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THE RUPTURE TEST CHARACTERISTICS OF SIX PRECISION-CAST AND THREE WROUGHT ALLOYS AT 1700° AND 1800° F

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THE RUPTURE-TEST CHARACTERISTICS OF SIX PRECISION-CAST AND THREE WROUGHT ALLOYS AT 1700° AND 1800° F

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SUMMARY

Rupture tests were conducted on nine alloys for time periods up to 400 hours duration at 1700° and 1800° F. Six of the alloys, 6059, Vitallium, 422-19, 61, X-40, and X-50, were in the form of precision castings; and three of the alloys, S590, S816, and Low-Carbon N155, were in the wrought form. An indication of the effect of heat treatment on wrought alloys was obtained by testing Low-Carbon N155 in both the hot-rolled and in the solution-treated conditions. Data on the relative properties of precision castings and wrought material were obtained by including precision castings of S590 and S816 alloys in the investigation.

The results obtained indicated the following ranges in rupture strength:

Type material	Temper-ature (°F)	Range in rupture strength				Highest strength alloy
		100-hour		1000-hour		
		Low (psi)	High (psi)	Low (psi)	High (psi)	
Precision castings	1700	11,000	17,000	8400	14,500	X-40
Wrought bar stock	1700	5,100	9,500	2500	6,600	S590, S816
Precision castings	1800	8,000	11,300	5400	9,800	X-40
Wrought bar stock	1800	4,900	5,600	2800	3,500	S590

All the alloys, except hot-rolled Low-Carbon N155 and cast 6059 at both temperatures, and cast X-40 at 1700° F, showed pronounced decreases in elongation and reduction of area as the time for rupture increased. This was accompanied by marked increases in hardness of the precision cast alloys during rupture testing.

At the temperatures considered in this investigation the better precision-cast alloys had much higher rupture strengths than the better wrought alloys for time periods longer than 10 hours for rupture. From the results on alloys S590 and S816 indications are that an alloy would have higher rupture strength for longer time periods when precision cast than when in the wrought condition. A solution treatment was very beneficial to the rupture strength of Low-Carbon N155 alloy.

Analysis of the results from the viewpoint of chemical composition indicates that such cast alloys as X-40, X-50, S816, and 422-19 are probably inherently stronger than the others. The indications, however, are that grain size and orientation will have a very pronounced effect on rupture strength. Most of the erratic rupture-test results could be attributed to variation in orientation of grains.

The rupture strengths found in this investigation should be considered as only a guide until further information has been obtained regarding possible ranges in strength for the compositions considered.

INTRODUCTION

Improvement of the efficiency of gas turbines is to a large extent dependent on increasing their temperature of operation. This can be accomplished by two means. One is to design turbines which will limit the maximum stress to that permissible at higher temperatures with available materials. The other is to develop alloys which will have better strength than can be obtained from known alloys.

The investigation on which this report is based provided data which can be used as a basis for estimating the maximum rupture strengths available from currently known alloys at 1700° and 1800° F. It also provides a background of properties on which to base future research for better alloys for use at these high temperatures.

The work was conducted for and with the financial assistance of the National Advisory Committee for Aeronautics as part of their program of metallurgical research sponsored at the University of Michigan. The primary purpose of this program is the development of better heat-resistant alloys than are now available for use in aircraft power plants. Prior investigations and the bulk of the current work have been limited to 1200° and 1350° F in accordance with the needs of most present-day service conditions. The intermediate temperatures between 1350° and 1700° F have been covered by the work of the War Metallurgy Committee of the National Defense Research Committee, Office of Scientific Research and Development. The extension of the temperature range to include 1700° and 1800° F was needed to provide data for current or future designs of gas turbines and jet propulsion engines contemplating operation at these temperatures.

The investigation was intentionally limited in scope to an exploratory degree. Nine alloys were arbitrarily selected as being the most promising for service at 1700° and 1800° F. These were relatively new and had not been investigated previously at these temperatures. This selection was based on known strength characteristics at lower temperatures and the chemical composition of the alloy. Only three wrought materials were included for comparison with cast alloys. One wrought material was studied in two conditions of heat treatment. Two alloys were included in both the wrought and the cast form. The type of testing was limited to rupture tests and the maximum time for rupture to 400 hours.

TEST MATERIALS

The chemical compositions of the nine alloys investigated are summarized in table I. They were essentially cobalt-nickel-chromium alloys with various amounts of tungsten, molybdenum and columbium added. The maximum amount of iron in any of the alloys was 30 percent, while in most of them it was only about 1-percent.

The chemical analyses in table I were reported by the suppliers of the test materials. The duplicate analyses for four of the precision-cast alloys do not check closely. This finding probably reflects the difficulty of accurately analyzing alloys of this type except by special techniques.

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The effect of heat treatment on the wrought alloys was checked by studying Low-Carbon N155 alloy in both the hot-rolled and solution-treated conditions. The relationship between wrought and cast alloys of the same composition was compared by testing S590 and S816 in both forms.

With the exception of Low-Carbon N155 alloy in the hot-rolled condition, test specimens of wrought materials were obtained from 0.25-inch-diameter rod. The precision-cast specimens were 0.160 or 0.250 inch in diameter. The specimens and the test materials were restricted to small sizes to approximate the conditions found in applications using sheets or small castings. The details of fabrication reported for each alloy are summarized as follows:

Hot-Rolled Low-Carbon N155: Hot-rolled by Universal-Cyclops Steel Corporation to 7/8-inch square bar stock, with a finishing temperature of about 1700° F. The bars were stress relieved at 1200° F and air cooled.

Solution-Treated Low-Carbon N155: 1/4-inch rod was solution treated at 1250° C (2282° F) for 1/2 hour and cooled in still air and submitted by the Union Carbide and Carbon Research Laboratories, Inc.

Wrought S590: The heat was melted in a 2-ton electric furnace and teemed to 8-inch ingots which were hammer cogged from 2250° F to 3-inch square billets. The billets were hot rolled from 2200° F to 9/32-inch round bars which were then centerless ground to 1/4-inch rounds and submitted by the Allegheny Ludlum Steel Corporation. The samples to be tested were solution treated at 2270° F for 1 hour and quenched in water. In addition the 1700° and 1800° F test specimens were aged for 16 hours at 1700° F and 1800° F, respectively, before testing.

Wrought S816: The heat was melted in an 800-pound electric arc furnace and teemed to 4-inch ingots. The ingots were forged from 2300° F to 2-inch square billets and rolled from 2250° F to 1/4-inch round bar stock. This material was submitted by the Allegheny Ludlum Steel Corporation. Before testing, 3-inch lengths were solution treated at 2300° F for 1 hour and quenched in water. The samples to be tested at 1700° F were aged at 1700° F for 16 hours and those tested at 1800° F were aged for the same time at 1800° F.

Vitallium, 6059, 422-19 and 61 alloys: Precision-cast specimens, made by Haynes Stellite Company, were submitted by the General Electric Company. The specimens were 0.160 inch in diameter with a 1-inch gage length and were cast with threads. All specimens were X-ray inspected for soundness by the General Electric Company. No heat treatment was used prior to testing.

Cast S590 and S816: Precision-cast specimens were made and submitted by the Allegheny Ludlum Steel Corporation. The specimens were 4 inches long and were cast with threads 1/2 inch in diameter. The gage section was 1/4 inch in diameter and 2 inches long. The specimens were checked for soundness by X-ray inspection. No heat treatment was used prior to testing.

X-40 and X-50: Precision castings were supplied by NDRC Project NRC-8. The castings were in the form of test specimens 1/4 inch in diameter with a 1-inch gage length and cast threads. The soundness of the test specimens was checked by X-ray inspection prior to submission. Tests were conducted on the material in the as-cast condition.

PROCEDURE

The experimental steps used to evaluate the alloys at 1700° and 1800° F included:

1. Rupture tests varying in duration from short-time tensile tests to an intended maximum of 400 hours

2. Metallographic examinations of original materials and specimens after rupture testing

3. Hardness tests before and after rupture testing.

The short-time tensile tests were run at a head speed of 0.03 inch per minute after holding the specimen at the test temperature for 1 hour. No strain data were obtained during these tests. The rupture tests of less than 10 hours duration were also conducted in the tensile machine after the specimen had been at the test temperature for 1 hour.

The rupture-test units used for the longer duration tests were of the single specimen type. The specimens were brought up to the test temperature and adjustments made during a period of approximately 24 hours prior to loading. The stress was applied through a simple-beam and knife-edge system. Time-elongation data were obtained by measuring the drop of the beam during the test.

Vickers hardness determinations were made on the samples used for metallographic examination. Samples of the original material and the longest duration-rupture specimens at each temperature were polished, etched, and photographed at 100X and 1000X. Specimens of the cast alloys were macroetched to show grain size.

RESULTS

Sufficient rupture tests were conducted to give data from tests up to at least 400 hours duration. Time-elongation curves are reported for the longer time tests. Metallographic structures and hardness test data were obtained, and an attempt was made to relate these factors to rupture strengths.

Rupture Tests

The usual logarithmic curves of stress versus rupture time (figs. 1 through 12) were plotted from the rupture-test data shown in table II. The rupture strengths defined by these curves are summarized as table III. The probable best rupture strengths at 1700° and 1800° F from available cast and wrought alloys will be of the order shown in the following tabulation:

Type alloy	Temperature (°F)	Highest rupture strengths		
		10-hour (psi)	100-hour (psi)	1000-hour (psi)
Precision castings	1700	20,000	17,000	14,500
Wrought bar stock	1700	15,000	9,500	6,600
Precision castings	1800	14,500	11,300	9,800
Wrought bar stock	1800	9,600	5,600	3,500

The complete range in properties for the alloys considered are compared graphically in figure 13.

The precision-cast alloys with the highest strengths were X-40, X-50, S816, and 422-19, although at 1700° F alloys 61 and Vitallium were nearly as strong. In the wrought condition S816 and S590 had similar strengths which were somewhat higher than those of Low-Carbon N155 alloy. All of the alloys, except hot-rolled Low-Carbon N155 and cast 6059 at both temperatures, and cast X-40 at 1700° F, decreased in elongation and reduction of area as the time for fracture increased. Erratic stress-rupture time and ductility data occasionally were obtained.

Short-time tensile test data were supplied by the General Electric Company from tests in their laboratory on the same four precision-cast alloys which they supplied for this investigation. The General Electric data, included in table IV, show considerable variation in room-temperature properties of the as-cast materials. The ductility at room temperature decreased to a considerable extent after aging at 1700° F. The tensile strength values at 1700° and 1800° F are all considerably higher than those found in the University of Michigan tests. (See table II.) This difference, however, could have been caused by a difference in testing technique.

Comparison of the rupture strengths obtained for cast and wrought S590 and S816 alloys indicated that alloys will have considerably higher strength at 1700° and 1800° F when in the cast condition. A solution treatment has a beneficial effect on the strength of wrought alloys at these temperatures as judged by its effect on Low-Carbon N155 alloy.

Most of the cast specimens had irregular fractures and necked down in other portions of the gage section as well as at the location of the fracture. These nonuniform deformation characteristics were due to variation in the amount and the direction of the deformation in the few relatively large grains composing the gage section of the specimens. Apparently the strength of the individual grains varied with various orientations in relation to the direction of the stress. The irregularities of the broken specimens made it difficult to measure ductility values accurately.

The time-elongation curves (figs. 14 through 36) offer additional information regarding the deformation characteristics of the alloys. In some cases high total elongation was associated with high rupture strength even though the deformation rate was faster than in other alloys with lower total elongation to fracture. Alloy X-40 (fig. 33), in comparison with X-50, 422-19, and cast S816 at 1700° F (figs. 35, 25, 31) are examples. In general, the deformation rate of the cast specimens decreased or remained constant during most of their life. The wrought specimens underwent increasing deformation rates from the beginning of the tests. As is usual, the specimens with low total elongation generally had low rates of deformation and short periods of so-called first-stage creep.

While conducting these tests, trouble was encountered with creep in adapters and threads. The strength of most of the alloys investigated exceeded that of available adapter materials. Many of the time-elongation curves were omitted because of known deformation in the adapters. In several tests fractures occurred through the gage marks with the result that the measured elongation after fracture was low in comparison to the total deformation over the reduced section of the specimen as indicated by the time-elongation curves. The curves shown are reliable for rates of deformation and as to shape of curve. The total deformations shown by the curves do not include initial deformation during the application of the load and may be in error due to movement in the adapter system, especially during the early part of the test.

Metallographic Examination

The original specimen and the most prolonged rupture-test specimens of each alloy were examined metallographically. Typical microstructures before and after testing are shown in figures 37 through 50[1]. Macrostructures showing grain size of cast specimens are included as figure 51.

The higher rupture strength of solution-treated Low-Carbon N155 alloy was associated with a larger grain size, fewer precipitated particles in the grain boundaries, and a more uniform distribution of excess constituents than in the weaker hot-rolled samples. (See figures 37 and 38.) Reprecipitation during testing probably resulted in a distribution of particles more favorable to rupture strength than existed in the hot-rolled condition. After rupture-testing, the hot-rolled specimens showed evidence of considerable oxygen penetration by the presence of easily etched areas near the fracture and along the surface of the specimens. The oxidation resistance apparently was increased by the solution treatment with a consequent improvement in properties.

There seemed to be little relationship between structure and rupture strength in the case of wrought S590 and wrought S816 alloys. (See figures 39 through 42). The S590 material had lower short-time rupture strengths and higher long-time rupture strengths than S816 alloy. Aging S590 at 1700° F produced heavy general precipitation; while aging at 1800° F agglomerated the precipitates to a considerable extent. Relatively little general precipitation occurred in S816 after aging at 1700° F and a large amount after 1800° F. On the other hand, the rupture specimens of S590 alloy did not show so much agglomeration as did those of S816 alloy. The precipitated constituents in S590 may have contributed to the long time strength of the alloy through greater stability.

The four Vitallium type alloys, 422-19, 61, Vitallium and 6059, in general, had the same microstructures. (See figures 43 through 46.) The rupture strengths were also remarkably similar in spite of wide chemical composition differences. The original structure of the Vitallium specimen indicated that its casting conditions were such as to promote precipitation since the normal structure for this alloy in the as-cast condition is similar to that of the other three alloys rather than that shown by figure 44a. Since the structure of the rupture test specimens appeared normal, it is not known if the structure of the original specimen examined was typical of the specimens tested.

A number of specimens were examined to ascertain if there was a relationship between microstructure and the erratic rupture test results. In general, it appeared that the orientation of the relatively few large grains composing the gage section (fig. 51) was the cause of variable results. A more favorable orientation of some grains appeared to cause greater resistance to deformation. Some orientations of the dendritic pattern also appeared to promote crack propagation and low strength more than others. One other structural characteristic associated with low strength was the small grains outlined by envelopes of excess constituent shown in figure 44b for the abnormally low strength Vitallium specimen tested at 1800° F under a stress of 5500 psi.

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[1]The photomicrographs were reduced to 69.5% of their original size during printing.

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The structures of cast S590 and S816 (figs. 47 and 48, respectively) were considerably different from the other cast alloys. In the as-cast condition the dendritic pattern of the excess constituents was finer and more complete. While there were differences in structure between S590 and S816 castings, there is no basis for attributing the higher rupture strength of S816 to this cause. The higher cobalt content of S816 apparently introduced an eutectic into the structure and suppressed the completeness of the dendritic pattern in comparison to S590.

The high rupture strengths of X-40 and X-50 alloys were due apparently to the general precipitation which occurred during rupture testing. (See figures 49 and 50.) This structural change did not occur in the Vitallium type alloys, although they had the same type of initial structure. The lower strength at 1800° F of X-50 alloy may have been due to a less stable precipitate since agglomeration of precipitates after rupture testing was more extensive than in X-40.

The dendritic pattern in the X-50 samples was more pronounced and continuous than in the other cast alloys. Consequently it is probable that the orientation of the grains had a considerable effect on the crack propagations during failure in rupture tests. This condition probably accounts for the erratic rupture times for tests on this alloy.

Vickers Hardness Surveys

The Vickers hardness of the samples used for metallographic examination were measured and included with the photomicrographs. In general the results of the hardness tests did not correlate with rupture strengths. It must be concluded, as is often the case, that the factors which increase rupture strength at elevated temperatures do not necessarily increase hardness at room temperature.

The Low-Carbon N155 specimens in the hot-rolled condition originally had a higher hardness than those which had been solution treated. After rupture testing the hardness of the hot-rolled specimens was the lowest. This might be interpreted to indicate that the higher rupture strength of the solution-treated specimens was due to a more stable structure.

Wrought S816 alloy had considerably higher hardness before and after rupture testing than did S590 alloy. The S590 alloy, however, had higher rupture strengths for the longer time periods.

All the Vitallium type alloys increased in hardness to a marked extent during testing at 1700° and 1800° F. Inspection of the hardness data, however, shows that no correlation exists with rupture strength. It was noted that the hardness of the Vitallium specimen tested at 1800° F varied widely in two different grains.

The higher strength cast S816 alloy had higher hardness and remained higher in hardness after prolonged testing at 1700° and 1800° F than was the case for cast S590 alloy.

Both X-40 and X-50 were quite hard initially and showed pronounced increases during testing at 1700° and 1800° F. The latter alloy, however, did not have quite such high hardness after testing at 1800° F, which adds to the instability theory for its lower strength at this temperature. The hardness values were not greatly different from those for the other cast alloys; so here again higher strength cannot be correlated with hardness.

DISCUSSION OF RESULTS

The main value of this investigation is the resultant data showing the order of magnitude of the rupture strengths at 1700° and 1800° F for the better alloys now available. As such, this information can serve as a guide in the use or selection of alloys for service at these temperatures. The data should be used conservatively until further investigations disclose the uniformity with which the properties can be reproduced.

Several important general guiding principles are disclosed by the data:

1. The better cast alloys have considerably higher rupture strength than the better wrought alloys at 1700° and 1800° F.

2. As judged by the results obtained on S590 and S816 alloys, a precision casting of a given analysis will have higher rupture strength at these temperatures than the wrought form, except for very short time periods.

3. The beneficial effect of solution treatment on Low-Carbon N155 alloy indicates that the rupture strengths of alloys at these temperatures can be changed by heat treatment.

These findings are in agreement with the general metallurgical principles that the cast form of an alloy has better strength at the more elevated temperatures than the wrought form, and that a solution treatment is beneficial to strength at the higher temperatures.

There are relatively few possible generalizations regarding the effect of alloy content on strength. Alloy S816 differs from S590 only by the substitution of cobalt for iron. This did not result in any improvement in rupture strength for alloy S816 in the wrought condition. In the cast form, however, the high cobalt resulted in a very decided improvement in rupture strength. Comparison of Low-Carbon N155 with wrought S590 and S816 seems to indicate a slight advantage for the higher carbon, tungsten, molybdenum, and columbium of the latter two alloys at 1700° F with little effect at 1800° F. The pronounced effects of heat treatment on these wrought alloys, however, indicate that heat treatment may have been more effective in influencing strength than alloy content.

The relative strengths of the cast alloys varied both with the time period and the temperature considered. The four cast alloys which had the highest rupture strengths in most cases, X-40, X-50, 422-19, and S816, all carried from 10 to 20 percent nickel as a consistent variable from the lower strength alloys. Alloy X-40 indicates that a combination of 10 percent nickel and slightly more tungsten than the tungsten or molybdenum in the other alloys is an optimum combination.

The analyses represented by X-40 and X-50 were outstanding at 1700° F; while those containing high nickel or iron, represented by 6059 and cast S590, were poor. At 1800° F X-40, X-50, and cast S816 were the best alloys with the others being somewhat weaker down to 61 alloy which seems to have been abnormally weak on the basis of the composition differences. At 1700° F the difference between 422-19 and 6059 on the basis of 100- and 1000-hour rupture strengths was only 3000 psi. These two alloys were the strongest and weakest of the four alloys tested in the form of 0.160-inch-diameter castings. At 1800° F the difference between alloy 61 and 422-19 was only 1700

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psi. Considering the variation of test points from straight line stress-rupture time curves, these differences are surprisingly small for the wide variation in composition considered.

The higher strength alloys seemed to be associated with general precipitation in the structure during testing. The microstructures and macrostructures also indicate that it is entirely possible that grain size and orientation may have been responsible for many of the variations between alloys. These structural characteristics appeared to account for the erratic results. It may well be that, if more tests had been run so as to obtain better average rupture strengths, more definite trends from composition differences would have been observed. In certain cases the orientation of the dendritic pattern apparently allowed cracks to propagate and cause failure earlier than other orientations. An example of this type finding is shown by the two time-elongation curves for alloy X-50 at 1700° F under 16,000 psi (fig. 35). In other cases the variation in orientation resulted in wide variations in resistance to deformation. One grain deforming more than an adjacent grain is evidence of this effect. The various grains also would deform more in one direction than the other. This also is reflected by the time-elongation curves, of which the two curves for alloy X-50 at 1700° F under 14,000 psi are an example.

The effect of specimen size and casting factors could not be evaluated, although these may have influenced the relative rupture strengths. The Vitallium, 6059, 61, and 422-19 specimens were made at the same place and were 0.160-inch in diameter. The X-40 and X-50 specimens were cast as 0.250-inch-diameter specimens with a 1-inch-gage section by the same organization that made the 0.160-inch-diameter specimens. While improbable, it is possible that the superiority of these two alloys in the rupture test may be due partially to the variation in section size. The cast S590 and cast S816 samples were made by a different organization and were cast with a gage section 0.250 inch in diameter and 2 inches long. In this case the superiority of S816 over S590 definitely appears to be one of analysis. There is no way, however, to evaluate possible section size effects in their relation to the other alloys.

In many of the alloys the total elongation to fracture was only 2 to 4 percent in the longer duration tests. This finding indicates that it would be possible that brittle failures might occur during service at these temperatures. The pronounced increase in hardness of most of the cast alloys during testing at 1700° and 1800° F suggests that a precipitation hardening effect was the cause of the low ductility to fracture. Alloy X-40 at 1700° F was an exception to this since it developed high hardness but retained its ductility. The solution-treated Low-Carbon N155 and wrought S816 and S590 alloys had surprisingly high hardness values after testing at 1700° and 1800° F, indicating that precipitation also occurred in these wrought alloys. Hot-rolled Low-Carbon N155 alloy, however, decreased in hardness during testing at 1700° F. It is instructive to note that hot-rolled Low-Carbon N155 alloy had high ductility, even though it was the only material which suffered appreciable surface attack during testing.

The metallographic examinations did not explain the relative rupture strengths of the various alloys. Further experience with the structures and properties probably will provide a better basis for relating the findings.

CONCLUSIONS

Data have been obtained at 1700° and 1800° F for nine alloys which indicate the probable range in the rupture-test properties for the known alloys which were considered to offer the greatest promise for service at these temperatures. Analysis of the data led to the following general conclusions:

1. The better precision-cast alloys have higher rupture strengths than the best wrought alloys. Precision castings of a given analysis are better in this test than the same alloy in the wrought form.

2. Rupture-test characteristics of a wrought alloy at 1700° and 1800° F can be drastically altered by heat treatment. A solution treatment at high temperatures promotes high rupture strength.

3. There are relatively few general relationships, as yet, between chemical composition and rupture-test properties at 1700° and 1800° F. The compositions represented by X-40, X-50, 422-19, and cast S816 alloys, however, do appear to be outstanding.

4. The present state of knowledge was not sufficient to relate microstructures to the rupture-test results. Grain orientations in the cast alloys did appear to have a pronounced effect on rupture-test-deformation characteristics.

5. All of the alloys were unstable during the rupture tests. Most of the cast alloys underwent pronounced increase in hardness, apparently due to a precipitation reaction. The same reason was probably the cause for the low total deformations of most of the longer time rupture specimens.

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TABLE I

CHEMICAL COMPOSITION OF ALLOYS TESTED AT 1700° AND 1800° F

Alloy	Heat number	Chemical composition (percent)										
		C	Mn	Si	Cr	Ni	Co	Mo	W	Cb	Fe	N_2
Low-Carbon N155 Hot-rolled	A11534	0.15	1.74	0.37	21.66	19.40	20.48	2.76	1.90	0.79	[a](30)	.14
Low-Carbon N155 Solution-treated	------	.13	----	----	21.5	19.0	19.0	3.10	2.10	1.05	[a](32)	.15
Wrought S590	41572	.47	1.35	.82	19.40	19.07	19.26	4.03	4.00	3.87	[a](27)	---
Wrought S816	50757	.38	.82	.25	18.87	19.70	45.64	4.04	4.71	3.43	2.94	---
6059	6437	[b].50 [c].396	.35 .20	.77 .76	26.85 24.61	35.15 33.50	Bal. 33.55	5.00 5.34	---- ----	---- ----	1.04 1.00	--- ---
Vitallium	V-4705	[b].24 [c].21	.98 ----	.63 ----	27.60 26.66	3.06 [d](Co + Ni	Bal. 65.56%)	5.13 5.57	---- ----	---- ----	1.76 ----	--- ---
422-19	1	[b].48 [c].54	.52 .32	.63 .63	29.10 26.24	13.90 10.81	Bal. 53.89	6.20 6.21	---- ----	---- ----	1.20 1.10	--- ---
61	6285	[b].44 [c].43	.54 .64	.48 .58	23.70 24.49	0.16 (Co + Ni	Bal. 69.90%)	---- ----	5.18 5.15	---- ----	1.04 1.00	--- ---
Cast S590	------	.57	.67	.63	20.11	20.64	20.04	3.63	4.50	4.02	[a](25)	---
Cast S816	------	.41	.42	.56	19.43	19.80	42.81	3.61	3.42	4.48	[a](5)	---
X-50	------	.76	----	----	22.57	20.05	40.70	----	12.17	----	[a](2.5)	---
X-40	------	.48	.64	.72	25.12	9.69	55.23	----	7.23	----	0.55	---

[a]By difference.

[b]General Electric Co., River Works, analysis.

[c]Haynes-Stellite Co. analysis.

[d]By difference from 98 percent.

