:10: |w nna |a White, James Edward, |d 1918-
670/1:  : |a Author's Applied seismic waves, 1983: |b CIP t.p. (J. E. White;
Dept. of Geophysics, Colorado Sch. of Mines, Golden, Colo.) CIP data sheet (b.
5/10/18)
670/2:  : |a LCCN 64-22962: His Seismic waves, 1965 |b (hdg.: White, James
Edward, 1918- ; usage: J.E. White)
```

```
LTUL DONE
                                                                      BAB3496
                                                        NOTIS CATALOGING         ADA8
UL- BAB3496 FMT B RT a BL m DT 10/06/92 R/DT none      STAT nn E/L   DCF    D/S D
SRC d PLACE miu LANG eng MOD ? T/AUD ? REPRO ? D/CODE s DT/1 1957 DT/2 ????
CONT ???? ILLUS ???? GOVT ? BIOG ? FEST ? CONF ? FICT ? INDX ? ME/B ?

040:   : |a MiU |c MiU
099/1:   : |a UMR0589
100:10: |a Decker, Raymond Frank.
245:10: |a Survey on creep damage to metals / |c prepared by Raymond F. Decker,
Jeremy V. Gluck, James W. Freeman.
260:   : |a Ann Arbor, Mich. : |b Engineering Research Institute, University of
Michigan, |c [1957?]
300/1:   : |a 25 leaves ; |c 28 cm.
650/1: 0: |a Metals |x Creep.
700/1:10: |a Gluck, Jeremy V. |q (Jeremy Victor)
700/2:10: |a Freeman, James W.
710/3:20: |a University of Michigan. |b Engineering Research Institute.
```

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

The objective of the present work is to correlate structures with creep-rupture properties in three types of heat-resistant alloys for the case where the structural differences are introduced by systematically varying the hot-working conditions. Two of the materials, "A" Nickel and A-286 alloy, were hot worked and creep-rupture tested under a previous WADC contract (Ref. 2). The third, "17-22-A" V ferritic steel, was hot-worked only recently; specimens are now being machined for creep-rupture testing. The selection of the hot-working conditions for "17-22-A" V is discussed, and photomicrographs of the resulting structures are presented in this report.

Commercially Pure Metal ("A" Nickel)

"A" Nickel was selected as a "simple material" for the hot-working study on the assumption that its relatively simple microstructure would permit identification of the effects of such variables as strain hardening, recovery, substructures, and recrystallization as influenced by prior working conditions. In most commercial high-temperature alloys precipitation effects or other phase changes complicate, or could even mask completely, the influence of these "matrix variables" on the creep-rupture properties.

Microscopic Examination after Rupture at 1100°F and 20,000 psi

Rupture samples which had been tested at 1100°F and 20,000 psi were sectioned longitudinally and examined for evidence of recrystallization during testing over a 1/2-inch length of the specimen, beginning at the fracture. Samples which had been rolled at 1800°F were all partially recrystallized by the rolling operation. Since it is difficult to distinguish between recrystallization by hot working and recrystallization during testing, the degree of recrystallization during testing after rolling at 1800°F has not been established.

Recrystallized areas were observed in all of the samples. Specimens with rolling reductions from zero to about 27 percent exhibited only minor amounts of recrystallization along the grain boundaries. Because this type of recrystallization was found in the annealed samples it was presumed to be induced by the large creep strains before fracture. In samples with higher rolling reductions, however, recrystallization was not restricted to the old grain boundaries, but extended more or less into the old structure, depending upon the amount and temperature of reduction. The amount of recrystallization increased with percent reduction and decreased, for a given reduction, as the rolling temperature was raised. This suggests that the cases of widespread recrystallization resulted from the combined effect of creep strains and rolling strains. Approximate values of percent recrystallized area are presented beside representative points on the log rupture life versus percent reduction curves in Figure 1 (formerly Fig. 3 in Ref. 2). The values were obtained by a lineal analysis. It appears that the leveling-off of the rupture time versus percent reduction curves (Fig. 1) and the final falling-off of some of the curves were probably caused by the more extensive recrystallization during testing which was observed in the samples with the larger reductions.

X-Ray Diffraction Studies

In the First Progress Report (Ref. 1) X-ray diffraction techniques were discussed in connection with substructure studies. The "spotty ring back reflection technique" used by Gay and Kelly (Ref. 3) was chosen as the most adaptable to the present work in view of the available equipment, time, and funds. A series of preliminary back-reflection X-ray photograms were taken using the smallest commercially made collimator available (about 0.5 mm diameter). The results were summarized in the First

Progress Report (Ref. 1), and the suggestion was made that more information could be obtained by repeating the series of exposures using a smaller X-ray beam.

Accordingly, the size of the X-ray beam was drastically reduced by means of a collimator constructed by co-axially mounting two 0.004 in. (0.1mm) diameter pinholes 2.75 inches apart. For the specimen-to-collimator distance used, the diameter of the beam at the surface of the specimen was calculated to be 0.35 mm or about three times the average grain diameter. Under these conditions, the number of arcs per Debye Ring was small enough that little or no overlapping of arcs occurred even at the highest reduction. This permitted the measurement of individual arc lengths which are directly related to the total range of orientations of subgrains within a single grain. Unfortunately, however, the substructural feature of greatest interest, the subgrain size, was not revealed. It was hoped that the small beam would reveal individual spots within the arcs, permitting a mean subgrain size to be calculated by the method of Gay and Kelly (Ref. 3). Apparently, some other means for measuring subgrain sizes must be found.

Nevertheless, to obtain a general picture of the influence of hot working on the diffraction patterns, a series of back reflection photograms were taken on the samples rolled at 1600°F. Representative photograms from this series are shown in Figure 2. A beam twice the size of the one described above was used so that the exposure time would be more reasonable. As a result, overlapping of arcs occurred for reductions exceeding about 9 percent.

Two additional types of information from back-reflection photograms are possible: the misorientation between adjacent subgrains and the relative amounts of lattice distortion. The requisite for calculating the misorientation between adjacent subgrains is that the diffraction arcs be resolved into a series of discrete spots, or at least a series of intensity maxima. The misorientation or angle of tilt is calculated from the distance between the spots or intensity maxima, the diameter of the Debye Ring, and the Bragg Angle. In the present work, subgrain spots were found only in samples rolled small amounts at high temperatures. Subgrain spots were best revealed in photograms of a sample rolled 5.8 percent reduction at 1800°F. Examples of these are presented in Figure 3.

Figure 4 shows how large differences in lattice distortion can, by mere inspection, be seen in the relative (radial) widths of the Debye Rings. Cold working, which is known to produce severe lattice distortions in nickel, caused the rings in Figure 4a to be so broad and diffuse that the α_1/α_2 doublet was not resolved. Rolling the same amount at 1400° or 1600°F., however, left considerably less lattice strains as indicated by the sharp rings in Figures 4b and 4c. This was probably the result of "simultaneous recovery" immediately after rolling while the material was still hot.

The possible usefulness of another X-ray technique was explored. The technique, which is commonly referred to as Berg-Barrett X-Ray Microscopy, involves (a) irradiating a large area of the sample by a wide beam of characteristic radiation striking the sample at a glancing angle, (b) recording the diffraction spots on a very fine grain photographic plate mounted only about 1 mm away from the sample, and (c) enlarging the images up to 200 diameters. Unless a grain is perfect its diffraction spot will exhibit an internal structure which is an indication of the presence of a substructure.

Berg-Barrett micrographs were taken from several annealed and hot-worked nickel samples. Evidently the substructure introduced by hot working is on a scale too fine to be revealed by the Berg-Barrett technique. This is demonstrated by the fact that the fine substructure which has been revealed by etching in the sample rolled 5.8 percent at 1800°F (Figs. 17 and 18, Ref. 2) could not be seen in a Berg-Barrett micrograph of the same sample. One of the reasons is that light microscopy has a much greater resolving power than the X-ray technique. A maximum magnification of 2000 diameters is possible with light microscopy; whereas, 200 diameters is about the greatest magnification obtainable with the Berg-Barrett technique.

The Berg-Barrett micrographs revealed a coarse type of substructure typical of metals annealed at temperatures well below their melting point. The only influence of hot working that could be seen in the Berg-Barrett micrographs was a sort of smearing of the pre-existing, coarse substructure.

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

The study of the effect of hot working on the creep-rupture strength of "17-22-A"V steel is complicated by the phase transformation which occurs on cooling from rolling by the secondary hardening reaction which occurs on tempering in the range of 1200°F. As indicated in Reference 1, it was at first reasoned that the most likely prospect for improving the creep-rupture strength of steels like "17-22-A"V would be to combine the high strength associated with high austenitizing temperatures and the high ductility associated with fine bainitic structures. Under ordinary conditions, high heat-treating temperatures result in coarse structures which lack good ductility.

To produce the desired structure it was decided to use a high temperature of heating followed by hot working the bar as it cooled down to slightly above the minimum temperature for complete simultaneous recrystallization. To determine the minimum recrystallization temperature a series of bars were heated to 2200°F for 1 hour, air cooled to various temperatures and then rolled 50 percent in two quick passes. The rolling temperatures were controlled by varying the time of air cooling from 2200°F according to a master cooling curve established for this particular barstock with a thermocouple at the center of the bar. The results are given in Figure 5 where the austenite grain size is plotted versus rolling temperature. For a 50-percent reduction, simultaneous recrystallization began at about 1950°F and was complete at about 2100°F.

On the basis of these data three bars of "17-22-A"V were heated to 2200°F for 1 hour, air cooled to 2100°, 2000°, and 1900°F., respectively, and reduced 50 percent by rolling. The resulting structures were all 100 percent bainite, but the prior austenite had experienced complete, partial, and no recrystallization, respectively. The three rolled bars were tempered at 1250°F for 6 hours in accordance with the accepted treatment of this material. The tempered structures are shown in Figures 6, 7, and 8. The relative creep-rupture strengths will be determined next.

FUTURE WORK

Further attempts will be made to solve the problem of revealing the fine substructure in severely rolled nickel samples. If a lack of sharpness and/or an insufficient concentration of impurity atoms is the reason why some subgrain boundaries are not revealed by etching, then heating--especially under a low stress--might help by allowing polygonization (sharpening of subgrain boundaries) and diffusion of impurity atoms to subgrain boundaries. Spot checks will be made on specimens which have been subjected to creep at 1100°F and 11,000 psi to see whether fine substructures can be revealed by etching.

The electron microscope study of A-286 structures will be resumed immediately.

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

Because the constant strain rate tests that were run during the previous period at a strain rate of 0.25 percent per hour did not show the expected differences in strength for aging and non-aging steel in the temperature range for creep, it was concluded that further testing should be carried out at a considerably slower strain rate. A strain rate of 0.004-percent per hour was selected as being sufficiently low to show peaks in the aging steel in the creep range if they were present. This is in accordance with the hypothesis that the high creep resistance aging steels should show peaks in stress versus temperature curves for small amounts of deformation at slow strain rates.

