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DEVELOPMENT OF LOW ALLOY Ti-B STEELS FOR HIGH TEMPERATURE SERVICE APPLICATIONS

CORNELL AERONAUTICAL LABORATORIES

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WRIGHT AIR DEVELOPMENT CENTER

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DEVELOPMENT OF LOW ALLOY Ti-B STEELS FOR HIGH TEMPERATURE SERVICE APPLICATIONS

Cornell Aeronautical Laboratories

April 1952

Materials Laboratory
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FOREWORD

This report was prepared by the Cornell Aeronautical Laboratory, Inc., Buffalo, New York under Contract No. W33-038 AC-21094. Work was initiated as a research and development project identified by Research & Development No. R605-227, High Temperature Alloy Research and Development. It was administered under the direction of the Materials Laboratory, Research Division, Wright Air Development Center, with Lt. M. A. Piekutowski acting as project engineer.

ABSTRACT

Conservation of strategic metals may be accomplished in the production of jet aircraft engine parts provided suitable lean alloy substitutes are made available, capable of operating at elevated temperatures. In this respect, the Ti-B type of ferritic steels have been investigated and their creep and rupture properties evaluated and improved for service temperatures in the neighborhood of 1200°F. It was demonstrated, through testing the properties of a 600 pound heat of 3 Cr- 1 Mo-Ti-B sheet steel with a 2.2 Ti/C ratio, that such a composition not only could be steel mill processed satisfactorily on a semi-commercial basis, but also that 1200°F rupture and creep strength properties, equivalent to the Cr-Ni stainless steels, could be obtained for several hundred hours of life.

A detailed investigation of this steel provided design type creep data for several conditions of heat treatment and hot rolling procedures. Other tests indicated that high hot strength properties could be retained in light gage sheet material provided surface decarburization was minimized during processing. Ceramic coated creep and rupture test specimens of this alloy displayed a life advantage over uncoated specimens at temperatures above 1200°F because of protection against oxidation. Heliarc welded joints of the 3 Cr- 1 Mo-Ti-B sheet steel had inferior hot rupture strength with respect to the parent metal when no filler rod was used. A reasonable approach to parent metal high temperature strengths was obtained with the use of a 347 stainless steel filler rod with the weld bead left on.

As the result of studies made on a variety of compositions of 30 pound laboratory heats of Ti-B steels with varying C, Ti, B, Mo, Cr, W, and V, it was possible to determine the effect of these alloys on the high temperature strengths of the ferritic steels. With increasing quantities of titanium and carbon, marked gain in hot strength was obtained in both the plain C-Ti-B and 3 Cr- 1 Mo-Ti-B steels, provided the Ti/C ratio was maintained in the neighborhood of two to five. Boron variations from 0.010 to 0.10 percent produced no significant effect on the high temperature strength properties at 1200°F. Molybdenum was most effective in improving the hot strength properties of the Ti-B steels while chromium served more in providing resistance to scaling and oxidation. Tungsten and vanadium, added to the 3 Cr- 1 Mo-Ti-B steels to the maximum extent of about 0.15 percent individually or in combination, caused only minor gain in creep and rupture strength properties.

The unusual high temperature strength of the Ti-B steels is

the result of their ability to form a hardened low temperature transformation product of acicular ferrite exceptionally resistant to tempering. This is accomplished as a result of the following:

(a) Solid solution hardening and diffusion interference effects of both boron and titanium.
(b) The retention of titanium and carbon in the supersaturated ferrite.
(c) The dispersion hardening effect of the precipitated titanium carbide.

PUBLICATION REVIEW

Manuscript copy of this report has been reviewed and is approved for publication.

FOR THE COMMANDING GENERAL:

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INTRODUCTION

The utilization of low alloy materials is of timely interest, for example, to those concerned with the specification and procurement of metals for high temperature jet engine service. While a few of the component parts of jet aircraft operate under conditions requiring highly alloyed heat resistant materials, a number of instances exist where conservation of strategic alloys may be realized through lean alloy substitutes. Where the operating temperature is in the neighborhood of 1200°F, the Ti-B type of ferritic steel has been shown to be potentially useful, demonstrating high temperature strengths comparable to the Cr-Ni stainless steels for life periods of several hundred hours.

The Ti-B steels were originally introduced by the Titanium Alloy Manufacturing Division of the National Lead Company and described by Comstock in 1949 (1)*. With the addition of approximately 0.02 percent boron to low carbon titanium steels, it was found possible to appreciably harden such a composition. Its high resistance to softening at temperatures of the order of 1200°F led to the discovery of its unusual hot strength properties. The hot-short characteristics commonly associated with steels containing boron in excess of 0.007 percent were not found in the Ti-B steels where sufficient titanium was present to combine with the carbon and thereby prevent the formation of a low melting Fe-B-C eutectic constituent.

Additional study at the Cornell Aeronautical Laboratory of the Ti-B steels, primarily in the form of sheet stock, revealed the sensitivity of the high temperature strength properties of this alloy to variations in the ratio of the titanium to carbon contents (2). In the case of a series of Cr-Mo-Ti-B heats with 0.05 percent carbon and Ti/C ranging from 0.5 to 7.0, stresses to produce rupture in 100 hours at 1200°F varied from 12,000 to 36,000 p.s.i. with the maximum strength occurring with the Ti/C ratios in the range of 2 to 4. While considerable gain in rupture strength was achieved with the lower Ti/C values, it has proved necessary with these compositions to restrict hot working operations to temperatures below 1850°F to avoid hot shortness.

The disclosure of the outstanding high temperature strength properties obtained from 30 pound laboratory heats of the Ti-B type of steels warranted an extension of this development to the semi-commercial production stage wherein study could be made of the properties

* See bibliography

of the alloy prepared according to steel mill practice. It has been the purpose of the current investigation to examine one such 600 pound heat of 3 Cr- 1 Mo-Ti-B steel produced by the Simmonds Saw Company of Lockport, N.Y. under the sponsorship of the Titanium Alloy Manufacturing Division of the National Lead Company. Of particular interest has been the determination of the specific creep and fracture design stress values of such a mill heat in the form of sheet stock subjected to various heat treatments and processing procedures, including hot rolling, welding, and ceramic coating treatments.

While the initial data obtained have established the high hot strength properties of the Ti-B steels, little information has been available concerning the metallurgical nature of these alloys and the factors predominantly responsible for their behavior. A parallel objective of the present study has been to investigate the effect of composition variations of B, C. Ti, Cr, and Mo on high temperature strength and attempt to systematize these findings into a working theory describing the high temperature strengthening mechanism of the Ti-B steels.

DESCRIPTION OF STEELS

600 Pound 3 Cr- 1 Mo-Ti-B Steel

This semi-commercial heat of steel was prepared in a 600 pound induction furnace at the Simmonds Saw Plant and poured into a single eight-inch square big end up-hot top mold. Deoxidation consisted of the addition of $1\frac{1}{2}$ to 2 pounds of aluminum per ton. Low carbon ferro-titanium and ferro-boron were added after deoxidation to permit proper control of the titanium and boron contents. All working of the ingot was done hot at temperatures in the neighborhood of but not exceeding 1850°F. A 2 x 8 foot slab was formed by forging and this was subsequently rolled and cross rolled to produce 0.090-inch plate approximately 20 inches wide. Additional hot rolling at 1850°F was done by C.A.L. to reduce the 0.090-inch plate to sheet stock of thicknesses varying from 0.030 to 0.075 inch.

The composition of the heat is listed below:

Simmonds Heat No.	C.A.L. No.	C	Mn	Si	Cr	Mo	Ti	B	Ti/C
F-4286	H.T. 73	.065	.34	.28	2.91	1.06	.14	.022	2.2

While the Ti/C ratio of this heat is lower than that recommended for good hot working properties, it was specified because of the much better high temperature strength obtainable as compared to heats with ratios near 10. The fact that sound sheet stock was prepared from this heat would indicate that the lower Ti/C ratios could be handled commercially provided too high forging and rolling temperatures are not used.

Except for material tested in the as-rolled or as-welded condition, the sheet stock was usually given a simple normalizing treatment prior to machining of the test specimens. For normalizing temperatures of the order of 2100°F, recommended to develop the high hot strength properties, a hydrogen atmosphere furnace was used to minimize scaling and decarburization of the sheet material. A conventional air furnace was used for normalizing temperatures of 1900°F and lower.

30 Pound Laboratory Heats

For purposes of investigating the effects of composition variables on the high temperature strength properties of the Ti-B steels, a number of 30 pound laboratory induction furnace heats were used. All heats were prepared at the Research Laboratories of the Titanium Alloy Division of the National Lead Company.

Each 30 pound melt was split three ways and poured into three-inch diameter cylindrical graphite molds at which time the alloying element being varied in a series was adjusted for each ingot. The 10 pound ingots were hammered and received by the Cornell Aeronautical Laboratory in this form. The slabs were hot rolled in the temperature range of 1750 to 1850°F to 0.050 to 0.075 inch sheet stock, requiring approximately 10 to 12 roll passes and reheatings. In a few cases, the stock was pickled and cold rolled from 0.090 inch down to the final thickness before the final normalizing heat treatment. Heat treating procedures used were similar to those described for the semi-commercial 600 pound sheet stock, with a hydrogen atmosphere furnace being employed for 2100°F normalizing temperatures.

Approximately 30 compositions were used in the course of determining the variation of hot strength with changes in C, B, Ti, Ti/C, Mo, Cr, V, and W. Table I itemizes the analyses of these steels arranged in groups of three according to the split heat series. In general, the alloy variations of these steels were specified with the objective of demonstrating the effect produced on the high temperature strength properties by changes in carbon, boron at several carbon levels, titanium and Ti/C at several carbon levels, Ti/C in C-Mo-Ti-B steel, Ti/C in 3 Cr- 1 Mo-Ti-B steel at high carbon, and additions of tungsten and vanadium to the 3 Cr-1 Mo-Ti-B grade.

TEST METHODS AND EQUIPMENT

For determining both the creep and fracture properties of the sheet material, lever loading type machines were used as illustrated in Figure 1. A standard shaped sheet material tension test specimen was dead weight loaded through the 10-1 lever arm with an accuracy of stress determination of at least 1 percent. Suitable precautions in the design of the specimen holder system were taken to minimize friction and specimen bending through the use of knife edges, pin joints, and trunnion assemblies.

All creep and rupture test specimens were 16 inches long by 1 inch wide with a reduced gage length section 0.500 inch wide by approximately 4 inches from shoulder to shoulder. Specimen strips were sheared from the original sheet such that their longitudinal axis was parallel to the rolling direction of the sheet. Final contour of the specimen was achieved by a grinding operation leaving the edges free from a cold worked condition.

TABLE I

CHEMICAL COMPOSITIONS OF 30 POUND LABORATORY HEATS OF Ti-B STEELS

Material	C	Mn	Si	B	Ti	Cr	Mo	V	W	Ti/C	T.A.M. No.	C.A.L. No.
	.03	.41	.39	.023	.06					2.0	1502-1	64
	.05				.24					4.8	-2	65
C-Ti-B	.06				.87					14.5	-3	66
Increasing Ti or	.08	.34	.34	.010	.27					3.4	1595-1	80
Ti/C with constant	.09				.94					13.3	-2	81
C and B at three	.08				1.65					20.6	-3	82
C levels.	.14	.16	.17	.024	.31					2.2	1596-1	83
	.14				1.12					8.0	-2	84
	.14				1.74					12.4	-3	85
C-Ti-B	.03	.51	.44	.022	.04					1.3	1503-1	67
Increasing C and Ti	.05				.12					2.4	-2	68
with constant Ti/C and B.	.12				.17					1.4	-3	69
C-Ti-B	.04	.48	.28	.028	.07					1.8	1511-1	70
Increasing B with				.059							-2	71
constant C and Ti				.12							-3	72
at two Ti or Ti/C	.05	.34	.24	.018	.27					5.4	1594-1	77
levels.	.04			.044	.26					6.5	-2	78
	.05			.10	.22					4.4	-3	79
C- 1 Mo-Ti-B	.06	.30	.31	.018	.085		1.13			1.4	1593-1	74
Increasing Ti or Ti/C	.05				.29					5.8	-2	75
with constant C and B	.04				.44					11.0	-3	76
3 Cr- 1 Mo-Ti-B	.20	.48	.19	.031	.27	2.63	.93			1.4	1597-1	86
Increasing Ti or Ti/C with	.12				.92					7.7	-2	87
constant B and high C.	.13				1.46					11.3	-3	88

TABLE I (Cont'd.)

CHEMICAL COMPOSITIONS OF 30 POUND LABORATORY HEATS OF Ti-B STEELS

Material	C	Mn	Si	B	Ti	Cr	Mo	V	W	Ti/C	T.A.M. No.	C.A.L. No.
3 Cr- 1 Mo-Ti-B + V	.15	.49	.45	.025	.36	3.13	1.04			2.4	1627-1	90
Increasing V with con-	.14				.33			.06		2.4	-2	91
stant C, Ti, and B	.15				.35			.12		2.3	-3	92
3 Cr- 1 Mo-Ti-B + W	.14	.45	.46	.028	.35	3.20	1.00		.08	2.5	1628-1	93
3 Cr- 1 Mo-Ti-B + W	.13				.35				.14	2.7	-2	94
3 Cr- 1 Mo-Ti-B + W + V	.16				.34			.10	.17	2.1	-3	95

Temperature over the two-inch specimen gage length was maintained within 3°F of the nominal value and continuously recorded on a multipoint recorder system. Frequent checking of the specimen test temperature was made with a precision potentiometer connected successively to each of the three thermocouples attached to the specimen gage length.

Strain measurements were made by means of extensometers attached to the reduced section of the specimen and working in conjunction with a cantilever SR-4 strain gage detector. The strain reading, represented electrically by the unbalance in a bridge circuit, is obtained either as a manual reading or continuously recorded on a Dynalog type strain recorder. The gage permitted a reading sensitive to a length change of 0.00005 inch with the over-all accuracy of the multiple switching and recording system equivalent to a strain of approximately 0.0002 over the two-inch gage length.

In tabulating and plotting the time-elongation creep data, total permanent deformation was the value of strain considered. Thus the elastic strain obtained during initial loading and the thermal expansion resulting from heating the specimen to the test temperature are not included in the permanent deformation quantities referred to in the charts and discussion of this report.

600 POUND HEAT OF 3 Cr- 1 Mo-Ti-B STEEL

Metallurgical Characteristics

In the course of testing the Ti-B type of steels, the need for proper heat treatment has been noted in order to develop the inherent high temperature strength properties of this material. In general, sufficiently high hardening temperatures must be used not only to gain full austenitization but also to promote maximum solution of carbides. Full hardness was obtainable in the sheet thicknesses used by air cooling from the normalizing temperature.

For purposes of establishing the particular response to heat treatment of the 600 pound material, a number of one-inch square specimens of 0.050-inch sheet stock were quenched from various temperatures and their room temperature hardness determined. Figure 2 is a plot of the results obtained and indicates that full hardness is attained at approximately 1700°F. At austenitizing temperatures above 1700°F, it is likely that additional solution of carbides occurs but is not reflected in higher room temperature hardness values due to the compensating effect of increased grain size.

Microstructures representative of the sheet alloy normalized from 1700, 1900, and 2100°F are shown in Figure 3. Not only is the grain coarsening with increased temperature evident but also the tendency towards the formation of a transformation product more acicular in nature.

Effect of Heat Treatment on High Temperature Strength

It was of interest to determine the importance of the austenitizing temperature of the 3 Cr- 1 Mo-Ti-B steel with respect to its high temperature strength properties. With the hardness vs: temperature data of Figure 2 as a guide, 0.050 inch sheet material suitable for creep and rupture testing was normalized from the temperatures of 1700, 1900, and 2100°F. The stress-time to rupture results obtained at temperatures of 1100, 1200, and 1400°F are represented in Figure 4. A very pronounced increase in high temperature rupture strength was associated with increasing normalizing temperature. Fracture ductility decreased markedly in progressing from the 1700 to 1900°F treatment with little further drop occurring with the 2100°F temperature.

While the most prominent microstructural difference apparent in the photomicrographs of Figure 3, which represent the three normalizing treatments, was that of grain size, it is unlikely that this variable accounts for the rupture strength differences. As a matter of fact at the temperatures of 1100 and 1200°F for the range of rupture times considered, it would be expected that fine grain size would give the better rupture life in this type of steel. It is quite likely that the most contributing factor to the better rupture strength properties of the material normalized from 2100°F as compared to 1700°F is the solution of greater quantities of titanium, boron, and carbon in the austenite resulting in stronger ferrite after cooling. In the case of one specimen normalized from 1600°F and tested at 1200°F, a most inferior rupture life resulted, since very little of the hardening constituents were dissolved in the partially austenitic steel at this normalizing temperature.

In commercial practice it would undoubtedly be more convenient and economical to utilize the sheet material in the as-rolled condition without further normalizing or additional heat treatments. For this reason, some indication was obtained of the creep and rupture properties of this alloy in the hot rolled condition. Since the 2.2 Ti/C ratio of the 600 pound heat material permitted a maximum rolling temperature of about 1850°F, full advantage normally would not be taken of the alloying constituents because of the partial solution accomplished at the relatively low rolling temperature of 1850°F. Accordingly, two hot rolling procedures were used to provide sheet material for test. In both cases, 0.092-inch stock was rolled to 0.075 inch in one pass at 1850°F. However, one set of the 0.092-inch material was preheated to 2100°F for 10 minutes in H_2 and hot rolled after cooling to 1850°F while the other was heated directly to 1850°F. Figure 5 compares the microstructures resulting from these two treatments. The relatively coarse grain size characteristic of the 2100°F austenitizing temperature was only partially broken down by the 1850°F hot rolling operation. The room temperature hardness of the two differently treated hot rolled sheets was about the same and only slightly higher than the normalized material thus indicating little residual cold working remaining in the material.

The hot rolled material preheated at 2100°F before rolling at 1850°F had definitely superior creep and fracture strength to that heated directly to the 1850°F hot rolling temperature (Figure 6). Approximately the same creep and rupture strengths were obtained at the 1200°F temperature level for the hot rolled stock preheated to 2100°F before rolling as for that given a subsequent normalizing treatment. The hot rolled material because of grain refinement does display a ductility advantage with respect to the normalized stock for the same rupture strength level. These results further confirm the desirability of using the 2100°F solution treatment. Where various processing operations such as hot rolling or ceramic coating must be performed at temperatures below 2100°F, the stock should first be preheated to 2100°F and then cooled to the processing temperature if maximum hot strength properties are required.

Room temperature tensile properties for the 3 Cr-1 MO-Ti-B sheet steel in three conditions of heat treatment are represented in the following tabulation:

TABLE II

ROOM TEMPERATURE TENSILE PROPERTIES OF
600 POUND HEAT OF 3 Cr-1 Mo-Ti-B SHEET STEEL

Condition	Ultimate Strength P. S. I.	Yield Strength 0.2% P. S. I.	% Elongation In 2 Inches
Normalized 1900°F	127,500	105,500	5.0
Normalized 2100°F	124,000	96,500	7.0
Preheated at 2100°F Hot Rolled at 1850°F	143,000	110,000	9.0

Better room temperature ductility at comparable strength levels was obtained in the hot rolled material relative to the normalized stock as was the case at elevated temperatures.

Creep and Fracture Design Data

A sufficiently detailed number of high temperature creep and rupture tests were made with 0.050-inch sheet stock prepared from the 600 pound heat of 3 Cr- 1 Mo-Ti-B steel to define limiting creep and rupture stresses at temperatures of 1100, 1200, and 1400°F for times ranging from several minutes to several hundred hours. Such data are presented for two initial normalizing temperature treatments of 1900 and 2100°F in Figures 7 and 8.

It is of interest to compare some of the hot strength properties obtained for the 600 pound Ti-B heat at 1200°F with those available for several austenitic Cr-Ni stainless steels in sheet form. The possibility of substitution for the richer alloy steels is apparent from Table III.

Effect of Gage Thickness on High Temperature Strength

It had been noted previously (2) that the high temperature strength properties of Ti-B steel sheet were quite sensitive to gage thickness. This peculiarity has been investigated with the objective of determining the cause of such variation so that the high strength properties of this alloy might be retained in the lighter gages of the order of 0.030 to 0.050 inch.

A series of rupture tests was conducted at 1200°F with sheet stock from the 600 pound heat hot rolled to the sizes of 0.081, 0.072, 0.050, and 0.036 inch. All test stock after finish rolling was heat treated at 2100°F in H_2 for 10 minutes and air-cooled. The results shown in Figure 9 were obtained and as replotted in Figure 10 reveal the downward trend of rupture strength with decreasing sheet thickness.

Two possibilities were considered as contributing to this size effect; namely, oxidation of the specimen during the course of the high temperature test and original surface composition changes produced during the rolling and heat treating operations. Either or both effects, being related to the steel surface, would influence the over-all strength to a greater extent in the thinner sheet since the surface deterioration would constitute a greater percentage of the cross section in the lighter gage material. To investigate these possibilities, 0.050-inch specimens were prepared from 0.064-inch stock by grinding approximately 0.007 inch from each side. The usual 2100°F heat treatment was performed on the 0.064-inch sheet prior to grinding of the surfaces.

The time-for-rupture results for the surface ground specimens

TABLE III

COMPARISON OF CREEP AND RUPTURE PROPERTIES AT 1200°F FOR
600 POUND HEAT OF 3 Cr- 1 Mo-Ti-B STEEL AND SEVERAL Cr-Ni STAINLESS STEELS

	P.S.I. Stress For							
	1% Creep In				Rupture In			
Material	1 Hr.	10 Hrs.	100 Hrs.	300 Hrs.	1 Hr.	10 Hrs.	100 Hrs.	300 Hrs.
600 Pound Heat 3 Cr- 1 Mo-Ti-B 2100°F - H_2 - 10 Min. AC	50,000	40,500	22,000	15,000	52,100	42,000	28,000	17,500
(Type 302) 18 Cr- 8 Ni Annealed	17,000	16,500	16,200	16,000	38,000	31,000	21,000	17,000
(Type 316) 18 Cr-13 Ni- 2 Mo Annealed	17,000	17,000	16,500	16,500	37,000	33,000	28,000	25,500

are spotted on the stress-time chart of Figure 9 for comparison with the sheet material with the as-heat treated surfaces. The fact, that the 0.050-inch material, ground from the 0.064-inch stock, displayed strengths significantly greater than the regular 0.050-inch stock and approximately equal to the 0.072-inch material, indicated that the size effect was due primarily to a composition or strength gradient in the cross section of the as-heat treated material. The data further showed that surface oxidation or scaling, occurring during the rupture test, was not a contributing factor to the size effect for the time and temperature conditions of interest.

