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# FIRST 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

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## SUMMARY

Progress is reported for research carried out under Air Force Contract No. AF 33(616)-5466 covering the work period March 15, 1958 to June 30, 1958.

The present research is a continuation of work started earlier concerning basic factors affecting the creep-rupture properties of heat-resistant alloys. The two current phases of the work are studies of (a) the influence of hot working on structure and creep-rupture properties, and (b) the relationship between strain aging phenomena and high-temperature strength.

Under the previous contract, creep-rupture properties of A-286 alloy and "A" Nickel were surveyed after the material had been rolled controlled amounts at various temperatures. The present work, on these alloys is primarily concerned with correlating the creep-rupture properties with structural changes introduced by the working. The results of three preliminary hot-rolling experiments are presented for "17-22-A"V steel.

Constant strain rate tension tests are being used to investigate the influence of strain aging type reactions on the high-temperature strength of alloys. As a means of learning how the strain aging reactions influence properties, the reactions of 1020 carbon steel are being studied. The relative creep properties of 1020 steel show a marked superiority for conditions subject to strain aging. Because the compositional and structural relations accompanying this difference are better known than for any other alloy, it provides an excellent case to develop an understanding of the reaction. The role of strain aging in a heat-resistant alloy is being determined concurrently using a heat of A-286 alloy studied by Captain Domian at the Materials Laboratory, WADC for strain aging characteristics using hot-hardness tests. The general objective is to define and understand the role of strain-aging reactions in creep-rupture properties of alloys to the extent that it can be applied to obtain optimum strength in aircraft heat-resistant alloys.

# INTRODUCTION

This report, the first quarterly progress report issued under Air Force Contract No. AF 33(616)-5466, covers work done from March 15, 1958 to June 30, 1958.

The present investigation is a continuation of work initiated at the University of Michigan for the Wright Air Development Center concerning the relationship between microstructure and creep-rupture properties of heat-resistant alloys. In the early work, creep-rupture properties of several ferritic alloys were surveyed after their microstructures were systematically varied by heat treatment. Last year the study was extended to include a precipitation-strengthened, austenitic alloy (A-286) and a commercially pure metal ("A" Nickel). The structures of these alloys were varied by systematic hot working instead of heat treatment. The creep-rupture properties of the nickel were surveyed for the as-rolled conditions, and the A-286 properties were surveyed after a standard solution and aging heat treatment was given to the rolled bars.

The current emphasis of the hot-working study is to correlate the creep-rupture properties with the structural changes induced by the hot-working operations. When the basic relationships between working conditions and structure and between structure and properties are established, it will be possible to use controlled hot working to produce wrought products with reproducible, high strengths.

In addition to the hot-working experiments, considerable time was spent last year on the design and construction of a high-temperature, constant-strain-rate, tension testing machine. It is planned to use the machine this year in a study of the relationship between strain aging phenomena and high-temperature strength. The results of some preliminary constant strain rate tests are presented in this report.

To facilitate reporting the progress of the hot working and strain aging studies, the present and future progress reports will be divided into two parts:

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

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

## 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 basic survey of creep-rupture properties over a range of rolling temperatures and reductions was completed last year for "A" Nickel and A-286 alloy (1)*. For these two materials the objective of this year's work will be to correlate the properties with the structure as observed through ordinary-light and electron microscopy and with x-ray analysis. The hot rolling of "17-22-A"V steel planned for last year was delayed because the results of preliminary experiments were not conclusive. Current data and future plans are discussed below for each material.

## Commercially Pure Metal ("A" Nickel)

Considerable emphasis is being placed on the study of "A" Nickel. The main reason for this is that, having a simple microstructure, the structural variables involved in hot-working should be amenable to identification. At present it is considered that the variables which most likely control the creep-rupture strength are limited to strain hardening, substructures, recovery, recrystallization, and grain size.

### Rolling and Creep-Rupture Testing

It was not possible to complete the processing and creep-rupture testing of samples for two conditions last year. The conditions were heating at 1400°F without reduction and a small (about 5 percent) reduction at 80°F. Samples for the former will be prepared and tested in the near future; samples for the small cold reduction have been prepared and tested at 800°F and 18,000 psi and at 1100°F

*Numbers in parentheses pertain to references listed at the end of the report.

and 20,000 psi. The rolling, hardness, and creep-rupture data obtained were as follows:

Reduction of Area by Rolling at 80°F ------------ 4.5 percent

As-Rolled Hardness ---------------------------- 127 DPH

Rupture Data at 1100°F and 20,000 psi

Rupture Life ----------------------------- 39.2 hours

Elongation ------------------------------- 46.3 percent

Reduction of Area ------------------------ 53.8 percent

Stress to Produce Minimum Creep

Rate of 0.0001 percent per hour at 800°F -------- 19,000 psi

The creep strength at 800°F for nickel cold reduced 35.5 percent was missing from the data in Reference 1 because a duplicate test was necessary when the report was prepared. The test result is now available and is presented in Table I. Other minor changes are also presented and explained in Table I and its footnotes. The general conclusions originally drawn from these data in Reference 1 are not altered by the additional data.

Figures 1 through 4 have been redrawn from Reference 1 (formerly Figures 6, 8, 9, and 12, respectively) because they experienced the greatest improvement from the additional data. Figures 1 and 4 show how the addition of the two points at 4.5 and 35.5 percent reduction and at 127 and 218 DPH helps to establish the shapes of the curves for the cold-rolled series. In Figure 2 the addition of the point at 127 DPH and 39.2 hours establishes the straight-line correlation for the cold worked samples in the range 127 to 198 DPH. In Figure 12 of Reference 1 the three close points (171 to 198 DPH) merely suggested that a straight-line correlation might exist. The complete elongation-time curve at 1100°F and 20,000 psi for the sample cold rolled 4.5 percent is presented in Figure 3, along with the two curves originally drawn in Figure 9 of Reference 1. The three curves are plotted together to illustrate how, for a given as-rolled hardness, the temperature of working markedly affects the creep-rupture strength of "A" Nickel.

Recovery Experiments

The influence of recovery time on hardness at 800° and 1100°F was determined for five widely different conditions of rolling. The main purpose was to gain some idea as to the relative stability of the various, cold- and hot-worked structures at the temperatures used for creep-rupture testing. Another objective was to determine whether any major hardness changes occurred within the usual time of preheating for a creep-rupture test. Samples in the base condition (annealed from 1600°F) were also run to check on the structural stability of that condition. All specimens were slugs cut from the same rolled bars from which the creep-rupture samples were machined. They were heated in air without stress for times up to 100 hours. The results are presented in Figure 5. Hardness values after testing obtained from the low-stress threaded region of completed creep and rupture samples are also plotted in Figure 5 for comparison. These points approximate unstressed recovery treatments of about 1000 hours.

The scatter of the hardness data is sufficient, as can be seen from the range of closed circles for the as-rolled or "zero time" samples, that care must be exercised in drawing conclusions from these limited data. The apparent trends are as follows:

1. Hot-worked nickel is more stable than when cold worked.
2. Hot-worked nickel experiences very little softening at 800°F up to 1000 hours, contrary to the cold-worked nickel.
3. The softening of hot-worked and cold-worked nickel at 1100°F occurs rapidly during the first hour followed by a more gradual softening from 1 hour to 1000 hours.
4. No change in hardness could be found with the given hardness test for annealed nickel heated at 800° or 1100°F for times up to 2300 hours.

One conclusion to be drawn from these results is that an attempt to correlate creep-rupture properties with initial structure must allow for structural changes occurring during the preheat period of the creep-rupture test before the stress is applied.

Structural Studies

The microscopic identification of substructures in "A" Nickel in this investigation to date has been successful in only a few hot rolled samples, namely those reduced small amounts at high temperatures. At lower rolling temperatures and/or higher reductions it has not yet been possible to reveal the substructures by etching. Considerable time has been spent, therefore, in searching the literature for x-ray methods of measuring substructures in polycrystalline samples. Perhaps the most exhaustive and precise method is the one devised by Weissman (2) which uses the high resolving power of the double crystal diffractometer principle. This procedure consists of (a) analyzing x-ray reflections from individual grains by using a grazing angle of incidence and varying the specimen-film distance from about 1 mm (a Berg-Barrett x-ray micrograph) to the usual distances used for ordinary Debye-Sherrer x-ray photograms, and then (b) determining the misorientation parameters by appropriate rocking curves. A device recently developed for use with this procedure permits the exact specimen area being examined by x-rays to be examined also by an ordinary-light microscope.