PROCEDURE

Processing of Materials

Steel F, an aluminum-deoxidized 1020 carbon steel was tested under two different conditions of heat treatment. One rendering the material susceptible to strain aging, the other minimizing it. The first treatment, designated by FA, was air cooling after one hour at 2150°F. This heat treatment is above the coarsening temperature of the austenite which results in solution of nitrogen. Air cooling from this temperature assures that sufficient nitrogen will remain in solution to render this steel susceptible to strain aging. Furnace cooling this same material, designated FN, after one hour at 1650°F results in most of the nitrogen being precipitated from solution as aluminum nitride causing the steel to be essentially immune to strain aging.

Steel C, a silicon-deoxidized 1020 steel, is susceptible to strain aging in all conditions of heat treatment. To reduce variables due to heat treatment, steel C was also furnace cooled after one hour at 1650°F. This is to assure that the main difference observed between steel C and the non-aging condition of steel F will be due to strain aging phenomenon.

The A-286 alloy concurrently being tested was given a one hour solution treatment at 1800°F and oil quenched, plus aged 16 hours at 1325°F.

Testing

During this period constant strain rate tests were run on the carbon steel at a strain rate of 0.004 percent per hour over the temperature range of 400° to 1000°F. These tests were run in the same manner as the previous 0.25 percent per hour tests in standard creep units maintaining the strain rate by manual adjustment of the load. The tests were run to 2-percent total deformation which involved a testing time of 500 hours per test.

Tests of the same nature were run on the A-286 alloy at 1000°, 1100°, and 1200°F. Tensile tests were also run on the A-286 alloy during this period to acquire a complete set of tensile data from room temperature to 1200°F as well as to get some indication as to the way strain aging manifests itself at higher strain rates in this alloy.

RESULTS AND DISCUSSION

Test data from the constant strain rate tests at 0.004-percent per hour revealed that inflection points in the stress-temperature curves occur between 800° and 900°F for the aging carbon steels. Tensile test data obtained during this period on A-286 revealed how strain aging manifests itself at high strain rates in this alloy. Tests run at a strain rate of 0.004-percent per hour on A-286 resulted in brittle fracture with less than 2-percent total strain at 1000° and 1100°F.

Carbon Steel

1. A very definite inflection occurred in the strength versus temperature

curves at about 800°F for the carbon steels susceptible to strain aging (Figs. 9 and 10) when treated at 1650°F to reduce strain aging, steel FN showed no evidence of these inflections at about 800°F (Fig. 11). The inflections for the aging steels are even more evident by observing Figures 12 through 15 which compare the curves for both aging and non-aging steels at total deformations of 0.25, 0.5, 1.0, and 2.0 percent. The maximum difference in strength between aging steel C and non-aging steel FN appeared at about 850°F.

The solid points in Figure 9 represent tests run at 0.004 percent per hour on aging steel C which was air cooled after 1 hour at 1650°F. The previous series of constant strain rate tests at 0.25 percent per hour were conducted on steel C in this condition. The two tests were carried out at 600° and 800°F to illustrate the effect of heat treatment, by altering the carbide structure, on the strength under conditions of constant strain. The high strength of steel FA air cooled from 2150°F compared to steels C and FN which were both furnace cooled from 1650°F is attributed mainly to the stronger carbide structure which forms on faster cooling.

2. Figures 12-15 also illustrate that the strengthening phenomenon prevalent in aging steels C and FA is strongly dependent on the amount of strain as well as the temperature of testing. Looking at steels C and FN the maximum different at 0.25 percent deformation occurs at about 900°F (Fig. 12). As the total strain is increased the temperature of maximum difference in stress moved toward lower temperatures, reaching 800°F at a deformation of 2 percent (Fig. 15). The same inflection characteristics are observed for steel FA only at slightly lower temperatures than for steel C.

In Figure 10 it is evident that the strengthening phenomenon has become inactive where the strength drops off rapidly for the 2 percent total strain at about 800°F. In this temperature range it is observed that the strength for 2 percent deformation falls below that for smaller deformations. This condition arises because the material has become weakened to the point where load must be removed to maintain the constant strain rate. This is equivalent to third-stage creep for constant load creep tests.

3. It is evident from Figures 12-15 that after passing the inflection point the strengths of the aging steels drop off rapidly approaching the lower strength of non-aging steel FN with increasing deformations and temperatures. The exact mechanism which is responsible for the strain-activated strengthening seems to become inactive or lose its effectiveness when either the temperature or plastic strain becomes too high. Although several theories as to the exact nature of this mechanism can be put forth at this time more work must be done in this area to pin down this phenomenon conclusively as well as to evaluate its applicability to materials other than carbon steels.

A-286 Alloy

Several important observations can be made from the results of the tensile tests conducted on A-286 as to the way strain aging manifests itself in this alloy. From the hot-hardness testing program carried out by Captain Domian on this same heat of A-286 at the Materials Laboratory, Wright Air Development Center, it was evident that strain aging phenomena are present in this alloy. The tensile test results show that the manifestations for this alloy are not the same as for carbon steels. The yield point, often considered synonomous with strain aging, is not present in this alloy over the investigated temperature range of room temperature to 1200°F. The tensile curve at room temperature not only exhibited no yield point but also displayed no serrations of any kind in the plastic region which was somewhat surprising. However, as the temperature was increased to 500°F definite serrations did begin to occur in the plastic region. The serrations increased in both frequency and intensity as the temperature was raised, reaching a maximum at about 900°F. Above this temperature the serrations became less pronounced, disappearing almost entirely at 1200°F. The temperature range for maximum serration corresponds to the temperature where Captain Domian found strain aging manifesting itself in his hot hardness investigations.

Strain was recorded by a mirror extensometer system out to 2 percent wherever possible. The resulting values of stress versus temperature for 0.5, 1.0, and 2.0 percent elongation are plotted in Figure 16. The peaks are very pronounced at 1.0 and 2.0 percent, but due to the fact that the jagged serrations encountered at 900°, 1000°, and 1100°F caused the mirror readings to be somewhat erratic, these curves must be viewed with some reservation.

The curve of tensile strength versus temperature (Fig. 17) shows no marked peak as do curves for aging carbon steels; however, a hesitation or inflection does occur over the temperature range from 700° to 900°F which may be a manifestation of strain aging in the alloy. The breaking strength plotted on this same figure, drops off sharply at 800°F, then remains essentially constant out to 1200°F. This is attributed to the sharp drop-off in the percent reduction of area over this temperature range.

The reduction of area versus temperature curve (Fig. 18) shows a somewhat unexpected increase at 800°F which is surprising in that the region of maximum strain aging effect is usually associated with the minimum values of elongation and reduction of area. The major decrease in ductility in A-286 alloy is reported at about 1300°F. No abnormal strengthening effect has, however, been observed in this region.

Constant strain rate tests on A-286 alloy at a strain rate of 0.004 percent per hour show no evidence of the strengthening mechanism between 1000° and 1200°F (Fig. 19) that was observed for carbon steels at this strain rate. Figure 19 does show that for all tests run the strengths decrease at the total deformation is increased. This means that the alloy becomes weaker or that the stress must be lowered to maintain the same strain rate as the test progresses. This condition is equivalent to third-stage creep where a material exhibits continuously increasing creep rate under constant load.

Of the three tests run on this alloy at a strain rate of 0.004 percent per hour two of them at 1000° and 1100°F fractured with slightly less than two percent total strain with a brittle failure.

FUTURE WORK

The results on carbon steel demonstrate that high creep resistance in alloys suceptible to strain aging is associated with a retention of strength to higher temperatures as the strain rate and total strains considered are reduced. Future work will be mainly concerned with identification of the mechanism and checking the observation for steel on other alloys.

REFERENCES

1. A. P. Coldren, J. E. White, and J. W. Freeman, "Studies of Heat-Resistant Alloys," First Progress Report to Wright Air Development Center, Contract No. AF 33(616)-5466 (June 30, 1958).

2. A. P. Coldren and J. W. Freeman, "An Investigation of the Relationship Between Microstructure and Creep-Rupture Properties of Heat-Resistant Alloys," proposed Wright Air Development Center Technical Report 58-204, 1958.

3. P. Gay and A. Kelly, "X-Ray Studies of Polycrystalline Metals Deformed by rolling. I. The Examination of the Harder Metals, Copper, Nickel and Iron," Acta Cryst., 6 (February, 1953) 165.