TABLE II

RUPTURE TEST DATA AT 1700° AND 1800° F FOR TWELVE ALLOYS

Alloy	Temperature (°F)	Stress (psi)	Rupture time (hr)	Elongation (percent)	Reduction of area (percent)
Wrought Low-Carbon N155	1700	23,800	S.T.T.T.	47	54.3
Solution-treated		15,000	4.25	53	55.3
		8,500	60.0	28	26.7
		7,000	134.0	16	17.8
		5,500	516.0	10	13.3
Wrought Low-Carbon N155	1800	16,100	S.T.T.T.	67	58.6
Solution-treated		11,000	4.12	44	44.7
		5,000	108.0	11	6.7
		4,000	192.0	[a]9.7	12.0
		3,500	415.0	8	11.5
		3,000	Broke in threads after 603 hr		
Low-Carbon N155	1700	23,000	S.T.T.T.	48	61.3
Hot-rolled		17,000	0.84	41.5	59.8
		10,000	11.0	28.5	38.2
		8,000	24.5	31	34.7
		6,000	49.5	32	37.5
		4,500	169.5	25	27.5
		3,500	318.0	32	28.5
Wrought S816	1700	31,300	S.T.T.T.	[a]13	16.7
		22,000	1.18	[a]9	24.7
		12,000	63.0	11	13.3
		11,000	64.0	18	18.4
		9,000	168.0	[a]8	5.0
		7,500	111.0	6	5.0
		7,000	390.0	15	16.7
Wrought S816	1800	22,200	S.T.T.T.	32	33.4
		14,000	2.51	[a]19	20.2
		6,000	66.5	[a]7	14.0
		5,000	119.0	10	9.7
		4,000	323.5	[a]7	9.1
Wrought S590	1700	24,100	S.T.T.T.	52	56.1
		16,000	2.43	59	62.5
		11,833	36.0	40	60.9
		9,000	94.5	31	44.7
		8,000	252.0	28	31.9
		7,200	600.0	[a]14	17.8
Wrought S590	1800	18,500	S.T.T.T.	63	61.7
		12,000	2.65	53	54.3
		7,000	36.0	33	38.8
		5,000	185.5	23	21.2
		4,200	374.0	[a]3.5	12.1
Cast 6059	1700	29,300	S.T.T.T.	27	40.8
		20,000	1.18	44	66.2
		13,000	60.0	20	54.3
		11,000	190.0	13	31.9
		10,000	362.5	12	32.4

See footnotes at end of table.

TABLE II.- (CONTINUED)

Alloy	Temperature (°F)	Stress (psi)	Rupture time (hr)	Elongation (percent)	Reduction of area (percent)
Cast 6059	1800	22,500	S.T.T.T.	25	53.6
		14,000	4.45	30	73.1
		10,000	44.0	23	45.0
		9,000	156.0	10	26.7
		8,000	281.0	19	54.3
Cast Vitallium	1700	34,000	S.T.T.T.	19	66.2
		25,000	0.43	35	52.8
		14,000	35.0	31	58.0
		13,000	144.0	8	27.7
		10,500	625.0	6	21.0
Cast Vitallium	1800	25,200	S.T.T.T.	24	54.3
		15,000	2.28	48	64.0
		10,000	76.0	17	42.2
		9,000	97.5	13	27.7
		8,000	132.0	10	16.7
		7,500	527.0	[a]5	15.6
		6,800	Discontinued after 1244 hr		
		5,500	895.0	12	15.6
Cast 422-19	1700	35,000	S.T.T.T.	18	41.8
		21,000	5.2	15	39.8
		18,000	16.5	13	29.0
		15,000	47.0	6	28.2
		14,000	170.0	7	16.0
		13,000	346.0	6	13.3
Cast 422-19	1800	24,700	S.T.T.T.	28	45.6
		16,000	3.37	24	36.9
		11,000	44.0	12	28.8
		10,000	161.0	10	17.0
		9,000	204.0	11	39.0
		8,000	460.0	4	10.9
Cast 61	1700	29,800	S.T.T.T.	25	51.9
		18,000	7.15	32	53.6
		14,000	48.0	4	4.4
		13,000	156.0	7	9.7
		12,000	717.0	5	5.0
Cast 61	1800	24,200	S.T.T.T.	25	47.4
		15,000	2.27	19	36.0
		11,000	29.5	19	51.9
		10,000	268.0	[a]5	8.5
		9,000	100.0	9	12.1
		8,000	146.0	7	7.5
		7,000	283.5	9	6.2
Cast S590	1700	26,450	S.T.T.T.	13.5	29.5
		25,400	S.T.T.T.	16.5	32.2
		16,000	8.6	16.5	40.5
		12,000	29.0	15.5	39.8
		10,000	200.0	14	20.5
		9,500	378.5	8	13.5

See footnotes at end of table.

TABLE II.- (CONTINUED)

Alloy	Temperature (°F)	Stress (psi)	Rupture time (hr)	Elongation (percent)	Reduction of area (percent)
Cast S590	1800	18,600	S.T.T.T.	27	40.5
		13,000	4.43	25.5	34.2
		8,000	96.0	14	33.0
		7,500	185.0	11	22.6
		7,000	137.0	12	27.7
		6,500	480.0	5	14.6
Cast S816	1700	34,000	S.T.T.T.	13	24.8
		21,000	4.31	7	21.3
		15,000	60.0	7	19.0
		14,500	84.0	14.5	22.0
		14,000	290.5	3.5	5.5
		13,000	327.0	6	8.7
Cast S816	1800	25,900	S.T.T.T.	12	21.3
		17,000	3.5	16	26.6
		11,000	74.0	11	17.7
		10,000	156.0	9	16.7
		9,000	385.0	7	16.2
Cast X-40	1700	32,600	S.T.T.T.	38	43.4
		23,000	1.03	45	55.4
		18,000	40.0	26	40.0
		17,000	169.0	20	43.2
		14,000	828.0	25	43.2
Cast X-40	1800	25,950	S.T.T.T.	28	45.7
		14,000	Discontinued after 9.64 hr		
		12,000	41.0	34	55.5
		11,000	196.0	[a]19	31.9
		10,000	688.0	[a]9	17.8
Cast X-50	1700	20,000	8.67	27	50.1
		18,000	23.0	32	34.5
		17,000	52.0	20	36.0
		16,000	241.5	[a]5	13.4
		16,000	426.0	[a]4	2.5
		15,000	227.0	14	27.2
		14,000	204.0	13	23.6
		14,000	848.0	2	5.6
Cast X-50	1800	15,000	5.48	36	30.5
		10,000	75.5	17	38.4
		9,000	276.0	[b]---------	12.2
		8,000	665.0	9	9.1

S.T.T.T. short-time tensile test

[a]Fractured through gage mark.

[b]Part of specimen near fracture missing.

TABLE III

RUPTURE STRENGTHS OF CAST AND WROUGHT ALLOYS AT 1700° AND 1800° F

Alloy	Temperature (°F)	Stress for rupture in indicated time periods				
		1-hour	10-hour	100-hour	400-hour	1000-hour
PRECISION CASTINGS						
X-40	1700	24,000	20,000	17,000	15,500	14,500
X-50	1700	22,000	19,000	16,000	15,000	14,000
422-19	1700	25,000	19,000	15,000	12,500	11,500
S816	1700	24,000	19,000	14,500	12,500	11,500
61	1700	21,000	17,000	14,000	12,500	11,500
Vitallium	1700	22,500	17,000	13,000	11,000	10,000
6059	1700	20,500	16,000	12,000	10,000	8,600
S590	1700	19,500	14,500	11,000	9,400	8,400
WROUGHT ALLOYS						
S816	1700	23,000	15,000	9,500	7,300	6,100
S590	1700	18,500	13,000	9,400	7,600	6,600
Solution-treated Low-Carbon N155	1700	20,000	12,500	7,600	5,800	4,800
Hot-rolled Low-Carbon N155	1700	16,000	10,000	5,100	3,300	2,500
PRECISION CASTINGS						
X-40	1800	15,500	13,000	11,300	10,300	9,800
X-50	1800	18,000	13,500	10,000	8,600	7,700
422-19	1800	19,000	14,000	10,000	8,000	7,100
S816	1800	20,000	14,500	10,500	9,000	7,800
61	1800	16,500	12,500	8,600	6,500	5,400
Vitallium	1800	16,500	12,500	9,400	7,900	7,000
6059	1800	17,000	12,500	9,300	7,700	6,800
S590	1800	16,000	11,500	8,000	6,600	5,800
WROUGHT ALLOYS						
S816	1800	17,500	9,600	5,300	3,700	3,000
S590	1800	15,000	9,200	5,600	4,200	3,500
Solution-treated Low-Carbon N155	1800	15,500	8,700	4,900	3,500	2,800

TABLE IV

SHORT-TIME TENSILE PROPERTIES OF FOUR PRECISION-CAST ALLOYS
(General Electric Company Data)

Alloy	Treatment	Temperature (°F)	Tensile strength (psi)	Yield strength 0.02% offset (psi)	Elongation % in 1 in.	Reduction of area (percent)
Vitallium	As cast	Room	106,000	66,000	16	22.4
	-----do-----	1700	42,470	--------------	27	52.4
	-----do-----	1800	33,265	--------------	35	52.4
	Cast and aged at 1700° F - 16 hours A.C.	Room	99,150	57,650	[a] 4	11.2
	-----do-----	1800	32,910	--------------	49	63.1
422-19	As cast	Room	98,100	55,100	5	11.9
	-----do-----	1700	45,180	--------------	17	26.6
	-----do-----	1800	36,290	--------------	24	33.7
	Cast and aged at 1700° F - 16 hours A.C.	Room	94,500	56,200	1	0.65
	-----do-----	1700	47,135	--------------	18	33.3
	-----do-----	1800	37,800	--------------	21	38.7
6059	As cast	Room	82,550	46,900	7	10.3
	-----do-----	1700	42,950	--------------	23	26.5
	-----do-----	1800	33,400	--------------	24	50.3
	Cast and aged at 1700° F - 16 hours A.C.	Room	76,800	41,100	3	3.4
	-----do-----	1700	45,430	--------------	16	34.0
	-----do-----	1800	33,690	--------------	[b]26	41.7
61	As cast	Room	103,400	58,350	7	11.2
	-----do-----	1700	37,475	--------------	[c] 7	35.7
	-----do-----	1800	33,115	--------------	32	40.6
	Cast and aged at 1700° F - 16 hours A.C.	Room	108,500	53,900	6	14.2
	-----do-----	1700	43,580	--------------	18	35.7
	-----do-----	1800	33,050	--------------	27	39.5

[a]Bar broke at measured section.

[b]Bar broke near measured section.

[c]Bar broke outside of measured section.

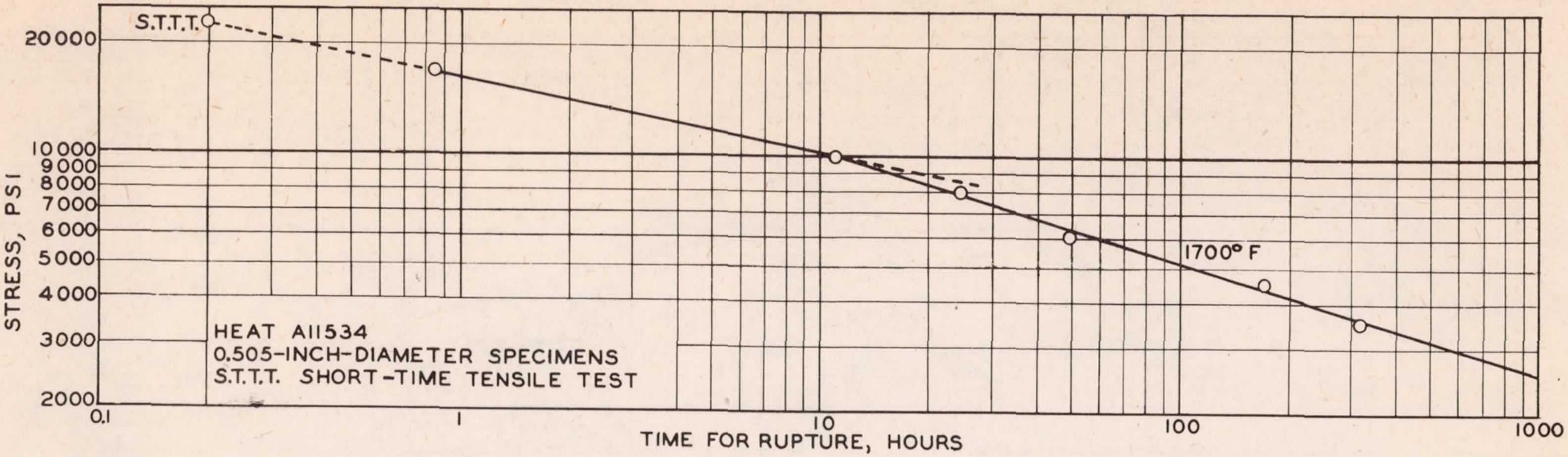

FIGURE 1.- STRESS-RUPTURE TIME CURVES AT 1700° F FOR HOT-ROLLED LOW-CARBON N155 ALLOY.

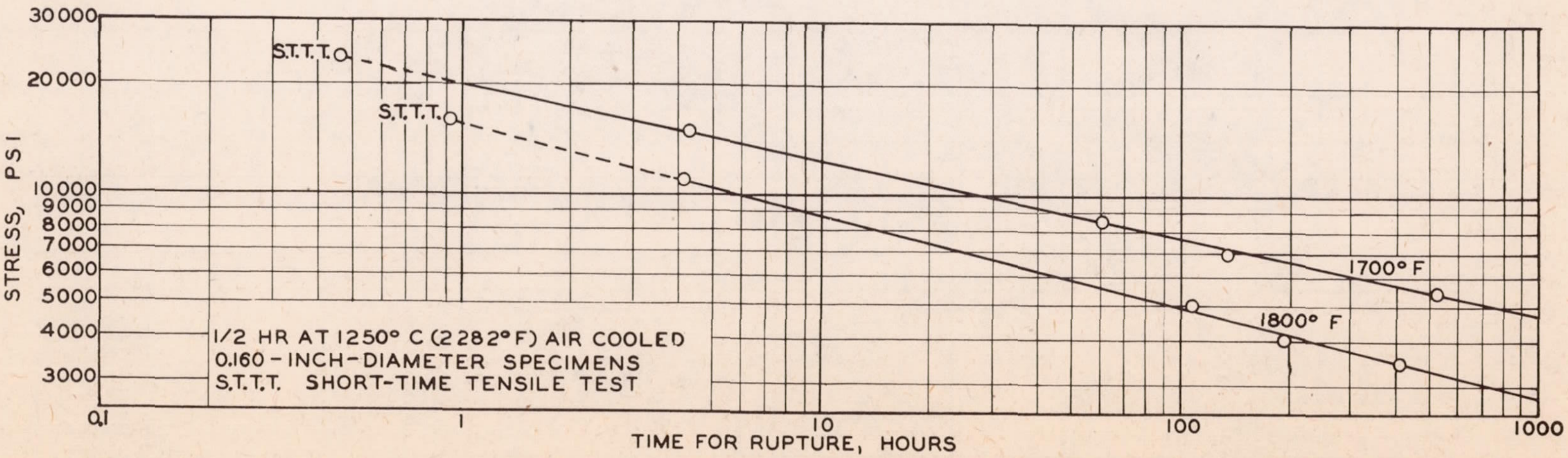

FIGURE 2.- STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR SOLUTION-TREATED LOW-CARBON N155 ALLOY.

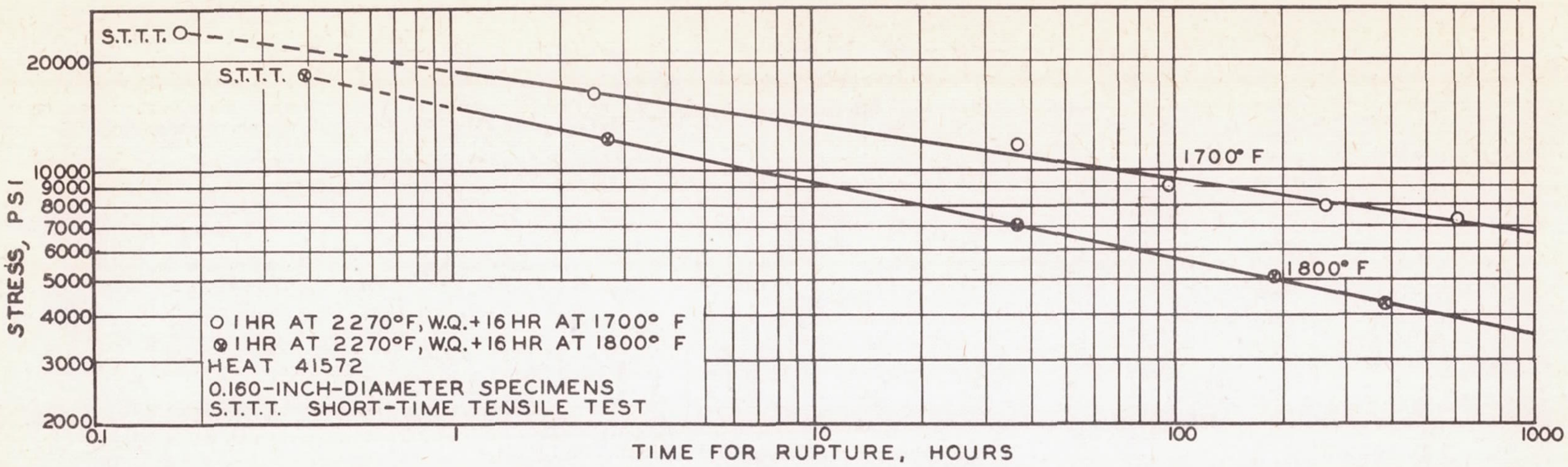

FIGURE 3.- STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR WROUGHT ALLOY S590.

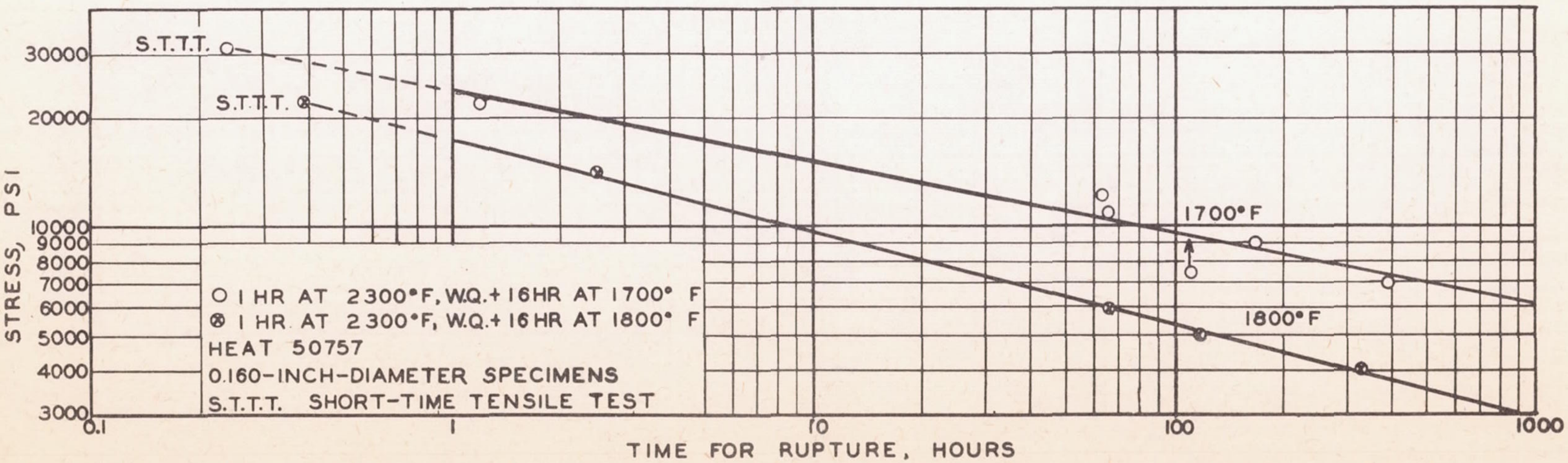

FIGURE 4.- STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR WROUGHT ALLOY S816.

C1-M

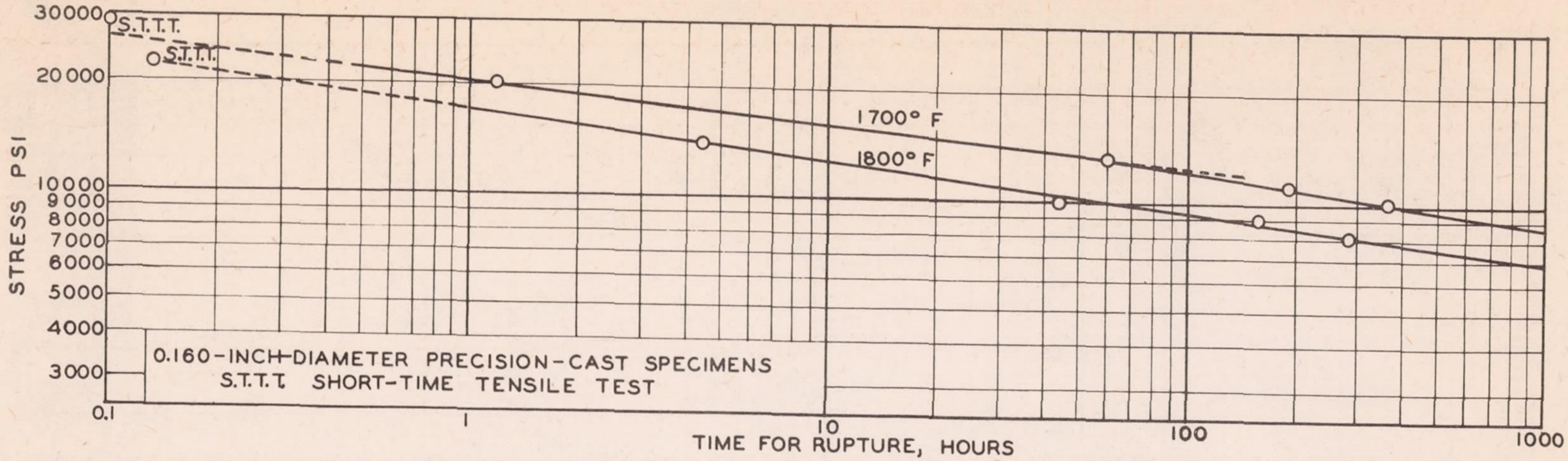

FIGURE 5.- STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR CAST ALLOY 6059.

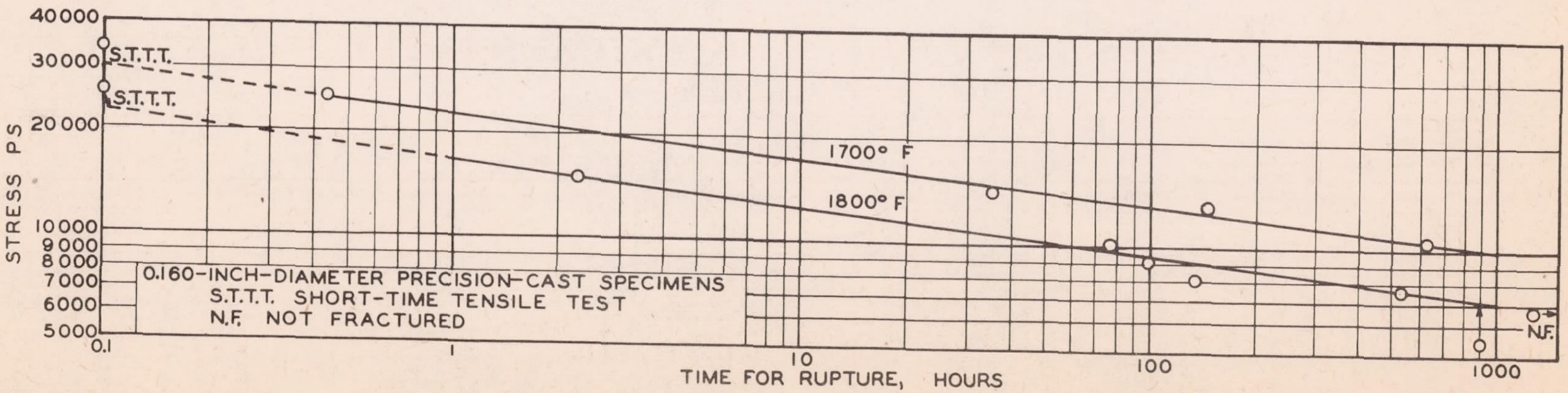

FIGURE 6.- STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR CAST ALLOY VITALLIUM.

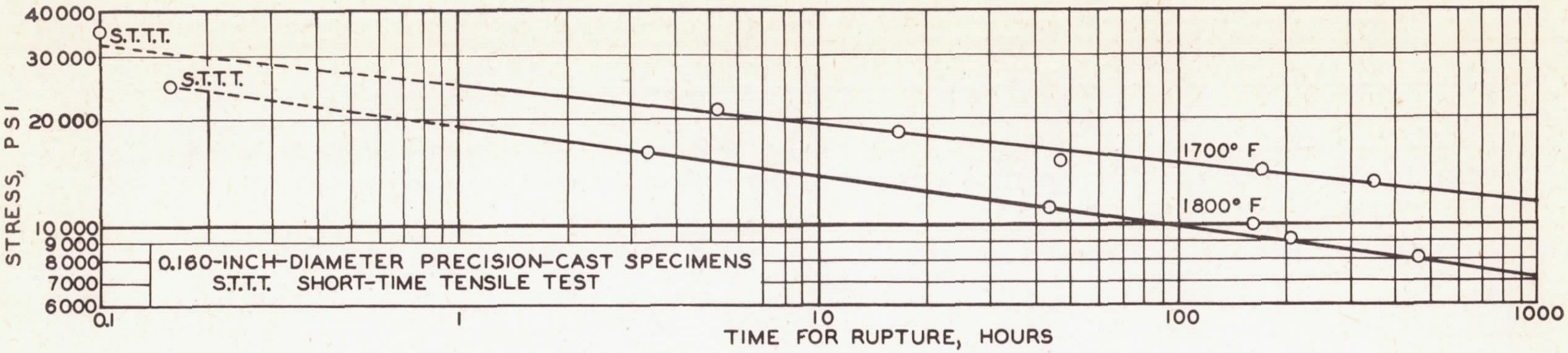

FIGURE 7. - STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR CAST ALLOY 422-19.