The surface deterioration was further investigated by means of microhardness surveys and metallographic examination. The variation in hardness across the sections of the various thicknesses is represented in Figure 11 and clearly indicates a gradation of properties from center to edge. The most likely cause of this cross-sectional strength gradient is decarburization resulting from the hydrogen atmosphere heat treatment of scale bearing stock. In the 0.036-inch material, decarburization extended completely through the cross section as indicated by the relatively low peak hardness at the center of the section. In the material that is hot rolled only, decarburization is not severe as indicated by the cross-sectional hardness valuation.

In Figure 12 are shown the unetched edges at 200X of the 3 Cr-1 Mo-Ti-B steel representative of several methods of processing. During the manual hot rolling procedure used in the Laboratory, requiring a number of reheating operations, an appreciable scale layer forms as shown in the top photomicrograph. The middle photo shows that when the scaled steel is heat treated in a hydrogen atmosphere the iron oxide is reduced to form a porous iron layer. The water vapor, which is a product of this reaction, is effective in producing the decarburization previously noted. Where the hot rolled sheet is descaled by pickling prior to hydrogen heat treatment the iron oxide reduction and formation of water vapor is minimized resulting in little decarburization.

The decarburization of the Ti-B steels is undesirable from two standpoints. It has been established that the hot strength properties, at temperatures in the neighborhood of 1200°F are dependent upon both the carbon content level and the Ti/C ratio. Loss of carbon from the original analysis causes both of these factors to change in the direction of poorer creep and rupture strength. It is understandable, therefore, why the high temperature strength properties of Ti-B steels, which have been subjected to decarburizing conditions, should be sensitive to section size. On the basis of these findings, it was possible to prepare 0.030-inch sheet stock with improved high temperature rupture properties, as illustrated in Figure 13, by limiting the extent of surface decarburization during rolling and heat treating.

In reviewing the laboratory methods which were used to process the 600 pound heat of 3 Cr- 1 Mo-Ti-B steel plate of 0.092 inch to the usual test sheet size of 0.050 inch, it should be recalled that the decarburizing conditions have been severe. For this reason, the results obtained on the heavier sheet sizes of 0.070 to 0.080 inch are more representative of the inherent strength of this material. In commercial practice, where the sheet could be prepared in a continuous rolling operation without repeated heatings, considerable advantage would be had. Utilization of the sheet stock in the hot rolled condition without subsequent heat treatment would further minimize the effect of surface decarburization and thus permit the production of thin gage material with approximately the same high temperature strength properties as the thicker sizes.

Effect of Welding on High Temperature Strength

Because of the frequent use of welding in the fabrication of jet components from sheet stock material, an evaluation was made of the high temperature rupture properties of the 3 Cr- 1 Mo-Ti-B steel in the welded condition. Two sets of test panels were prepared from 0.050-inch sheet stock normalized from 2100°F using a Heliarc welding method with a copper back up strip to minimize the formation of a heat-tempered zone.

One set of weld panels was prepared without filler rod by flanging the joined edges about 3/32 inch and fusing with the Heliarc torch. Test strips were cut at right angles to the weld bead so as to locate the weld joint at the center of the specimen gage length. The weld bead was gound flush with the surface to produce rupture test specimens as shown in Figure 14. A decline in rupture strength at both 1100 and 1200°F was observed for the welded stock as compared to the parent material as shown in Figure 15. Failure occurred predominantly in the edge of the weld metal (Figure 16). This, together with the fact that reheat treatment of the welded stock did not improve its hot strength properties, proved that the decline was not due to a heat-tempered zone. It is possible that a composition change occurred in the weld joint leading to an alteration in the Ti/C ratio and therefore lowering of rupture strength.

In a second series of weld tests, panels were butt welded together, using a filler rod of 347 stainless steel. Fracture again occurred in the weld metal zone (Figure 17) both in those cases where the weld bead was left on and where ground flush. The rupture strength properties at 1200°F were again inferior to the parent material as shown in Figure 18. However, for rupture life above approximately 100 hours, the welded stock with the bead left on closely approached the strength properties of the parent metal.

In spite of the fact that the welded joint high temperature rupture strengths were not equivalent to the parent material, they were still retained at a respectably high level. Inasmuch as failure was confined to the weld metal in both cases, this situation conceivably could be bettered through trial of additional filler rod materials.

Effect of Ceramic Coating on High Temperature Strength

Because of the lean alloy content of the Ti-B steels, they are not notably resistant to corrosion or oxidation at high temperatures. At 1200°F for time durations of the order of several hundred hours, air oxidation is not prohibitive for sheet stock material of the 3 Cr- 1 Mo-Ti-B composition. However, at higher temperatures scaling becomes a significant factor in causing premature failure. Ceramic coatings have been used satisfactorily on other low alloy steels to provide protection against surface attack and accordingly an attempt was made to determine the effect of a ceramic coating procedure on the hot strength properties of the 600 pound heat of 3 Cr- 1 Mo-Ti-B steel.

Four creep test specimens, machined from 0.050-inch sheet stock normalized from 2100°F, were ceramic coated at the University of Illinois according to the detailed procedure of the reference laboratory report (3). Coating Number 327-1, with a thickness of approximately 0.004 inch, was applied with a dip treatment of 4 minutes at 1900°F. Maturing temperatures of at least 1900°F would be desirable for use with this alloy in order to obtain the high hot strength properties potentially available.

Three of the four ceramic coated specimens were tested at 1200°F and one at 1400°F, providing the creep and rupture data shown in Figure 19. Comparison curves, representing results from uncoated sheet stock heat treated at 1900°F are included in the chart. At 1200°F, where scaling conditions are not too severe for the several hundred hour time periods, no advantage was shown by the ceramic coated material. It is also of interest to note that no deterioration resulted from the 1900°F ceramic coating process relative to the 1900°F heat treated bare material. The one specimen tested at 1400°F would indicate that a definite gain in rupture time is had from the ceramic coating treatment as the result of preventing premature failure due to oxidation.

A qualitative indication of the extent of surface protection provided by the ceramic coating is shown in the photograph of Figure 20. The coating was fairly well preserved after fracturing of the 1200°F specimens even after 336 hours at load and stretching 4.5 percent. The flaked appearance of the 1400°F specimen resulted partially from the required handling of the specimen following its rupture.

C-Ti-B Steel With Varying Ti or Ti/C and Constant C and B at Three Levels of C.

The importance of the Ti/C ratio in affecting the high temperature strength properties of the Ti-B steels had been noted previously in the case of the 3 Cr- 1 Mo-Ti-B grade. To examine this variable more systematically, a series of plain carbon steels with Ti-B additions were used. At each of three carbon levels a split heat was prepared with three different titanium or Ti/C contents to give the nine compositions 64-65-66, 80-81-82, and 83-84-85 shown in Table I.

In order that the various compositions could be properly prepared for high temperature testing and the hot strength results intelligently interpreted, some familiarization was first obtained of the metallurgical characteristics of these steels. Their response to heat treatment, as measured by room temperature hardness and appearance of microstructure, correlated in a qualitative way with the Fe-C-Ti phase diagrams of Tofaute and Büttinghause (4) as shown in reference (5). Sufficient carbon of the order of 0.05 percent was found to be a necessary minimum to insure a reasonable extent of austenitization during heating and ample hardenability during air cooling of sheet material. Because of the reaction of titanium with carbon to form titanium carbide and also the inherent effect of titanium itself to restrict the gamma field in iron, relatively high titanium contents were found undesirable from a heat treating standpoint.

The hardness versus heat treating temperature chart of Figure 21 and the photomicrographs of Figure 22 illustrate the case for increasing titanium at the 0.08 percent carbon level. With the composition of the highest titanium content of 1.65 percent, practically no transformation occurs and the microstructure remains as alpha iron + titanium carbide at temperatures up to 2100°F. With the 0.94 percent titanium, partial transformation and hardening develop, while fairly complete austenitization and solution of carbides is attained at 2100°F for the 0.27 titanium composition. On the basis of such observations, all creep and rupture test specimens were heat treated by normalizing at 2100°F.

A comparison of the high temperature creep and rupture results obtained at 1200°F (Figures 23, 24, and 25) for the three series of steels, indicates the general superiority of those compositions with the lower titanium contents. In the case of the split heat series, with the lowest carbon contents, the split with the lowest titanium displayed inferior hot strength primarily because of its low carbon of 0.03 percent and thus its inability to respond to heat treatment.

With only three titanium contents available at each carbon level, a quantitative relationship was not defined for the variation of high temperature strength with titanium for a given carbon. However, from the data of Table IV which includes results from several other C-Ti-B heats used in this investigation, it was concluded that maximum high temperature strengths are obtained with such titanium contents which give Ti/C values in the neighborhood of 3 to 6. Where too much titanium is present, the steel cannot be appreciably strengthened by heat treatment nor can the ferrite matrix retain any strength at elevated temperatures due to the rapid precipitation of both titanium and carbon as titanium carbide. If the titanium content is too low, insufficient strengthening of the ferrite is obtained from solid solution hardening and optimum dispersion of titanium carbide.

C-Ti-B Steel With Increasing C and Ti at Constant Ti/C and B

The creep and fracture results obtained from the previously described series of heats with three ranges of carbon, indicated improved properties with increasing carbon, particularly with respect to retention of 1200°F rupture strength to longer periods of time. This trend was further investigated with a split heat of three compositions with carbon contents of 0.03, 0.05, and 0.12 percent. Titanium increased correspondingly in each case to maintain the Ti/C ratio approximately two. While a Ti/C ratio nearer 4 to 5 would have given higher over-all strength values, this was not appreciated at the time of preparation of this test heat series. The 2100°F normalized microstructures together with their respective hardnesses as shown in Figure 26, conform to the trends previously noted for these compositions. With increasing carbon, the ferrite structure becomes more acicular in nature with practically 100 percent austenitization having occurred in the 0.12 carbon steel.

The 1200°F creep and fracture results of Figure 27 indicate the superiority of the higher bearing carbon and titanium compositions of this series. With these results, together with those obtained from the other C-Ti-B heats, it was possible to demonstrate more effectively the relationship of high temperature rupture strength with increasing titanium and carbon contents. Such data were plotted in Figure 28 for those heats in which the Ti/C ratio was similar (1.3 to 3.4). A definite upward trend in the stresses required to produce rupture in 100 hours was found with increasing carbon and titanium. As carbon increases, larger quantities of titanium are permissible for a given Ti/C ratio without excessively depleting the ferrite transformation product of carbon due to precipitation. The resulting microstructure benefits thereby from the greater quantity of dispersed carbides as well as from the maximum quantity of titanium and carbon retained in the ferrite solid solution.

TABLE IV

EFFECT OF Ti/C RATIO ON 1200°F RUPTURE STRENGTH AT DIFFERENT CARBON LEVELS IN C-Ti-B STEEL

H.T. No.	Composition				Stress (PSI) For Fracture At 1200°F In	
	C	Ti	B	Ti/C	10 Hrs.	100 Hrs.
70	.04	.07	.028	1.8	16,000	10,000
68	.05	.12	.022	2.4	14,000	10,500
65	.05	.24	.023	4.8	31,000	11,000
77	.05	.27	.018	5.4	22,000	9,000
66	.06	.87	.023	14.5	16,000	10,500
80	.08	.27	.010	3.4	21,000	13,000
81	.09	.94	.010	13.3	16,000	11,000
82	.08	1.65	.010	20.6	17,500	10,000
83	.14	.31	.024	2.2	21,000	16,500
84	.14	1.12	.024	8.0	17,000	12,000
85	.14	1.74	.024	12.4	18,000	12,000

C-Ti-B Steel With Varying B and Constant C and Ti at Two Ti (or Ti/C) Levels

In spite of the small amounts of boron used in the Ti-B steels, the original work reported by Comstock (1) demonstrated its need in developing the unusually high hot strength properties of this grade of steel. Two series of split heats were prepared for this investigation to determine the effect of boron variations on the high temperature strength. In both series, boron varied from approximately 0.010 to 0.10 percent with one set at a Ti/C value of 1.8 (70, 71, and 72) and the other at about 5 (77, 78, and 79).

The results shown in Figures 29 and 30 indicate the 1200°F creep and fracture strengths to be relatively insensitive to the range of boron contents investigated, particularly for times of the order of 100 hours. It would appear that whatever strengthening effect is contributed by the boron addition is fully realized in the steels containing as little as 0.010 percent with no further gain being experienced with boron increasing to 0.12 percent. Further testing would be warranted to the extent of determining the hot strength properties of sheet steels containing boron in the range of 0 to 0.010 percent.

The minor variations of creep and rupture strength with boron variation was surprising in view of the noticeable microstructural differences observed in these series of steels. While the acicular ferrite matrix structure appeared normal and alike in a given boron series, increasing quantities of an unknown constituent appeared with increasing boron. This was revealed to best advantage in the unetched condition as shown in Figures 31 and 32. Most likely this is a low melting eutectic constituent responsible for the hot-short characteristics of certain grades of the Ti-B steels. Mention has been made in the literature of an Fe_2B-Fe_3C eutectic (6, 7).

In spite of the profuse quantity and network arrangement of the low melting constituent in the higher boron steels, no loss in room temperature ductility was noted as measured by bend tests on the sheet stock material. Ninety degree bend reversals were made with 0.050-inch sheet first to one side and then the other around a diameter of approximately twice the sheet thickness. All three of the boron compositions of heats 70, 71, and 72 sustained six bend reversals before failure indicating the network eutectic constituent to be fairly ductile at room temperature.

C- 1 Mo-Ti-B Steel With Varying Ti or Ti/C and Constant C and B

In progressing from the plain C-Ti-B steels to the 3 Cr- 1 Mo-Ti-B grade which has shown excellent high temperature strength in both the laboratory and 600 pound semi-commercial heats, a series was studied containing the molybdenum addition without chromium. Three different titanium additions were made to the split heat to provide Ti/C values of 1.4, 5.8, and 11 (No. 74, 75, and 76). All three compositions yielded approximately the same room temperature hardness with the microstructures as shown in Figure 33 when normalized from 2100°F.

The stress-time for rupture and time-for-1 percent-deformation curves of Figure 34 display the typical effect of Ti/C ratio as previously identified for the other grades of Ti-B steels. While all three compositions approach a common strength level at the short time end of the scale, the major difference exists in their respective abilities to retain this high strength with extended time of exposure to temperature. In the higher titanium bearing splits, degeneration occurs more rapidly because of the mass-action accelerated precipitation of carbon and titanium from the ferrite matrix. It is quite likely that at longer times than those used for testing, the three compositions would again approach a common strength level corresponding to complete tempering of all three of the microstructures.

The importance of molybdenum in contributing to the 1200°F strength properties of the Ti-B steels can readily be appreciated from the comparison data available. With the 1 percent molybdenum addition to the plain C-Ti-B steels, the stress for rupture in 100 hours at 1200°F is raised from approximately 11,000 p.s.i. to 22,000 p.s.i., a percentage gain equivalent to that resulting from the Ti-B addition to carbon steel. This value of molybdenum in improving the hot strength properties of ferritic steels has been previously recognized and full advantage of it should be made in formulating lean alloy steel compositions for high temperature service.

3 Cr- 1 Mo-Ti-B Steel With Varying Ti or Ti/C and Constant C and B

All previous study of the 3 Cr- 1 Mo-Ti-B steels, prepared as experimental laboratory heats and as the 600 pound semi-commercial heat, had been made with compositions having carbon contents in the neighborhood of 0.05 to 0.06 percent. Since subsequent work has indicated, in the case of the plain C-Ti-B steels, that their high temperature strength properties were improved with increased carbon, a split heat series of 3 Cr- 1 Mo-Ti-B steel (86, 87, and 88) was prepared at a higher carbon level with three titanium contents. Although the carbon of split 86 was higher than intended, it was possible to note trends due both to carbon and Ti/C variation.

The microstructures of Figure 35 indicate again the tendency for suppression of austenite formation during heat treatment with the higher titanium compositions. In the 0.20 carbon-0.27 titanium composition, practically full austenitization was attained at the 2100°F heat treatment resulting in the formation of a low temperature transformation product of substantial hardness.

The 1200°F creep and rupture results of Figure 36 show the marked hot strength superiority for the lower Ti/C ratio heats as was found to be the case for the 3 Cr- 1 Mo-Ti-B steels of lower carbon. In comparing the 1200°F rupture strength of the lower Ti/C ratio split of this group with available data from other 3 Cr- 1 Mo-Ti-B heats of comparable Ti/C but lower carbon, the advantage of increased carbon is evident. (See Table V) This advantage is particularly significant at the longer time periods. With the higher carbon composition, the 3 Cr- 1 Mo-Ti-B grade becomes equivalent to the better stainless steels for times out to 1000 hours rather than the several hundred hours previously obtained with the low carbon 3 Cr- 1 Mo-Ti-B steels. It would appear from the results obtained with both the C-Ti-B and the 3 Cr- 1 Mo-Ti-B steels, that the carbon level should be raised from the usual 0.05 percent to the neighborhood of 0.15 percent if maximum high temperature strength is desired.

3 Cr- 1 Mo-Ti-B Steel With Additions of V and W

Further investigation of possible gain in high temperature strength properties was made using the 3 Cr- 1 Mo-Ti-B steel with approximately 0.15 percent carbon and Ti/C ratio of two as a base material. Additions of vanadium and tungsten were tried to the extent of 0.12 percent vanadium and 0.17 percent tungsten as listed for heats 90 through 95 in Table I. Good hardenability was displayed by all heats together with similar suppressed transformation microstructures as shown in Figure 37.

While the rupture data, summarized in Figure 38 indicated a minor gain resulting from the vanadium and tungsten additions when compared to the plain 3 Cr- 1 Mo-Ti-B split heat of this series, no significant over-all improvement of hot strength was obtained. Since vanadium and tungsten are both strong carbide forming elements, it is possible that the ratio of Ti/C of two was not optimum for maximum rupture strength properties. Further investigation of the effects of vanadium and tungsten would be desirable in the 3 Cr- 1 Mo-Ti-B steels using other titanium or Ti/C contents as well as higher solution temperatures for heat treatment.

TABLE V

COMPARISON OF 1200°F RUPTURE STRESSES FOR SEVERAL 3 Cr- 1 Mo- Ti-B AND Cr-Ni STAINLESS STEELS

Type	C	Ti	B	Ti/C	1200°F Stress For Rupture 10 Hrs.	100 Hrs.	300 Hrs.	1000 Hrs.	Heat No.
3 Cr- 1 Mo-Ti-B	.056	.11	.027	2.0	-	30,000	20,000	15,000	52 *
0.050" sheet	.065	.14	.022	2.2	42,000	28,000	17,500	13,000	73
	.20	.27	.031	1.4	45,000	33,000	28,000	21,000	86
Cr-Ni Stainless	Type 302				31,000	21,000	17,000	15,000	Reference 9
Steels	Type 316				33,000	28,000	25,500	23,000	"
0.050" sheet	Type 347				35,000	28,000	24,000	19,000	"

* Reference (2); Data corrected for sheet thickness of 0.050 inch.

MECHANISM OF HIGH TEMPERATURE STRENGTHENING IN THE Ti-B STEELS

The Ti-B type of ferritic steels have been shown to possess unusually good high temperature strength properties. This is of particular interest in view of the limited amounts of titanium and boron used for alloying purposes. It is possible, through review of the data obtained, to gain some understanding of the metallurgical nature of these alloys and its relationship to the hot strength properties observed.

The formation, by suitable heat treatment, of a low temperature acicular transformation product which is remarkably resistant to tempering is basically responsible for the high temperature strength properties of the Ti-B steels. Ordinarily, martensite or a hardened acicular ferrite structure in steel is associated with extremely low creep strength. However, the martensite of the Ti-B steels differs from the ordinary variety in its high degree of stability when exposed to elevated temperatures. Thus, the inherent strength of such a microstructure may be utilized at temperatures well above those normally capable of producing rapid tempering.

The role of titanium and boron in promoting these characteristics appears to be twofold: First, solid solution strengthening occurs by hindering the diffusion of iron atoms during both the tempering and the deformation process. Secondly, and probably of even greater effect, is the precipitation and dispersion type hardening produced primarily by carbides of titanium (5).

It will be recalled that in the heat treating or processing of the Ti-B steels, sufficiently high austenitizing temperatures must be used in order to realize fully their desirable hot strength properties. This is related to the extent of solution of the titanium, carbon, and boron in the austenite, a reaction requiring temperatures in the range of 1900 to 2100°F for reasonable fulfillment. If the temperature is too low, relatively poor properties are developed since the titanium, boron, and carbon remain ineffective in the form of spheroidized particles of titanium carbide and undissolved boron. A similar situation also results where the titanium content or Ti/C ratio is too great even though high austenitizing temperatures are used, since, with such compositions, the austenite phase is suppressed and the structure remains ferritic at all temperatures.

With the alloying elements of carbon, titanium, and boron properly dissolved and distributed in the austenite, it is possible, during subsequent cooling to room temperature, to develop the characteristic low temperature transformation product displayed by these steels. The X-ray diffraction patterns of Figure 39 for the C-Ti-B split heat series 86-87-88 illustrate the more intense internal lattice strain and supersaturation of the ferrite with increasing quantities of titanium and carbon. The degree of resolution of the $K\alpha_1$ and $K\alpha_2$ doublet lines for the 310 plane further indicates the extent of tempering and relief of lattice strain for these compositions when exposed to 1200°F. The ability of the Ti-B sheet steels to suppress the ferrite transformation to the martensite or lower bainite temperature range upon air cooling is quite striking in view of their low carbon and alloy contents.

Of even greater interest is the high resistance to tempering of this martensitic structure shown by certain compositions - namely, those with the highest creep and rupture strengths. The tempering curves of Figure 40 for the C-Ti-B split heat series of increasing Ti and C content (67-68-69) demonstrate this effect when related to the corresponding creep and rupture properties as shown in Figure 27. In the case of the C-Ti-B steel series with varying boron (70-71-72), where no significant difference in hot strength was found, the rate of softening of all three compositions is equivalent as indicated in the tempering curves of Figure 41.

Thus boron and titanium, when present in suitable proportions relative to carbon, not only serve to suppress the austenite-to-ferrite transformation to sufficiently low temperatures to retain in solution the carbon and titanium in supersaturated quantities but they also aid in preserving this hardened condition upon exposure to high temperatures. The function of boron would appear primarily to be one of diffusion interference, hindering not only the precipitation and coalescence of TiC but also the movement of dislocations and self diffusion of iron during the creep process. In steels of the compositions considered, boron would not be expected to contribute to their good hot strength properties through precipitation hardening (8) since full solubility in the ferrite should have existed for the boron contents used.

While titanium may be effective in improving the creep strength of carbon steel by solid solution hardening, its major effect results from critical dispersion of titanium carbide throughout the ferrite matrix. However, this precipitation process cannot be considered without noting that the carbon used in such a reaction is supplied from within the supersaturated ferrite solution where it exists as an important hardening agent. With too high a titanium content, too much carbon is robbed from the ferrite at an accelerated rate to form coarser and less effective carbides of titanium. With too low a titanium addition, full advantage is not taken of the solid solution and dispersion hardening effects potentially available.

