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DEVELOPMENT OF WROUGHT AND CAST ALLOYS FOR HIGH TEMPERATURE APPLICATIONS

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FOREWORD

This report was prepared by the Watervliet Research Laboratory, Allegheny Ludlum Steel Corporation under USAF Contract No. AF 33(616)-2463. This contract was initiated under Project No. 7351, "Metallic Materials", Task No. 73512, High Temperature Alloys", formerly RDO No. 615-13, "High Temperature Alloys", and was administered under the direction of the Materials Laboratory, Directorate of Research, Wright Air Development Center with Captain C. M. Hollyfield acting as project engineer.

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ABSTRACT

Developmental studies were conducted on wrought Fe-base and both wrought and cast Co-base alloys for applications at high temperatures. A heat treatable, Fe-base, austenitic alloy containing Mn and Cr was modified with B to give excellent stress-rupture properties at 1200°F. Oxidation resistance was greatly improved by small Al additions. A wrought Co-base alloy with good stress-rupture properties at 1600 and 1700°F and improved oxidation resistance was developed. Composition levels giving optimum properties were determined for the cast Co-base alloys. Modifications involving B were investigated in both wrought and cast Co-base alloys.

PUBLICATION REVIEW

This report has been reviewed and is approved.

FOR THE COMMANDER:

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SUMMARY

Alloy development studies aimed at development of better materials for application at elevated temperatures were carried out on wrought Fe-base alloys and wrought and cast Co-base alloys.

Wrought Fe-Base Alloys

An extensive study of heat treatment, hot working, and cold working on D-183 alloy resulted in an optimum heat treatment of solution at 1900°F and aging at 1300°F. Modifications containing .20 B* resulted in an alloy which had a 100-hour rupture strength of 67,000 psi at 1200°F. This alloy is designated R-71. Its optimum heat treatment consisted of a solution treatment at 2050°F followed by an age at 1300°F. Additions of .10 to .20 Al resulted in a marked improvement in oxidation resistance of D-183 alloy.

Wrought Co-Base Alloys

A wrought Co-base alloy, R-94, was found which combines good rupture properties at 1600 and 1700°F with adequate oxidation resistance. The composition of R-94 is .10 C, 1.20 Mn, .30 Si, 15 Cr, 10 Ni, 5 Mo, 10 W, 1 Cb+Ta, 3 Fe, balance Co. It also has comparatively good thermal shock resistance. Optimum C, Cr, and Cb+Ta levels were found to be those in the above analysis. A detrimental effect on rupture strength was found for B additions up to 0.20.

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* All analyses of elements are in terms of weight per cent. Percentage symbols have been deleted for brevity.

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Cast Co-Base Alloys

A cast Co-base alloy, F-538, containing .7 C, 20 Cr, 10 Ni, 10 W, 5 Mo, 1 Cb+Ta, 1 Fe was found to have optimum high temperature rupture and thermal shock properties in this study of cast R-17 and R-27 alloys. B had an effect on improving properties with an optimum B content around 0.6.

INTRODUCTION

This report describes research investigations in the development of forged and cast alloys for high temperature applications such as in aircraft gas turbines. It is a continuation of work carried out under a previous Air Force contract as reported under reference 1. Wrought-Fe-base and both wrought and cast Co-base alloys were studied.

The Fe-base alloys for service around 1200°F were based on an austenitic Mn-Cr steel, D-183, developed under the previous contract.. In the wrought and cast Co-base alloys for service around 1600 to 1700°F, work was concentrated on R-27, a low Cr, low Cb alloy which had outstanding rupture properties as reported in reference 1.

The alloys were mainly considered with the object of improving, by composition variations, the high temperature rupture properties. Other criteria used in the evaluation of the alloys were tensile, hardness, oxidation, thermal shock, and microstructural characteristics.

This is the final report presenting results of these studies under Contract AF 33(616)-2463. It covers the contract period from April 1, 1954 to December 31, 1954.

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WROUGHT ALLOYS

The experimental alloys in which composition variations were made were melted as small induction furnace heats and forged to bar stock. The composition of the base D-183 alloy was: .30 C, 18 Mn, .4 Si, 12 Cr, 3 Mo, .75 V, .15 N, and the balance Fe. Studies were made of the effect of heat treatment and cold work on the properties of this alloy. Alloy modifications included variations of Mo and C and additions of B and Al. Twenty-one heats were melted in this study of Fe-base alloys.

The composition of the base R-27 alloy was: .10 C, 1.2 Mn, .3 Si, 10 Cr, 10 Ni, 10 W, 5 Mo, 1 Cb+Ta, 1 Fe and the balance Co. Major alloying modifications were made in Cr, Cb+Ta, C, and B contents with limited modifications of Mo, W, and Ni. A few C and Cb+Ta modifications of R-17, made early in the program, are included in this presentation. A total of thirty-three wrought Co-base alloys were made.

Wrought Alloy Procedures

Melting

All of the initial modifications made under this contract were melted with virgin materials in a 40 KW Ajax induction furnace as 17-pound heats using magnesia crucibles. In some instances, where remelts of promising D-183 modifications containing B were made, a larger size crucible was employed which yielded 35-pound ingots. Pouring temperatures were between 2700 and 2800°F as measured by Pt, Pt-Rh immersion thermocouple. Both ingots were tapered and approximately 8 inches long. The 17-pound ingot was 2-3/4 inches square at the top and the 35-pound ingot

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was 3-5/8 inches square at the top.

For the study of the effects of treatment on D-183 alloy, the bar stock from a 12-inch square, 26-inch long ingot from a 1100-pound arc-furnace heat was used. This heat was melted and bar stock made under the previous Air Force contract.

Forging and Rolling

All the forging was done on a 2000-pound steam hammer between flat dies.

A forging temperature of 2050°F was used for all the D-183 modifications except those containing B. A good average forging temperature for the B-containing heats appeared to be 1950°F. The ingots were forged to 5/8 inch square bar stock in six to ten heatings, depending on the plasticity and workability of the modification.

Four 35-pound induction heats of D-183 containing B (R-102 to R-105) were forged on a 2000-pound steam hammer to 2-inch squares from 1950°F. These were then rolled from 1950°F on a 9-inch guide round production mill to 51/64-inch round in approximately twelve passes. This latter hot working was decided upon after a careful consideration of the hot working difficulties which were encountered in forging similar B modifications (17-pound heats R-68 to R-72). These latter heats forged well to an intermediate size, but subsequent finish forging resulted in bursting along diagonal planes of the square bar stock. These planes are the areas of maximum shear in forging. It was evident, from previous experience with similar hot working difficulties in other alloys, that this condition could be eliminated by roll-

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ing in closed passes to a round bar. On the 35-pound heats, this was successful in producing sound stock from heats R-102, R-103, and R-104. Heat R-105 (with the highest B), however, split badly after five rolling passes and was scrapped.