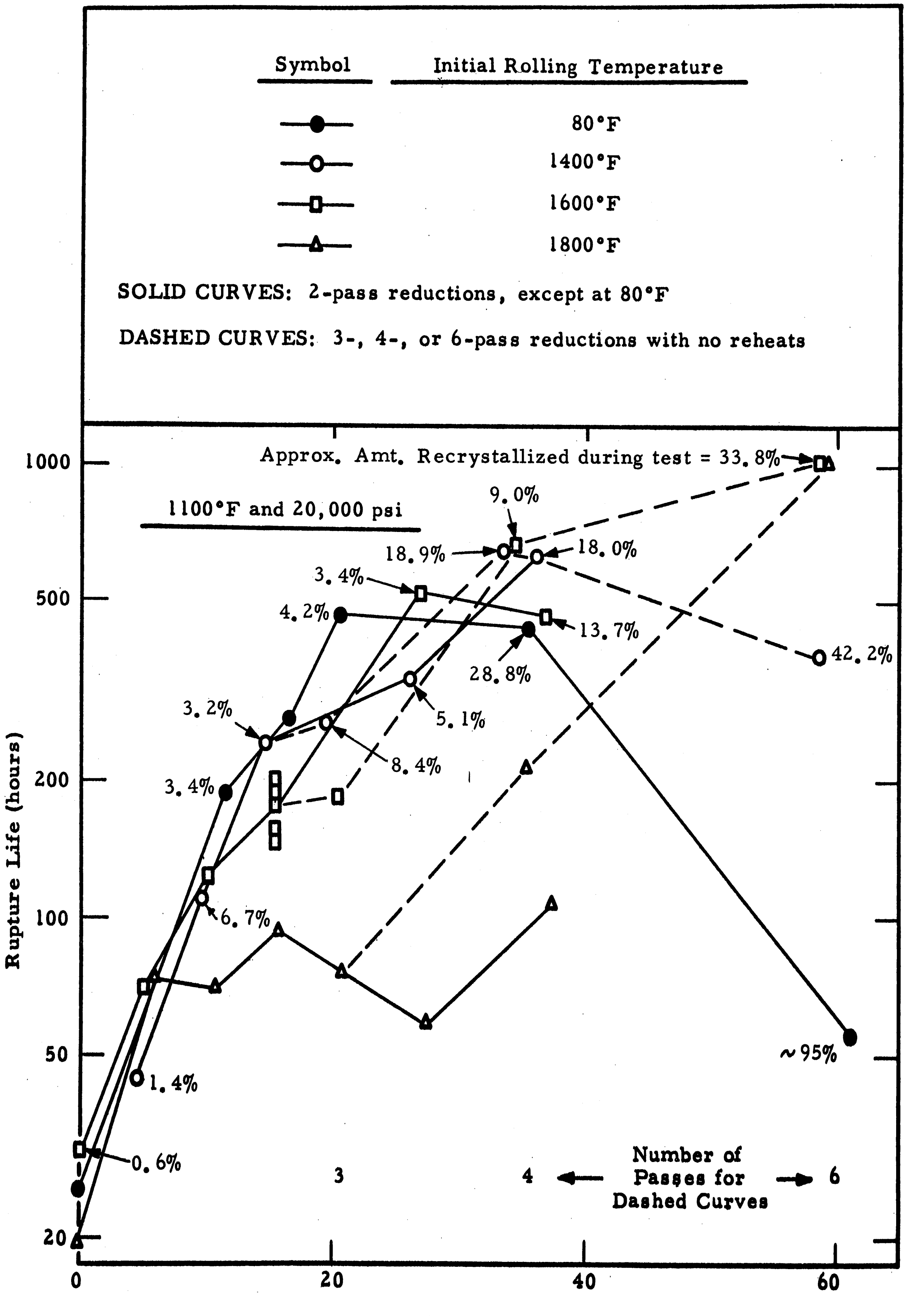
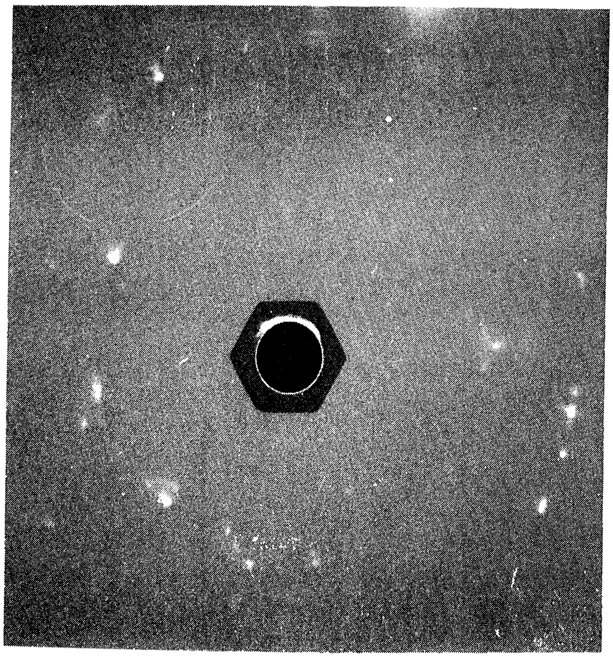
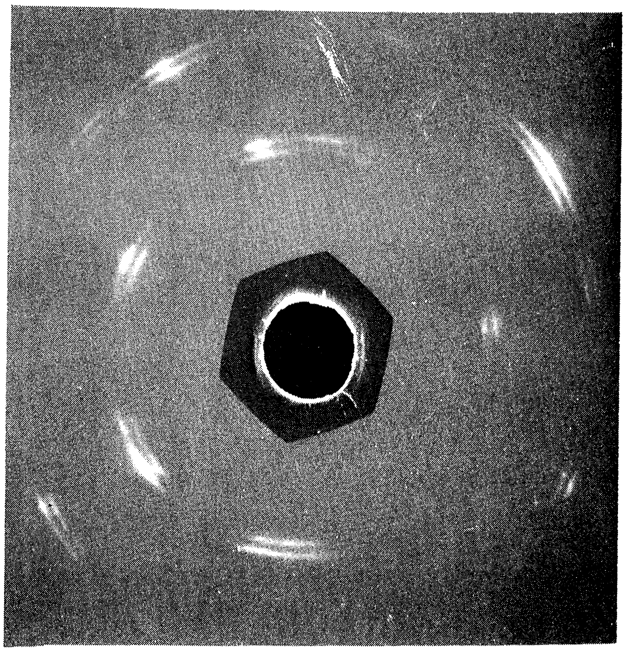


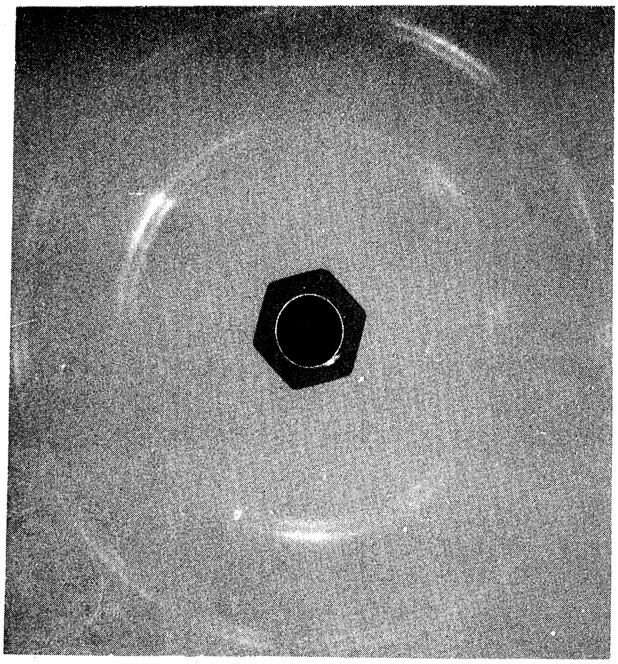
Figure 1. - Variation of log rupture life with rolling conditions for "A" Nickel tested at 1100°F and 20,000 psi, showing approximate extent of recrystallization during testing.



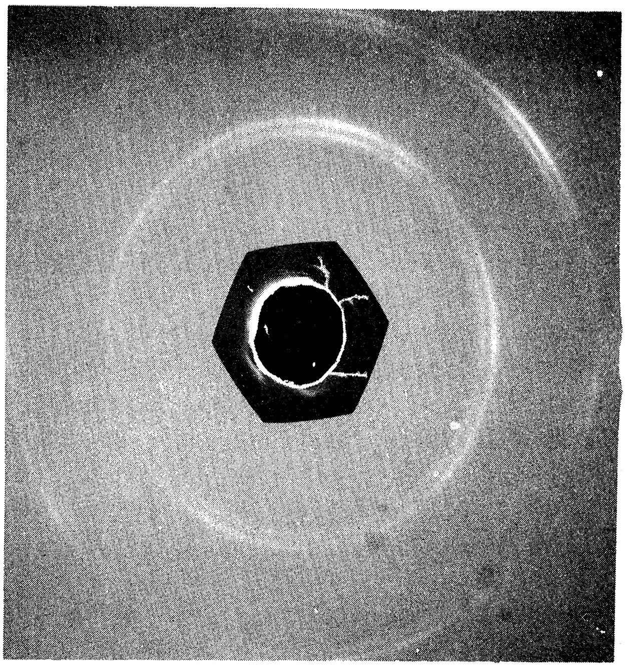
(a) Annealed at 1600°F



(b) 5.3 Percent Reduction at 1600°F



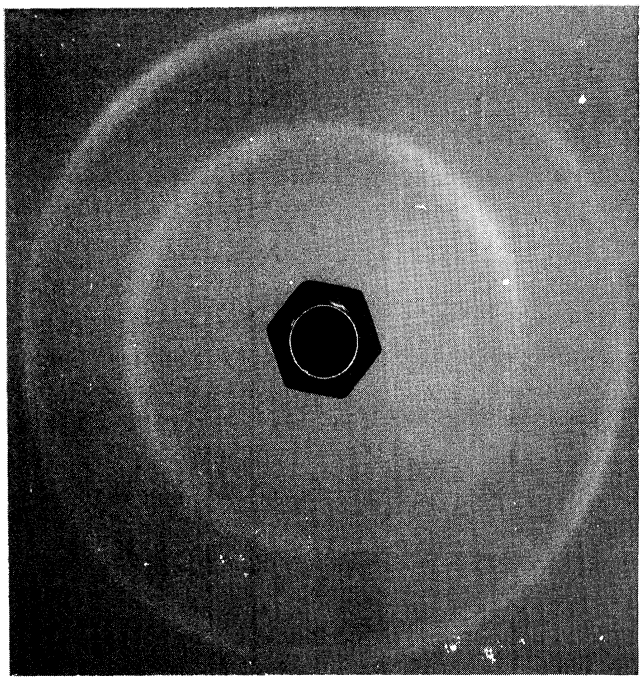
(c) 9.9 Percent Reduction at 1600°F



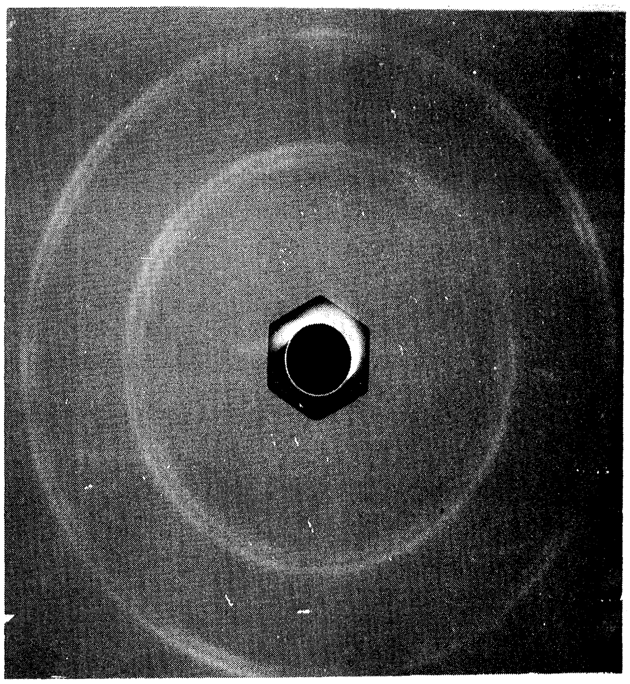
(d) 34.4 Percent Reduction at 1600°F

Figure 2. - Back-reflection X-ray Photograms showing the effect of percent reduction of area by rolling on the diffraction spots from the (331) planes (outer pair of circles) and the (420) planes (inner pair of circles) of "A" Nickel rolled at 1600°F. Radiation--CuK_{α}, specimen-to-film distance--6cm., diameter of area on specimen irradiated--0.70mm.

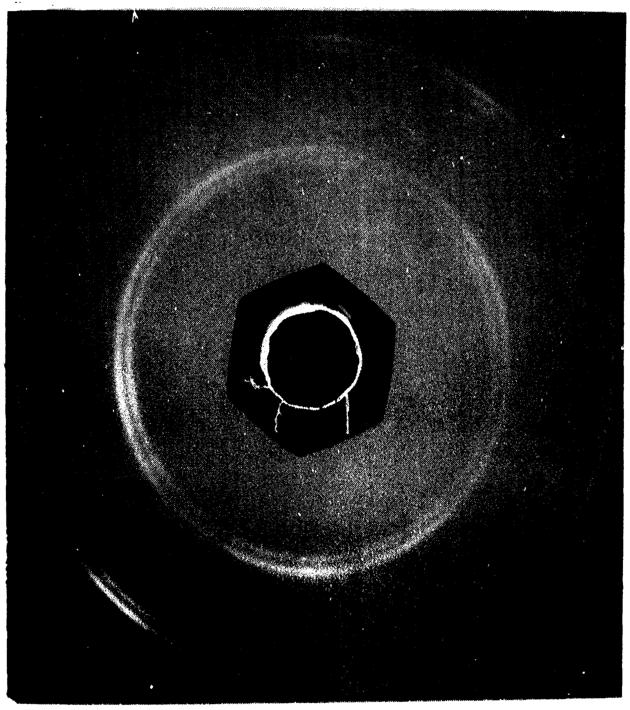
Figure 3. - X-ray evidence of substructure produced in "A" Nickel by rolling 5.8 percent reduction of area at 1800°F. The photographs above are enlarged (2.6X) portions of back-reflection photograms similar to the one shown in Figure 2 (b). Each arc or pair of arcs corresponds to one grain, and the small spots within each arc correspond to subgrains of slightly differing orientations. In the photographs above, small spots 1 mm apart would correspond to adjacent subgrains having an angle of tilt of 20 minutes between them.