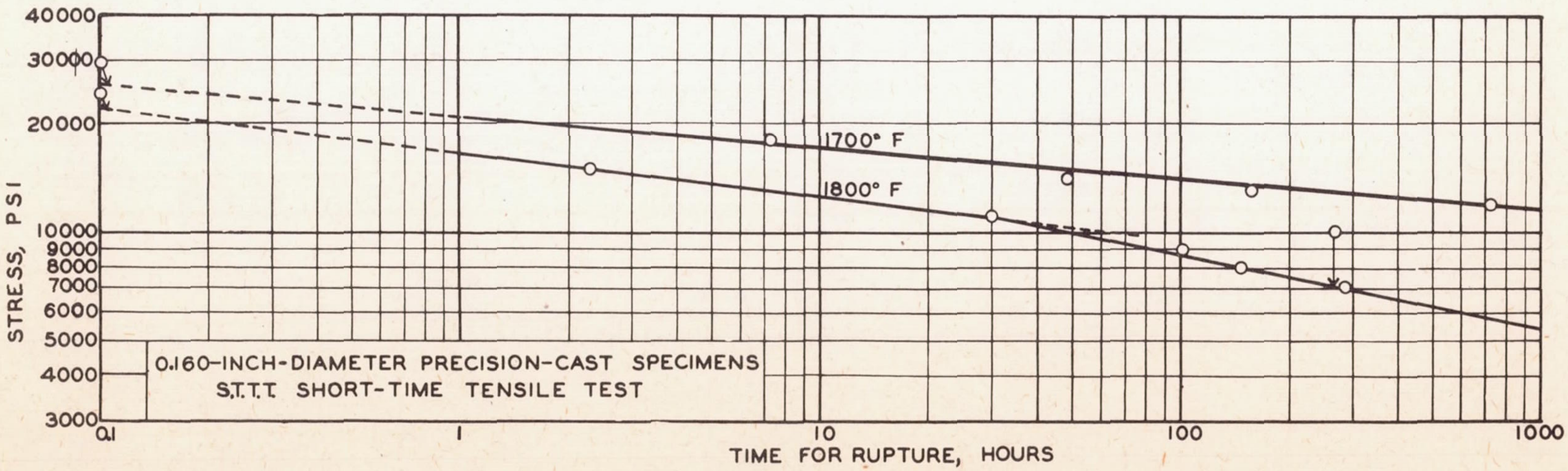

FIGURE 8. - STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR CAST ALLOY 61.

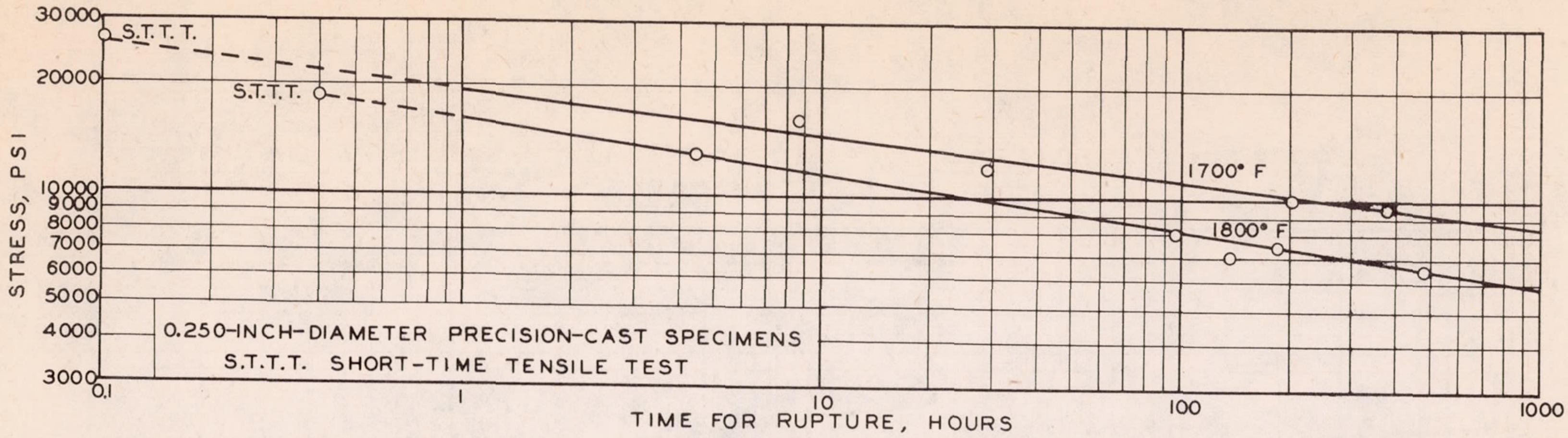

FIGURE 9.- STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR CAST ALLOY S 590.

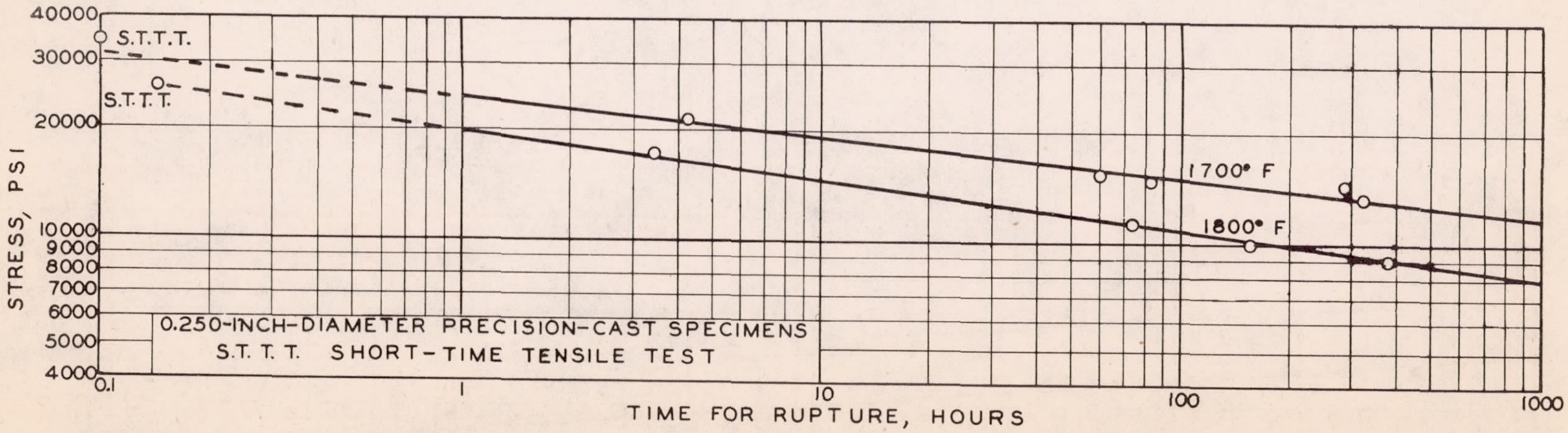

FIGURE 10.- STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR CAST ALLOY S 816.

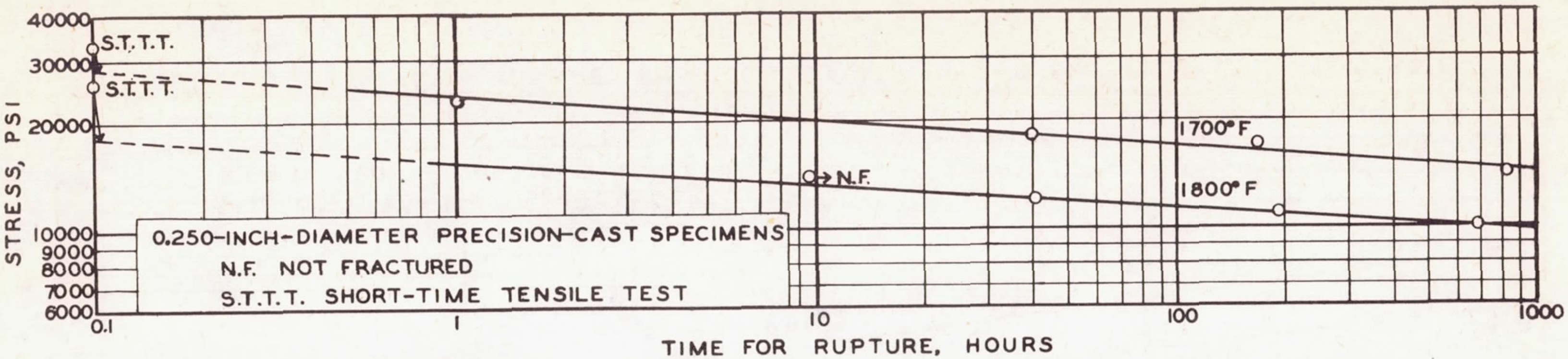

FIGURE 11.- STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR ALLOY X-40.

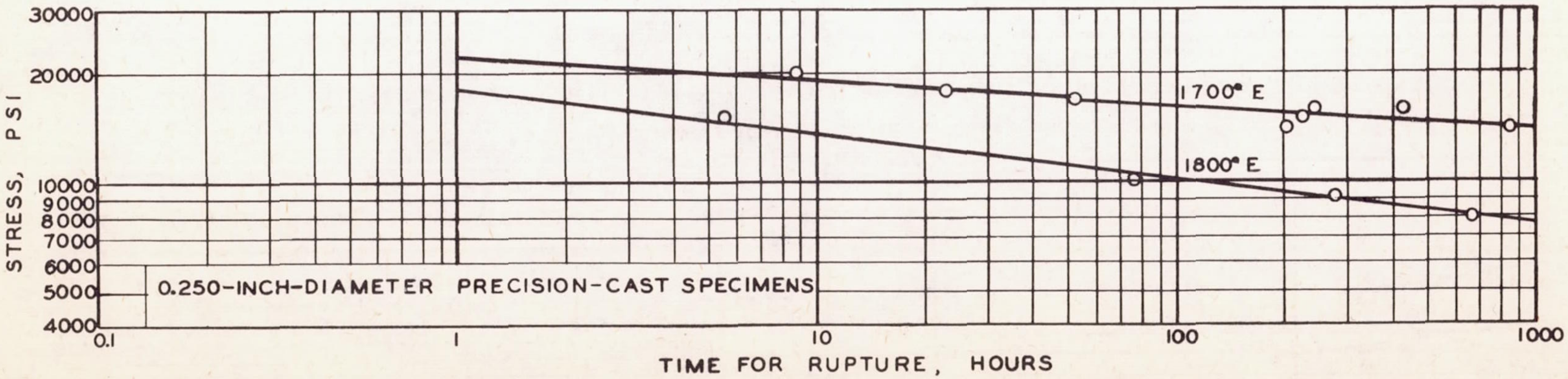

FIGURE 12.- STRESS-RUPTURE TIME CURVES AT 1700° AND 1800° F FOR ALLOY X-50.

	PRECISION CASTINGS								WROUGHT ALLOYS			
ALLOY	X-40	X-50	422-19	S816	61	VITALLIUM	6059	S590	S816	S590	SOL.-TREATED LOW-CARBON N155	HOT-ROLLED LOW-CARBON N155
CHEMICAL COMPOSITION PERCENT — C	0.48	0.76	0.54	0.41	0.43	0.21	0.40	0.57	0.38	0.47	0.13	0.15
Mn	.64	------	.32	.42	.64	.98	.20	.67	.82	1.35	------	1.74
Si	.72	------	.63	.56	.58	.63	.76	.63	.25	.82	------	.37
Cr	25.12	22.57	26.24	19.43	24.49	26.66	24.81	20.11	18.87	19.40	21.5	21.66
Ni	9.69	20.05	10.81	19.80		3.06	33.50	20.64	19.70	19.07	19.	19.40
Co	55.23	40.07	53.89	42.81	69.90	61 –	33.55	20.04	45.64	19.26	19.	20.48
W	7.23	12.17	------	3.42	5.15	------	------	4.50	4.71	4.00	2.1	1.90
Mo	------	------	6.21	3.61	------	5.57	5.34	3.63	4.04	4.03	3.1	2.76
Cb	------	------	------	4.48	------	------	------	4.02	3.43	3.87	1.05	.79
N_2	------	------	------	------	------	------	------	------	------	------	.15	.14
Fe	.55	2.5 –	1.10	5 –	1.00	1.76	1.00	25 –	2.94	27 –	32 –	30 –
HEAT TREATMENT	AS CAST	AS CAST	AS CAST	AS CAST	AS CAST	AS CAST	AS CAST	AS CAST	W.Q. 2300° F 16 HR AGING AT TEST TEMP.	W.Q. 2270° F 16 HR AGING AT TEST TEMP.	1/2 HR AT 2280° F AIR-COOLED	AS HOT-ROLLED

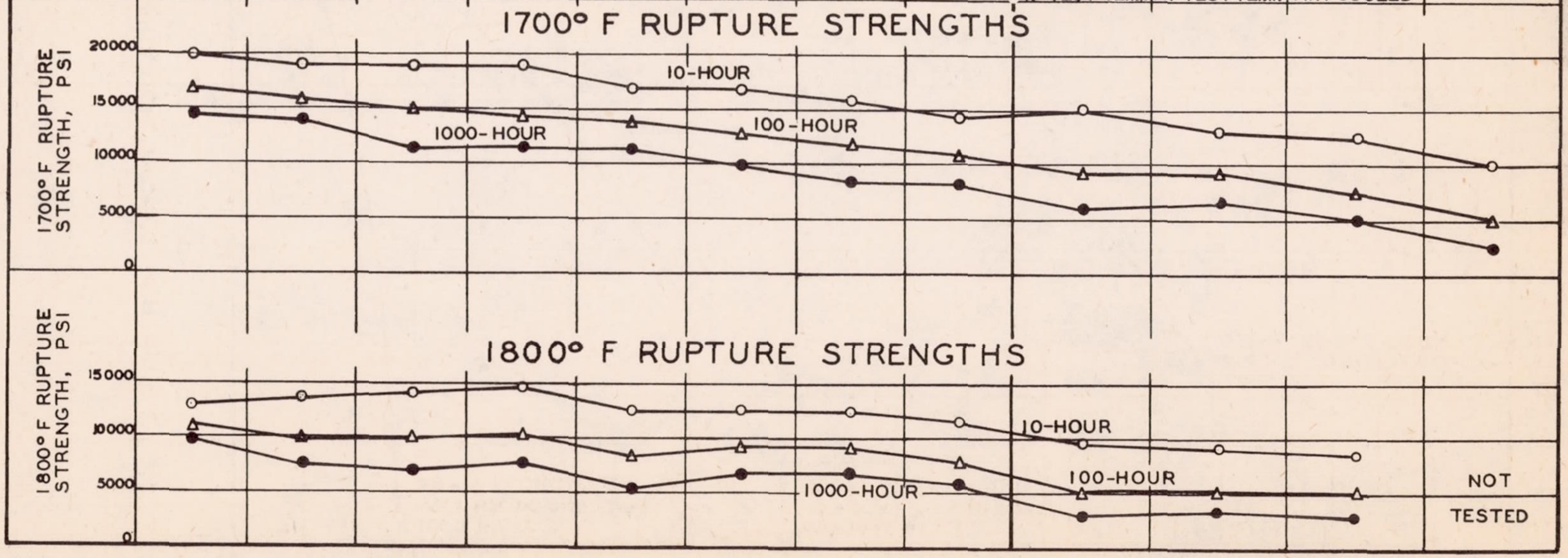

FIGURE 13.- COMPARATIVE RUPTURE STRENGTHS AT 1700° AND 1800° F FOR INDICATED ALLOYS.

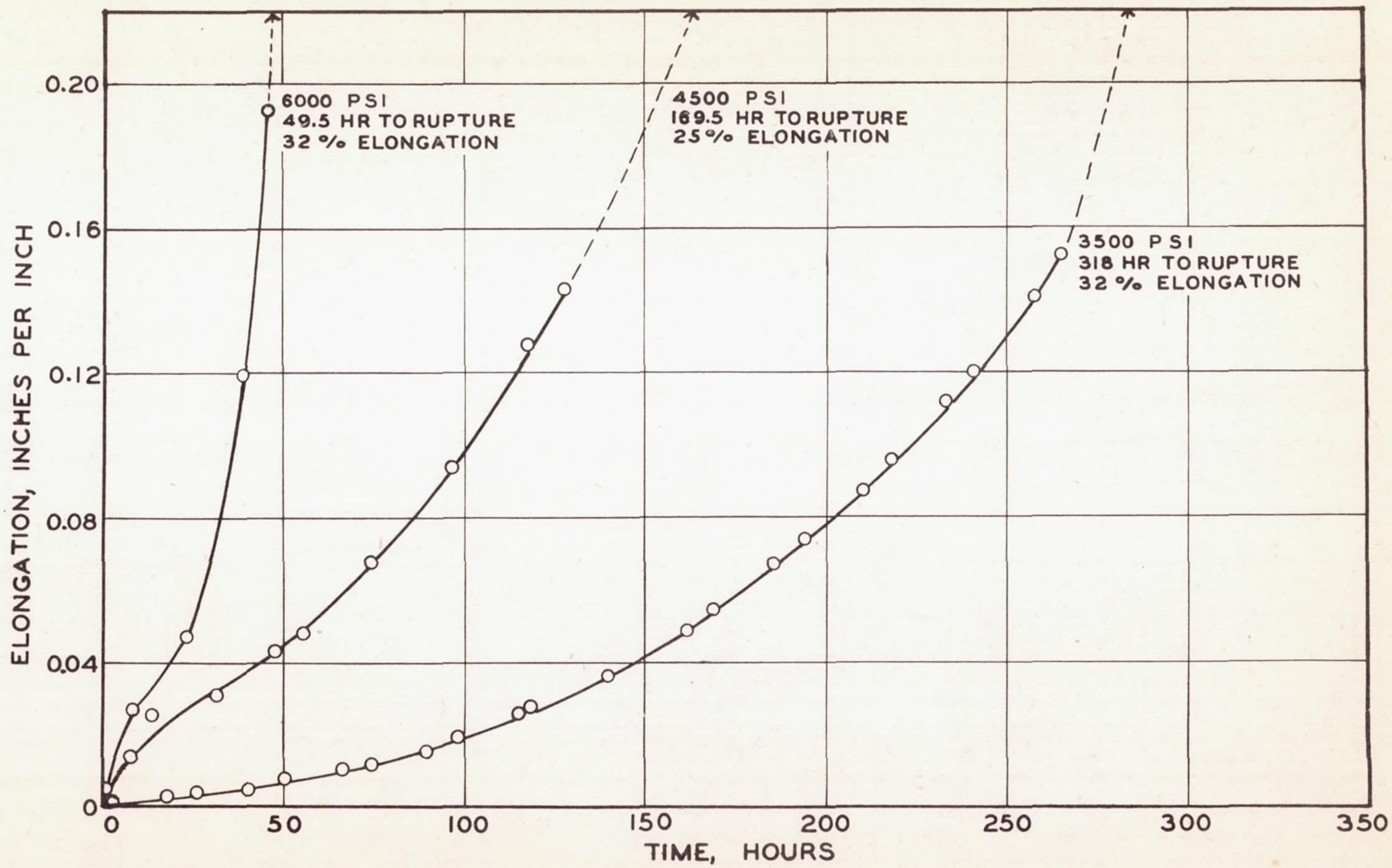

FIGURE 14.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON HOT-ROLLED LOW-CARBON N155 ALLOY.

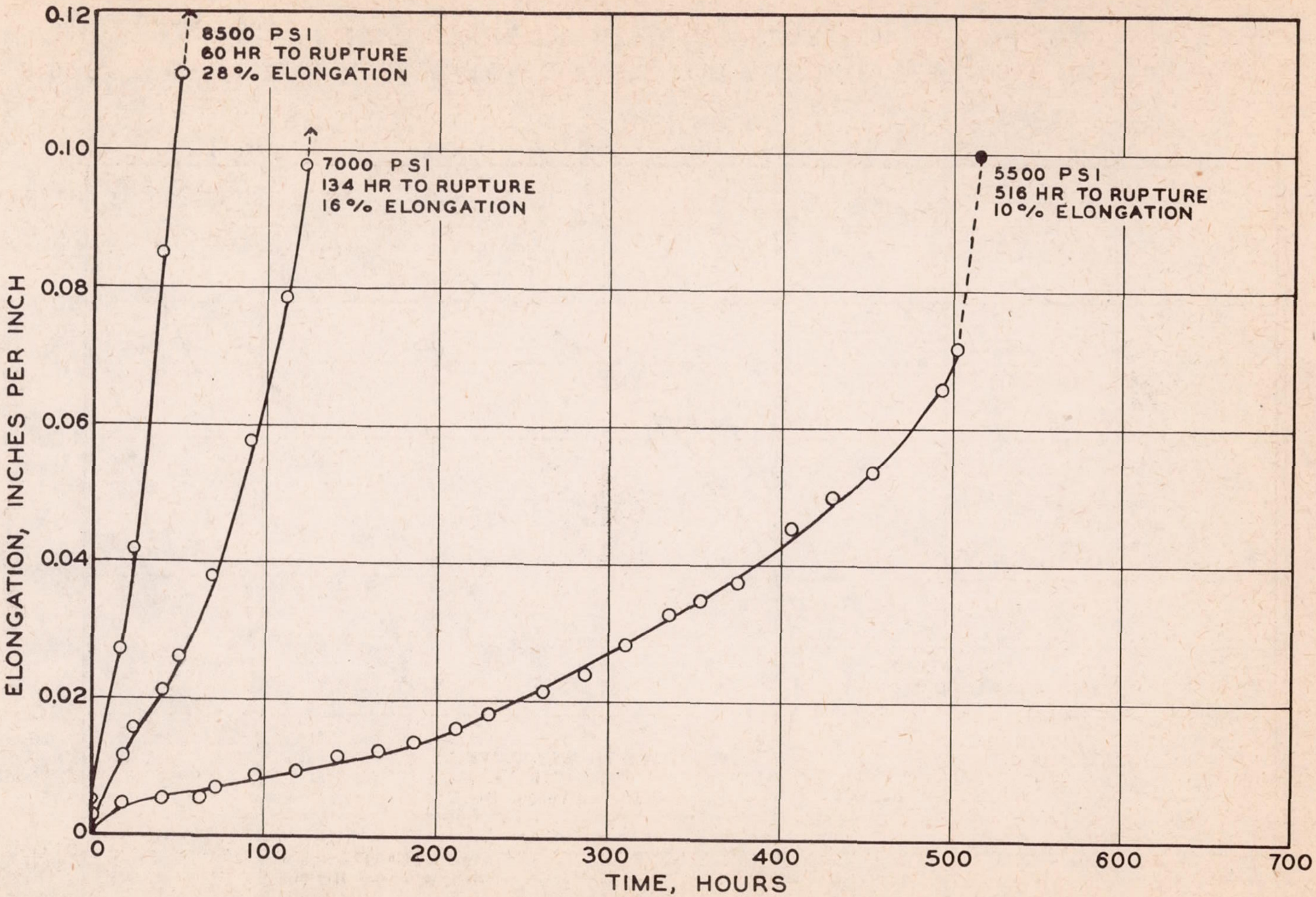

FIGURE 15.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON SOLUTION-TREATED LOW-CARBON N155 ALLOY.

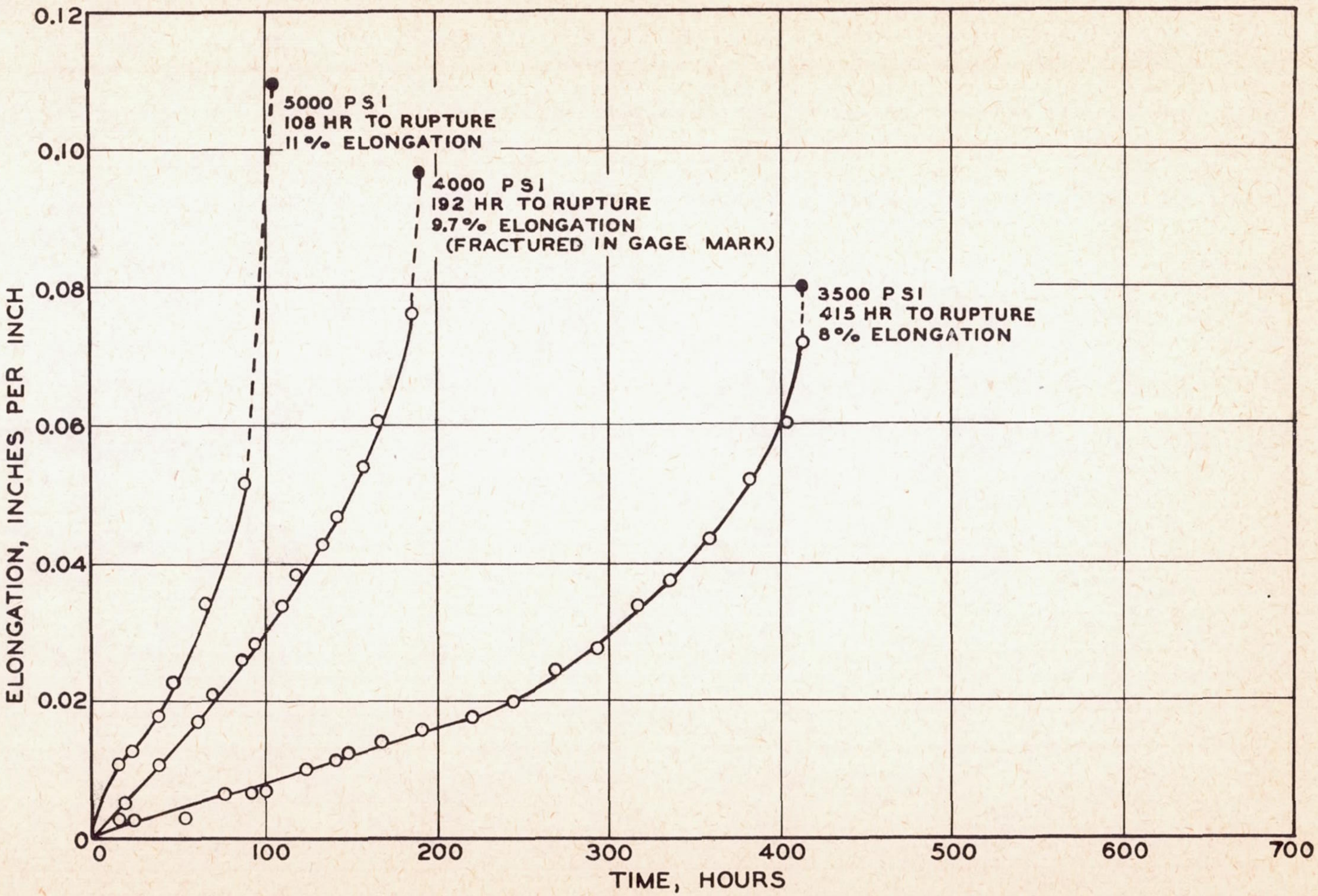

FIGURE 16.- TIME-ELONGATON CURVES FOR RUPTURE TESTS AT 1800° F ON SOLUTION-TREATED LOW-CARBON N155 ALLOY.