Another x-ray method which is somewhat simpler and yields proportionately less information is the "Spotty Ring Back-Reflection Technique" used by Gay and Kelly (3) in studying the structure of various cold rolled metals including nickel. The principle employed by this method is that when a sufficiently small x-ray beam of characteristic radiation is used in taking a "stationary film-stationary specimen" back reflection photogram, the number of diffracting crystallites will be small enough that spotty rings will be obtained. From the number of spots on a ring the mean diameter of the diffracting crystallites (subgrains in a worked material) can be calculated. When the individual spots are spread tangentially into arcs, as well as being broadened radially, estimates of misorientations and internal strain can be obtained.

This laboratory does not presently have the equipment needed for Weissmann's technique. Therefore, a series of preliminary x-ray photograms of representative rolled nickel samples were taken using the Spotty Ring Back Reflection Technique used by Gay and Kelly. The smallest collimator (about 0.5 mm diameter) presently available in this laboratory was not quite small enough, but the following observations could be made.

1. In the samples reduced 5 percent at 1800°F and 5 percent at 1600°F several of the diffraction arcs were resolved into a series of spots connected by background radiation. According to Gay and Kelly (3) these spots correspond to fragments or crystallites (subgrains) into which the original grain was broken during working. The background radiation between the spots corresponds to distorted lattice regions between the crystallites, i.e., subgrain boundaries which are not perfectly sharp.
2. In the samples reduced 35 percent at 80°, 1400°, 1600°, and 1800°F it was obvious from the relative breadths of the diffraction rings that considerably less internal lattice distortion remained in the hot-worked samples than in the cold-worked sample.
3. In addition, extensive rolling at room temperature or at elevated temperatures without recrystallization spreads an initial pattern of a few diffraction spots into a pattern of continuous diffraction rings.

4. Diffraction spots from strain-free grains are easily distinguished from spots arising from grains that have experienced plastic deformation. This situation arises when partial recrystallization occurs during the last pass. Diffraction spots from deformed grains are spread more or less into arcs; whereas, spots from strain-free grains are sharp and black, with practically no tangential spreading into arcs.

It is planned to make a collimator which will reduce the x-ray beam size to about 0.1 mm and repeat the series of back reflection exposures described above. It is hoped that the results will be good enough to permit at least a good qualitative -- if not semi-quantitative -- evaluation of substructures in the rolled nickel samples.

For good quantitative results for the higher deformations, the resolution obtainable only with x-ray beam diameters of the order of 0.01-0.04 mm may be needed (2,3). Such a small beam can be obtained by collimating a beam from an ordinary x-ray generator, but the intensity of the collimated beam would be so low that unreasonably long exposure times would have to be used. This problem can be solved only by using a special, high-intensity x-ray generator. This laboratory is not equipped with such a generator at present.

In the meantime, a search is still being made to find a way of revealing the substructure by metallographic techniques. As mentioned earlier, partial success has been attained by etching electrolytically in aqueous $H_3PO_4$ solution. Only a very few samples respond to this technique.

## Precipitation-Strengthened, Austenitic Alloy (A-286)

The basic survey of creep-rupture properties of heat treated A-286 alloy was carried out under the previous contract (Ref. 1). The survey covered rolling temperatures of 80°, 1700°, 1950°, and 2200°F, and reductions up to 40 percent.

### Structural Studies

Grain size measurements were made on samples from all the rolled and heat treated bars of A-286 alloy prepared last year. The measurements were made by counting grains in a 4 in. x 5 in. area marked on the "frosted" glass of a metallograph. The data were converted to A.S.T.M. grain size numbers.

To check for a possible correlation between grain size and high-temperature strength a plot was made of log-rupture life at 1350°F and 40,000 psi versus grain size (Fig. 6). The 1350°F data were used for the test because if grain size effects are present they would most likely be found at high test temperatures. The plot (Fig. 6) shows that extreme differences in grain size, at least in the range of 0 to 8, can exist without much change in strength at 1350°F and 40,000 psi.

It would seem that structural variables in A-286 most likely to affect the creep-rupture strength are solution effects and the types of precipitate particles. As yet, electron microscope techniques for showing fine structural details after the standard heat treatment have not been established. Work along this line will proceed simultaneously with work on the other materials.

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

### Preliminary Hot Rolling

The study of "17-22-A"V steel was delayed last year because of indefinite results of preliminary experiments. Originally it was reasoned that the most likely prospect for improving the creep-rupture strength of hardenable, ferritic materials such as "17-22-A"V steel would be to attain a high degree of solution of excess phases within a fine-grained structure. Ordinarily, conditions producing good solutioning result in coarse-grained structures.

Three preliminary experiments were carried out. The first was the determination of the austenite grain size as a function temperature. The results are shown in Figure 7 as the range of grain size (shaded area) resulting from heating the as-received stock to various temperatures from 1 hour and air cooling. The divided area between 1930° and 1980° was the result of abnormal grain growth in isolated regions.

The purpose of the second and third experiments was to determine the minimum temperature for simultaneous recrystallization of the austenite during rolling. The first approach was to heat for 1 hour at various temperatures and then reduce 25 percent by rolling. It was expected that above a certain temperature the grain size would be reduced below the size produced by simple heating (shaded area). The resulting average grain sizes, denoted by closed circles in Figure 7, indicate that no grain refinement was achieved up to 2200°F. Perhaps a 25 percent reduction is too small to cause appreciable simultaneous recrystallization. The results of the second attempt to determine the minimum temperature for recrystallization were also unsatisfactory. In this case the as-received stock was heated to 2200°F for 1 hour, air cooled to various temperatures, and then rolled 25 percent. The purpose of heating to 2200°F was to coarsen the grains so that the beginning of recrystallization could be detected more easily. The rolling temperatures were measured by varying the time of air cooling from 2200°F according to a master cooling curve established for this particular stock with a thermocouple at the center of the bar. The grain sizes produced by this treatment are indicated in Figure 7 by open circles. Only at 2100°F was some recrystallization evident, and that was only partial as indicated by the two mean grain sizes reported for that temperature. Again it seems that a 25-percent reduction is too small to achieve complete grain refinement at any temperature.

It is planned to repeat the experiment described above, where the bars are preheated to 2200°F and then cooled to the rolling temperature, except that a much larger reduction will be used in an effort to obtain complete grain refinement.

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

The objective of the constant-strain-rate investigation is to provide information relating strain aging characteristics to properties at high temperatures. For a material known to be subject to strain aging it has been observed that aging in the alloy is closely associated with high creep strengths at elevated temperatures. Unfortunately strain aging characteristics are usually identifiable only at temperatures lower than those at which creep-rupture properties are of interest. Susceptibility to strain aging is evidenced by the presence of yield points, serrated stress-strain curves, or most conveniently, by "inflections" in curves of hot hardness versus temperature or of stress for specific deformations versus temperature under conditions of constant strain. The differences between two conditions of an alloy, one subject to strain aging and the other free from strain aging disappear when the rapid strain rate testing conditions are raised to the temperature of the creep-rupture range. Yet if the same conditions which cause evidence of strain aging at lower temperatures are responsible for higher creep-rupture strengths, they must continue to be effective at the lower strain rates of creep-rupture tests, even though they are no longer evident in rapid strain rate tensile tests at elevated temperatures.

The first objective of the research is to experimentally determine the way the strength of materials susceptible to strain aging varies from the lower temperatures where creep does not occur to the creep-rupture range as a function of strain rate and amount of strain. This should then indicate the way the interstitial elements such as nitrogen and carbon which cause strain aging help to strengthen an alloy under creep-rupture conditions.

# PROCEDURE

The general procedure being used is to establish curves of stress versus temperature for deformations up to 2 percent as a function of strain rate. The range of strain rates being covered vary from those of tensile tests to slow creep tests. The material being used first is 1020 carbon steel. This material was selected because it is very susceptible to strain aging when deoxidized with silicon. When deoxidized with aluminum it is relatively free from strain aging (unless specially heat treated to retain nitrogen in solution). The silicon-deoxidized type steel will have nearly double the creep-rupture strength that aluminum deoxidized type steels exhibit in the temperature range of 850° to 1000°F. There is also considerable creep data and nitrogen analysis data available from previous work done on the 1020 steels being used.

A-286 alloy is concurrently being studied. The same material was investigated for strain-aging characteristics by Captain Henry Domian at the Materials Laboratory, WADC. He used hot hardness tests and a very rapid strain rate test in his investigation. At present, there is no definite information regarding variations in creep-rupture properties of A-286 alloy arising from strain aging.

It is probable, however, that a considerable increment of creep-rupture strength is imparted to A-286 alloy by the same conditions which cause susceptibility to strain aging in the alloy. Moreover, it is quite possible that the wide variations in the creep-rupture properties of A-286 alloy with a given final heat treatment is due to the influence of hot-working conditions on the state of the interstitial elements which participate in strain aging.