(a) 35.5 Percent Reduction at Room Temperature



(b) 36.1 Percent Reduction at 1400°F



(c) 34.4 Percent Reduction at 1600°F

Figure 4. - Back-reflection X-ray Photograms showing the effect of temperature of reduction on the diffraction rings from the (331) planes (outer pair of circles) and the (420) planes (inner pair of circles) of "A" Nickel rolled 35 percent reduction of area. Radiation--CuK_{α}, specimen-to-film distance--6 cm., Diameter of area on specimen irradiated--0.70mm.

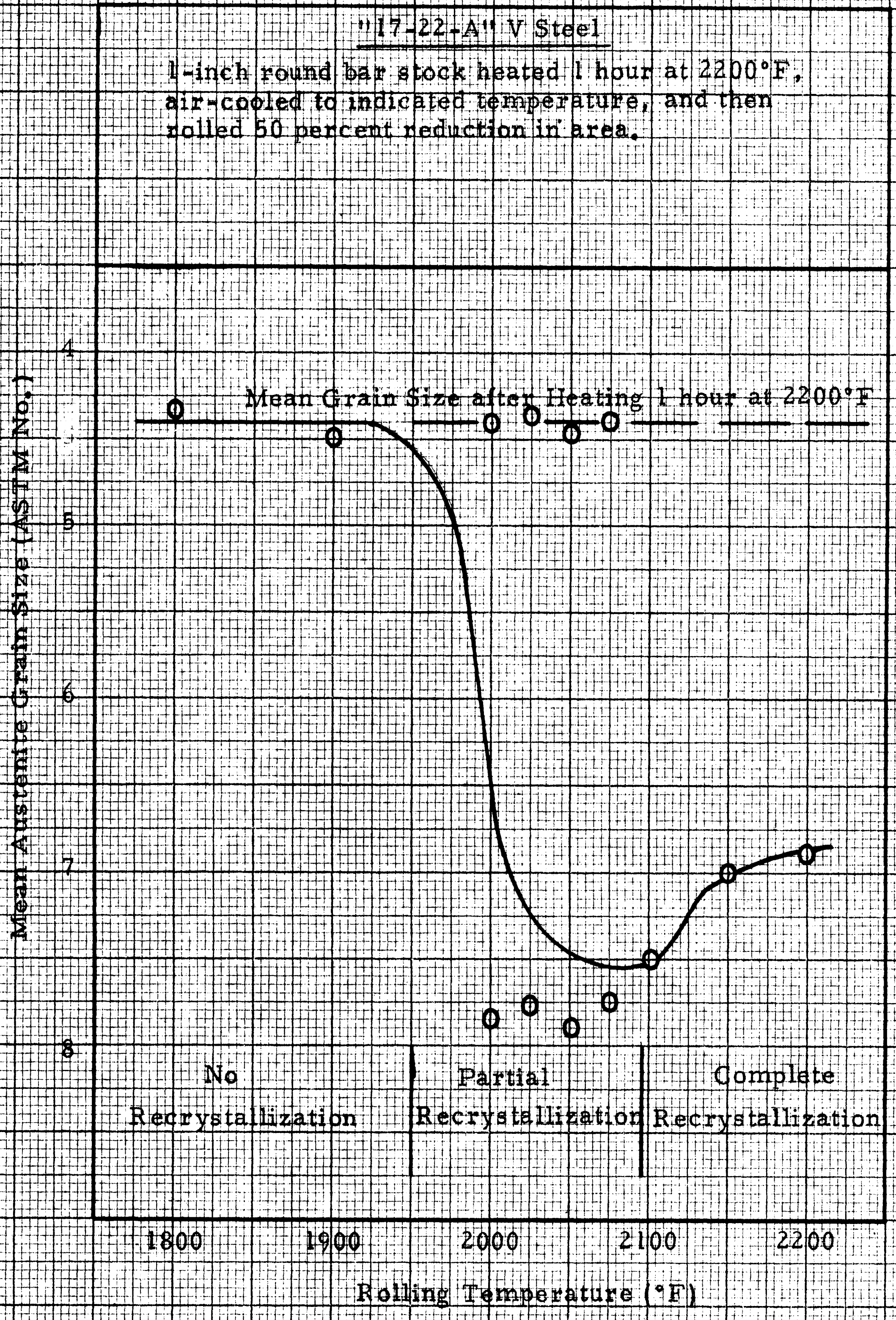
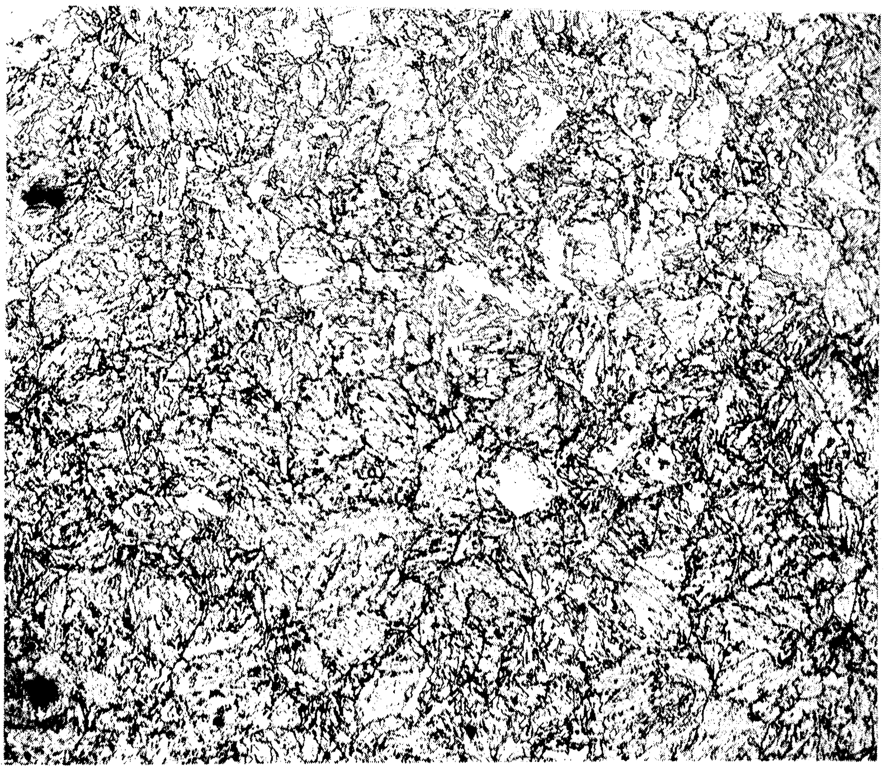
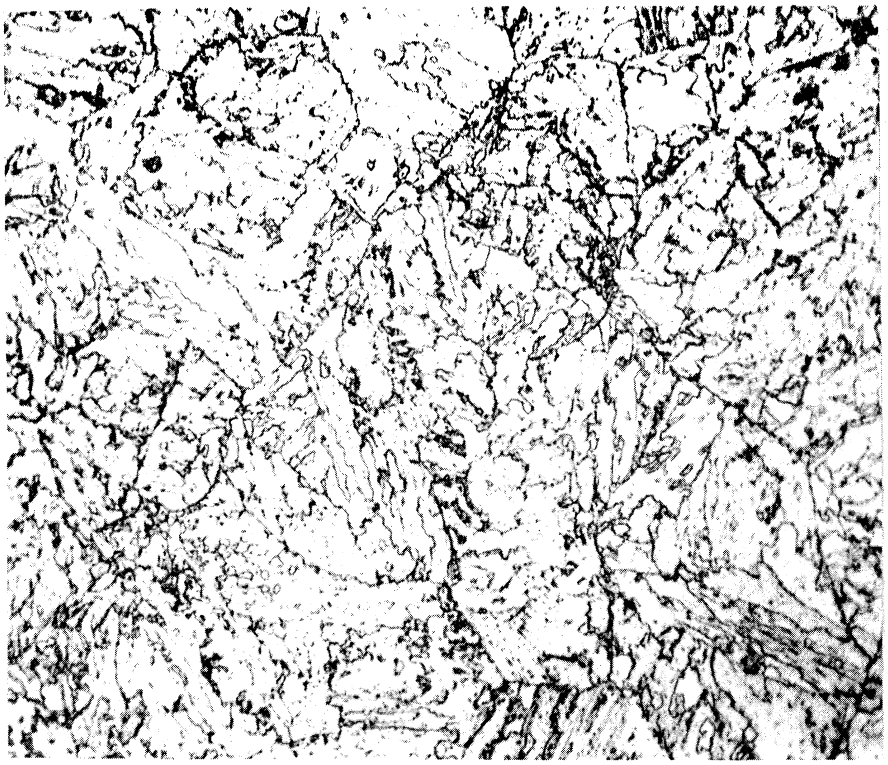


Figure 5. - Determination of the Minimum Temperature for Simultaneous Recrystallization of Austenite in "17-22-A" V Steel after Heating 1 Hour at 2200°F. The Reduction of Area by Rolling was 50 percent.