The cold rolling and hot-cold working on standard D-183 was done on a laboratory 2-high rolling mill. The rolls measured 5-3/4 inches diameter by 3-1/4 inches long. The stock used in this program was taken from a 1000-pound ingot which was melted and processed to 7/8-inch round under the previous contract. Prior to rolling, the 7/8-inch round was forged to .725-inch square. A maximum reduction of .020 inch (2.7 per cent reduction) was maintained for the individual rolling passes. Control of final size in any amount of reduction varied ± .001 inch. In hot-cold working, four passes per heating were made, one on each side of the square. In hot-cold working, four reheatings were required for 10 per cent reduction, eight for 20 per cent reduction, and thirteen for 30 per cent reduction. Furnace temperatures were controlled to ± 10°F. There was little drop in temperature during working.

On the previous contract, the Co-base alloys were forged at 2150°F to sizes ranging from 5/8-inch to 7/8-inch square. Under the present contract, it was decided to use a 2200°F forging temperature whenever possible to minimize stiffness and corner cracking. Final stock size was 5/8-inch square. Ten to fifteen heatings were required. On a few of the R-27 modifications, it was necessary to discontinue forging at 3/4-inch square because of corner cracks. Because of hot shortness, the B modifications of R-27 were forged from a temperature of 2150°F.

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Rupture Testing

The standard rupture specimen used in these tests was 3 inches long with a 1-inch reduced section and a .252-inch diameter. Specimens were generally machined from the center of the 5/8-inch bar stock. However, in certain cases where center bursts were observed in either completed test specimens or in original bar stock, the test specimens were machined away from the unsound center. Such specimens are noted in the tabulations of results.

The "V-notch" rupture specimens had an outside diameter of .275-inch with a 60° notch at the center of the reduced section, a .005-inch radius at the base of the notch, and a .195-inch diameter across the base of the notch. The notches were produced by wet grinding. The notch geometry was such as to produce a theoretical stress concentration factor (K_t) of 4.2 for the notched bar in tension. This type of test was used as a measure of notch sensitivity in rupture life. The criterion for freedom from notch sensitivity was that the "V-notch" rupture life would be at least as long as the standard or smooth bar rupture life under a given stress.

A lever arm, constant load, rupture testing unit was used. The specimens were loaded into an 18-inch long resistance furnace with a 2-inch diameter muffle. The lever arm, attached to the bottom of the specimen holder, had a 10:1 ratio. The temperatures were controlled by Brown electronic potentiometer recording controllers. Temperature control varied ± 3°F at 1200°F testing temperatures, and ± 5°F at 1500 to 1700°F. A drop switch on the lever arm shut off the furnace current and controller at fracture.

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Tensile Testing

The tensile specimens were machined with a .252-inch diameter and a 2-inch reduced section. The tests were run in a 60,000-pound capacity mechanical Riehle testing machine. Stress strain curves were recorded from a Peters extensometer. High-temperature tests were run in a Marshall furnace with a temperature control of ± 3°F at 1200°F and ± 5°F at 1500°F.

Thermal Shock Testing

The thermal shock test involved repeatedly heating the edge of a triangular-shaped sample in a propane-air flame from a Selas burner to a certain maximum temperature and then suddenly cooling this edge in a blast of compressed air.

The thermal shock apparatus consists of a steel framework to which are attached a control box that houses a cycle counter, relays, and switches necessary for automatic operation. Solenoid air valves and pneumatic pistons are used to move the burners and air cooling nozzles back and forth in front of the test piece and to turn the air blast on and off as required. The temperature-time-cycle of the test is controlled automatically by a Brown radiamatic head focused on the test edge acting through a high-speed Brown circular chart electronic potentiometer. The burner and the controller are adjusted to obtain the desired edge temperature, 1800°F, which was measured with a Pyro optical pyrometer.

In this apparatus, two specimens can be tested at one time at the same test temperatures. All tests were made by cycling the specimen from 800 to 1800°F in a 20-second cycle: 16 seconds to heat to 1800°F and 4 seconds to air cool to 800°F.

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The specimen used was a 45° angle wedge 3/8-inch deep with a 1/32-inch wide test edge. It was two inches long. The burner heated about a 1-inch length of the specimen.

Thermal shock failures were determined by inspecting the specimens under a 20-power binocular microscope every 500 cycles until cracking began and approximately every 50 cycles after the appearance of the initial cracks. Failure was considered reached when a crack had proceeded completely across the 1/32-inch test edge.

Oxidation Tests

These tests were run on 1/2-inch diameter x 1/2-inch long polished samples in still air in a muffle furnace with line contact between sample and holder. The round sample was set in a tray which had round grooves in the surface to prevent the specimens from rolling. This provided not only stability for the sample but the least area of contact between tray and sample. Temperature variation in the furnace was ± 3°F. Each test was run in duplicate. The completed tests were cleaned electrolytically in a **sodium** hydroxide - sodium carbonate salt bath at 880°F. Weighings made on samples before and after testing were accurate to ± .002 grams.

Wrought Alloy Results and Discussions

Wrought Fe-Base Alloys

Table I lists the chemical analyses of the modifications of D-183 alloy. Tables II, III, IV, and V present the tensile, rupture, and hardness results of investigations involving heat treatment and mechanical working of the basic D-183 alloy. Table VI

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gives the room temperature tensile results and Table VII the elevated temperature tensile results on D-183 and its modifications. Rupture results are given in Table VIII.

In Figures 1 through 4 are shown the results of heat treatment investigations involving solution and aging temperatures.

Figure 1 indicates that an optimum combination of properties can be obtained with an 1800°F solution temperature in the absence of subsequent aging. In Figure 2, a 1900°F solution temperature with a 16 hour age at 1300°F gave comparable rupture properties and hardness. This latter treatment appeared best, since the alloy has shown itself to be responsive to aging, and since after this treatment it is notch ductile in the rupture test.

Results of studies of the effect of aging temperatures after solution treatments at 2100 and 1900°F are shown in Figures 3 and 4. Although aging after a 2100°F solution treatment gave a marked increase in hardness, there was no outstanding rupture life improvement. Rupture ductilities were also relatively low.

In Figure 4, D-183 alloy hardens during aging at 1200°F to a Brinell hardness of 255, drops some in hardness after aging at 1300°F, and then rises to a Brinell hardness of 277 after aging at 1400 to 1600°F. The rupture life increases with aging temperatures up to 1300°F and then rapidly decreases after higher aging temperatures. Rupture ductility drops slightly with aging at 1200°F and rises with aging temperatures above 1200°F. Microstructural examinations indicated the reason for this response to aging to be the increasing amounts of an excess constituent obtained at and above 1300°F which, above a critical amount, lowered rupture life although hardness and rupture ductility

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continued to increase. It thus appears the alloy gains some strength from solution effects as well as by an aging response.

