FINAL REPORT

TO

MATERIALS LABORATORY

WRIGHT AIR DEVELOPMENT CENTER

ON

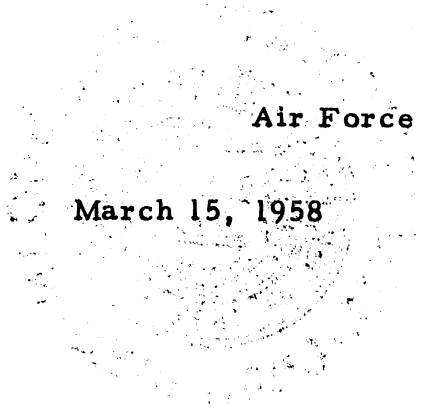
AN INVESTIGATION OF THE RELATIONSHIP BETWEEN MICROSTRUCTURE AND CREEP-RUPTURE PROPERTIES OF HEAT-RESISTANT ALLOYS

by

A. P. Coldren

J. W. Freeman

Engineering Research Institute
The University of Michigan



Project 2460

Air Force Contract No. AF 33(616)-3239
Task No. 73512

March 15, 1958

FOREWORD

This report was prepared by the Engineering Research Institute of the University of Michigan under USAF Contract Number AF 33(616)-3239. The contract was initiated under Task Number 73512, with Mr. C. B. Hartley acting as project engineer for the Materials Laboratory, Wright Air Development Center. This report covers work done from December 15, 1956 to March 15, 1958.

The A-286 and the "A" Nickel used in this research were supplied gratis by the Allegheny Ludlum Steel Corporation and the International Nickel Company, respectively.

SUMMARY

Progress is reported for an investigation of the influence of conditions of hot working on the properties of alloys at high temperatures. Conditions of working can be used to control microstructural variations in a manner which cannot be obtained in any other procedure and is capable of developing structures superior in properties to any other treatment. The major objective is to define the basic principles involved so that they can be applied to the general problem of developing optimum properties in any alloy. The relatively simple structure of "A" Nickel is being used as an experimental material for study of the role of working for properties in the as-worked condition. A-286 alloy is being used as an example of a material whose properties are influenced after solution and aging treatments by the conditions of prior working.

The results reported cover the initial surveys of the relationships of working conditions to creep and rupture properties. Structural analyses to define the basic principles involved were confined to preliminary partial studies. The investigation is being continued with emphasis on the structural studies.

TABLE OF CONTENTS

	Page
INTRODUCTION	1
TEST MATERIALS	2
EXPERIMENTAL PROCEDURE	3
Preparation of Stock	3
Conditions of Working	3
Heat Treatment	4
Temperature Measurements	4
Hardness Measurements	5
Evaluation of Creep and Rupture Properties	5
RESULTS AND DISCUSSION	6
Commercially Pure Metal ("A" Nickel)	6
Rupture Properties at 1100°F and 20,000 psi	6
Creep Properties at 1100°F and 11,000 psi	7
Creep Characteristics at 800°F	8
Relation of Rupture and Creep Properties to Hardness	8
Relation of Microstructure to Rupture and Creep Properties	11
General Discussion	12
Age-Hardenable, Austenitic Alloy (A-286)	12
Rupture Properties at 1200°F	13
Rupture Properties at 1350°F	13
Creep Resistance at 1200°F and 35,000 psi	13
Relation of Properties to Hardness	14
Relation of Properties to Microstructure	14
General Discussion	16
CONCLUSIONS	17
"A" Nickel	17
A-286 Alloy	17
REFERENCES	19

LIST OF TABLES

Table

I Rolling Conditions and Rupture, Creep, and Total Deformation Data for "A" Nickel Rolled Isothermally and Non-Isothermally with no Reheats

II Rolling Conditions and Rupture, Creep, and Total Deformation Data for A-286 Alloy Rolled Isothermally and Non-Isothermally with no Reheats

LIST OF ILLUSTRATIONS

Figure

1 Cooling Curves for 7/8-Inch Square Bars of "A" Nickel Air Cooled from 1400°, 1600°, and 1800°F.

2 Cooling Curves for 3/4-Inch Diameter Bars of A-286 Alloy Air Cooled from 1700°, 1950°, and 2200°F.

3 Variation of Log Rupture Life with Rolling Conditions for "A" Nickel Tested at 1100°F and 20,000 psi

4 Variation of Rupture Ductility with Rolling Conditions for "A" Nickel tested at 1100°F and 20,000 psi

5 Variation of Log Minimum Creep Rate with Rolling Conditions for "A" Nickel Tested at 1100°F and 11,000 psi

6 Influence of Rolling Conditions on the Stress to Produce a Minimum Creep Rate of 0.0001 Percent per Hour for "A" Nickel at 800°F

7 Relationship Between Hardness and Rolling Conditions for "A" Nickel

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

9 Total Strain Versus Time Curves for "A" Nickel at 1100°F and 20,000 psi Showing the Effect of Two Different Sets of Rolling Conditions at a Constant Hardness of 130 DPH

10. Total Strain Versus Time Curves for "A" Nickel at 1100°F and 20,000 psi Showing the Effect of Two Different Sets of Rolling Conditions at a Constant Hardness of 118 DPH

11 Correlation of Log Minimum Creep Rate with As-Rolled Hardness for "A" Nickel Tested at 1100°F and 11,000 psi

12 Correlations Between As-Rolled Hardness and the Stress to Produce a Minimum Creep Rate of 0.0001 Percent per Hour for "A" Nickel at 800°F.

13 Microstructure of "A" Nickel in the (a) Base Condition and As Subsequently Cold Rolled (b) 16.5 Percent, (c) 35.5 Percent and (d) 61.2 Percent.

14 Microstructure of "A" Nickel Rolled at 1400°F.

Figure

15. Microstructure of "A" Nickel Heated to 1600°F and (a) Air Cooled, (b) Rolled 20.3 Percent in 3 Passes, (c) Rolled 59.1 Percent in 6 Passes, and (d) Rolled 36.8 Percent in 2 Passes.

16. Microstructure of "A" Nickel Heated to 1800°F and (a) Air Cooled, (b) Rolled 20.7 Percent in 3 Passes, (c) Rolled 59.3 Percent in 6 Passes, and (d) Rolled 37.5 Percent in 2 Passes.

17 Appearance of Substructure in "A" Nickel at Different Magnifications

18 Effect of Orientation and Inhomogeneous Strain on the Appearance of Substructure in a Single Sample of "A" Nickel.

19 Variation of Log Rupture Life with Prior Rolling Conditions for Heat Treated A-286 Alloy Tested at 1200°F and 65,000 psi

20 Influence of Prior Rolling Conditions on the Ductility of Heat Treated A-286 Alloy Tested at 1200°F and 65,000 psi

21 Variation of Log Rupture Life with Prior Rolling Conditions for Heat Treated A-286 Alloy at 1350°F and 40,000 psi

22 Influence of Prior Rolling Conditions on the Ductility of Heat Treated A-286 Alloy Tested at 1350°F and 40,000 psi

23 Effect of Percent Reduction of Area by Rolling on the Creep Curve for A-286 Alloy Rolled at Room Temperature, Heat Treated, and Creep Tested at 1200°F and 35,000 psi

24 Effect of Percent Reduction of Area by Rolling on the Creep Curve for A-286 Alloy Rolled at 1700°F, Heat Treated, and Creep Tested at 1200°F and 35,000 psi

25 Effect of Percent Reduction of Area by Rolling on the Creep Curve for A-286 Alloy Rolled at 1950°F, Heat Treated, and Creep Tested at 1200°F and 35,000 psi

26 Effect of Percent Reduction of Area by Rolling on the Creep Curve for A-286 Alloy Rolled at 2200°F, Heat Treated, and Creep Tested at 1200°F and 35,000 psi

27 Effect of Heating Without Rolling on the Creep Curve for Heat Treated A-286 Alloy Creep Tested at 1200°F and 35,000 psi

28 Effect of Rolling Temperature on the Creep Curve for A-286 Alloy Reduced Isothermally 9.4 ($\pm$1.0) Percent, Heat Treated, and Creep Tested at 1200°F and 35,000 psi

29 Effect of Rolling Temperature on the Creep Curve for A-286 Alloy Reduced Isothermally 20.0 ($\pm$2.0) Percent, Heat Treated, and Creep Tested at 1200°F and 35,000 psi

LIST OF ILLUSTRATIONS (continued)

Figure

30 Effect of Rolling Temperature on the Creep Curve for A-286 Alloy Reduced Isothermally 35.7 (±4.0) Percent, Heat Treated, and Creep Tested at 1200°F and 35,000 psi

31 Effect of Prior Working Conditions on the Heat-Treated Hardness of A-286 Alloy

32 Microstructure 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.

33 A-286 Alloy Hot Rolled from 2200°F and Subsequently Solution Treated and Aged

34 Microstructure of A-286 Alloy Solution Treated and Aged after Rolling 20.0 (±2.0) Percent at (a) Room Temperature, (b) 1700°F, (c) 1950°F, and (d) 2200°F

INTRODUCTION

Ewing and Freeman (1)* demonstrated that creep and rupture strengths for low carbon N-155 could be raised appreciably by controlled hot rolling. Frey and Freeman (2) studied the effects of cold working on the same alloy. Others (3-11) have reported similar work on several ferrous and non-ferrous alloys. The general conclusion of these studies was that for all materials both strengthening and weakening are possible effects of working. Frequently it was found that maximum strengths were obtained with an optimum degree of working which became lower as the test temperature was raised. Working much beyond the optimum amount sometimes reduced strength below that of the annealed or base condition. In a few instances the optimum reduction was reduced to zero with the result that any working at all was detrimental. Except for the work by Ewing and Freeman (1) the vast majority of the work cited above involved working only at room temperature. Very few data are available relating high temperature properties to hot-working conditions where the results are amenable to the development of basic metallurgical principles.

The results presented in this report consist mainly of creep and rupture data shown as a function of working conditions. Except for a few preliminary structural observations already made, the correlation of structures with properties will be done as a part of the next phase of the investigation.

The investigation was sponsored by the Materials Laboratory, Wright Air Development Center under Contract AF 33(616)-3239.

The objective of the investigation is to provide the basic knowledge necessary to apply controlled hot-working conditions to provide optimum properties in alloys at high temperatures. It is known that maximum, all-around engineering properties are observed for the hot-worked condition of many alloys. The principles involved are not, however, understood well enough to enable predictable application to alloys. Secondly, the properties of alloys at high temperatures are known to vary to a marked extent in some cases with the conditions of working. This is the source of much of the common variation between such products as sheet, plate, forgings, and bars. The investigation has been undertaken with the expectation that the results can be applied to reduce variation in properties and to develop better properties than by customary production procedures.

In the investigation general principles are being sought which will have general application. The principles should be adaptable to any alloy from aluminum or titanium to super alloys or high melting point alloys based on molybdenum, tungsten, or niobium. In selecting metals for experimental purposes two requirements were established. One material should have a simple microstructure, preferrably a pure metal, where extraneous reactions such as precipitation would not complicate the study. "A" Nickel came closest to meeting this requirement. The other material was required to show pronounced effects from working conditions which persisted after standard heat treatments. A-286 alloy was chosen for this purpose because of its known wide variability in properties which appeared to be due to variation in prior hot-working conditions.

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

General principles are being developed by determining the basic cause of observed effects through study of the microstructural characteristics as influenced by hot working. The research evolved naturally from previous research which sought to evaluate similar general principles using heat treatment to vary microstructures. Variable response to a given heat treatment due to prior working history of alloys is a source of major variation.

Due to the normal variations in properties of alloys at high temperatures, both power plants and airframes have to pay a considerable penalty in weight. Designs must be based on minimum statistically expected properties. This becomes of even more importance in missles and space vehichles due to the ever increasing penalty for excess weight. Even if the results only raise the minimum assured properties and provide a basis for avoiding the unexpected low values so often found in fabricating unfamiliar parts, the investigation will be very worthwhile. There is, however, a greater potential in having to utilize working conditions to develop even better properties.

TEST MATERIALS

For the study of the influence of working conditions on properties in the as-worked condition, certain metallurgical requirements were needed for the experimental material. The ideal material would have a simple, single-phase microstructure free from precipitates and other complicating structural changes. Because substructures as controlled by working and testing conditions were expected to be the main structural variable by which working conditions would influence properties, the material should be one in which substructures could be studied with a minimum of difficulty. "A" Nickel was selected as the nearest approach to the ideal material which was readily available in the quantities needed for the required creep and rupture testing program. It was recognized that "A" Nickel does contain a number of impurities which could lead to the formation of precipitates.

The second material needed for the investigation had to be subject to pronounced variation in properties after a solution treatment as a result of variable prior working conditions. The Ti + Al strengthened alloys widely used in aircraft gas turbines and airframes appeared to be the most subject to this variation. A-286 alloy was selected as a typical material for the type of basic information required from the investigation.

The "A" Nickel was supplied gratis by The International Nickel Company as 1-inch diameter cold-drawn bars. The A-286 alloy was supplied by the Allegheny Ludlum Steel Corporation as hot-rolled 3/4-inch diameter bars. The bars had been rolled from consumable electrode, vacuum melted ("Consutrode") Heat Number 21030. The chemical compositions of the two materials reported by the manufacturers were as follows;

Alloy	C	Mn	Si	Cr	Ni	Mo	V	Fe	Other
"A" Nickel	0.06	0.27	0.06	--	99.46(Ni+Co)	--	--	0.09	0.03 Cu; 0.008 S
A-286	0.06	1.35	0.47	14.58	25.3	1.38	0.21	Base	2.00 Ti; 0.17 Al; 0.004 B

The history of the A-286 stock prior to the time it was supplied for the investigation was reported by Allegheny Ludlum as follows:

1. The 20-inch diameter ingot was pressed and cogged from 2150°F to a 4-1/4-inch square.
2. Recogged to 2-7/8-inch square from 2100°F.
3. Rolled to 3/4-inch diameter round from 2100°F.
4. No straightening or other conditioning was applied to the bars before shipment.

EXPERIMENTAL PROCEDURE

Preparation of Stock

The "A" Nickel barstock was put into a standard, base condition as follows: The 1-inch diameter rods were cold rolled 20 percent into 7/8-inch round cornered square rods and batch annealed at 1600°F for 1 hour. The annealed grain size was 2 to 4 on the ASTM scale, and the hardness was 88 DPH. The annealing temperature was limited to 1600°F to avoid excessive grain growth.

The as-received plus heat-treated condition of the A-286 was used as the base condition. By special order the A-286 rods were shipped in the unstraightened, as-hot rolled condition to ensure maximum lineal uniformity of microstructures and residual stresses.

Conditions of Working

In this report the meaning of the term "hot working" is extended beyond the classical definition to include working of heated bars regardless of whether or not recrystallization occurs during the operation.

The mode of working was limited to closed-pass rolling because the temperatures and amounts of working are fairly easily controlled, and the lineal unfformity of working is good. The rolling was done on a two-high mill equipped with 9-1/2-inch diameter rolls driven at 30 RPM. Closed, square passes were used without a lubricant. Heating for rolling at elevated temperatures was done in a gas-fired furnace adjacent to the mill.

Two types of rolling schedules were used for each alloy. One schedule was designed to approximate isothermal working conditions and consisted of two passes in rapid succession. The maximum reduction under these conditions was determined by the capacity of the mill to deform the alloys at their lowest hot rolling temperatures. The other type of rolling schedule was intended to simulate commercial practice where the last stage of hot working is usually carried out over a range of falling temperature. This involved successive reductions without reheats using 3, 4, 5, or 6 passes, depending upon the total reduction.

To obtain a rather broad survey of working conditions a wide range of rolling temperatures was used for both alloys. The highest temperature in each case was selected to give increasing amounts of recrystallization with increasing reduction. For purposes of comparison a series of samples rolled at room temperature were included for both the "A" Nickel and the A-286 alloy.

The nominal temperatures and reductions were as follows:

Material	Initial Rolling Temperature	Nominal Percent Reduction of Area	
		Isothermal	Falling Temperature
"A" Nickel	80°, 1400°, 1600°, 1800°F	0, 5, 10, 15, 27, 37	20, 34, 59
A-286	80°, 1700°, 1950°, 2200°F	0, 5, 10, 20, 26, 34	12, 19, 25, 39

The degree of working was not completely uniform through the bar cross sections. Hardness decreased from the surface to center in a manner which varied with the temperature and degree of reduction. The degree of reduction is, therefore, not a perfect indicator of what happened to the material during rolling. Due consideration of this effect, however, indicated that the non-uniformity of working was too small to affect the general conclusions drawn from the data.

Heat Treatment

The "A" Nickel was tested in the as-rolled condition while the A-286 alloy was solution treated and aged after rolling. The A-286 heat treatment was: 1 hour at 1650°F, oil quench + 16 hours at 1325°F, air cool.

Temperature Measurements

Temperatures were measured with Chromel-Alumel thermocouples using proven pyrometric practices. In addition to the usual temperature measurements, a considerable amount of experimentation was carried out to enable measurement of finishing temperatures during rolling. A method was developed involving a two-tyned thermocouple fork, but the reliability of the method has not yet been established.

The decrease in temperature of the nickel during 2 passes through the mill from 1800°F is now estimated at about 140°F. This arises from air cooling as is shown by Figures 1 and 2, as well as the chilling effect of the rolls. The air cooling causes a drop of about 60°F and the chilling effect of the rolls amounts to about 80°F. This has been checked using a two-tyned thermocouple fork to measure surface temperatures and with a bar carrying a thermocouple in the center of the bar. At present the measurements have only been completed for nickel rolled from 1800°F. No measurements have been obtained for A-286.

The measurements of temperature changes during rolling made to date indicate that the so-called isothermally rolled bars actually decreased in temperature about 150°F during rolling.

Hardness Measurements

Diamond Pyramid Hardness values were obtained for every condition prior to testing. Loads of 30 and 50 Kg. were used for the "A" Nickel and A-286 samples, respectively. The measurements were made on representative longitudinal sections mounted in bakelite and prepared for metallographic examination. Each value was derived from the average of six diagonal measurements on three identations. Care was exercised in locating the three indentations to give an average hardness across a 1/4-inch wide area corresponding to the 1/4-inch diameter reduced section of the test specimens.

Evaluation of Creep and Rupture Properties

Three different measures of high temperature strength were used for each material in order to evaluate the effects of working both for temperature of testing and for strain rate effects. The properties were evaluated using test temperatures relatively low and relatively high in the creep range. Relatively slow strain rate creep tests were used as well as high strain rate rupture tests. The specific test conditions used to evaluate the high temperature properties were as follows:

Material	Test Temperature	Applied Stress	Data Used in the Evaluation
"A" Nickel	1100°F	20,000 psi	Rupture Life and Ductility
	1100°F	11,000 psi	Minimum Creep Rate
	800°F	Variable	Stress to Produce Minimum Creep Rate of 0.0001 Percent per Hour
A-286	1350°F	40,000 psi	Rupture Life and Ductility
	1200°F	65,000 psi	Rupture Life and Ductility
	1200°F	35,000 psi	Creep Strain vs. Time Curve

All of the creep and rupture tests were run at a constant load which was applied by a dead weight through a cantilever beam system. Standard test specimens having a 0.250-inch diameter and a 1-inch gage length were used with a Martens-type optical extensometer. The least reading of the extensometer on these specimens was about 0.00001 inches per inch. Each test was started by placing the specimen into a pre-heated testing unit, making any adjustments necessary to bring the center and ends of the gage section all to within $\pm 3°F$ of the nominal test temperature, and then applying the load. The entire starting procedure usually lasted about four hours. Extension readings were taken one, two, or three times daily, depending upon the rate of elongation.

RESULTS AND DISCUSSION

The experimental results for "A" Nickel and A-286 alloy are presented separately in the following two sections of the report. In each case the data are analyzed and discussed from the viewpoint of the objective of establishing the basic mechanisms by which working conditions can influence properties at high temperatures.

Commercially Pure Metal ("A" Nickel)

The influence of working conditions on hardness, rupture, and creep properties in the as-rolled condition was evaluated with specimens from samples of "A" Nickel rolled as detailed in Table I. Preliminary microstructural studies are also reported.

Rupture Properties at 1100°F and 20,000 psi

The log rupture time (Fig. 3) increased with percent of isothermal reduction nearly independent of the temperature of reduction for rolling at 80°, 1400°, 1600°, or 1800°F until either recrystallization during rolling or during testing limited the strengthening. Recrystallization during rolling at 1800°F caused the rupture time to level off for reductions of more than 6 percent. Recrystallization during testing caused the rupture time of the material rolled at 80°F to fall off for reductions of more than 20 to 30 percent. Both factors were probably operating for reductions of 30 to 40 percent at 1600°F with some possibility that slightly more isothermal reduction at 1400°F would be required for recrystallization during testing. The independence of log rupture time from temperature of reduction and the dependence mainly on degree of reduction, in the absence of recrystallization, was unexpected and, as will be discussed later, is probably a very significant factor in the role of working conditions related to creep-rupture properties.

Five rupture tests were conducted on material from five separate bars rolled 15 percent at 1600°F to establish a measure of the reproducibility of rupture times. When plotted on Figure 3 these tests suggest that the material rolled at 1600°F was slightly weaker than those rolled at 80° or 1400°F. Statistically, however, there was no significant difference based on the available test results. It is concluded, therefore, that any effect of temperature of reduction was relatively small and that the main factor was percent reduction. This apparently applied for reductions up to 20 to 30 percent at 80°, 1400° and 1600°F and up to 6 percent at 1800°F.

When the temperature was allowed to decrease as rolling proceeded, the strength of material rolled from 1800° or 1600°F was increased to higher levels than was obtained by isothermal reduction. Apparently this occurred because reduction was continued at temperatures below which recrystallization continued, and the resulting structures were resistant to recrystallization during testing. The strength was not built up by the larger reductions over a falling temperature range from 1400°F, apparently because recrystallization occurred during testing. The reduction in strength from the larger reductions was not as severe for rolling from 1400°F as for equivalent reductions at room temperature.

In general, rolling under the conditions of this investigation influenced ductility in the rupture tests (Fig. 4) in the opposite direction to the rupture time. Those conditions of rolling which increased rupture time reduced ductility. When rupture time fell off for the larger reductions, the ductility increased. The ductility values were not, however, as independent of the temperature of reduction for the smaller reductions.

These results for "A" Nickel suggest that a build-up in strength for a large total reduction from a number of small reductions on a falling temperature requires an initial working temperature high enough to have allowed at least partial recrystallization during equivalent isothermal reduction. A similar build-up in strength was observed for N-155 alloy in Reference 1 for a number of small reductions on a falling temperature. The suggested relation to potential recrystallization during isothermal reduction was not observed in Reference 1. A review of that data now indicates, however, that the initial temperature for rolling had to be high enough to induce recrystallization during isothermal reduction. Further study of the A-Nickel should clarify the type of microstructure having unusual strength and stability which is developed in this manner.

Creep Properties at 1100°F and 11,000 psi

The effect of rolling conditions on the log minimum creep rate at 1100°F and 11,000 psi is shown in Figure 5. The general pattern of creep resistance was similar in many respects to the rupture resistance (Fig. 3). The creep resistance was apparently somewhat more sensitive to initial structural variations and structural changes during testing. For instance, the total range of creep rates covered about 2.7 orders of magnitude while the overall range of rupture lives (Fig. 3) extended over about 1.7 orders of magnitude. For isothermal reductions of 10 to 25 percent there seemed to be a significant difference between cold-rolled material and material rolled isothermally at 1400° and 1600°F. There was no significant difference in rupture resistance at 1100°F and 20,000 psi in this region.

Creep Characteristics at 800°F

The very large increase in strength at 800°F from working made it impossible to use a single-stress type survey. The use of a median stress would have resulted in excessive yielding on loading in tests on annealed material (See first two lines of Table I); whereas, there would have been no appreciable secondary creep in material in the strongest condition. Therefore, the stress to produce a minimum creep rate of 0.0001 percent per hour was determined approximately for each condition. Each value was obtained by extrapolation of the creep rate of a single test parallel to a typical stress versus log creep rate curve. The typical curve was determined with three tests on nickel rolled 15 percent at 1600°F.

The extrapolated values are presented in Figure 6 as a function of rolling conditions. The strengths continued to increase with percent reduction up to the maximum reduction except for isothermal reductions at 1800°F where recrystallization limited the strength. Apparently the absence of recrystallization during testing at 800°F was the reason why the cold worked material maintained its superior strength even at the highest reduction.

The temperature of rolling had a definite effect on strengths at 800°F. No effect of rolling temperature could be seen in tests at 1100°F and 20,000 psi, and very little effect was found in tests at 1100°F and 11,000 psi so long as no recrystallization occurred either during rolling or testing.

With an appropriate change in the vertical scale, the plot of creep strength versus percent reduction (Fig. 6) would bear a close resemblance to the pattern of as-rolled hardness versus percent reduction (Fig. 7). It would, therefore, be expected that a correlation could be found between creep strength at 800°F and as-rolled hardness. As will be shown in the next section, three separate correlations were found.

Relation of Rupture and Creep Properties to Hardness

Hardness values for the as-rolled samples (Fig. 7) showed that hardness increased with amount of reduction over the whole range of reductions and temperatures of rolling except 1800°F. For isothermal rolling at 1800°F, the hardness did not increase for reductions beyond about 10 percent, apparently because recrystallization during rolling limited the hardness. Rolling over a falling temperature (dashed curves, Fig. 7) resulted in a higher hardness for a given reduction than for isothermal reduction. The first 10-percent reduction resulted in considerably more increase in hardness than subsequent equivalent reductions. Furthermore, the rate of increase in hardness for reductions of more than 10 percent was the same for isothermal reductions at 1400° and 1600°F. The rate of increase was also the same for all four temperatures of rolling with a falling temperature for reductions larger than approximately 15 percent.

The rupture times at 1100°F and 20,000 psi were correlated with the as-rolled hardness (Fig. 8) to obtain insight into the factors of working which govern properties. Figure 8 may be interpreted as follows:

1. The relationship between rupture time and hardness varies with the temperature of rolling. The characteristics of Figure 3 and 7 indicate that separate correlation curves could have resulted from the difference in hardness resulting from the first 10 percent of reduction. The correlations are nearly parallel for larger reductions because the hardness tended to increase at a rate which was nearly independent of the rolling temperature. Figure 3 indicates that the amount of reduction was the controlling factor and not the temperature of reduction. The differences between hardness change and rupture time change with amount of reduction, therefore, might explain the separate correlations of Figure 8.

2. There were very wide differences in rupture time for a given hardness, as is emphasized by the creep curves of Figures 9 and 10, for tests on materials with equal hardness. This illustrates two important factors:

(a) If hardness measures the initial strain hardening, some other factor than strain hardening is a controlling factor. Figure 3 indicates that it is mainly related to the degree of reduction. It has not yet been established if the pronounced changes in rupture time with hardness for a given temperature of rolling indicates that rupture time is either highly dependent on initial hardness or that some other factor not measured by hardness influenced rupture time to a marked extent.

(b) The correlations amply illustrate that initial hardness cannot be relied upon to indicate rupture strength at 1100°F for the as-worked condition of "A" Nickel.

3. The correlations of Figure 8 indicate that finishing temperature was not important and that from the viewpoint of correlation with hardness the initial temperature of rolling controlled. This is not, however, established beyond question by the available data. The points on Figure 8 for rolling over a falling temperature tend towards higher hardness for their rupture time than the correlations indicate. This may, therefore, represent an effect of finishing temperature. A single solid curve has been used to correlate data for material rolled from 1600° or 1800°F. The data for isothermal rolling at 1800°F may actually be a separate curve and the points for rolling on a falling temperature from 1800°F may plot on the curve for 1600°F data as a result of the lower finishing temperature. At least it does not seem possible to explain the high strengths at higher hardness levels obtained on rolling with a falling temperature from hardness values.

4. Those specimens rolled at 80°F and from 1400°F on a falling temperature to the higher hardness levels recrystallized during testing and did not fall on the correlations of Figure 8. It may well be that when the other specimens are examined in more detail, it will be found that recrystallization may have been contributing more to the lower rupture life than the correlations indicate. In that event it would point to an independence of rupture time (in the absence of recrystallization during testing) as Figure 8 suggests.

5. The overall indications of Figures 3, 7, and 8 are that other factors than strain hardening are the predominant factors by which working conditions influence rupture time at 1100°F and 20,000 psi. Figure 3 indicates that it is most closely allied to degree of reduction (in the absence of recrystallization during working or testing) than temperature of working or amount of hardening from working.

The correlation between initial hardness and minimum creep rate at 1100°F and 11,000 psi (Fig. 11) did not separate the samples rolled at different temperatures as well as it did for rupture time at 20,000 psi (Fig. 8). Those samples rolled from 1800° or 1600°F, however, generally had the highest creep resistance and those rolled from 1400° or 80°F, the lowest creep resistance for a given hardness. Again there were remarkably wide variations in strength for a given hardness. Also those samples which recrystallized extensively during testing were far off the correlation. Also those samples which were rolled to large reductions on a falling temperature tended to have faster creep rates for a given hardness than thos which were rolled at a more nearly constant temperature. While Figure 11 suggests more dependence of creep strength on hardness than was indicated by Figure 8 for rupture time, it is probable that there was really little difference. More data better defining the effects of hardness as controlled by rolling temperature would probably show no more slope to the curves of Figure 11 than for Figure 8.

When the creep strengths at 800°F were correlated with as-rolled hardness (Fig. 12) the following features were evident:

1. The creep strength correlated with hardness on the basis of rolling temperature as did rupture strength at 1100°F (Figs. 8).

2. The creep strengths correlated on the basis of initial temperature of rolling for all tests whether rolled isothermally or on a falling temperature. Finishing temperature did not appear to influence the correlation with hardness. Because recrystallization did not occur during testing at 800°F and the strength did correlate with the hardness values for a given initial temperature of rolling, it is suggested that more detailed examination of the specimens tested at 1100°F will show that recrystallization rather than finishing temperature caused deviation from the correlation of Figure 8.