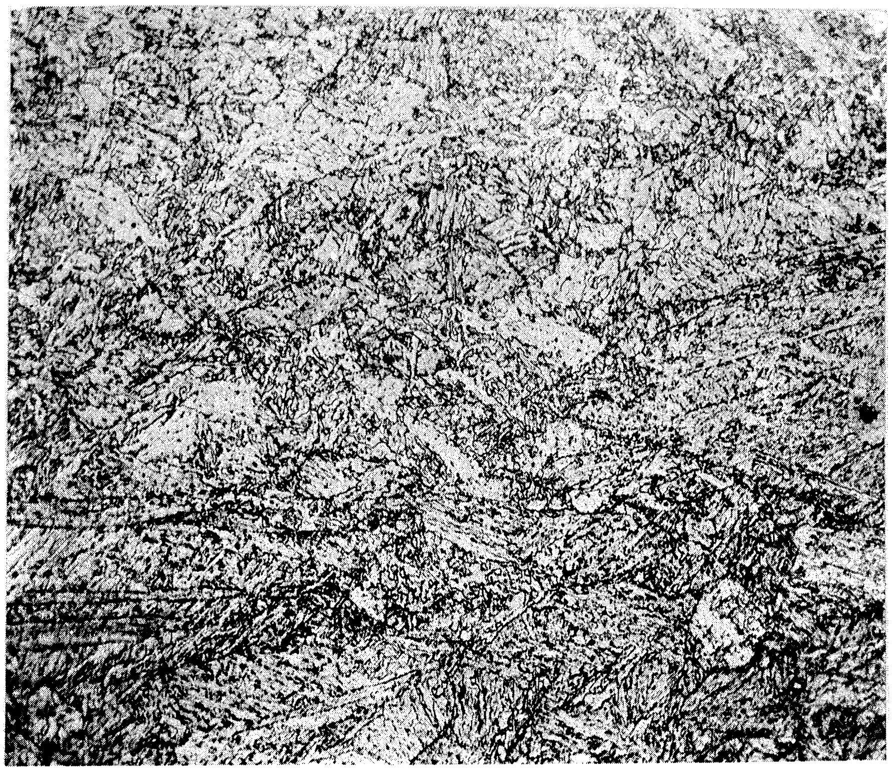


X250D

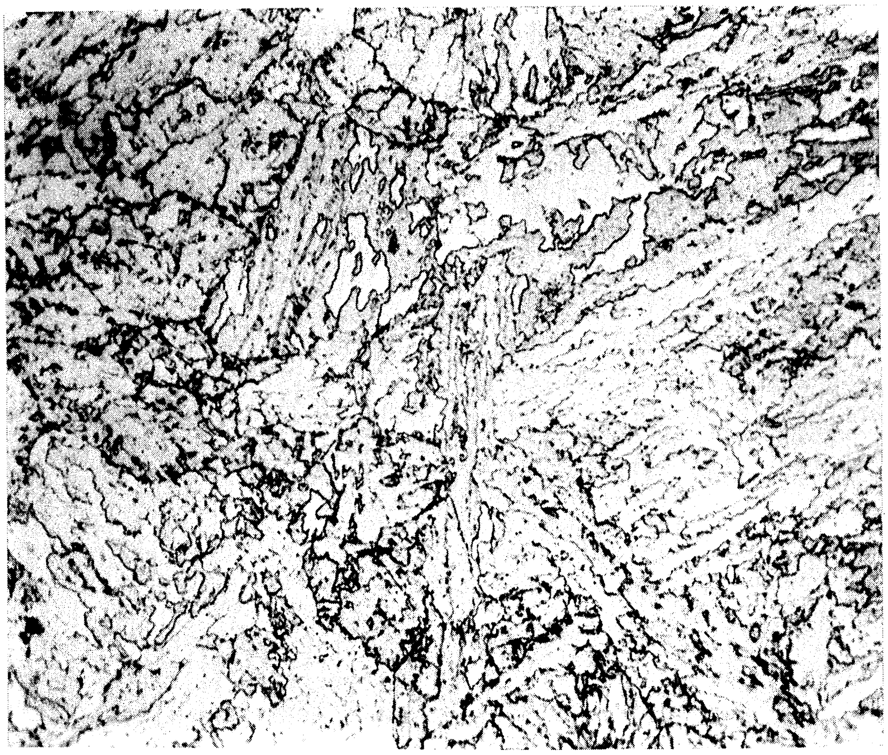


X1000D

Figure 6.- "17-22-A" V steel heated 1 hour at 2200°F., air cooled to 2100°F., rolled 50 percent reduction in 2 passes, and air cooled to room temperature + tempered 6 hours at 1250°F. Complete recrystallization during rolling.

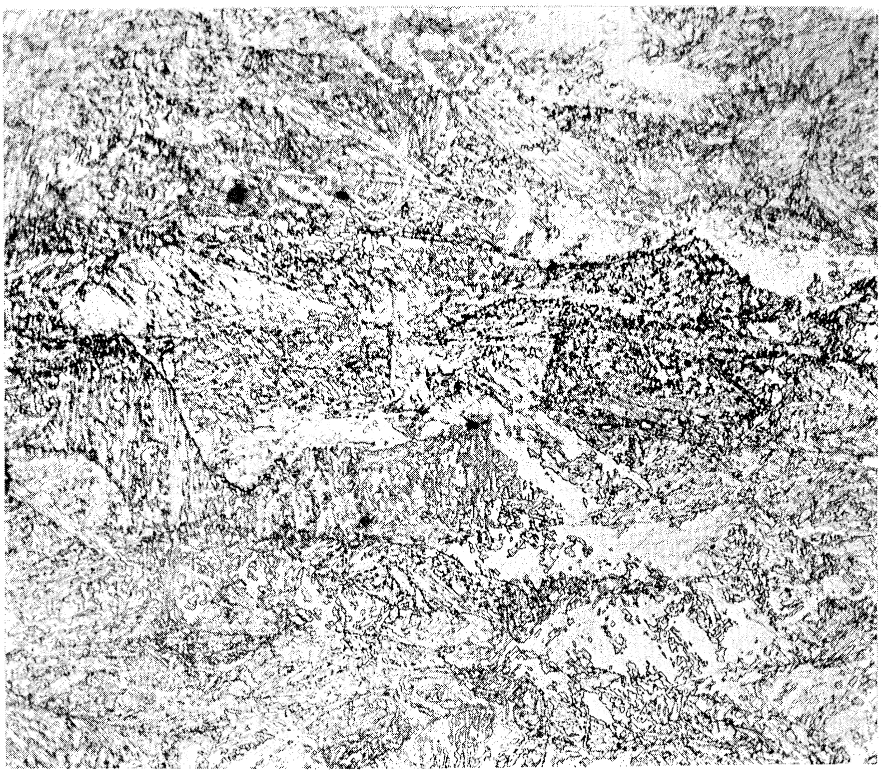


X250D



X1000D

Figure 7. - "17-22-A" V steel heated 1 hour at 2200°F., air cooled to 2000°F., rolled 50 percent reduction in 2 passes, and air cooled to room temperature + tempered 6 hours at 1250°F. Partial recrystallization during rolling.



X250D

X1000D

Figure 8. - "17-22-A" V steel heated 1 hour at 2200°F., air cooled to 1900°F., rolled 50 percent reduction in 2 passes, and air cooled to room temperature + tempered 6 hours at 1250°F. No recrystallization during rolling.

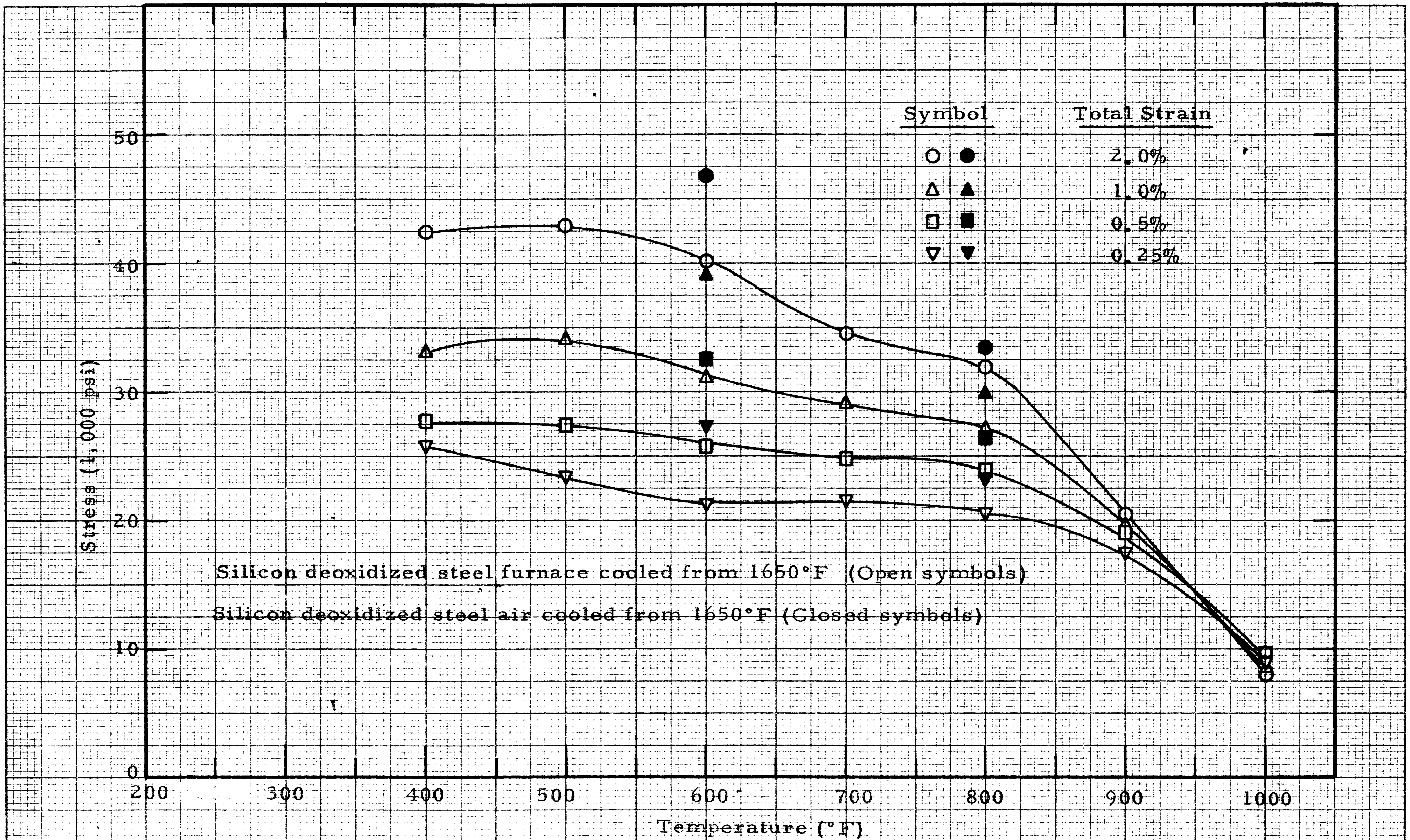


Figure 9. - Stress-Temperature Curves for Steel C for a Strain Rate of 0.004 Percent per Hour at Total Deformations of 0.25, 0.5, 1.0, and 2.0 percent.

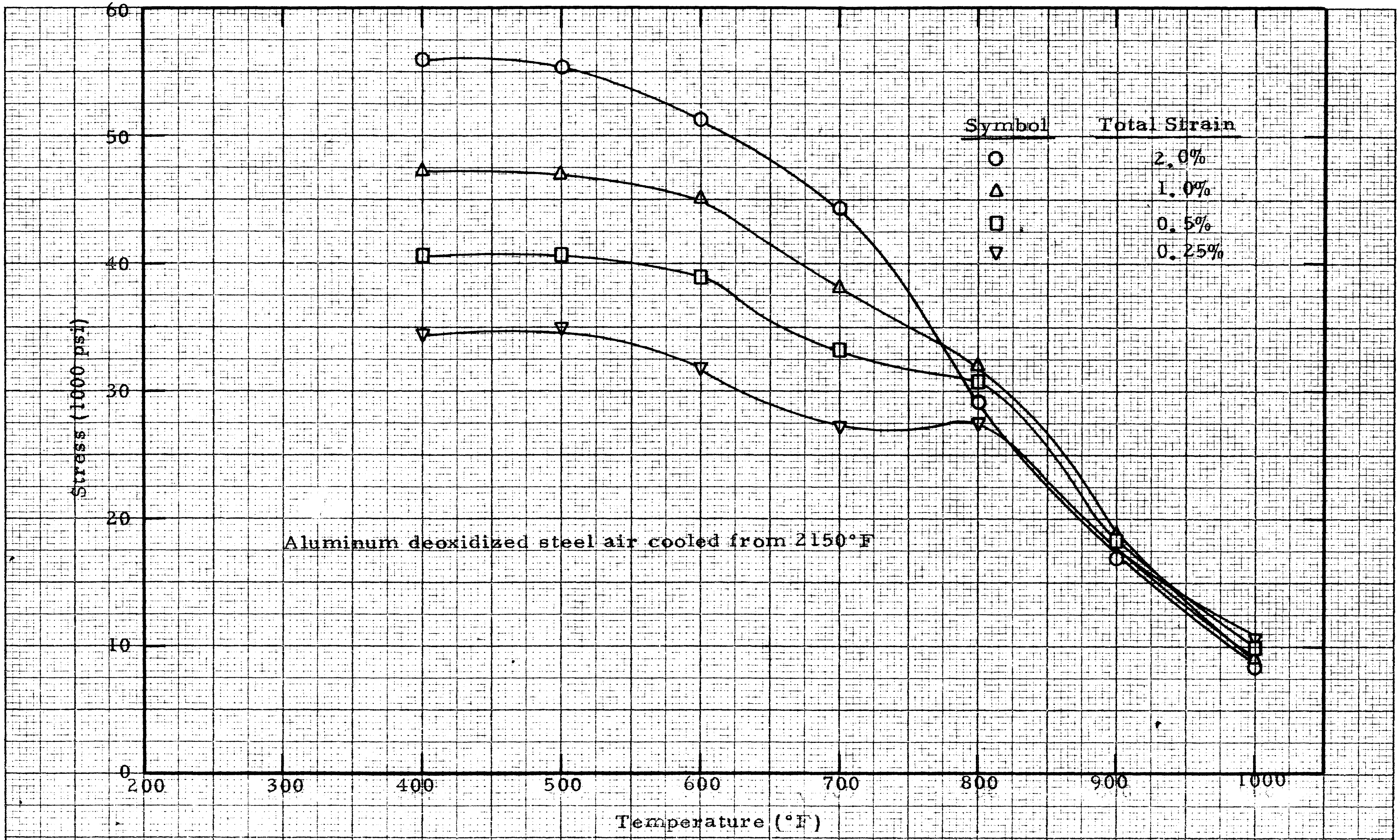


Figure 10. - Stress-Temperature Curves for Steel FA at total deformation of 0.25, 0.5, 1.0, and 2.0% for a Strain Rate of 0.004 Percent per Hour.