Figure 5 shows the effect of test temperature on the tensile properties of D-183 alloy. Tensile strength drops off rapidly with increasing temperatures. Little drop in yield strength or change in ductility was noted for the alloy between 1000 and 1500°F.

Figures 6 and 7 show the effect of mechanical working on rupture properties of D-183. The most outstanding improvement in rupture life resulted from hot-cold working at 1300°F although both cold working and hot-cold working at any temperature increased rupture life. Rupture ductilities, also shown in Figures 6 and 7, were lowered seriously enough to overshadow this increased rupture strength. With these low ductilities the notch rupture sensitivity would be high.

The room temperature hardness increase from cold working and hot-cold working was reflected in the large increase in tensile and yield strengths. This is shown in Figure 8. Room temperature ductility exhibited some variation but was not reduced below 20 per cent elongation.

Figure 9 presents the influence of B on the rupture properties at 1200°F of D-183 alloy. It is noted that an optimum rupture life was obtained at about .20 B. There was also a marked increase in the ratio of the notch to smooth bar rupture life with increasing B. These simultaneous improvements indicated that possibly a further increase in rupture strength could be obtained with heat treatment while maintaining satisfactory notch life.

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The results of varying solution and aging temperatures in such studies are shown in Figures 10 and 11. The results in Figure 10 indicate that the minimum aging temperature for satisfactory notch rupture life was 1300°F. Greater smooth bar lives were obtained at the lower aging temperature of 1200°F but notch life was very poor. There is also an indication in Figure 10 that a 2000°F solution temperature resulted in a higher level of rupture strengths.

Further studies of solution treatment temperature with a 1300°F aging treatment (Figure 11) indicated an added improvement in rupture life with increasing solution temperatures up to 2100°F. Rupture ductility was not impaired by this increased solution temperature. Results in Table IX indicate the alloy is notch ductile after a 2050°F solution plus a 1300°F aging treatment. However, it had a tendency toward notch sensitivity when the solution temperature was raised to 2100°F. The former treatment, 2050°F plus 1300°F treatment, appeared to be best for the alloy.

This B modification exhibited the presence of a low-melting phase after treating at temperatures of 2150°F and above as shown by the microstructures in Figure 12. This is another factor in limiting the maximum practical solution temperature to 2050°F.

This modified D-183 alloy containing .2 B has been designated R-71 alloy. With its optimum heat treatment, it has a 100-hour rupture strength of 67,000 psi at 1200°F. This represents an improvement of approximately 15,000 psi over that of the base D-183 alloy. The 1000-hour rupture strength of R-71 is around 52,000 psi.

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A limited amount of investigation of B-containing D-183 modified with lower C and lower Mo did not yield any modification of significance.

Another object in the study of this relatively low Cr alloy, D-183, was directed at improving its oxidation resistance. Increasing the Cr content in the alloy was limited by the ferrite-forming tendencies of this element. In Table X are shown the results of oxidation tests run at 1200 and 1500°F. For comparison, similar tests were run on 16-25-6 and A-286 alloys. The results at 1200°F are of such small magnitude that experimental error in testing and descaling may have overshadowed any significant differences. The tests at 1500°F showed D-183 to have the poorest oxidation resistance of the alloys.

The pronounced improvement in oxidation resistance at 1500°F caused by small Al additions to D-183 is shown in Table X and Figure 13. With an Al addition of only .10 to .20, the oxidation resistance was increased to the same level as 16-25-6 alloy. Further Al additions up to 2.2 gave little further improvement.

The rupture properties of the alloys containing .10 to .20 Al are at least as good as those of D-183 alloy. With higher Al there was a marked decrease in rupture strength. It is therefore concluded that the oxidation resistance of D-183 can be markedly improved by slight additions of Al. While the limits of the investigation did not permit the melting of the B-modified D-183 (R-71) with Al, it is believed that this alloy could be satisfactorily modified by the addition of around .15 Al. However, both D-183 and R-71 without Al probably have sufficient oxidation

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resistance for their intended maximum operating temperature of around 1200°F.

Commercial production of D-183 alloy containing B or B+Al additions has not been tried. Melting of a 1000-pound heat of either of these alloys would be the next step in investigating the mechanical properties and hot workability problems which may be encountered in producing large ingots of this material.

The results of three fatigue tests in a pneumatic fatigue machine indicated that the fatigue strength at 10^8 cycles is 38,000 psi at 1200°F for D-183 after a heat treatment of 1 hour at 1900°F, water quenched plus 16 hours at 1300°F and then air cooled. This compares favorably with those of other Fe-base superalloys.

Wrought Co-Base Alloys

The chemical compositions of the thirty-three heats made in the study of wrought Co-base alloys are given in Table XI. Total Cb+Ta is reported. The ratio of Cb to Ta was about 2.5 to 1. Results of rupture, oxidation and thermal shock tests on these alloys are given in Tables XII, XIII, and XIV. The important results obtained are presented graphically.

The effect of C on rupture life of R-27 at two test conditions is shown in Figure 14. An optimum rupture life is indicated at C contents of .10 to .15. This range was used in the remainder of the alloys investigated.

The inherent oxidation resistance of R-27 containing 10 Cr was poor. Part of the present program was aimed at improving this oxidation resistance without impairing the outstanding rupture

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properties of R-27. Raising the Cr to 15 and 20 was tried first with success. The effect of Cr on rupture strength at three test temperatures is shown in Figure 15. At 1500°F, increasing the Cr impairs the rupture life, but an optimum rupture life was obtained for 15 Cr at 1600 and 1700°F. It will be shown later that this 5 per cent increase in Cr was sufficient to raise the oxidation resistance to a level equivalent to that of other superalloys. There was an increase in rupture ductility with increasing Cr content at all three temperatures as shown in Figure 15. This increase was relatively greater at temperatures above 1500°F. The ductility level, however, decreased with increasing test temperatures. The alloy designation R-94 has been given this R-27 modification with improved oxidation resistance.

In the above alloys Cr replaced Co. In two other modifications, 5 Cr replaced 5 Mo (Heat R-96) which lowered rupture properties and 5 W (Heat R-95) which gave the same outstanding rupture life as did R-27. R-95 had relatively low rupture ductility, however, so it was concluded that a Cr replacement of Co offered the most promise for improving oxidation resistance of R-27 without impairing rupture properties.

In a series of R-27 type alloys with variable Cb+Ta, the optimum rupture properties were obtained at 1 Cb+Ta as shown in Figure 16. Cb+Ta variations were also made in a low C (.04 per cent) modification of R-17. Figure 17 shows that rupture life at 1600°F appeared to approach an optimum at 2 Cb+Ta. These heats were made to investigate the relation of Cb+Ta/C ratio to rupture strength. Work under a previous contract had indicated the

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optimum ratio to be approximately 10, as is the case in standard S-816 alloy. The results in Figure 17 indicate an optimum ratio of around 50. It is noted that rupture lives are very short in these latter tests. However, it appears that the rupture strength of this type alloy is more affected by the individual variations of these elements rather than by the Cb+Ta/C ratio.