3. Because separate correlations of creep strength with hardness (Fig. 12) were obtained depending on the initial temperature of rolling, it seems evident that other factors in addition to strain hardening influenced strength.

Simply heating to the working temperature did not change strength or hardness. The unidentified factors influencing strength then must arise from the influence of temperature of heating in combination with reduction. This effect appears to be responsible for the greater increase in strength for a given hardness when the initial temperature of reduction was raised from 80° to 1400° to 1600°F.

The slopes of the curves of Figure 12 decreased with temperature of reduction. This indicates that the gain in strength from a given amount of hardening was less the lower the temperature of working. This appears to be due to an increase in strength with amount of reduction over and above that due to hardness since a larger reduction was required to attain a given hardness the higher the rolling temperature. At 1100°F the amount of reduction was the controlling factor.

Relation of Microstructure to Rupture and Creep Properties

The photomicrographs shown in Figures 13 through 18 were taken of samples which were electropolished in a solution of 10 percent perchloric acid plus 90 percent glacial acetic acid and etched electrolytically in a 40 percent aqueous solution of ortho-phosphoric acid.

The major microstructural features observed were:

1. The grain size did not vary appreciably as a result of heating for working without reduction (Figs. 13a, 14a, 15a, and 16a).

2. Considerable precipitation apparently occurred during cooling after heating without reduction. It outlined the grain boundaries and appeared within the grains in small angular outlines. The exact nature of this precipitate has not been established. "A" Nickel has several contaminating elements present which would be expected to form precipitates. In addition, the precipitate may form within the grains at locations which may be related to the substructures to be described later resulting in an apparent exaggeration of the amount of precipitate.

3. Reductions of about 20 percent at 80°, 1400° or 1600°F (Figs. 13b, 14b and 15b) noticeably distorted the grains and made the grain boundaries irregular. Partial recrystallization occurred on working at 1800°F (Fig. 16b). The amount of angular precipitate was reduced by rolling 20 percent at the three elevated temperatures but not at room temperature (Figs. 13b, 14b, 15b, and 16b).

4. Approximately 60 percent reduction over a falling temperature range exaggerated the microstructural changes described under Item 3 (Figs. 13d, 14c, 15c, and 16c). Even the recrystallized grains were distorted in the material reduced from 1800°F. The materials rolled 60 percent at 80°F and from 1400°, 1600°, or 1800°F showed the more usual spherical form for the precipitates.

5. A nearly isothermal reduction of about 30 percent at 80°, 1400°, or 1600°F (Figs. 13c, 14d, and 15d) resulted in extensive grain distortion, with an apparent increase in the amount of precipitate after reduction at 1600° and 1800°F. The grains were nearly completely recrystallized (Fig. 16d) as a result of reduction at 1800°F and the etching effects accompanying the precipitate resembled the unworked condition. Distorted grain boundaries were observed at all temperatures of working.

6. Some of the conditions of working resulted in material which showed extensive substructures, as the examples of Figures 17 and 18 show. Two problems have been found to complicate the determination of substructures. After some conditions of working, the etching procedures used to date fail to reveal a substructure, although it seems certain that it exists. In those cases where substructures are revealed there is a wide variation in its appearance from grain to grain (Fig. 18).

7. A number of specimens have been examined after testing. This part of the study is still incomplete and is continuing. It has previously been noted that the two largest reductions at 80°F and the largest reduction on a falling temperature from 1400°F resulted in materials which underwent extensive recrystallization during testing at 1100°F but not at 800°F. It is suspected that some of the other conditions of working will show partial recrystallization as a result of testing at 1100°F.

The relationships of properties to the observed microstructural conditions have not yet been sufficiently well evaluated to allow the establishment of definite conclusions. Only the following statements can be made:

1. There seems to be no relationship between as-rolled grain size and properties. The amount of recrystallization from rolling at 1800°F did not appear to correlate with properties.

2. It is believed that a proper evaluation of substructures will be closely related to properties. With the techniques used to date it has not, however, been possible to properly evaluate the substructures to check this hypothesis or even obtain a qualitative trend.

3. Grain distortion observed as a result of working has been consistent with the temperature and degree of reduction and with the hardness values. There does not seem to be any relation of this feature of the microstructure of any more significance than the hardness values.

4. The role of the precipitates has not been clarified. As yet no significant correlation has been found.

General Discussion

The research completed to date has fairly well outlined the relationship of creep and rupture properties to rolling conditions. Future research will be aimed at establishing the basic mechanisms involved with emphasis on the evaluation of substructure effects. In addition, it seems necessary to study the structural changes which occur during heating for testing as well as those which occur while the specimens are exposed to creep.

It is difficult to anticipate the exact experimental steps which will be taken. Creep and rupture testing will be carried out only to fill in missing points which seem significant or to check various hypotheses that develop as the structural studies progress. As previously mentioned the main factor is believed to be substructures and the major emphasis will be placed in checking this hypothesis. This will be difficult, because it is already known that the techniques used by others to study substructures do not work well for the large reductions involved in this investigation. In addition, the role of the precipitates and the minor elements present in the material must be established.

Age-Hardenable, Austenitic Alloy (A-286)

The influence of working conditions on the creep and rupture properties of A-286 was measured for the case where a solution treatment at 1650°F plus aging at 1325°F was applied after working. The conditions of rolling and creep-rupture properties are detailed in Table II. The general effects of prior working conditions on the hardness and microstructure of heat treated A-286 are presented below.

Rupture Properties at 1200°F

The major variable influencing rupture properties at 1200°F and 65,000 psi (Figs. 19 and 20) was the temperature of heating for working. The major increase resulted from raising the temperature of working from 1700° to 1950°F. Raising the temperature to 2200°F resulted in slightly lowered strength except for reductions of more than 20 percent. The rupture time tended to fall off for reductions of more than 20 percent at 1950°F and for more than 10 percent at 1700°F. There was no significant difference between samples rolled isothermally and those rolled with the temperature falling. It should be recognized that in those cases where the properties did vary with percent reduction, the variation was small in comparison to the effect of heating alone, except possibly for rolling from 1700°F.

Elongation and reduction of area (Fig. 20) varied opposite to the rupture times. The samples rolled at 2200°F had the lowest ductility. The ductilities of the samples rolled at 1700° or 1950°F tended to overlap. The reduction of area for the samples rolled at 80°F tended to be high.

For the material investigated, the major variable was the temperature of heating for rolling, or at least increasing the rolling temperature from 1700° to 1950°F. Degree of reduction was secondary to the temperature effect. Apparently the temperature of heating for working was more influential than the finishing temperature.

Rupture Properties at 1350°F

The results for rupture tests at 1350°F and 40,000 psi (Figs. 21 and 22) were generally similar to those at 1200°F except that the maximum rupture time was obtained for rolling from 2200°F. Reductions up to 10 percent at 1950° and 2000°F reduced rupture time. However, continued reduction from 2200°F again raised strength. Again there was little if any significant difference from isothermal and falling-temperature reduction.

Ductility values were more dependent on temperature of rolling (Fig. 22) than at 1200°F. The ductilities generally decreased with temperatures of rolling except that the larger reductions at 80°, 1700°, and 1950°F all gave similar and high ductility.

As at 1200°F the major variable was the temperature of heating for rolling with relatively minor effects from degree of reduction. At 1350°F the strength continued to increase with temperature up to 2200°F, whereas it was lower for 2200°F than for 1950°F when tested at 1200°F.

Creep Resistance at 1200°F and 35,000 psi

The strain rate for evaluating the effects of rolling was reduced by lowering the stress to 35,000 psi at 1200°F and measuring the creep characteristics. The results of these tests are shown as the creep curves of Figures 23 through 30. Actual curves are reported because decreases in volume and continuously increasing creep rates with time in most of the tests made it unfeasible to report minimum creep rates. Only a few of the conditions of rolling exhibited first and second stages of creep.

Qualitatively the results were similar to those obtained from the rupture tests. The effect of amount of reduction (Figs. 23 through 26) was small in comparison to the effect of temperature of rolling (Figs. 27 through 30). When rolled at room temperature all of the creep curves (Fig. 23) were within the expected degree of reproducibility. At 1700°F a reduction of 8.4 percent (Fig. 24) gave the highest strength as it did in the rupture tests. The creep tests after rolling at 1950°F (Fig. 25) were also qualitative the same as the rupture tests and in addition showed a decrease in length (negative creep) in the early portions of most of the tests. When rolled at 2200°F (Fig. 26) the influence of amount of reduction, was small and a number of the specimens underwent negative creep.

The creep curves for specimens heated to the rolling temperatures without reduction (Fig. 27) show the pronounced effect of temperature of heating. As in the rupture tests the material heated to 1950°F was slightly stronger than heated to 2200°F. These effects were maintained for reductions of about 9.4 and 20 percent (Figs. 28, 29 and 30). Furthermore, these figures suggest that the negative creep was associated with reduction at 1950° and 2200°F with the intermediate reductions being most effective.

In evaluating Figures 23 through 30 attention should be given to the scales used for showing strain. For instance, the strain scale for Figure 26 is 10 times as sensitive as for Figure 23. Thus, the variation for conditions of rolling at 2200°F was considerably less than for rolling at room temperature. Similar variations occur in the other figures. Furthermore, only creep strain is shown on the figures since no plastic deformation occurred on loading.

Relation of Properties to Hardness

The hardness values after rolling and heat treating (Fig. 31) gave an overall range of 284 to 335 DPH with most values being between 300 and 320 DPH. The peak values for material rolled 10 percent at 1950° or 2200°F seem significant as does the high values for material rolled on a falling temperature from 2200°F. No relation between hardness values and creep and rupture properties at 1200° or 1350°F was evident.

Relation of Properties to Microstructure

The photomicrographs presented in Figures 32, 33, and 34 were taken of samples which were electropolished in a solution of 90 percent glacial acetic acid plus 10 percent perchloric acid and etched electrolytically in a 5 percent aqueous solution of sulphuric acid.

The grain size of the original stock after solution treatment at 1650°F and aging (Fig. 32a) was very fine (ASTM grain size <8). Heating to the working temperatures without reduction resulted in the grain sizes shown by Figures 32b, 32c and 32d. These were rated as follows:

1700°F	ASTM grain size 8
1950°F	ASTM grain size 4-6
2200°F	ASTM grain size 0-2

The rupture and creep tests at 1200°F previously presented showed that there was some increase in strength from heating to 1700°F; a large increase from heating to 1950°F; and heating to 2200°F resulted in a somewhat lower strength than for 1950°F. When rupture tested at 1350°F, the material heated to 2200°F was slightly the strongest. Research carried out under Contract AF 33(616)-3380 indicated that solution treatment at 2100°F actually gives maximum rupture strength at 1200°F and 65,000 psi. These results indicate, therefore, that the strength did not increase regularly with grain size. If grain size in itself was a major factor, a very doubtful possibility, the optimum size apparently would be close to ASTM 5.

The microstructures after a reduction of about 20 percent at all four temperatures considered and solution treatment at 1650°F and aging (Fig. 34) show the following:

1. The grain sizes after heat treatment were slightly larger after reduction at 80° or 1700°F (Fig. 34a and 34b) than when no reduction was involved. Considerable intergranular precipitation was present. The very slight change in grain structure seems consistent with the observed absence of change in properties.

2. When reduced about 20 percent at 1950°F, the grains were not changed in size (Fig. 34c) but did appear to be deformed. Precipitation had occurred at the grain boundaries and along twin boundaries. The rupture and creep data indicate that these changes in microstructure did not alter the properties at 1200°F over simply heating alone. At 1350°F the rupture strength was somewhat lower.

3. When reduced 20 percent at 2200°F and heat treated there were fine recrystallized grains along the grain boundaries of the large grains (Fig. 34d) established by heating to 2200°F. There seemed to be considerable precipitation at the boundaries of the fine recrystallized grains and no precipitation was evident in their interior. The larger unrecrystallized grains appeared to contain a considerable amount of fine precipitates. Again these changes in microstructure were not accompanied by an appreciable change in creep and rupture properties at 1200° or 1350°F.

The reductions of about 20 percent at the four temperatures did induce changes in microstructure without appreciable changes in properties.

The microstructures after various reductions at 2200°F followed by solution treatment at 1650°F and aging (Fig. 33) show the recrystallized grains at the boundaries of the larger grains established by heating to 2200°F. The distortion of the larger grains is also shown. An isothermal reduction of 35.1 percent from 2200°F resulted in about 20 percent recrystallization. The precipitation within the large grains may have been intensified by increasing reduction. The fine recrystallized grains were outlined by precipitation at the boundaries. The changes in creep and rupture properties as a result of this working were hardly significant.

General Discussion

The data obtained to date indicate that the major factor influencing the properties of the A-286 studied was the temperature of heating for working. The amount of reduction either had no effect or was secondary in magnitude compared to the effect of heating. The generality of this observation is open to considerable question. The stock used had a very fine grain size when heat treated at 1650°F. There is considerable doubt that the temperature of heating would have been so influential if the original stock had been produced under conditions which would have resulted in a larger grain size when heat treated at 1650°F.

The study of the relationships between structure and properties is very incomplete at this stage of the investigation. So far it has not been possible to correlate properties with the observed microstructures. Grain size in itself does not seem to be the answer. The development of fine grains along the grain boundaries of coarse grains resulting from reductions at a high temperature did not appear to appreciably influence properties.

Further work will initially be concentrated on evaluation of the $Ni_3(Al, Ti)$ precipitate as influenced by heating and working. This study will also evaluate the influence of the carbide type precipitates. Additional experiments involving the variation of the structure before working will also probably be necessary. The general objective will be to determine if temperatures of heating rather than degree of reduction are really the controlling variable which causes wide variations in the properties of A-286 alloy when given a standard heat treatment.

One feature of the indications that heating temperature for working is the major variable influencing properties arises for the reported processing of the stock by the Allegheny Ludlum Steel Corporation. As was described in the section on TEST MATERIALS, the stock was finish rolled from 2100°F. According to the results of these experiments this should have given high strength. Some other factor which has not been included in the investigation to date must have been involved. The original bars were reduced from 2-7/8 inch square to 3/4 inch round, a reduction of about 95 percent. This is a much larger reduction than was used in the experiment. Both Figures 19 and 21 suggest that strengths do fall off with the larger reductions from 1950°F, but not from 2200°F. It may be that when the material is reduced from 2100°F a similar decrease in strength occurs. Certainly the material recrystallized during working or during heat treatment to a very fine grain size. In any event, the low initial strength has not yet been explained in view of the temperature of heating for rolling of bars and should be considered as a further restriction on the generality of the results of the experiments which suggested that heating temperatures controlled the strength of A-286 alloy.

CONCLUSIONS

The progress attained to date in the study of the relationship between properties at high temperatures and conditions of hot working is still too incomplete to permit many conclusions. The relationships between properties and conditions of working have been outlined for a simple microstructure material in the as-worked condition using "A" Nickel as the experimental material. Similar data have been established for the Ti + Al strengthened alloy A-286 when solution treated at 1650°F and aged after working. These together with partial structural studies indicate the future research to be carried out.

"A" Nickel

The results obtained for "A" Nickel indicate that unidentified factors over and above strain hardening govern properties at 1100°F and are about as influential as strain hardening at 800°F. In general, the properties at 1100°F appeared to be mainly a function of degree of reduction independent of the initial or final temperatures of working. Recrystallization during rolling or during testing were the two limitations on this generality. At 800°F, the degree of reduction was still an important factor although the increased influence of strain hardening made properties more dependent on the temperature of working. The properties at both temperatures increased markedly for a given as-worked hardness with temperature of working up to 1600°F. There appeared to be little difference in properties for working at 1600° or 1800°F on the basis of hardness. Higher strengths at higher hardness levels for larger reductions were developed by rolling on a falling temperature from 1800° or 1600°F than could be obtained by nearly isothermal reductions. No direct relationship to grain size has been observed. The investigation is being continued to determine if the major unidentified factor governing properties is the influence of working conditions on substructures.

A-286 Alloy

The results for the A-286 alloy indicated that the major factor was the temperature of heating for working and not the degree of reduction or the finishing temperature. The generality of this conclusion is open to doubt because it may have been due mainly to the prior history of the stock resulting in initially low strength and very fine grain size when heat treated at 1650°F. In the experiments, working at 1950°F gave maximum or near maximum strengths. Working at 80° or 1700°F did not raise strengths much above the low strength of the original stock. Compared to strengths obtained on working at 1950°F, working at 2200°F resulted in slightly lower strength at 1200°F and slightly higher strength at 1350°F.

Research carried out under another Air Force contract on the same A-286 stock indicated that the maximum strength was associated with heating to about 2100°F. Therefore, the results for working at 1950°F probably do not give true maximum strengths. No correlations between grain size and other microstructural characteristics have developed from the very incomplete

structural studies. The investigation is being continued with emphasis on determining if the major factor in the hot working of A-286 which influences properties is the temperature of heating, and determining the microstructural effects which induce the observed effects.

REFERENCES

1. J. F. Ewing and J. W. Freeman, "Influence of Hot-Working Conditions on High-Temperature Properties of a Heat-Resistant Alloy," NACA Technical Note 3727, August, 1956.

2. D. N. Frey and J. W. Freeman, "Fundamental Effects of Cold Working on the Creep Resistance of an Austenitic Alloy," Trans. AIME, 191 (1951) 755-760.

3. H. Zschokke, "Improvement of Creep Resistance by Cold Work," The Engineers' Digest (American Edition), 4 (1947) 82-85.

4. O. D. Sherby and J. E. Dorn, "The Effect of Cold Rolling on the Creep Properties of Several Aluminum Alloys," Trans. ASM, 43 (1951) 611-634.

5. G. T. Harris and W. H. Bailey, "Effect of Warm-Working on an Austenitic Steel (G 18 B)," Symposium on High-Temperature Steels and Alloys for Gas Turbines, The Iron and Steel Institute, London, July, 1952, pp. 60-67.

6. N. J. Grant and A. G. Bucklin, "Creep-Rupture and Recrystallization of Monel from 700° to 1700°F," Trans. ASM, 45 (1953) 151.

7. N. J. Grant and A. G. Bucklin, "Creep-Rupture Properties of Cold-Worked Type 347 Stainless Steel," Trans. ASM, 45 (1953) 151-170.

8. P. Shahinian, "Influence of Cold Work on Strength of Steel at Elevated Temperatures," Trans. ASM, 48 (1956) 952-967.

9. N. J. Grant, A. G. Bucklin and W. Rowland, "Creep-Rupture Properties of Cold-Worked Type 347 Stainless Steel," Trans. ASM, 48 (1956) 446.

10. R. A. Lula, A. J. Lena, and H. M. Johnson, "Effect of Cold Work on Elevated-Temperature Properties of Types 381, 305, and 310 Stainless Steel," Trans. ASME 79 (1957) 921-926.

11. F. B. Cuff, Jr. and N. J. Grant, "Effect of Cold Work on Creep-Rupture Properties of Simple 18-8 Stainless Steels," J. Iron and Steel Inst., 186 (1957) 188.

12. T. H. Hazlett and R. D. Hansen, "Influence of Substructure on the Shape of the Creep Curve," Trans. ASM, 47 (1955) 508-519.

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

TABLE I

ROLLING CONDITIONS AND RUPTURE, CREEP, AND TOTAL DEFORMATION DATA FOR "A" NICKEL ROLLED ISOTHERMALLY AND NON-ISOTHERMALLY WITH NO REHEATS

Rolling Conditions			Rupture, Creep, and Total Deformation Data											
										Time to Reach Indicated Total Deformation (hrs)				
Temperature (°F)[a]	Reduction of Area (%)	Number of Passes	Temperature (°F)	Stress (psi)	Time to Rupture (hrs)	Elongation (% in 4D)	Reduction of Area (%)	Deformation on Loading (%)	Minimum Creep Rate (% per hr)	0.2%	0.5%	1.0%	2.0%	5.0%
80	0.0	0	800	35,000	> 525.6	--	--	>5.00	0.00120	b	b	b	b	b
			800	32,000	> 525.4	--	--	>5.00	0.00106	b	b	b	b	b
			800	20,000	> 982.0	--	--	1.70	0.000420	b	b	b	15.0	>1000.0
			800	14,000	>1269.7	--	--	0.383	0.000265	b	< 10.0	1500.0 g	--	--
			1100	20,000	25.1	61.6	74.8	3.46	2.34	b	b	b	b	< 1.0
			1100	17,000	102.1	51.8	52.8	1.92	d	b	b	b	d	d
			1100	14,000	359.0	29.1	30.7	0.908	0.0587	b	b	< 1.0	~2.0	23.0
			1100	11,000	2309.7	e	28.8	0.240	0.00827	b	< 1.0	12.0	50.0	240.0
			1100	8,000	>1079.6	--	--	0.110	0.00225	7.0	60.0	190.0	570.0	>1000.0
			1200	15,000	10.2	81.5	67.3	1.25	d	b	b	b	d	d
	11.3	4	800	30,000	>1289.5	--	--	0.414	0.000072	b	95.0	>>1000.0	--	--
			1100	20,000	187.9	23.6	31.0	0.117	0.00700	~2.0	32.0	71.0	107.0	152.0
			1100	11,000	>1348.7	--	--	0.055	0.000118	140.0	>1000.0	--	--	--
	16.5	6	800	34,000	> 986.2	--	--	0.184	0.000088	< 5.0	>>1000.0	--	--	--
			1100	20,000	269.9	27.8	31.5	0.112	0.00735	~2.0	33.0	85.0	145.0	219.0
			1100	11,000	>1153.0	--	--	0.054	0.000117	215.0	>1000.0	--	--	--
	20.5	6	800	33,000	>1147.7	--	--	0.195	0.000043	< 1.0	>>1000.0	--	--	--
			1100	20,000	470.5	20.9	29.8	0.141	0.00512	10.0	62.0	136.0	250.0	392.0
			1100	11,000	>1296.0	--	--	0.035	0.000063	650.0	>>1000.0	--	--	--
	35.5	8	800											
			1100	20,000	442.1	30.0	42.1	0.100	0.00624	9.0	50.0	130.0	223.0	329.0
			1100	11,000	> 863.0	--	--	0.057	0.000960	60.0	330.0	550.0	745.0	--
	61.2	12	800	50,000	>1100.0	--	--	0.274	0.00011	b	50.0	>>1000.0	--	--
			1100	20,000	56.3	53.7	86.5	0.109	0.0628 d	~1.0	6.0	12.0	18.0	26.0
			1100	11,000	> 428.8	--	--	0.049	0.0297	12.0	27.0	50.0	80.0	170.0
1400	4.7	2	800	20,000	>1171.2	--	--	0.185	0.000076	10.0	>>1000.0	--	--	--
			1100	20,000	43.9	40.0	58.1	0.171	0.40 d	< 1.0	< 1.0	2.0	4.5	--
			1100	11,000	> 861.1	--	--	0.081	0.000680	55.0	300.0	880.0 g	--	--
	9.7	2	800	20,000	> 934.5	--	--	0.105	0.000022	>1000.0	--	--	--	--
			1100	20,000	110.9	41.8	54.5	0.172	0.0710	< 1.0	4.0	11.0	21.0	45.0
			1100	11,000	> 838.3	--	--	0.059	0.000315	120.0	780.0	>1000.0	--	--
	14.7	2	800	29,000	>1180.2	--	--	0.152	0.00019	40.0	~2000.0 g	--	--	--
			1100	20,000	240.0	30.0	40.8	0.104	0.0119	~1.0	25.0	54.0	90.0	150.0
			1100	11,000	> 742.0	--	--	0.081	0.000277	90.0	1000.0 g	--	--	--
	26.1	2	800	32,000	>1247.3	--	--	0.188	0.00011	< 5.0	~2000.0 g	--	--	--
			1100	20,000	335.4	20.0	33.0	0.144	0.0114	~1.0	14.0	60.0	120.0	237.0
			1100	11,000	>1108.8	--	--	0.077	0.00022	20.0	750.0	>1000.0	--	--
	36.1	2	800	34,000	>1226.6	--	--	0.157	0.000077	48.0	>>1000.0	--	--	--
			1100	20,000	627.2	30.9	56.1	0.076	0.00431	10.0	80.0	185.0	245.0	550.0
			1100	11,000	> 886.8	--	--	0.047	0.000448	105.0	710.0	~2000.0 g	--	--
	19.5	3	800	31,000	>1146.8	--	--	0.233	0.000183	b	790.0	--	--	--
			1100	20,000	265.0	30.0	40.8	0.142	0.0127	2.0	24.0	56.0	97.0	162.0
			1100	11,000	>1227.9	--	--	0.091	0.000056	170.0	>1000.0	--	--	--
	33.8	4	800	40,000	>1304.3	--	--	0.216	0.000144	b	640.0	>>1000.0	--	--
			1100	20,000	647.8	22.7	41.7	0.114	0.0093	< 1.0	17.0	60.0	143.0	340.0
			1100	11,000	>1136.9	--	--	0.072	0.000148	70.0	1000.0	>>1000.0	--	--
	58.7	6	800	49,000	>1228.1	--	--	0.277	0.00017	b	50.0	>>1000.0	--	--
			1100	20,000	379.8	36.4	76.4	0.129	0.00352	~3.0	68.0	152.0	233.0	324.0
			1100	11,000	> 676.2	--	--	0.089	0.000726	45.0	380.0	600.0	--	--
1600	0.0	0	800	16,000	> 721.4	--	--	0.708	0.000294	b	b	600.0	>1000.0	--
			1100	20,000	30.5±0.5	46.4	53.0	3.47	0.480	b	b	b	b	< 1.0
			1100	11,000	> 982.0	--	--	0.436	0.00845	b	< 1.0	8.0	45.0	235.0
	5.3	2	800	20,000	> 937.3	--	--	0.174	0.000178	10.0	~1300.0 g	--	--	--
			1100	20,000	71.3	40.7	49.5	3.90	0.218	b	b	b	b	~1.0
			1100	11,000	> 863.5	--	--	0.089	0.000990	65.0	335.0	830.0	>1000.0	--
	9.9	2	800	24,000	> 725.1	--	--	0.396	0.000195	b	410.0	>1000.0	--	--
			1100	20,000	125.2	38.2	51.0	0.108	0.0735	< 1.0	~5.0	11.0	22.0	56.0
			1100	11,000	>1084.1	--	--	0.061	0.000250	220.0	1100.0 g	--	--	--
	15.2	2	800	26,000	> 654.0	--	--	0.325	0.00026	b	440.0	>>1000.0	--	--
			800	24,000	> 868.7	--	--	0.161	0.00013	10.0	>>1000.0	--	--	--
			800	22,000	>1197.4	--	--	0.179	0.000080	10.0	>>1000.0	--	--	--
			1100	20,000	142.3	38.2	52.0	0.148	0.0487	<1.0	~3.0	10.0	28.0	67.0
			1100	20,000	153.2	27.8	47.9	0.113	0.0350	~3.0	11.0	25.0	47.0	92.0
			1100	20,000	184.4	30.0	44.2	0.154	0.0242	<1.0	11.5	30.0	62.0	120.0
			1100	20,000	199.1	26.4	42.6	0.122	0.0224	~2.0	15.0	36.0	69.0	134.0
			1100	20,000	177.5	36.4	49.2	0.170	0.0304	<1.0	9.0	25.0	49.0	99.0
			1100	11,000	>1368.7	--	--	0.061	0.00027	160.0	740.0	~2000.0 g	--	--

TABLE I (concluded)

ROLLING CONDITIONS AND RUPTURE, CREEP, AND TOTAL DEFORMATION DATA FOR "A" NICKEL ROLLED ISOTHERMALLY AND NON-ISOTHERMALLY WITH NO REHEATS

Rolling Conditions			Rupture, Creep, and Total Deformation							Time to Reach Indicated Total Deformation (hrs)				
Temperature (°F)[a]	Reduction of Area (%)	Number of Passes	Temperature (°F)	Stress (psi)	Time to Rupture (hrs)	Elongation (% in 4D)	Reduction of Area (%)	Deformation on Loading (%)	Minimum Creep Rate (% per hr)	0.2%	0.5%	1.0%	2.0%	5.0%
1600	26.7	2	800	32,000	>1177.2	--	--	0.333	0.000242	b	450.0	>1000.0	--	--
			1100	20,000	521.8	29.1	35.2	0.096	0.00862	6.0	40.0	95.0	188.0	360.0
			1100	11,000	>1320.0	--	--	0.052	0.000227	105.0	1030.0	>2000.0	--	--
	36.8	2	800	32,000	>1246.5	--	--	0.251	0.00015	b	960.0	--	--	--
			1100	20,000	463.7	28.2	39.6	0.076	0.0088	4.0	30.0	85.0	175.0	329.0
			1100	11,000										
	20.3	3	800	28,000	>1036.1	--	--	0.205	0.000182	b	~1200.0 g	--	--	--
			1100	20,000	180.9	28.2	56.7	0.117	0.0248	~1.0	13.0	33.0	63.0	118.0
			1100	11,000	>1083.7	--	--	0.095	0.000231	30.0	740.0	>1000.0	--	--
	34.4	4	800	40,900	>1052.5	--	--	0.698	0.000415	b	b	205.0	>1000.0	--
			1100	20,000	663.6	27.3	42.0	0.106	0.00476	~3.0	50.0	130.0	255.0	480.0
			1100	11,000	>1203.6	--	--	0.048	0.000091	330.0	>1000.0	--	--	--
	59.1	6	800	61,900	0.4	36.5	72.9	1.58	d	b	b	b	< 0.1	--
			800	48,000	>1009.9	--	--	0.242	0.000165	b	460.0	>1000.0	--	--
			1100	20,000	1030.4	16.4	53.6	0.088	0.00179	15.0	145.0	420.0	720.0	960.0
			1100	11,000	>1057.8	--	--	0.060	0.000107	145.0	>1000.0	--	--	--
1800	0.0	0	800	15,000	> 886.0	--	--	0.832	0.000158	b	b	570.0	>>1000.0	--
			1100	20,000	19.5	50.9	62.9	4.50	1.30	b	b	b	b	0.03
			1100	11,000	>1173.5	--	--	0.109	0.0087	~1.0	10.0	25.0	70.0	280.0
	5.8	2	800	19,000	> 871.2	--	--	0.335	0.00011	b	510.0	>>1000.0	--	--
			1100	20,000	72.8	43.4	49.1	1.06	0.190	b	b	b	< 1.0	~4.0
			1100	11,000	> 702.0	--	--	0.082	0.00129	55.0	230.0	610.0	~1500.0 g	--
	10.8	2	800	19,000	>1197.4	--	--	0.286	0.000206	b	930.0	--	--	--
			1100	20,000	69.9	47.3	64.3	0.988	0.282	b	b	< 1.0	~1.0	7.0
			1100	11,000	> 702.0	--	--	0.078	0.00121	55.0	240.0	650.0	~1600.0 g	--
	15.7	2	800	19,700	>1267.3	--	--	0.284	0.000103	b	950.0	--	--	--
			1100	20,000	93.3	48.2	63.7	0.786	0.180	b	b	< 1.0	~1.0	8.0
			1100	11,000	> 843.0	--	--	0.073	0.00092	65.0	295.0	785.0	--	--
	27.3	2	800	18,000	> 844.9	--	--	0.233	0.00020	b	~1000.0 g	--	--	--
			1100	20,000	58.7	43.6	59.7	1.04	0.254	b	b	b	~1.0	7.0
			1100	11,000	> 988.2	--	--	0.084	0.000355	60.0	620.0	~2000.0 g	--	--
	37.5	2	800	19,000	> 856.4	--	--	0.395	0.000245	b	< 10.0	~1800.0 g	--	--
			1100	20,000	109.0±0.9	41.8	62.7	0.771	0.160	b	b	< 1.0	4.0	19.0
			1100	11,000	> 886.9	--	--	0.190	0.00306	~1.0	40.0	180.0	525.0	~1500.0 g
	20.7	3	800	19,000	> 973.2	--	--	0.212	0.000115	b	>>1000.0	--	--	--
			1100	20,000	75.2±1.5	53.6	63.9	0.438	0.0222	b	< 0.4	0.4	1.5	13.0
			1100	11,000	>1010.6	--	--	0.092	0.00049	35.0	460.0	~1500.0 g	--	--
	35.2	4	800	29,000	> 748.2	--	--	0.509	0.000303	b	b	~1400.0 g	--	--
			1100	20,000	214.3	27.3	48.6	0.145	0.0174	~2.0	15.0	38.0	73.0	132.0
			1100	11,000	> 701.4	--	--	0.087	0.000298	135.0	1050.0 g	--	--	--
	59.3	6	800	34,000	>1056.8	--	--	0.176	0.000075	~10.0	>>1000.0	--	--	--
			1100	20,000	1028.3	e	33.4	0.137	0.00208	~5.0	115.0	350.0	660.0	935.0
			1100	11,000	> 917.8	--	--	0.078	0.000099	525.0	>1000.0	--	--	--

> "Greater than." (Indicates in the "Time to Rupture" column the time at which the test was discontinued.)
>> "Much greater than."
< "Less than."
~ "Approximately."
a Initial temperature; no reheats were used.
b Indicated deformation exceeded upon application of load.
d Value unknown or uncertain because of insufficient time-elongation data.
e Calculation impossible because a piece of the gage section was lost.
g Value obtained by extrapolation or interpolation.