In view of the composition and microstructural equilibriums being sought between the titanium and carbon of these steels when exposed to sub-critical temperatures, it is not surprising that the high temperature strength properties were found to be critically dependent upon the Ti/C ratio. The simultaneous presence of other alloying elements such as Mo and Cr seemed to shift the optimum Ti/C ratio to lower values since these elements, because of their carbide forming tendencies, made some demand on the available carbon.

CONCLUSIONS

In the course of investigating the Ti-B type of ferritic steels for use as lean alloy materials in high temperature service, a number of generalizations have evolved which may be itemized as follows:

1. Compositions of the 3 Cr- 1 Mo-Ti-B type of steels have displayed high temperature creep and rupture strength properties in sheet form equivalent to the Cr-Ni stainless steels at 1200°F for times approaching 1000 hours.

2. These maximum strength properties are obtained with the Ti/C ratio maintained in the neighborhood of 2 to 4. With these Ti/C values, some precautions must be taken to hot work the alloy at temperatures below 1850°F to prevent defects due to hot shortness.

3. It has been demonstrated that the 3 Cr- 1 Mo-Ti-B steel with low Ti/C ratio can be handled in a commercial mill as the result of the processing of a 600 pound heat to sheet material.

4. The high hot strength properties of the Ti-B steels may be obtained to the same extent in heats up to at least 600 pounds as have previously been characteristic of 30 pound laboratory melts.

5. Heat treatment of such steels should be carried out at temperatures in the range of 1900 to 2100°F. As the normalizing temperature increases, better high temperature properties result because of the more complete solution of the titanium, boron, and carbides of the alloying elements.

6. Equivalent hot strength properties with higher ductility

may be obtained from hot rolled sheet without subsequent normalizing. If the rolling operation is conducted at 1850°F, the slab should be preheated to 2100°F and then processed after cooling to the rolling temperature.

7. While some decline in creep and rupture strength may be obtained in light gage Ti-B sheet material, this can be avoided by prevention of surface decarburization during the rolling operation.

8. High temperature failure of Heliarc welded joints of 3 Cr-1 Mo-Ti-B sheet steel has been found to occur within the fusion zone both in the case where no filler rod was used and where 347 stainless steel was tried. Rupture strengths at 1200°F equivalent to the parent metal were not achieved although close approach was made with the 347 filler rod joint when the weld bead was not removed.

9. The 3 Cr- 1 Mo-Ti-B steel may be ceramic coated satisfactorily yielding high temperature strength properties characteristic of the finishing temperature used in the processing. Such coatings are of value in protecting against air oxidation at temperatures near 1400°F although their need is doubtful at 1200°F temperatures for several hundred hours of life.

10. Increasing quantities of carbon and titanium with constant Ti/C of 3 to 6 in the C-Ti-B steels and Ti/C of 2 to 4 in the 3 Cr- 1 Mo-Ti-B grade improves the high temperature properties up to carbon contents of 0.15 to 0.20 percent, the maximum quantities investigated.

11. Boron, varying from 0.10 to 0.010 percent causes negligible change in the creep properties of the C-Ti-B steels at several Ti/C levels in spite of the occurrence of large quantities of a eutectic network appearing in the higher boron steel.

12. Molybdenum was very effective in improving the hot strength properties of the plain C-Ti-B steel, with a 1 percent addition causing a 100 percent increase in 1200°F rupture strength for 100 hours.

13. The phenomenal high temperature strengths of the Ti-B steels is related to their ability to form a hardened low temperature transformation product of acicular ferrite remarkably resistant to tempering. This is the result of solid solution hardening and diffusion interference effects of both boron and titanium, the retention of some carbon and titanium

in the supersaturated ferrite, and the dispersion hardening effect of the precipitated titanium carbide.

RECOMMENDATIONS FOR FUTURE WORK

The many favorable test results obtained on both laboratory and steel mill heats of the 3 Cr- 1 Mo-Ti-B type of steels warrants trial of this alloy in actual jet engine construction. While additional laboratory work may prove desirable to solve specific problems concerned with the production and utilization of this lean alloy material, the most significant results that can be obtained at this stage of the development will be had through actual service operation and testing. It is recommended that a mill size heat of the order of 10 tons, of 3 Cr- 1 Mo-Ti-B steel with 0.15 percent carbon, 0.45 percent Ti, and 0.010 percent boron be fabricated into sheet material for use in such jet engine parts as tail cones and combustion chambers.

Accelerated utilization of this type alloy could be carried further through additional study of optimum methods of processing, particularly as related to methods of welding, joining, surface protection, and hot working techniques. One of the features requiring improvement, from the standpoint of commercial production, is that of the hot working properties in the compositions of the low Ti/C ratios. Since this difficulty is related to the boron addition, it would seem definitely desirable to investigate the effect of boron in these steels at lower quantities than have previously been used. In general the objective should be to obtain good hot working properties at the low Ti/C levels.

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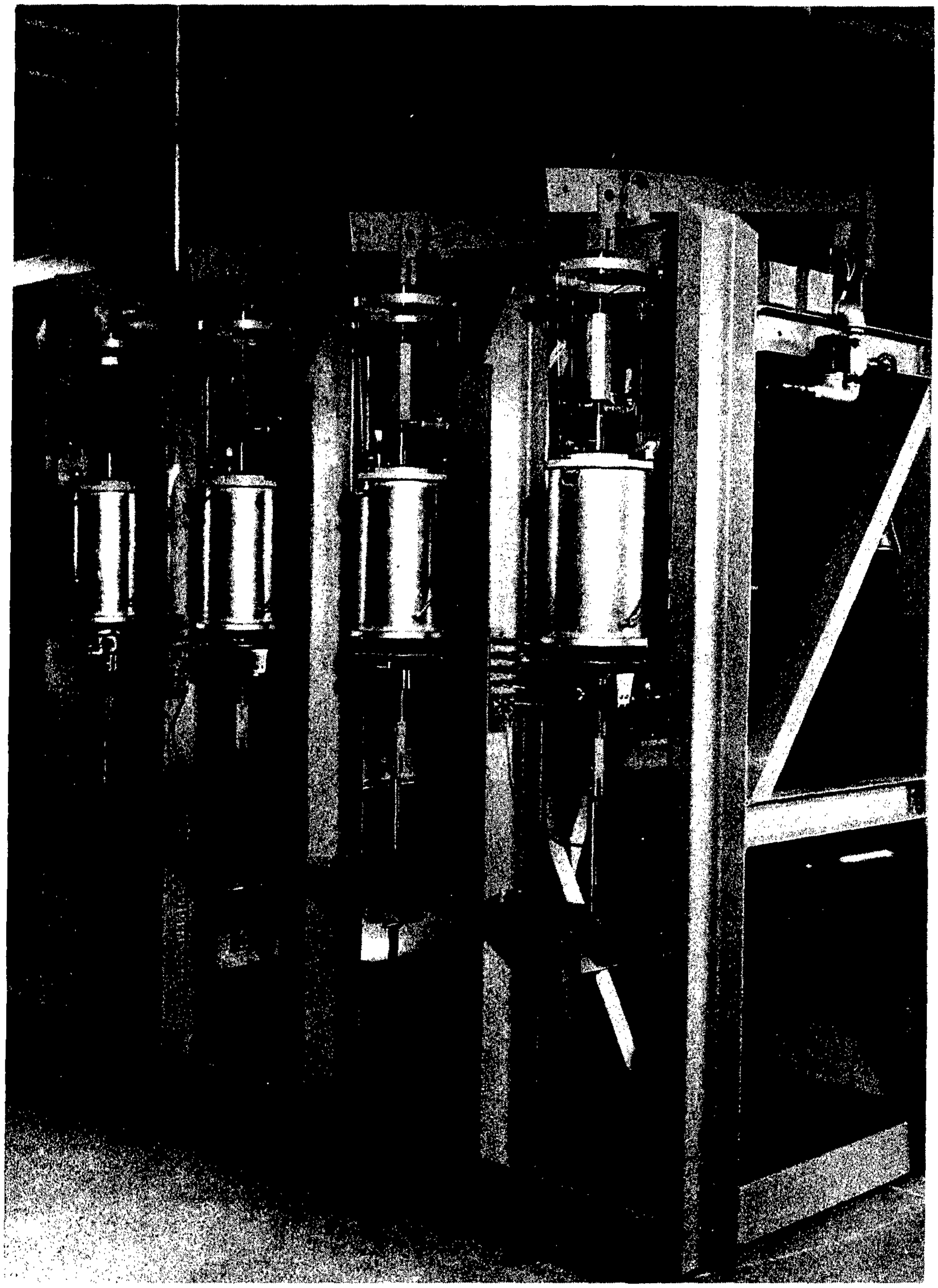


Figure 1

Creep Testing Equipment

WATER QUENCH HARDNESS CHARACTERISTICS OF A 600 LB. HEAT OF 3 Cr - 1 Mo - Ti - B STEEL ORIGINALLY TREATED AT 1600°F - 5 HRS - AIR COOLED

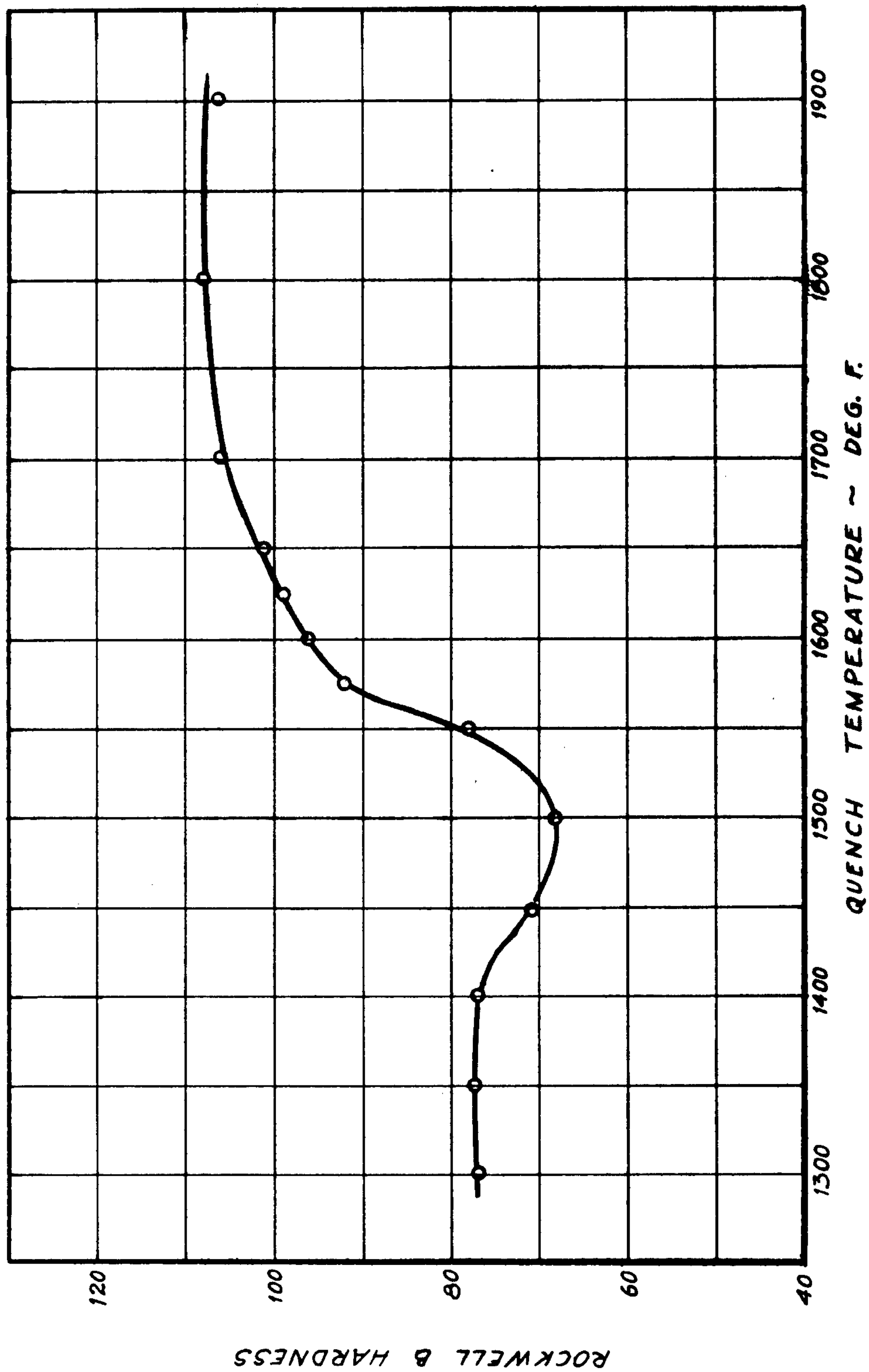
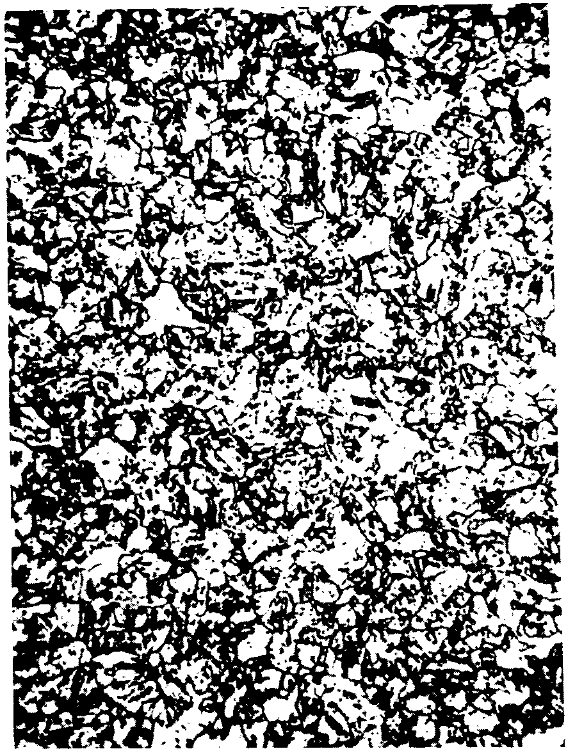


Fig. 2

Normalized Microstructures of

600 Pound Heat of 3 Cr- 1 Mo-Ti-B Steel

Nital Etch 500X



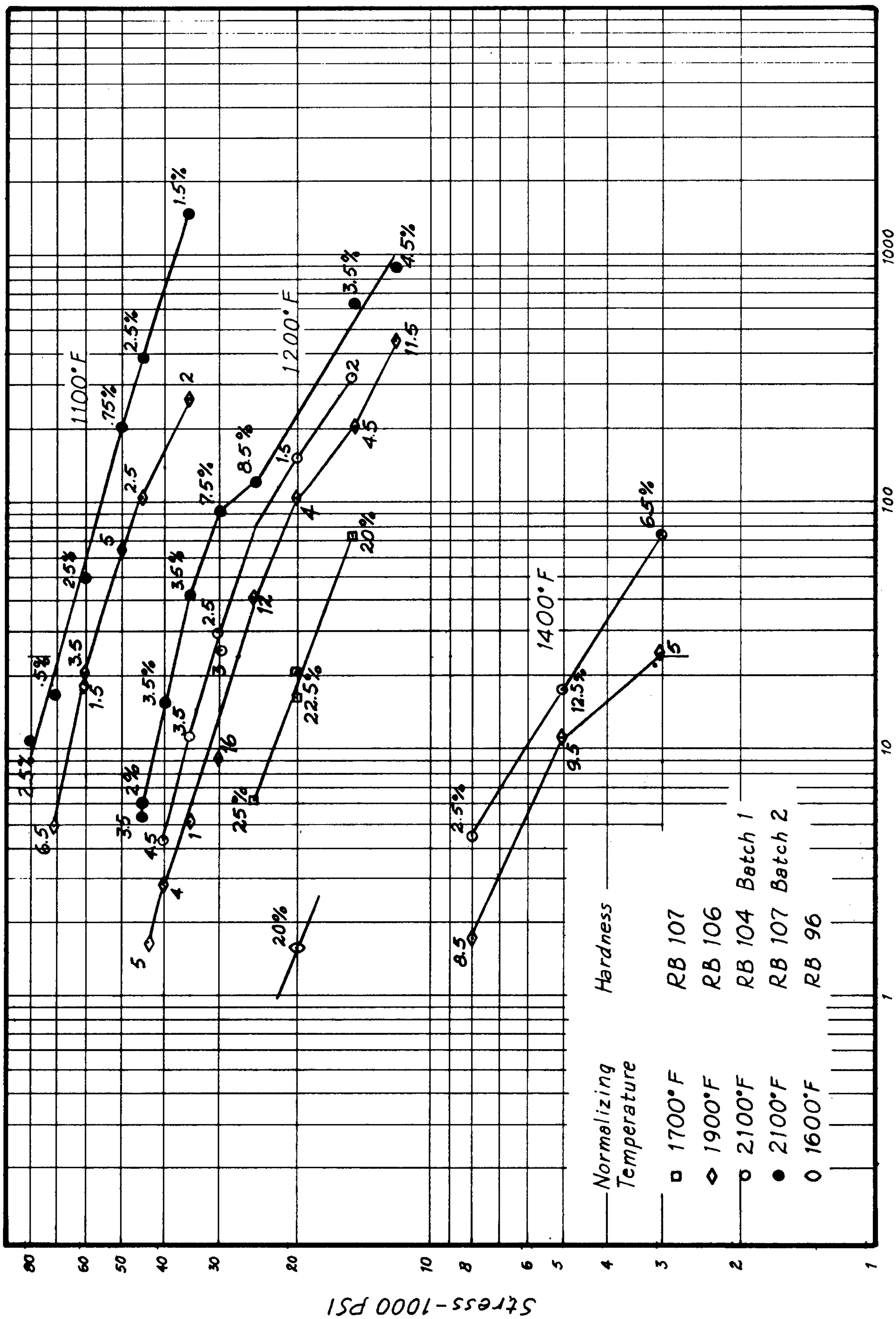
1700°F RB 109

1900°F RB 106

2100°F RB 107

Figure 3

EFFECT OF NORMALIZING TEMPERATURE ON THE RUPTURE CHARACTERISTICS OF 600 POUND HEAT OF 3 Cr - 1 Mo - Ti - B STEEL



Time for Fracture - Hours

Fig. 4

Hot Rolled Microstructures of

600 Pound Heat of 3 Cr- 1 Mo-Ti-B Steel

Nital Etch 500X

Hot Rolled at 1850°F
RB 111

Preheated at 2100°F
Hot Rolled at 1850°F
RB 110

Figure 5

COMPARISON OF 1200°F CREEP AND RUPTURE PROPERTIES OF HOT ROLLED AND NORMALIZED 600 # HEAT 3 Cr - 1 Mo - Ti - B 0.075" SHEET STEEL