Two experimental approaches are being used. Tests are being conducted on specimens in regular creep units using manual adjustment of the load to maintain constant-strain rates. An automatic constant-strain rate machine was constructed under Contract AF 33(616)-3239. When the operational problems with this unit are solved it will be utilized to obtain data under conditions which are too difficult for manual tests.

## Processing of Materials

Steel C (see page 3 ) is a silicon-deoxidized 1020 carbon steel. The stock was heated at 1650°F for 1 hour and air cooled. This provided material strongly subject to strain aging. Steel F (see page 3 ) was deoxidized with sufficient aluminum so that it is relatively free from strain aging when annealed from 1650°F. The aluminum combines with nitrogen in the steel removing it from solution and drastically reduces the susceptibility of the steel to strain aging.

The A-286 alloy was solution treated at 1800°F for 1 hour, oil quenched, and then aged at 1325°F for 16 hours. This treatment was selected as being the highest commercial solution-treating temperature and, therefore, most likely to induce strain aging characteristics with standard heat treatment alone. The aging treatment was standard for the alloy. This treatment also duplicates the treatment used by Captain Domian prior to conducting hot-hardness tests.

Standard specimens were machined from stock after heat treatment.

# RESULTS

To date, manual constant-strain rate tests have been conducted on the two carbon steels at a rate of 0.25 percent per hour. Tests covered the range from 350° or 400° to 900° or 1000°F. Sporadic yielding occurred up to 500°F.

At 350°F this was so prevalent in the silicon-deoxidized Steel C that it was virtually impossible to maintain a constant-strain rate. The effect was quite small at 500°F. The change in stress for total deformations of 0.5, 1.0 and 2.0 percent for the two steels are shown by Figures 8, 9, 10 and 11. These curves show the following:

1. Steel C (subject to strain aging) decreased in strength with increasing temperature at about the same rate as Steel F (aluminum deoxidized to reduce strain aging) for temperatures above 600°F. There might be a somewhat faster rate of decrease for Steel F for deformations of 0.5 and 1.0 percent.

2. Steel C showed a peak in strength at least for deformations of 1.0 and 2.0 percent at the usual temperature for strain aging of about 400°F. At a total strain of 0.5 percent the peaks at 400°F were not evident due to the fact that only a very small amount of plastic strain was involved in this amount of total strain.

3. Although Steel F was heat treated to eliminate strain aging the test conducted at 400°F exhibited a somewhat erratic or serrated stress-strain curve. The effect of the irregularities was small, however, and the resulting absence of peaks in Figure 9 is as would be expected for this condition of minimum susceptibility to strain aging.

These data are now being reviewed for significance. The 0.25 percent per hour rate is quite rapid and may not be low enough to show the expected difference. This is suggested by the slightly slower rates of decrease in strength at the smaller deformations for Steel C. Test data at the high rates used for tensile tests are needed for comparison purposes. Slower strain rate tests are also needed to indicate strengths in the creep testing range.

The curves for Steel C lie at a higher stress level than those for Steel F. This may be the most significant part of the results. Before this is accepted, however, the influence of the difference in heat treatment must be resolved. When the tests were undertaken it was expected that there would be a shift in the peak observed in the stress versus temperature curve toward higher temperatures as the strain rate and total strains were reduced. If no peak, as such, is observed at least the strength for a given deformation would be expected to drop off more slowly with increasing temperature for a material which was subject to strain aging than for one which was not. Therefore, the heat treatments were selected to make as wide a difference in the strain aging characteristics as possible.

Similar data at 1000° to 1200°F are shown for the A-286 alloy. Sporadic yielding was so large at 800°F that the deformation strengths could not be established. Values for 0.5 percent deformation were so close to the proportional limit that they are not shown. The values for 1.0 percent deformation are above those for 2.0 percent at 1100° and 1200°F due to a steady decrease in load necessary to maintain the strain rate at 0.25 percent per hour after the deformation reached 1.0 percent. These data are shown in Figure 12.

The data on A-286 are being analyzed and further work planned. The alloy with a treatment at 1800°F definitely shows evidence of strain aging at temperatures below 1000°F. The problem remains to integrate this with the properties at 1200°F. The data are also being correlated with the hot-hardness values reported by Captain Domian.

## REFERENCES


1. 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.

2. S. Weissmann, "Quantitative Study of Substructure Characteristics and Correlation to Tensile Properties in Nickel and Nickel Alloys," J. Appl. Phys., 27 (April, 1956) 1335.

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.

TABLE I

CREEP DATA USED FOR ESTIMATING THE CREEP STRENGTH OF

"A" NICKEL AT 800°F

| Rolling Conditions | | | Creep Data at 800°F | | |
|---|---|---|---|---|---|
| Initial Temp. (°F) | Red. of Area (%) | No. of Passes | Stress (psi) | Min. Creep Rate(%/hr) | Extrapolated Stress for 0.0001%/hr M.C.R. (psi)** |
| | | | 15,000 a | 0.000265 | |
| 80 | 0 | 0 | 13,000 * | 0.000277 | 9,500 |
| | | | 9,000 * | 0.000085 | |
| | 4.5 | 2 | 18,000 * | 0.000070 | 19,000 |
| | 11.3 | 4 | 30,000 | 0.000072 | 31,000 |
| | 16.5 | 6 | 34,000 | 0.000088 | 34,500 |
| | 20.5 | 6 | 33,000 | 0.000043 | 36,000 b |
| | 35.5 | 8 | 42,000 * | 0.000060 | 43,500 |
| | 61.2 | 12 | 50,000 | 0.000110 | 50,000 c |
| 1400 | 4.7 | 2 | 20,000 | 0.000076 | 21,000 |
| | 9.7 | 2 | 20,000 | 0.000022 | 25,000 |
| | 14.7 | 2 | 29,000 | 0.000190 | 27,000 |
| | 26.1 | 2 | 32,000 | 0.000110 | 31,500 |
| | 36.1 | 2 | 34,000 | 0.000077 | 34,500 |
| | 19.5 | 3 | 31,000 | 0.000145 d | 30,000 |
| | 33.8 | 4 | 40,000 | 0.000144 | 38,000 |
| | 58.7 | 6 | 49,000 | 0.000170 | 47,500 |
| 1600 | 0 | 0 | 16,000 | 0.000294 | 10,000 |
| | 5.3 | 2 | 20,000 | 0.000178 | 18,000 |
| | 9.9 | 2 | 24,000 | 0.000195 | 22,000 |
| | | | 26,000 | 0.000260 | |
| | 15.2 | 2 | 24,000 | 0.000130 | 23,000 |
| | | | 22,000 | 0.000080 | |
| | 26.7 | 2 | 32,000 | 0.000242 | 29,000 |
| | 36.8 | 2 | 32,000 | 0.000150 | 30,500 |
| | 20.3 | 3 | 28,000 | 0.000182 | 26,000 |
| | 34.4 | 4 | 40,900 | 0.000415 | 36,000 |
| | 59.1 | 6 | 48,000 | 0.000165 | 46,500 |
| 1800 | 0 | 0 | 15,000 | 0.000158 | 11,000 |
| | 5.8 | 2 | 19,000 | 0.000110 | 18,500 |
| | 10.8 | 2 | 19,000 | 0.000206 | 16,500 |
| | 15.7 | 2 | 19,700 | 0.000103 | 19,500 |
| | 27.3 | 2 | 18,000 | 0.000200 | 15,500 |
| | 37.5 | 2 | 19,000 | 0.000245 | 16,000 |
| | 20.7 | 3 | 19,000 | 0.000115 | 18,500 |
| | 35.2 | 4 | 29,000 | 0.000303 | 25,500 |
| | 59.3 | 6 | 34,000 | 0.000075 | 35,000 |

Data obtained after Reference 1 was written.

* Extrapolations were carried out using the slope determined by the three tests on material reduced 15.2% at 1600°F.

Incorrectly reported in Reference 1 as 14,000 psi

Misplotted in Figures 6 and 12 of Reference 1 as 35,000 psi.

Misplotted in Figures 6 and 12 of Reference 1 as 49,000 psi.

Reported in Reference 1 as 0.000183% per hour.