The addition of B to an R-27 modification was initially made in Heats R-115 and R-116. It was discovered after heat treating rupture pieces of these heats at the usual solution temperature of 2250°F that intergranular melting had occurred. A metallographic study was then made of Heats R-117, R-118, and R-119 containing respectively .05, .10, and .20 B. These were solution treated at temperatures ranging from 2050 to 2250°F in 50°F increments. Microstructural results of this study were similar to those for D-183 alloy, which are shown in Figure 12, except that the massive grain boundary phase indicating incipient melting occurred at 2200°F. The amounts of this phase at a given solution temperature increased with increasing B content. At 2150°F there was no indication of this phase. This temperature was then used for heat treating all of the rupture specimens tested from B-containing heats.

In Figure 18 are shown the results of rupture tests on B modifications of R-94 alloy. It should be noted that the B modifications shown were solution treated at a lower temperature than the R-94 alloy. The results in Figure 18 indicate an appreciable decrease in rupture life and a large increase in rupture ductility with increasing B. The conclusion is therefore drawn

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that B up to .20 per cent is an impractical addition to this wrought alloy. Higher B additions would involve serious hot working problems.

The results of oxidation tests on R-27 are compared with those for S-816, V-36, and X-40 alloys in Figure 19. The poor resistance of R-27 is noted. Also included are results for the improved R-94, the modification with 15 Cr. This effect of Cr additions to R-27 is further shown in Figure 20. Increasing the Cr above 15 per cent, as was done in Heat R-55 (20 Cr), offered little further advantage.

Only a limited thermal shock program was conducted on the wrought alloys. This concerned the Cr modifications. The results of these tests are given in Table XIV and graphically in Figure 21. There was considerable scatter in results for the base R-27 composition. With increasing Cr content from 10 to 20 per cent, there was a large increase in thermal shock life. Also evident was the correlation of this increase in shock life with rupture ductilities which were obtained from tests made at 1600 and 1700°F.

It was noted in work under the previous Air Force contract (1) that certain of the high Co-Ni modifications exhibited a degree of magnetism. Some of the alloys in the present work also exhibited magnetism. Generally this was very low but could be picked up with a powerful permanent magnet on a sample suspended on a thread as a pendulum. Increasing amounts of Cr, Cb, Mo, and W have appeared to suppress the magnetic response, while increasing C, Ni, and Co tend to enhance this response. At 15 Cr and above there was little magnetism found in any of the alloys.

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It is noted that the observed magnetic response to composition changes is opposite to the effects of these elements generally attributed to such changes in lower alloy austenitic-type steels. However, in these Co-base alloys the influence of composition on the Curie temperature of the basic alloy is believed to be mainly responsible for the magnetic changes.

CAST ALLOYS

The initial series of twelve cast Co-base modifications melted under this program were of the R-17 base analysis (.40 C, 1.2 Mn, .4 Si, 20 Cr, 20 Ni, 4 Mo, 4 W, 4 Cb+Ta, balance Co). These included C and Cb+Ta variations. Only limited work was done on these alloys because emphasis was shifted to R-27 modifications to parallel the wrought Co-base alloy program. Alloy modifications of R-27 included variations in C, Cr, Mo, and W with additions of B. Thirty-one heats were precision cast in this latter program. Rupture and thermal shock properties were the main evaluation criteria on the cast alloys.

Cast Alloy Procedures

Melting and Casting

All melts were made with virgin materials in 17-pound magnesia crucibles using a 40 KW Ajax induction furnace. The pouring temperatures were controlled at around 2750°F by using a silica covered Pt, Pt-Rh immersion thermocouple attached to a high-speed electronic recorder. Four of the heats tested (F-456, F-457, F-458, and F-459) were off analysis (low Cb+Ta and high W) due to an off-analysis lot of Fe-Cb-Ta.

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The precision casting of rupture, tensile, and thermal shock specimens was done by the conventional lost wax or investment method. Six thermal shock and six rupture test specimen wax patterns were assembled on a sprue. They were then encased in a steel mold can which was filled with a hydrolized silicate bonded material. The molds were de-waxed by steaming and then held for further de-waxing in an oven at 300°F. The molds were heated (200°F per hour) to 1600°F prior to casting. The castings were allowed to cool in air to room temperature prior to shake out. A total of twelve each of thermal shock and rupture samples were cast per heat. The sand blasted test bars were visually inspected for casting defects and the threads recut with a steel die.

Testing

As-cast rupture and tensile test specimens had a .252-inch diameter and a 2-inch gage length test section. The fractures were checked after tests to determine if the bars had internal defects which may have affected results. Generally, the specimens were found to be free from defects.

Rupture, tensile, and thermal shock testing procedures were the same as described under the Wrought Alloy section of this report.

Cast Co-Base Alloy Results and Discussion

The analyses of the cast heats investigated are shown in Table XV. The stress-rupture data are given in Table XVI. The room temperature tensile and thermal shock properties are in Tables XVII and XVIII respectively.

During the first part of this program the investigation was

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concerned with R-17 and its modifications. Three heats of the base alloy were made to determine the rupture, room temperature tensile, and thermal shock properties. From these heats (F-427, F-428, and F-456) it was determined that good reproducibility of rupture properties could be obtained on the as-cast, R-17 Co-base alloy. Limited data in Table XVIII indicate that thermal shock life is affected adversely if the cast samples are only partially ground or not ground at all prior to testing. The rupture life at 1600°F and 20,000 psi stress and the thermal shock life at 1800°F are included in the tabular presentations and are also shown graphically in Figures 22 and 23. The effects of the C content on the rupture properties and thermal shock life are also shown in Figures 22 and 23. It is seen that increasing the C content above .4 per cent continually increased the rupture life at 1600°F and 20,000 psi stress with a decrease in the rupture ductility. Below .4 C the rupture life sharply decreased as the elongation increased. The tensile strength (Table XVII) remained fairly constant with C variations while the tensile elongation decreased markedly with the increasing C content. From Figure 23 it can be seen that the maximum thermal shock life was obtained on the base alloy containing .4 C. One heat (F-474) was made with a 30 Cr variation upon which tests indicated much lower rupture and room temperature tensile properties. Since these properties were so low, thermal shock tests were not conducted on this alloy.

At this point in the program, the emphasis was changed from R-17 to a cast R-27 alloy which was the most promising wrought

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alloy. Two heats of the base R-27 alloy were cast to determine the as-cast properties for comparison with the wrought R-27 alloys and the cast R-17 alloys. The good rupture properties found for R-27 alloy in the wrought condition were not exhibited by the alloy when in the cast condition. It had such low rupture, room temperature tensile, and thermal shock properties that alloy modification was necessary to improve these properties. The first modification was that using C additions up to 1.5 per cent. These results are shown graphically in Figures 22 and 23. The best combination of rupture and thermal shock properties were obtained on an R-27 type alloy containing .7 C. The yield strength of this alloy was better than that of R-27, but there was a large decrease in the tensile ductility. Further increases of C lowered the rupture, thermal shock, and room temperature tensile properties. Increasing the C content of R-27 markedly increased the magnetic response of this alloy.