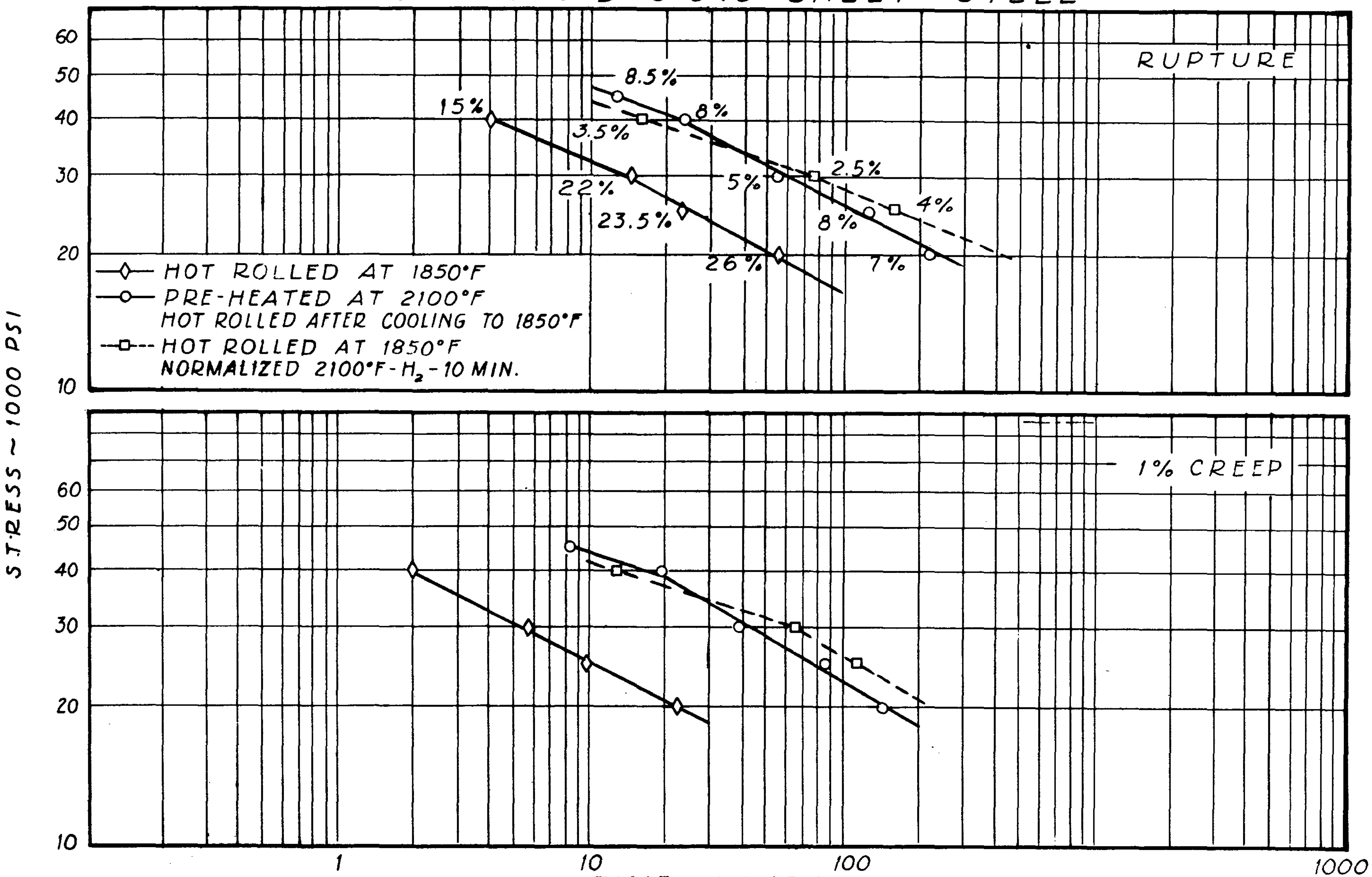


Fig. 6

HIGH TEMPERATURE CREEP AND RUPTURE PROPERTIES
600 POUND HEAT OF 3Cr-1Mo-Ti-B STEEL
0.050" SHEET NORMALIZED AT 1900°F

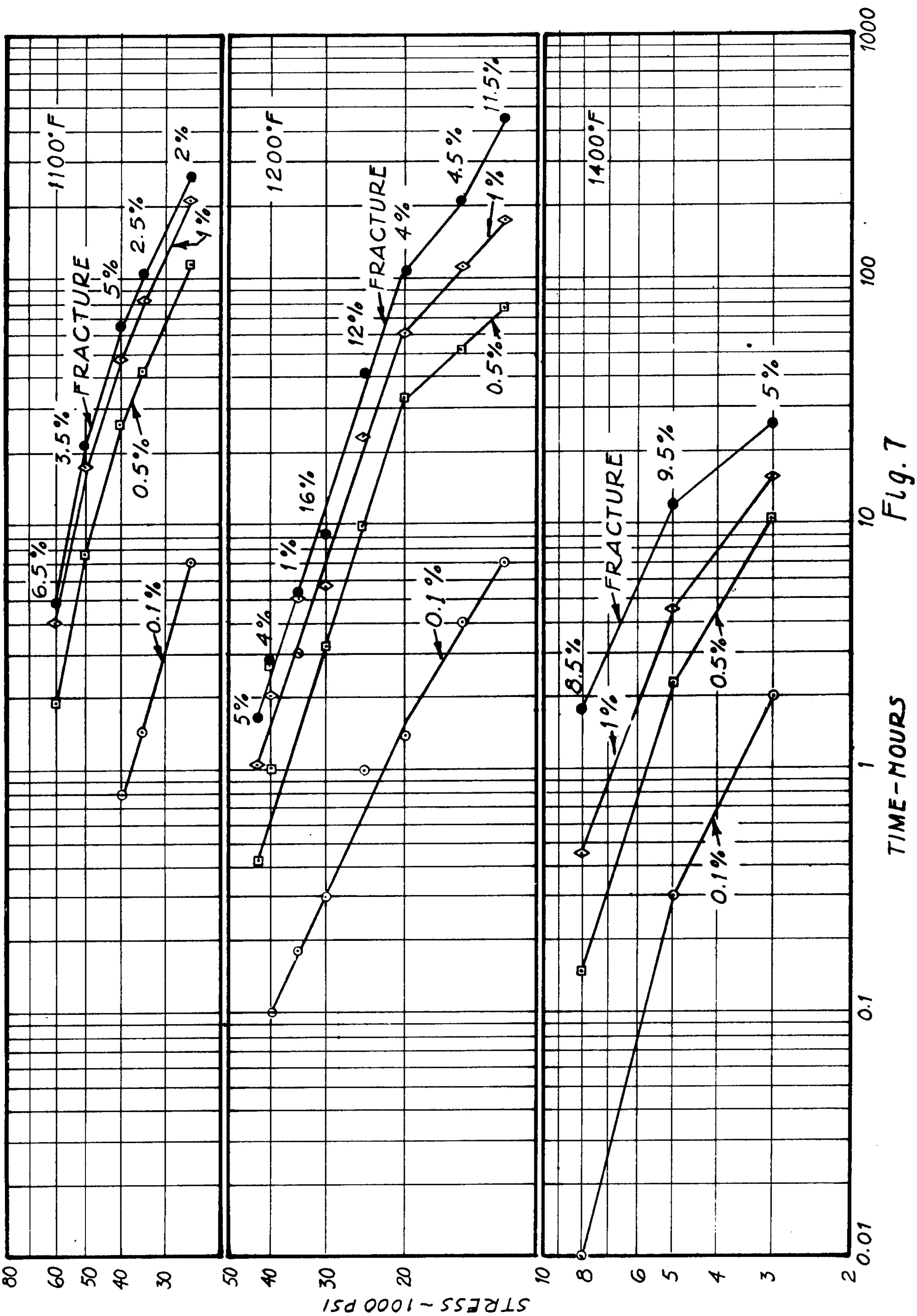
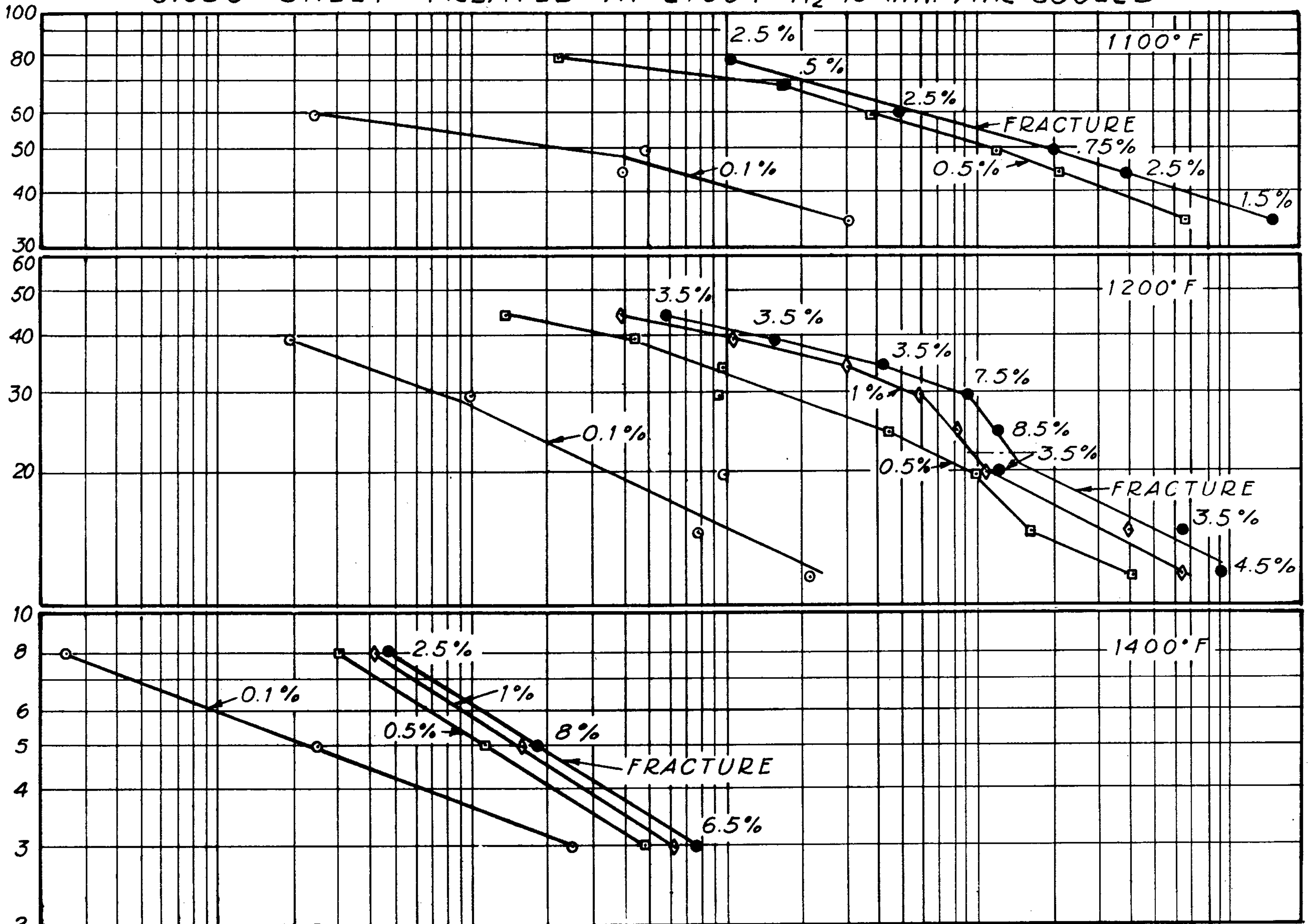


Fig. 7

HIGH TEMPERATURE CREEP AND RUPTURE PROPERTIES
600 POUND HEAT OF 3Cr-1Mo-Ti-B STEEL
0.050" SHEET TREATED AT 2100°F-H_2-10 MIN. AIR COOLED



TIME ~ HOURS Fig. 8

THE EFFECT OF SHEET THICKNESS ON THE 1200°F RUPTURE STRENGTH
600 POUND HEAT OF 3Cr-1Mo-Ti-B STEEL
HEAT TREATED AT 2100°F-H_2-10MIN. AIR COOLED

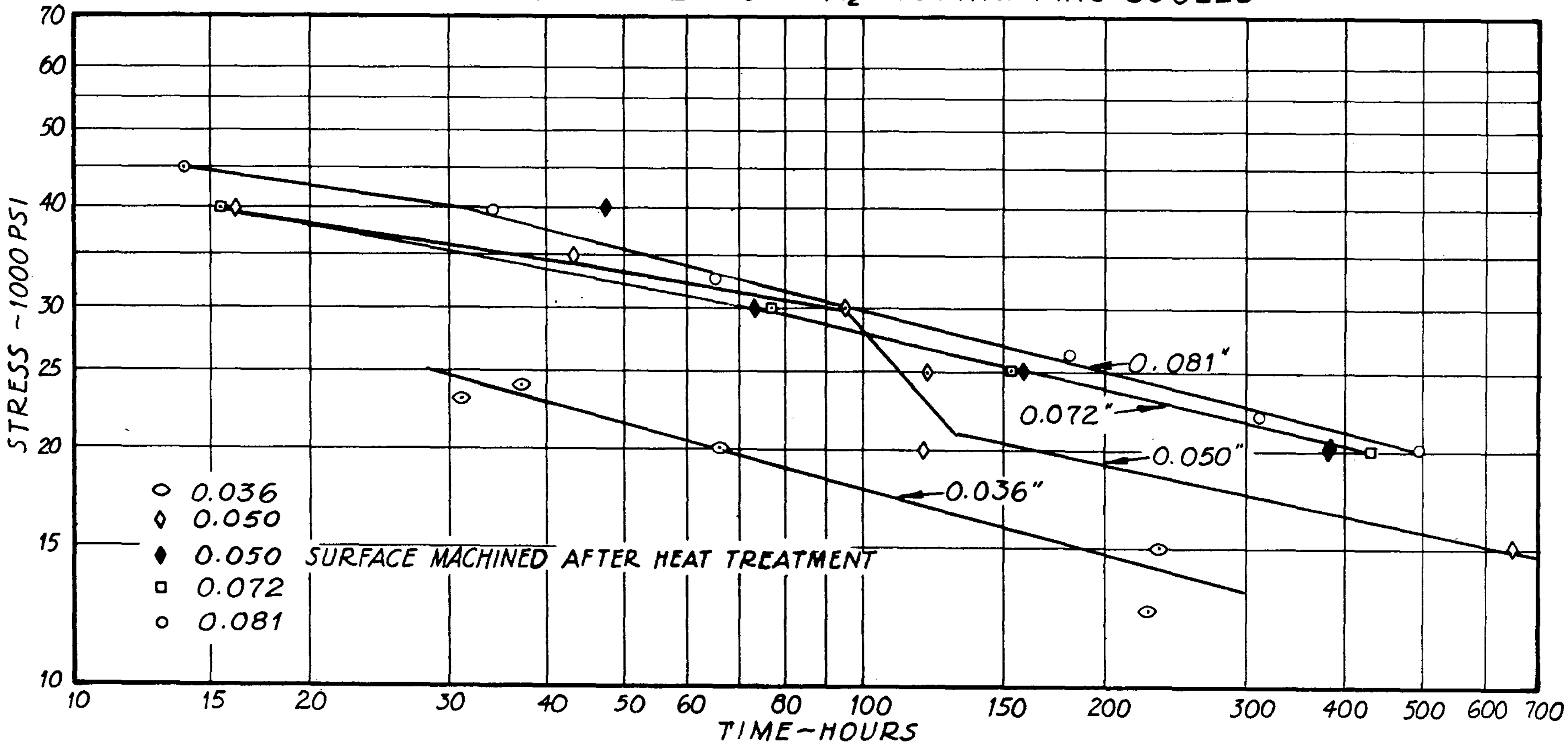


Fig. 9

VARIATION OF 1200°F RUPTURE STRENGTH WITH SHEET THICKNESS
600 POUND HEAT OF 3Cr-1Mo-Ti-B STEEL
HEAT TREATED AT 2100°F-H_2-10MIN.-AIR COOLED

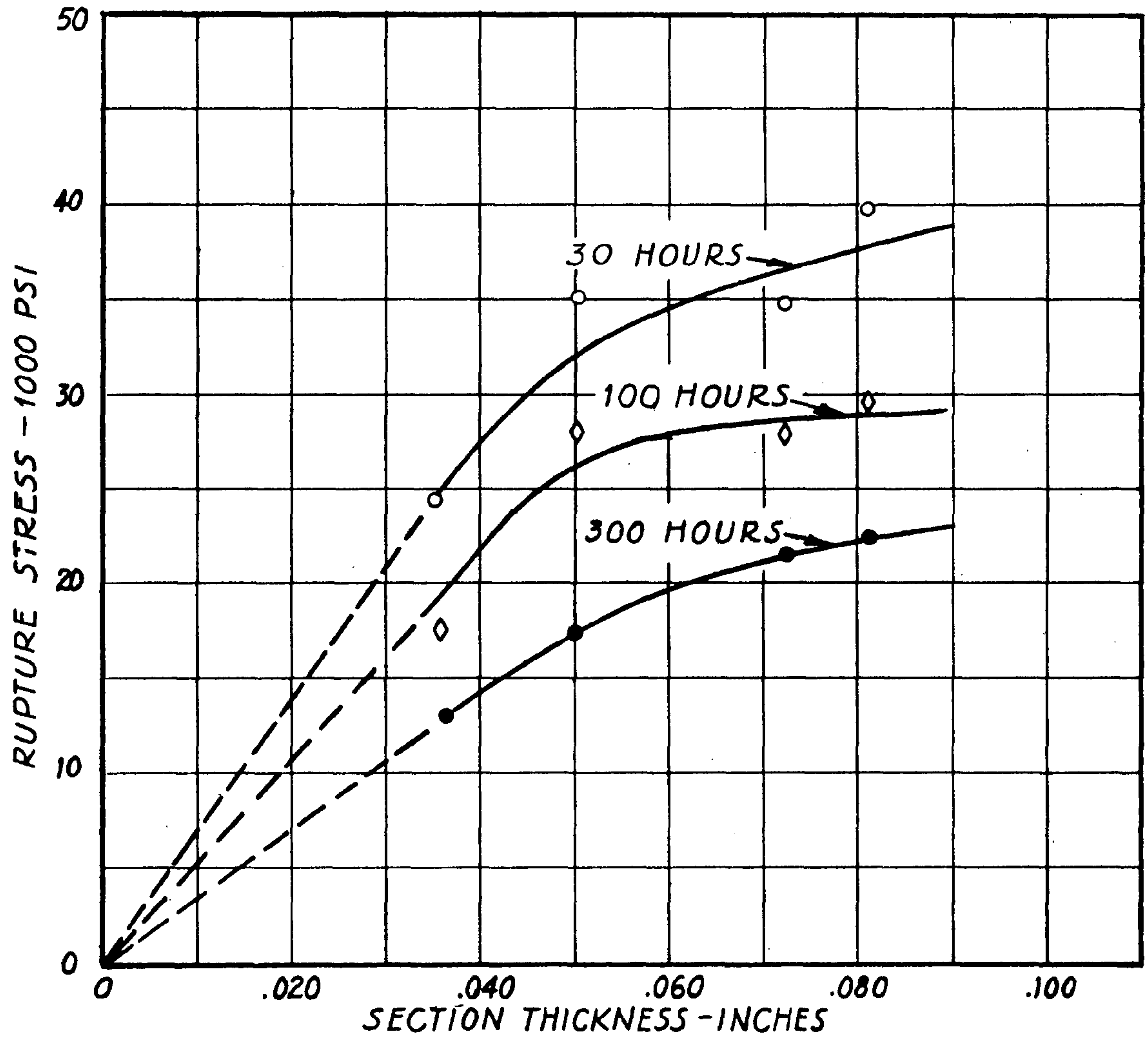


Fig. 10

EFFECT OF THICKNESS ON CROSS-SECTION MICROHARDNESS
600 POUND HEAT 3 Cr-1Mo-Ti-B STEEL
HOT ROLLED AT 1850°F-NORMALIZED 2100°F-H_2-10 MIN.

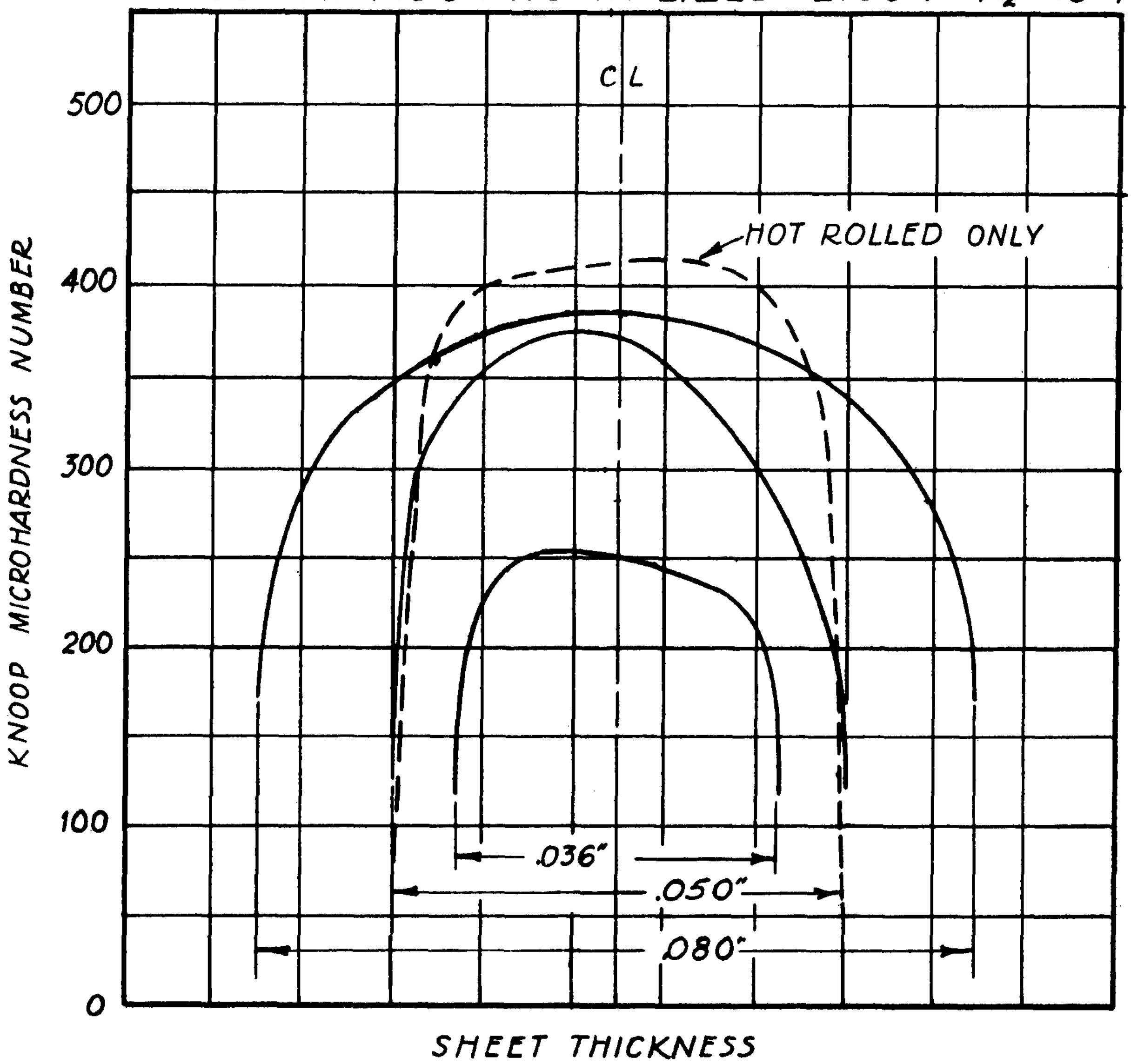
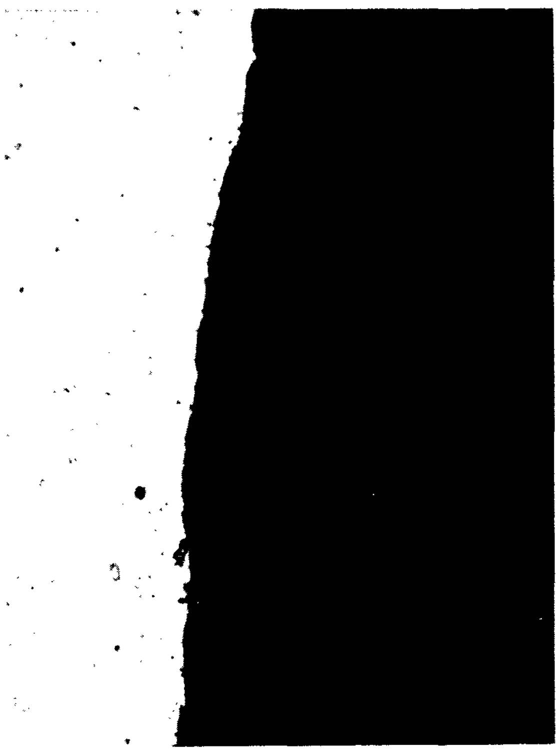


Fig.11

Surface Microcross-Sections of

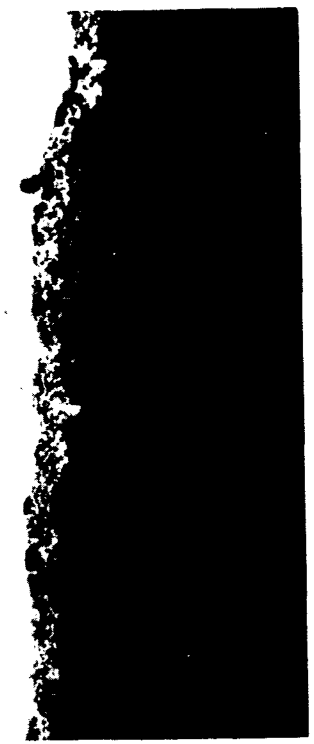
600 Pound Heat of 3 Cr- 1 Mo-Ti-B Steel

Unetched 200X



Hot Rolled at 1850°F

Hot Rolled at 1850°F
2100°F - H_2 - 10 Min. - Air Cool



Hot Rolled at 1850°F
Pickled
2100°F - H_2 - 10 Min. - Air Cool

Figure 12

THE STRESS-RUPTURE PROPERTIES OF THE 600 #HEAT 3Cr-1Mo-Ti-B STEEL HEAT TREATED AT 2100°F IN H_2-AIR COOLED. EFFECT OF SURFACE CONDITION PRIOR TO H_2 TREATMENT