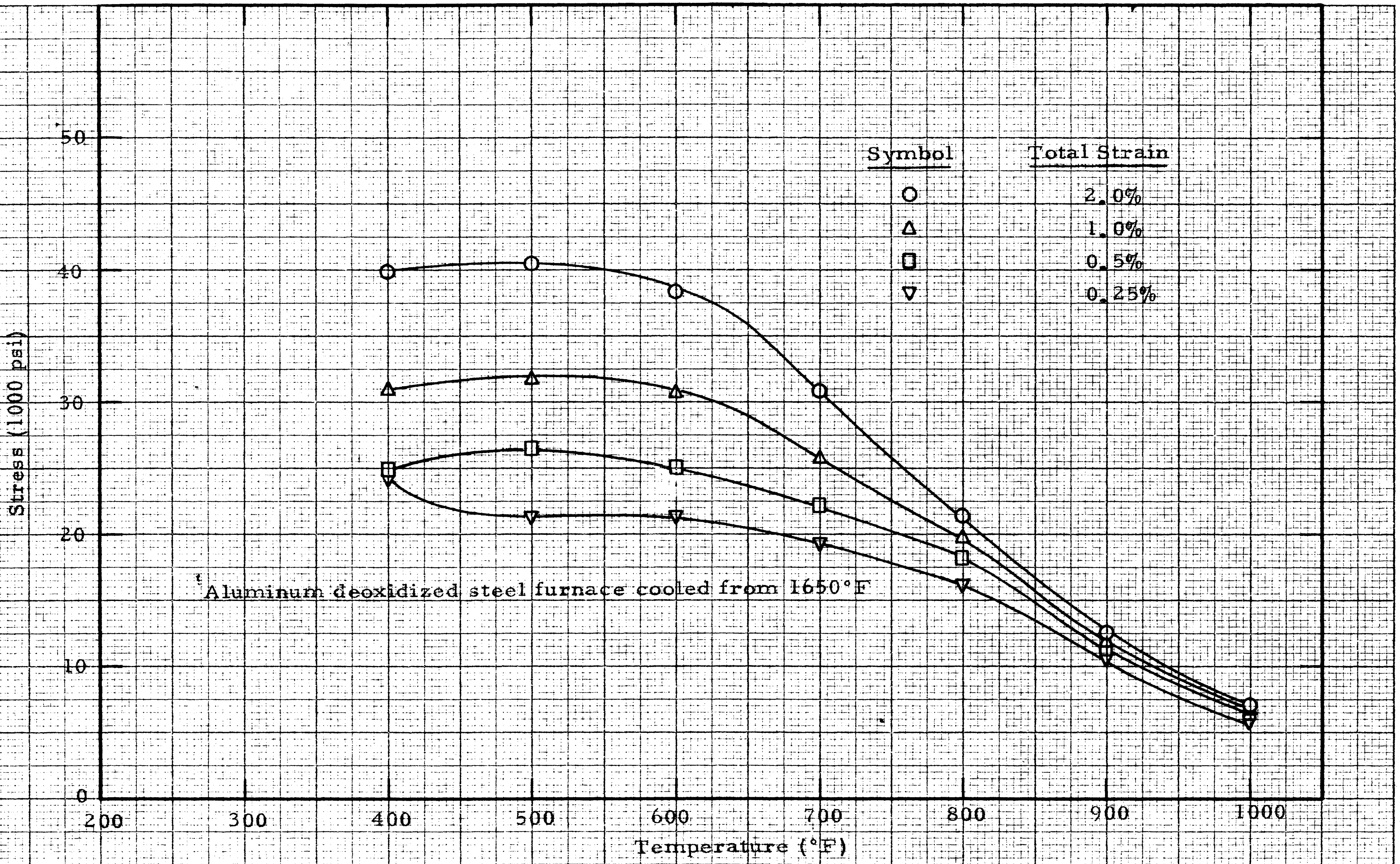


Figure 11. - Stress-Temperature Curves for Steel FN at Total Deformation of 0.25, 0.5, 1.0, and 2.0 Percent for a Strain Rate of 0.004 Percent per Hour.

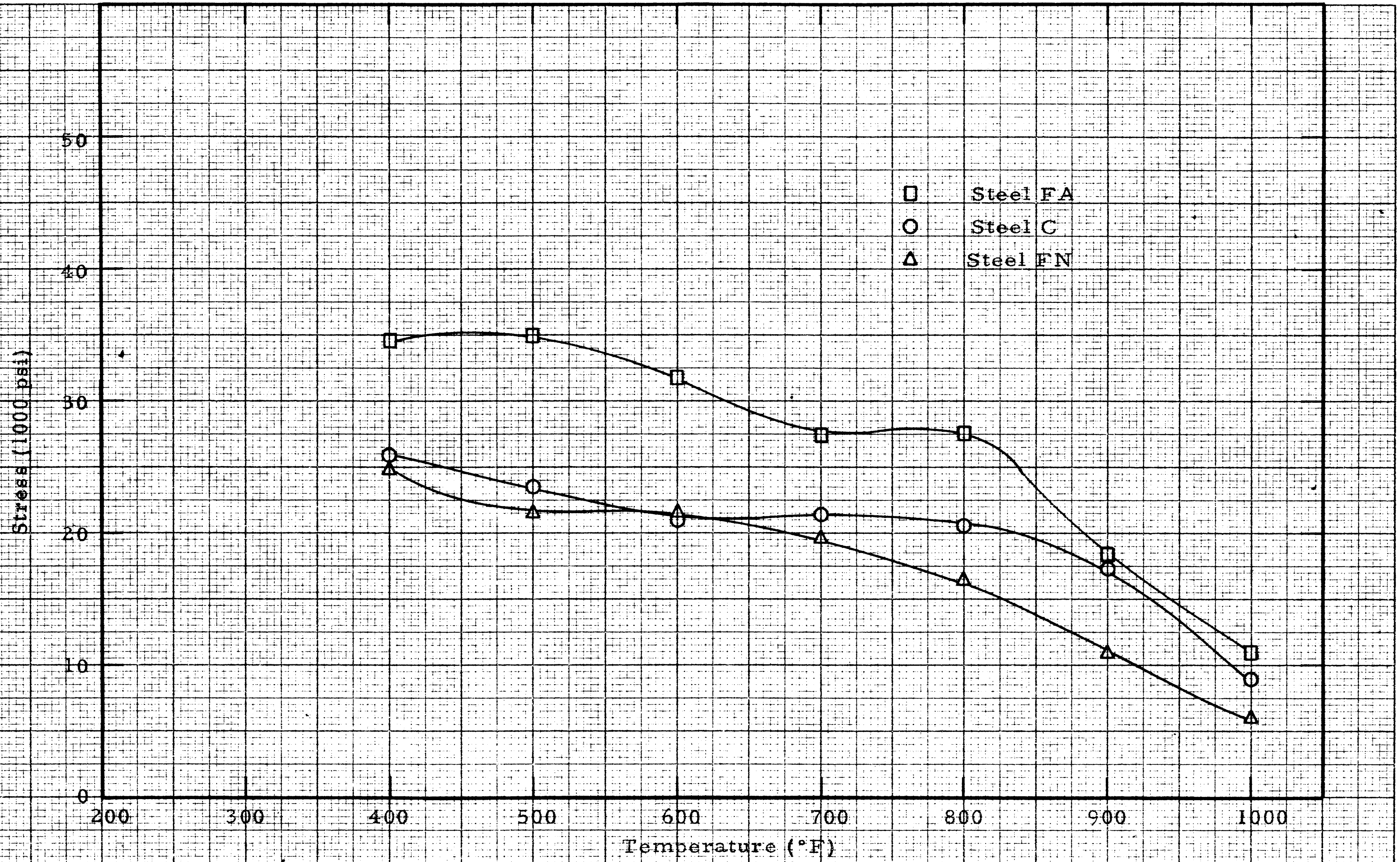


Figure 12. - Stress-Temperature Curves of Steel C, FA, and FN for 0.25 Percent Total Strain at a Constant Strain Rate of 0.004 Percent per Hour.