Increasing the C content in a 10 Mo - 5 W modified R-27 alloy, also shown in Figure 22, indicated that higher rupture and tensile properties could be obtained at around .4 C instead of at .7 C, the optimum for R-27 alloy. Thermal shock data were not obtained on the three 10 Mo - 5 W alloys.

For some alloys there was a considerable variation of thermal shock lives in check tests. This broad spread is not considered typical for results in thermal shock tests. However, the trends in shock life with composition which are shown in the graphs are believed to be significant.

The relation of the Cb+Ta/C ratio on the rupture life at

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1600°F and 20,000 psi stress shown in Figure 24 indicated that optimum rupture properties were obtained at ratios of 1.3 for the R-27 type alloys and around 3 for the R-17 and the 10 Mo - 5 W R-27 type alloys. The Cb+Ta/C ratio was varied by varying the C content in the three alloy systems.

The necessity of increasing Cr in R-27 for improving oxidation resistance and decreasing the magnetic response was pointed out in the section on wrought Co-base alloys. Increased Cr to 15 per cent in the cast alloy reduced the rupture life, as shown in Figure 25, and slightly increased the thermal shock life. Cr was also raised in the more promising .7 C alloy. The optimum combination of rupture, room temperature tensile, and thermal shock properties was obtained on a .7 C - 20 Cr modification. The 1600°F rupture life gradually increased to an optimum at 20 Cr. The 25 Cr alloy had rupture properties similar to those of the 10 Cr alloy. The thermal shock life greatly decreased with an addition of 15 Cr, then increased continually with increasing Cr to 25 per cent where it surpassed the 10 Cr alloy. This material was magnetic with up to 15 Cr but non-magnetic at the higher Cr levels. A better combination of rupture and thermal shock properties and higher rupture properties was obtained on the .7 C - 20 Cr alloy (F-538) than on any other alloy studied during this period.

As shown in Figure 26, the 1600°F rupture life of R-27 increased to an optimum with B additions of up to .6 per cent, above which the rupture life gradually decreased. The thermal shock life and room temperature tensile strengths did not

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vary greatly with the increasing B modifications although the tensile ductility decreased sharply with increasing B. B additions up to 1.0 per cent did not improve any of the properties of a .6 C, 16 Cr, 4 Mo, 4 W system studied, as can be seen in Figure 26. None of these properties are equal to those of the higher Cr alloy discussed above.

The limited data presented in Figure 27 indicate that increasing the W from zero to 15 per cent did not appreciably affect the 1600°F rupture life in the 5 and 10 Mo variations of the 20 Cr, .4 C R-27 series. The thermal shock life of this series did increase slightly with increasing W. One heat (F-599) of a 0 Mo, 20 Cr, .4 C, 10 W R-27 alloy indicated that the rupture, tensile, and thermal shock properties are superior to those of the 5 or 10 Mo series.

To determine the effect of heat treatment on the as-cast rupture properties, test bars from eight different heats of R-27 with the highest rupture life were heat treated and rupture tested. The heat treatment used was 1 hour at 2150°F water quenched solution treatment plus 16 hours at 1400°F air cooled aging treatment. Generally the rupture tests on the heat treated bars indicated that their rupture properties were equal to or slightly lower than the as-cast rupture properties.

From microstructures shown in Figure 28, it can be seen that generally the thermal shock failures were transgranular. Occasionally, the crack will follow a grain boundary when these grain boundaries are perpendicular to the test edge. If the grain boundary does not continue perpendicular to the test edge,

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then the thermal shock crack will proceed transgranularly. When there are massive carbides in the structure due to .7 C additions, the thermal shock fracture will follow the carbides perpendicular to the test edge. Increasing the C content from 0.1 to 0.7 in R-27 alloy resulted in a marked increase in excess constituents in the cast structure. There was also a slight increase in the excess constituents with Cr additions.

CONCLUSIONS

Given below are the conclusions indicated by the results of these investigations.

Wrought Fe-Base Alloys

On wrought Fe-base alloys of the D-183 type:

1. The optimum heat treatment for D-183 alloy is 1 hour at 1900°F water quenched, plus 16 hours at 1300°F and then air cooled.
2. Cold working and hot-cold working of D-183 raise the rupture life considerably but at the expense of ductility. In view of the high notch rupture sensitivity with low ductilities, neither of these methods of working are considered satisfactory for the alloy.
3. An outstanding increase in the rupture properties of D-183 can be obtained with a B addition of .20 per cent. The optimum heat treatment for this modification is 1 hour at 2050°F water quenched, plus 16 hours at 1300°F and then air cooled. This alloy has been designated R-71. It has a 100-hour rupture strength of 67,000 psi at 1200°F.

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4. A large increase in the oxidation resistance of D-183 can be obtained with Al additions of .10 to .20 per cent.

Wrought Co-Base Alloys

On wrought Co-base alloys of the R-27 type:

1. An optimum C content is .10 to .15 per cent.
2. An optimum Cb+Ta content is 1.0 per cent in standard R-27 alloy.
3. Raising the Cr content from 10 to 15 per cent improves the oxidation resistance of R-27 to a level with other 20 Cr high temperature alloys without impairing the 1600 and 1700°F rupture properties. R-94 alloy at this 15 Cr level possessed an optimum combination of properties in this investigation.
4. Modifications of R-94 containing B up to .20 per cent have lower rupture strength and high rupture ductility at test temperatures of 1600 and 1700°F.
5. Thermal shock life of wrought R-27 is raised by increasing the Cr content from 10 to 20 per cent. The shock life appears to correlate with an increase in rupture ductility.

Cast Co-Base Alloys

On cast Co-base alloys of the R-17 and R-27 types:

1. Increasing C content up to 1.4 per cent in R-17 alloy continually increases the 1600°F rupture life and decreases the rupture ductility and thermal shock life.

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2. An optimum C content for R-27 alloy is .7 per cent for the best rupture and thermal shock life.
3. The optimum C content for the 10 Mo - 5 W modification of R-27 alloy is .4 per cent.
4. The optimum Cb+Ta/C ratio for the cast Co-base alloys studied is between 1.3 and 3.2.
5. The highest rupture and thermal shock lives were obtained on a .7 C - 20 Cr R-27 alloy, F-538.
6. An optimum B content for R-27 is .6 per cent. B additions to the 10 Mo - 5 W R-27 modification did not improve the properties.
7. No appreciable changes in rupture and thermal shock properties were noted with the limited Mo and W modifications.
8. No improvement was noted in rupture life resulting from a solution and aging heat treatment.