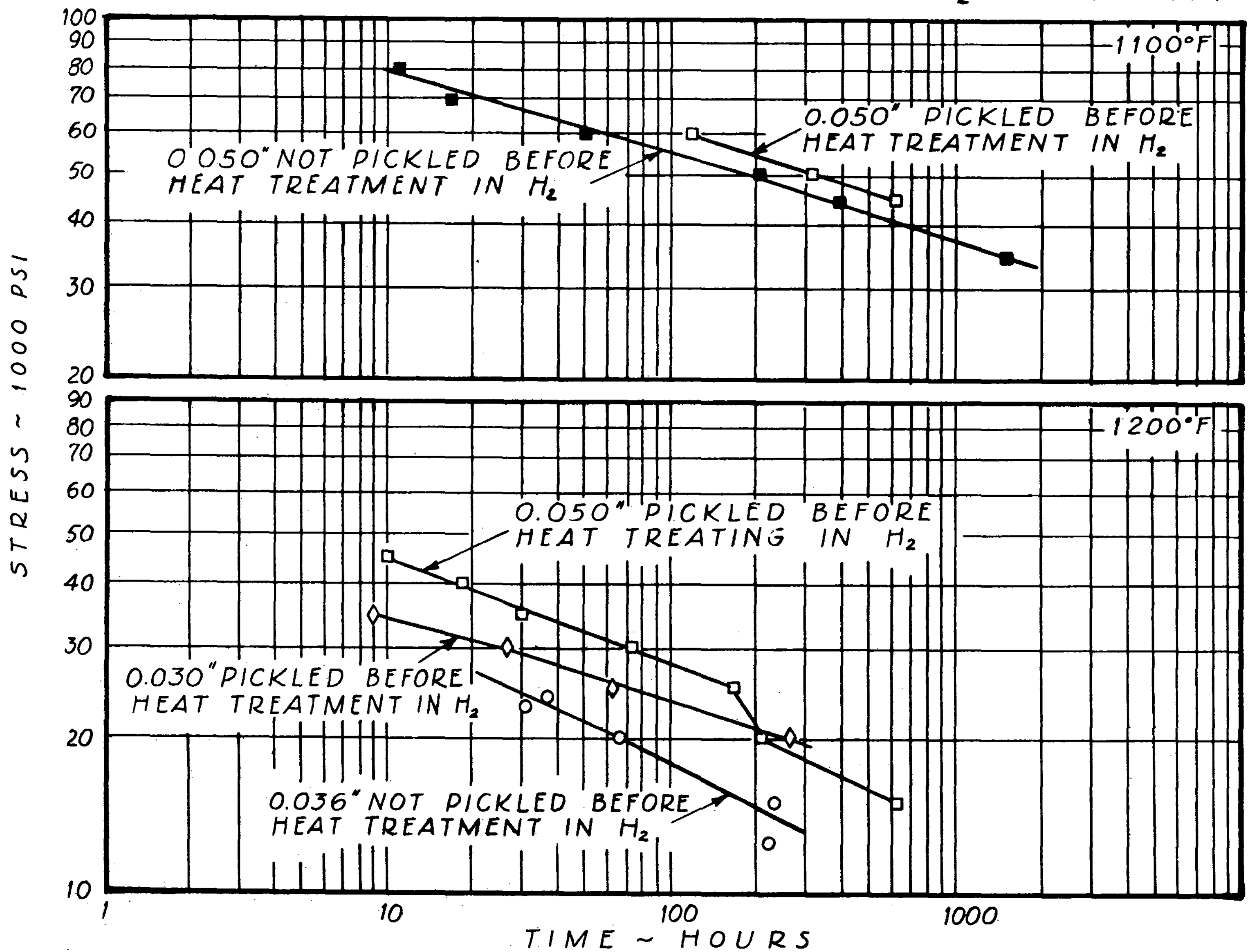
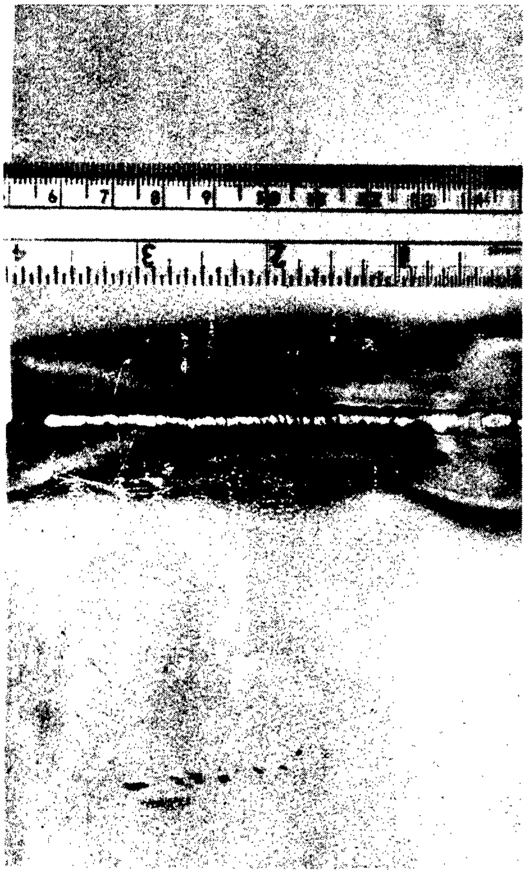
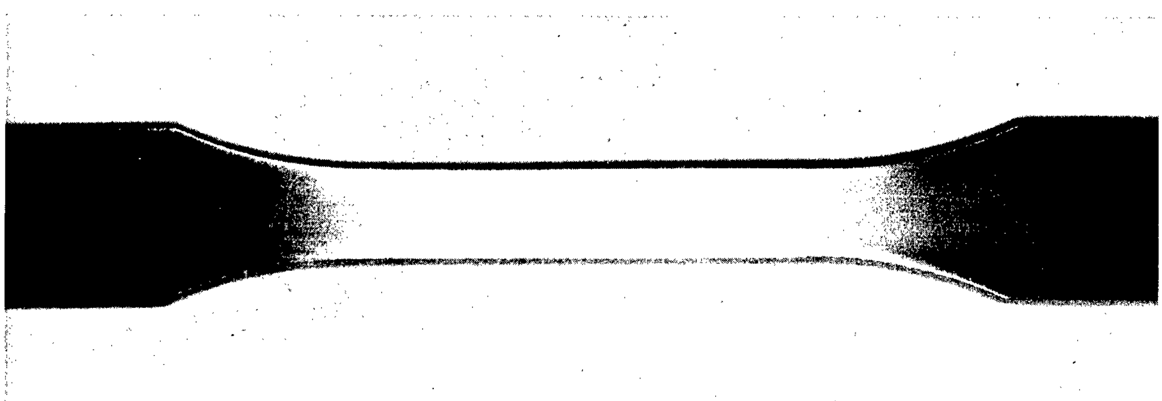


Fig. 13



Face Side of Weld

Back Side of Weld



Machined Rupture Specimen

Figure 14

Heliarc Welded Without Filler Rod

600 Pound Heat of 3 Cr- 1 Mo-Ti-B Steel

0.050-Inch Sheet Material

HIGH TEMPERATURE RUPTURE PROPERTIES OF HELIARC BUTT WELDED 3 Cr - 1 Mo - Ti - B SHEET STOCK

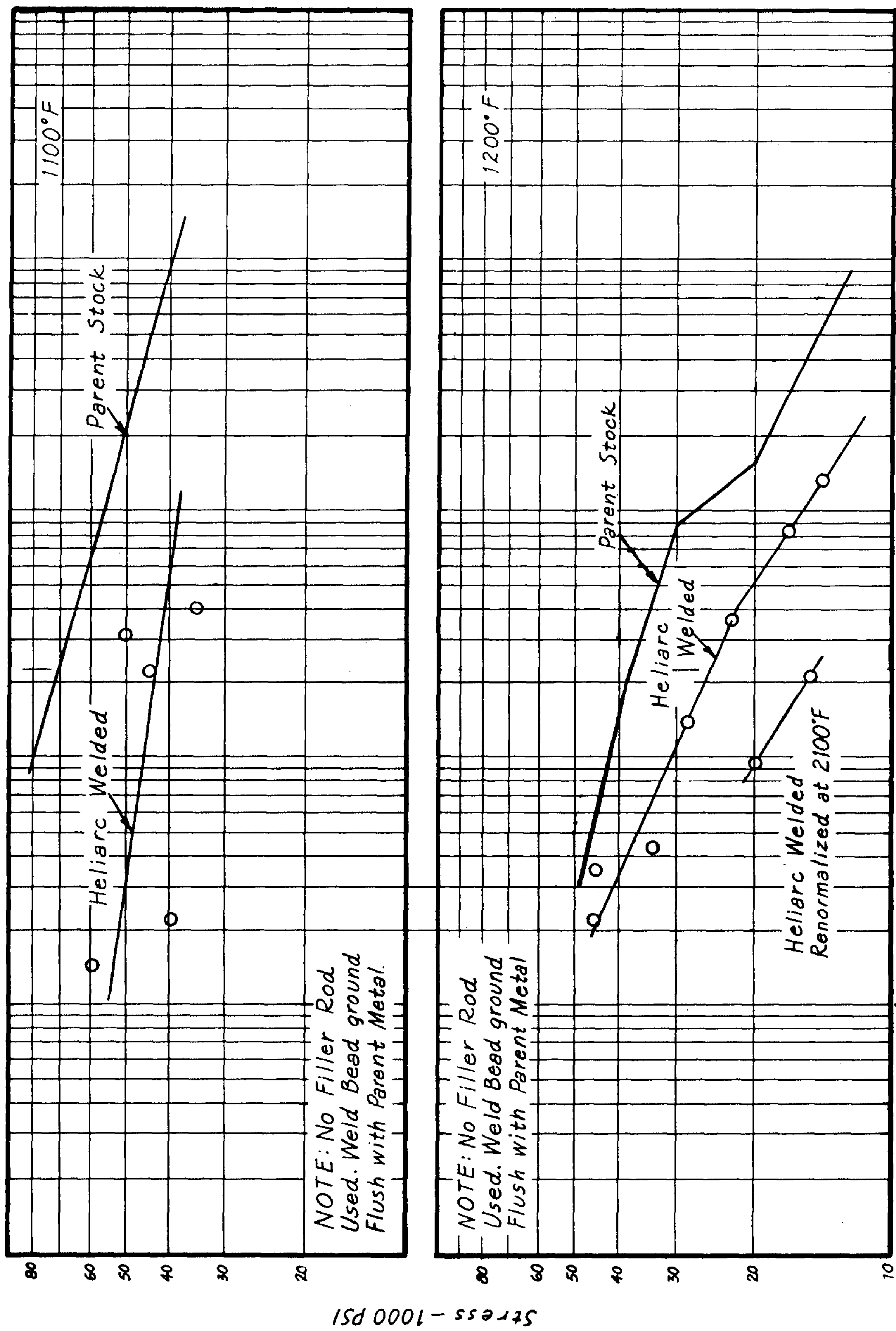


Fig. 15

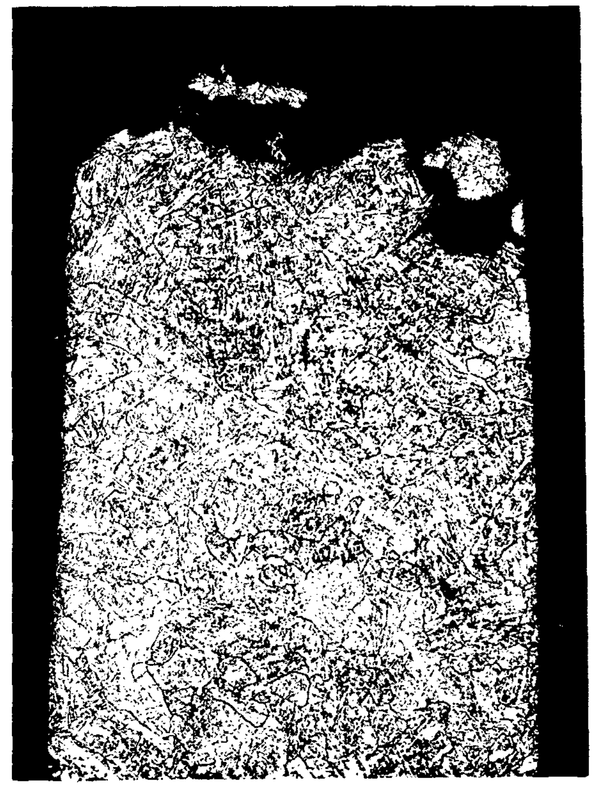
Microstructures of Fractured Heliarc Welds

600 Pound Heat of 3 Cr- 1 Mo-Ti-B Steel

No Filler Rod

Nital Etch 75X

$1100^{\circ}F$ - 35,000 PSI. - 39 Hrs.



$1200^{\circ}F$ - 12,500 PSI. - 131 Hrs.

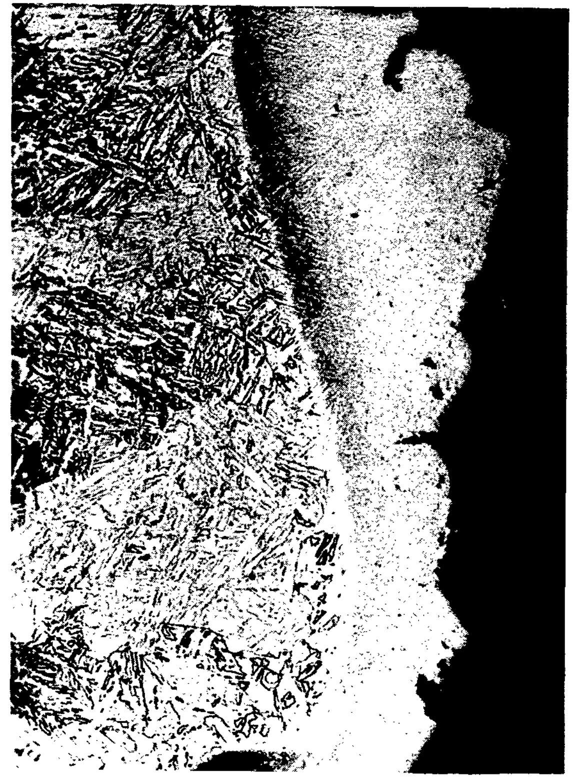
Figure 16

Microstructures of Fractured Heliarc Welds

600 Pound Heat of 3 Cr- 1 Mo-Ti-B Steel

347 Stainless Steel Filler Rod

Bead Ground Flush
1200°F - 20,000 PSI. - 55 Hrs.



Bead Left On
1200°F - 20,000 PSI. - 136 Hrs.

RUPTURE STRENGTH AT 1200°F
OF 3Cr-1Mo-Ti-B SHEET STEEL
HELIARC WELDED WITH 347 STAINLESS STEEL FILLER ROD

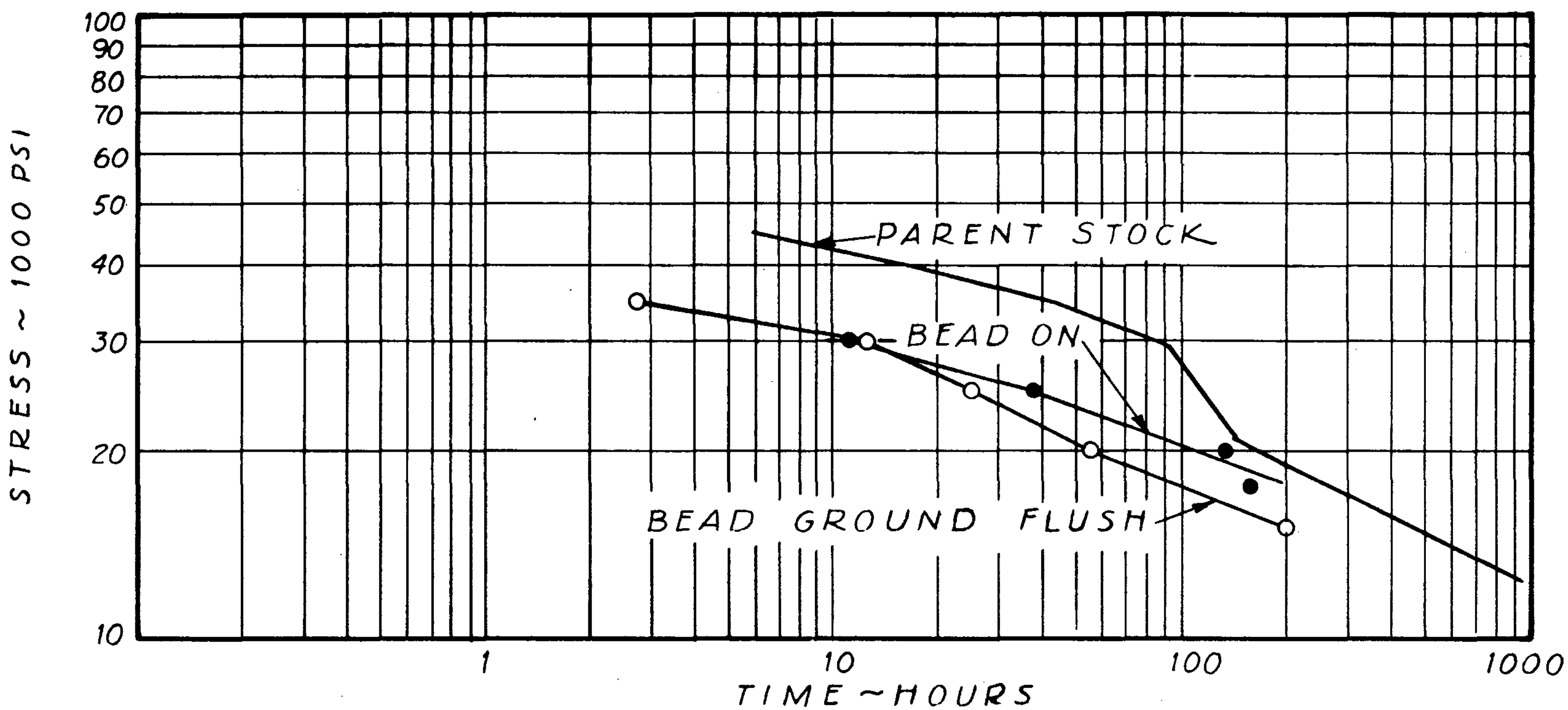


Fig. 18

CREEP AND RUPTURE STRENGTH OF CERAMIC COATED 0.050" SHEET
600 POUND HEAT OF 3Cr-1Mo-Ti-B STEEL

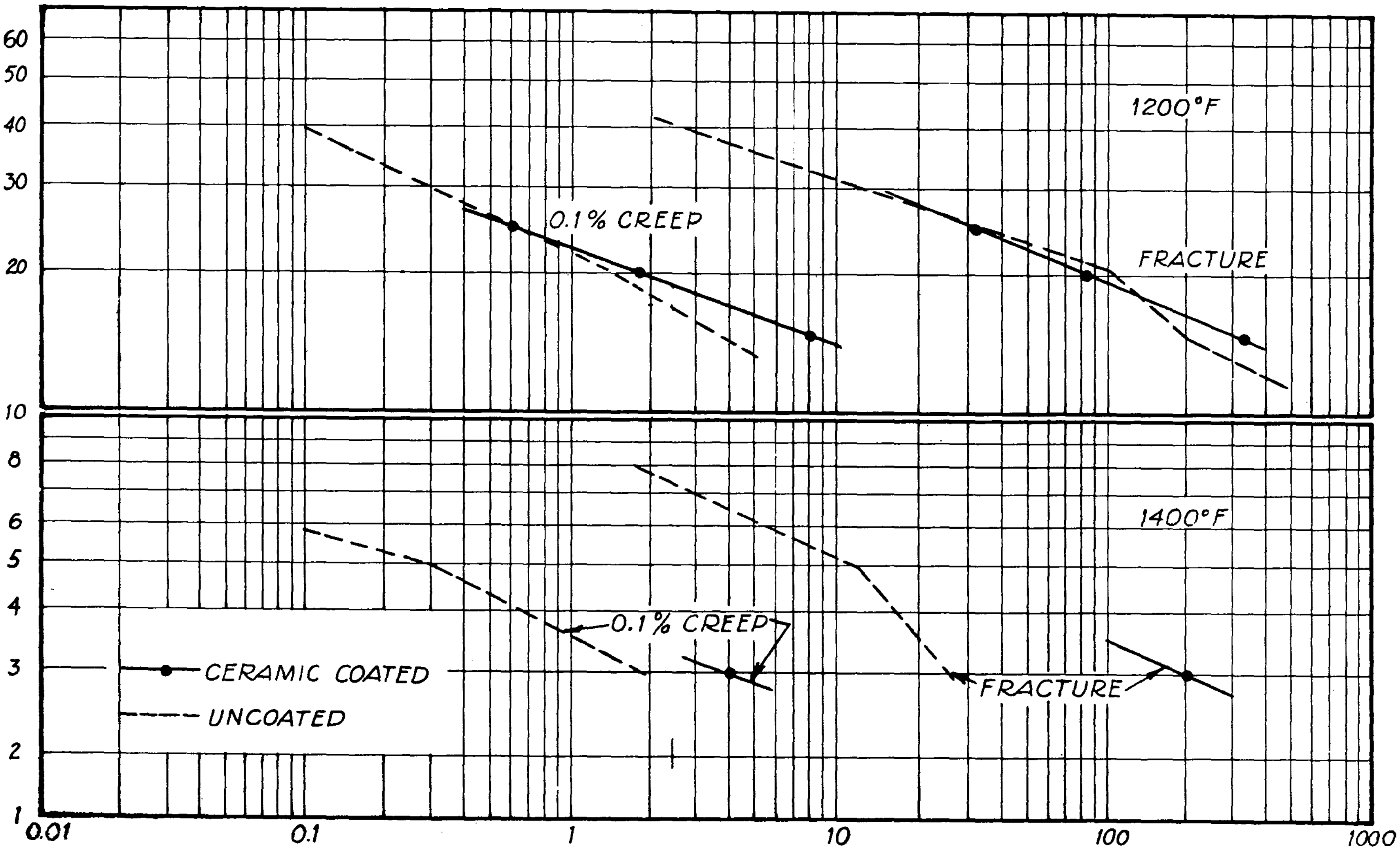


Fig. 19

Uncoated	Uncoated	Ceramic Coated	Ceramic Coated	Ceramic Coated	
1400°F	1200°F	1400°F	1200°F	1200°F	Original
3000 PSI.	15,000 PSI.	3000 PSI.	15,000 PSI.	25,000 PSI.	Ceramic
5% Elong.	4.5% Elong.	8% Elong.	4.5% Elong.	3% Elong.	Coating
25 Hrs.	205 Hrs.	200 Hrs.	336 Hrs.	32 Hrs.	