W-75

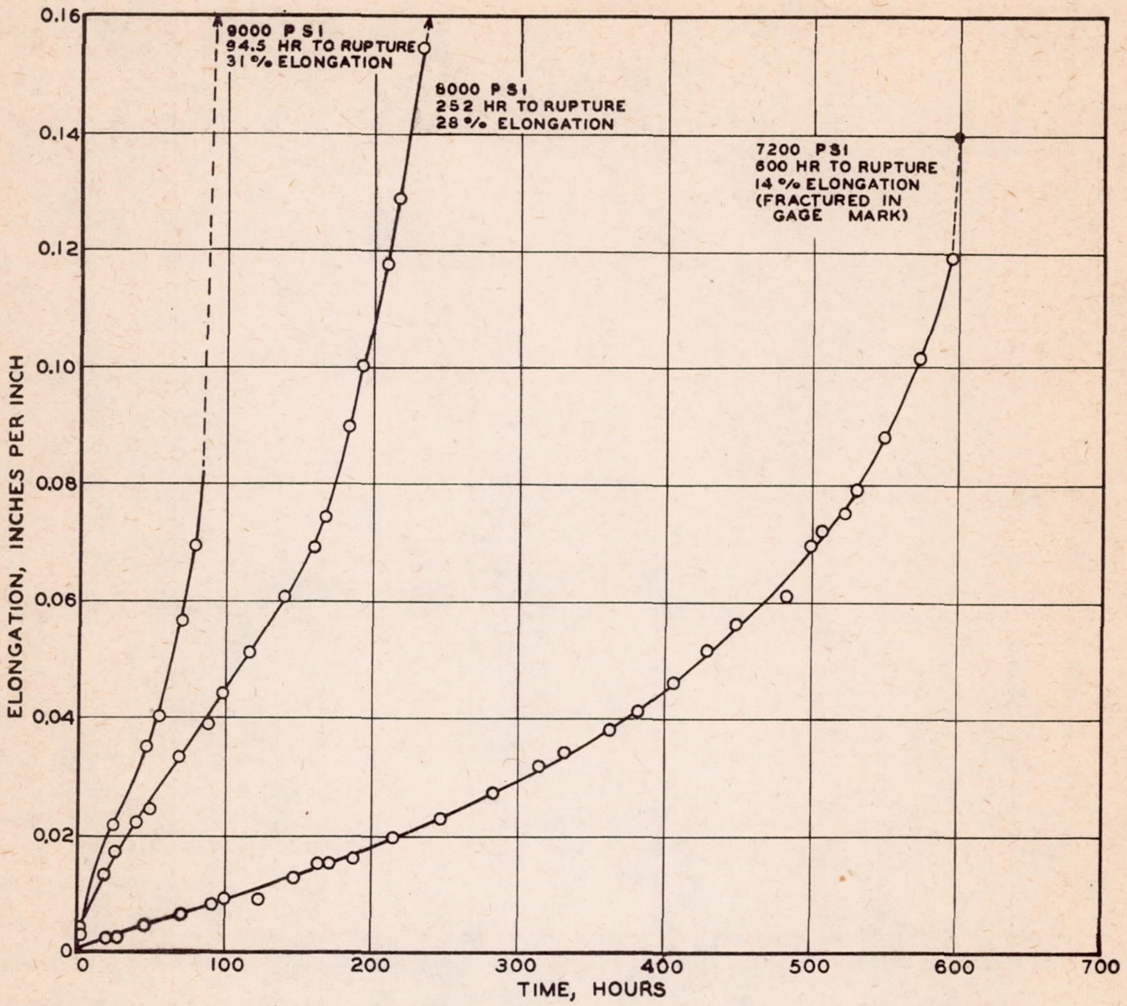

FIGURE 17.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON WROUGHT S590 ALLOY.

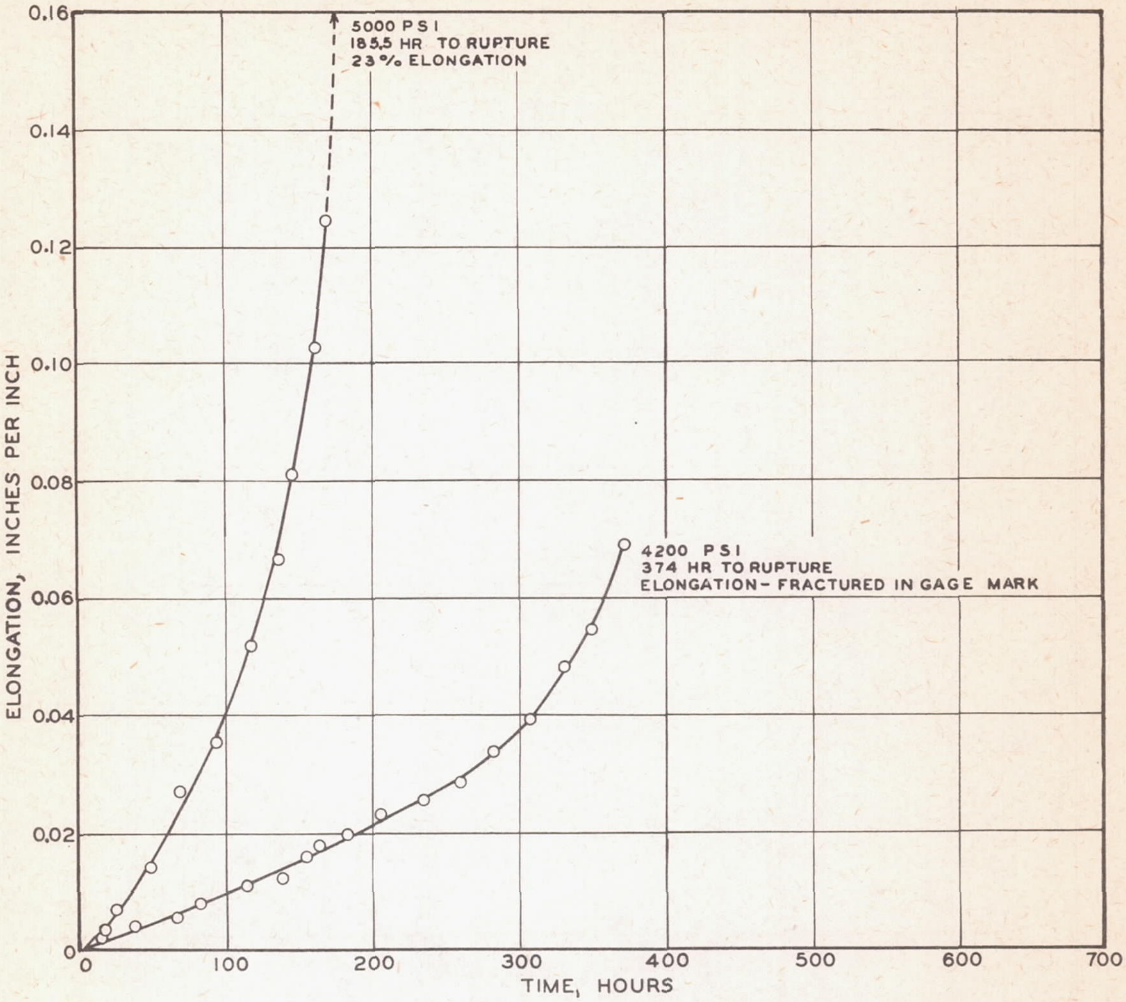

FIGURE 18.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON WROUGHT S590 ALLOY.

W-75

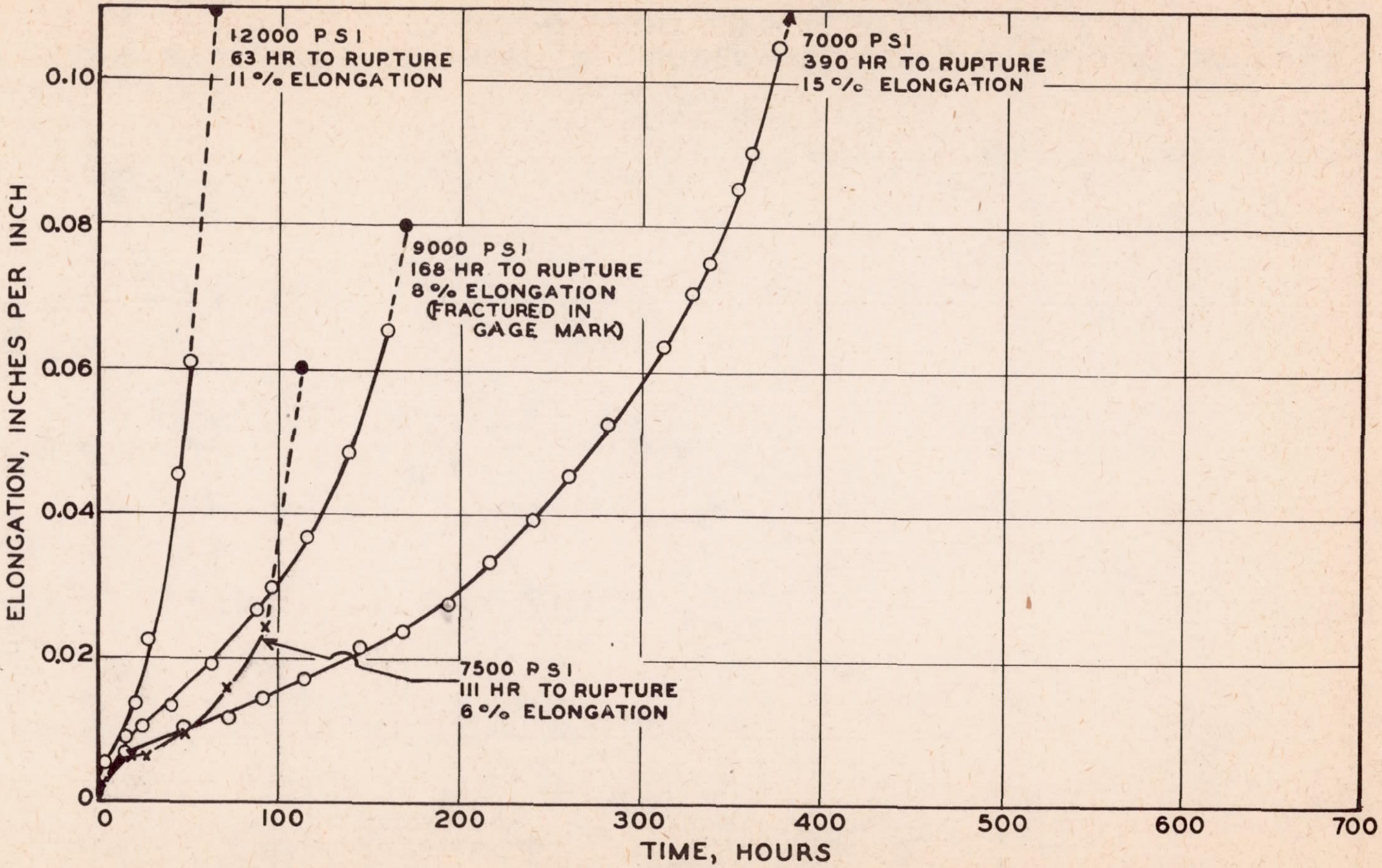

FIGURE 19.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON WROUGHT S816 ALLOY.

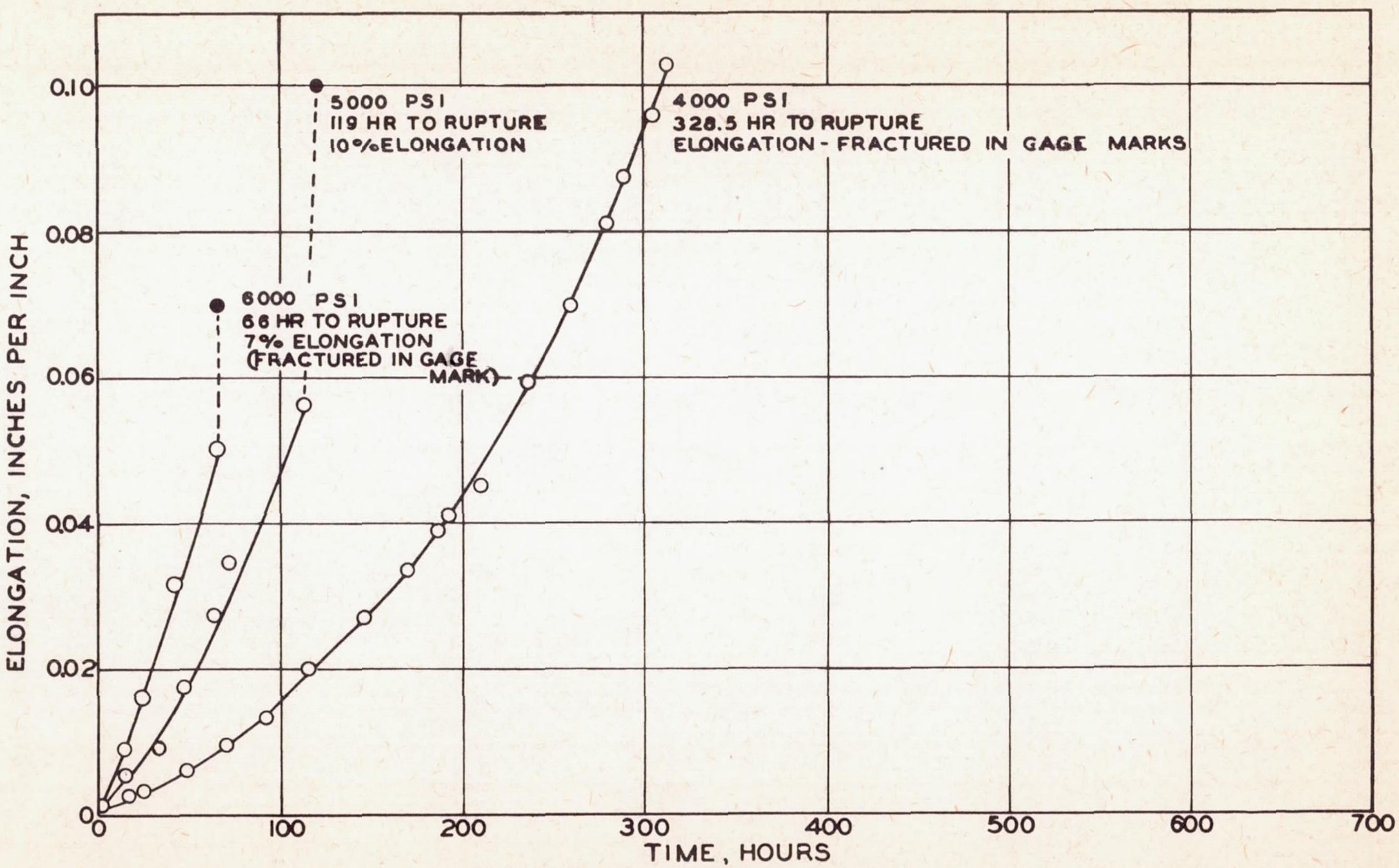

FIGURE 20.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON WROUGHT S816 ALLOY.

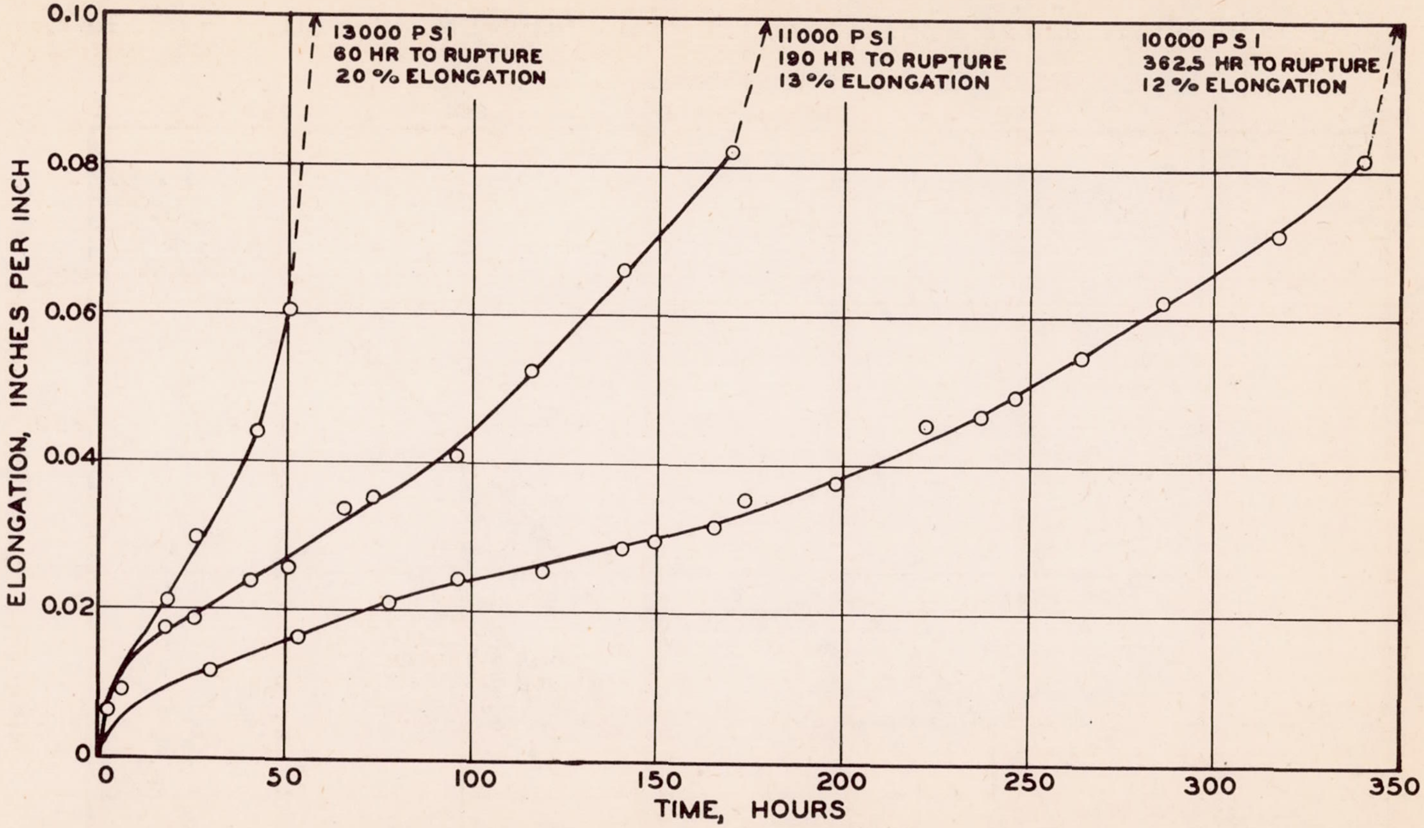

FIGURE 21.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON PRECISION-CAST 6059 ALLOY.

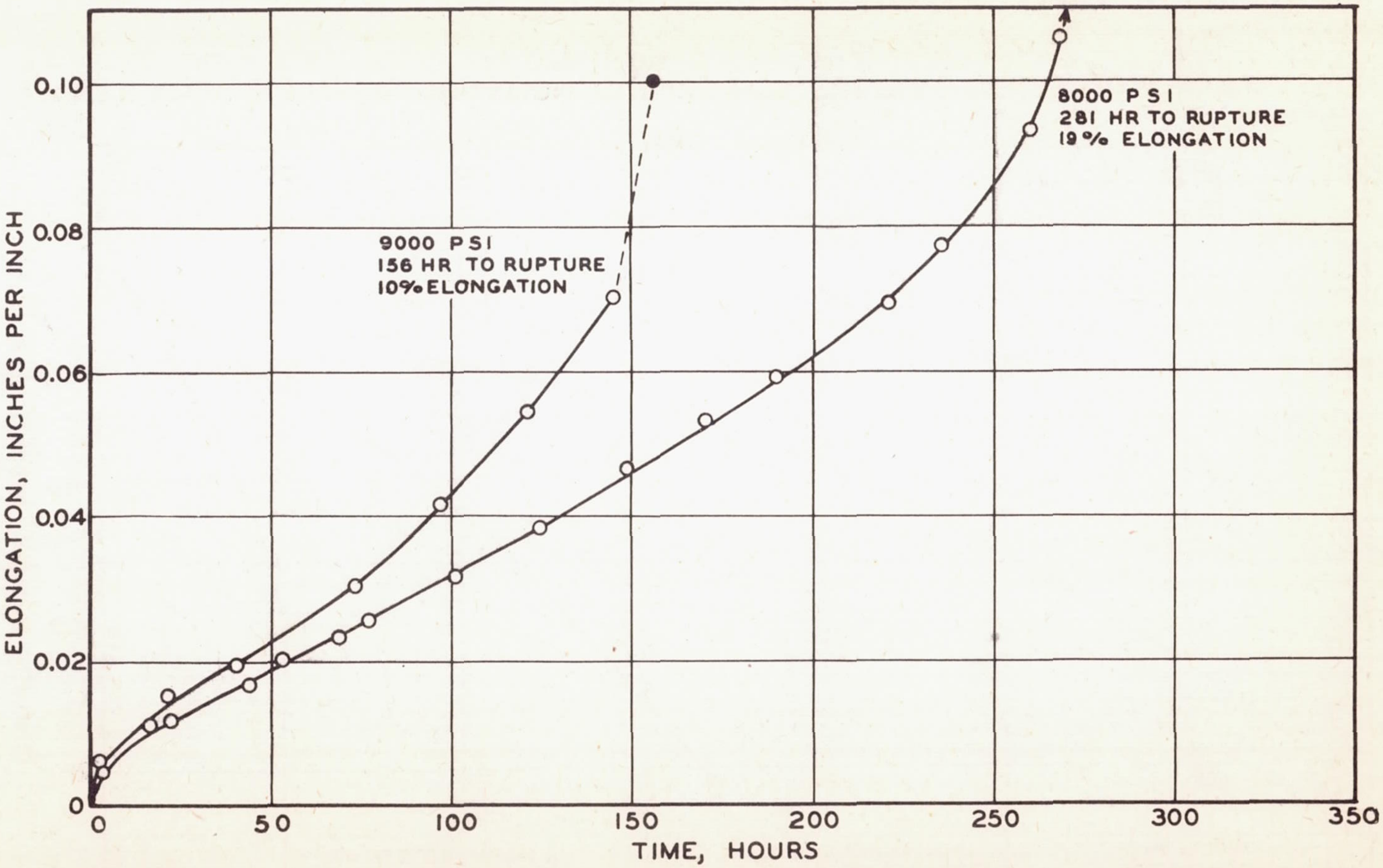

FIGURE 22.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON PRECISION-CAST 6059 ALLOY.