TABLE II

ROLLING CONDITIONS AND RUPTURE, CREEP, AND TOTAL DEFORMATION DATA FOR A-286 ALLOY ROLLED ISOTHERMALLY AND NON-ISOTHERMALLY WITH NO REHEATS

Rolling Conditions			Rupture, Creep, and Total Deformation Data							Time to Reach Indicated Total Deformation (hrs)				
Temperature (°F)[a]	Reduction of Area (%)	Number of Passes	Temperature (°F)	Stress (psi)	Time to Rupture (hrs)	Elongation (% in 4D)	Reduction of Area (%)	Deformation on Loading (%)	Minimum Creep Rate (% per hr)	0.2%	0.5%	1.0%	2.0%	5.0%
80	0.0	0	1200	65,000	19.1	13.6	17.6	0.421	0.860	b	0.8	6.0 g	--	--
			1200	45,000	399.9	11.8	11.0	0.229	0.00348	b	80.0	145.0	204.0	310.0
			1200	35,000	>1156.7	--	--	0.198	c					
			1350	40,000	14.1	49.1	48.4	0.285	0.680	b	0.3	1.0	2.1	5.0
	9.4	3	1200	65,000	27.1	10.9	15.2	0.346	0.055	b	2.9	8.5	13.0 g	--
			1200	45,000	463.8	10.9	11.0	0.272	0.00164	b	96.0	196.0	263.0	374.0
			1200	35,000	>453.6	--	--	0.254	c					
			1350	40,000	17.2	41.8	52.0	0.268	d	b	d	d	d	d
	19.7	5	1200	65,000	29.9	9.1	13.8	0.348	0.0264	b	4.0	9.5	15.0	25.0
			1200	45,000	413.2	10.0	11.0	0.232	0.0022	b	95.0	171.0	236.0	339.0
			1200	35,000	>597.5	--	--	0.186	c					
			1350	40,000	16.3±3.0	52.7	49.1	0.328	0.875	b	0.3	0.8	2.0	4.7
	28.8	7	1200	65,000	31.0	13.6	16.0	0.352	0.0455	b	3.0	9.0	14.0	23.0
			1200	45,000	397.6	13.6	15.2	0.215	0.00296	b	80.0	142.0	204.0	305.0 g
			1200	35,000	>891.2	--	--	0.227	c					
			1350	40,000	19.9	49.1	52.5	0.278	d	b	d	d	d	d
	39.5	9	1200	65,000	20.7	10.0	13.8	0.318	d	b	d	d	d	d
			1200	45,000	420.7	11.0	11.4	0.256	0.00280	b	71.0	145.0	216.0	334.0
			1200	35,000	>1504.7	--	--	0.270	c					
			1350	40,000	15.2	47.3	46.0	0.259	0.736	b	0.5	1.1	2.5	5.5
1700	0.0	0	1200	65,000	30.5	10.0	9.8	0.295	0.720	b	3.0	9.0	14.0	24.0
			1200	45,000	462.9	11.2	11.9	0.247	0.00420	b	48.0	142.0	220.0	340.0
			1200	35,000	>476.6	--	--	0.245	c					
			1350	40,000	18.9	39.1	47.5	0.236	0.356	b	0.4	2.0	4.5	8.5
	3.9	1	1200	65,000	38.5	16.3	13.3	0.371	0.0356	b	3.0	13.5	22.0	31.0
			1200	35,000	> 886.2	--	--	0.221	c					
			1350	40,000	21.0	37.2	43.9	0.262	0.228	b	1.0	2.7	4.8	9.6
	8.4	2	1200	65,000	43.9	6.4	6.4	0.316	0.186	b	6.5	14.0	23.0	--
			1200	35,000	>1396.4	--	--	0.207	c					
			1350	40,000	32.4	34.5	34.0	0.293	0.130	b	1.5	4.2	7.2	14.4
	21.5	2	1200	65,000	22.6	7.4	9.1	0.584	d	b	b	d	d	d
			1200	35,000	>601.6	--	--	0.187	c					
			1350	40,000	16.6	42.6	49.7	0.266	d	b	d	d	d	d
	31.7	2	1200	65,000	22.2	10.9	9.4	0.362	d	b	d	d	d	d
			1200	35,000	>625.1	--	--	0.214	c					
			1350	40,000	14.2	32.7	52.3	0.331	d	b	d	d	d	d
	11.5	3	1200	65,000	31.1	7.3	9.4	0.400	d	b	d	d	d	d
			1200	35,000	>840.5	--	--	0.198	c					
			1350	40,000	24.5	34.6	43.4	0.228	2.65	b	1.0	2.6	4.9	10.1
	17.7	3	1200	65,000	30.1	7.3	9.8	0.463	0.0472	b	0.4	9.0	14.5	26.3
			1200	35,000	> 887.2	--	--	0.110	c					
			1350	40,000	23.1	30.0	25.7	0.266	d	b	d	d	d	d
	37.6	5	1200	65,000	29.2	8.2	8.6	0.337	d	b	d	d	d	d
			1200	35,000	>912.8	--	--	0.190	c					
			1350	40,000	20.3	46.3	44.5	0.285	d	b	d	d	d	d

TABLE II(concluded)

ROLLING CONDITIONS AND RUPTURE, CREEP, AND TOTAL DEFORMATION DATA FOR A-286 ALLOY ROLLED ISOTHERMALLY AND NON-ISOTHERMALLY WITH NO REHEATS

Rolling Conditions			Rupture, Creep, and Total Deformation Data											
Temperature (°F)[a]	Reduction of Area (%)	Number of Passes	Temperature (°F)	Stress (psi)	Time to Rupture (hrs)	Elongation (% in 4D)	Reduction of Area (%)	Deformation on Loading (%)	Minimum Creep Rate (% per hr)	Time to Reach Indicated Total Deformation (hrs) 0.2%	0.5%	1.0%	2.0%	5.0%
1950	0.0	0	1200	65,000	127.2	6.4	4.8	0.323	0.00440	b	34.0	77.0	103.0	--
			1200	45,000	>1200.5	--	--	0.257	0.000068 e	b	1130.0	--	--	--
			1200	35,000	>1296.0	--	--	0.199	c					
			1350	40,000	102.2	18.2	22.6	0.233	0.00570	b	28.0	55.0	71.0	87.0
	3.9	1	1200	65,000	143.2	4.5	5.6	0.383	0.00250	b	47.0	96.0	118.0	--
			1200	45,000	>1203.2	--	--	0.212	0.000098	b	840.0	--	--	--
			1200	35,000	>647.7	--	--	0.186	c					
			1350	40,000	66.3	23.6	24.9	0.238	0.0154	b	3.5	26.0	36.0	--
	9.7	2	1200	65,000	127.4	6.2	8.3	0.360	0.0046	b	30.0	69.0	93.0	125.0 g
			1200	35,000	>1154.0	--	--	0.221	c					
			1350	40,000	48.9	20.0	26.2	0.223	0.0242	b	7.0	14.0	23.0	--
	18.0	2	1200	65,000	159.0	5.5	7.9	0.330	0.00170	b	60.0	73.0	117.0	165.0
			1200	35,000	>1395.2	--	--	0.197	c					
			1350	40,000	53.7	19.4	27.0	0.227	0.0508	b	5.0	14.0	23.0	34.0
	23.5	2	1200	65,000	53.9	7.3	14.9	0.448	0.0242	b	4.0	17.0	28.0	45.0
			1200	35,000	>1481.8	--	--	0.194	c					
			1350	40,000	42.1	32.7	37.2	0.246	d	b	d	d	d	d
	34.2	2	1200	65,000	78.5±2.0	9.1	12.8	0.382	0.0125	b	7.0	24.0	40.0	63.0
			1200	35,000	>1035.5	--	--	0.266	c					
			1350	40,000	42.2	30.9	45.6	0.229	0.118	b	2.0	6.0	13.0	25.0
	12.2	3	1200	65,000	100.2	7.3	7.9	0.350	0.00360	b	26.0	49.0	66.0	93.0 g
			1200	35,000	>960.5	--	--	0.228	c					
			1350	40,000	48.9	20.0	25.7	0.181	0.0428	~5.0	7.5	13.5	20.0	--
	18.8	3	1200	65,000	112.3	11.8	9.9	0.388	0.0041	b	21.0	53.0	73.0	--
			1200	35,000	>1464.0	--	--	0.251	c					
			1350	40,000	49.3	24.6	33.4	0.248	0.0708	b	2.5	9.5	18.0	31.0
	38.7	5	1200	65,000	62.1	10.0	8.6	0.390	0.0116	b	6.0	19.0	31.0	--
			1200	35,000	>914.6	--	--	0.241	c					
			1350	40,000	32.1	39.1	47.3	0.179	d	<1.0	d	d	d	d
2200	0.0	0	1200	65,000	118.7	2.7	3.2	0.426	0.00053	b	2.0	70.0	--	--
			1200	65,000	41.2	2.7	5.6	0.431	0.0246	b	0.4	14.4	--	--
			1200	35,000	>1318.3	--	--	0.416	c	b				
			1350	40,000	128.1	2.7	2.0	0.284	0.00266	b	70.0	107.0	--	--
	5.4	1	1200	65,000	80.7	2.7	4.8	0.398	0.00568	b	11.0	53.0	78.0 g	--
			1200	35,000	>1036.1	--	--	0.237	c					
			1350	40,000	72.4	8.0	10.3	0.170	0.0100 e	6.5	21.0	31.0	44.0	66.0
	10.4	2	1200	65,000	91.5	2.7	4.0	0.350	0.00383	b	31.0	61.0	--	--
			1200	35,000	>1386.0	--	--	0.173	c					
			1350	40,000	50.8	8.2	8.7	0.303	0.0244	b	7.5	17.0	26.5	43.0
	20.2	2	1200	65,000	97.0	3.6	7.1	0.382	0.0094	b	9.0	49.0	80.0	--
			1200	35,000	>1146.3	--	--	0.253	c					
			1350	40,000	78.4	10.9	14.4	0.252	0.0112	b	15.0	29.0	43.0	66.0
	26.1	2	1200	65,000	130.4	3.6	6.8	0.378	0.0025	b	39.0	85.0	118.0	--
			1200	35,000	>1536.0	--	--	0.252	c					
			1350	40,000	78.4	11.8	15.4	0.138	0.0072	10.0	23.0	37.0	56.0	--
	35.1	2	1200	65,000	121.9	4.5	8.6	0.341	0.00048	b	34.0	70.0	92.0	--
			1200	35,000	>1536.0	--	--	0.215	c					
			1350	40,000	66.1	13.6	16.7	0.247	0.0215	b	10.0	22.0	35.0	53.0
	14.5	3	1200	65,000	125.8	4.5	4.8	0.429	0.00207	b	29.0	84.0	113.0	--
			1200	35,000	>1171.0	--	--	0.250	c					
			1350	40,000	62.2	7.3	9.8	0.217	0.0180	b	11.0	20.0	31.0	54.0
	20.8	3	1200	65,000	132.2	4.5	6.0	d	0.00210	b	d	d	d	--
			1200	35,000	>1296.0	--	--	0.260	c					
			1350	40,000	66.6	9.1	12.7	0.283	0.0130	b	13.0	25.0	36.0	56.0
	39.8	5	1200	65,000	186.2	4.5	5.2	0.378	0.00178 e	b	67.0	108.0	148.0	--
			1200	35,000	>1564.6	--	--	0.202	c					
			1350	40,000	94.4	f	11.4	0.240	d	b	23.0	38.0	59.0	85.0

> "Greater than." (Indicates in the "Time to Rupture" column the time at which the test was discontinued.)
>> "Much greater than."
< "Less than."
~ "Approximately."
a Initial temperature; no reheats were used.
b Indicated deformation exceeded upon application of load.
c Refer to Figs. 23 through 30 where the entire creep curves are compared.
d Value unknown or uncertain because of either insufficient or faulty data.
e This rate of creep was preceded by an initial, brief period of either zero or negative creep.
f Calculation impossible because a piece of the gage section was lost.
g Value obtained by extrapolation or interpolation.

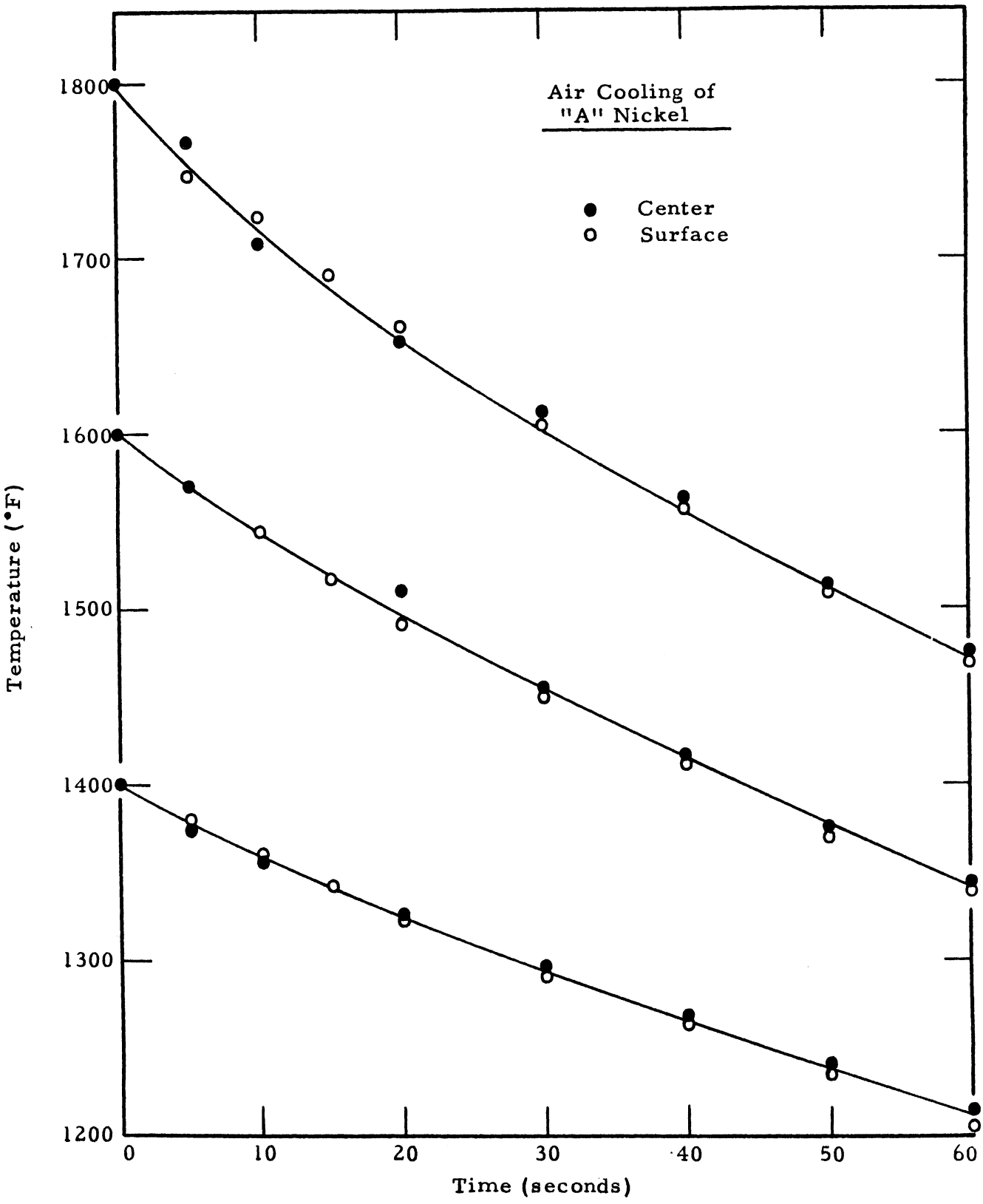


Figure 1 . - Cooling Curves for 7/8-Inch Square Bars of "A" Nickel Air Cooled from 1400°, 1600°, and 1800°F. The bars were held at temperature for 45 minutes before cooling.

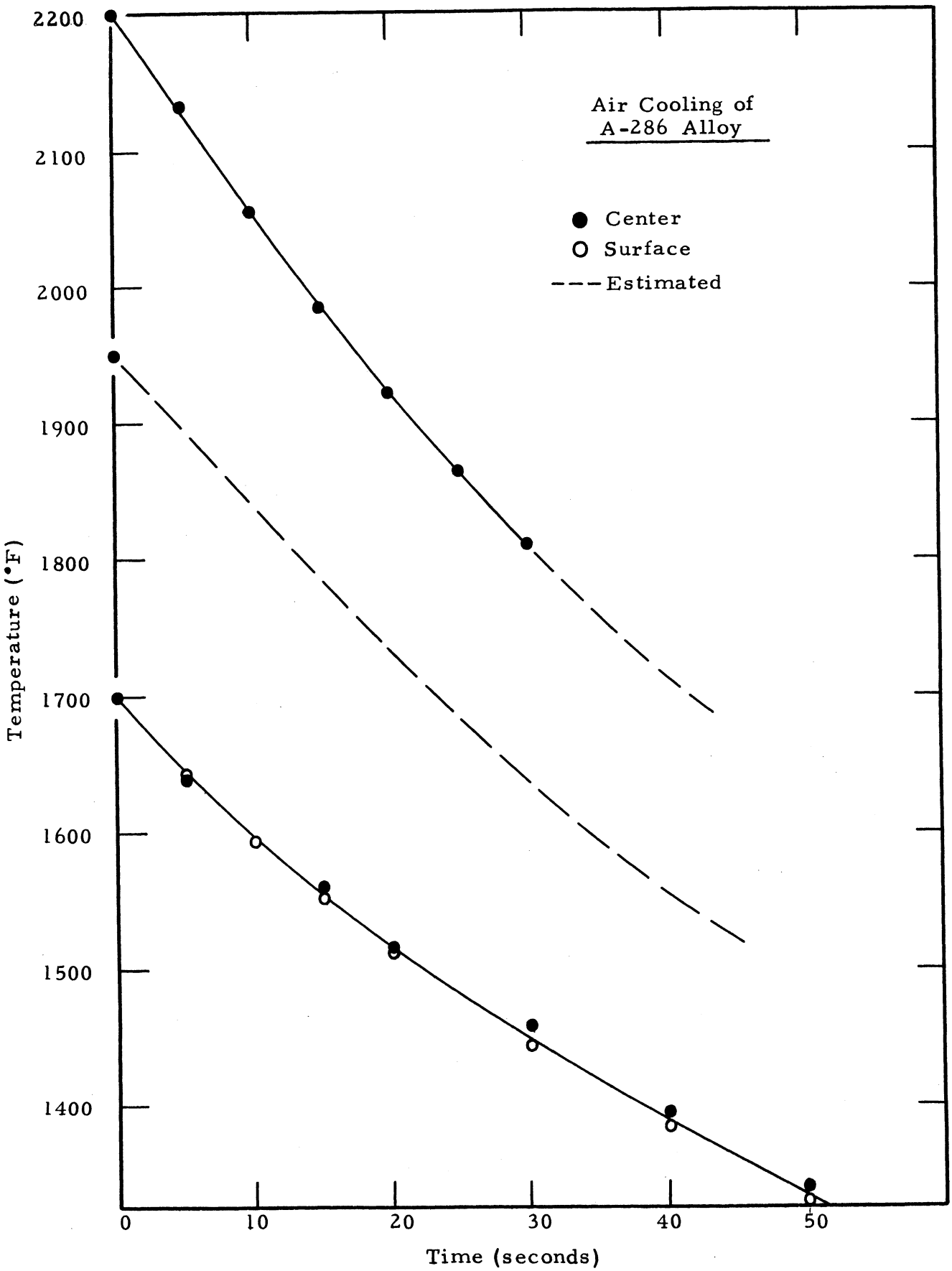


Figure 2 . - Cooling Curves for 3/4-Inch Diameter Bars of A-286 Alloy Air Cooled from 1700°, 1950°, and 2200°F. The bars were held at temperature for 45 minutes before cooling.

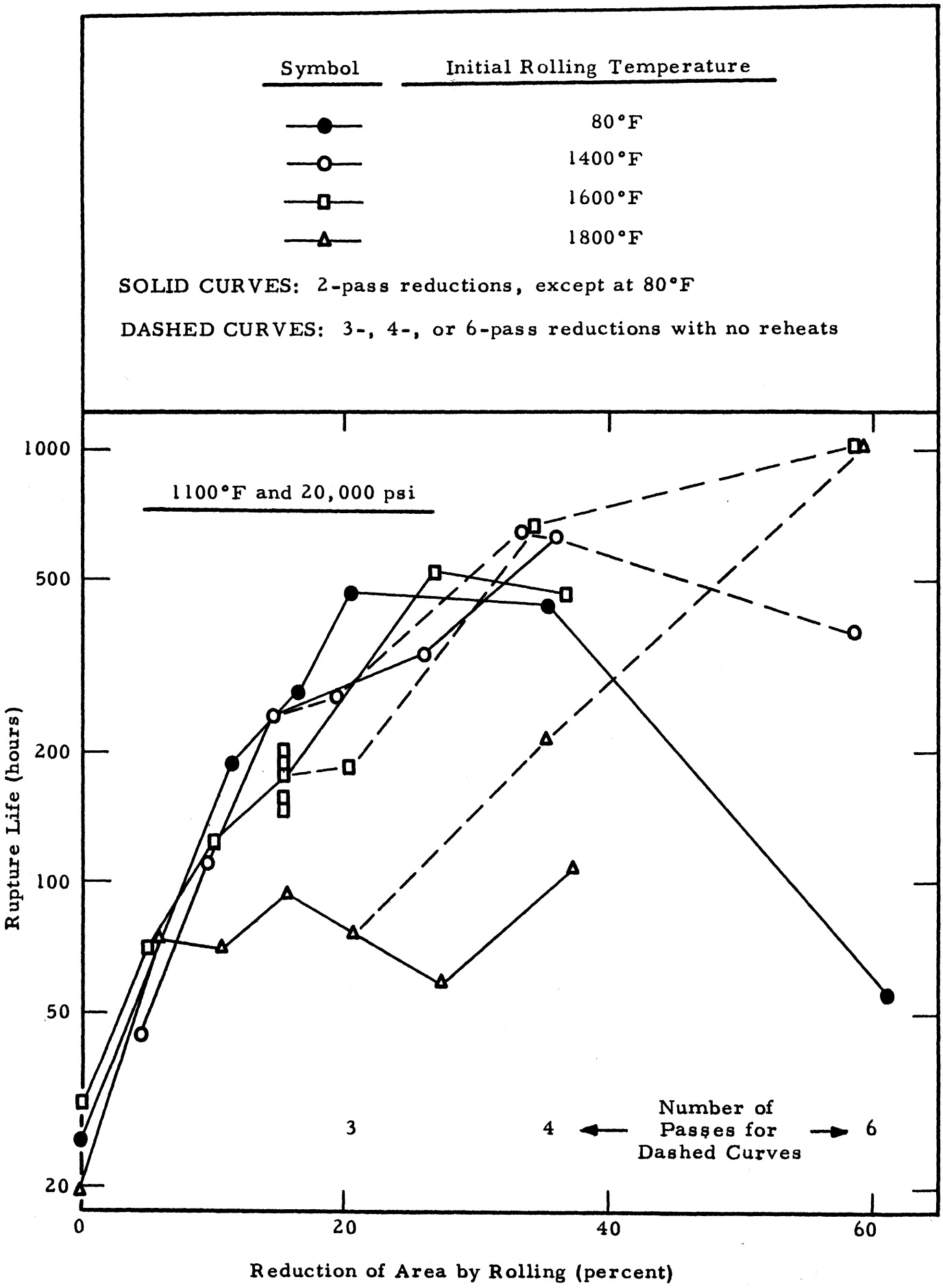


Figure 3. - Variation of Log Rupture Life with Rolling Conditions for "A" Nickel Tested at 1100°F and 20,000 psi.

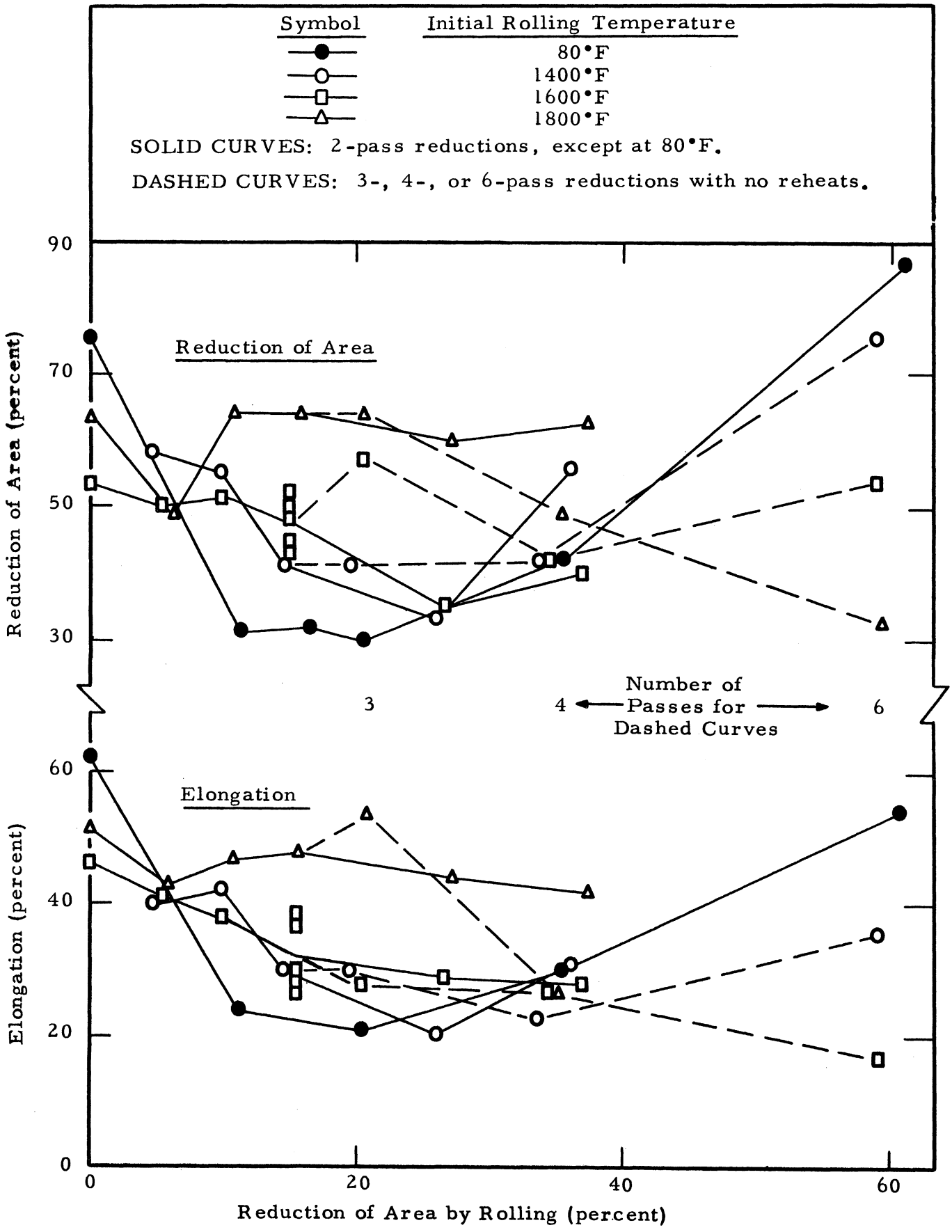


Figure 4 . - Variation of Rupture Ductility with Rolling Conditions for "A" Nickel tested at 1100°F and 20,000 psi.