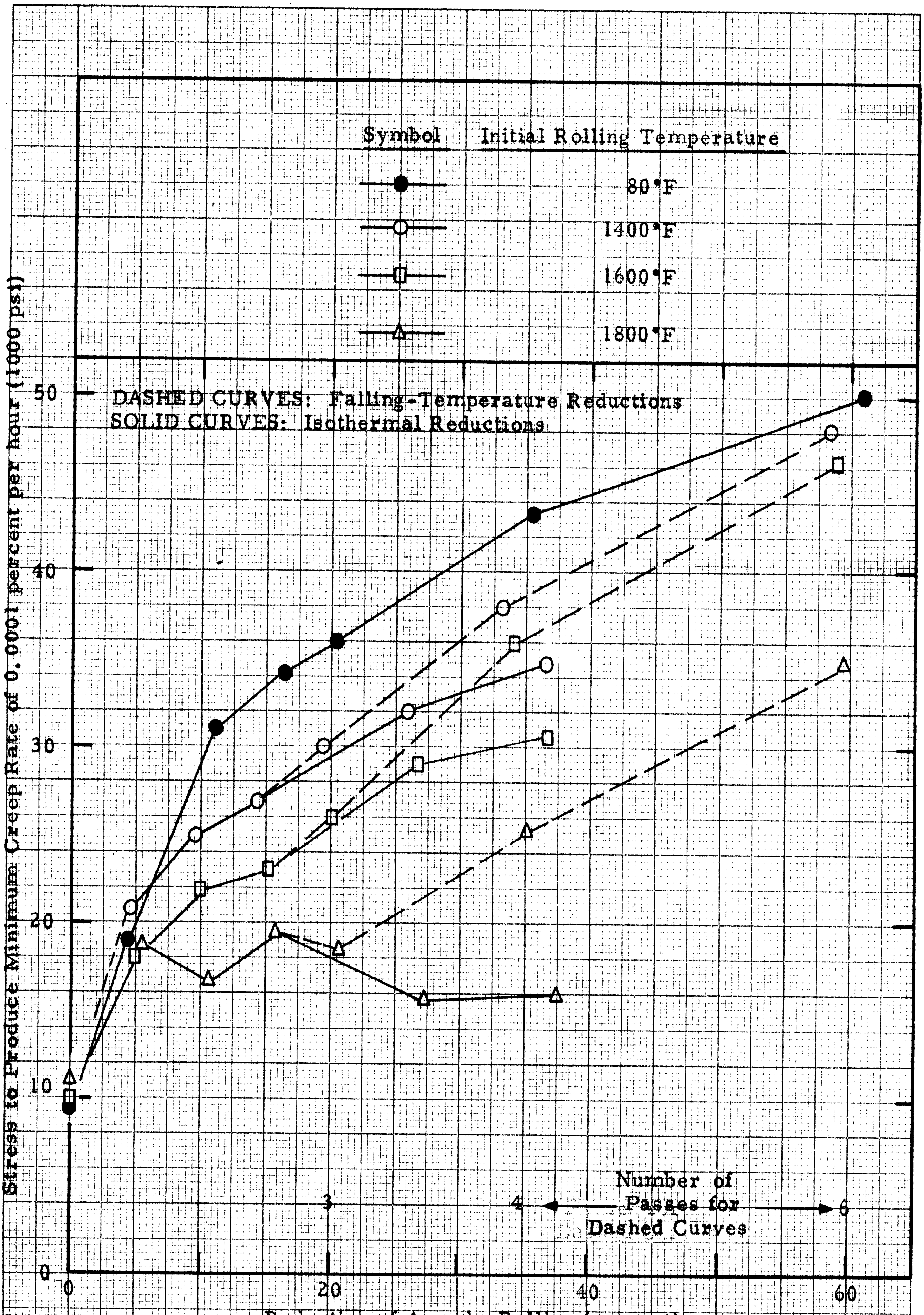


Figure 1. - Estimated Stress to Produce Minimum Creep Rate of 0.0001 percent per hour at 800°F for "A" Nickel.

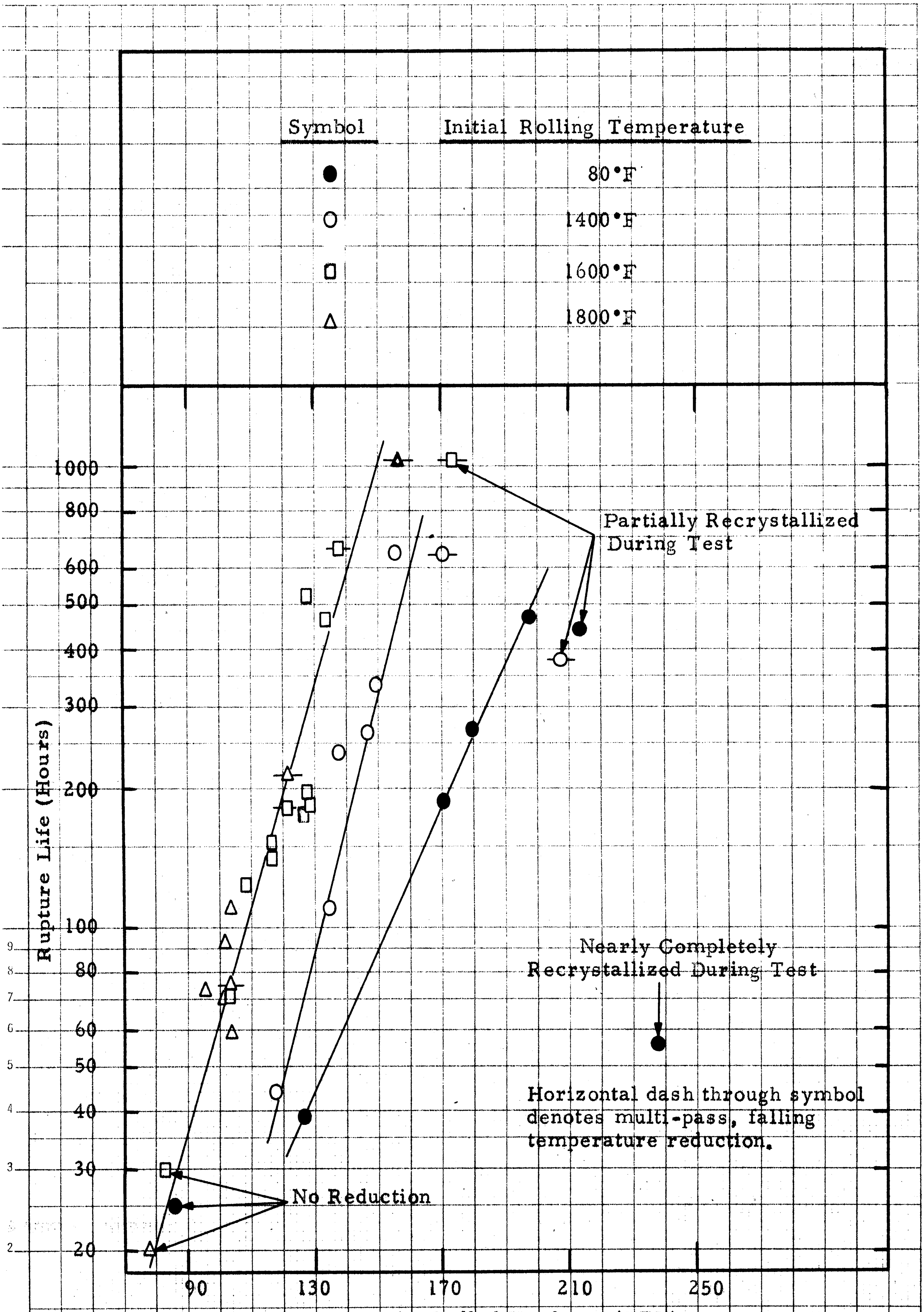


Figure 2. - Correlations of Log Rupture Life with As-Rolled Hardness for "A" Nickel. Tested at 1100°F and 20,000 psi.