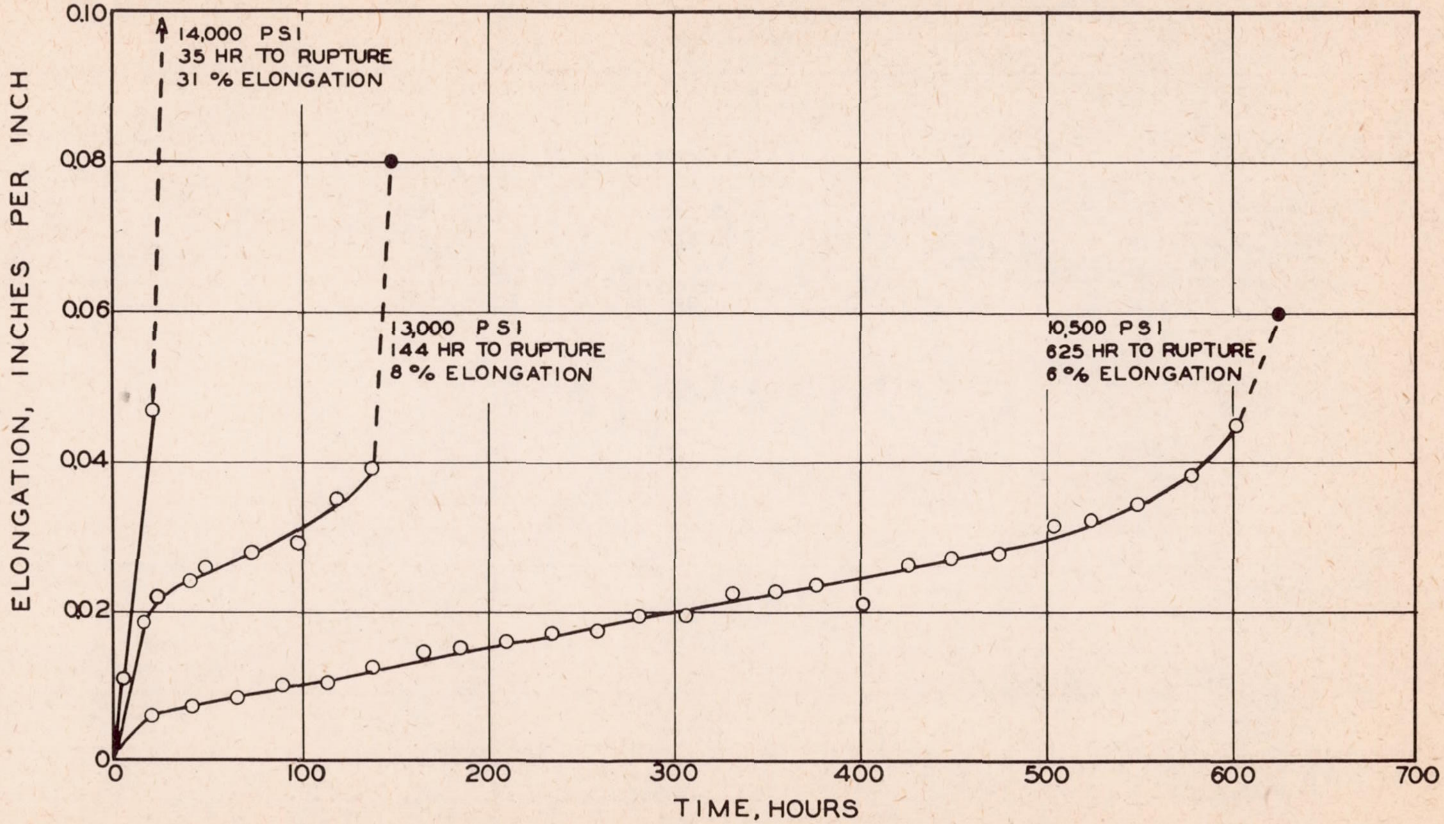

FIGURE 23.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON PRECISION-CAST VITALLIUM ALLOY.

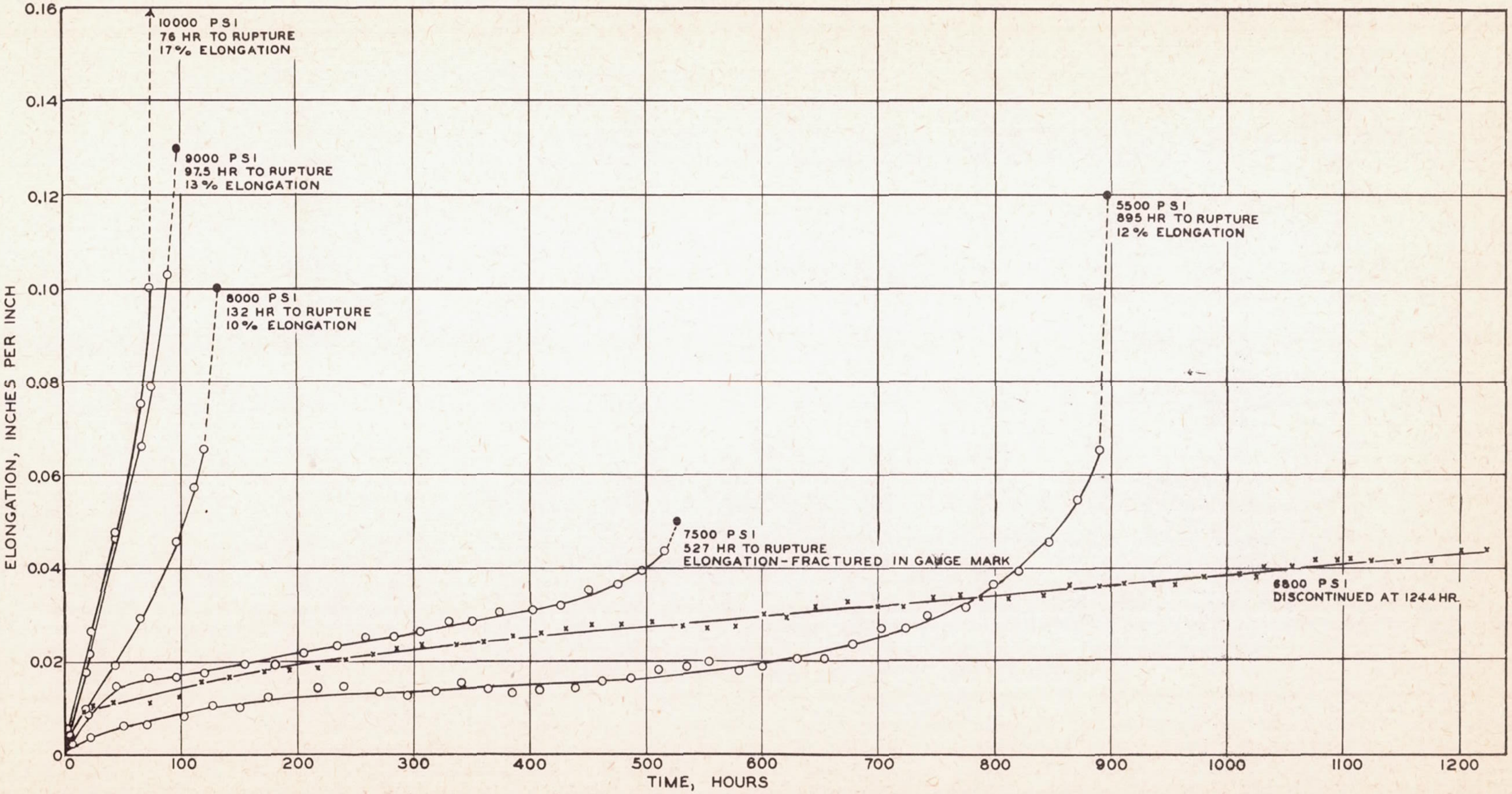

FIGURE 24.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON PRECISION-CAST VITALLIUM ALLOY.

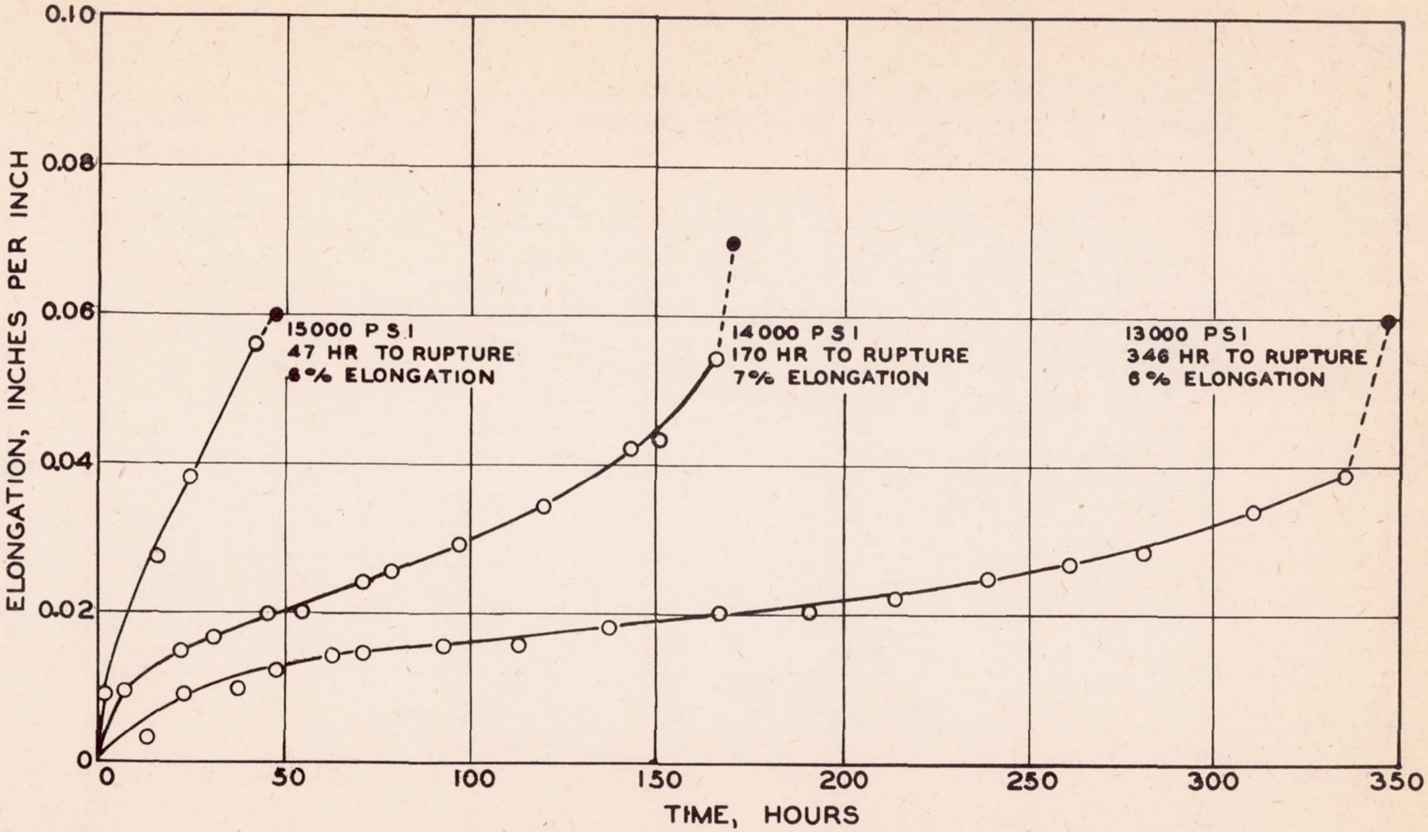

FIGURE 25.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON PRECISION-CAST 422-19 ALLOY.

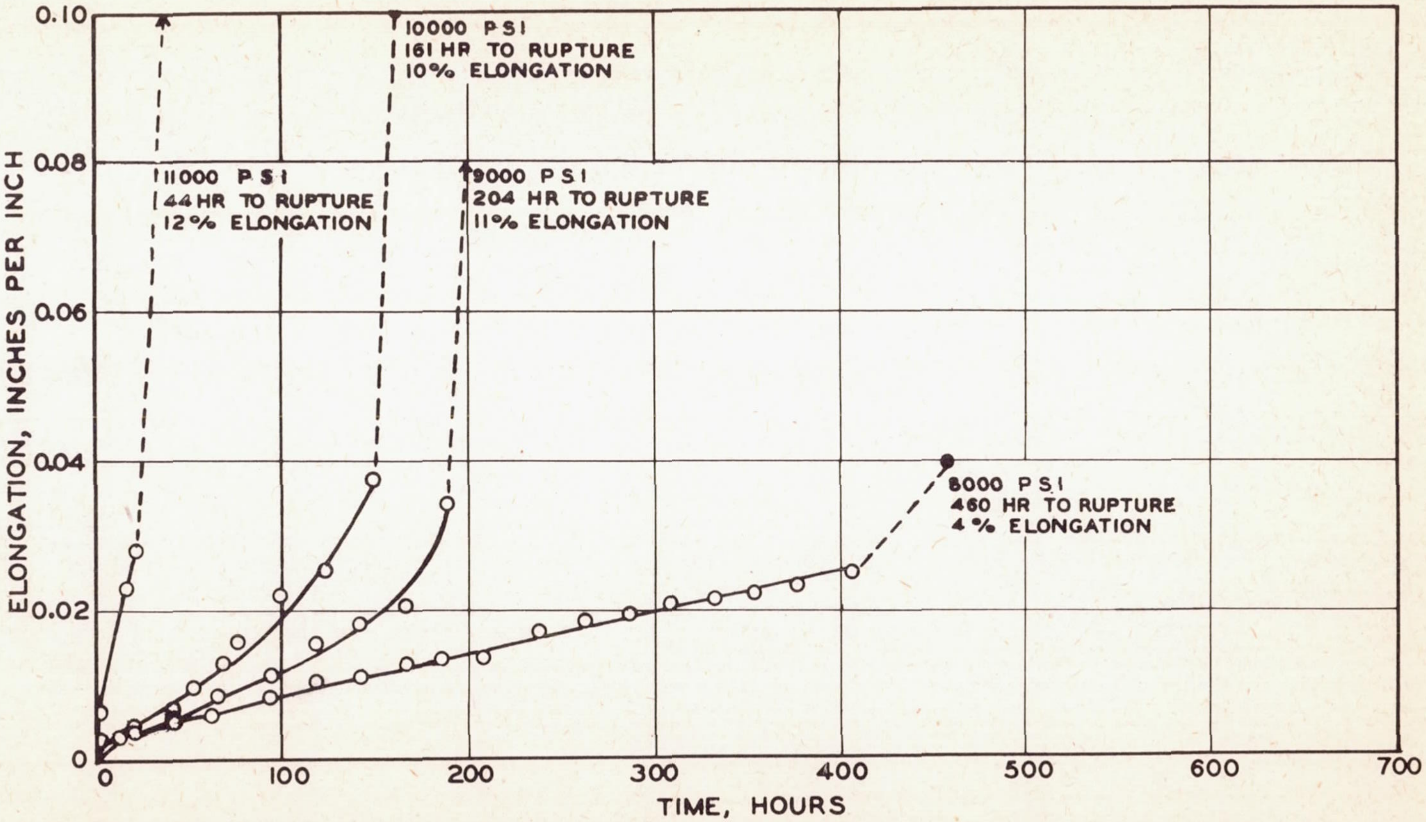

FIGURE 26.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON PRECISION-CAST 422-19 ALLOY.

W-75

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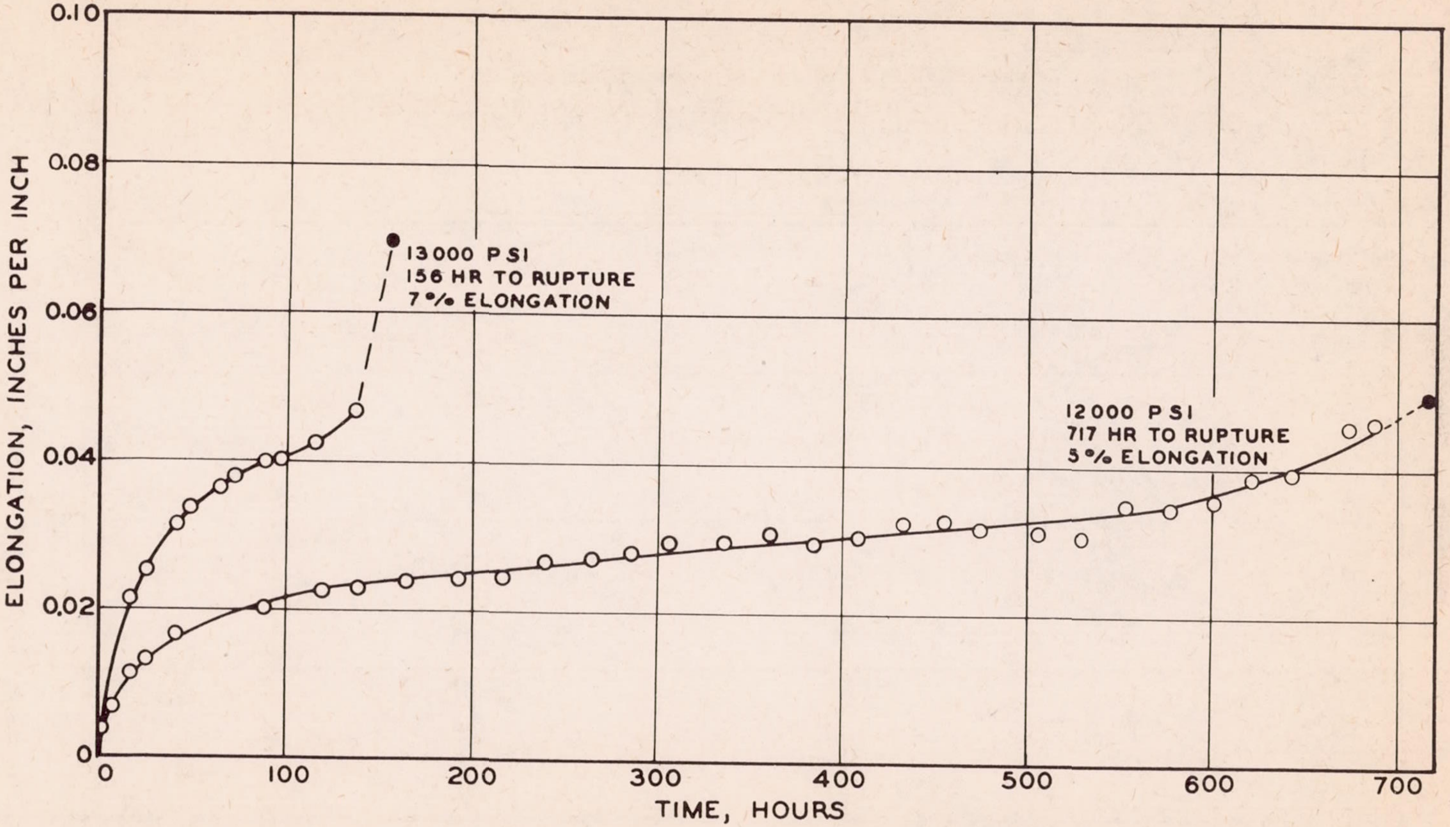


FIGURE 27.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON PRECISION-CAST 61 ALLOY.

FIG. 27

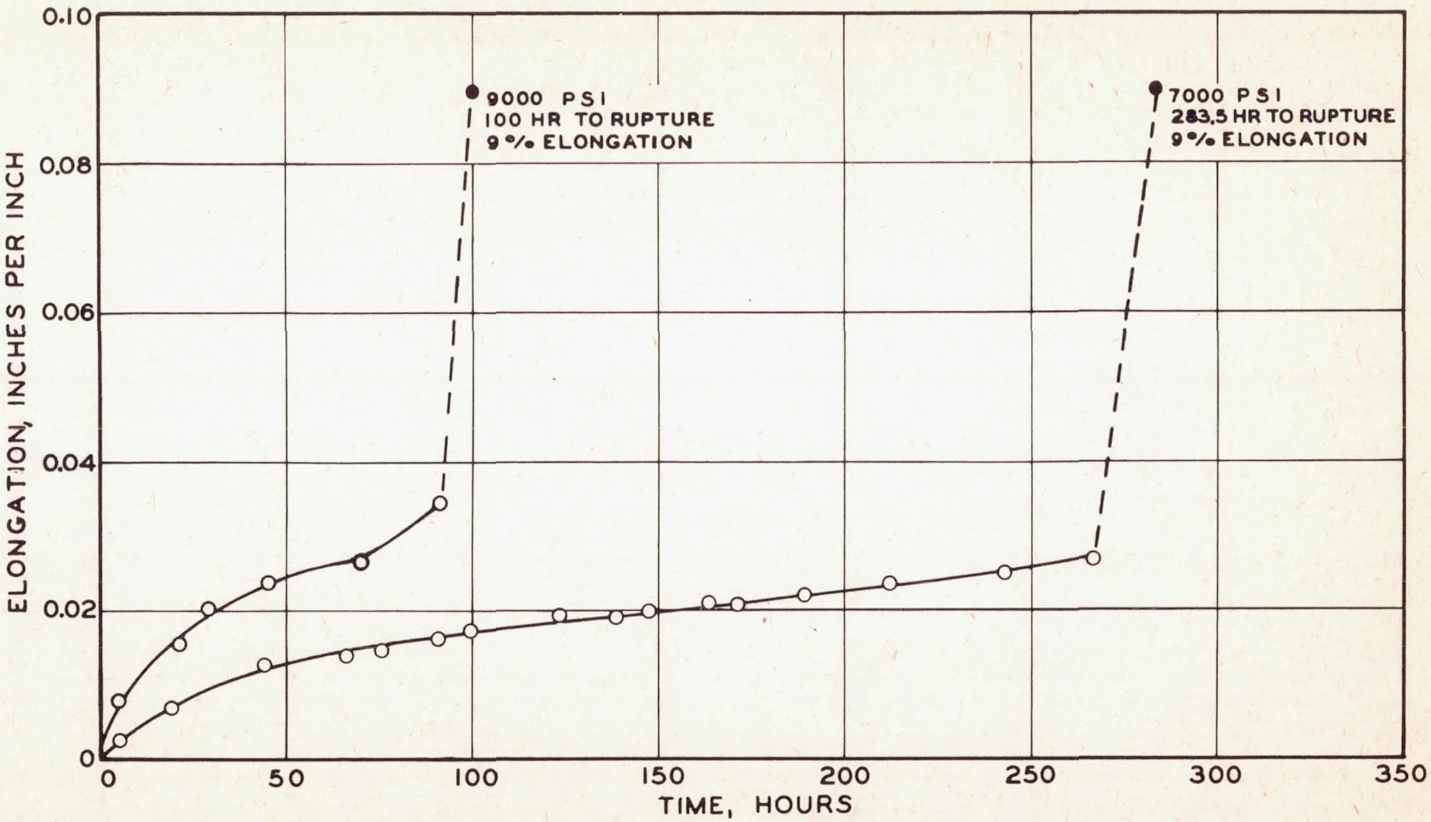


FIGURE 28.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON PRECISION-CAST 61 ALLOY.

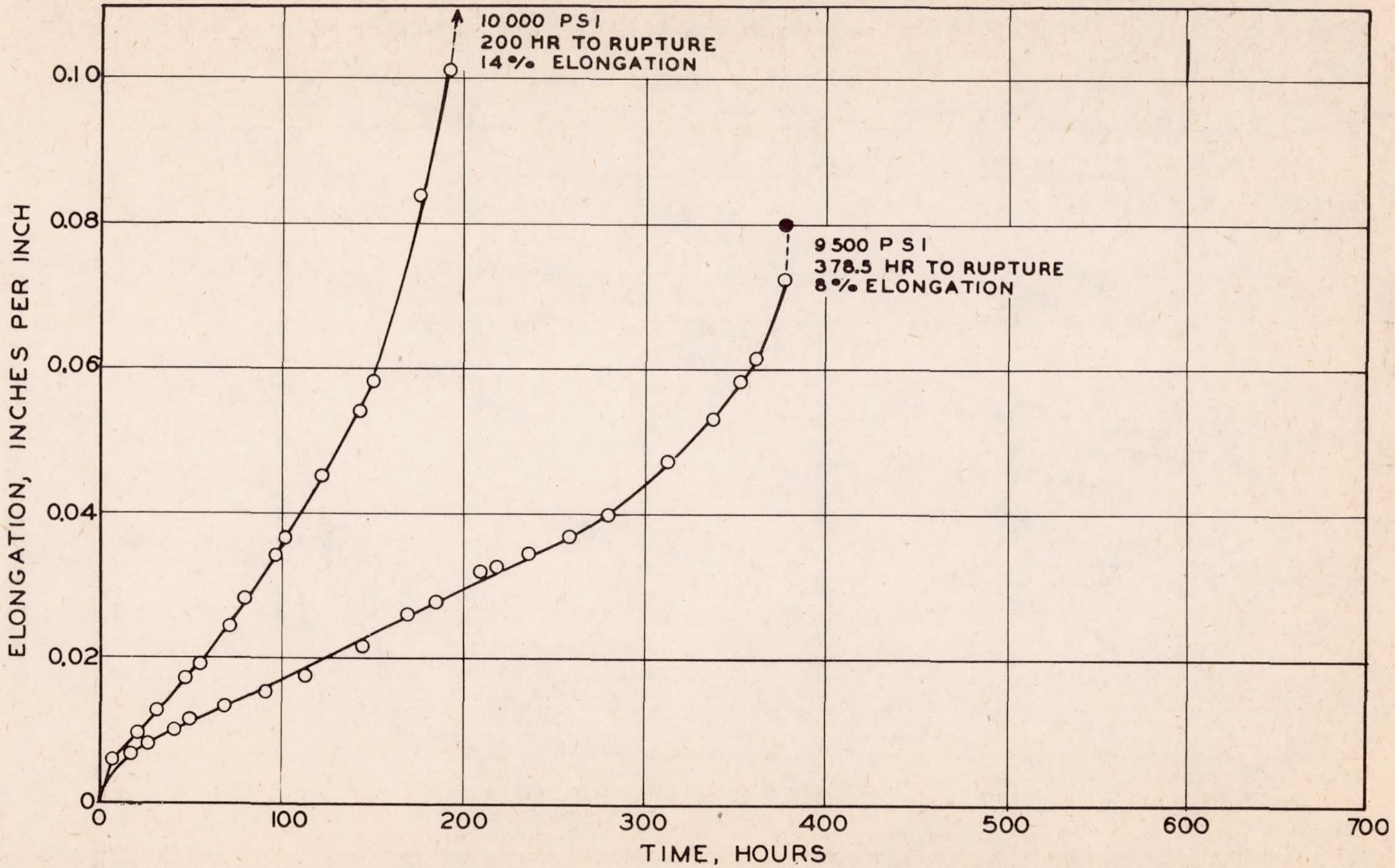


FIGURE 29.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON PRECISION-CAST S590 ALLOY.