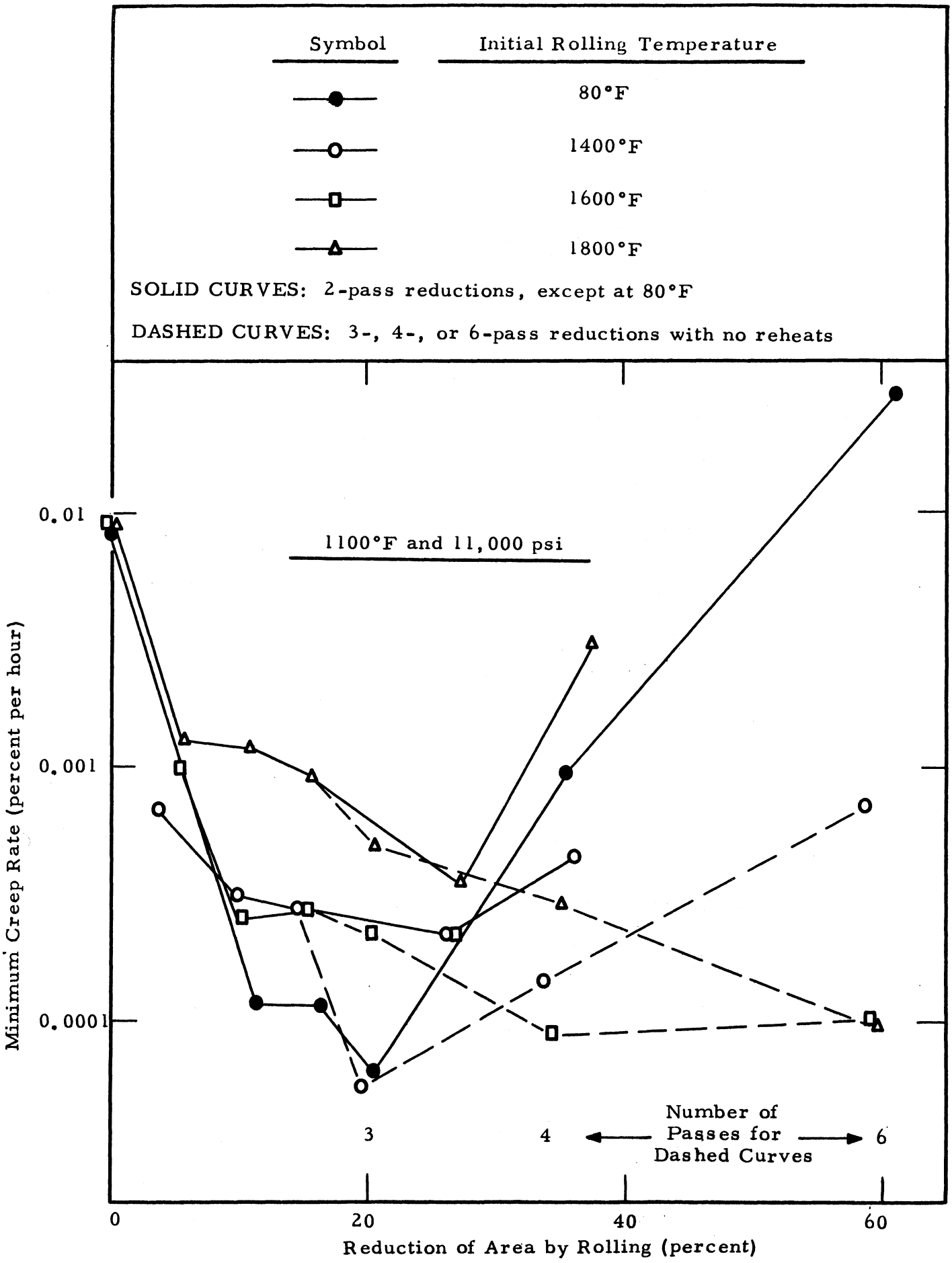


Figure 5. - Variation of Log Minimum Creep Rate with Rolling Conditions for "A" Nickel Tested at 1100°F and 11,000 psi.

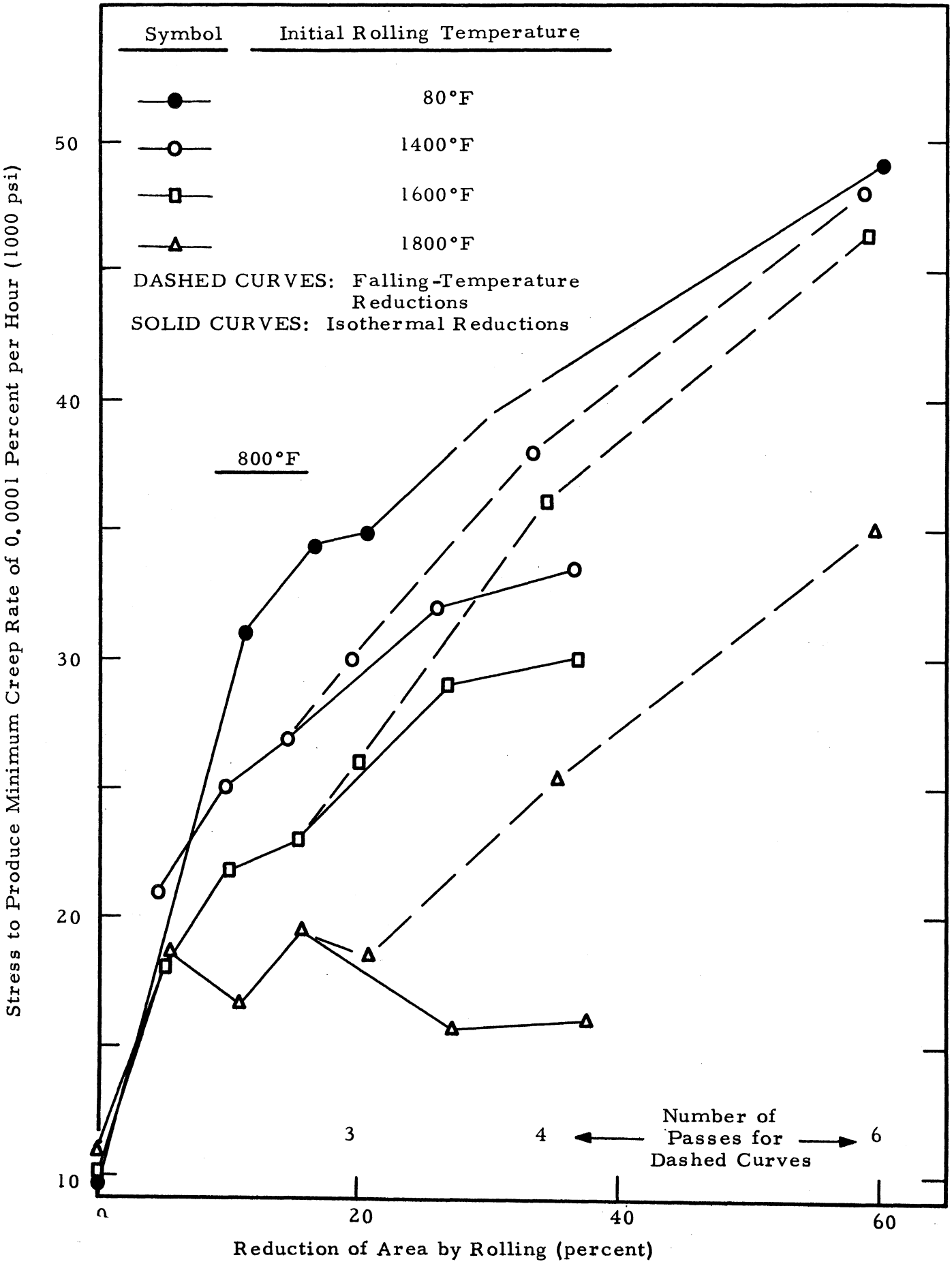


Figure 6 . - Influence of Rolling Conditions on the Stress to Produce a Minimum Creep Rate of 0.0001 Percent per Hour for "A" Nickel at 800°F.

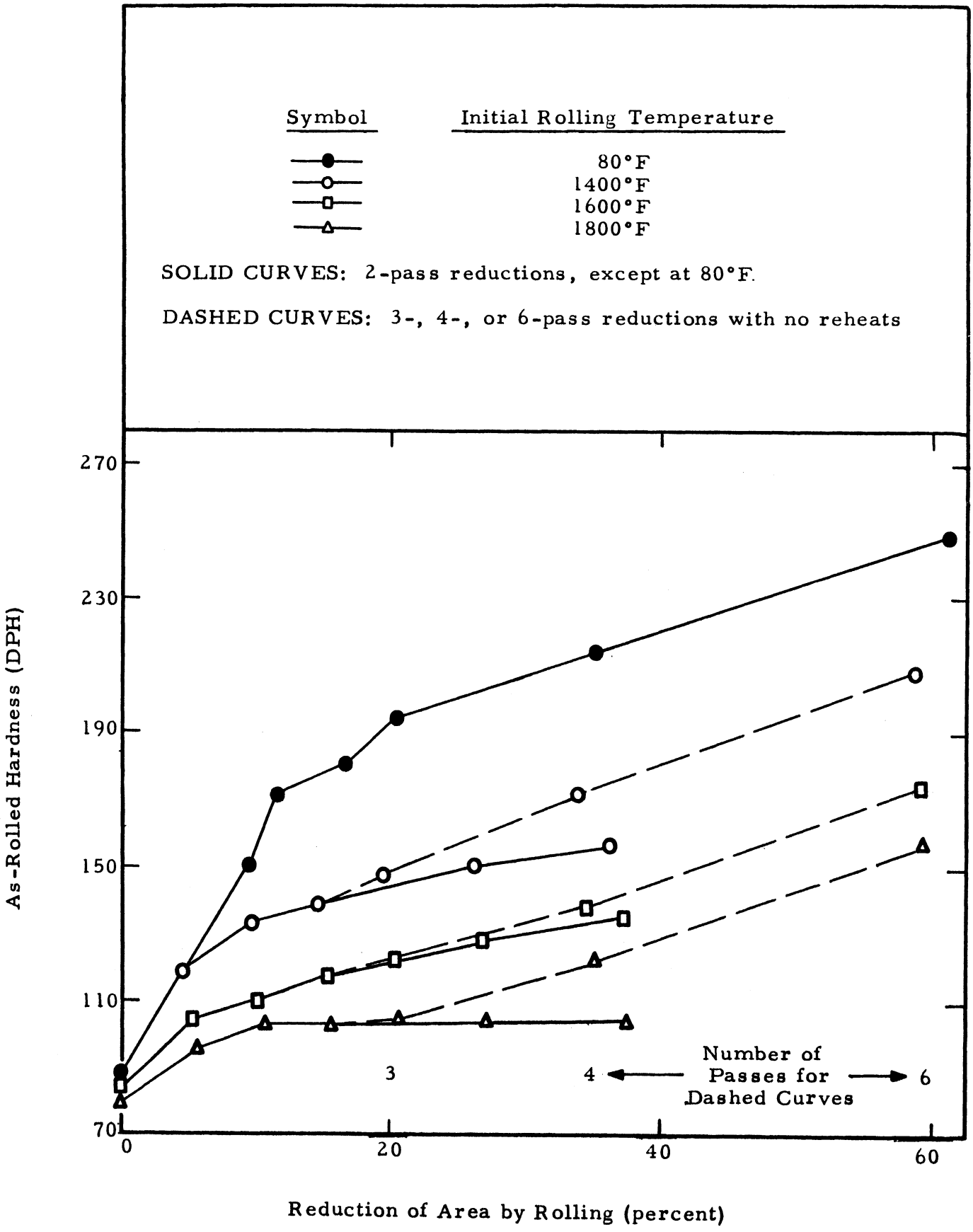


Figure 7 . - Relationship Between Hardness and Rolling Conditions for "A" Nickel.

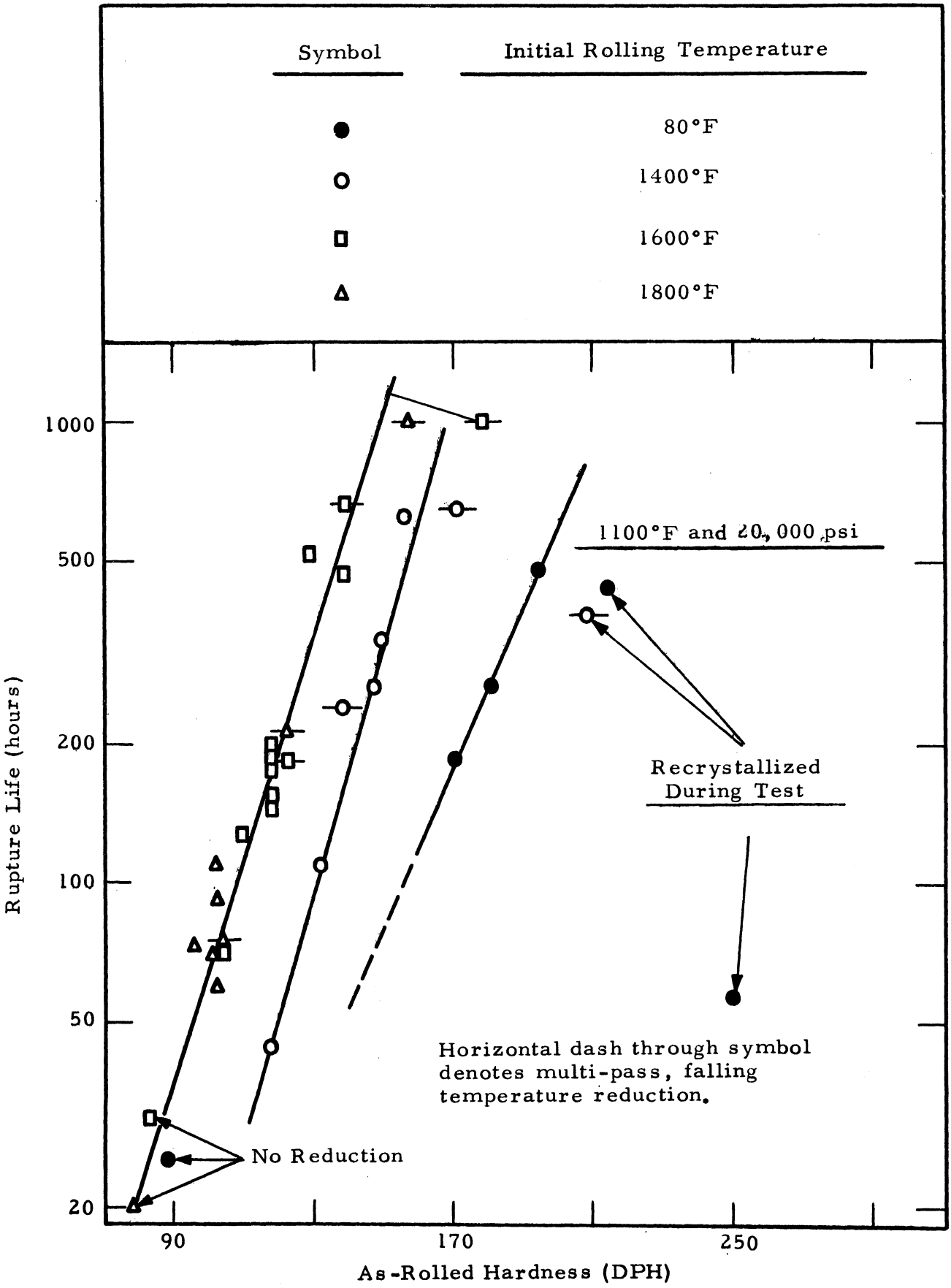


Figure 8 . - Correlations of Log Rupture Life with As-Rolled Hardness for "A" Nickel Tested at 1100°F and 20,000 psi.

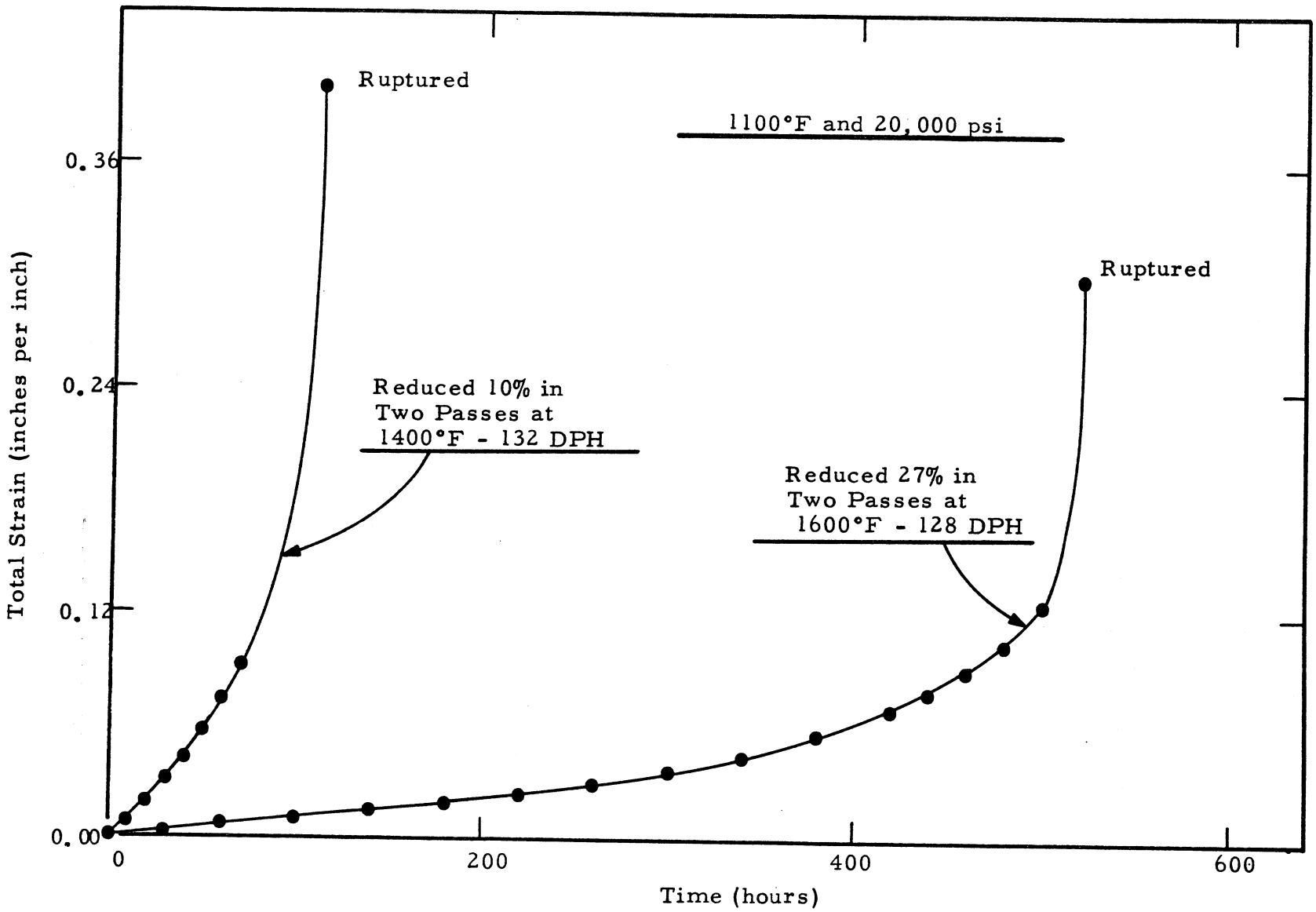


Figure 9 . - Total Strain Versus Time Curves for "A" Nickel at 1100°F and 20,000 psi Showing the Effect of Two Different Sets of Rolling Conditions at a Constant Hardness of 130 DPH.

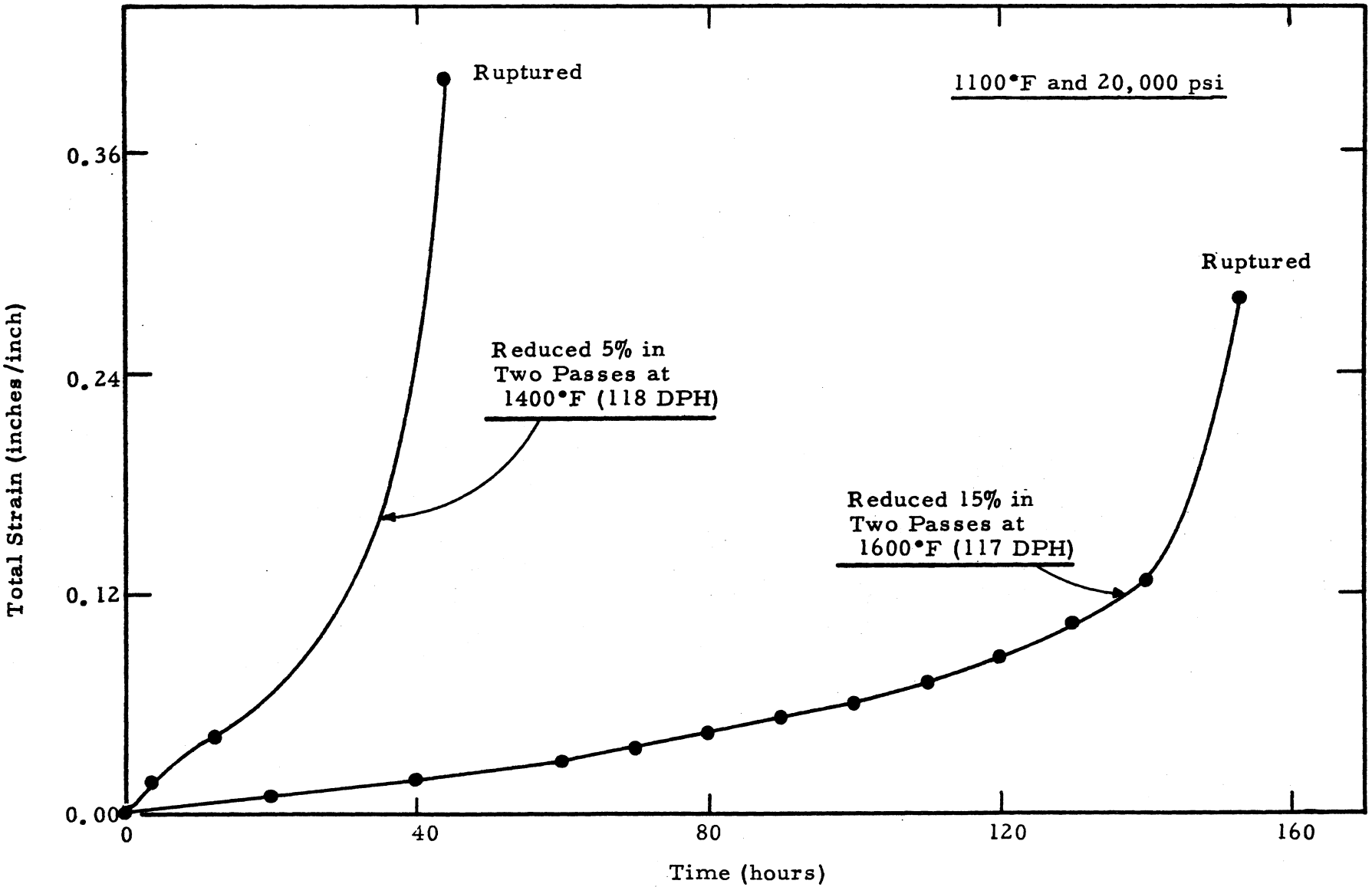


Figure 10. - Total Strain Versus Time Curves for "A" Nickel at 1100°F and 20,000 psi Showing the Effect of Two Different Sets of Rolling Conditions at a Constant Hardness of 118 DPH.

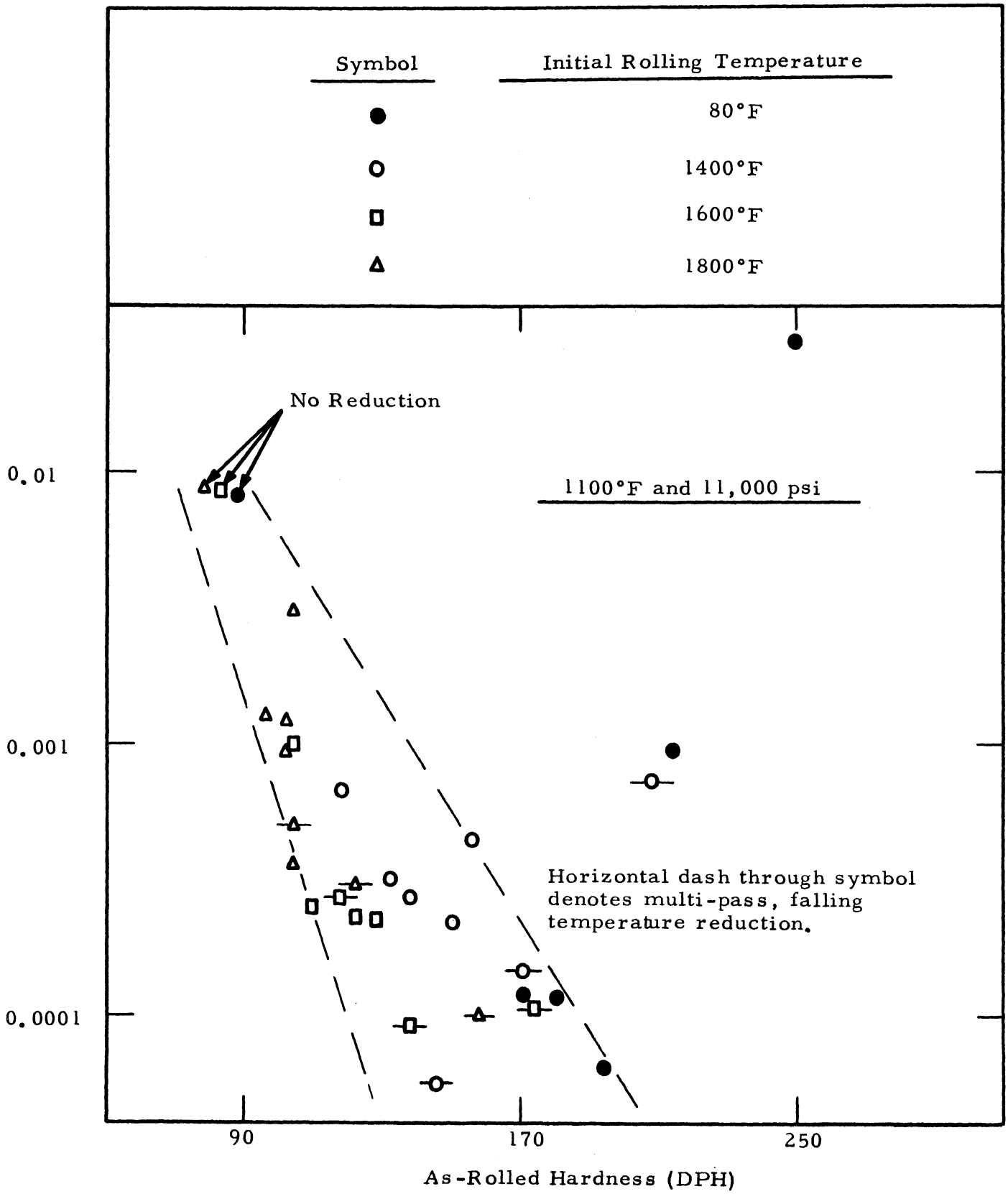


Figure 11. - Correlation of Log Minimum Creep Rate with As-Rolled Hardness for "A" Nickel Tested at 1100°F and 11,000 psi.