REFERENCES

1. R. R. MacFarlane, R. S. DeFries, E. E. Reynolds, and W. W. Dyrkacz. Research and Development of Wrought and Cast High Temperature Alloys. Final Report U.S.A.F. Contract No. AF 18(600)-149. June 1954. WADC TR 54-276.

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TABLE I

CHEMICAL ANALYSES OF D-183 ALLOY MODIFICATIONS

(Major modification underlined)

Chemical Analysis (%)

Heat	C	Mn	Si	Cr	V	N	Mo	Others
Base D-183	.30	18.00	.30	12.00	.75	.18	3.0	
9X-129	.29	18.06	.30	12.48	.70	.156	2.97	
R-57	.26	17.38	.18	12.18	.72	.23	2.96	Al- .12
R-58	.31	17.30	.37	12.10	.74	.22	2.98	Al- .21
R-68[a]	.27	17.80	.23	12.32	.80	.192	3.17	B - .014
R-69	.28	17.80	.13	12.16	.75	.168	3.10	B - .041
R-70	.32	18.24	.20	12.24	.84	.164	3.28	B - .086
R-71	.28	18.45	.18	12.34	.74	.180	3.20	B - .198
R-72	.31	17.55	.32	12.30	.66	.128	3.13	B - .437
R-81	.09	17.28	.40	12.38	.68	.168	--	B - .081
R-82	.09	17.49	.44	12.36	.63	.183	--	B - .153
R-83	.13	16.45	.29	12.64	.73	.171	1.38	B - .081
R-84	.27	17.82	.34	12.62	.71	.144	1.39	B - .081
R-85	.26	18.26	.19	12.84	.66	.216	1.01	Al-2.12
R-86	.30	18.08	.37	12.27	.66	.235	2.72	Al- .92
R-87	.32	18.11	.35	12.38	.53	.225	2.74	Al-2.22
R-102	.27	17.69	.25	12.42	.75	.220	3.13	B - .05
R-103	.27	17.97	.16	12.18	.77	.188	2.99	B - .09
R-104	.30	18.37	.24	12.60	.80	.190	3.02	B - .18
R-105[b]	.30	18.00	.30	12.00	.75	.18	3.0	B - .4
R-106	.31	19.36	.19	15.93	.95	.266	1.15	Al- .43
R-107	.29	17.02	.06	13.58	.74	.213	1.21	Al-1.11
R-108	.30	17.16	.17	13.66	.68	.221	2.96	Al- .94
R-120	.39	18.21	.13	12.17	.80	.168	3.66	Al- .09
R-121	.39	18.40	.12	12.25	.83	.162	3.29	Al- .13

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a - First heat melted under present contract

b - Cracked during rolling from 2" square to 3/4" round scrapped - melting aim analysis listed

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TABLE II

EFFECT OF HEAT TREATMENT ON RUPTURE PROPERTIES OF D-183 ALLOY

Heat 9X-129

Tested at 1200°F - 55,000 psi

Heat Treatment	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment
1800°F-1 Hr-W.Q.	62	21.8	25.6	255
1800°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.	22	32.8	40.1	286
1850°F-1 Hr-W.Q.	53	10.8	13.8	255
1850°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.	32[a]	30.5	40.5	277
1900°F-15 Min-W.Q. + 1300°F-16 Hr-A.C.	47	28.7	41.5	286
1900°F-30 Min-W.Q. + 1300°F-16 Hr-A.C.	47	23.2	38.4	277
1900°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.	41	18.7	33.4	277
1950°F-1 Hr-W.Q.	5.5	10.1	13.7	207
1950°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.	114	4.5	5.5	255
2000°F-1 Hr-W.Q.	3	13.5	18.5	196
2000°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.	56	4.1	6.7	248

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a - Test accurate to ± 8 hours

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TABLE III

EFFECT OF COLD WORKING AND HOT-COLD WORKING ON THE ROOM TEMPERATURE TENSILE AND HARDNESS PROPERTIES OF D-183 ALLOY

Heat 9X-129

Prior Treatment - 1900°F-1 Hr-W.Q.

Heat Treatment	Tensile Strength (psi)	Yield Strength (psi) .2% Offset	Yield Strength (psi) .02% Offset	Elong. (%)	Red. of Area (%)	Brinell Hardness
Cold rolled 5%	136,000	95,700	79,400	41.5	53.5	286
Cold rolled 10%	145,700	119,700	94,800	35.0	48.4	311
Cold rolled 0% + 1300°F-16 Hr-A.C.	134,900	79,100	64,300	34.5	28.9	258
Cold rolled 10% + 1300°F-16 Hr-A.C.	148,100	101,000	74,000	22.5	20.3	340
Hot-cold rolled 10% at 1200°F	157,200	134,800	120,300	26.5	42.3	364
Hot-cold rolled 10% at 1300°F	154,700	124,500	112,000	21.5	33.1	340
Hot-cold rolled 10% at 1500°F	150,100	114,000	99,000	24.5	21.6	321

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TABLE IV

EFFECT OF COLD WORKING AND HOT COLD WORKING ON THE RUPTURE PROPERTIES OF D-183 ALLOY

Heat 9X-129

Prior Treatment - 1900°F-1 Hr-W.Q.

Tested at 1200°F - 55,000 psi

Working and Heat Treatment		Rupture Time (Hr)	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment
Cold rolled	0%	22	6.6	10.0	217
Cold rolled	5%	163	4.3	8.1	286
Cold rolled	10%	156	3.1	5.5	311
Cold rolled	0% + 1300°F-16 Hr-A.C.	83	16.3	17.7	241
Cold rolled	5% + 1300°F-16 Hr-A.C.	58	25.6	49.5	321
Cold rolled	10% + 1300°F-16 Hr-A.C.	162	10.2	27.5	340
Cold rolled	20% + 1300°F-16 Hr-A.C.	79	5.8	9.9	375
Cold rolled	30% + 1300°F-16 Hr-A.C.	76	5.5	8.1	387
Hot-Cold rolled	5% at 1200°F	235	5.1	5.2	321
Hot-Cold rolled	10% at 1200°F	217	3.1	6.1	364
Hot-Cold rolled	5% at 1300°F	220	3.6	8.0	321
Hot-Cold rolled	10% at 1300°F	343	4.2	6.0	340

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TABLE IV (Continued)

Working and Heat Treatment	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment
Hot-Cold rolled 5% at 1500°F	187	8.0	26.5	302
Hot-Cold rolled 10% at 1500°F	125	10.6	34.0	321
Hot-Cold rolled 20% at 1500°F	122	17.6	33.6	340

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TABLE V

EFFECT OF COLD WORKING AND HOT-COLD WORKING ON THE RUPTURE PROPERTIES OF D-183 ALLOY

Heat 9X-129

Prior Treatment - 2050°F-1 Hr-W.Q.