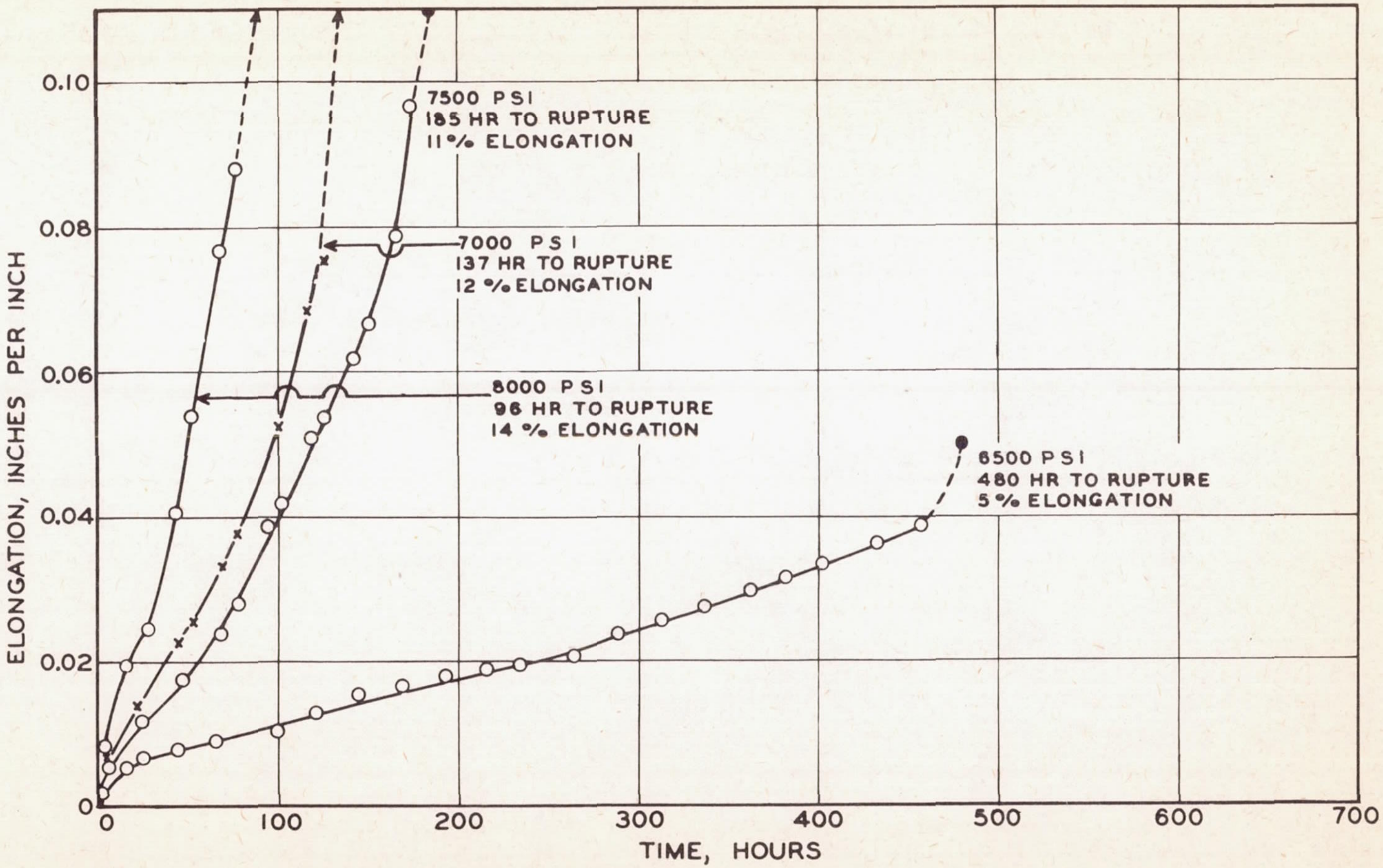


FIGURE 30.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON PRECISION-CAST S590 ALLOY.

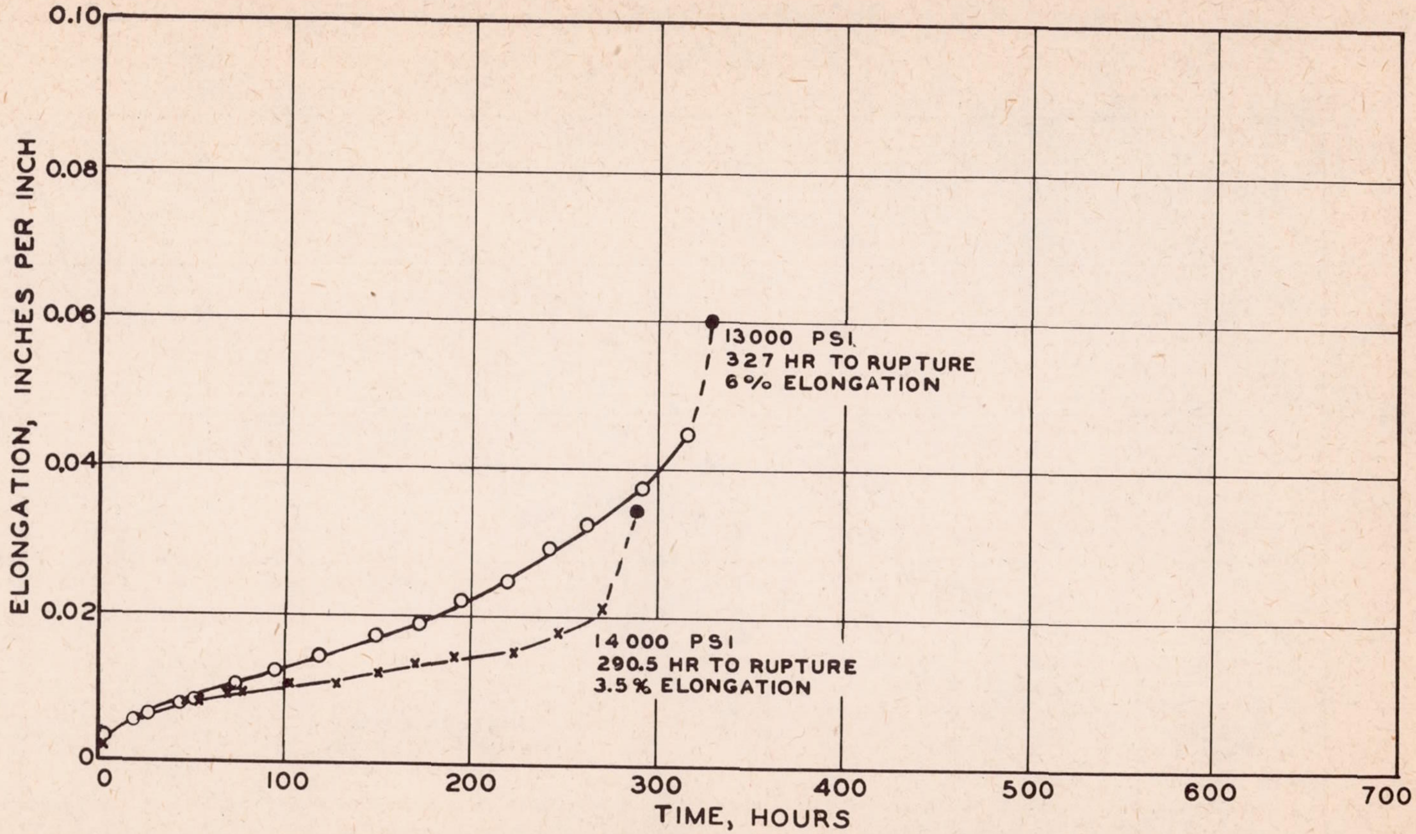


FIGURE 31.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON PRECISION-CAST S816 ALLOY.

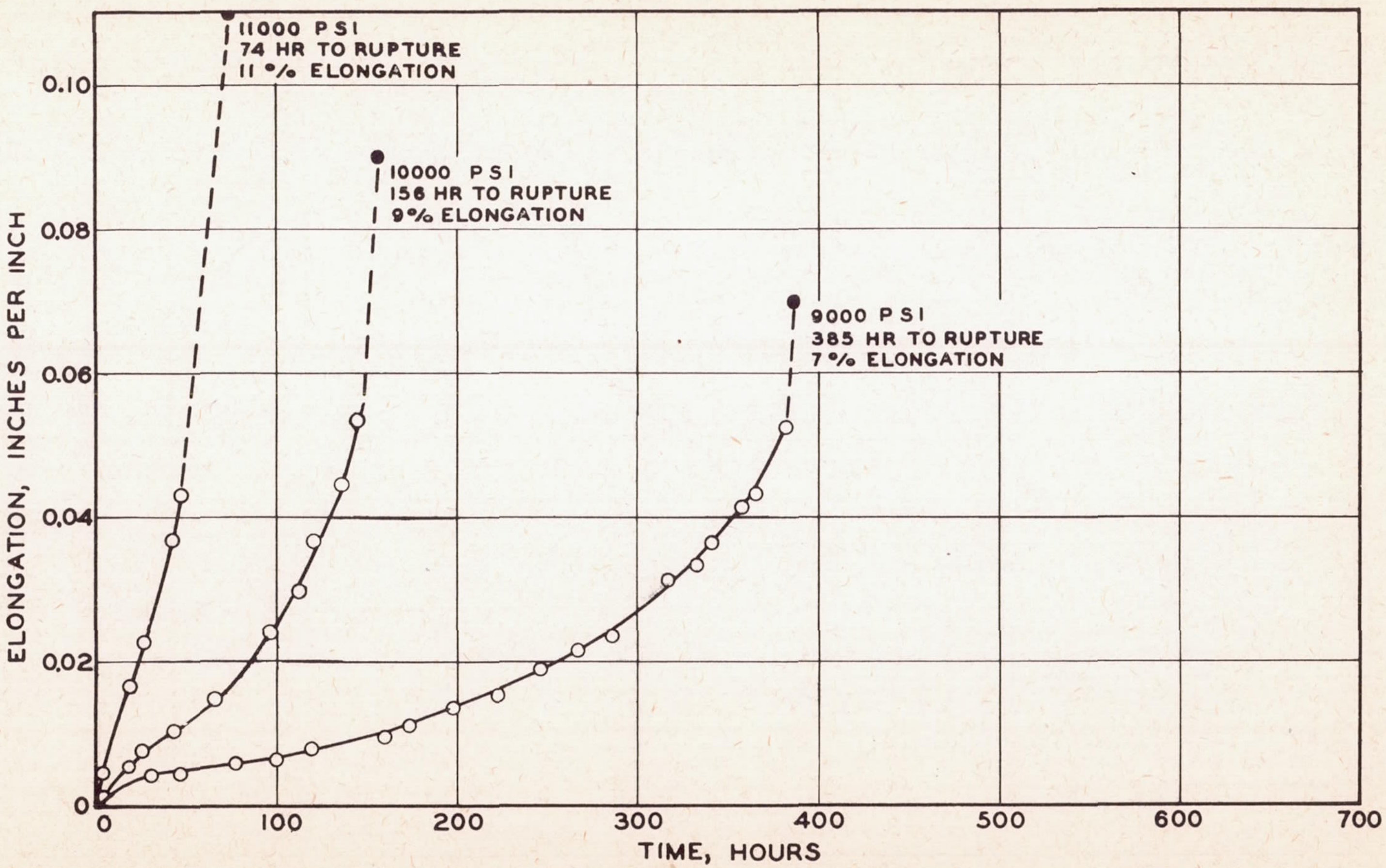


FIGURE 32.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON PRECISION-CAST S816 ALLOY.

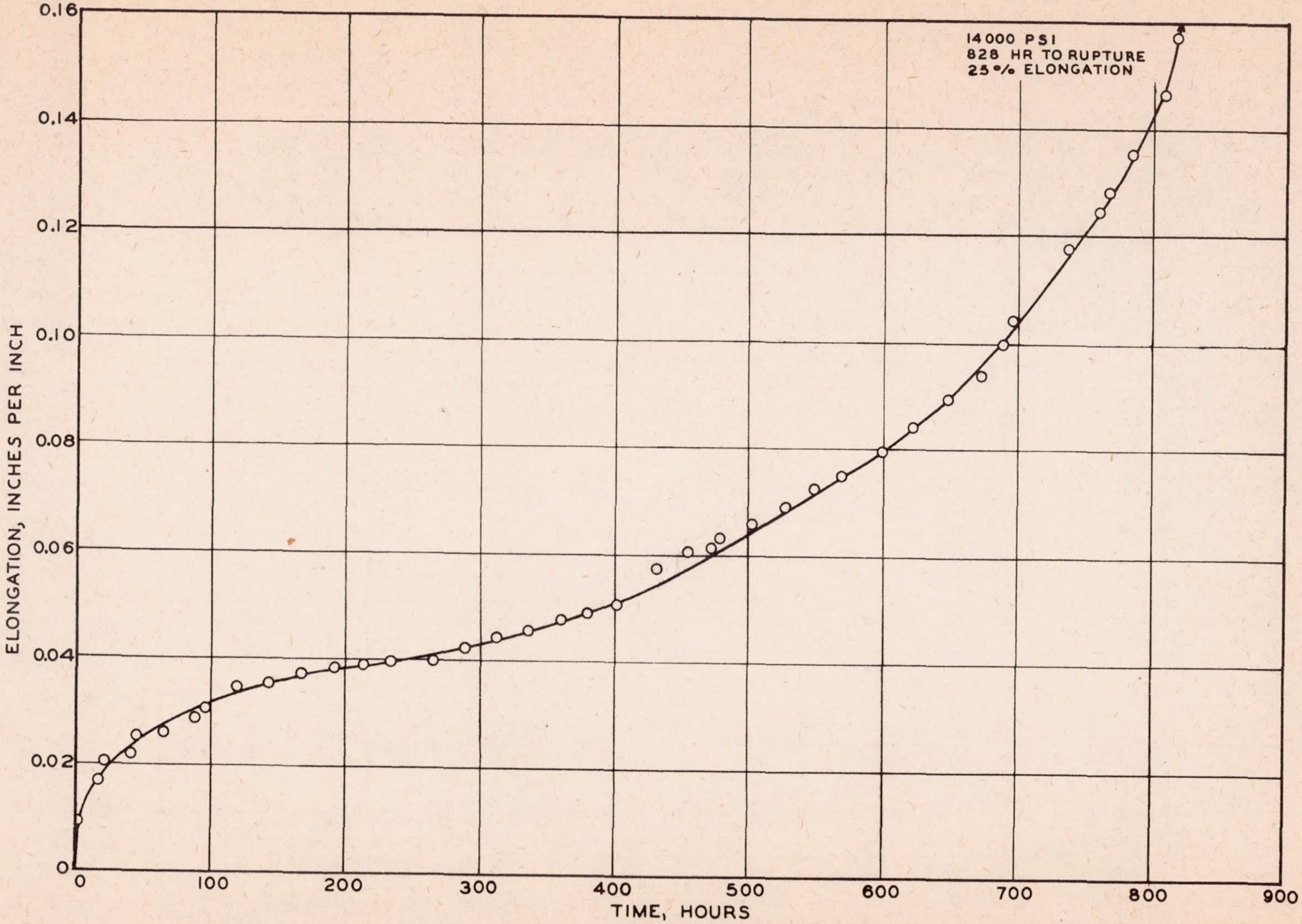


FIGURE 33.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON PRECISION-CAST X-40 ALLOY.

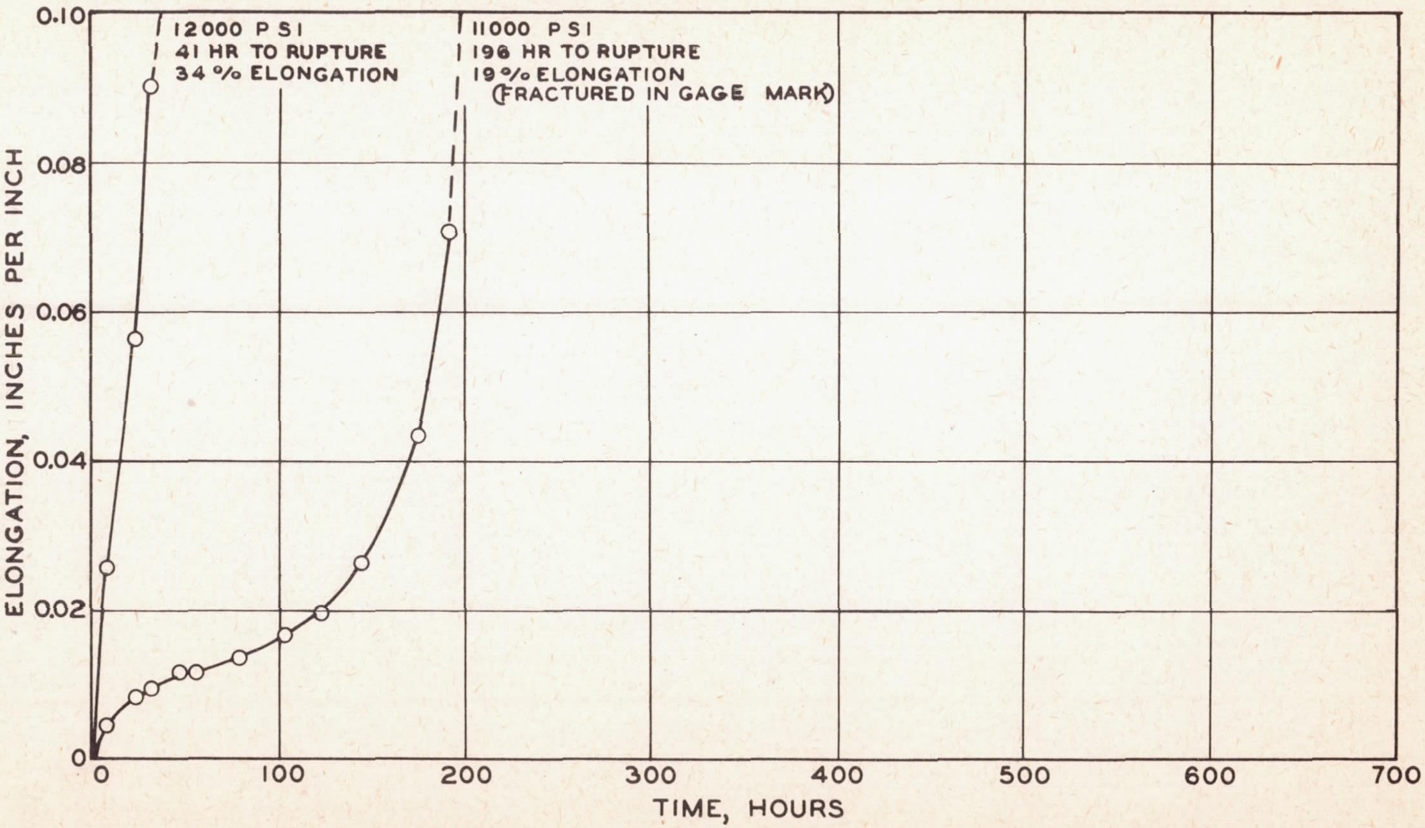


FIGURE 34.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON PRECISION-CAST X-40 ALLOY.

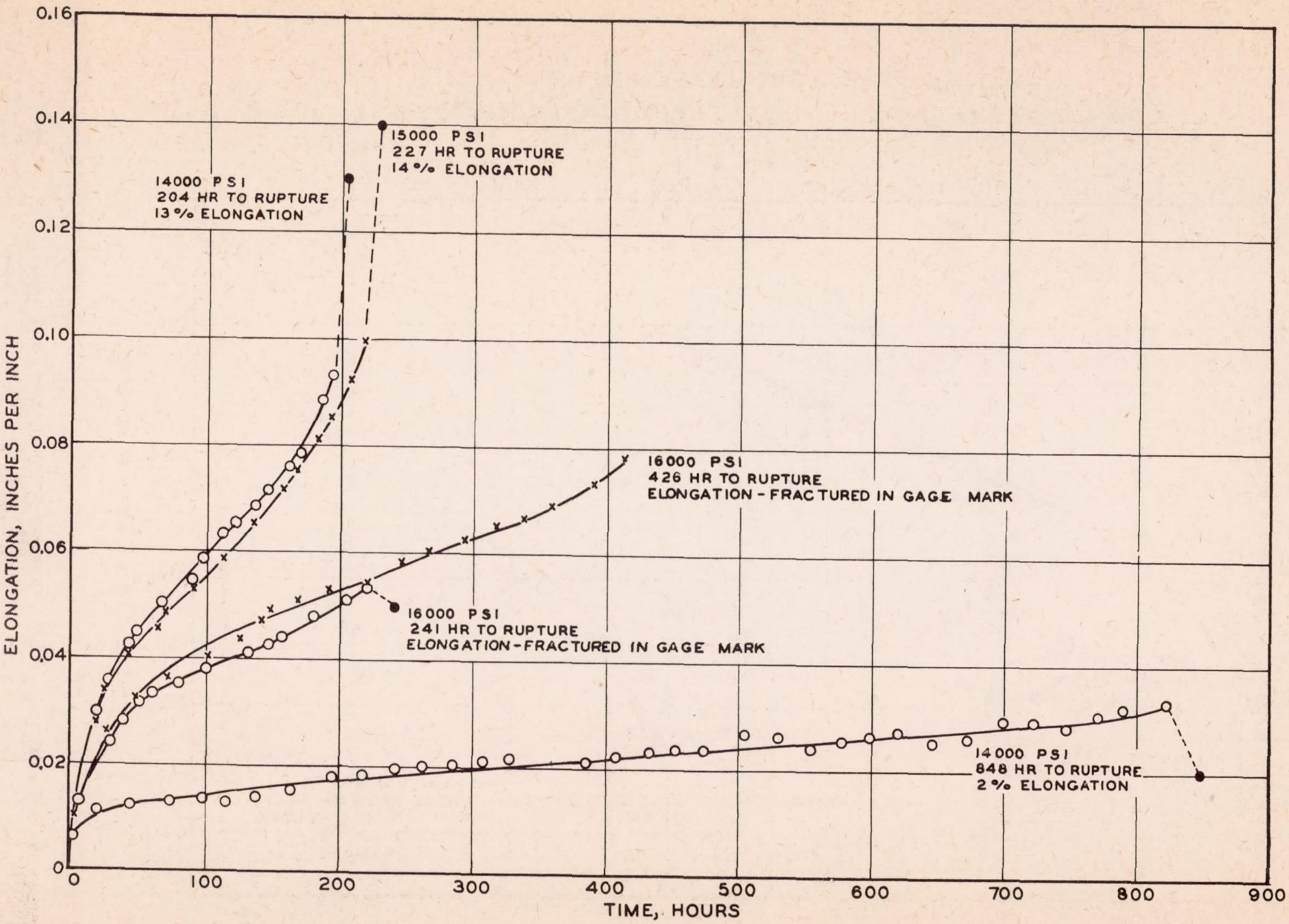


FIGURE 35.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1700° F ON PRECISION-CAST X-50 ALLOY.

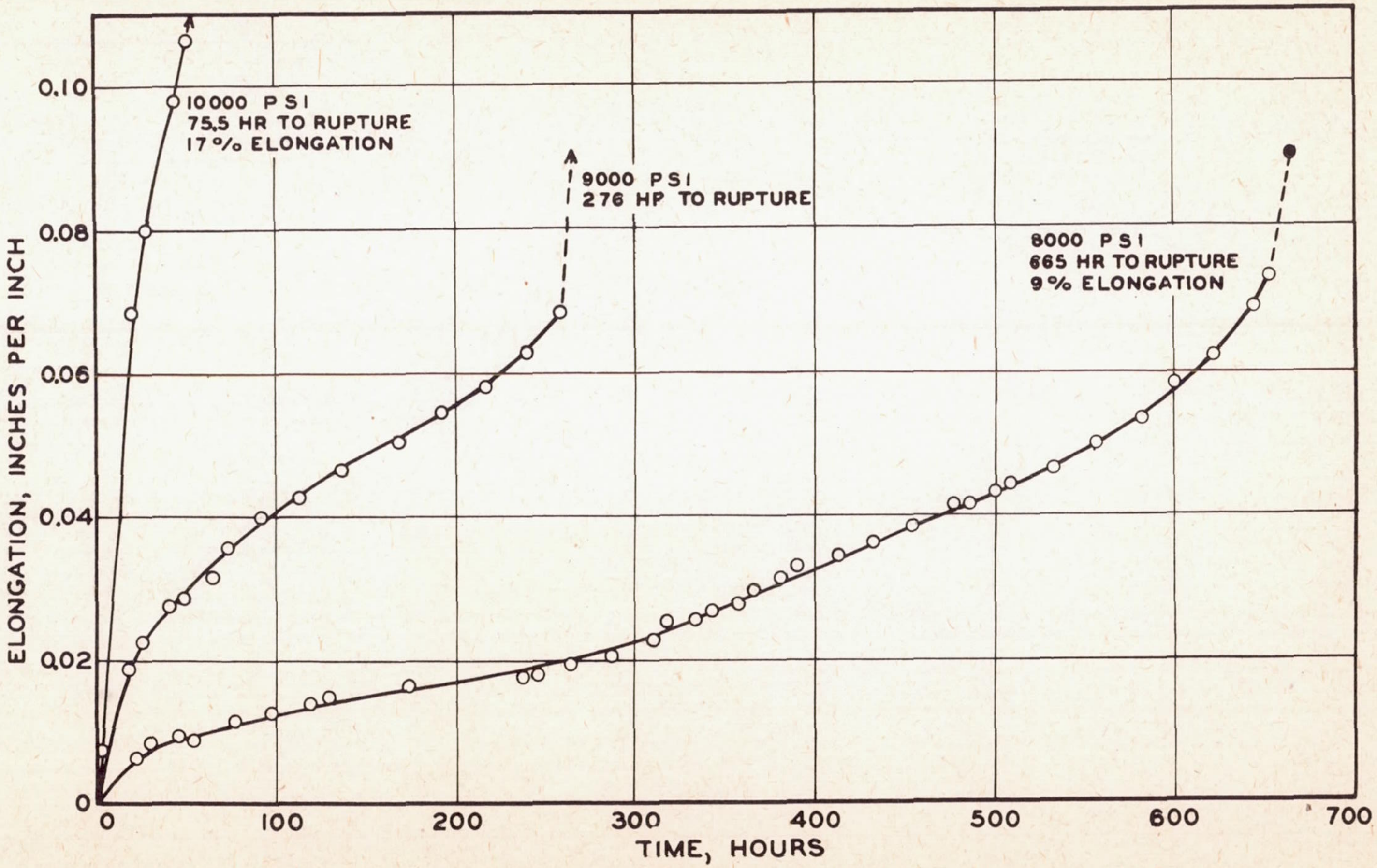
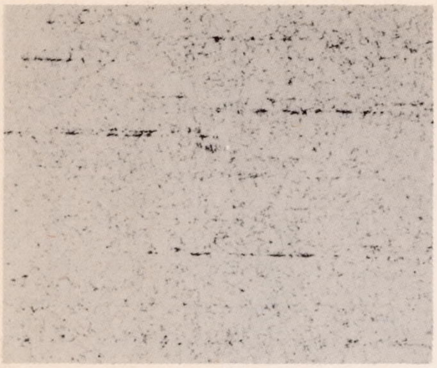


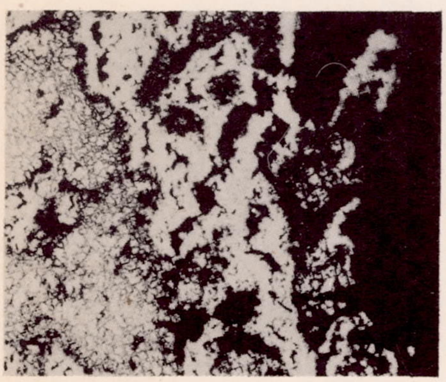
FIGURE 36.- TIME-ELONGATION CURVES FOR RUPTURE TESTS AT 1800° F ON PRECISION-CAST X-50 ALLOY.