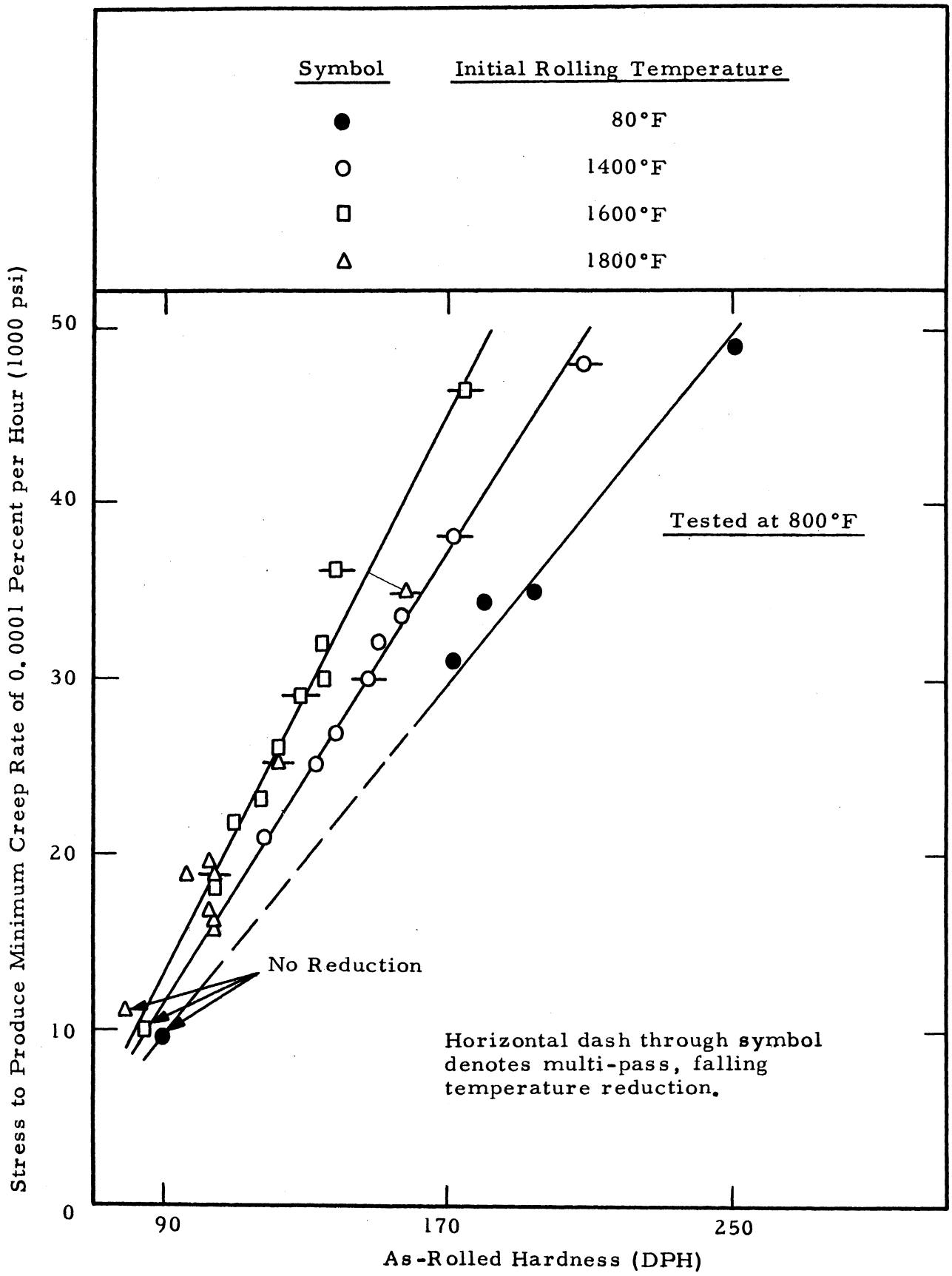
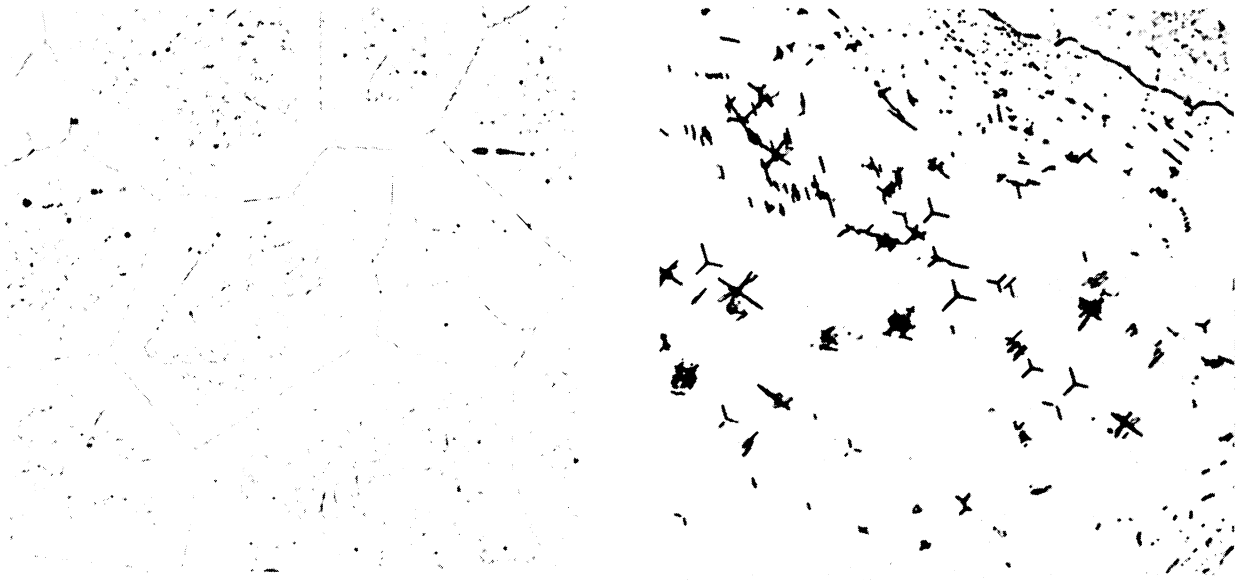
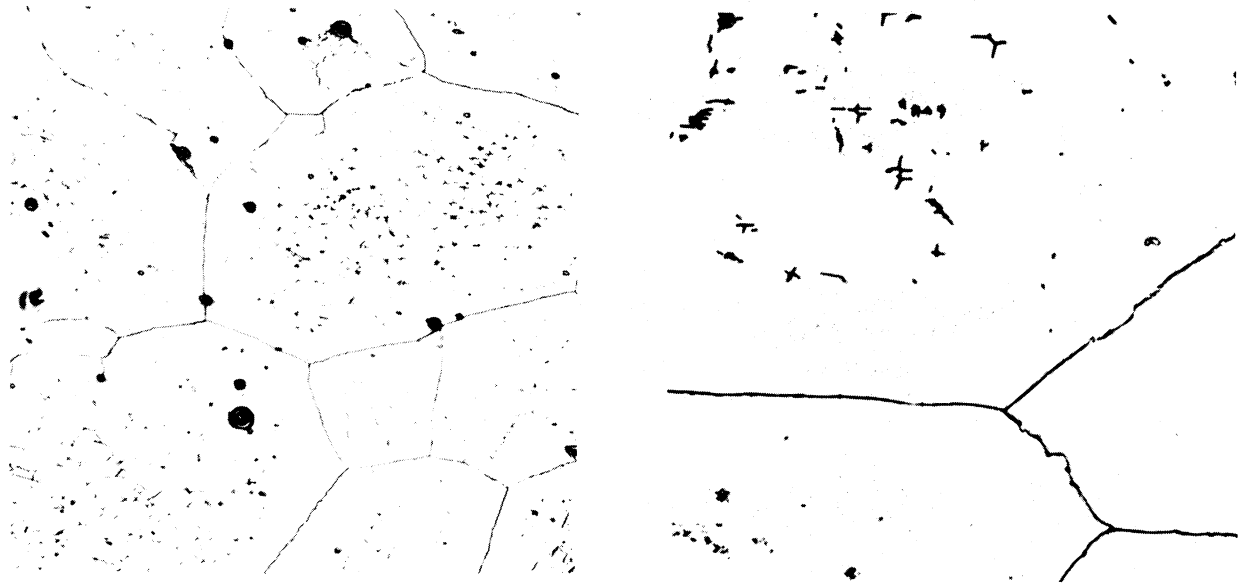


Figure 12. - Correlations Between As-Rolled Hardness and the Stress to Produce a Minimum Creep Rate of 0.0001 Percent per Hour for "A" Nickel at 800°F.

X250 X1000



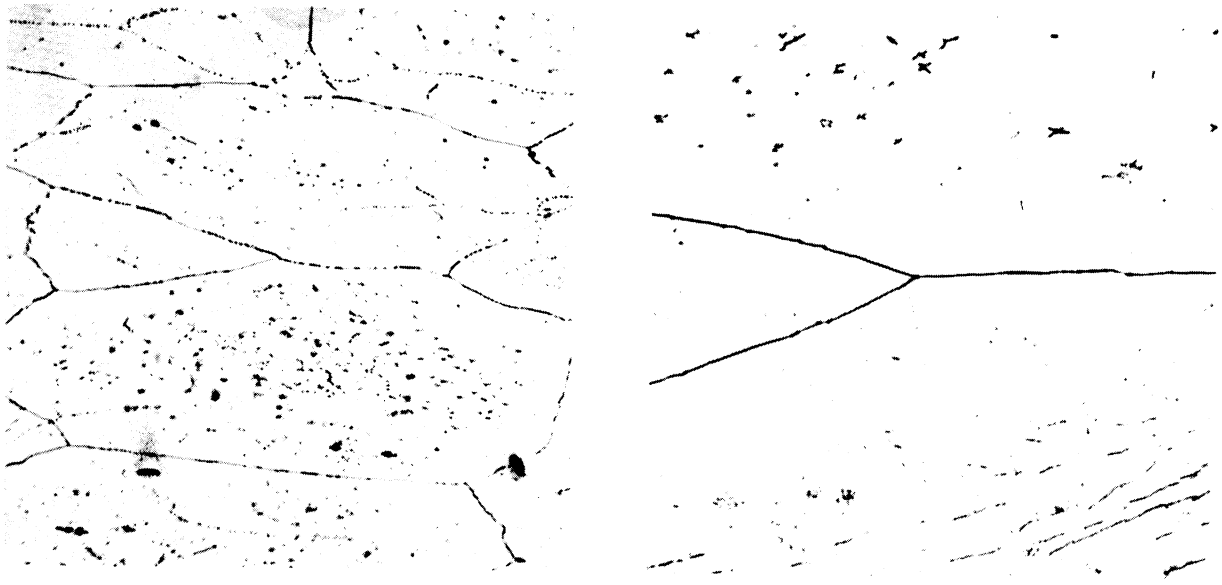
(a) Base Condition. (As Received + 20 Percent Cold Work + 1 Hour at 1600°F, Air Cooled.)



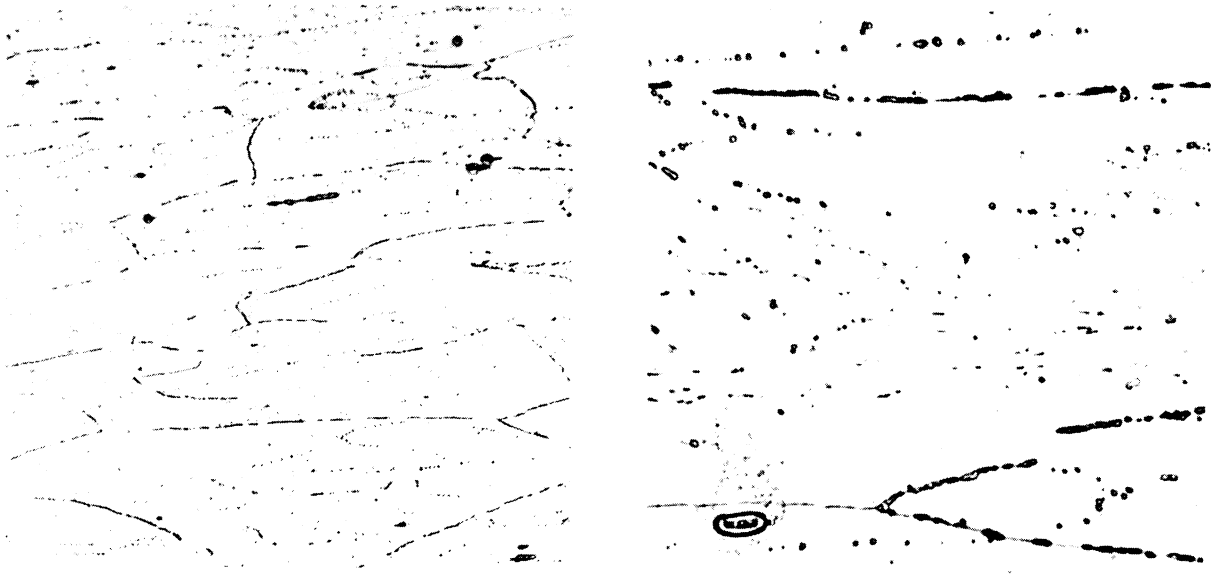
(b) Same as in (a) + 16.5 Percent Cold Work.

Figure 13 - Microstructure of "A" Nickel in the (a) Base Condition and As Subsequently Cold Rolled (b) 16.5 Percent, (c) 35.5 Percent, and (d) 61.2 Percent. Etched electrolytically in 40 percent H_3PO_4.

X250 X1000



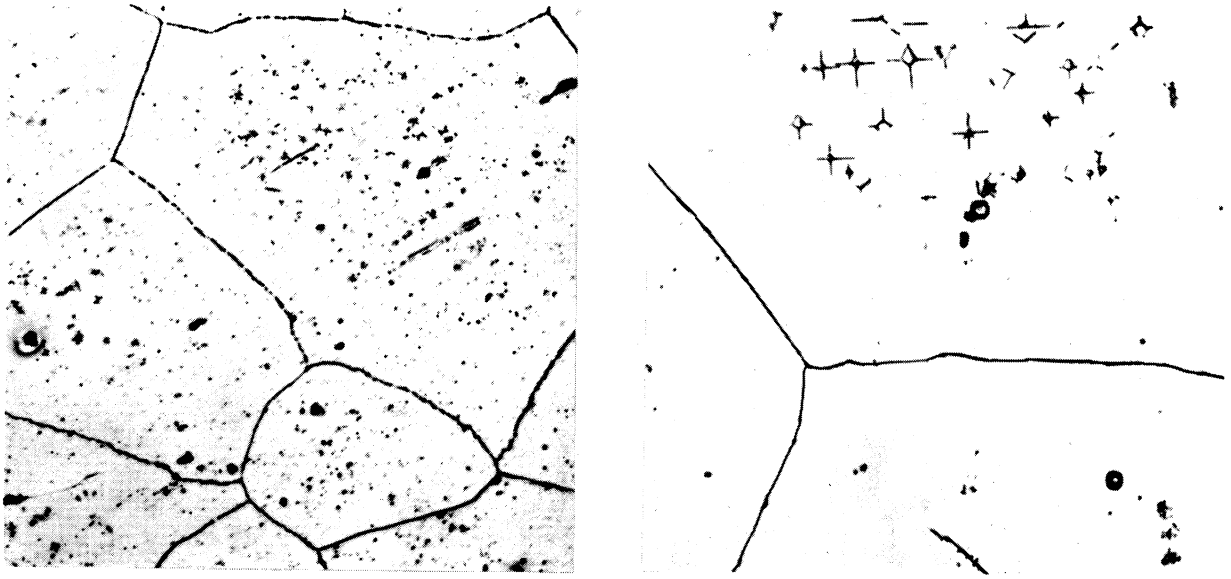
(c) Same as in (a) + 35.5 Percent Cold Work.



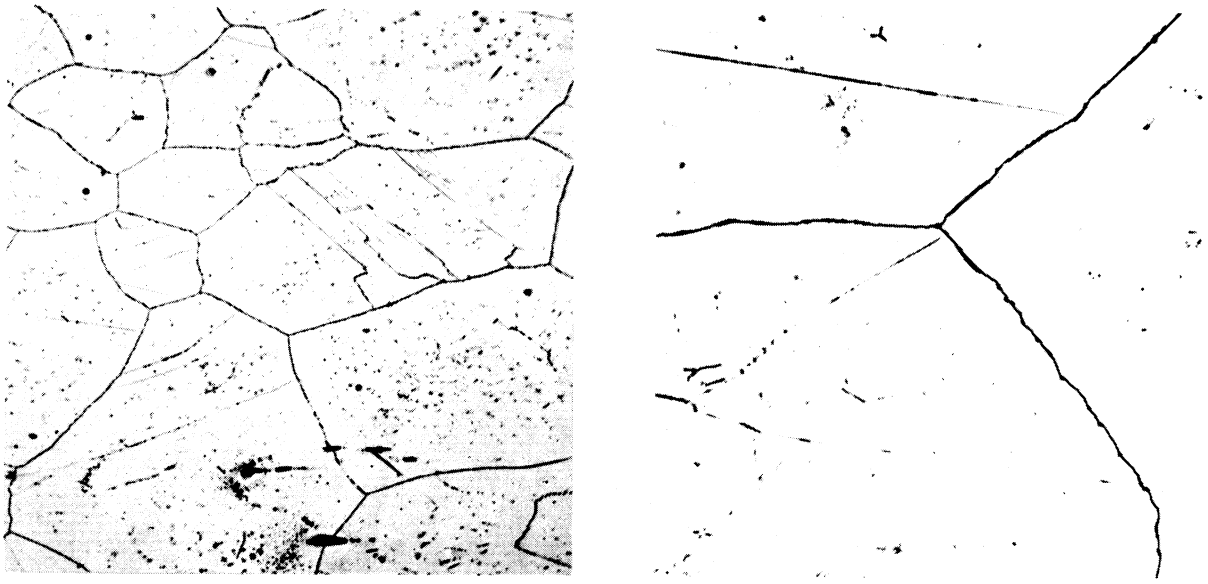
(d) Same as in (a) + 61.2 Percent Cold Work.

Figure 13 . - Concluded.

X250 X1000



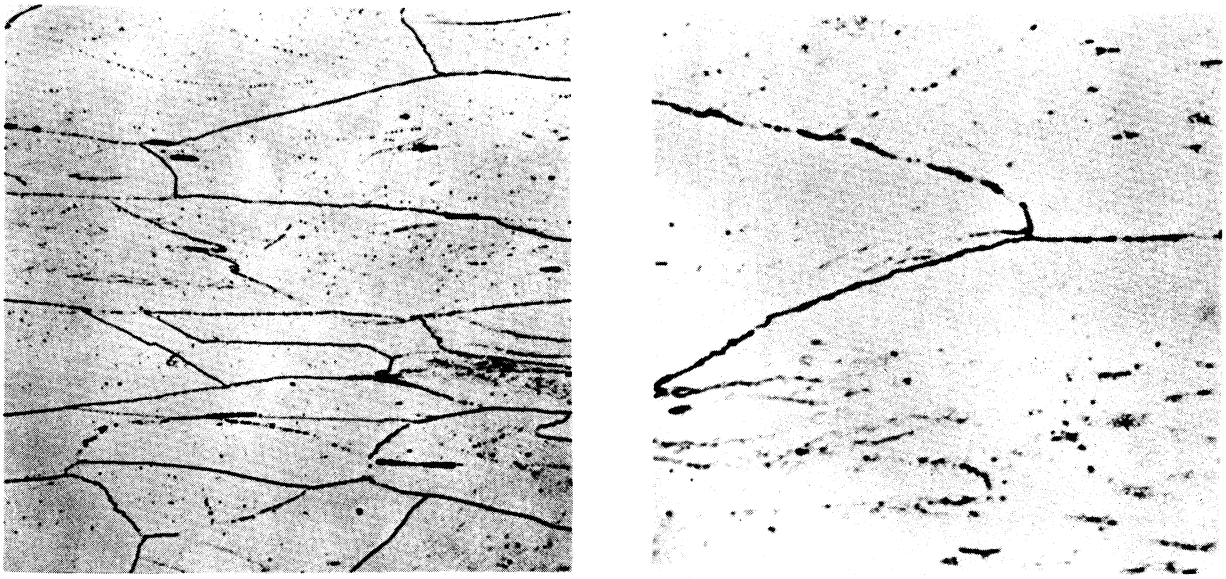
(a) Heated 45 Minutes at 1400°F, Rolled 4.7 Percent in 2 Passes, Air Cooled.



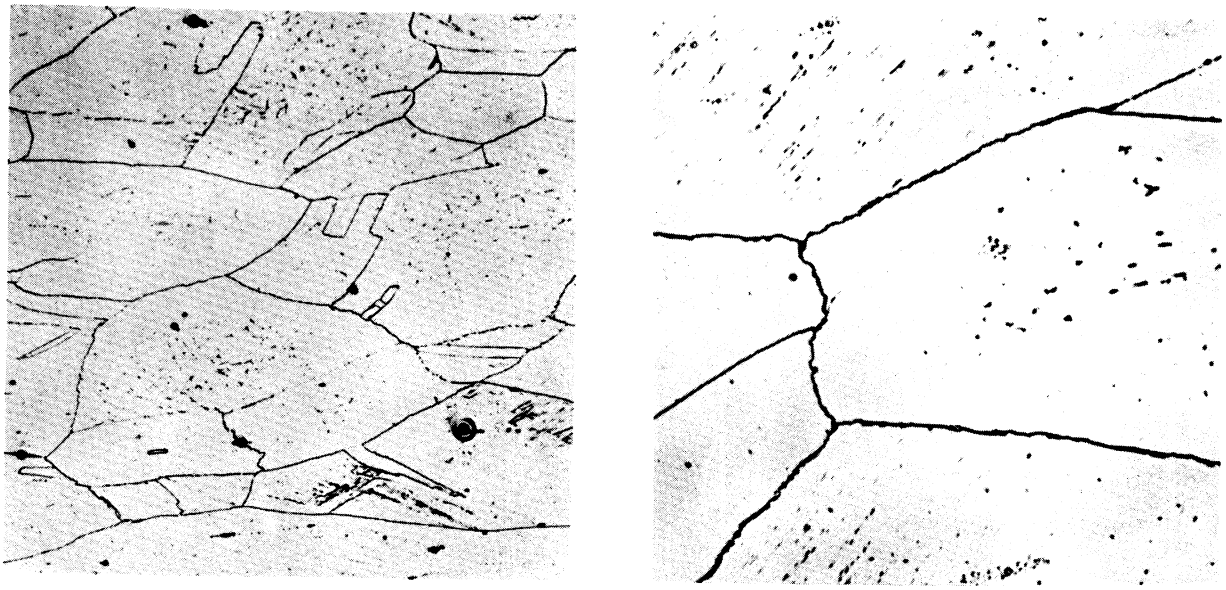
(b) Heated 45 Minutes at 1400°F, Rolled 19.5 Percent in 3 Passes, Air Cooled.

Figure 14. - Microstructure of "A" Nickel Rolled at 1400°F. The reductions were (a) 4.7 percent in 2 passes, (b) 19.5 percent in 3 passes, (c) 58.7 percent in 6 passes, and (d) 36.1 percent in 2 passes. Etched electrolytically in 40 percent H_3PO_4.

X250 X1000

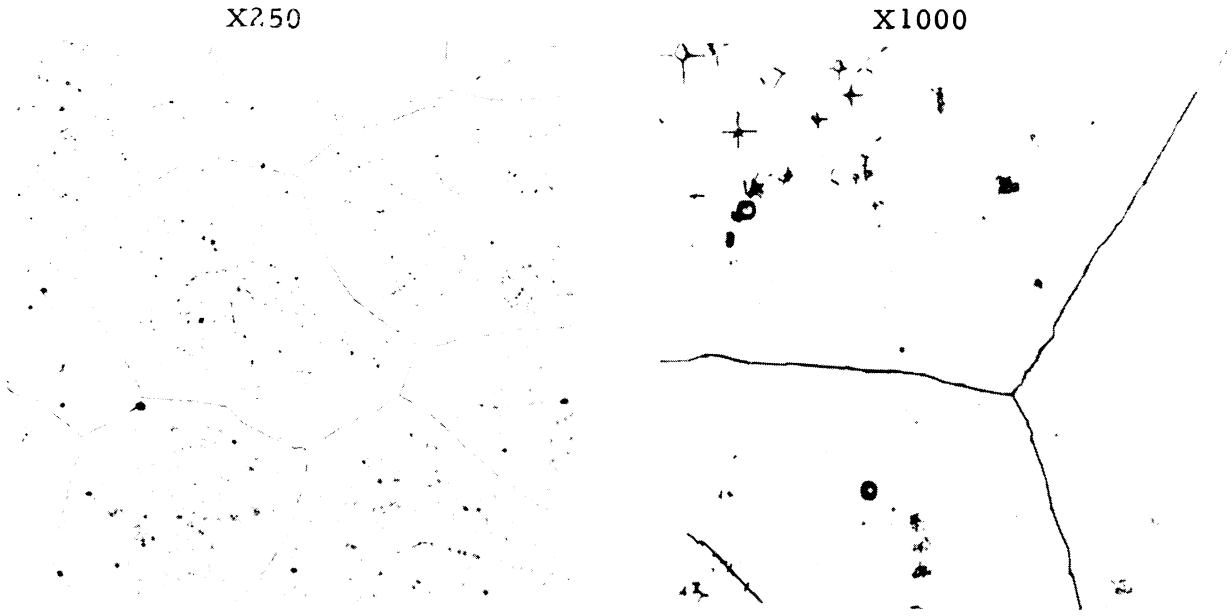


(c) Heated 45 Minutes at 1400°F, Rolled 58.7 Percent in 6 Passes, Air Cooled.

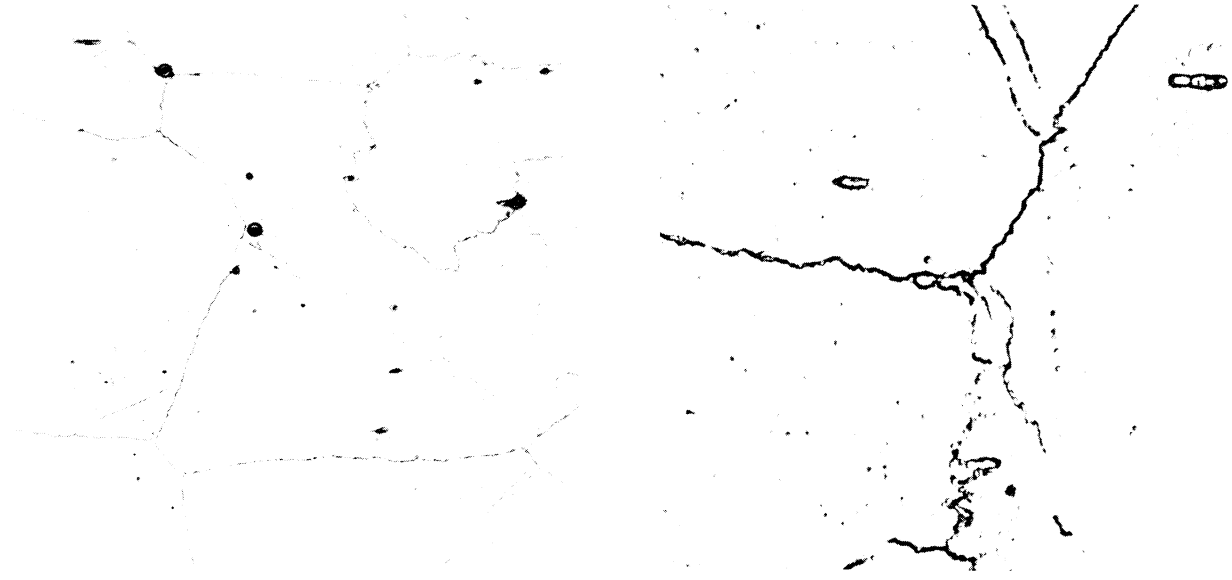


(d) Heated 45 Minutes at 1400°F, Rolled 36.1 Percent in 2 Passes, Air Cooled.

Figure 14. - Concluded.



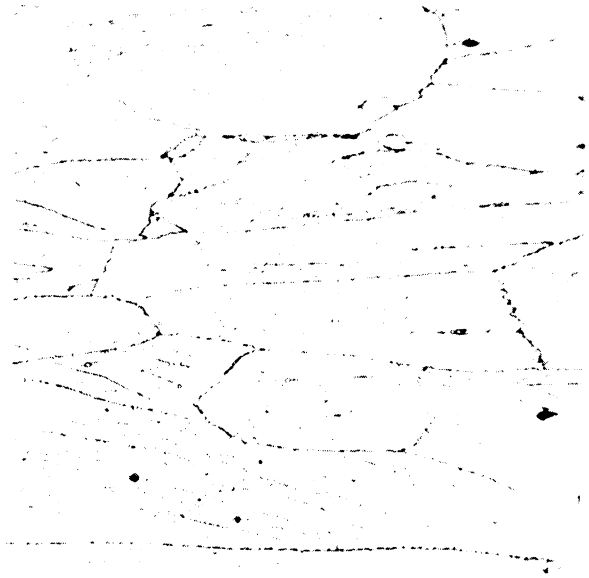
(a) Base Condition + Heated 45 Minutes at 1600°F, Air Cooled.



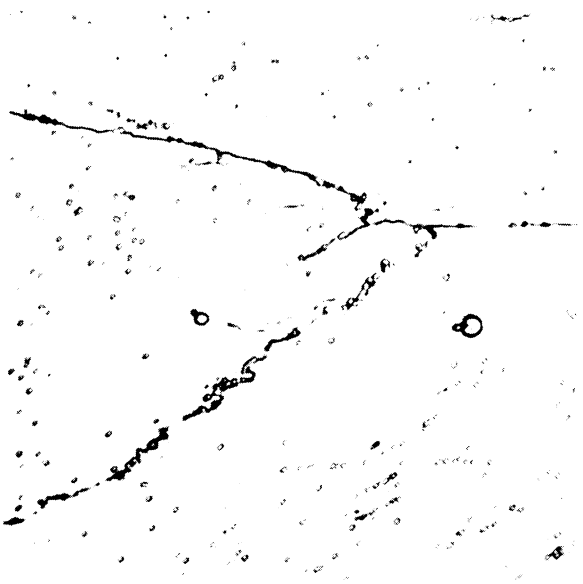
(b) Heated 45 Minutes at 1600°F, Rolled 20.3 Percent in 3 Passes, Air Cooled.

Figure 15. - Microstructure of "A" Nickel Heated to 1600°F and (a) Air Cooled, (b) Rolled 20.3 Percent in 3 Passes, (c) Rolled 59.1 Percent in 6 Passes, and (d) Rolled 36.8 Percent in 2 Passes. Etched electrolytically in 40 percent H_3PO_4.