Tested **at** 1200°F - 55,000 psi

Working and Heat Treatment		Rupture Time (Hr)		Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment
		Notch Bar	Smooth Bar			
Cold rolled	10%	--	5	4.3	12.8	281
Cold rolled	20%	.3	3.8	3.2	3.2	340
Cold rolled	30%	--	11	.9	1.0	402
Cold rolled	10% + 1200°F-16 Hr-A.C.	--	10	3.2	5.5	286
Cold rolled	20% + 1200°F-16 Hr-A.C.	.5	6.8	6.2	3.0	351
Cold rolled	30% + 1200°F-16 Hr-A.C.	--	17	1.2	3.1	410
Cold rolled	10% + 1300°F-16 Hr-A.C.	--	38	3.7	8.2	311
Cold rolled	20% + 1300°F-16 Hr-A.C.	11	48	.9	4.0	351
Cold rolled	30% + 1300°F-16 Hr-A.C.	--	134	1.1	3.4	387
Cold rolled	10% + 1400°F-16 Hr-A.C.	--	37	20.2	40.5	298
Cold rolled	20% + 1400°F-16 Hr-A.C.	66	107[a]	4.3	12.2	340
Cold rolled	30% + 1400°F-16 Hr-A.C.	--	19	2.6	7.3	387

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a - Time accurate to ± 8 hours

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TABLE V (Continued)

Working and Heat Treatment	Rupture Time (Hr)		Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment
	Notch Bar	Smooth Bar			
Hot-cold rolled 10% at 1200°F	--	3	2.1	9.1	312
Hot-cold rolled 20% at 1200°F	b	1.8	.7	2.9	364
Hot-cold rolled 30% at 1200°F	--	2.3	1.4	5.1	402
Hot-cold rolled 10% at 1500°F	--	123	1.0	1.5	316
Hot-cold rolled 20% at 1500°F	13	70	6.4	8.7	321
Hot-cold rolled 30% at 1500°F	--	67	3.5	5.0	340

---

b - Test broke on loading

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TABLE VI

ROOM TEMPERATURE TENSILE PROPERTIES OF D-183 AND ITS MODIFICATIONS

Heat	Tensile Strength (psi)	Yield Strength (psi) 0.2% Offset	Yield Strength (psi) .02% Offset	Elong. (%)	Red. of Area (%)
1900°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.					
9X-129	135,100 136,800 134,900	74,800 81,600 79,100	60,400 62,700 64,300	37.5 33.5 34.5	36.6 23.4 28.9
R-68[a]	135,900	72,100	55,000	38.5	33.2
R-69	135,000	72,100	56,400	41.0	45.7
R-81	113,200	47,900	41,800	52.5	46.6
R-82	112,400	45,900	38,700	38.0	33.8
R-83[b]	107,500	57,800	51,200	27.5[c]	9.3[c]
R-84[b]	111,600	58,300	39,200	17.5	21.0
R-85	93,900	42,700	17,300	22.0	24.5
R-86[b]	99,400	42,600	22,900	16.5[c]	16.7[c]
R-87	107,600	60,400	21,600	2.0	1.6
R-102	134,300	69,700	57,300	31.5	27.5
R-103	130,800	67,300	57,300	38.5	32.0
R-104	131,200	69,500	53,900	30.5	24.1
R-106	100,200	50,000	27,800	11.5	11.5
R-107	107,200	40,700	26,300	37.5	34.1
R-108	100,800	54,200	28,900	6.0	5.5
R-120	137,000	68,900	60,800	37.0	32.4
R-121	129,900	66,700	47,400	26.0	22.5
2100°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.					
R-104	148,300	98,200	75,200	34.5	40.4

---

a - Center burst discovered in specimen after fracture
b - Seam in specimen discovered after fracture
c - Estimated

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TABLE VII

HIGH-TEMPERATURE TENSILE PROPERTIES OF D-183 ALLOY AND ITS MODIFICATIONS

1900°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.

Heat	Temp (°F)	Tensile Strength (psi)	Yield Strength (psi) 0.2% Offset	Yield Strength (psi) .02% Offset	Elong. (%)	Red. of Area (%)
9X-129	1000	90,300	40,700	30,700	34.7	64.6
	1100	84,000	39,800	26,600	33.8	50.7
	1200	70,300	39,700	30,100	32.9	58.0
	1300	56,800	35,300	22,300	31.6	44.6
	1400	46,600	33,000	23,900	30.7	51.8
	1500	34,000	29,400	22,400	34.2	57.9
R-102	1200	70,800	34,600	24,300	32.0	55.2
R-103	1200	72,800	36,000	27,700	31.1	49.6
R-104	1200	72,000	36,100	25,300	29.3	45.6

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TABLE VIII

RUPTURE PROPERTIES AT 1200°F OF D-183 ALLOY AND ITS MODIFICATIONS

1900°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.

Heat	Stress (psi)	Rupture Time (Hr) Notch Bar	Rupture Time (Hr) Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment Solution	Brinell Hardness After Heat Treatment Solution + Age
R-68	50,000	--	289	9.3	24.0	228	269
	55,000	77	112	10.0	12.8		
	60,000	--	49	8.8	11.9		
R-69	50,000	--	441[a]	8.1	9.1	241	269
	55,000	53	54[a]	4.3	9.4		
	55,000	--	173	12.0	38.9		
	60,000[b]	44	30	8.4	13.7	255	269
	65,000[b]	--	30	12.0	15.4		
R-70	55,000[b]	470	268	12.0	29.7	241	277
	60,000[b]	--	174	10.4	13.9		
	65,000[b]	--	105	11.4	15.4		
R-71	55,000[b]	493	280	25.2	35.0	241	286
	60,000[b]	--	142	18.0	29.4		
	65,000[b]	--	89	8.8	9.4		
R-72	55,000[b]	1481	160	21.6	35.8	255	286
	60,000[b]	--	90	22.5	43.2		
R-81	50,000	--	18[c]	29.0	29.4	202	202
	55,000	7.8	d	32.0	49.0		
	60,000	--	7	37.9	62.4		
R-82	50,000	--	14	22.0	21.6	202	207
	55,000	8.3	d	35.0	54.8		
R-83	50,000	--	.3[a]	13.0	13.1	217	223
	55,000	--	8.5[a]	19.1	11.8		
	55,000[b]	52+[e]	102	15.4	15.4		217
	60,000[b]	--	9	13.2	18.5		

---

a - Center burst discovered in specimen after testing
b - These specimens were machined away from the center of bar stock material
c - Overheated to 1280°F-4 minutes and fractured
d - Broke on loading
e - Threads broke on rupture specimen - test discontinued

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TABLE VIII (Continued)