Figure 20

Comparison of Ceramic Coated and Uncoated

Rupture Test Specimens

EFFECT OF HEAT TREATING TEMPERATURE ON THE ROOM TEMPERATURE HARDNESS OF C - Ti - B STEELS

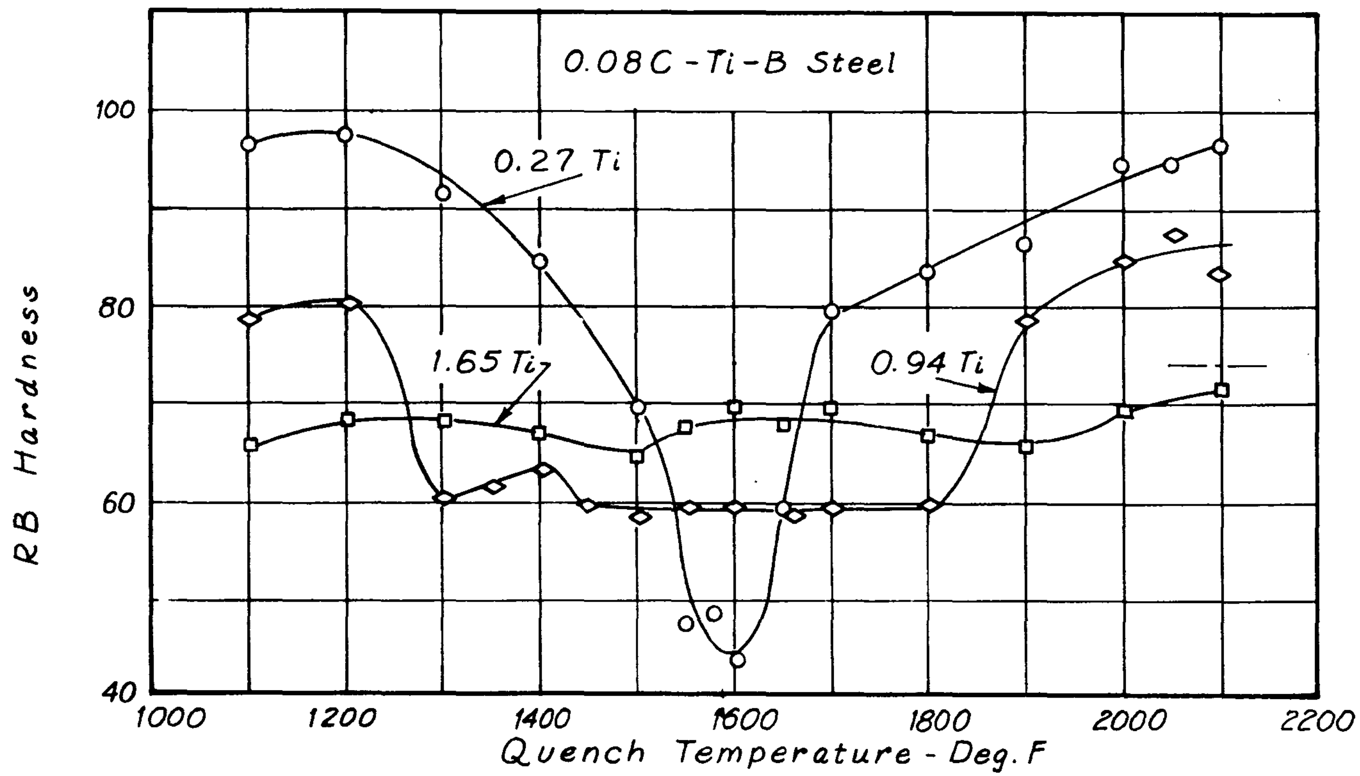
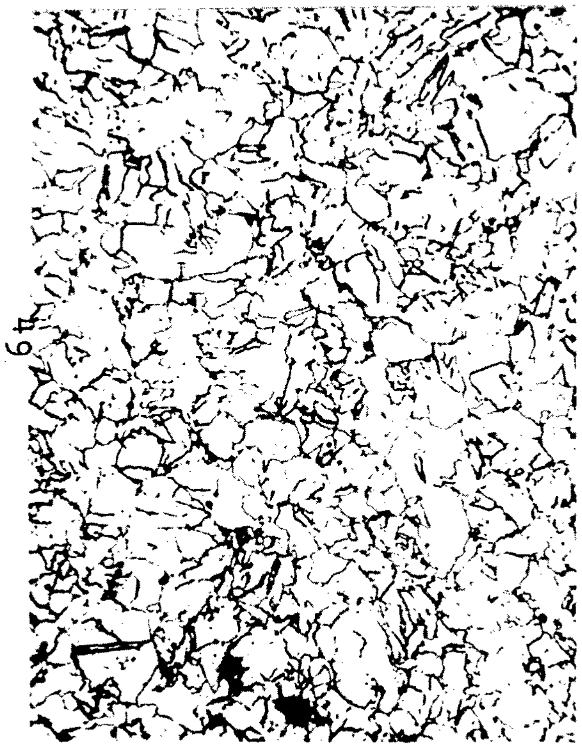


Fig. 21

Microstructures of 0.08 C-Ti-B Steel (80-81-82)

Heat Treated at 2100°F - H_2 - 10 Min. - Air-Cooled

Nital Etch 500X



.08 C
.27 Ti RB 96

.09 C
.94 Ti RB 88

.08 C
1.65 Ti RB 65

Figure 22

CREEP AND RUPTURE PROPERTIES OF C - Ti - B STEELS AT 1200° F.

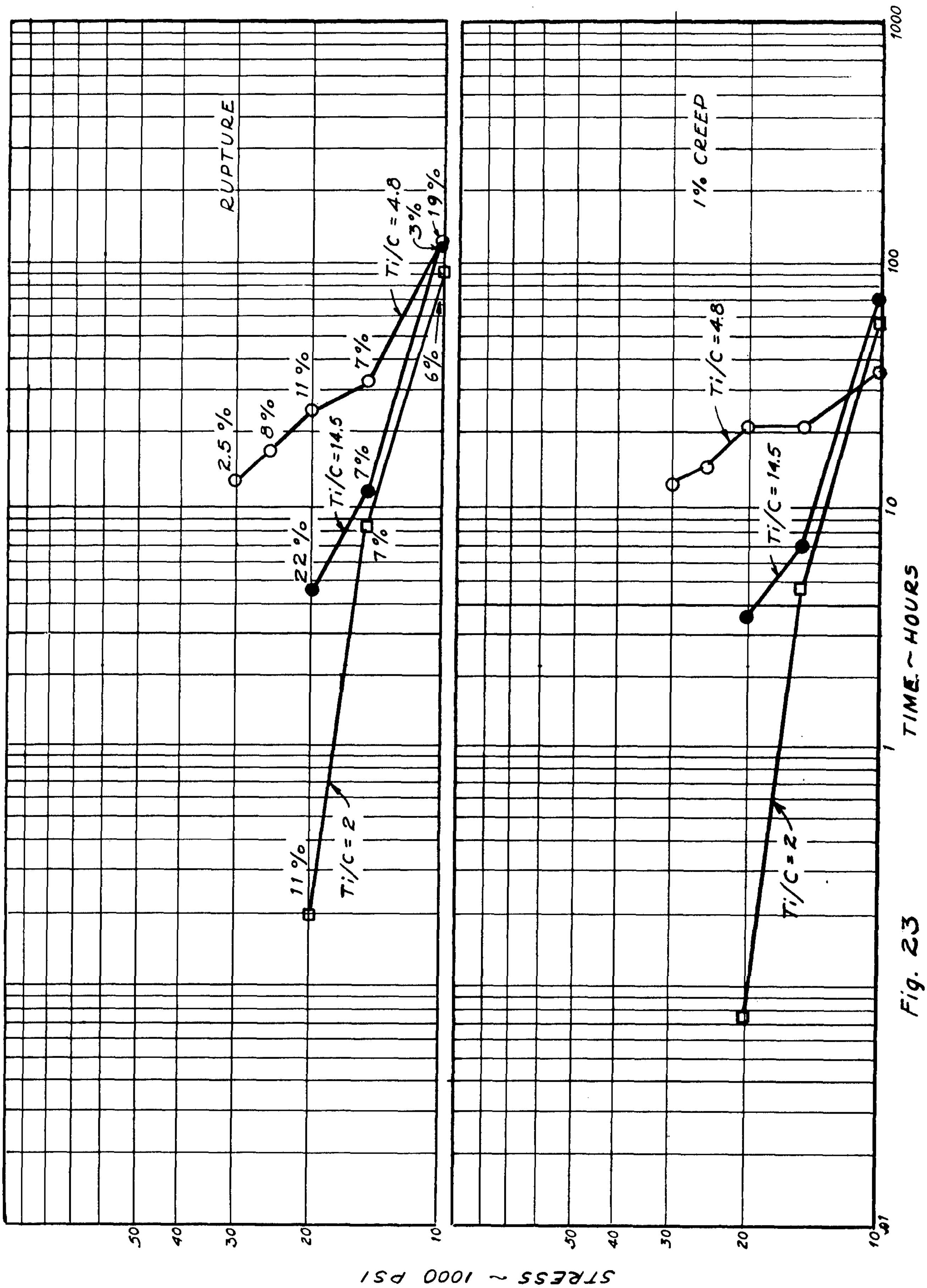


Fig. 23

CREEP AND RUPTURE PROPERTIES AT 1200°F FOR
0.08 C - Ti - B STEEL

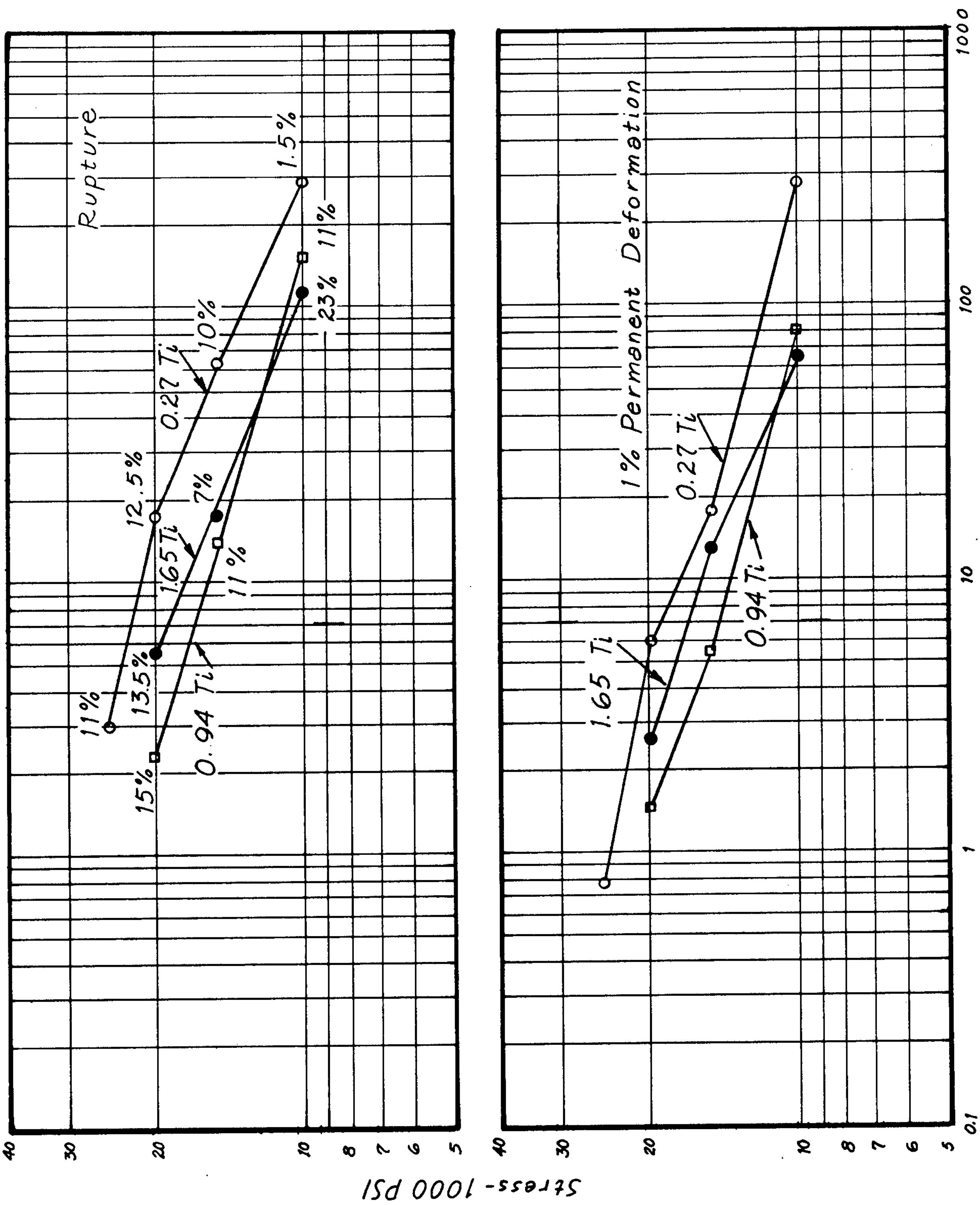


Fig. 24

CREEP AND RUPTURE PROPERTIES AT 1200°F FOR 0.14C-Ti-B STEEL

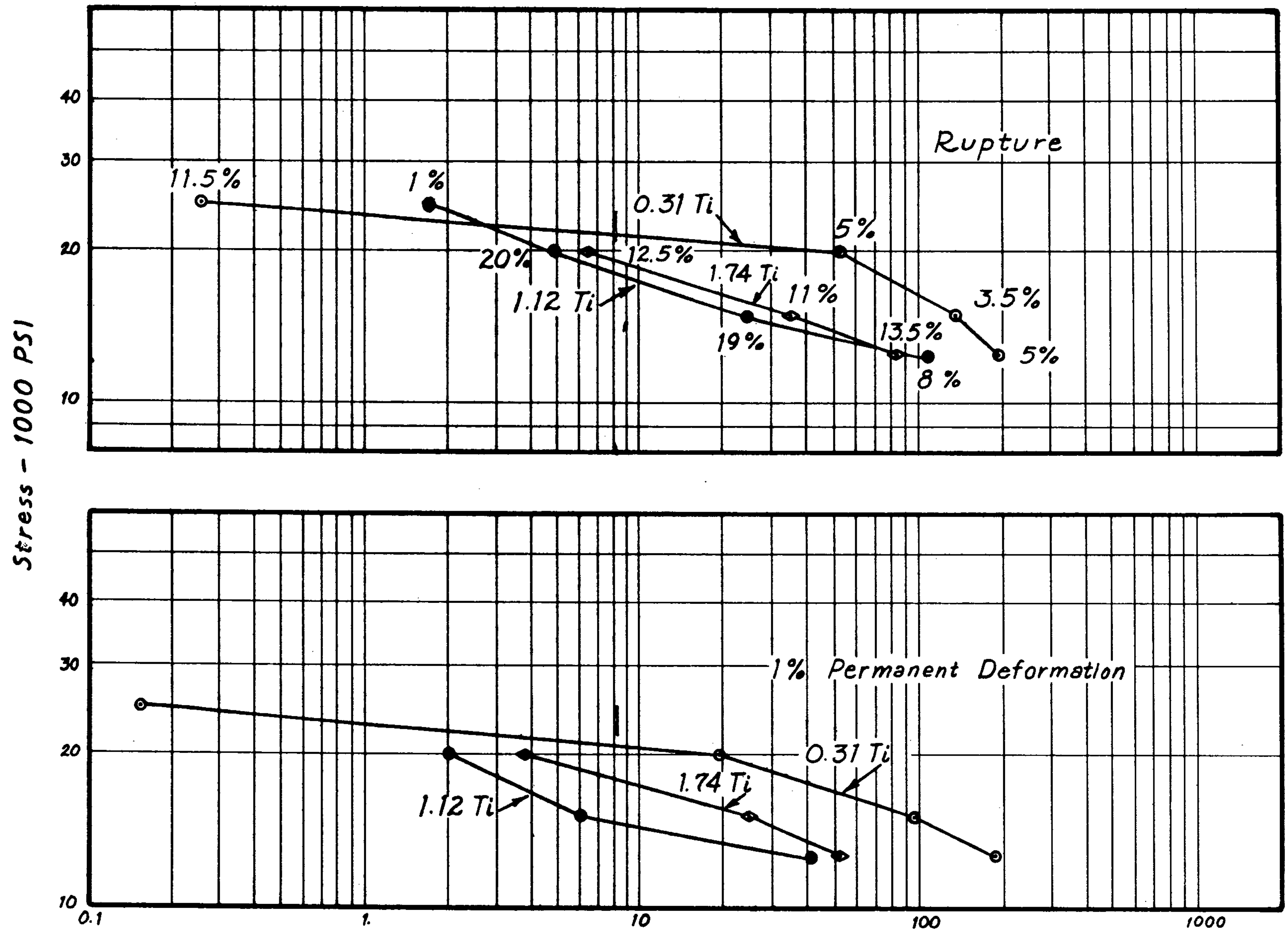
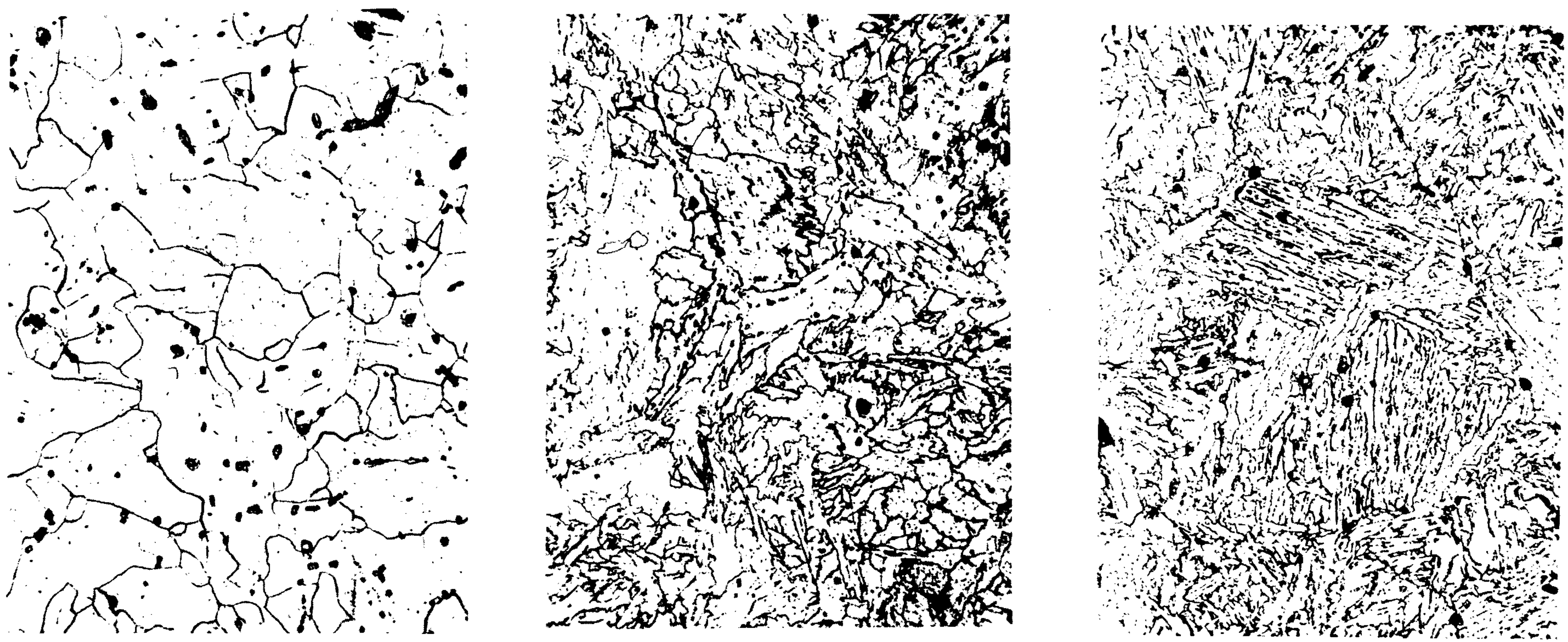


Fig. 25

Microstructures of C-Ti-B Steels With Varying Carbon (67-68-69)

$2100^{\circ}F$ - H_2 - 10 Min. - Air-Cool

Mital Etch 500X



.03 C RB 67 | .05 C RB 85 | 0.12 C RB 97

Figure 26

EFFECT OF CARBON CONTENT ON THE CREEP AND RUPTURE CHARACTERISTICS OF C-Ti-B STEEL AT 1200° F.