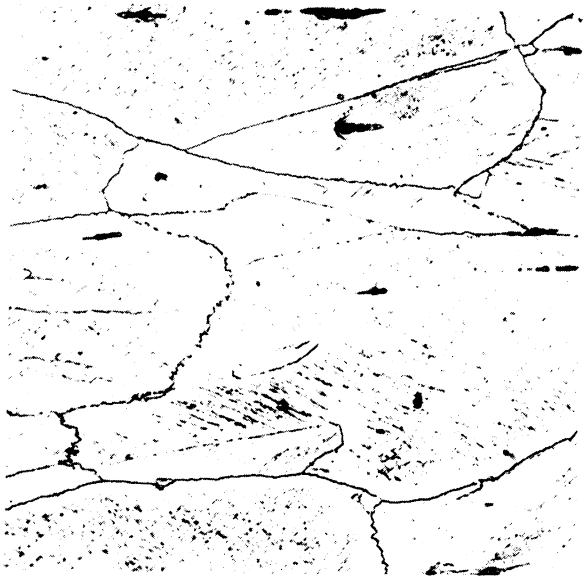
X250



X1000



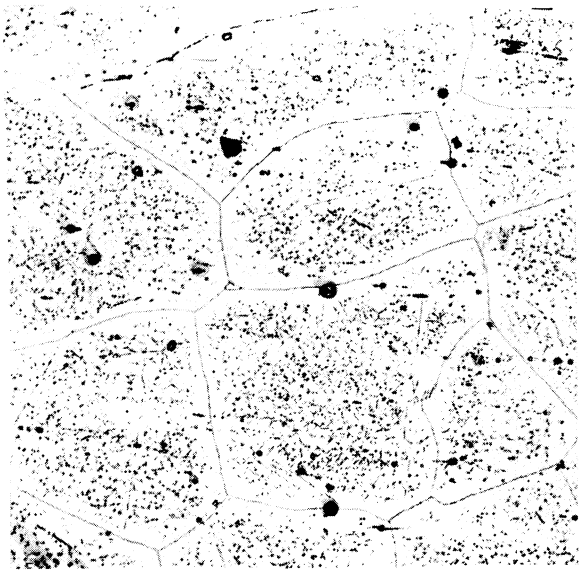
(c) Heated 45 Minutes at 1600°F, Rolled 59.1 Percent in 6 Passes, Air Cooled.



(d) Heated 45 Minutes at 1600°F, Rolled 36.8 Percent in 2 Passes, Air Cooled.

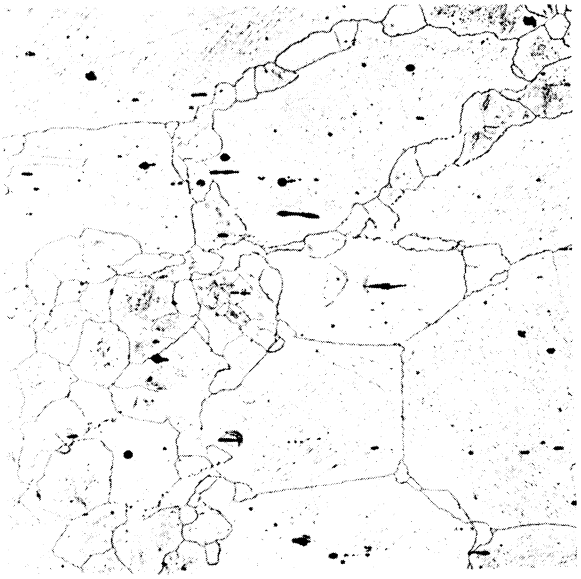
Figure 15. - Concluded.

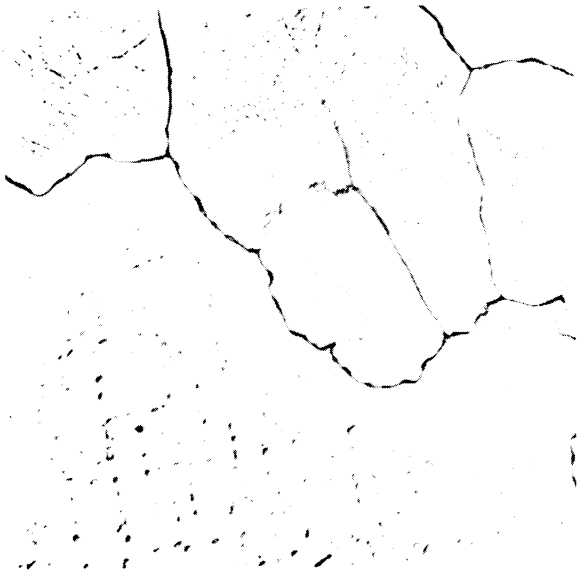
X100



X1000

(a) Base Condition + Heated 45 Minutes at 1800°F, Air Cooled.

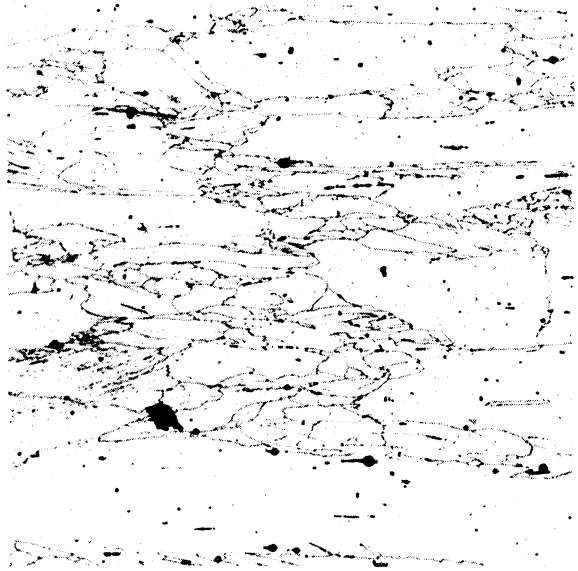




(b) Heated 45 Minutes at 1800°F, Rolled 20.7 Percent in 3 Passes, Air Cooled.

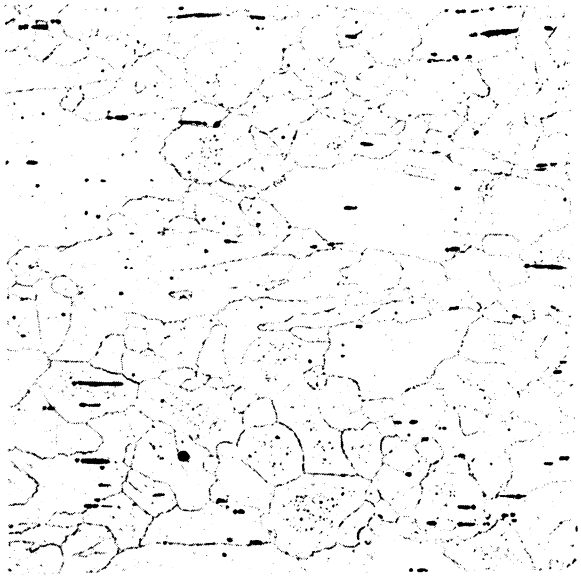
Figure 16. - Microstructure of "A" Nickel Heated to 1800°F and (a) Air Cooled, (b) Rolled 20.7 Percent in 3 Passes, (c) Rolled 59.3 Percent in 6 Passes, and (d) Rolled 37.5 Percent in 2 Passes. Etched electrolytically in 40 percent H_3PO_4.

X100



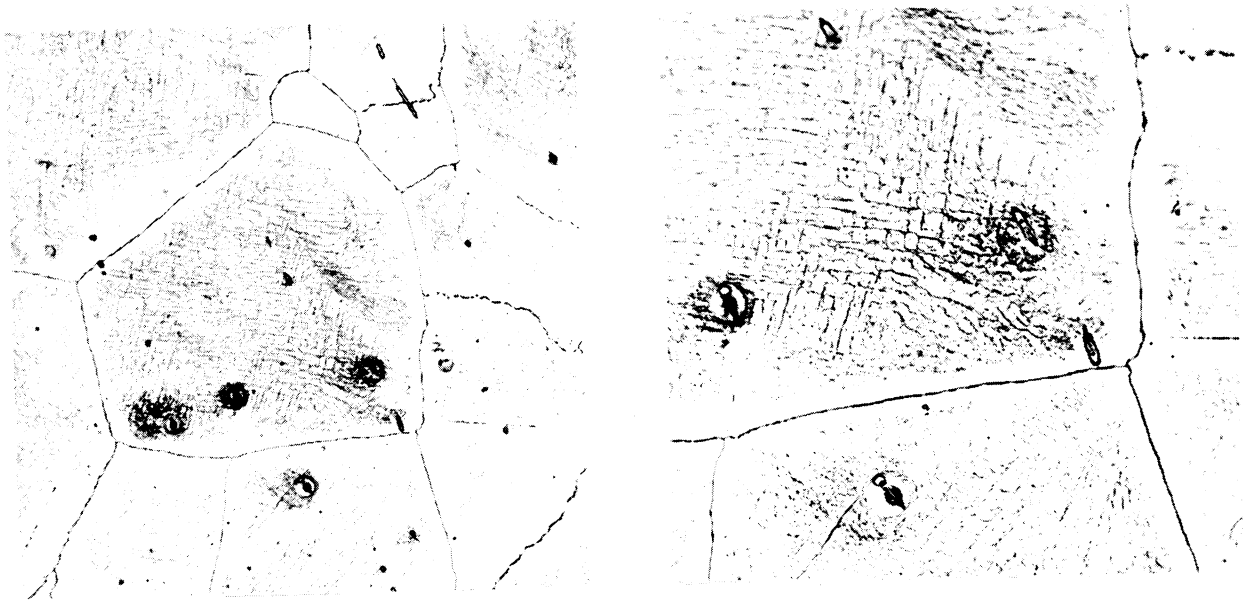
X1000

(c) Heated 45 Minutes at 1800°F, Rolled 59.3 Percent in 6 Passes, Air Cooled.

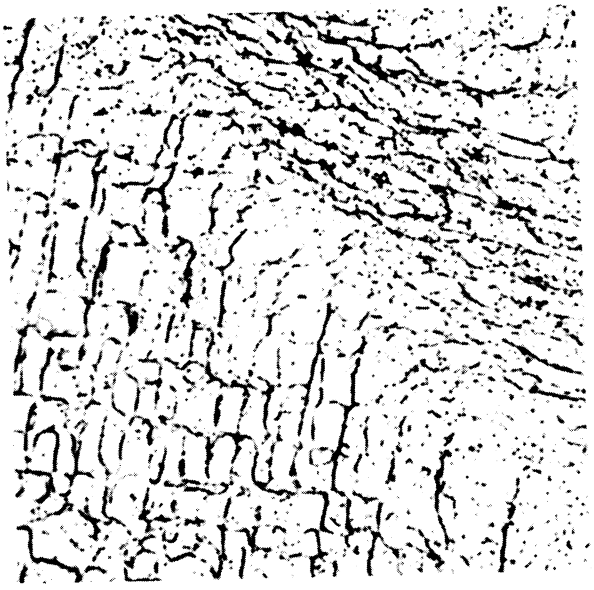


(d) Heated 45 Minutes at 1800°F, Rolled 37.5 Percent in 2 Passes, Air Cooled.

Figure 16. - Concluded.



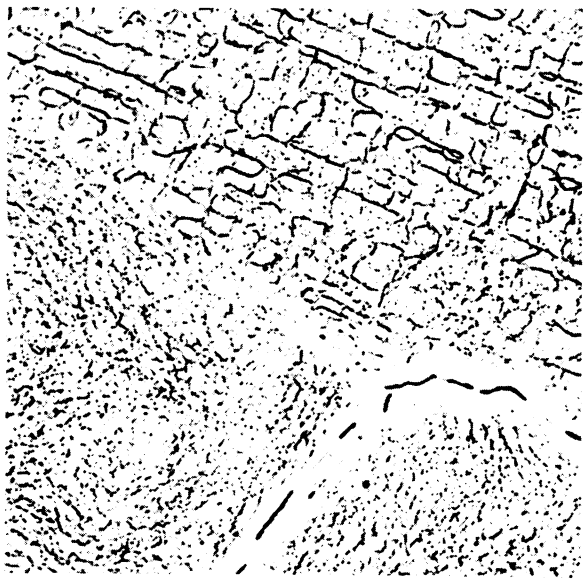
X250 X500



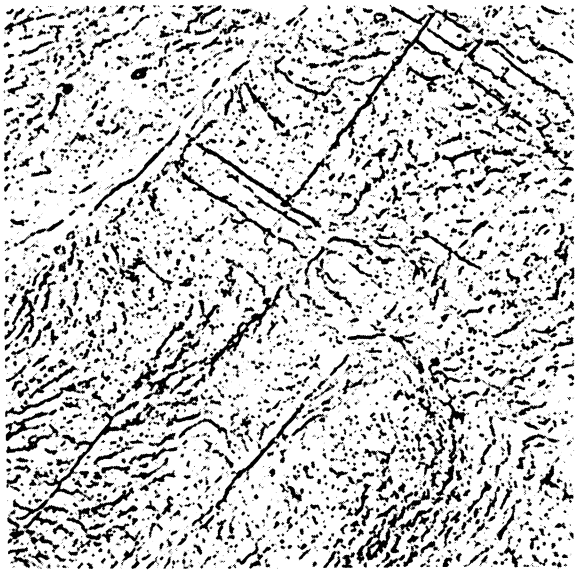
X2000

Figure 17. - Appearance of Substructure in "A" Nickel at Different Magnifications (Same Area). The sample was rolled 5.8 percent in 2 passes at 1800°F and air cooled. Electropolished and then etched electrolytically in 40 percent H_3PO_4.

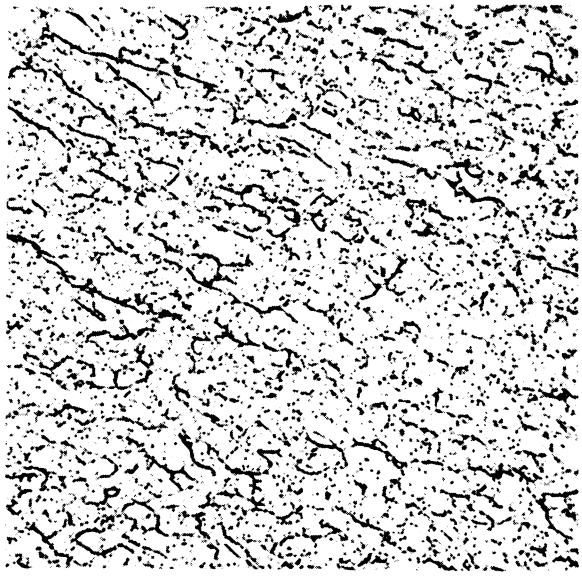
X1000



X1000



X1000



X1000

Figure 18. - Effect of Orientation and Inhomogeneous Strain on the Appearance of Substructure in a Single Sample of "A" Nickel. The sample was rolled 5.8 percent in 2 passes at 1800°F and air cooled. Electropolished and then etched electrolytically in 40 percent H_3PO_4.

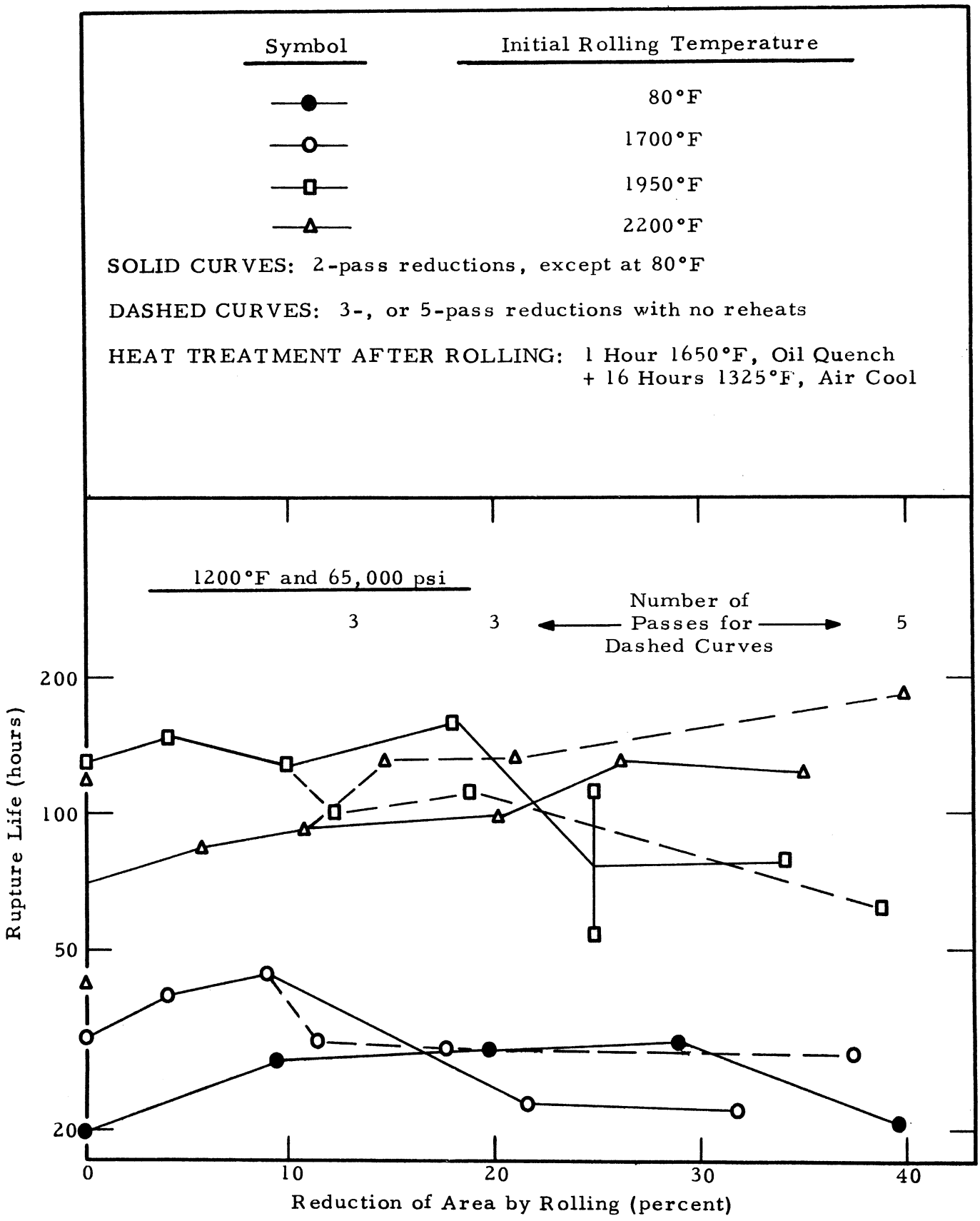


Figure 19 . - Variation of Log Rupture Life with Prior Rolling Conditions for Heat Treated A-286 Alloy Tested at 1200°F and 65,000 psi.

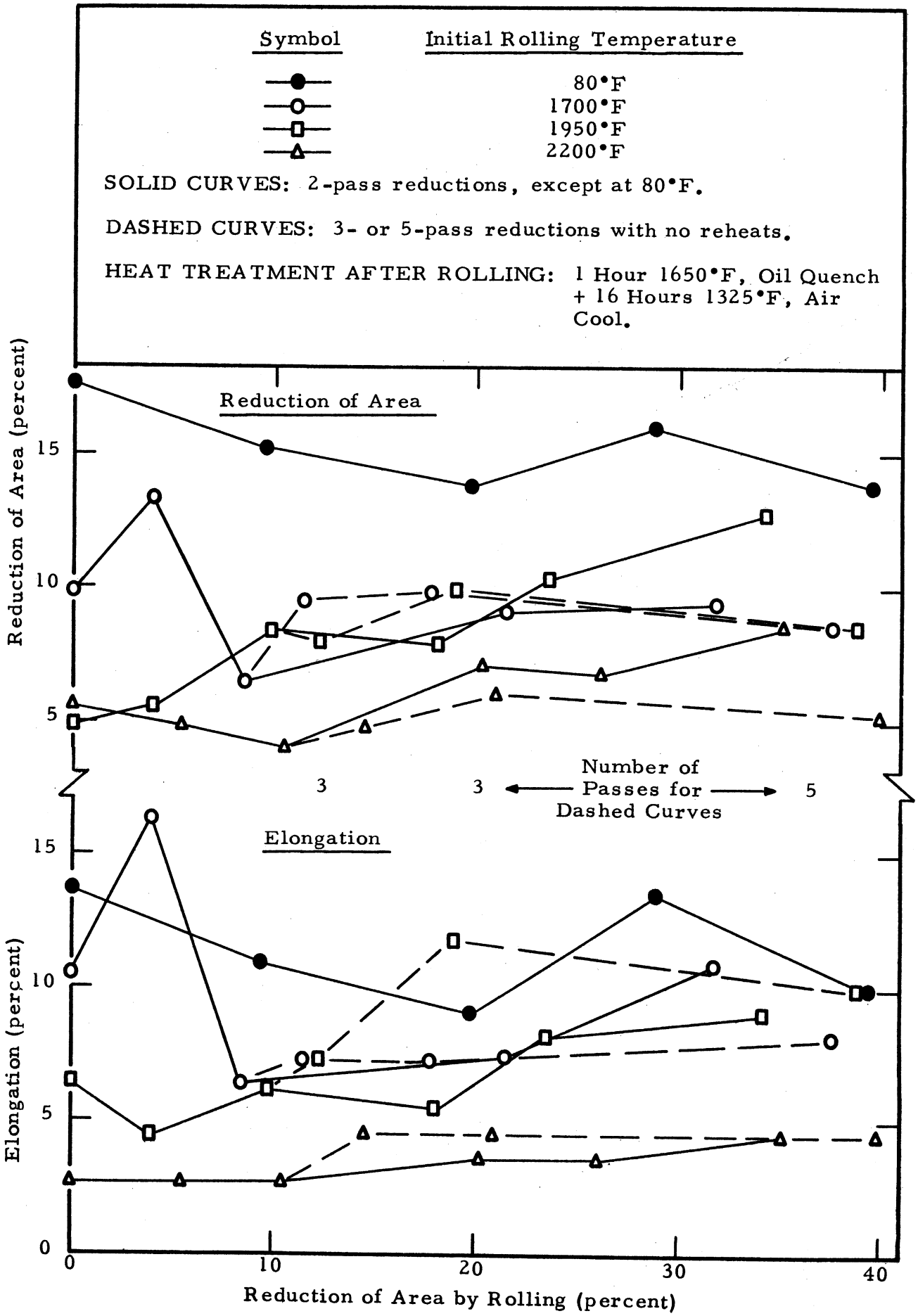


Figure 20. - Influence of Prior Rolling Conditions on the Ductility of Heat Treated A-286 Alloy Tested at 1200°F and 65,000 psi.

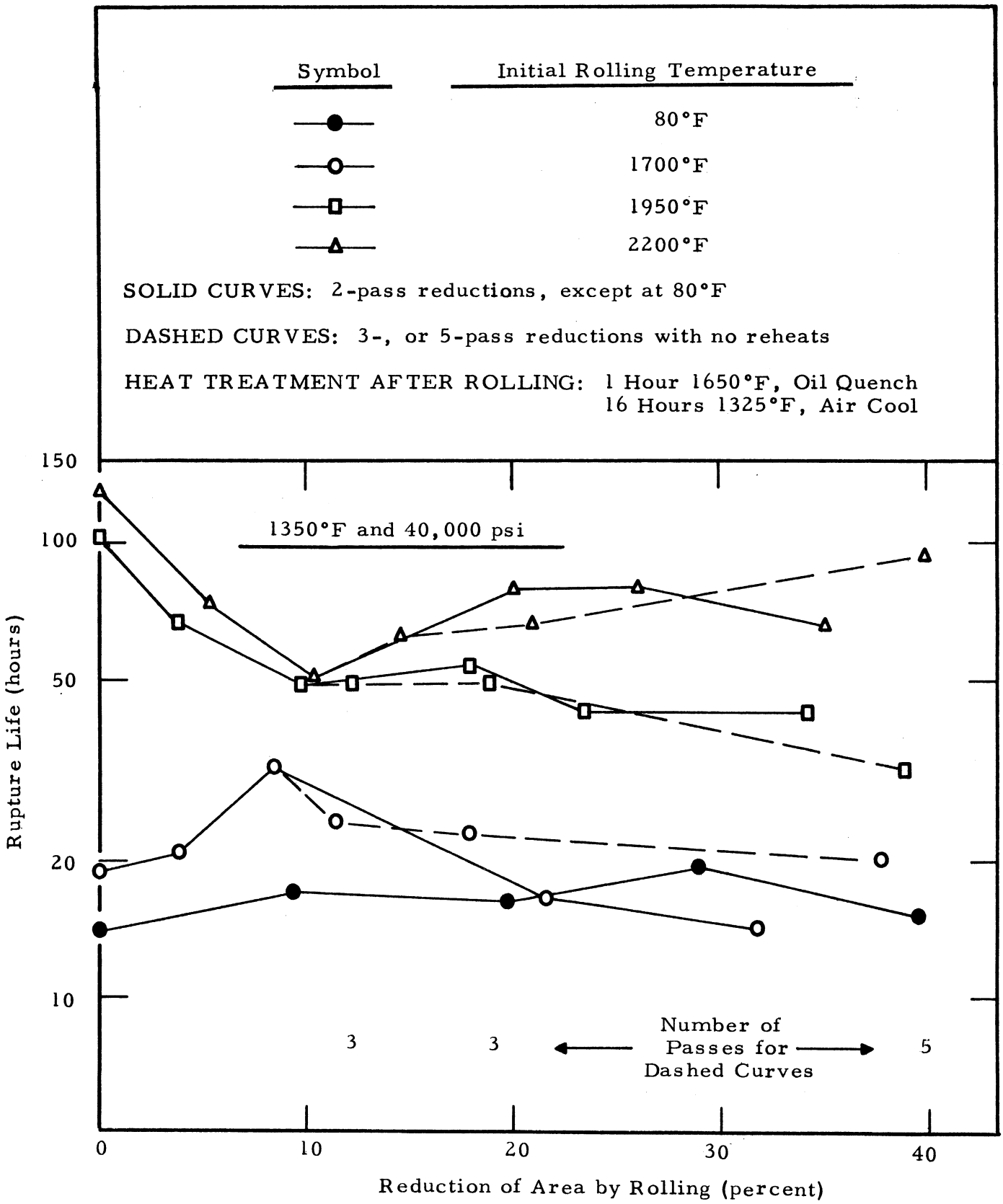


Figure 21. - Variation of Log Rupture Life with Prior Rolling Conditions for Heat Treated A-286 Alloy at 1350°F and 40,000 psi.

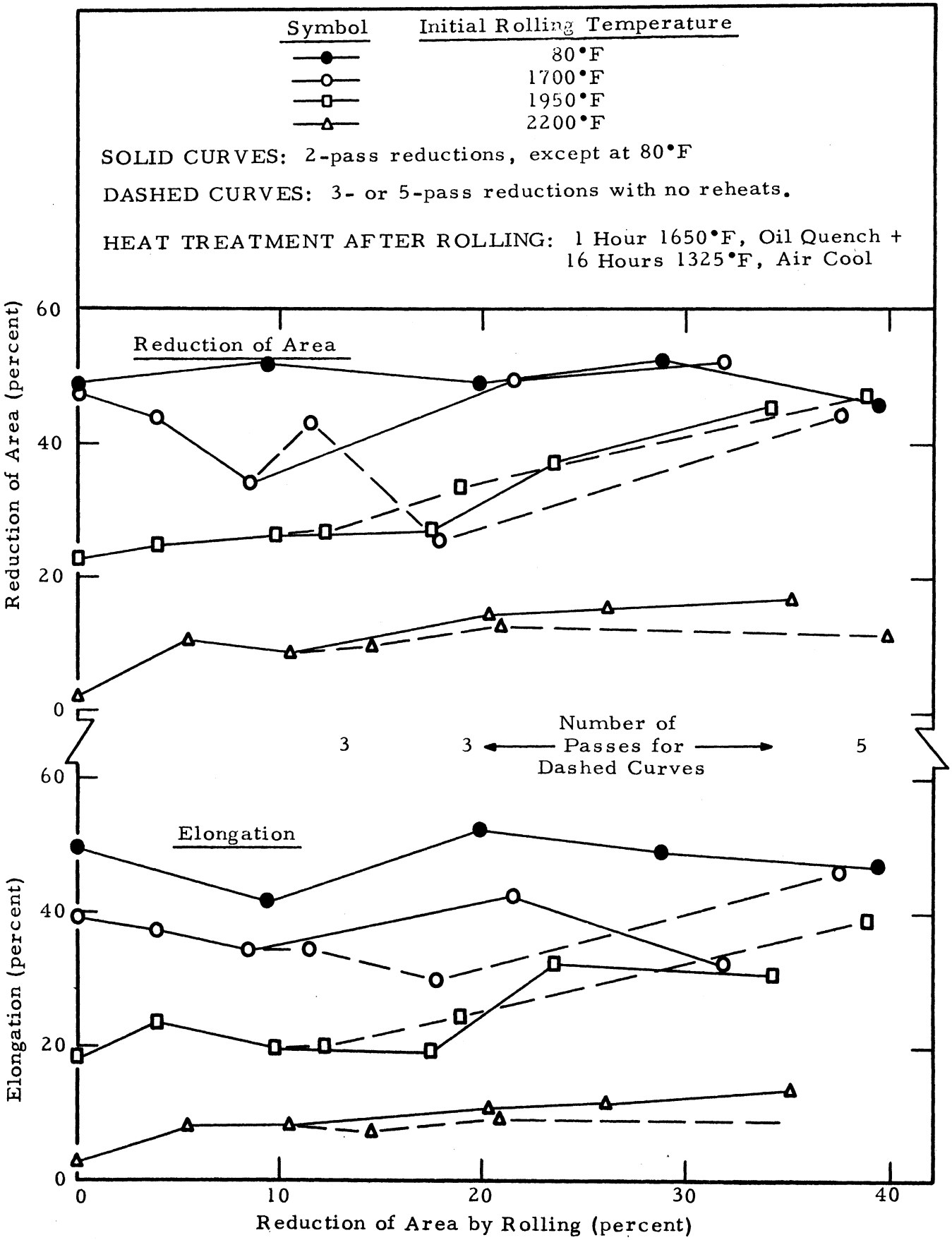


Figure 22 . - Influence of Prior Rolling Conditions on the Ductility of Heat Treated A-286 Alloy Tested at 1350°F and 40,000 psi.