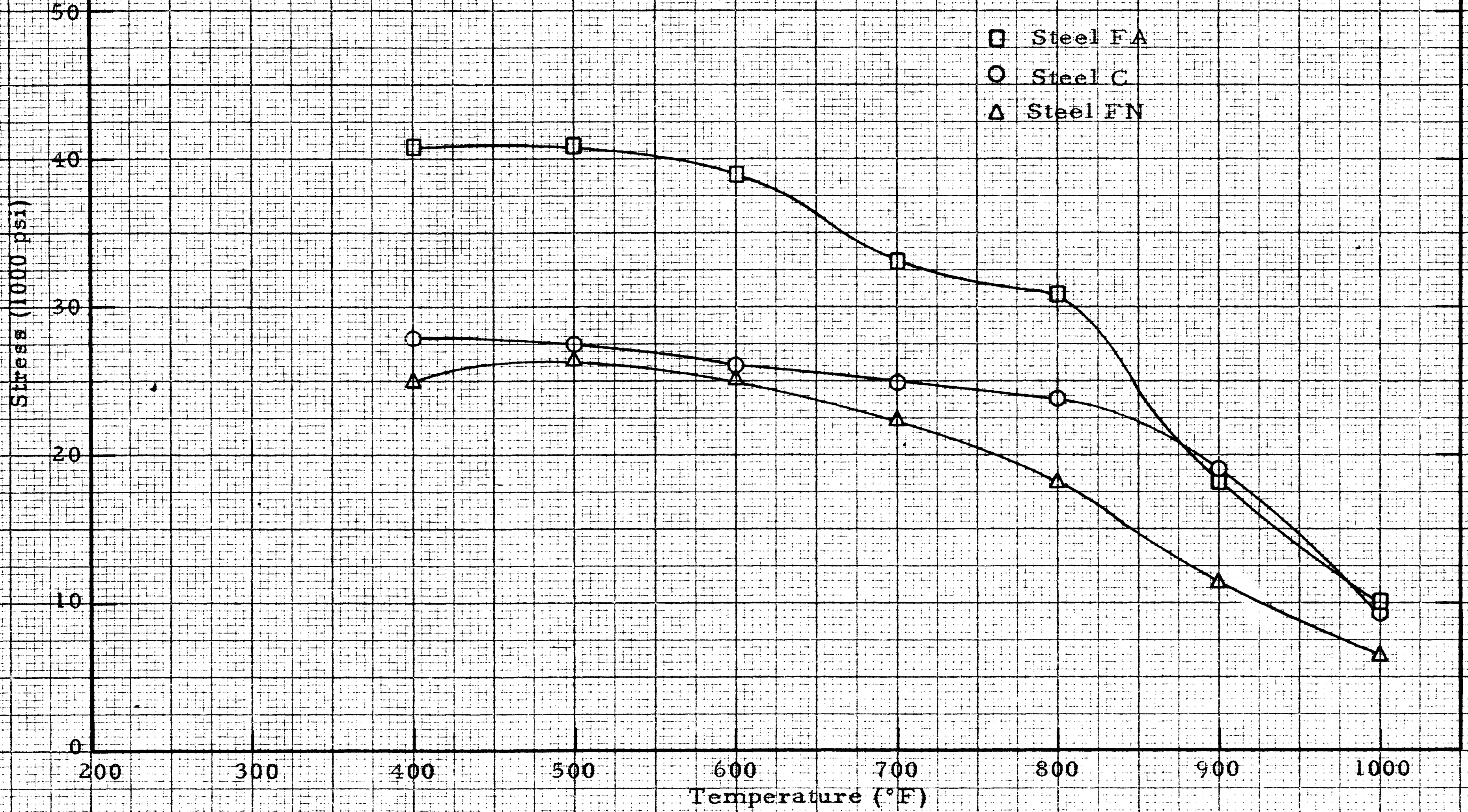


Figure 13. - Stress Temperature Curves of Steels C, FA and FN for 0.50 Percent Total Strain at a Constant Strain Rate of 0.004 Percent per Hour.

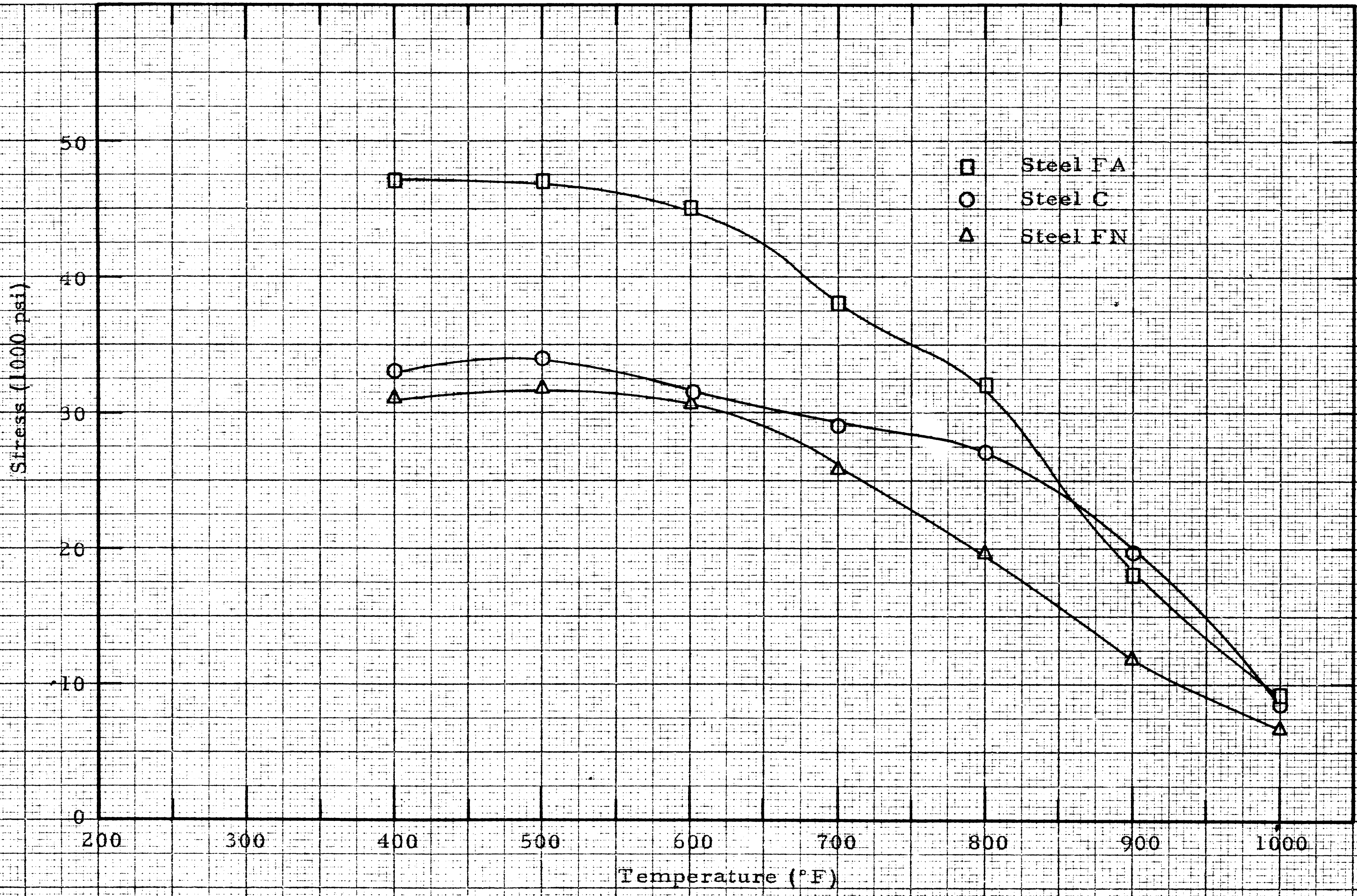


Figure 14. - Stress-Temperature Curves of Steels C, FA, and FN for 1.0 Percent Total Strain at a Constant Strain Rate of 0.004 Percent per Hour.

Figure 15. - Stress-Temperature Curves of Steel C, FA, and FN for 2.0 Percent Total Strain at a Constant Strain Rate of 0.004 Percent per Hour.

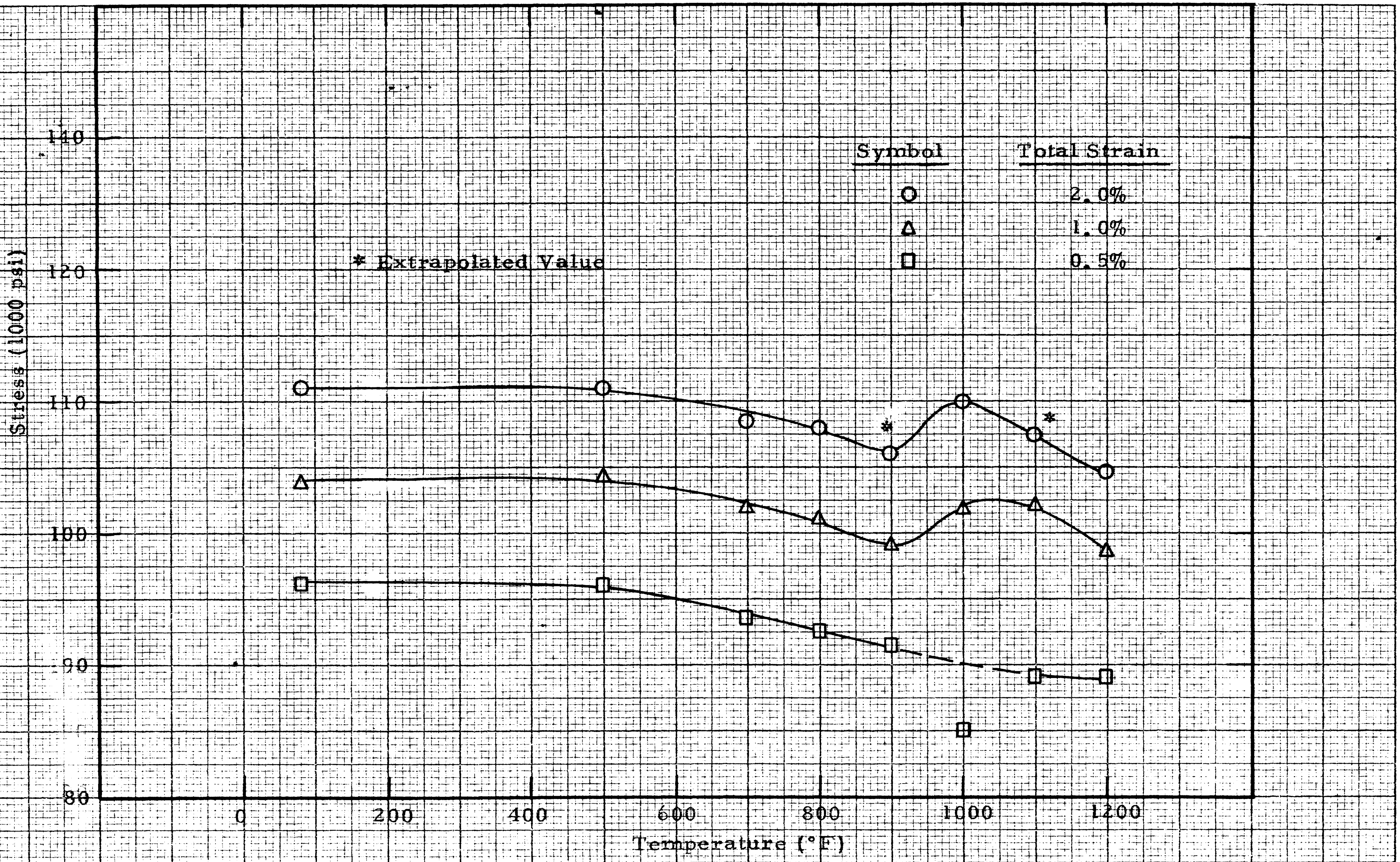


Figure 16. - Stress-Temperature Curves of A-286 Alloy for 2.0, 1.0, and 0.5 Percent Total Strain on Tensile Testing.