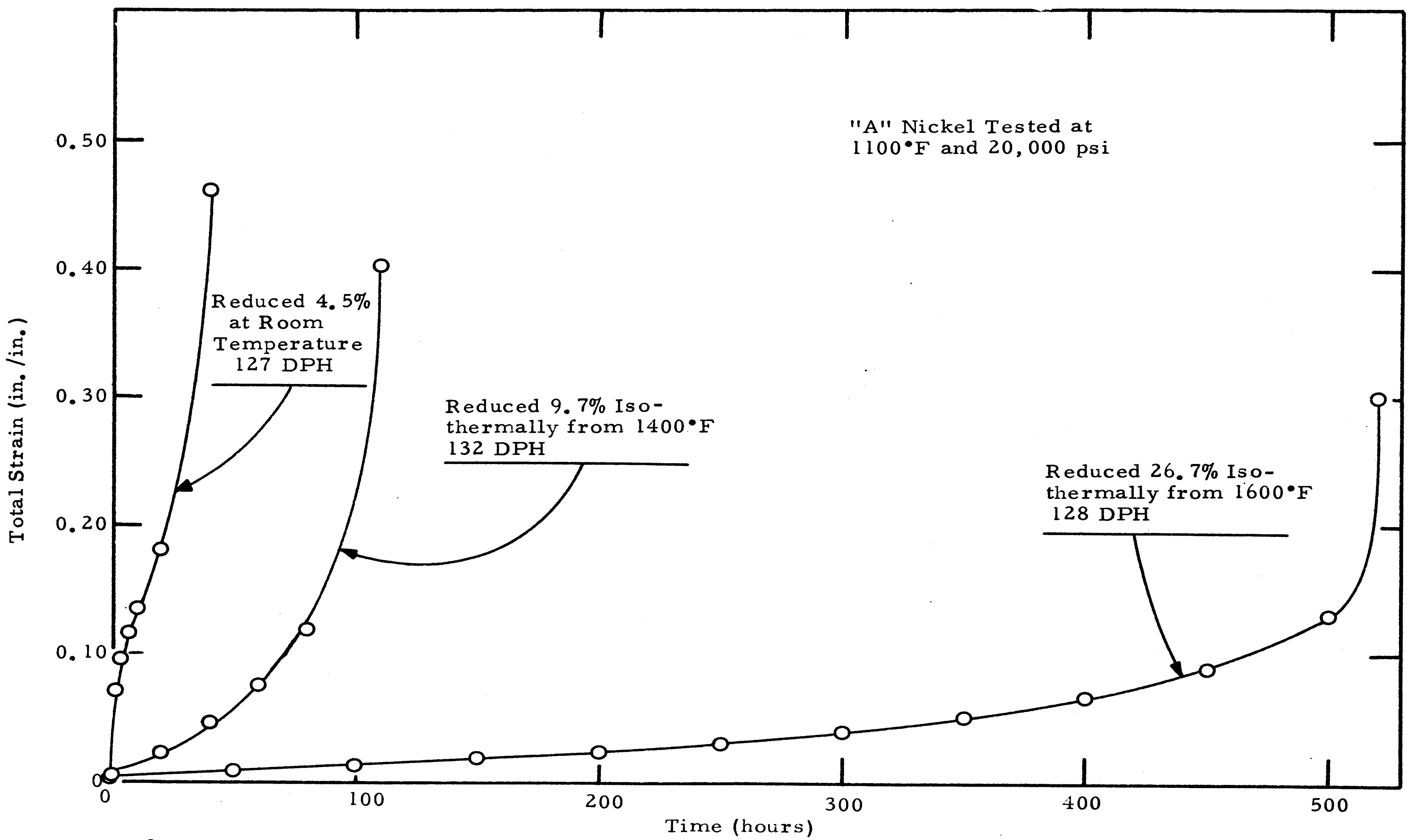


Figure 3. - Total Strain Versus Time Curves for "A" Nickel at 1100°F and 20,000 psi Showing the Effect of Initial Rolling Temperature for a Constant Hardness of 129 DPH.

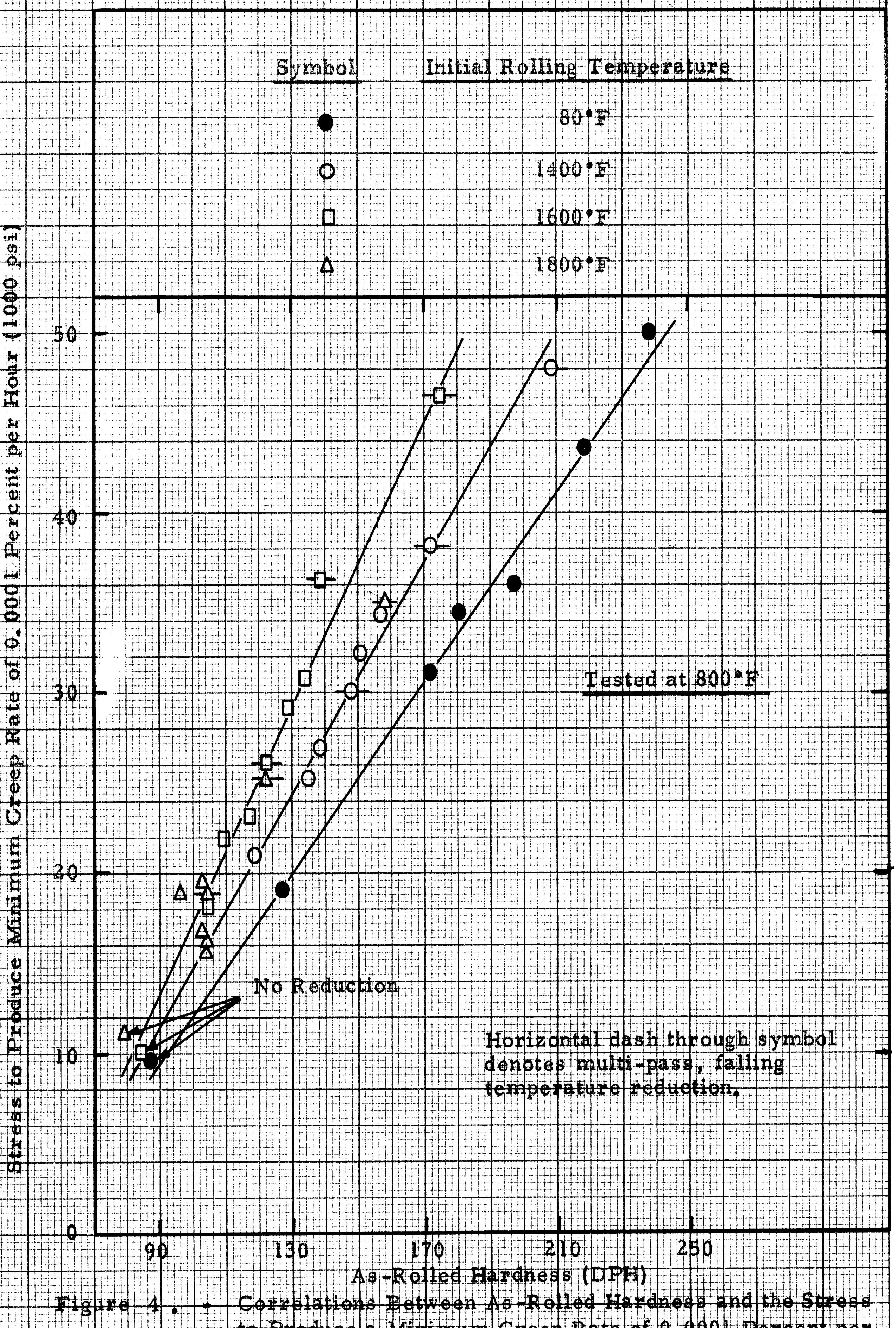


Figure 4. - Correlations Between As-Rolled Hardness and the Stress to Produce a Minimum Creep Rate of 0.0001 Percent per Hour for "A" Nickel Rolled at 80°, 1400°, 1600°, and 1800°F and Creep Tested at 800°F.