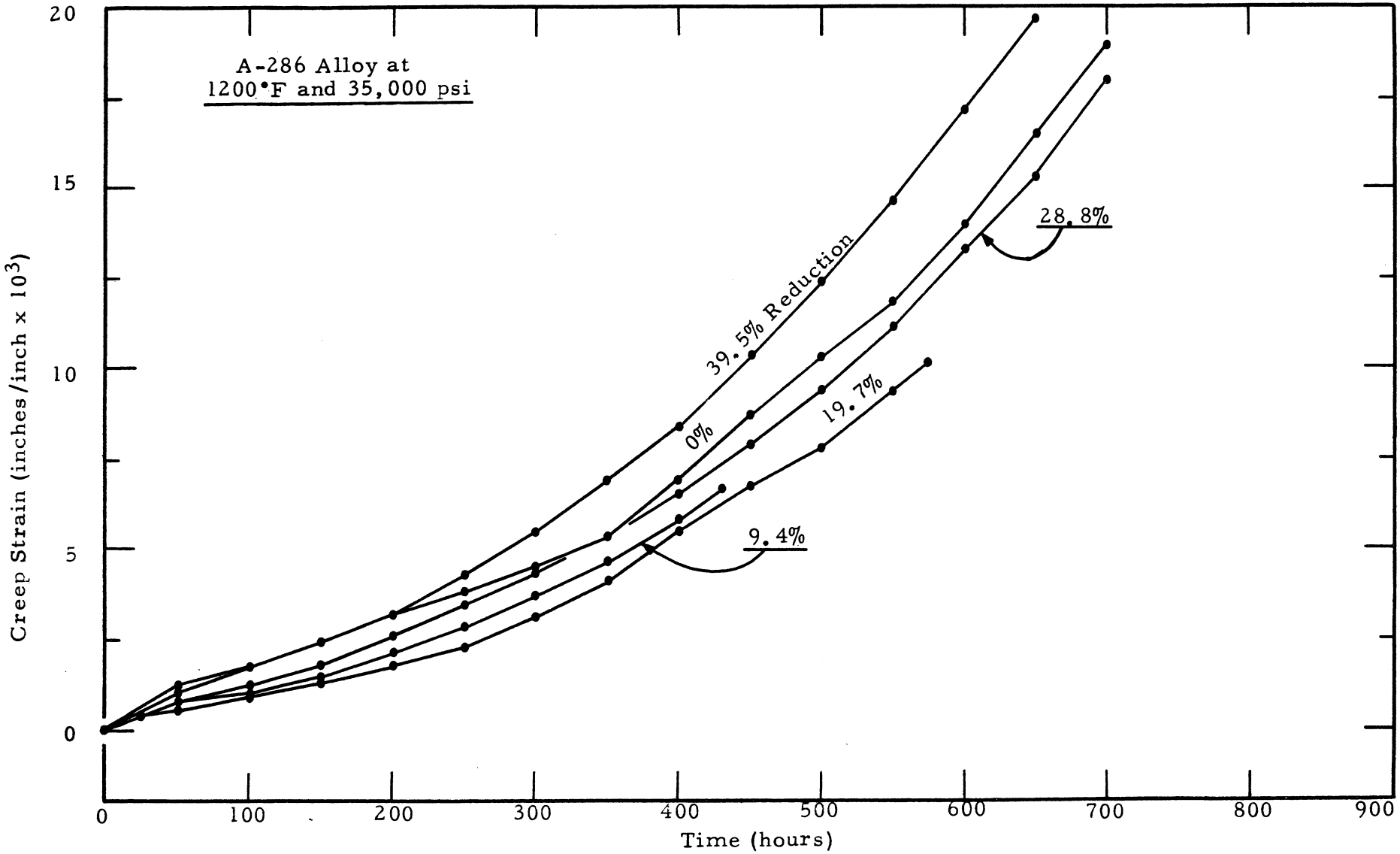


Figure 23. - Effect of Percent Reduction of Area by Rolling on the Creep Curve for A-286 Alloy Rolled at Room Temperature, Heat Treated, and Creep Tested at 1200°F and 35,000 psi.

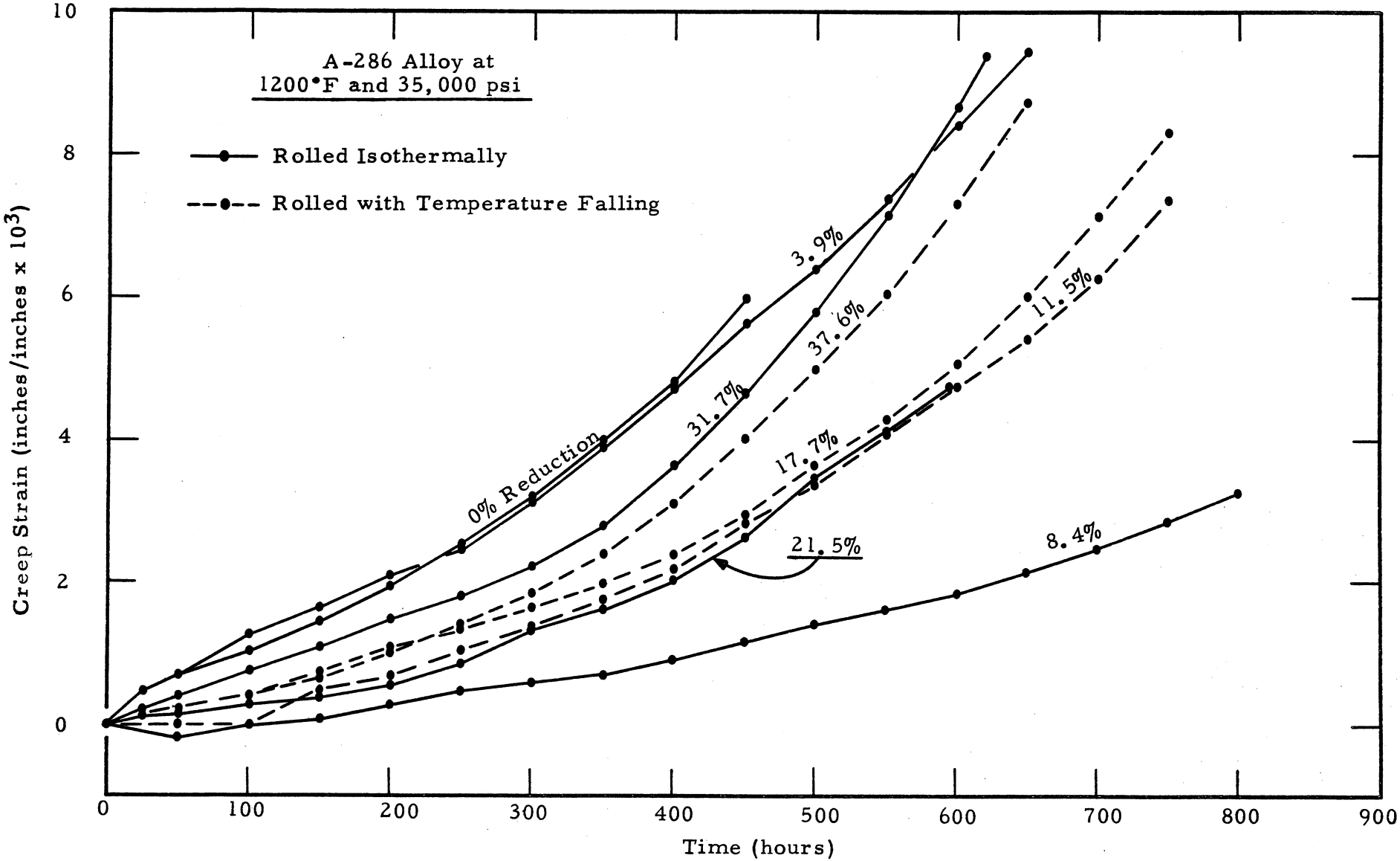


Figure 24. - Effect of Percent Reduction of Area by Rolling on the Creep Curve for A-286 Alloy Rolled at 1700°F, Heat Treated, and Creep Tested at 1200°F and 35,000 psi.

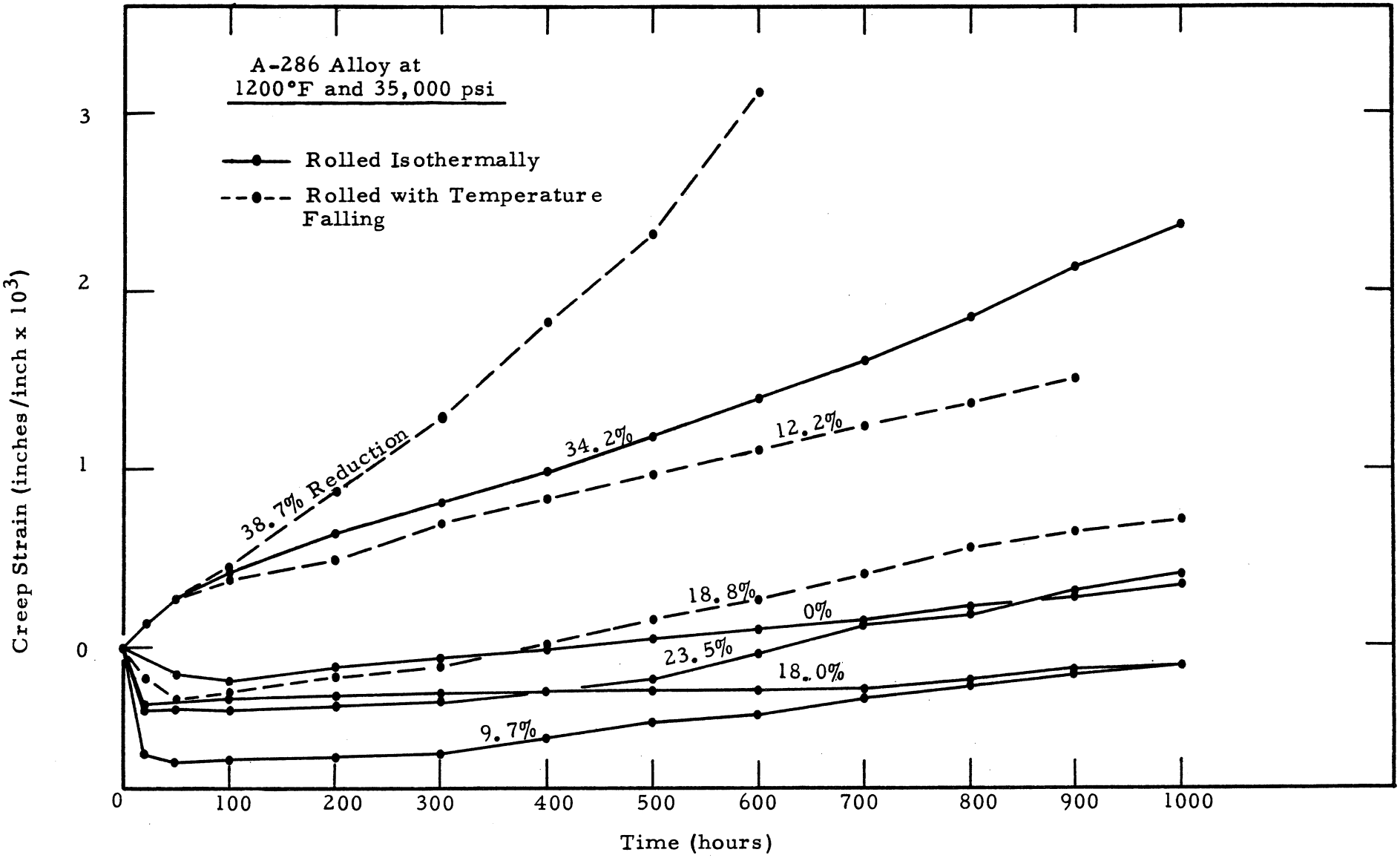


Figure 25 . - Effect of Percent Reduction of Area by Rolling on the Creep Curve for A-286 Alloy Rolled at 1950°F, Heat Treated, and Creep Tested at 1200°F and 35,000 psi.

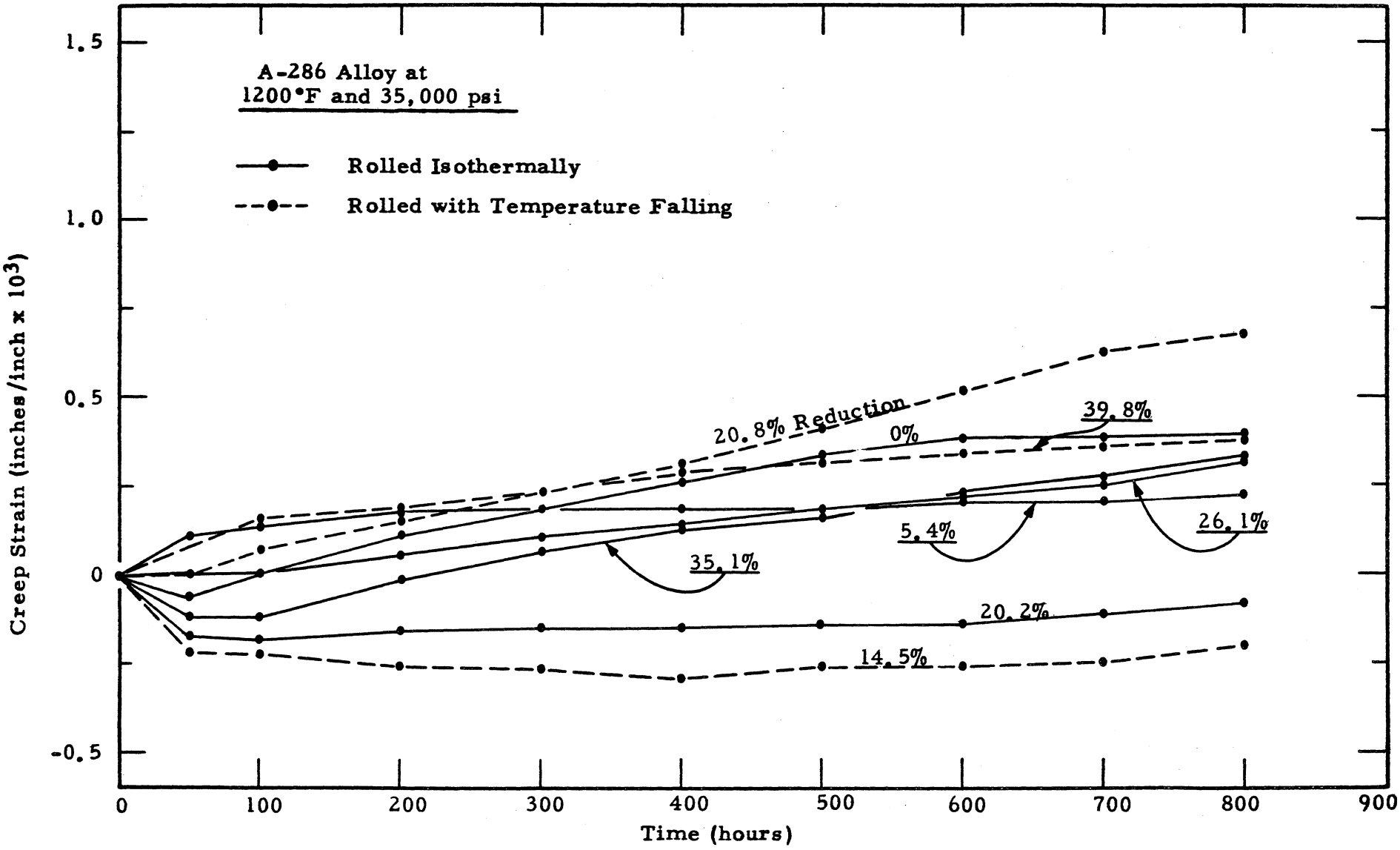


Figure 26 . - Effect of Percent Reduction of Area by Rolling on the Creep Curve for A-286 Alloy Rolled at 2200°F, Heat Treated, and Creep Tested at 1200°F and 35,000 psi.

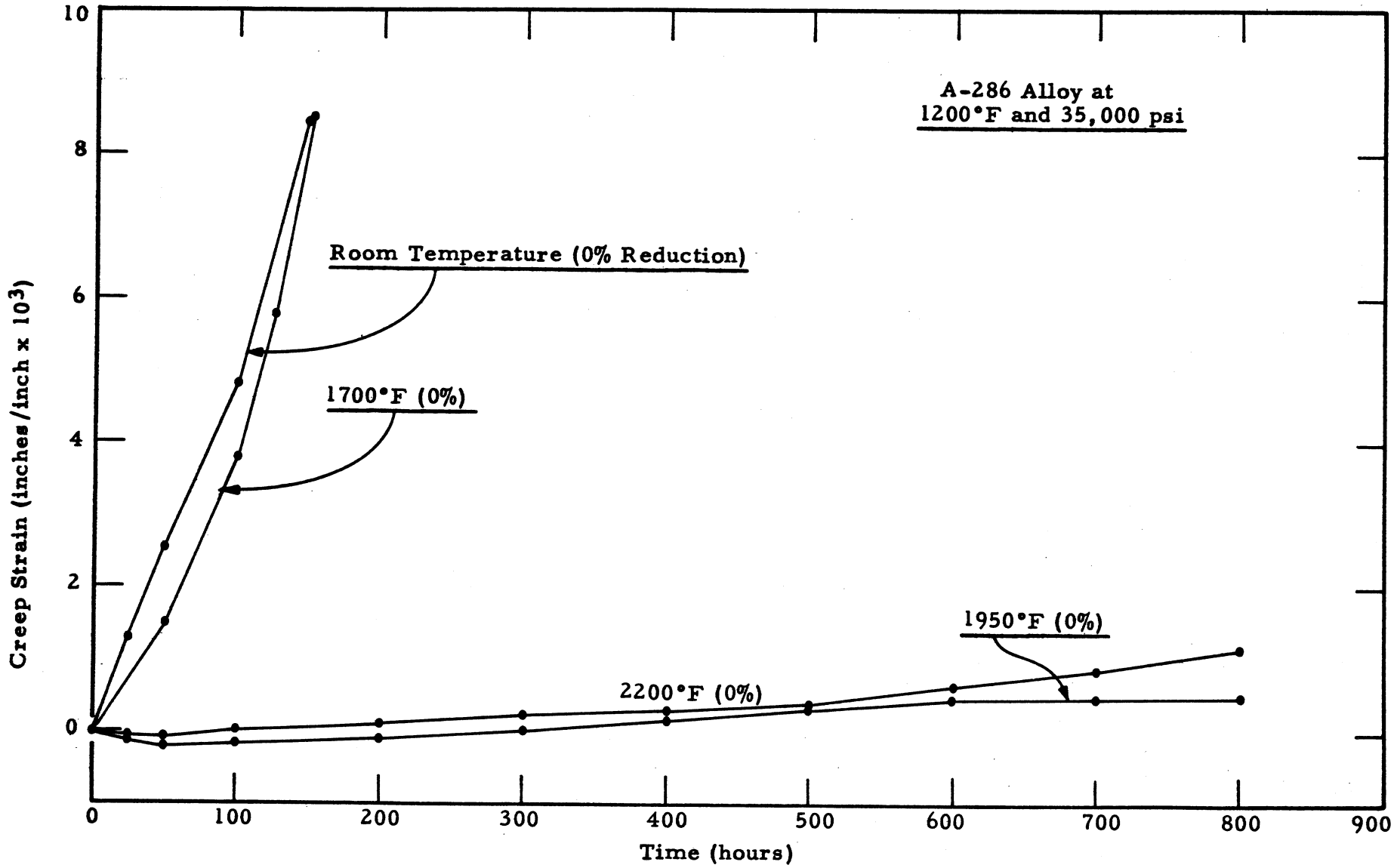


Figure 27. - Effect of Heating Without Rolling on the Creep Curve for Heat Treated A-286 Alloy Creep Tested at 1200°F and 35,000 psi.

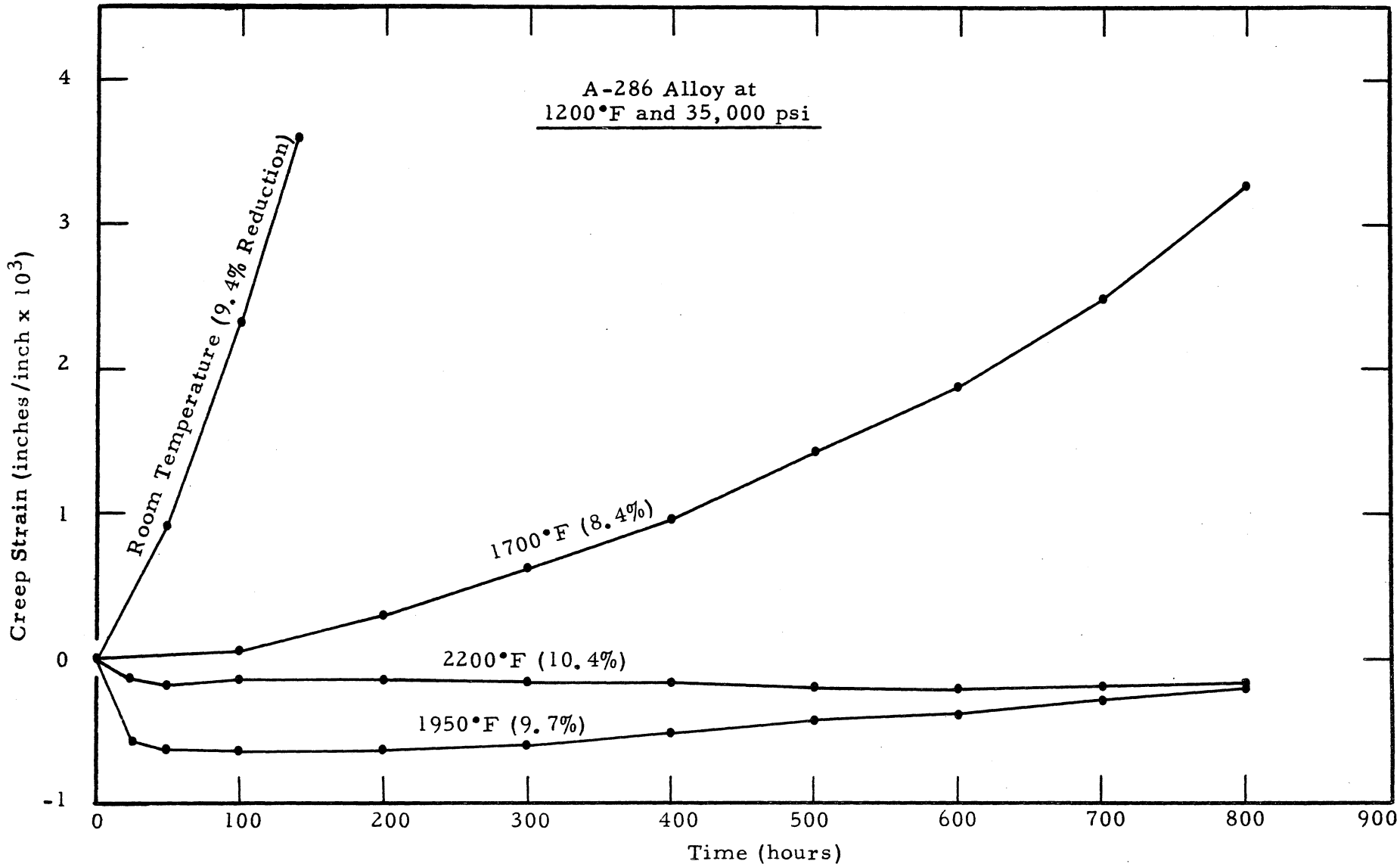


Figure 28. - Effect of Rolling Temperature on the Creep Curve for A-286 Alloy Reduced Isothermally 9.4 (±1.0) Percent, Heat Treated, and Creep Tested at 1200°F and 35,000 psi.

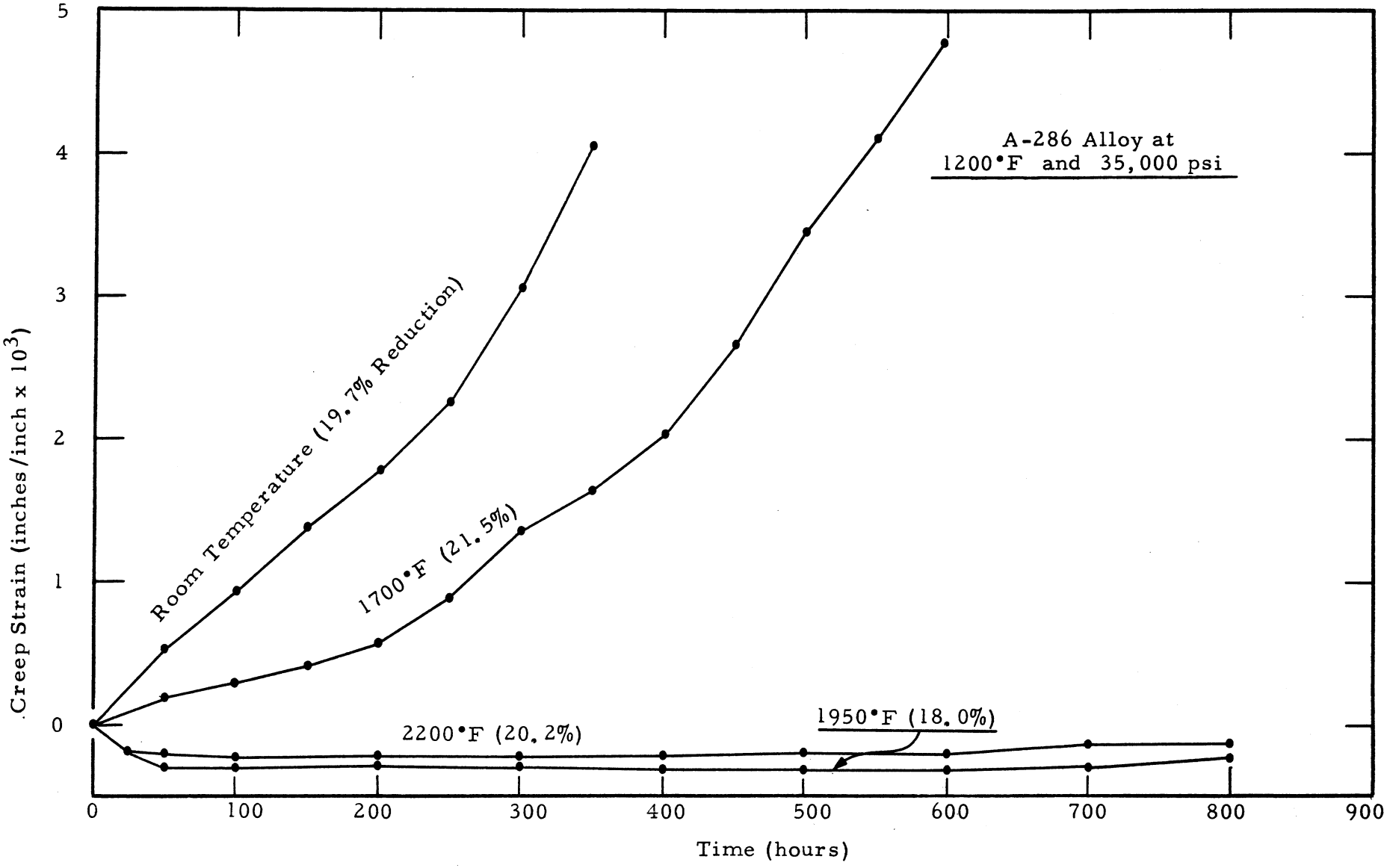


Figure 29 . - Effect of Rolling Temperature on the Creep Curve for A-286 Alloy Reduced Isothermally 20.0 (±2.0) Percent, Heat Treated, and Creep Tested at 1200°F and 35,000 psi.

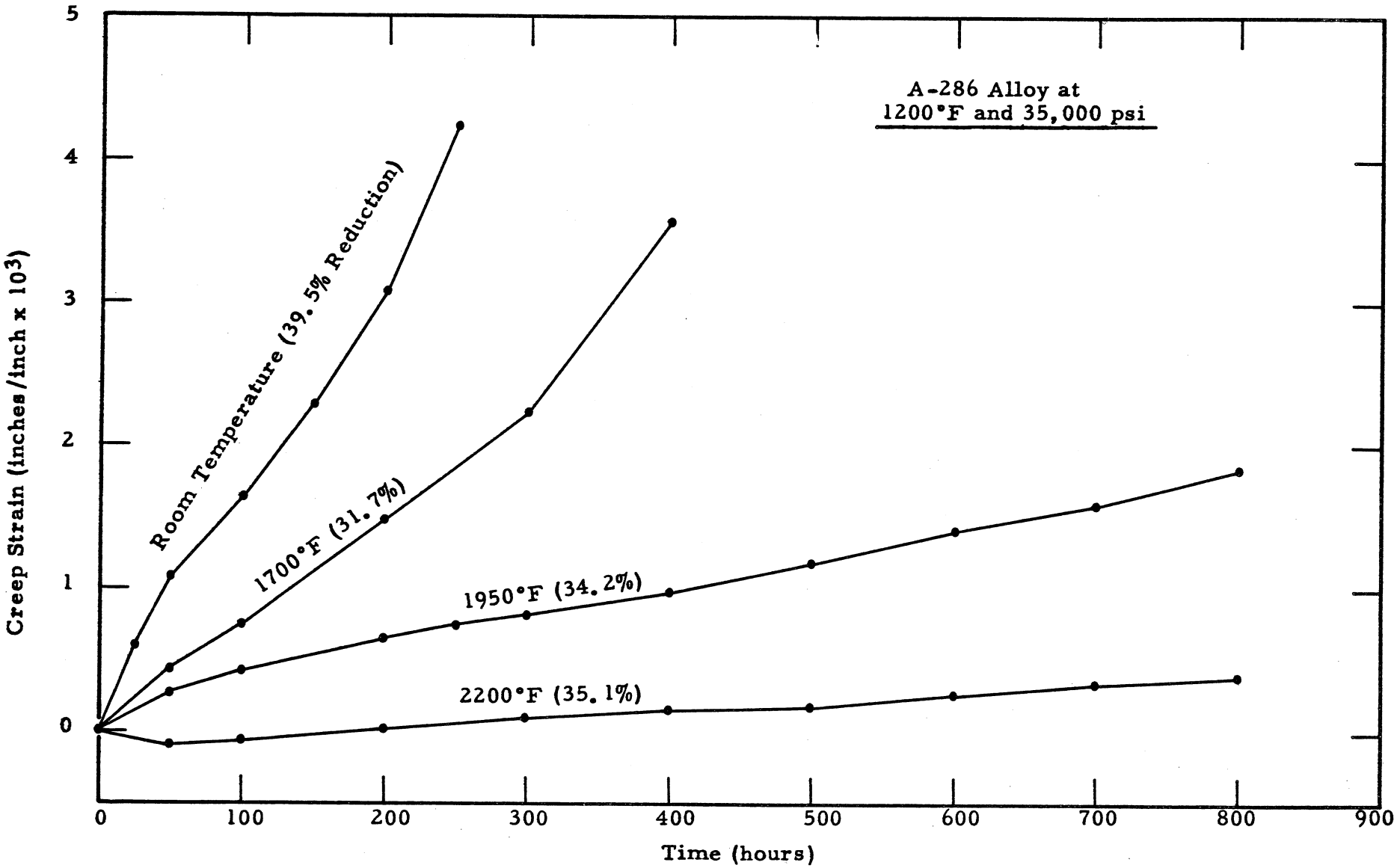


Figure 30 . - Effect of Rolling Temperature on the Creep Curve for A-286 Alloy Reduced Isothermally 35.7 (±4.0) Percent, Heat Treated, and Creep Tested at 1200°F and 35,000 psi.