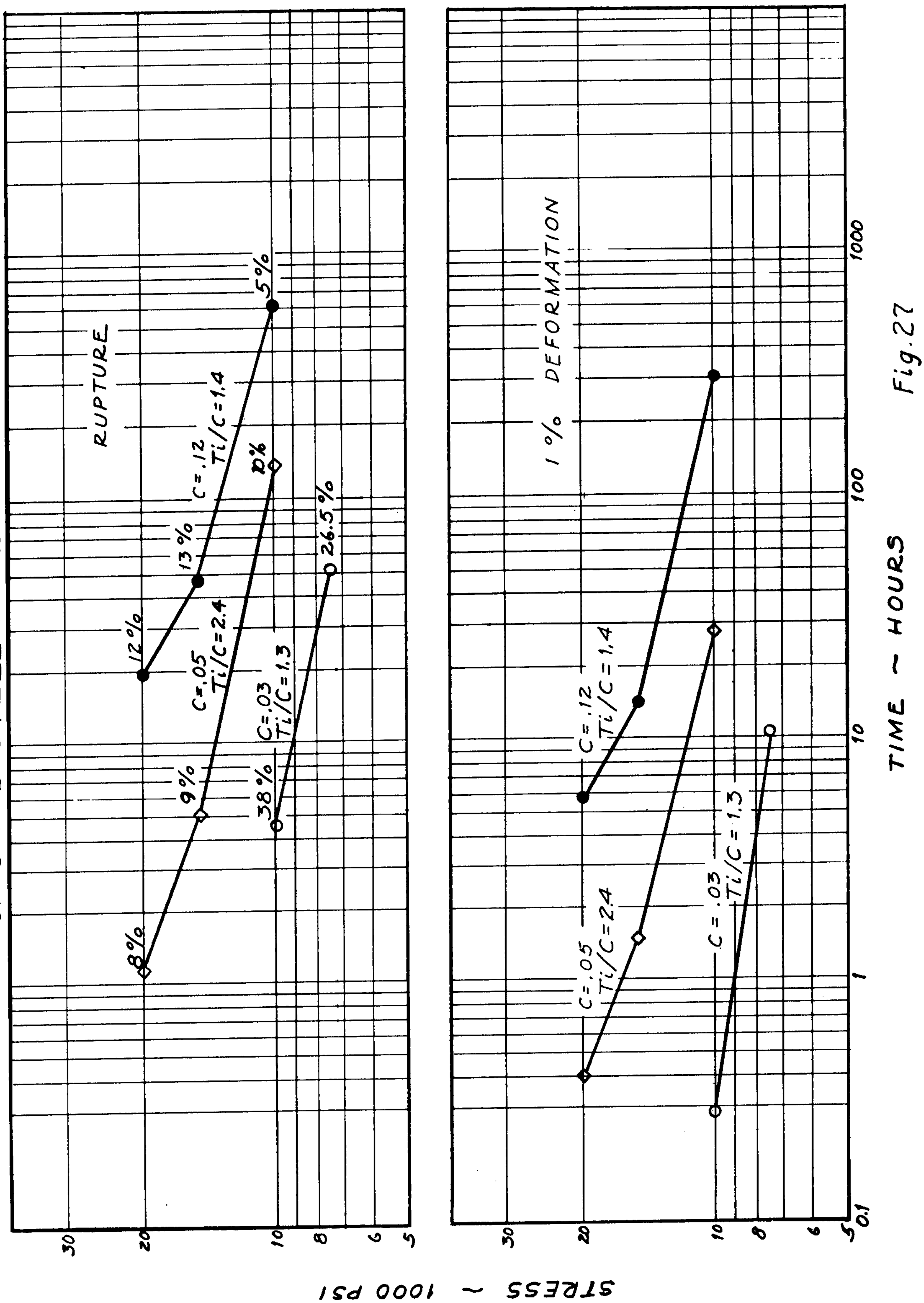


Fig. 27

EFFECT OF CARBON CONTENT ON THE STRESS FOR RUPTURE IN 100 HOURS AT 1200°F C-Ti-B STEELS

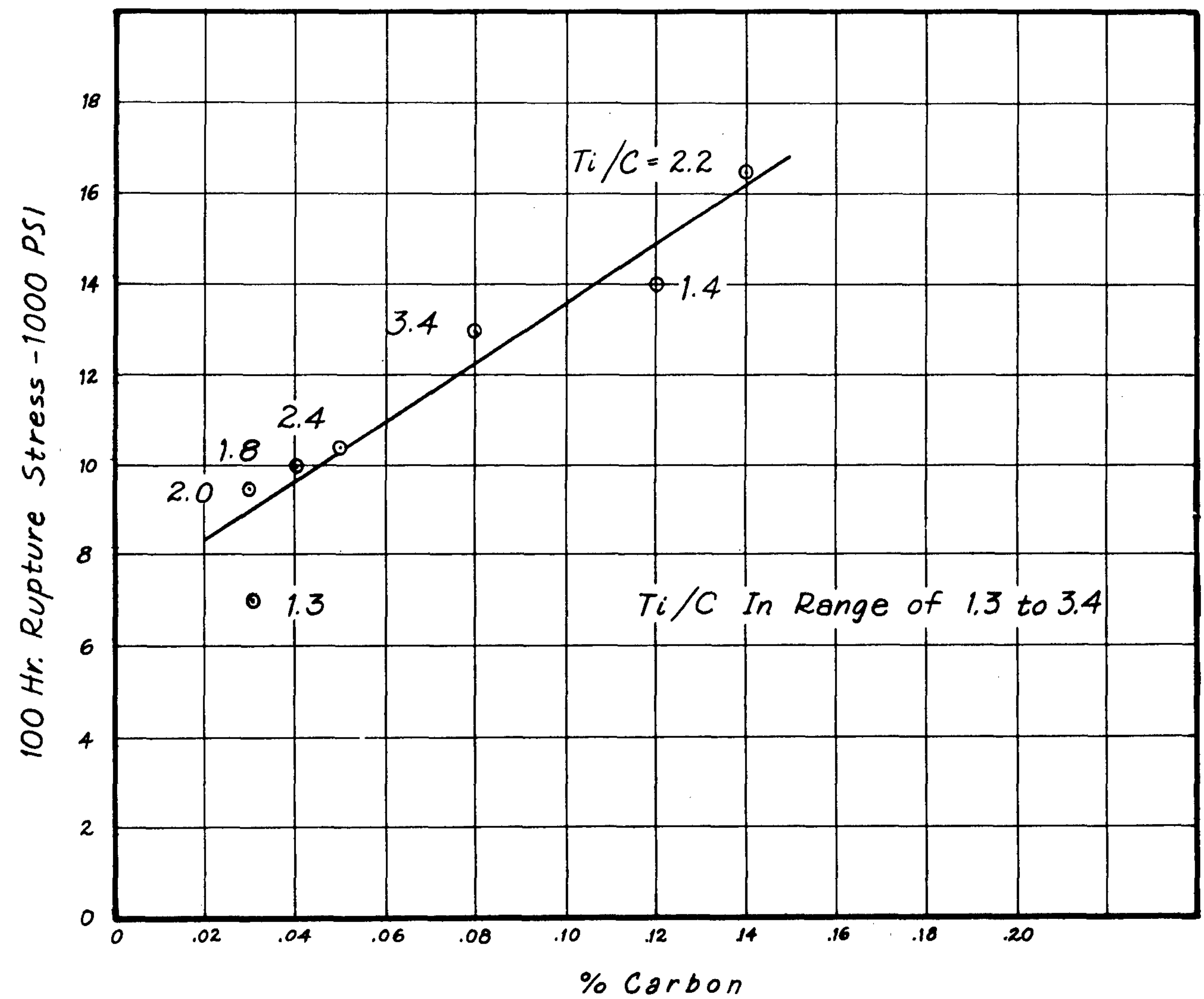


Fig. 28

EFFECT OF BORON CONTENT ON CREEP AND RUPTURE CHARACTERISTICS OF C-Ti-B STEEL AT 1200° F.
Ti/C RATIO = 1.8

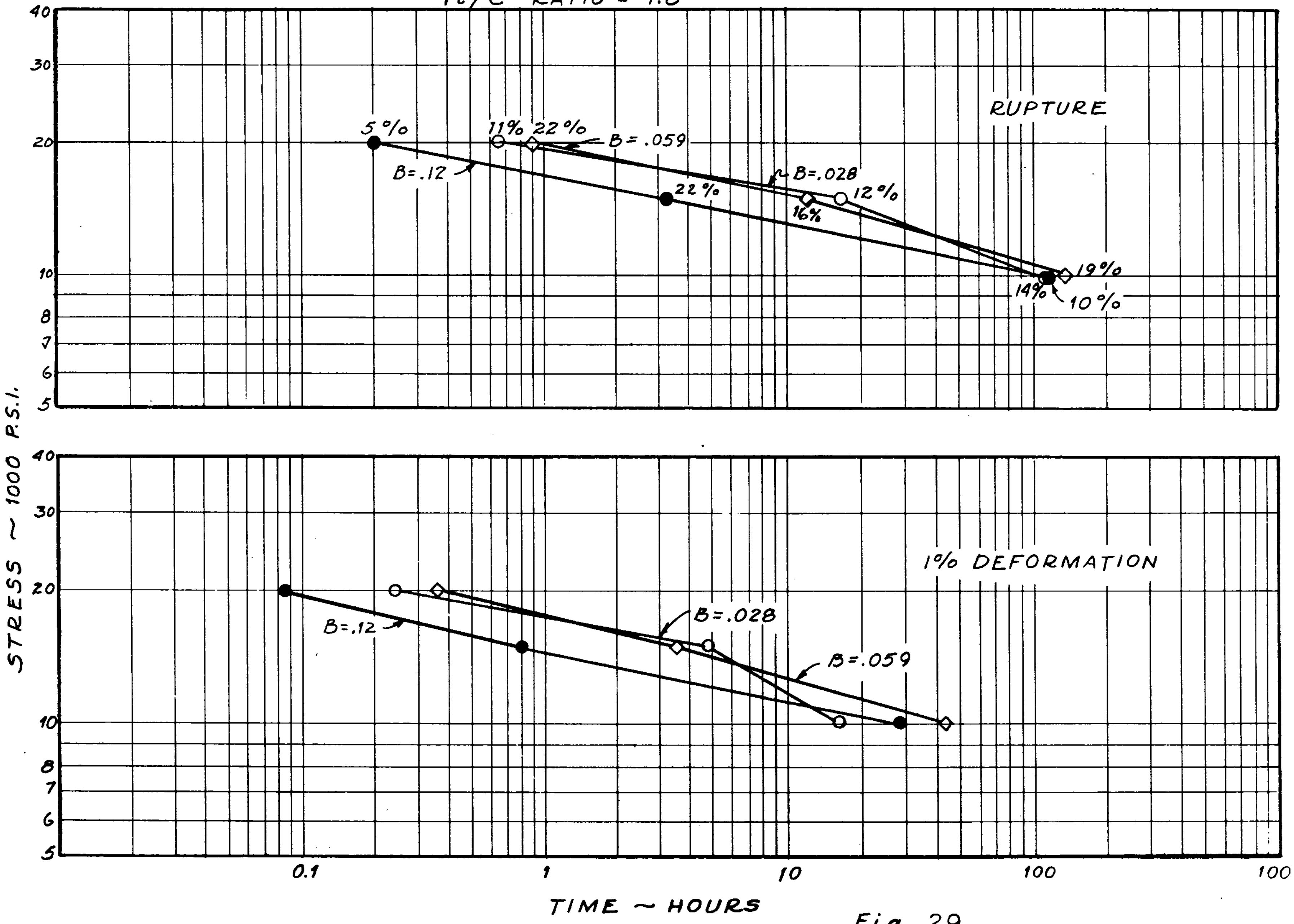


Fig. 29

EFFECT OF BORON CONTENT ON CREEP AND RUPTURE CHARACTERISTICS OF C-Ti-B STEEL AT 1200°F
Ti/C RATIO = 5

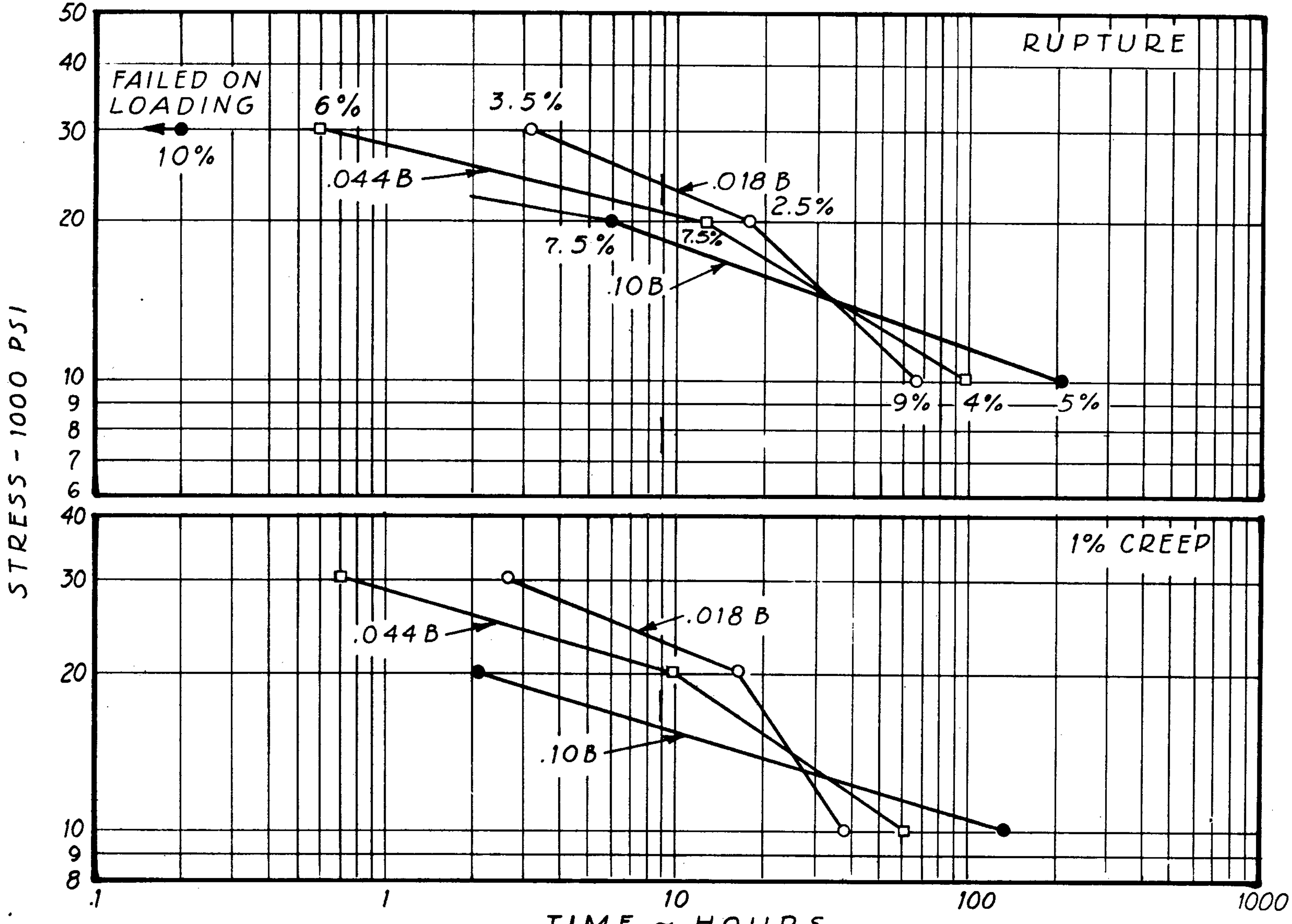


Fig. 30

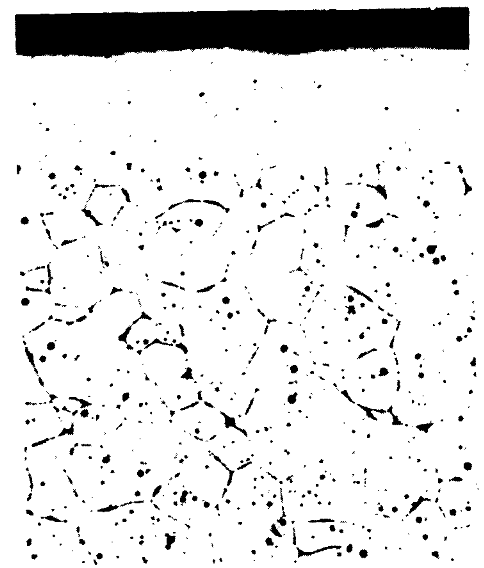
Microstructures of C-Ti-B Steels With Varying Boron (70-71-72)

2100°F - H_2 - 10 Min. - Air Cool

Unetched 100X

.028 B

.059 B



.12 B

Figure 31

Microstructure of C-Ti-B Steel With 0.12 Boron (72)

2100°F - H_2 - 10 Min. - Air-Cooled

Unetched 500X

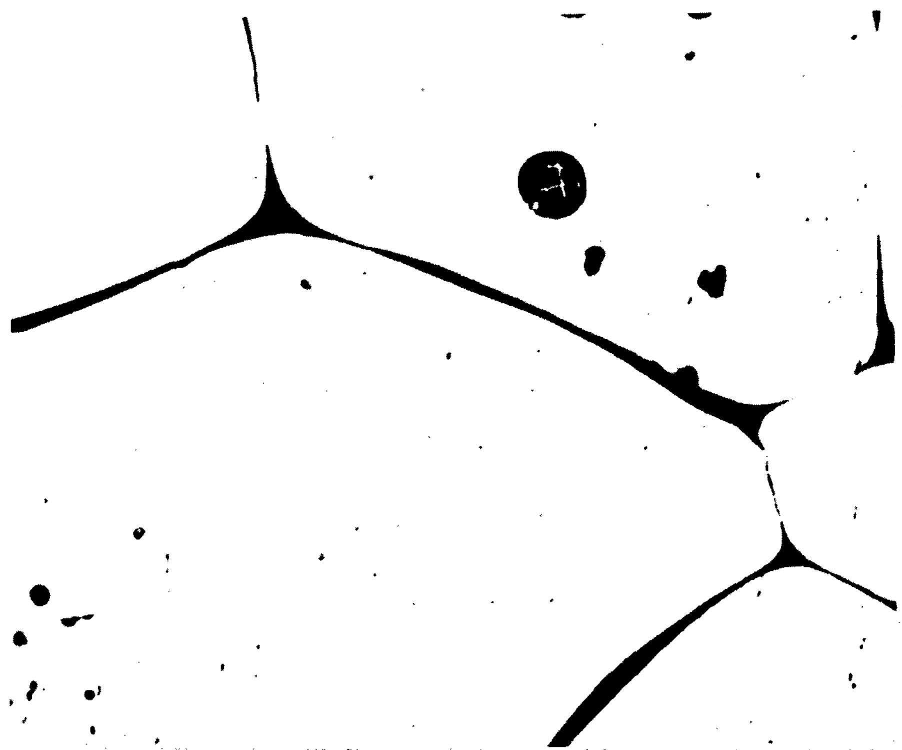
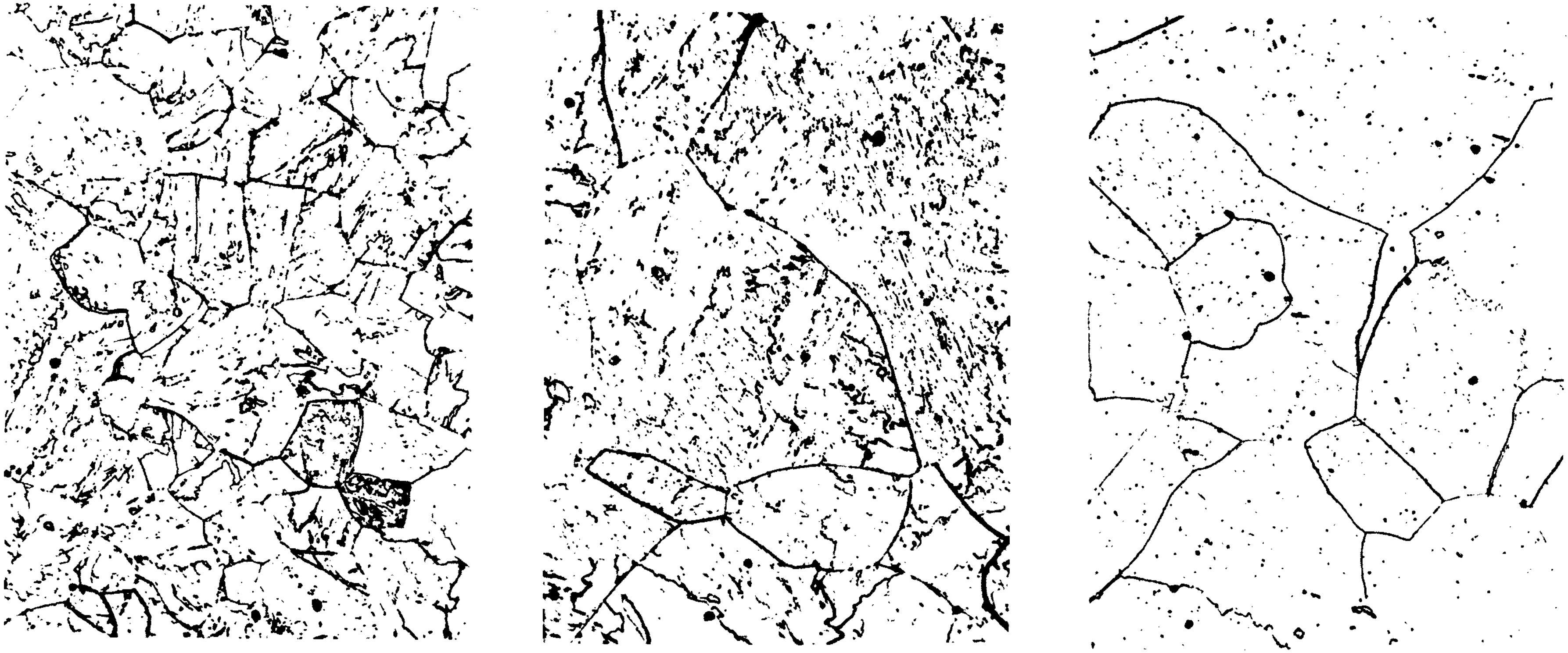


Figure 32

Microstructures of C- 1 Mo-Ti-B Steels (74-75-76)

$2100^{o}F$ - H_2 - 10 Min. - Air-Cool

Mital Etch 500X



Ti/C = 1.4 RB 95

Ti/C = 5.8 RB 97

Ti/C = 11 RB 98

Figure 33

THE EFFECT OF Ti ON THE CREEP & FRACTURE PROPERTIES OF A 0.05 C - 1 Mo - Ti - B STEEL AT 1200° F

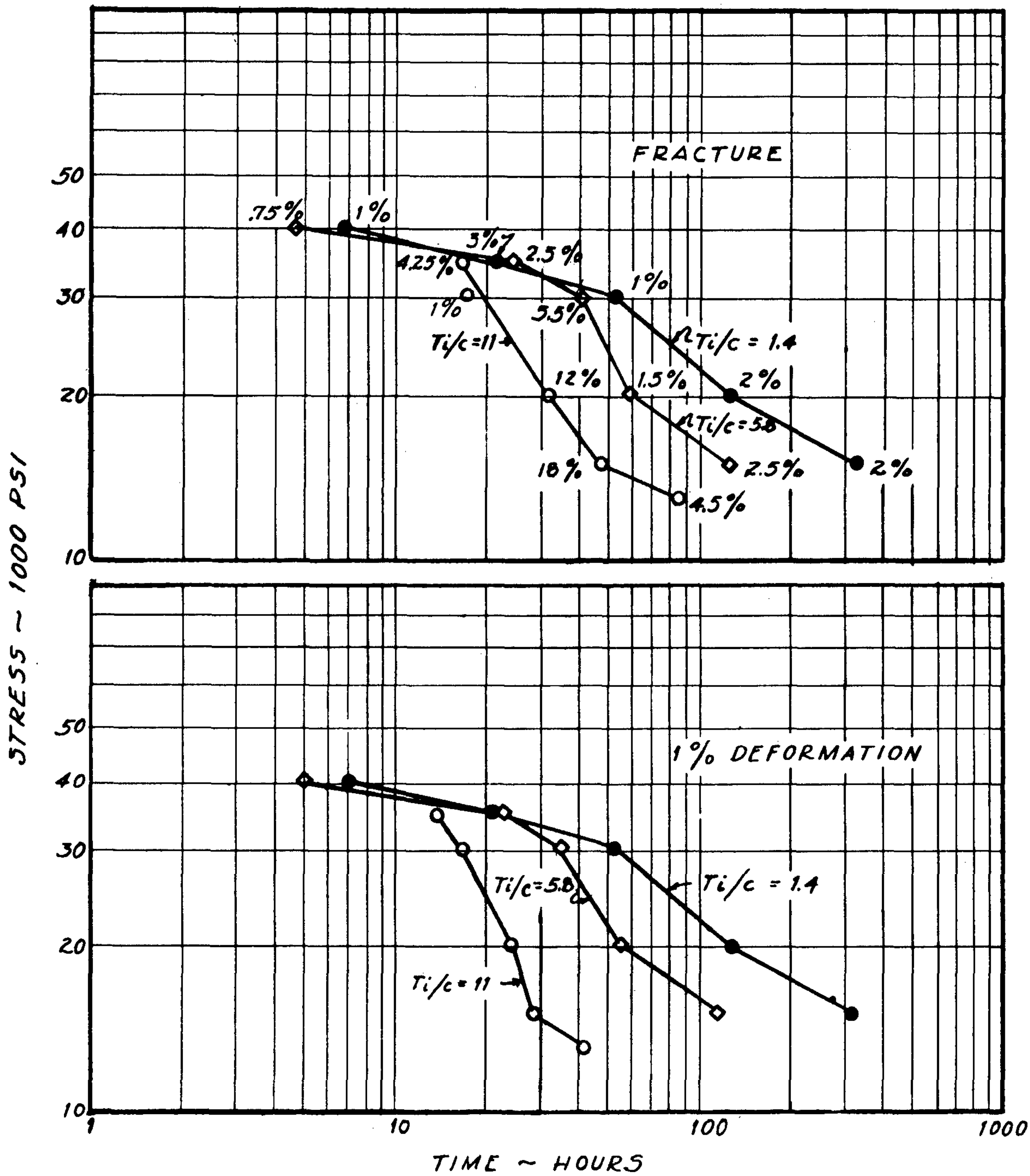


Fig. 34

MICROSTRUCTURES OF 3 Cr- 1 Mo-Ti-B Steels (86-87-88)