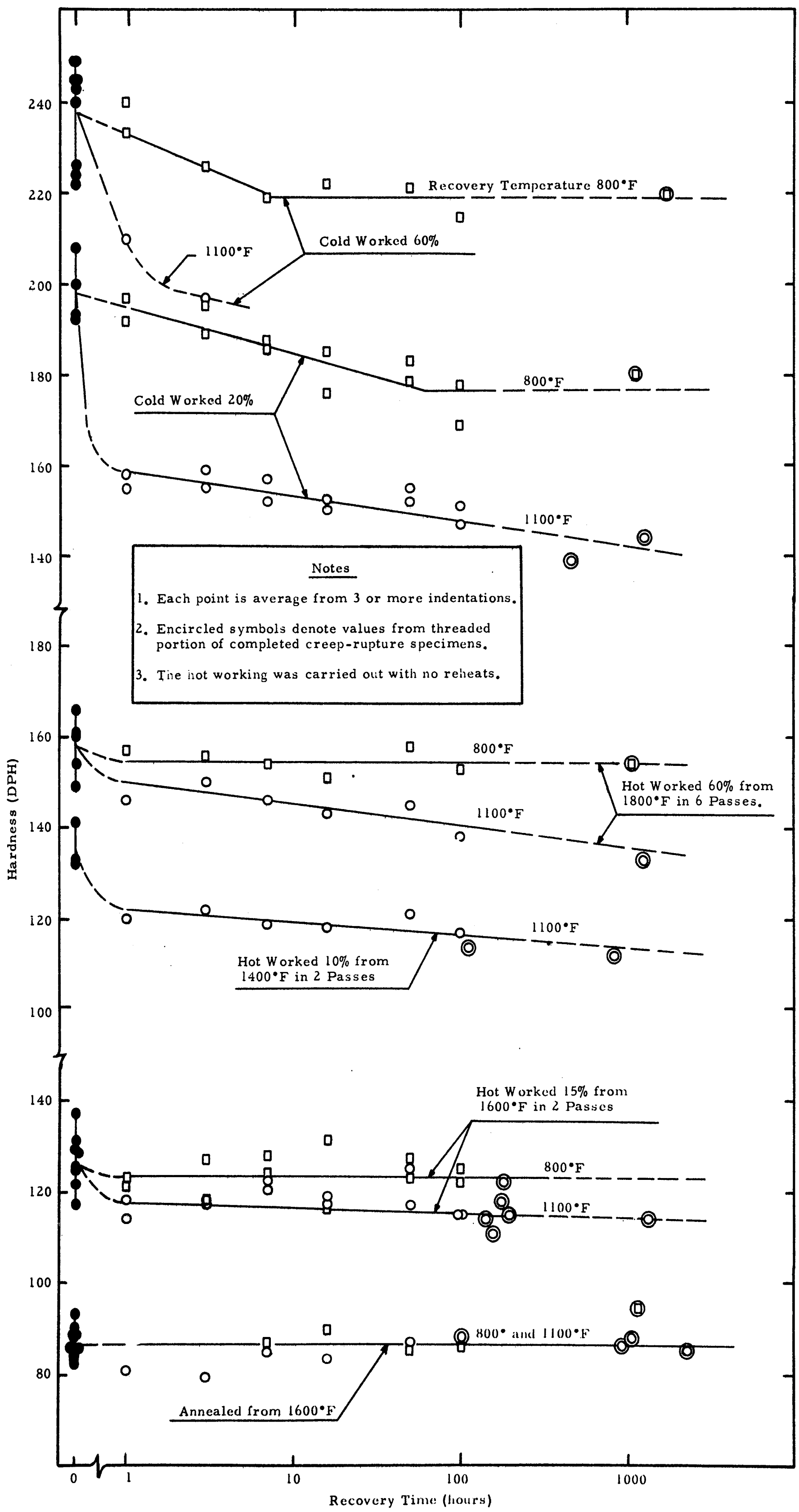


Figure 5. - Hardness Versus Recovery Time at 800° and 1100°F for "A" Nickel for Several Conditions of Prior Working.

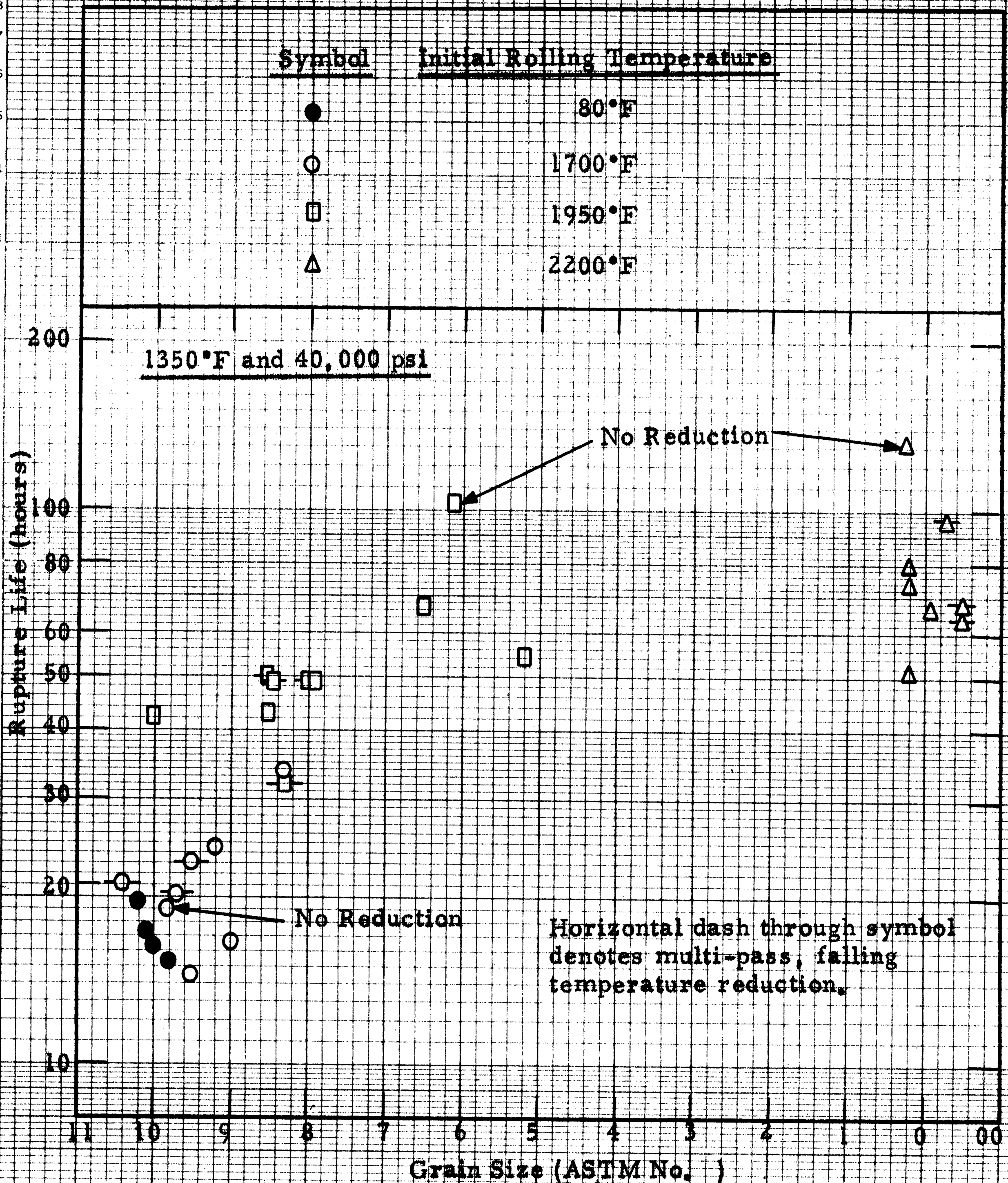


Figure 6. - Log Rupture Life Versus Grain Size for A-286 Rolled at 80°, 1700°, 1950°, and 2200°F, Heat Treated, and Tested at 1350°F and 40,000 psi. Heat treatment after rolling was 1 hour at 1650°F, oil quench, plus 16 hours at 1325°F, air cool.