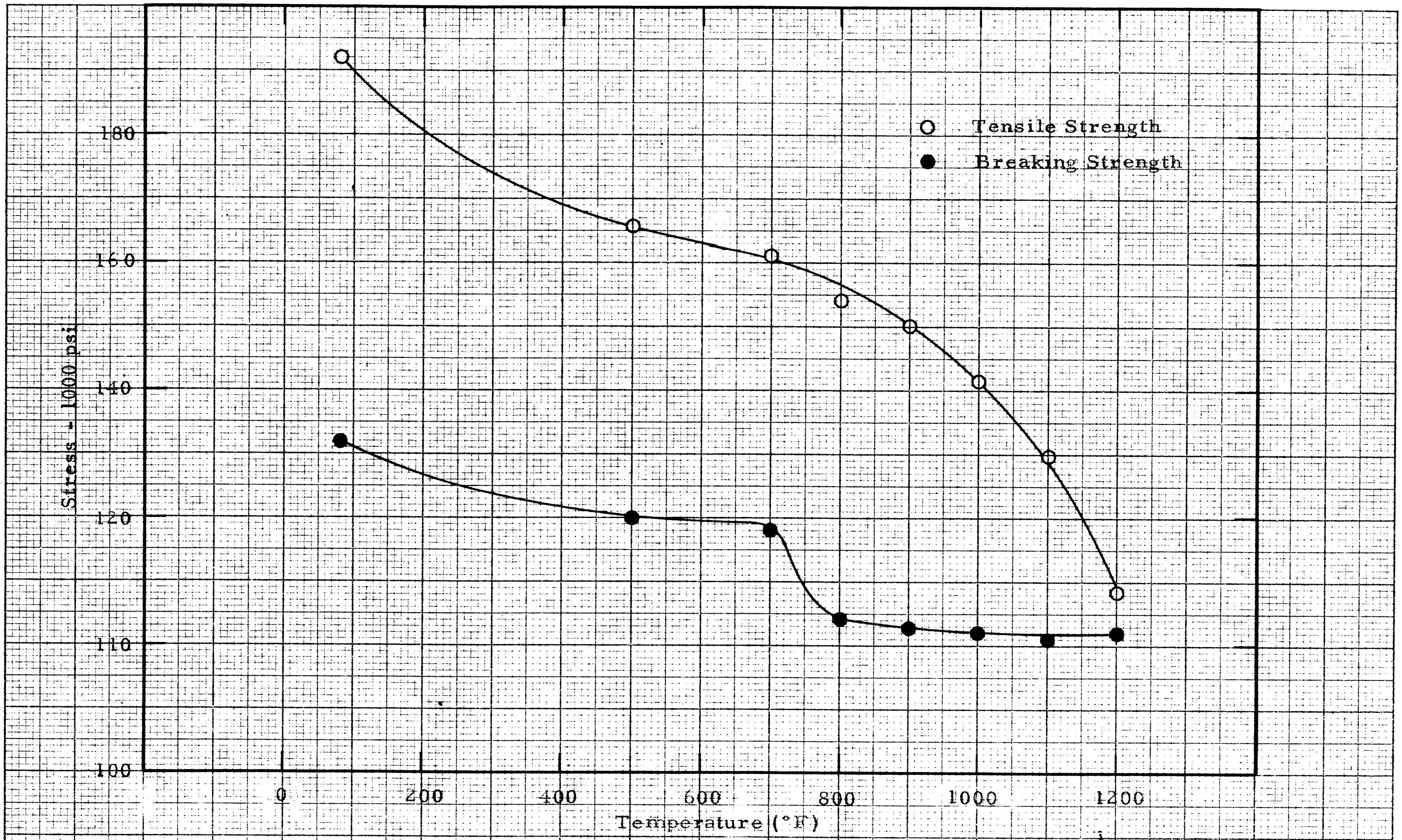


Figure 17. - Tensile and Breaking Strength Versus Temperature of A-286 Alloy on Tensile Testing.

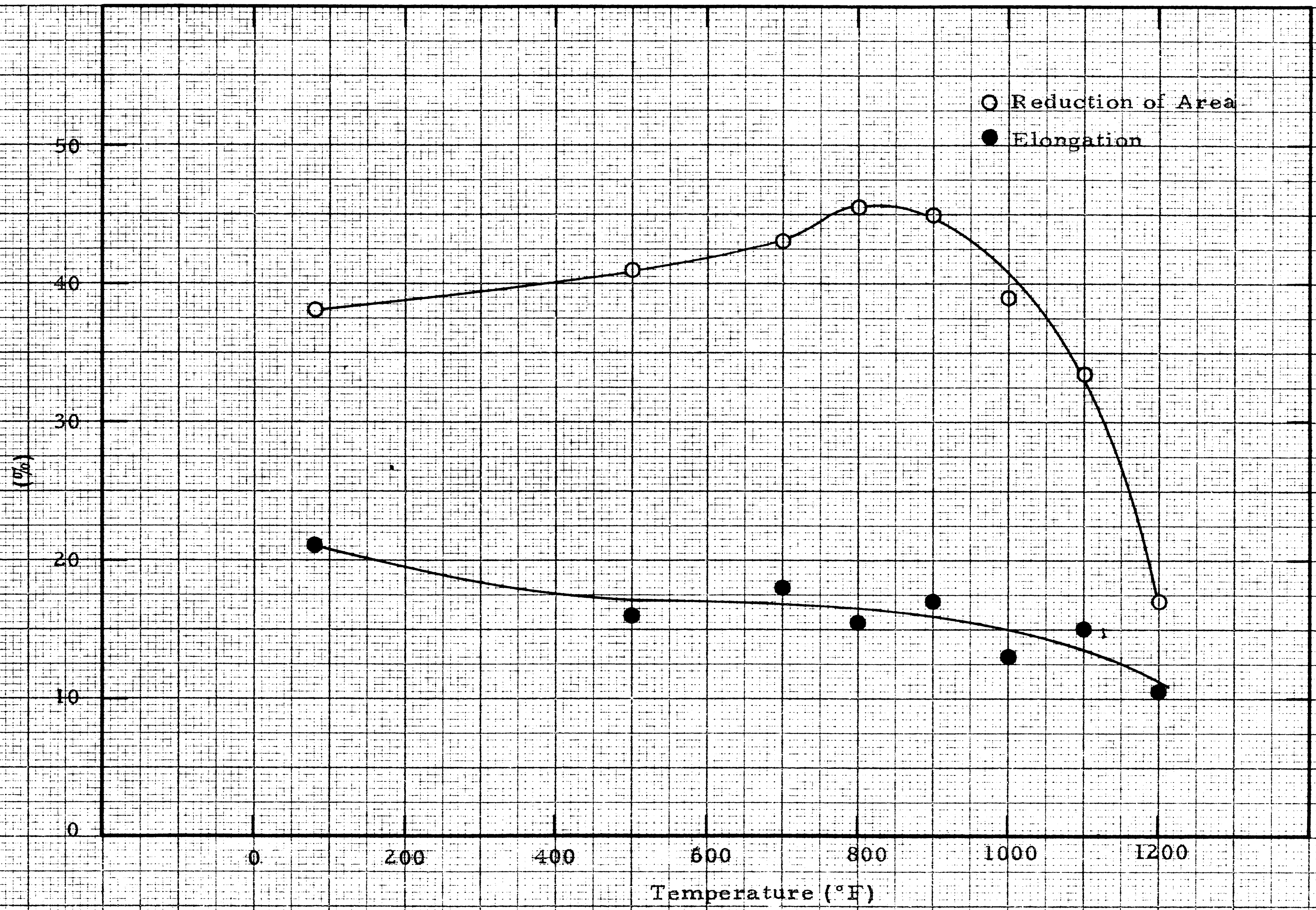


Figure 18. - Percent Elongation and Reduction of Area versus Temperature for A-286 Alloy on Tensile Testing.

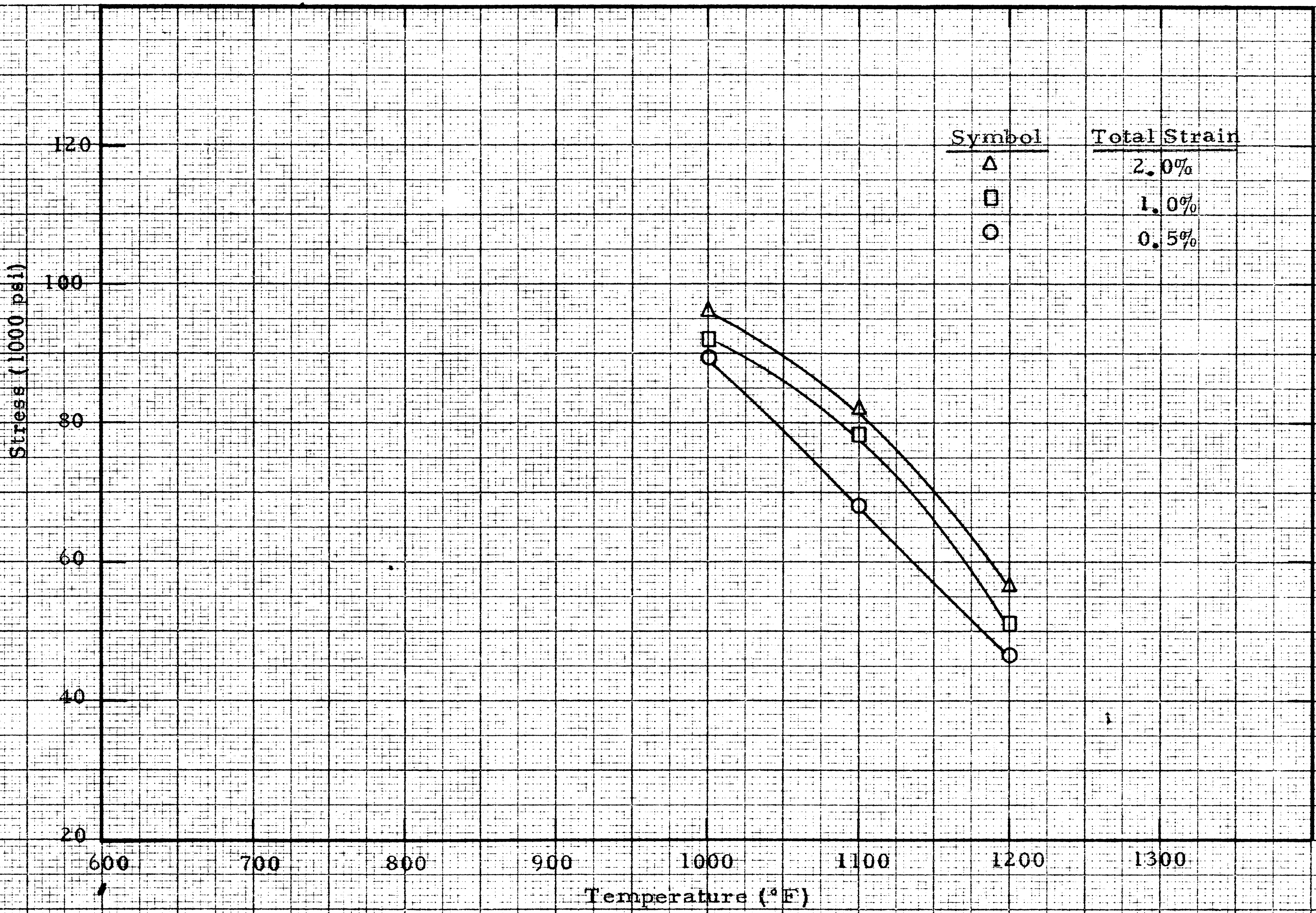


Figure 19. - Stress-Temperature Curves for A-286 Alloy at Total Strain of 0.5, 1.0, and 2.0% for a Strain Rate of 0.004 Percent per Hour.