$2100^{\circ}F$ - H_2 - 10 Min. - Air-Cool

Nital Etch 500X

.20 C
.27 Ti RB 111

.12 C
.92 Ti RB 99

.13 C
1.46 Ti RB 70

Figure 35

THE CREEP AND RUPTURE PROPERTIES AT 1200°F OF 3 Cr-1Mo-Ti-B STEEL WITH VARYING Ti/C RATIO

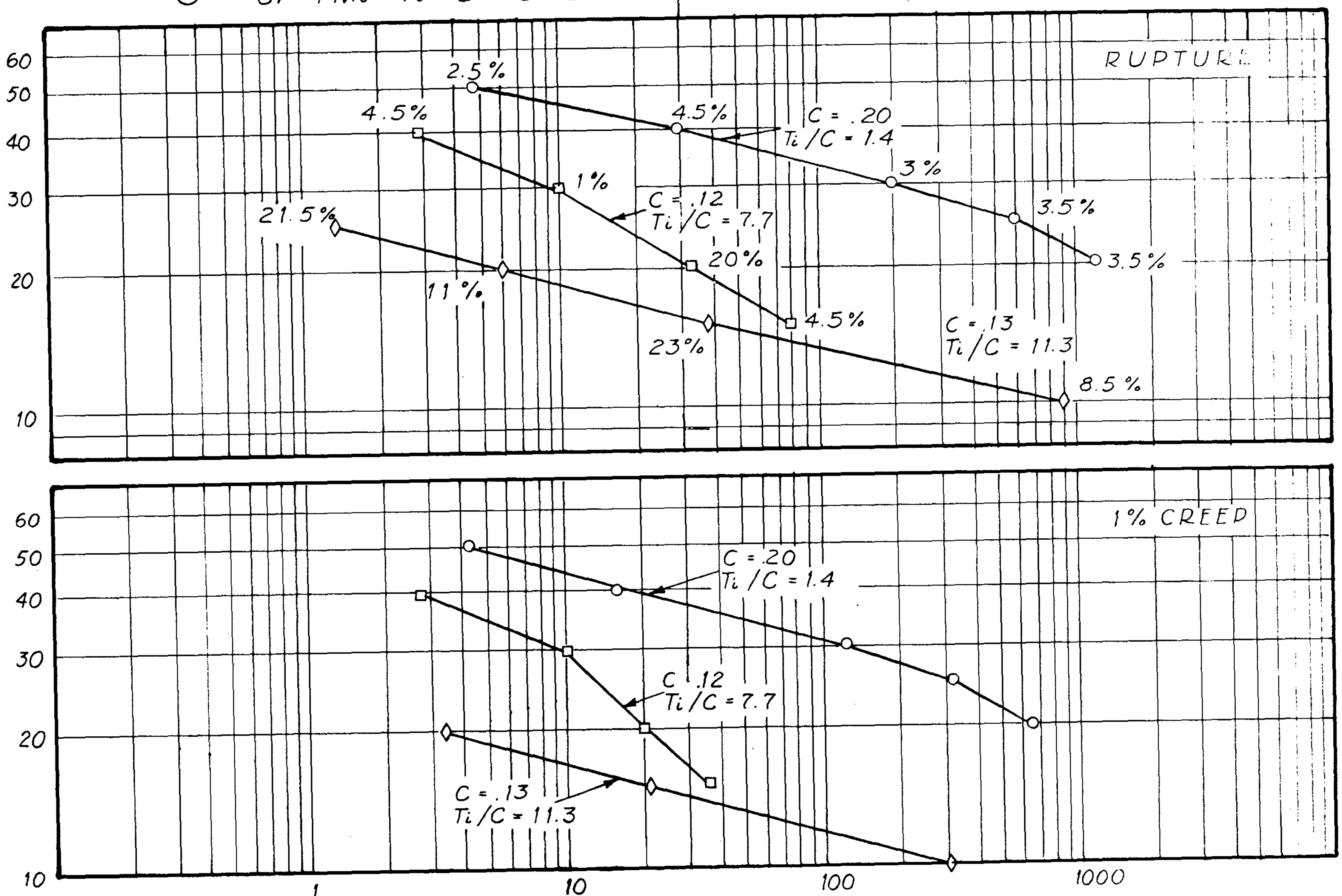


Fig. 36

Microstructures of 3 Cr- 1 Mo-Ti-B Steels With V and W Additions

$2100^{o}F$ - H_2 - 10 Min. - Air-Cool

Nital Etch 500X

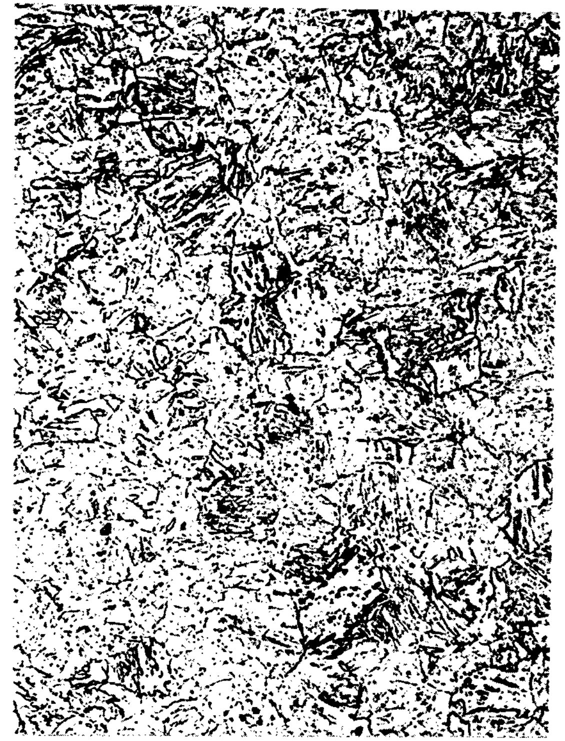
92

.12V RB 115

94

.14W RB 113

95



.10V
.17W RB 114

Figure 37

THE STRESS-RUPTURE PROPERTIES AT 1200°F OF 3 Cr-1Mo-Ti-B STEEL WITH V & W ADDITIONS

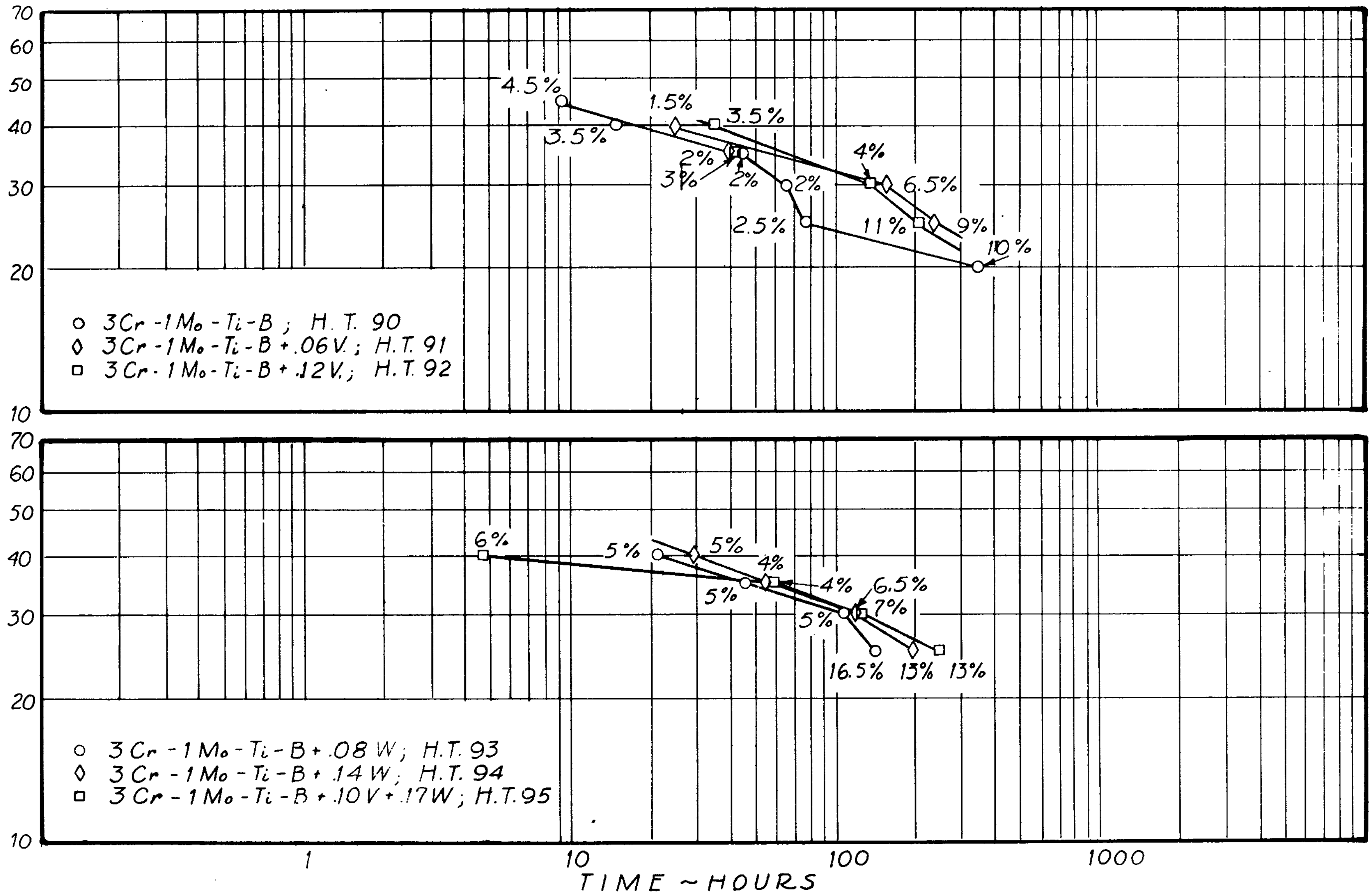


Fig. 38

Back Reflection X-Ray Diffraction Patterns of C-Ti-B Steels

With Varying Carbon

Cobalt Radiation - Film Distance = 5 cm.

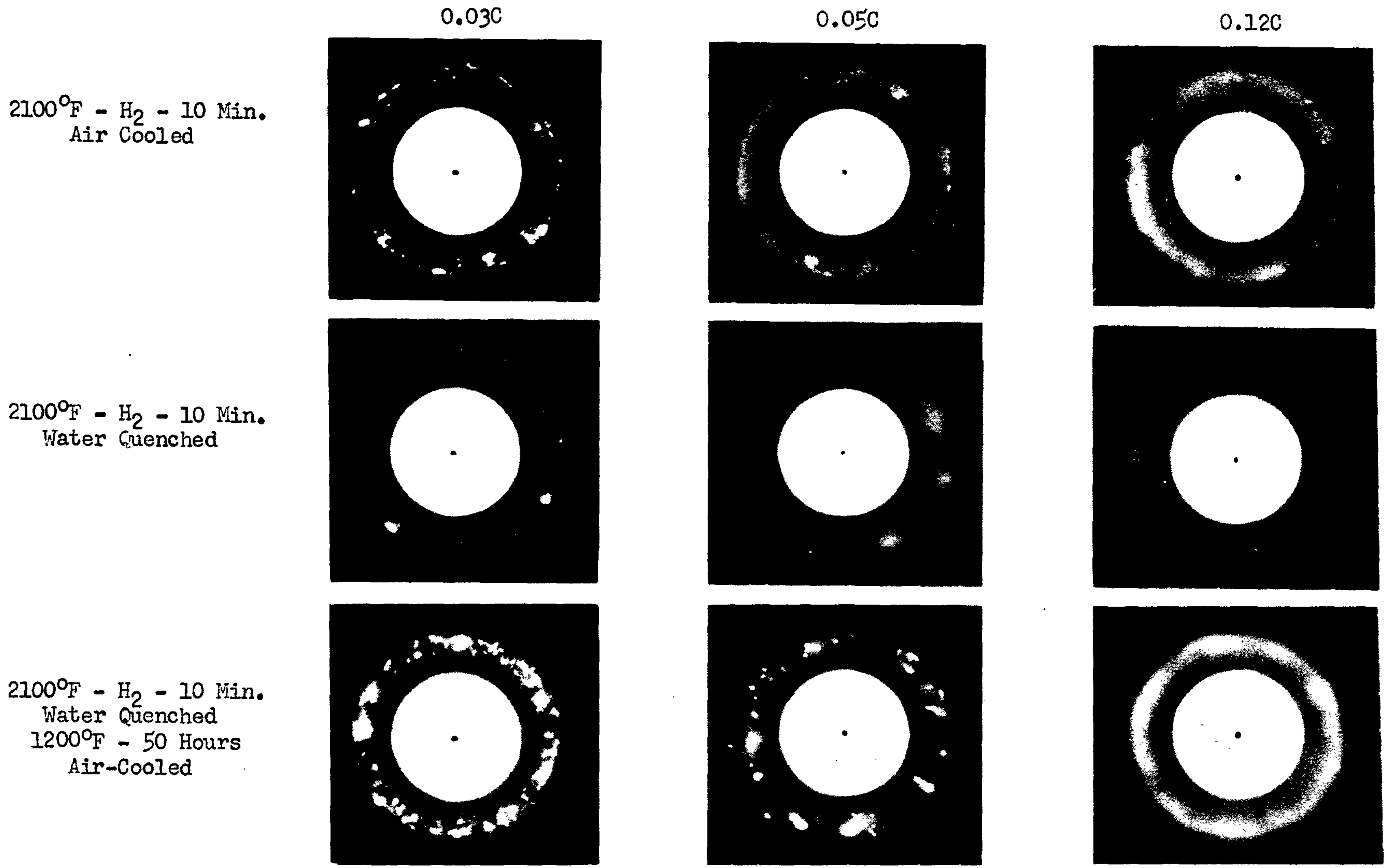


Figure 39

EFFECT OF HIGH TEMPERATURE AGING ON ROOM TEMPERATURE HARDNESS OF C-Ti-B STEEL WITH VARYING CARBON CONTENT

INITIALLY TREATED AT 2100°F - 10 Min. - WATER QUENCHED

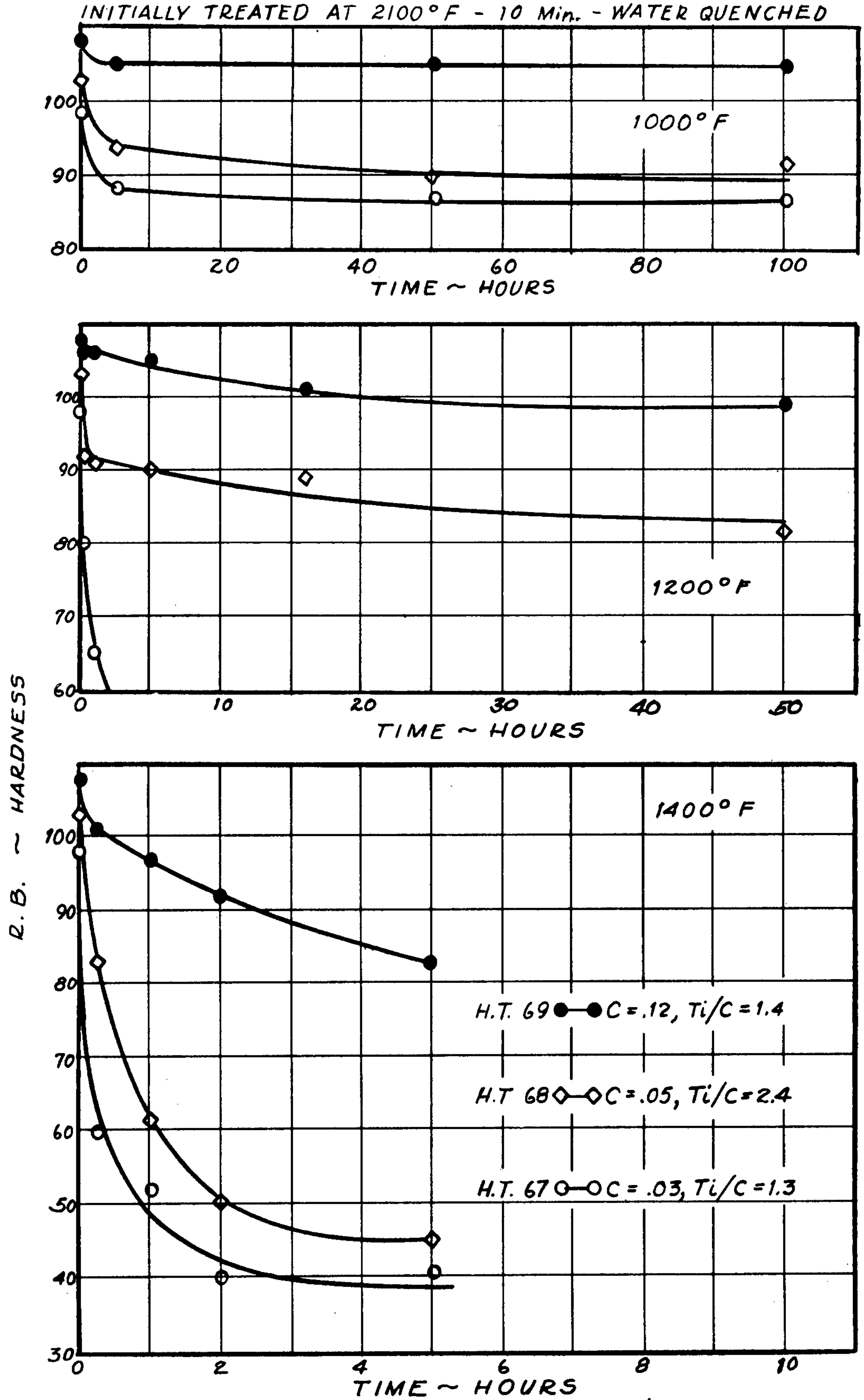


Fig. 40

EFFECT OF HIGH TEMPERATURE AGING ON ROOM TEMPERATURE HARDNESS OF C-Ti-B STEEL

WITH VARYING BORON CONTENT

INITIALLY TREATED AT 2100° F - 10 Min. - WATER QUENCHED

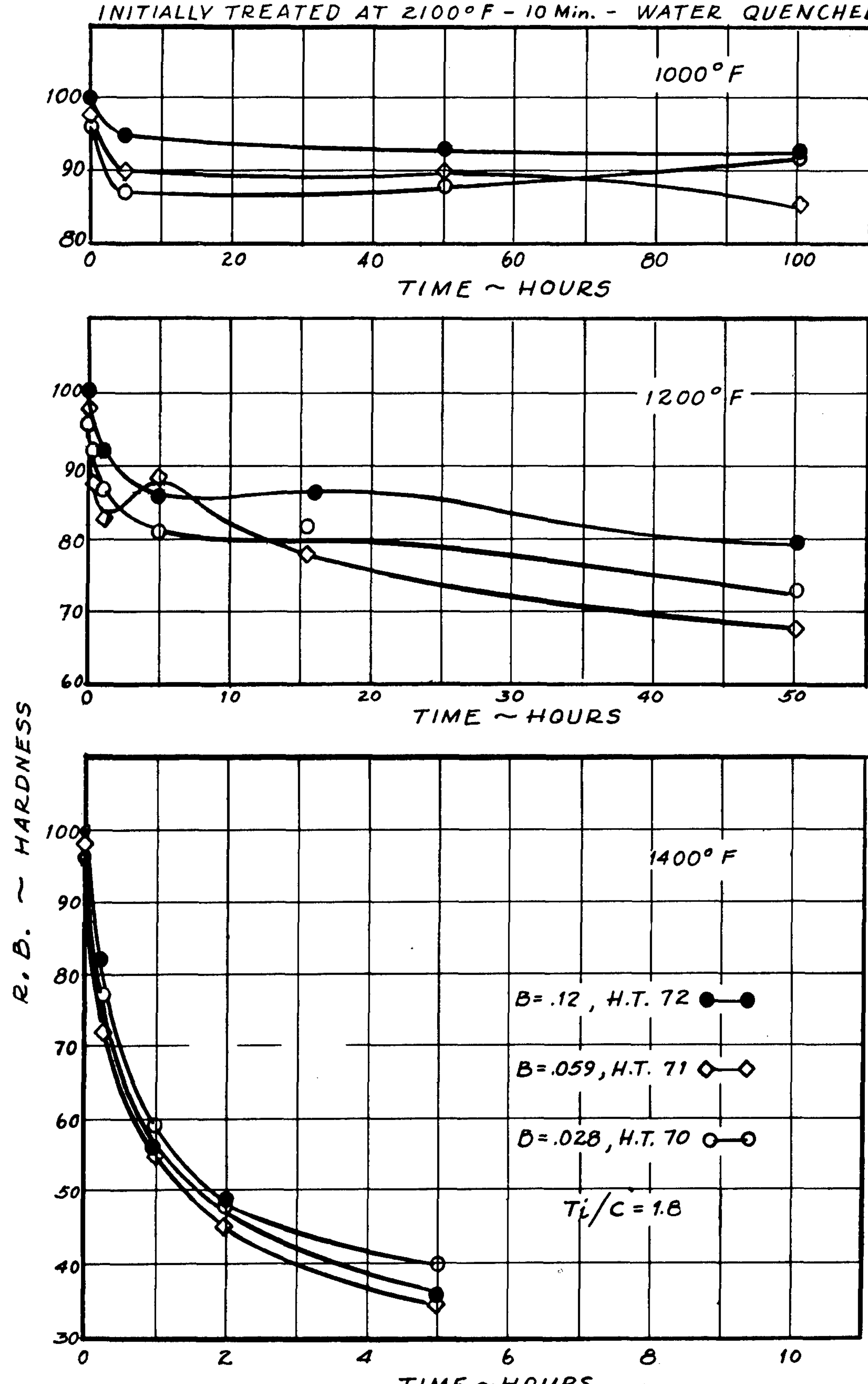


Fig. 41