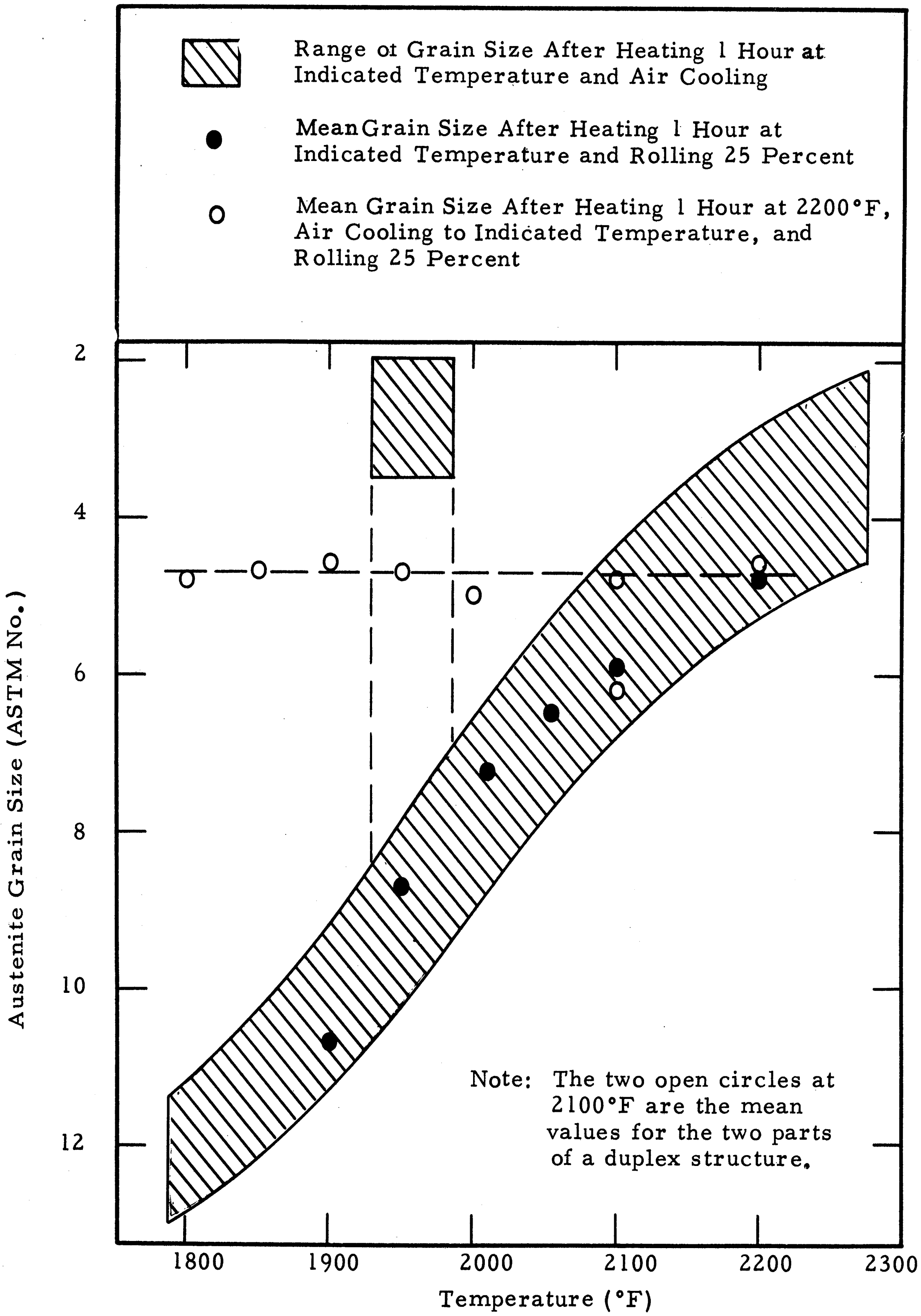


Figure 7. - Variation of Austenite Grain Size for Three Types of Experiments for "17-22-A"V Steel.

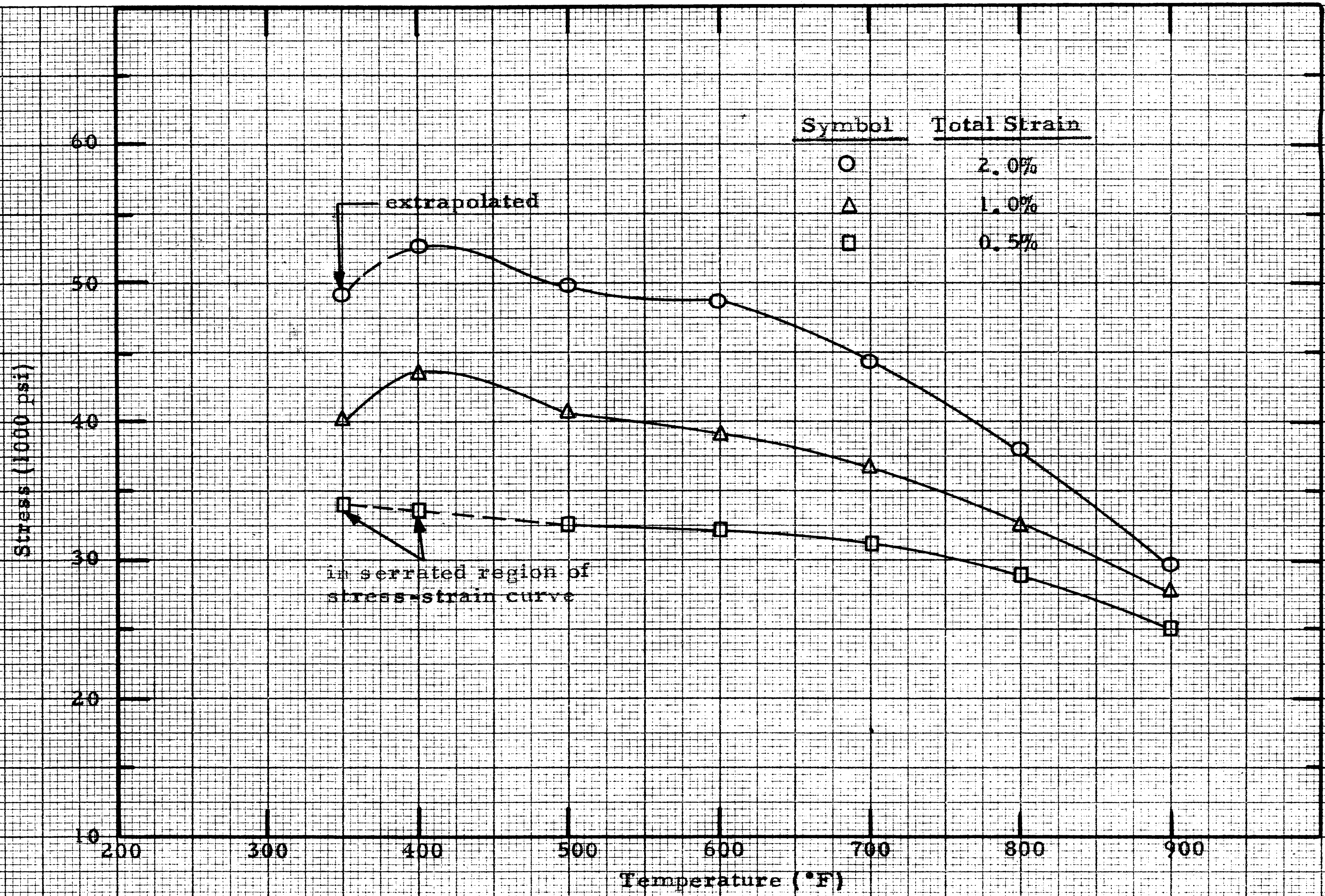


Figure 8. - Stress-Temperature Curves for Steel C at a Strain Rate of 0.25 Percent per Hour.

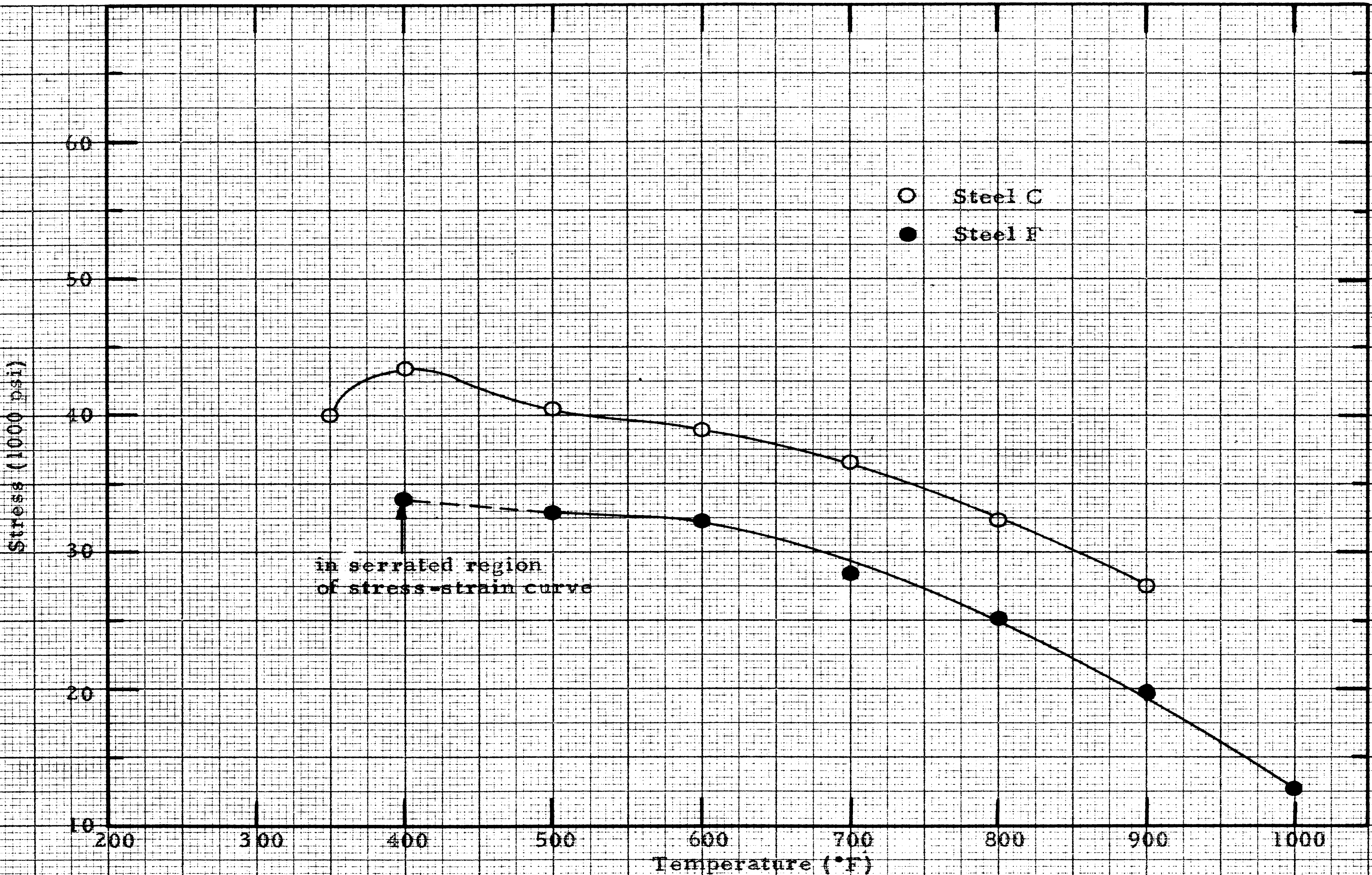


Figure 10. - Stress-Temperature Curves for Steels C and F at 1 Percent Total Strain for a Constant Strain Rate of 0.25 Percent per Hour.