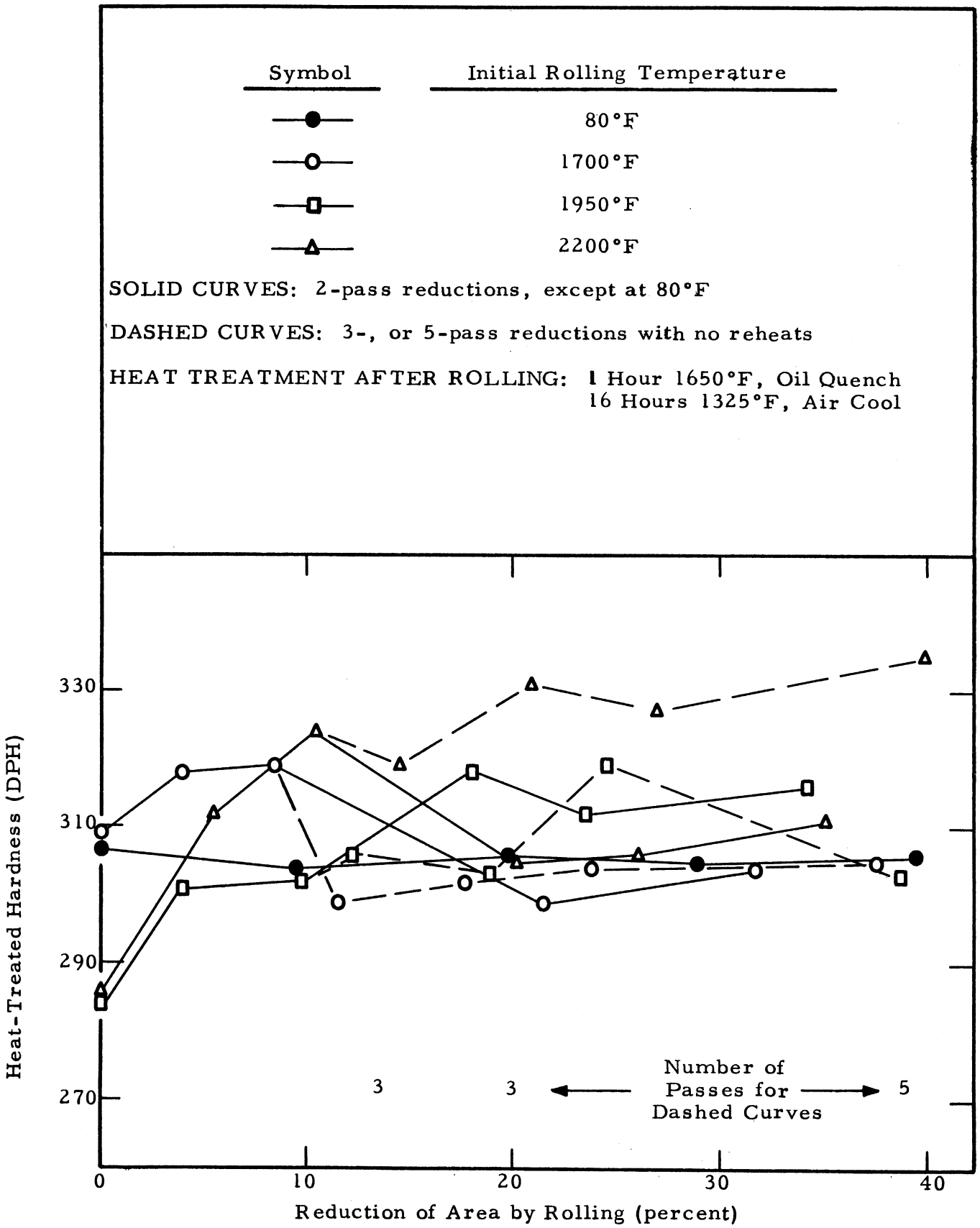
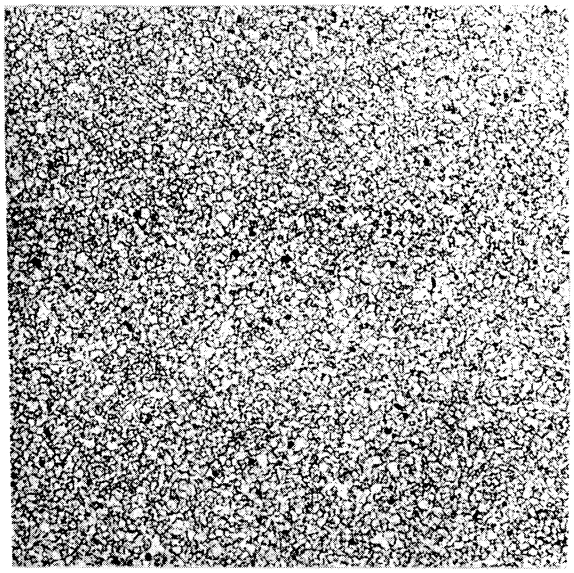
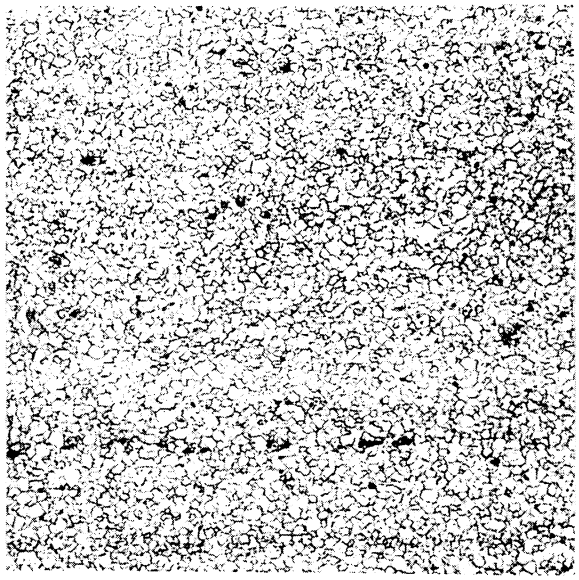


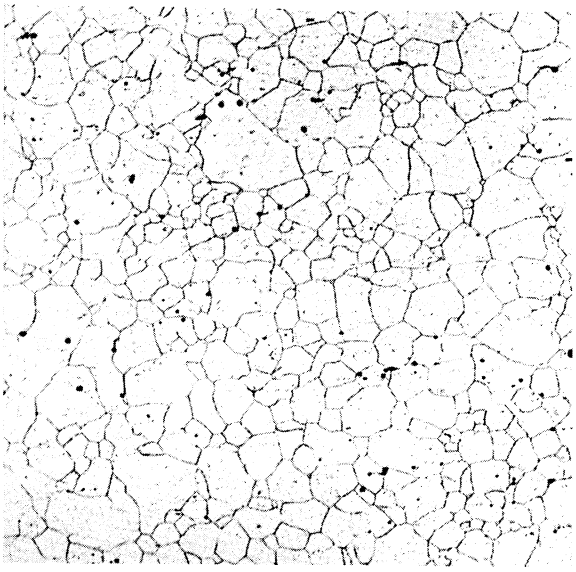
Figure 31. Effect of Prior Working Conditions on the Heat-Treated Hardness of A-286 Alloy.



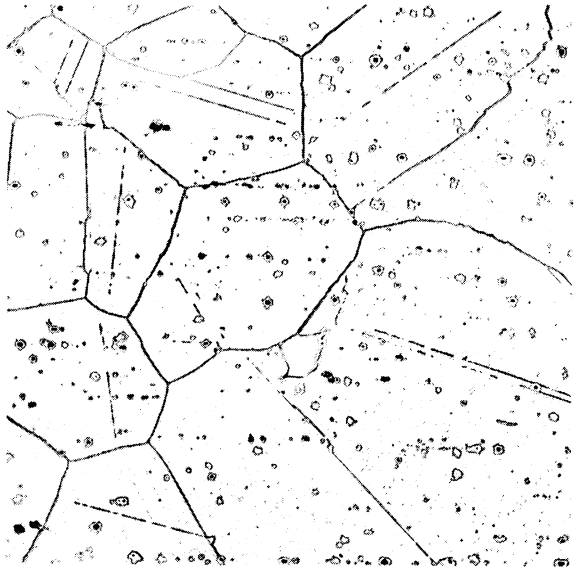
(a) Solution Treated and Aged.



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



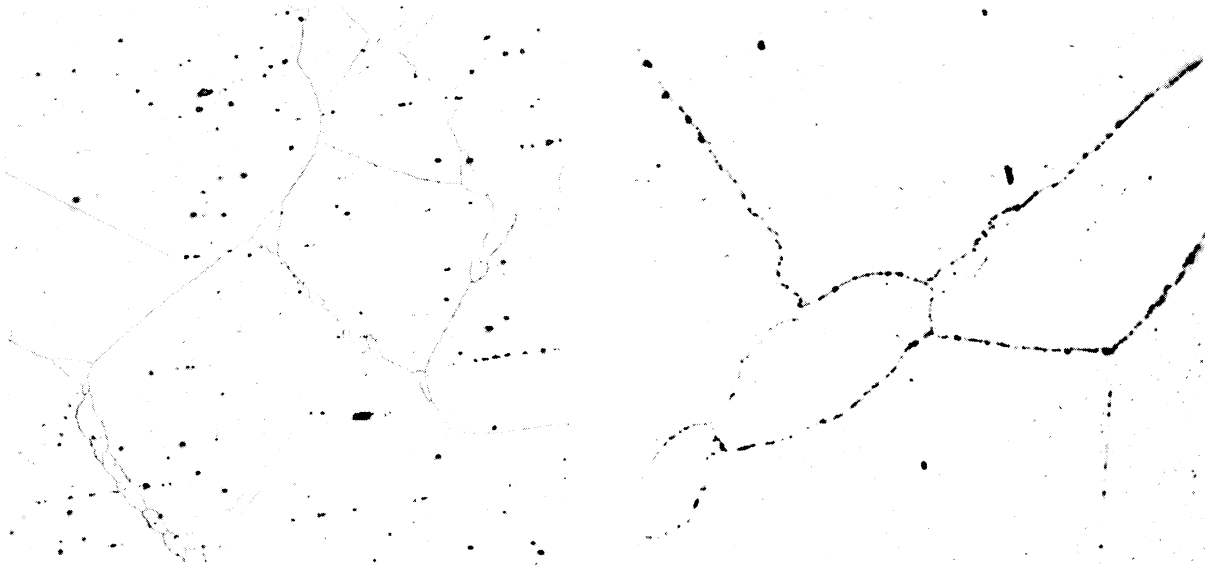
(c) 45 Minutes at 1950°F, Air Cooled + Solution Treated and Aged.



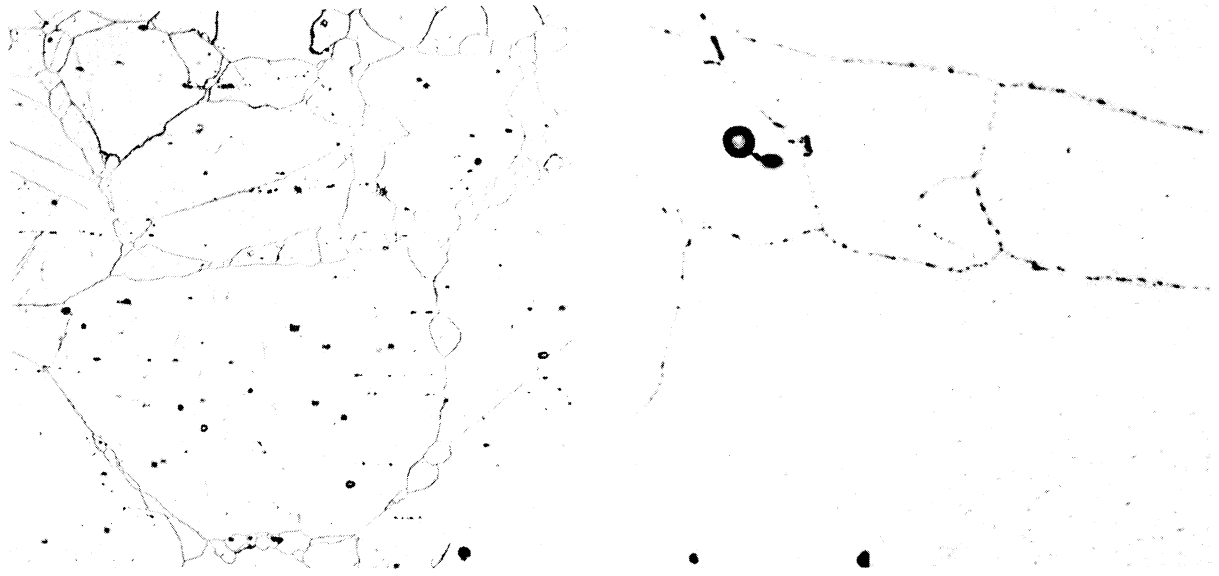
(d) 45 Minutes at 2200°F, Air Cooled + Solution Treated and Aged.

Figure 32. - Microstructure 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. Magnification X100D.

X100 X1000



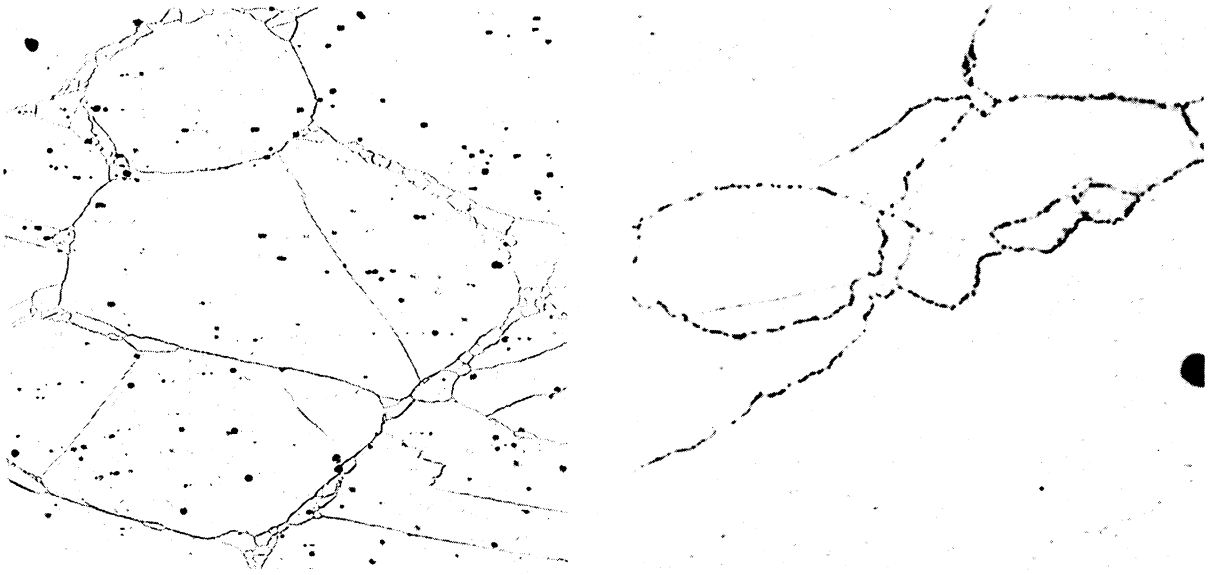
(a) Heated 45 Minutes at 2200°F, Hot Rolled 5.4 Percent in 1 Pass, Air Cooled + Solution Treated 1 Hour at 1650°F and Aged 16 Hours at 1325°F.



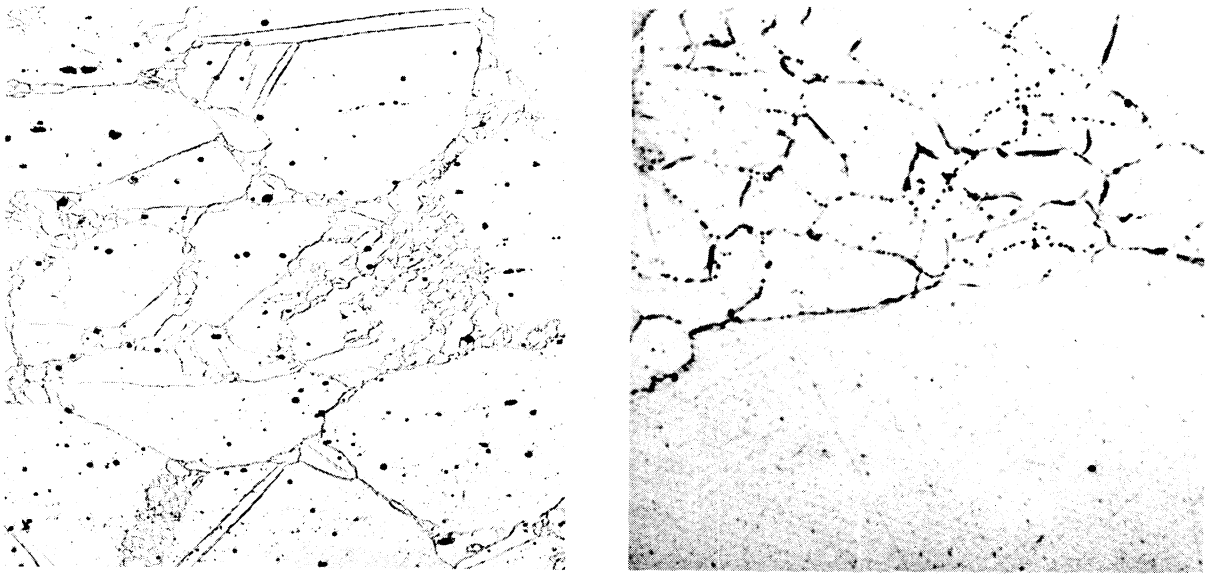
(b) Heated 45 Minutes at 2200°F, Hot Rolled 14.5 Percent in 3 Passes, Air Cooled + Solution Treated 1 Hour at 1650°F and Aged 16 Hours at 1325°F.

Figure 33. - A-286 Alloy Hot Rolled from 2200°F and Subsequently Solution Treated and Aged. The hot rolling was (a) 5.4 percent in 1 pass, (b) 14.5 percent in 3 passes, (c) 26.8 percent in 4 passes, and (d) 39.8 percent in 5 passes with no reheats.

X100 X1000



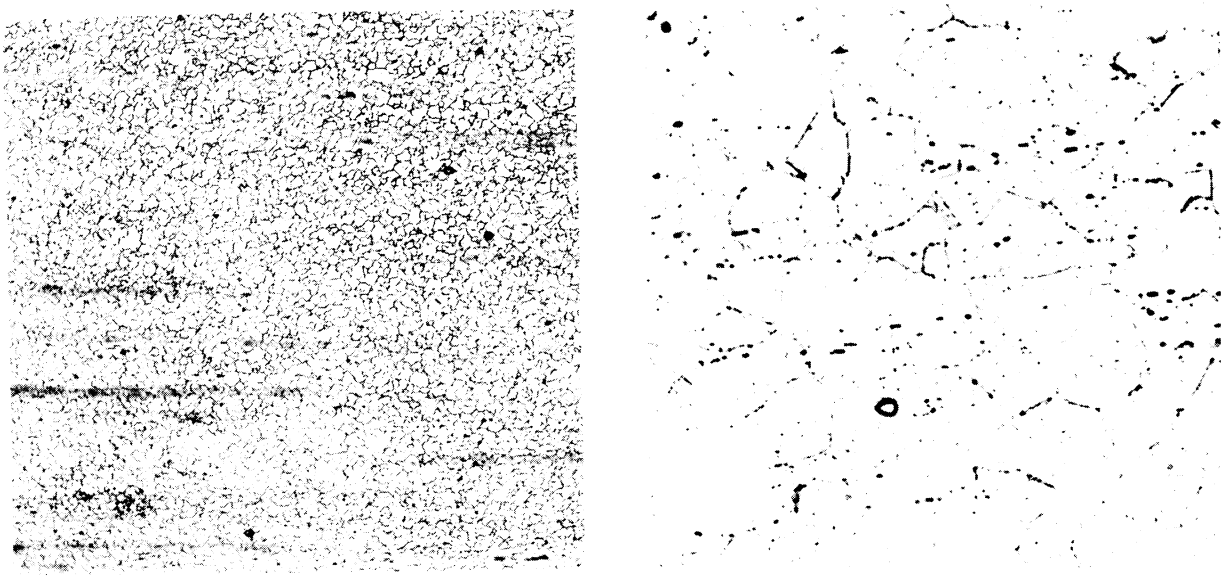
(c) Heated 45 Minutes at 2200°F, Hot Rolled 26.8 Percent in 4 Passes, Air Cooled + Solution Treated 1 Hour at 1650°F and Aged 16 Hours at 1325°F.



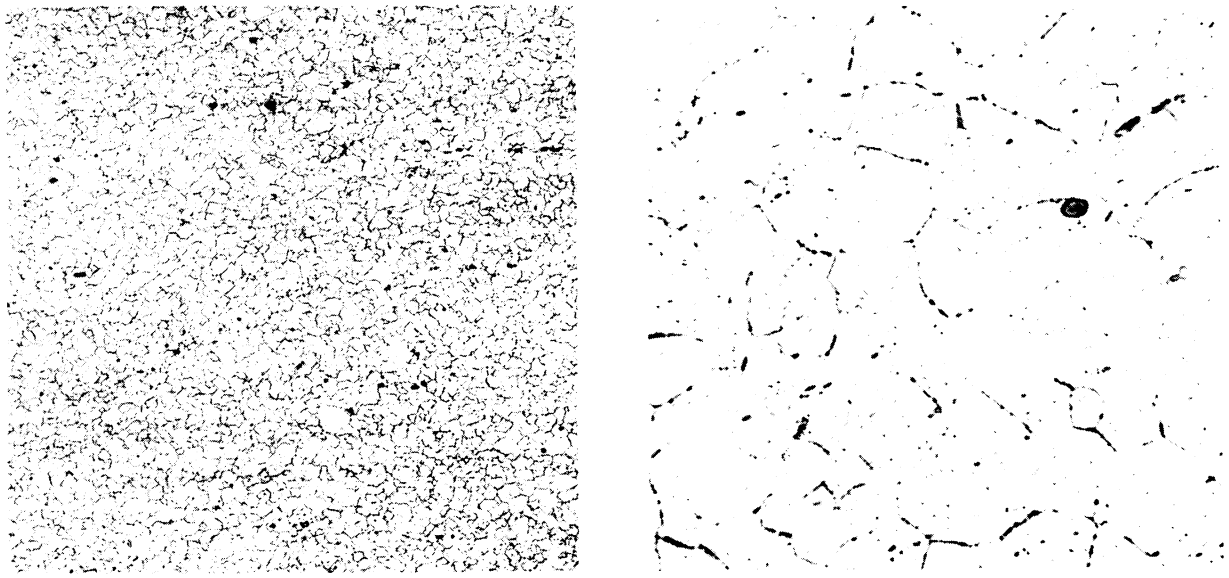
(d) Heated 45 Minutes at 2200°F, Hot Rolled 39.8 Percent in 5 Passes, Air Cooled + Solution Treated 1 Hour at 1650°F and Aged 16 Hours at 1325°F.

Figure 33.- concluded.

X100 X1000



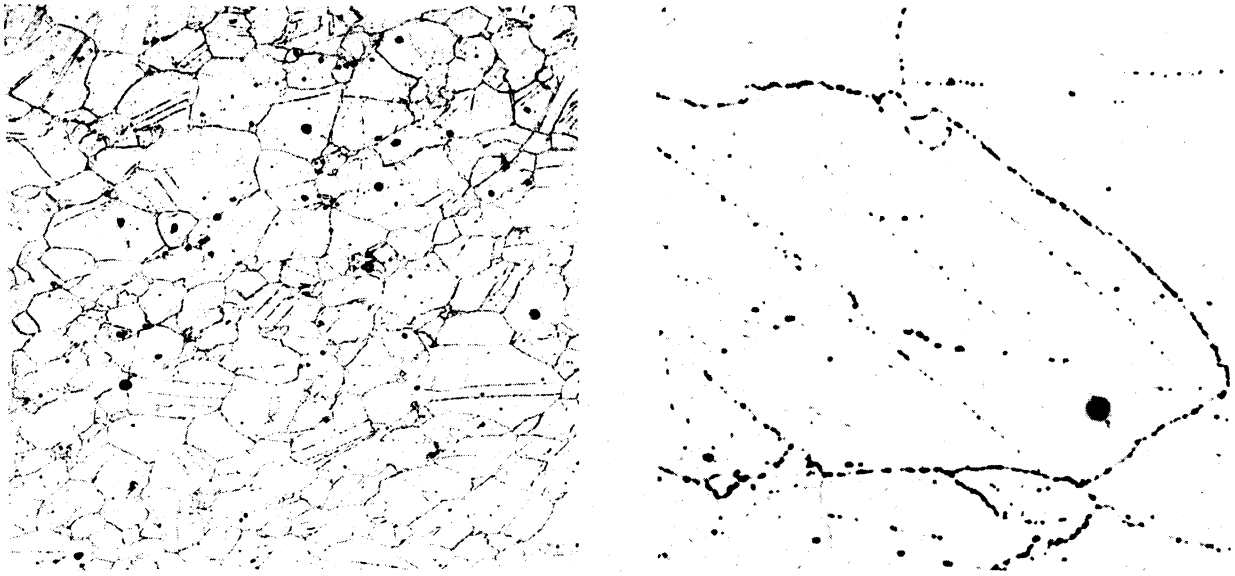
(a) Rolled 19.7 Percent at Room Temperature + Solution Treated 1 Hour at 1650°F, Oil Quenched + Aged 16 Hours at 1325°F, Air Cooled.



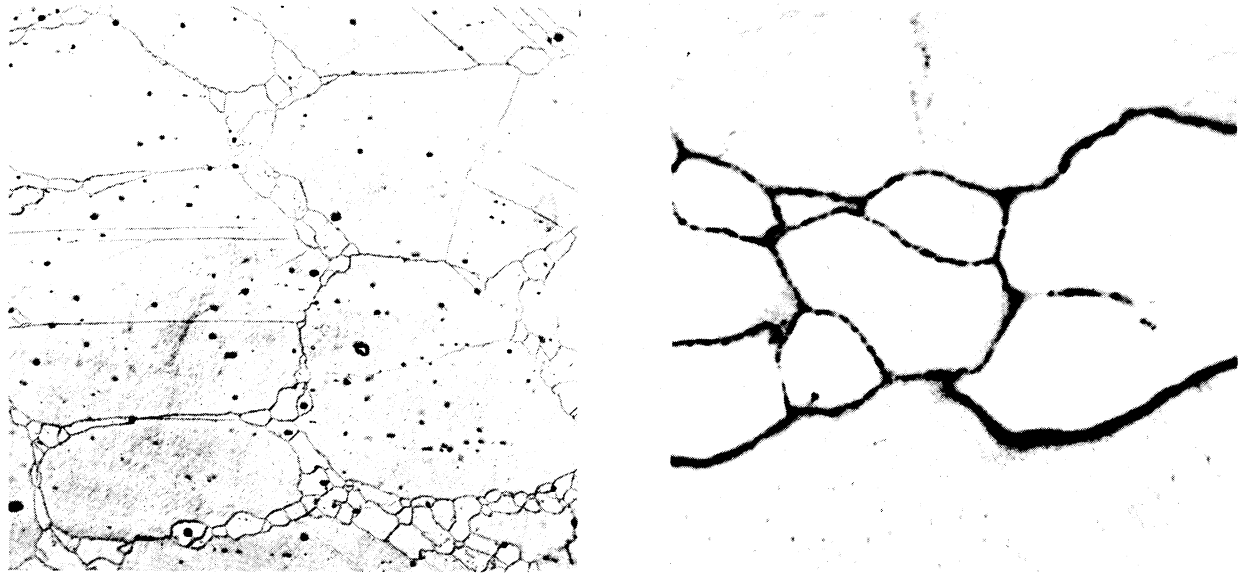
(b) Heated 45 Minutes at 1700°F, Rolled 21.5 Percent in 2 Passes, Air Cooled + Solution Treated 1 Hour at 1650°F, Oil Quenched + Aged 16 Hours at 1325°F, Air Cooled.

Figure 34. - Microstructure of A-286 Alloy Solution Treated and Aged after Rolling 20.0 ($\pm$2.0) Percent at (a) Room Temperature, (b) 1700°F, (c) 1950°F, and (d) 2200°F. Etched electrolytically in 5 percent H_2SO_4.

X100 X1000



(c) Heated 45 Minutes at 1950°F, Rolled 18.0 Percent in 2 Passes, Air Cooled + Solution Treated 1 Hour at 1650°F, Oil Quenched + Aged 16 Hours at 1325°F, Air Cooled.



(d) Heated 45 Minutes at 2200°F, Rolled 20.2 Percent in 2 Passes, Air Cooled + Solution Treated 1 Hour at 1650°F, Oil Quenched + Aged 16 Hours at 1325°F, Air Cooled.

Figure 34. - Concluded.