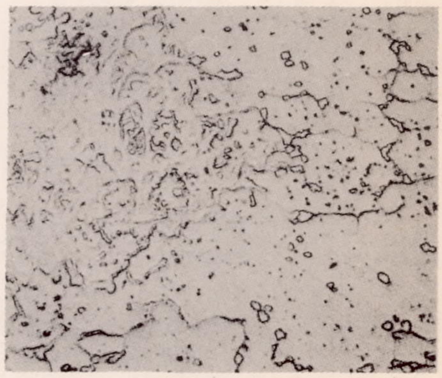
100X

1000X

(a) Original microstructure.
Aqua Regia in Glycerine etch - Vickers hardness 249.



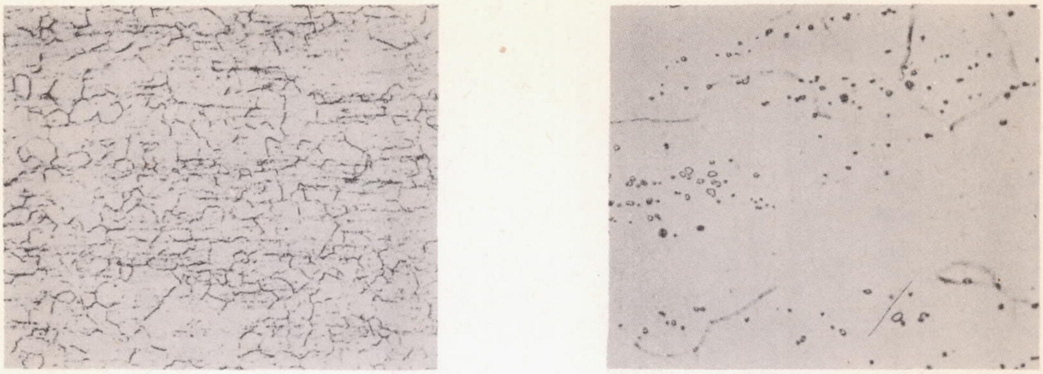
Fracture - 100X



Interior - 1000X

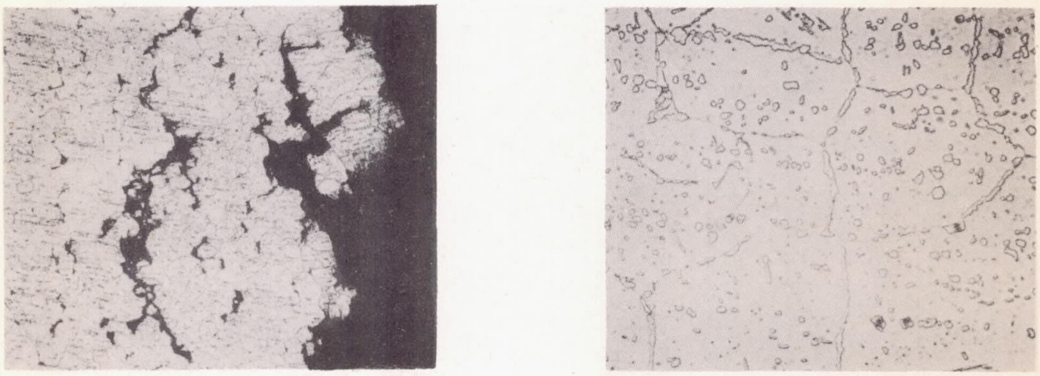
(b) 318 hours for rupture at 1700° F under 3500 psi.
Electrolytic Chromic acid etch - Vickers hardness 160.

FIGURE 37.- MICROSTRUCTURES OF HOT-ROLLED LOW-CARBON N155 ALLOY.



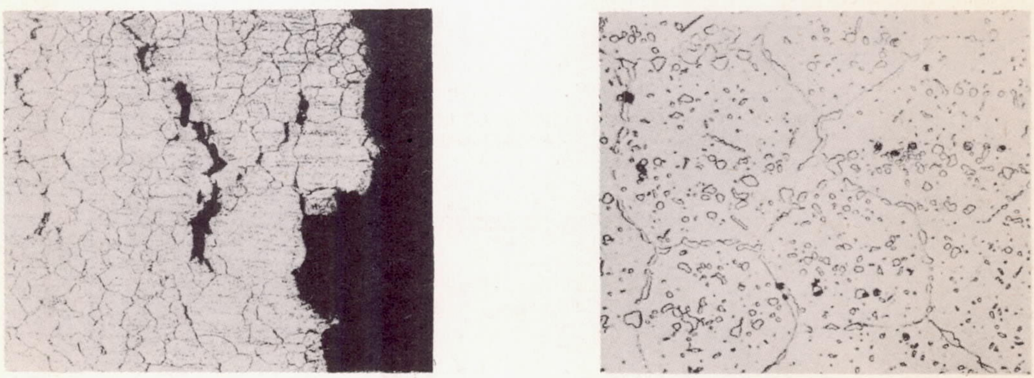
100X 1000X

(a) Original Microstructure.
Aqua Regia in Glycerine etch - Vickers hardness 201.



Fracture - 100X Interior - 1000X

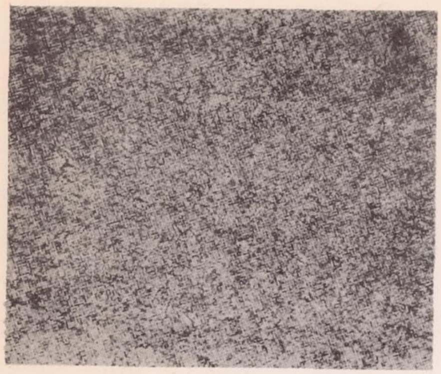
(b) 516 hours for rupture at 1700° F under 5500 psi.
Electrolytic Chromic acid etch - Vickers hardness 199.



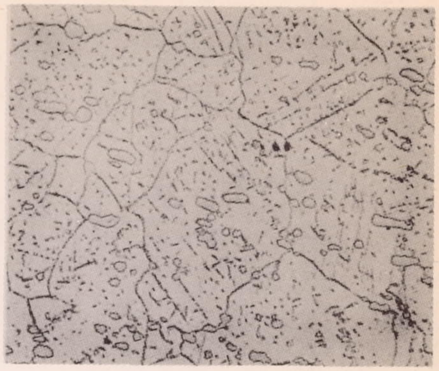
Fracture - 100X Interior - 1000X

(c) 415 hours for rupture at 1800° F under 3500 psi.
Electrolytic Chromic acid etch - Vickers hardness 201.

FIGURE 38.- MICROSTRUCTURES OF LOW-CARBON N155 ALLOY AIR COOLED FROM 2282° F.

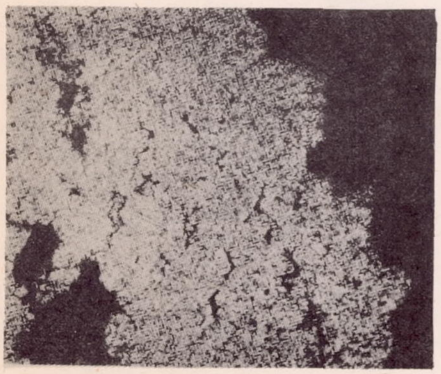


100X

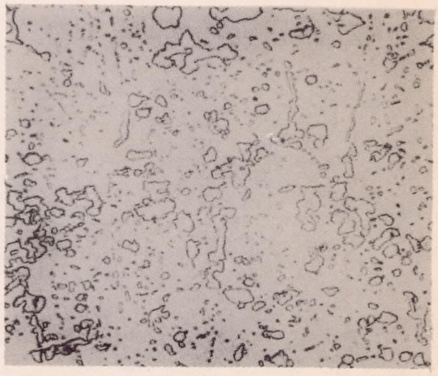


1000X

(a) Original microstructure.
Electrolytic Chromic acid etch - Vickers hardness 244.



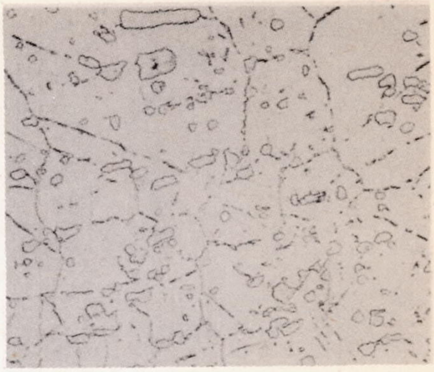
Fracture - 100X



Interior - 1000X

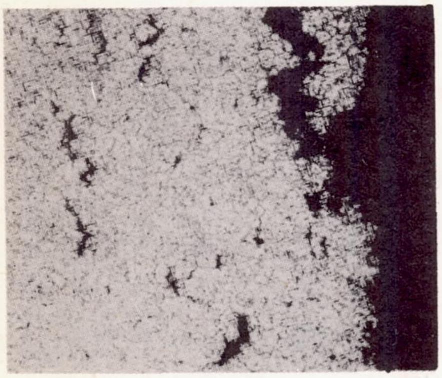
(b) 600 hours for rupture at 1700° F under 7200 psi.
Electrolytic Chromic acid etch - Vickers hardness 226.

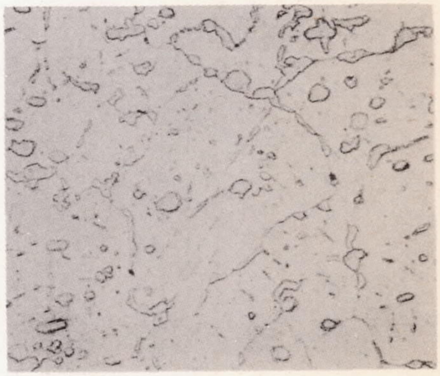
FIGURE 39.- MICROSTRUCTURES OF WROUGHT S590 ALLOY WATER QUENCHED FROM 2270° F AND AGED 16 HOURS AT 1700° F.



100X 1000X

(a) Original microstructure.
Electrolytic Chromic acid etch - Vickers hardness 231.



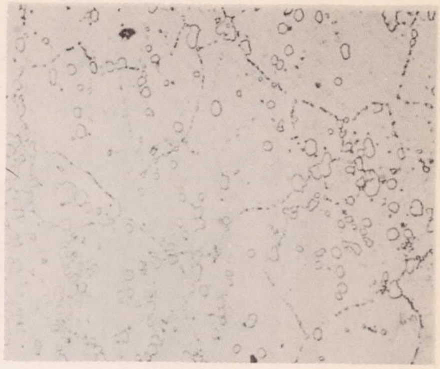


Fracture - 100X Interior - 1000X

(b) 374 hours for rupture at 1800° F under 4200 psi.
Electrolytic Chromic acid etch - Vickers hardness 216.

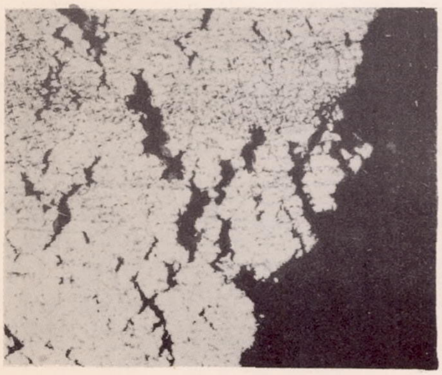
FIGURE 40.- MICROSTRUCTURES OF WROUGHT S590 ALLOY WATER QUENCHED FROM 2270° F AND AGED 16 HOURS AT 1800° F.

100X

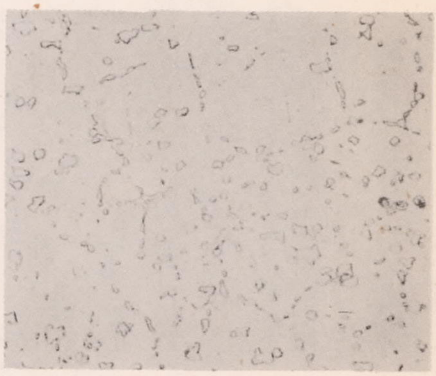


1000X

(a) Original microstructure.
Electrolytic Chromic acid etch - Vickers hardness 280.



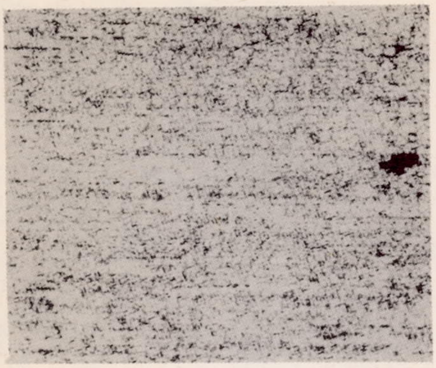
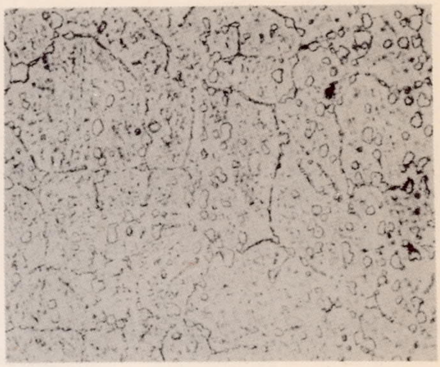
Fracture - 100X



Interior - 1000X

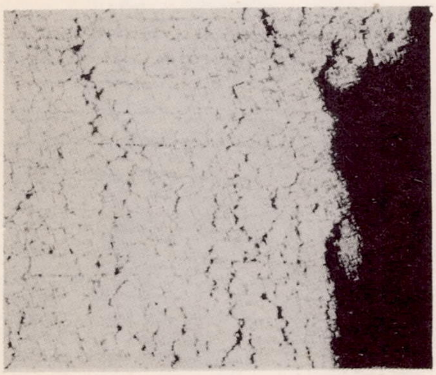
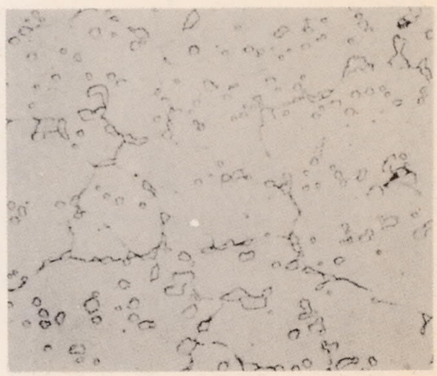
(b) 390 hours for rupture at 1700° F under 7000 psi.
Aqua Regia in Glycerine etch - Vickers hardness 263.

FIGURE 41.- MICROSTRUCTURES OF WROUGHT S816 ALLOY WATER QUENCHED FROM 2300° F AND AGED 16 HOURS AT 1700° F.

100X 1000X

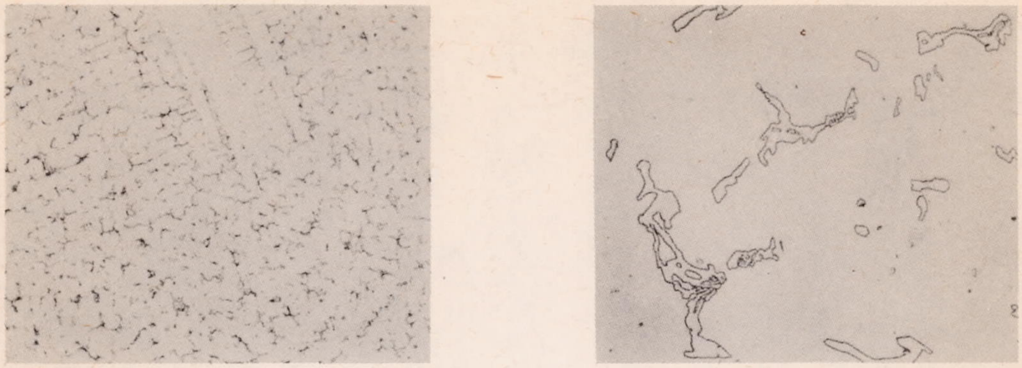
(a) Original microstructure.
Electrolytic Chromic acid etch - Vickers hardness 282.

Fracture - 100X Interior - 1000X

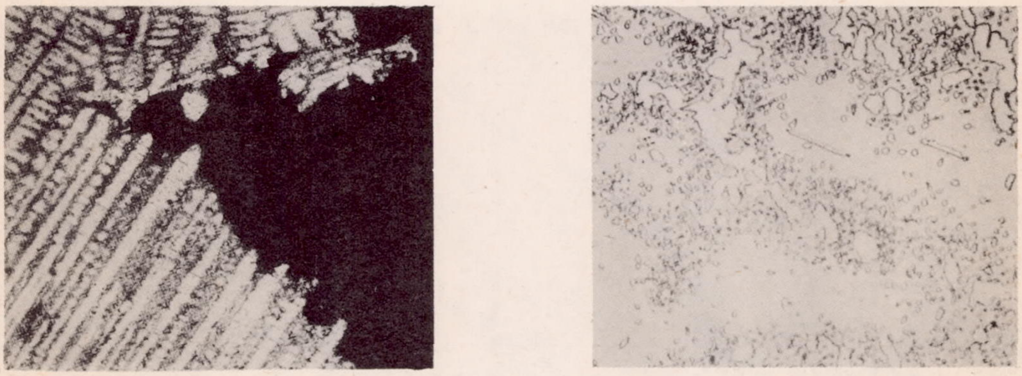
(b) 328.5 hours for rupture at 1800° F under 4000 psi.
Aqua Regia in Glycerine etch - Vickers hardness 270.

FIGURE 42.- MICROSTRUCTURES OF WROUGHT S816 ALLOY WATER QUENCHED FROM 2300° F AND AGED 16 HOURS AT 1800° F.



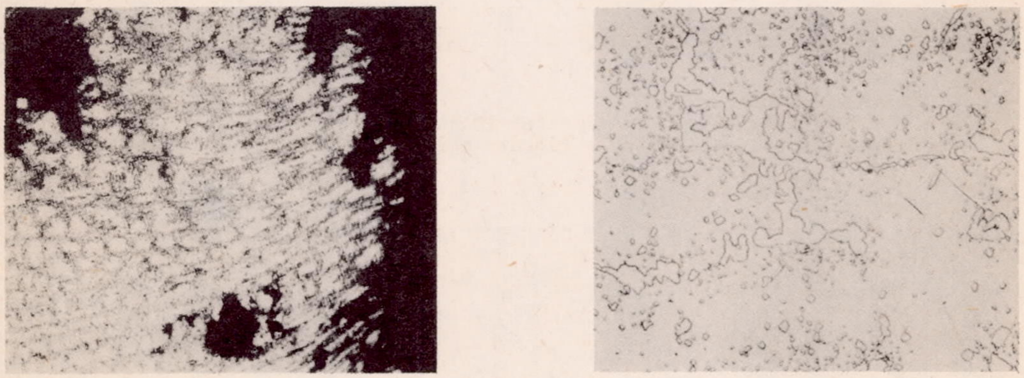
100X 1000X

(a) Original microstructure - As cast.
Electrolytic Chromic acid etch - Vickers hardness 234.



Fracture - 100X Interior - 1000X

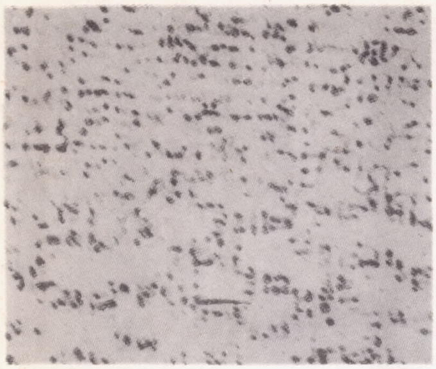
(b) 362.5 hours for rupture at 1700° F under 10,000 psi.
Electrolytic Chromic acid etch - Vickers hardness 290.



Fracture - 100X Interior - 1000X

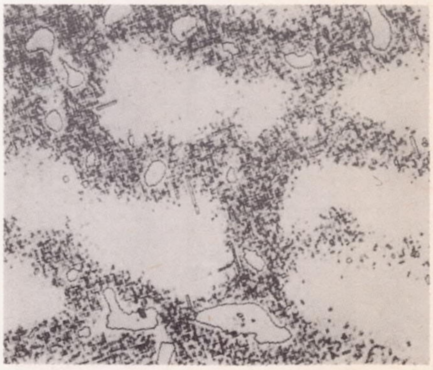
(c) 281 hours for rupture at 1800° F under 8000 psi.
Electrolytic Chromic acid etch - Vickers hardness 298.

FIGURE 43.- MICROSTRUCTURES OF PRECISION-CAST 6059 ALLOY.



100X 1000X

(a) Original microstructure - As cast.
Electrolytic Chromic acid etch - Vickers hardness 286.

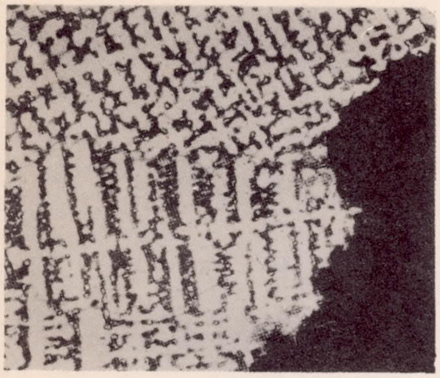


Fracture - 100X Interior - 1000X

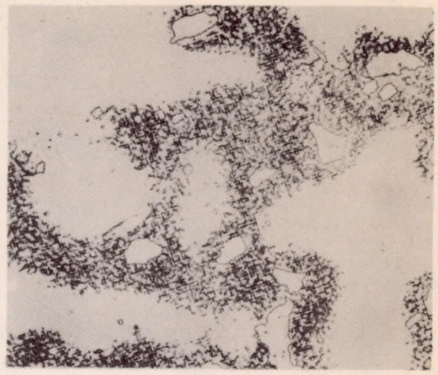
(b) 625 hours for rupture at 1700° F under 10,500 psi.
Electrolytic Chromic acid etch - Vickers hardness 398.

FIGURE 44a.- MICROSTRUCTURES OF PRECISION-CAST VITALLIUM ALLOY.

W-75

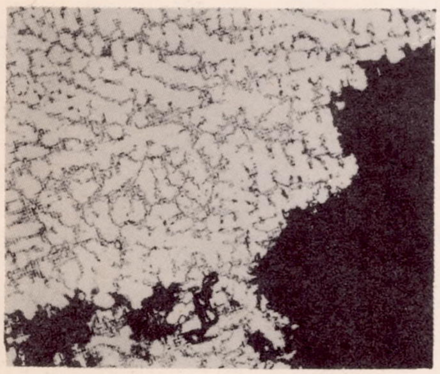


Fracture - 100X

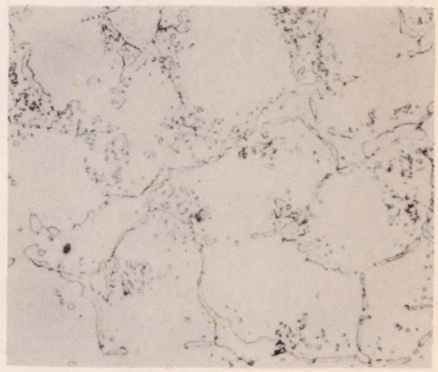


Interior - 1000X

(a) 527 hours for rupture at 1800° F under 7500 psi.
Electrolytic Chromic acid etch - Vickers hardness 354-386.



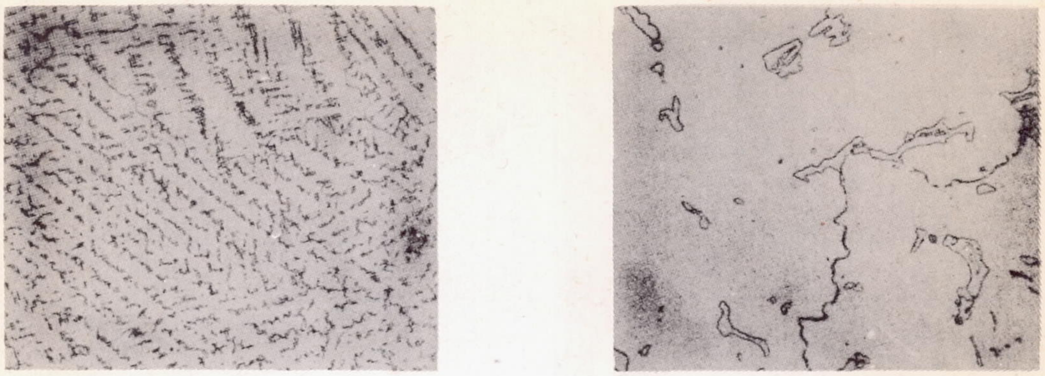
Fracture - 100X



Interior - 1000X

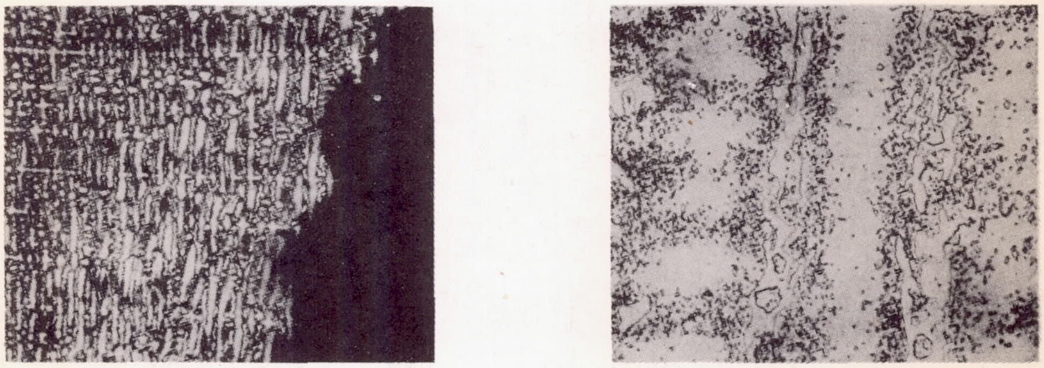
(b) 895 hours for rupture at 1800° F under 5500 psi.
Electrolytic Chromic acid etch - Vickers hardness 359.