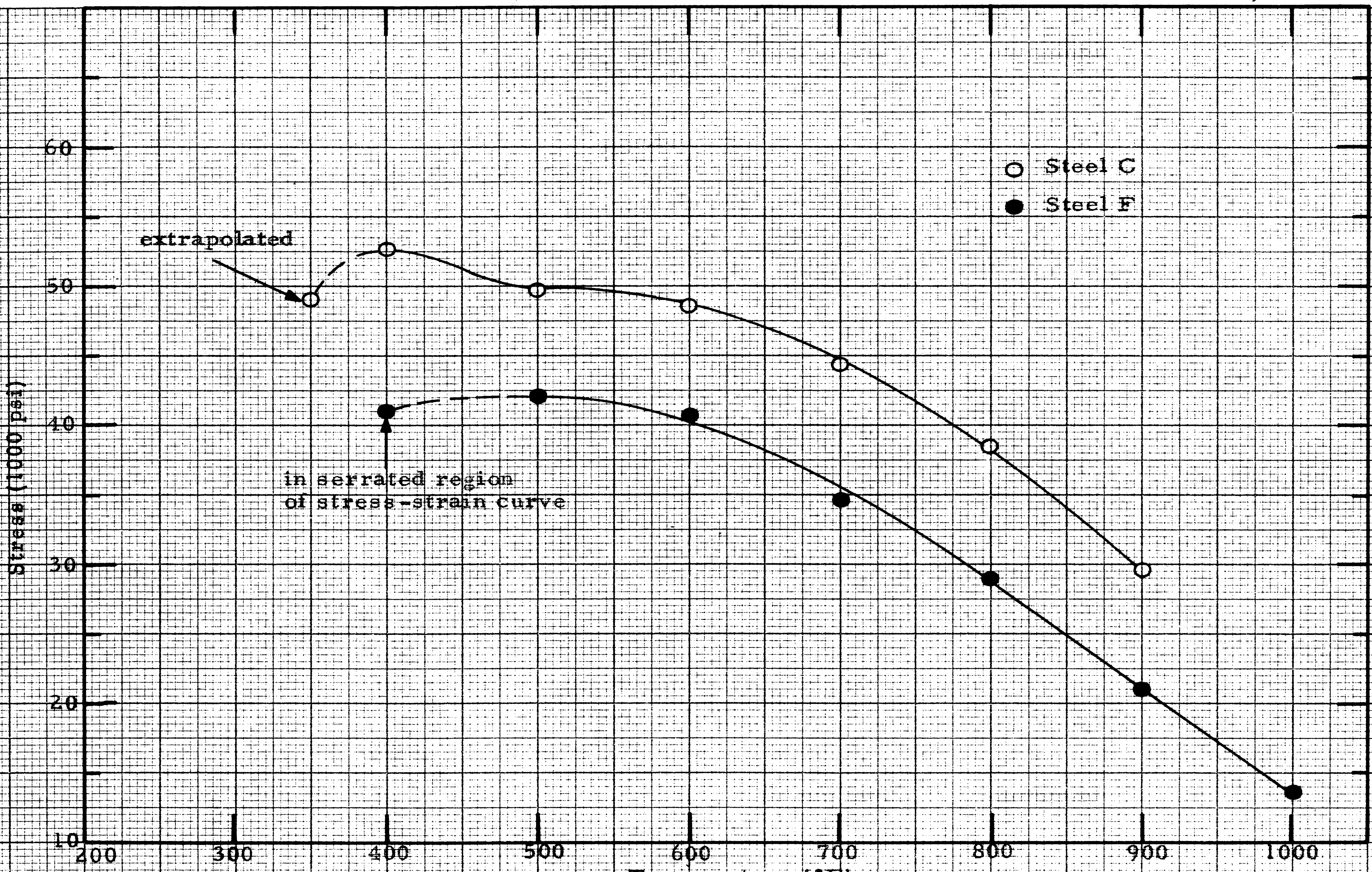


Figure 11. - Stress-Temperature Curves for Steels C and F at 2 Percent Total Strain for a Constant Strain Rate of 0.25 Percent per Hour.

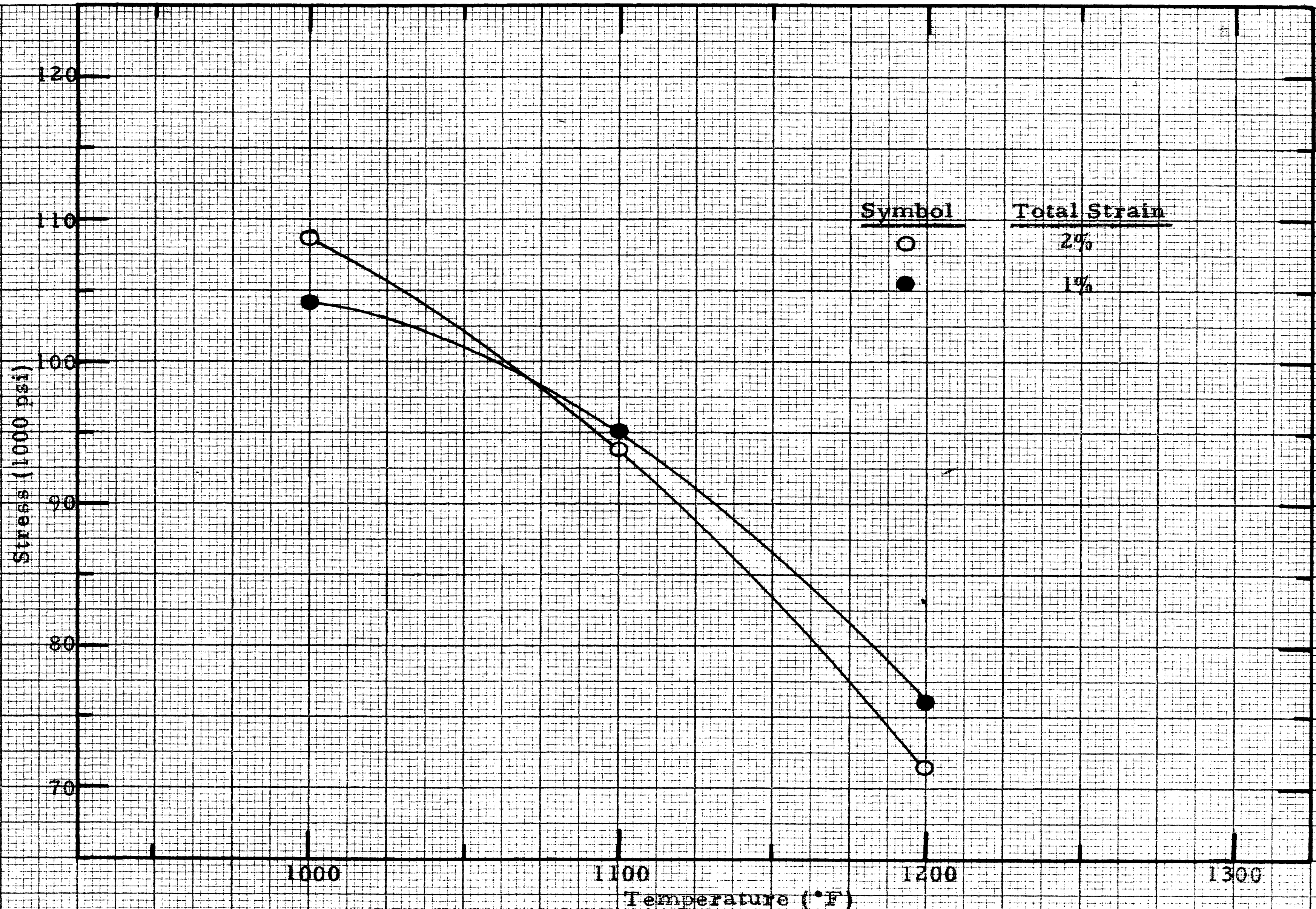


Figure 12. - Stress-Temperature Curves for A-286 Alloy at a Strain Rate of 0.25 Percent per Hour.