FIGURE 44b.- MICROSTRUCTURES OF PRECISION-CAST VITALLIUM ALLOY.



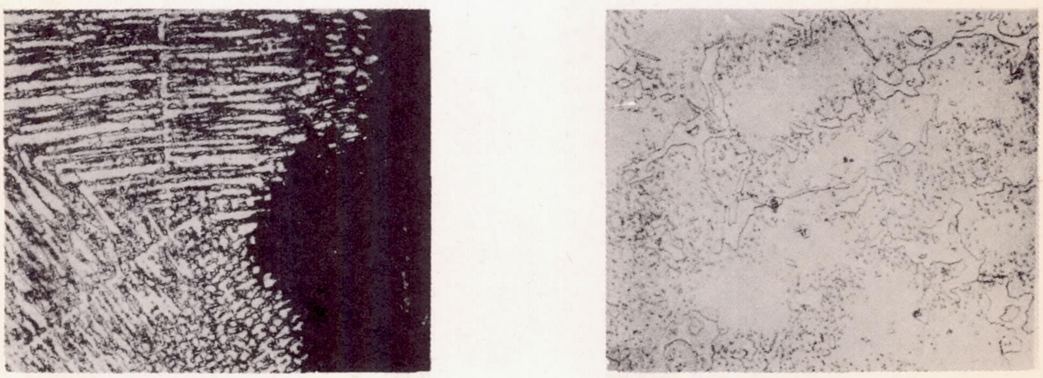
100X 1000X

(a) Original microstructure - As cast.
Electrolytic Chromic acid etch - Vickers hardness 309.



Fracture - 100X Interior - 1000X

(b) 346 hours for rupture at 1700° F under 13,000 psi.
Electrolytic Chromic acid etch - Vickers hardness 405.

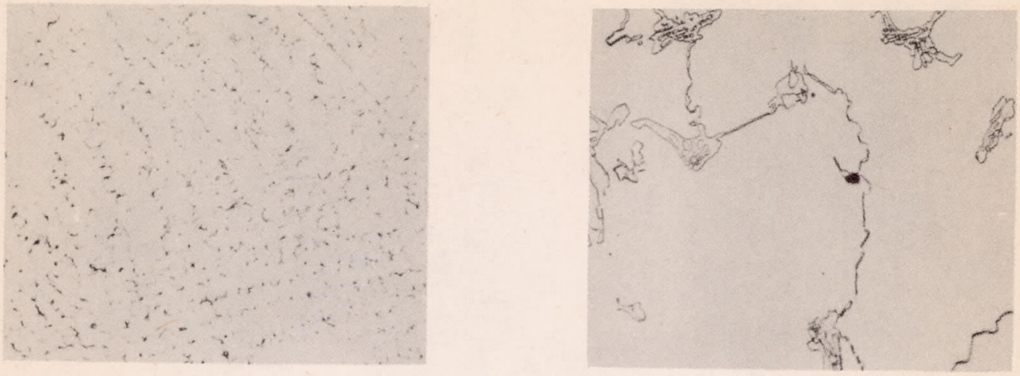


Fracture - 100X Interior - 1000X

(c) 460 hours for rupture at 1800° F under 8000 psi.
Electrolytic Chromic acid etch - Vickers hardness 368.

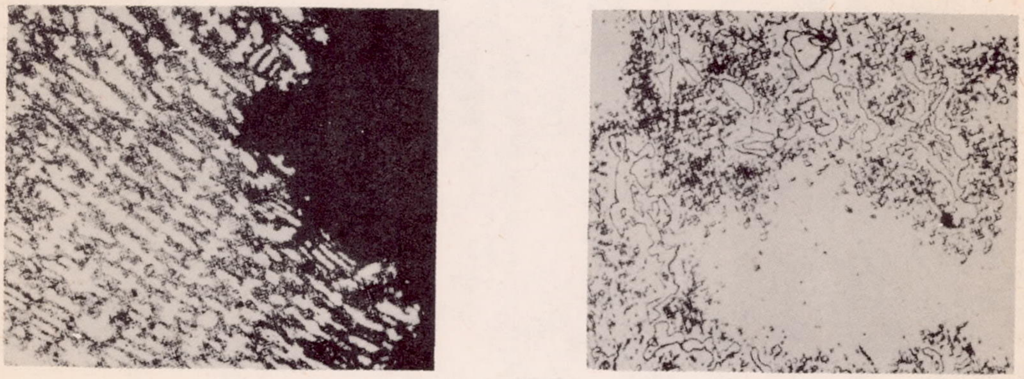
FIGURE 45.- MICROSTRUCTURES OF PRECISION-CAST 422-19 ALLCY.

W-75



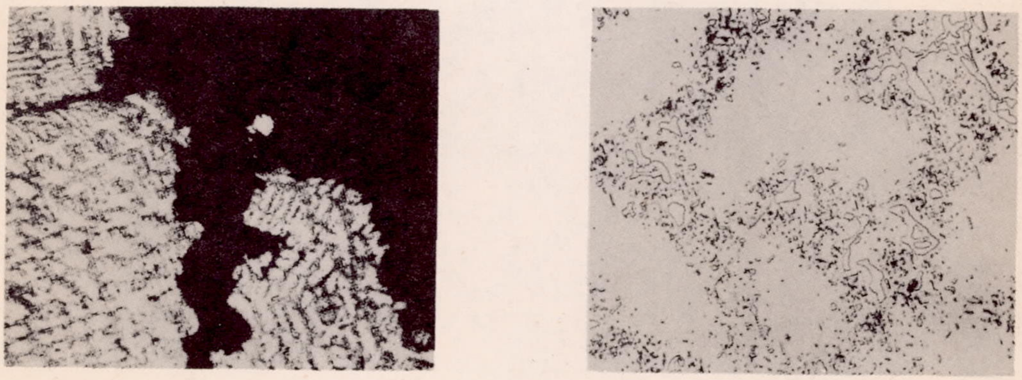
100X 1000X

(a) Original microstructure - As cast.
Electrolytic Chromic acid etch - Vickers hardness 333.



Fracture - 100X Interior - 1000X

(b) 717 hours for rupture at 1700° F under 12,000 psi.
Electrolytic Chromic acid etch - Vickers hardness 383.

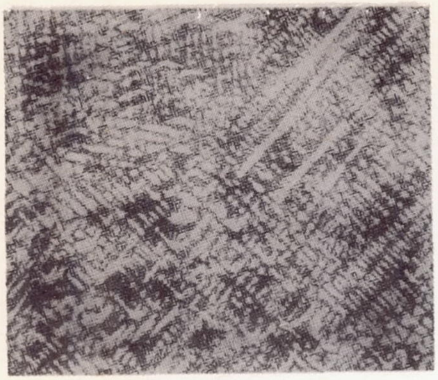


Fracture - 100X Interior - 1000X

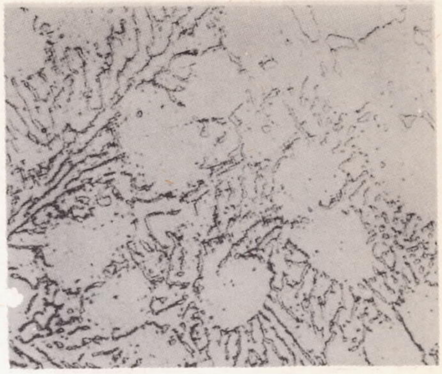
(c) 283.5 hours for rupture at 1800° F under 7000 psi.
Electrolytic Chromic acid etch - Vickers hardness 371.

FIGURE 46.- MICROSTRUCTURES OF PRECISION-CAST 61 ALLOY.

W-75

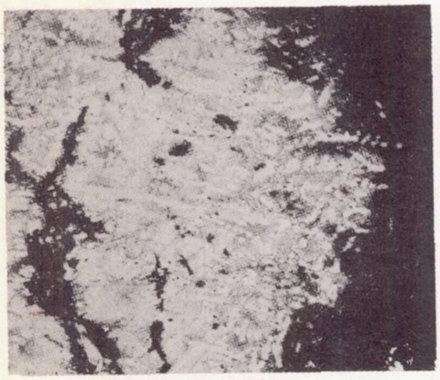


100X



1000X

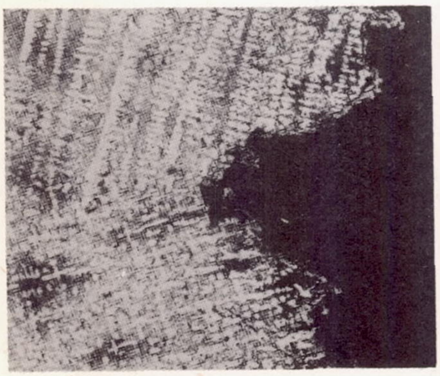
(a) Original microstructure - As cast.
Electrolytic Chromic acid etch - Vickers hardness 249.



Fracture - 100X

Interior - 1000X

(b) 378.5 hours for rupture at 1700° F under 9500 psi.
Electrolytic Chromic acid etch - Vickers hardness 282.



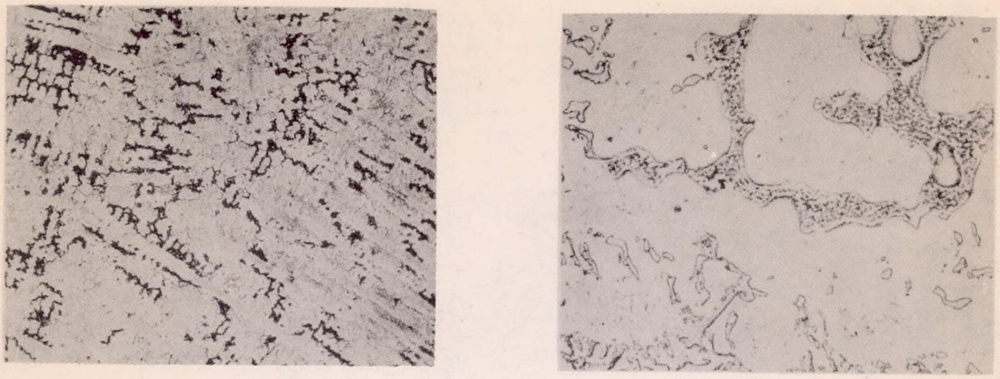
Fracture - 100X

Interior - 1000X

(c) 480 hours for rupture at 1800° F under 6500 psi.
Electrolytic Chromic acid etch - Vickers hardness 254.

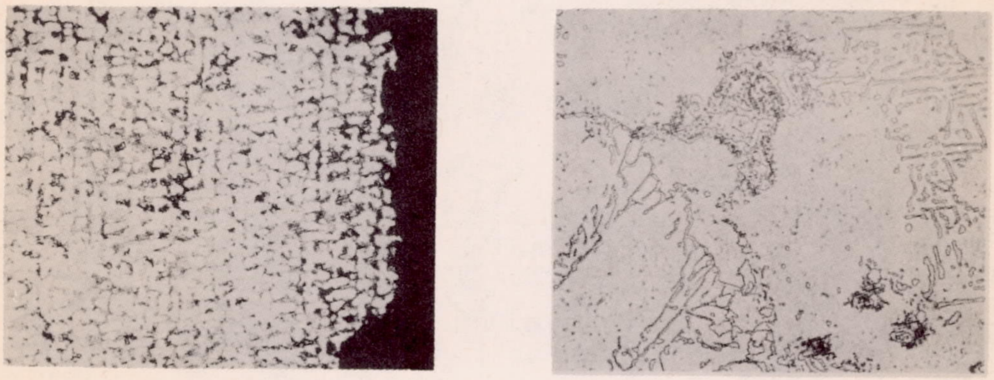
FIGURE 47.- MICROSTRUCTURES OF PRECISION-CAST S590 ALLOY.

W-75



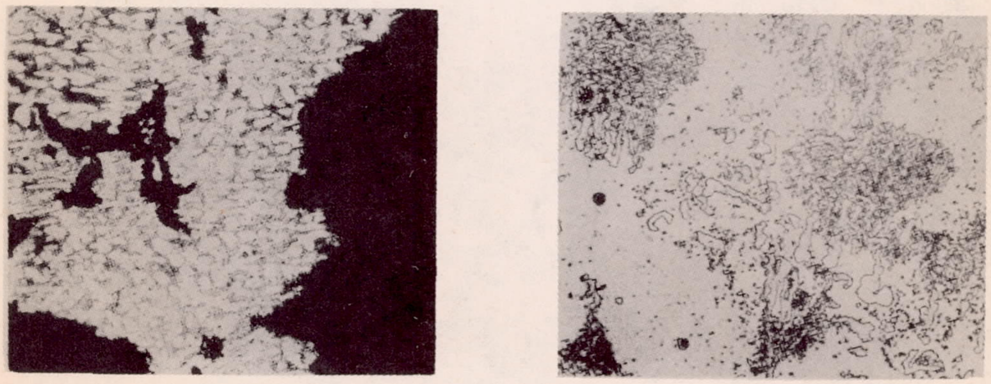
100X   1000X

(a) Original microstructure - As cast.
Electrolytic Chromic acid etch - Vickers hardness 333.



Fracture - 100X   Interior - 1000X

(b) 327 hours for rupture at 1700° F under 13,000 psi.
Electrolytic Chromic acid etch - Vickers hardness 320.

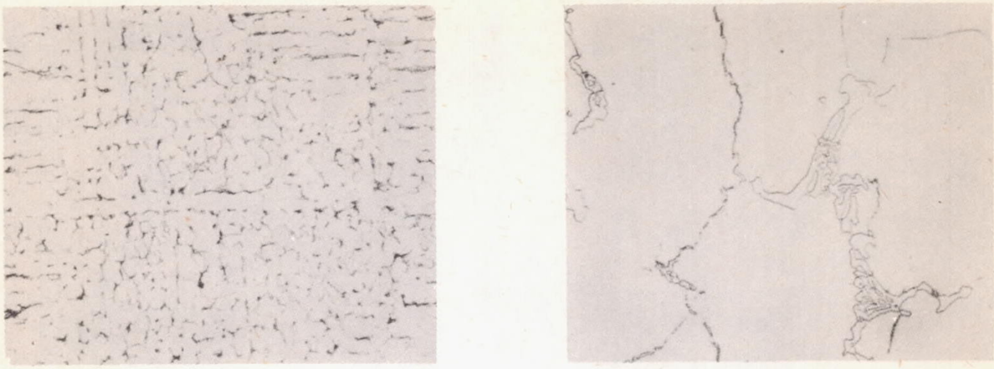


Fracture - 100X   Interior - 1000X

(c) 385 hours for rupture at 1800° F under 9000 psi.
Electrolytic Chromic acid etch - Vickers hardness 292.

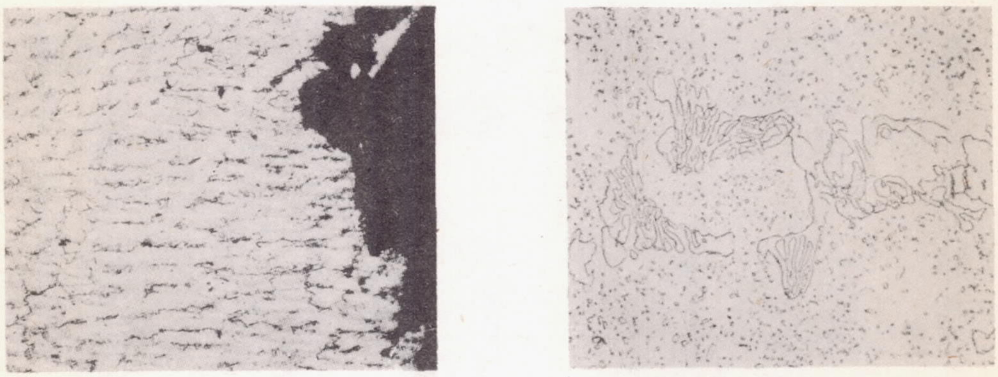
FIGURE 48.- MICROSTRUCTURES OF PRECISION-CAST S816 ALLOY.

W-75



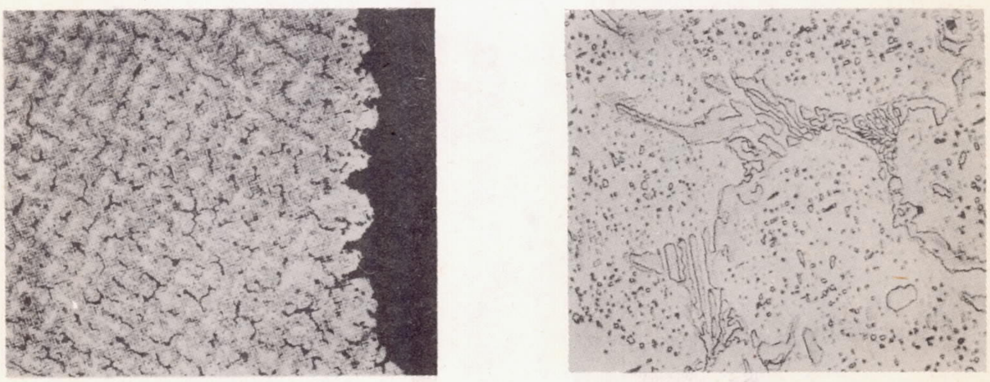
100X 1000X

(a) Original microstructure - As cast.
Electrolytic Sodium Cyanide etch - Vickers hardness 290.



Fracture - 100X Interior - 1000X

(b) 828 hours for rupture at 1700° F under 14,000 psi.
Electrolytic Sodium Cyanide etch - Vickers hardness 389.

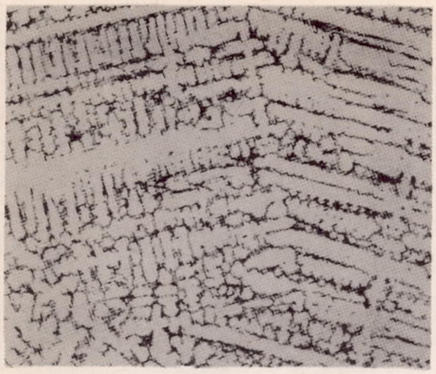


Fracture - 100X Interior - 1000X

(c) 688 hours for rupture at 1800° F under 10,000 psi.
Electrolytic Sodium Cyanide etch - Vickers hardness 371.

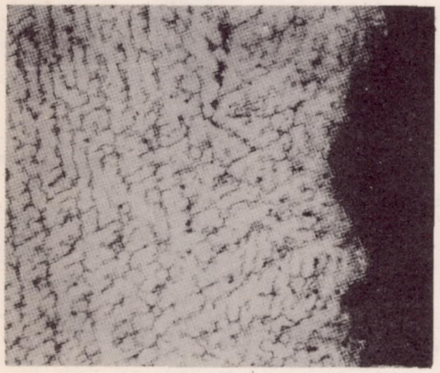
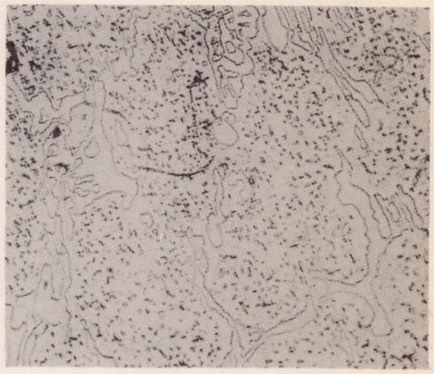
FIGURE 49.- MICROSTRUCTURES OF PRECISION-CAST X-40 ALLOY.

W-75



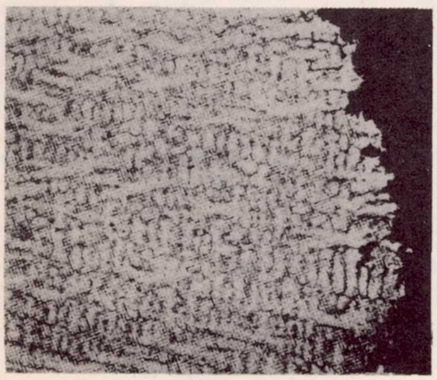
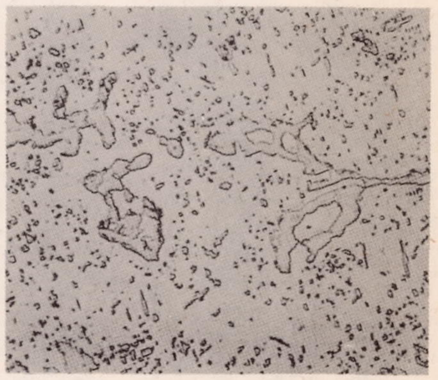

100X 1000X

(a) Original microstructure - As cast.
Electrolytic Sodium Cyanide etch - Vickers hardness 286.

Fracture - 100X Interior - 1000X

(b) 848 hours for rupture at 1700° F under 14,000 psi.
Electrolytic Chromic acid etch - Vickers hardness 390.

Fracture - 100X Interior - 1000X

(c) 665 hours for rupture at 1800° F under 8000 psi.
Electrolytic Sodium Cyanide etch - Vickers hardness 348.

FIGURE 50.- MICROSTRUCTURES OF PRECISION-CAST X-50 ALLOY.

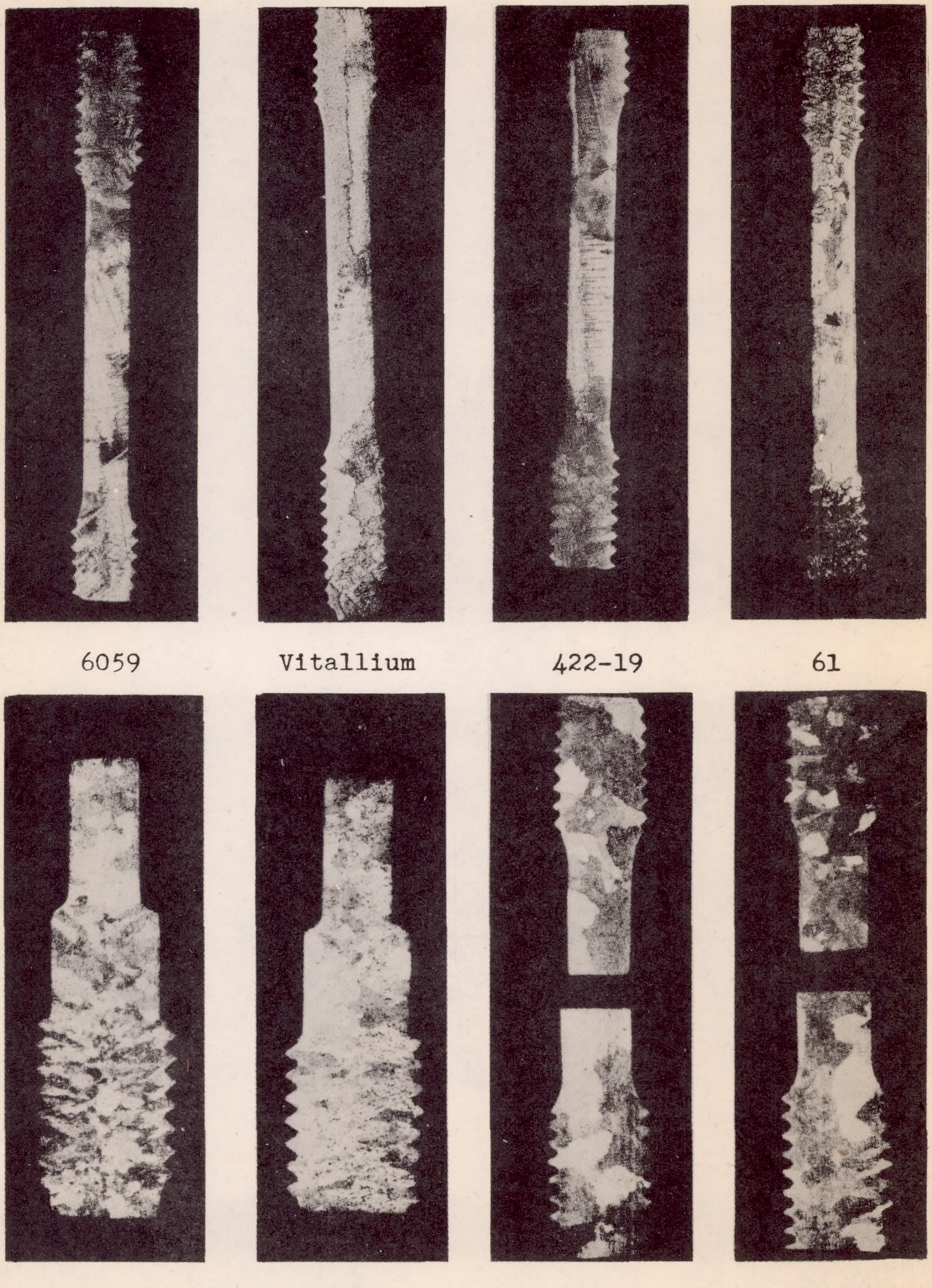


FIGURE 51.- MACROSTRUCTURES OF PRECISION-CAST ALLOYS.