Heat	Stress (psi)	Rupture Time (Hr) Notch Bar	Rupture Time (Hr) Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment Solution	Brinell Hardness After Heat Treatment Solution + Age
R-84	55,000	3[a]	12.3[a]	13.6	18.6	230	255
	55,000[b]	--	46	13.3	13.9	-	241
	60,000	--	1[a]	19.5	23.1		
	60,000[b]	--	24	18.5	22.4		
R-85	50,000	--	d	18.7	19.3	196	192
R-86	50,000	--	.3	35.2	47.6	207	217
R-102	55,000	--	104	14.8	20.2	248	269
	60,000	57	38	15.0	29.9		
R-103	55,000	--	189	15.0	39.3	255	269
	60,000	266	83	14.0	29.2		
	65,000	88	47	18.7	26.2		
R-104	55,000	--	292	30.0	47.6	255	277
	60,000	559	146	18.5	37.4	-	255
	65,000	179	65	17.0	26.0		
	65,000	--	32	14.7	16.6		
R-106	55,000	3.8	.5	35.9	38.4	228	286
R-107	50,000	--	.5	46.0	52.4	207	212
R-108	50,000	--	.5	39.6	40.5	217	302
R-120	55,000	282	87	20.3	23.9	255	286
R-121	55,000	213	60	26.2	56.5	243	262

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a - Center burst discovered in specimen after testing
b - These specimens were machined away from the center of bar stock material
d - Broke on loading

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TABLE IX

EFFECT OF HEAT TREATMENT ON THE RUPTURE PROPERTIES

AT 1200°F OF R-71 ALLOY (D-183 + .2 B)

Heat R-104

Heat Treatment	Stress (psi)	Rupture Time (Hr) Notch Bar	Rupture Time (Hr) Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment Solution	Brinell Hardness After Heat Treatment Solution + Age
1900°F-1 Hr-W.Q. + 1200°F-16 Hr-A.C.	65,000	51	133[a]	9.8	14.7	255	302
1900°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.	65,000	--	32	14.7	16.6	-	255
1900°F-1 Hr-W.Q. + 1375°F-16 Hr-A.C.	65,000	--	13	26.1	37.4	-	269
2000°F-1 Hr-W.Q. + 1200°F-16 Hr-A.C.	65,000	21/33	452	16.1	27.5	241	302
	75,000	--	87	6.3	8.1	228	293
2000°F-1 Hr-W.Q. + 1250°F-16 Hr-A.C.	65,000	70	--	--	--	-	302
2000°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.	65,000	1255+[b]	94	13.7	40.5	-	302
	75,000	570	--	--	--	-	321
2000°F-1 Hr-W.Q. + 1375°F-16 Hr-A.C.	65,000	--	38[a]	25.4	40.5	-	262
2050°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.	55,000	--	798	29.3	58.3	228	302
	60,000	--	516	20.7	52.0		
	65,000	1036+[b]	198	16.3	45.3		
	70,000	--	43	19.6	48.6		
2100°F-1 Hr-W.Q. + 1300°F-16 Hr-A.C.	65,000	--	292	13.1	32.5	223	293
	75,000	34	108	8.0	13.7	-	302

---

a - Time accurate to ± 8 hours
b - Test discontinued

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TABLE X

OXIDATION TESTS ON D-183 ALLOY MODIFICATIONS

Treatment: As Forged

Test Conditions: 100-hour exposure in still air
Tests were run in duplicate.

Heat	Alloy or Modification	Weight Loss (%) at: 1200°F	Weight Loss (%) at: 1500°F
G-126[a]	16-25-6	.007, .013	.464, .423
81240[a]	A-286	.028, .009	.083, .092, .087, .113
9X-129	D-183 (0 Al)	.130, .130	3.61, 3.48, 3.39, 3.42
R-120	D-183 + .09 Al		.632, .642
R-57	D-183 + .12 Al	.144, .215	.337, .337, .809, .798, .673, .696
R-121	D-183 + .13 Al		.500, .512
R-58	D-183 + .21 Al	.340, .211	.283, .218, .458, .593 .563, .672
R-86	D-183 + .92 Al	.182, .145	.313, .257, .205, .258
R-87	D-183 + 2.2 Al	.077, .079	.159, .167, .133, .143, .128, .122
R-106	D-183 + .5 Al + 1.0 Mo	.115, .096	.204, .154
R-107	D-183 + 1.0 Al + 1.0 Mo	.132, .102	.204, .188
R-85	D-183 + 2.1 Al + 1.0 Mo	.194, .074	.115, .119

---

a - Alloy Compositions

	C	Mn	Si	Cr	Ni	Mo	Ti	V	Al	N
16-25-6	.08	1.32	.69	16.10	25.11	7.05	--	-	-	.166
A-286	.08	1.46	.73	14.83	25.88	1.23	1.96	.38	.19	-

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TABLE XI

CHEMICAL ANALYSES OF WROUGHT CO-BASE ALLOY MODIFICATIONS

(Major modification underlined)

Chemical Analysis (%)

Heat	C	Mn	Si	Cr	Ni	Co[a]	Fe	W	Mo	Cb+Ta
R-27[b]	.10	1.2	.30	10.0	10.0	61	1.0	10.0	5.0	1.0
R-17[c]	.41	.69	.45	20.08	20.56	44.0	2.0	4.03	4.40	3.42
F-489[d]	.47	--	.37	25.26	10.28	Bal	.91	7.18	--	--
R-26	.11	1.08	.29	20.26	21.52	38.7	.97	10.24	5.00	.83
R-27	.11	1.20	.25	10.44	9.26	60.6	.97	10.22	5.00	.87
R-55	.14	1.16	.40	20.20	10.03	50.4	1.30	10.15	5.40	.82
R-73[e]	.02	1.10	.29	20.27	19.90	49.3	.53	4.39	4.19	--
R-74	.03	1.20	.36	20.14	20.07	48.7	.85	3.98	4.22	.46
R-75	.05	1.19	.32	20.04	20.21	47.8	1.14	3.85	4.25	1.13
R-76	.04	1.21	.29	20.04	19.86	47.6	1.30	4.19	4.13	1.35
R-77	.05	1.29	.40	20.09	20.07	46.3	1.58	4.04	4.11	2.04
R-78	.38	1.22	.28	19.51	19.94	50.0	.78	3.81	4.10	--
R-79	.39	1.29	.34	20.36	20.19	45.5	1.58	4.11	4.22	2.01
R-80	.39	1.25	.66	20.05	20.03	40.5	3.13	3.90	4.08	5.97
R-88	.12	1.15	.14	9.43	9.60	64.1	.59	9.95	4.88	--
R-89	.13	1.16	.36	9.95	10.02	62.5	.67	9.97	4.35	.32
R-90	.13	1.20	.34	9.90	9.95	62.2	.78	9.96	5.11	.47
R-91	.14	1.11	.39	9.83	10.02	61.4	1.30	10.02	5.08	.75
R-92	.12	1.25	.36	9.96	9.86	61.0	.98	10.21	5.21	1.09
R-93	.13	1.24	.32	10.18	10.37	59.9	1.32	10.17	5.28	1.12
R-94	.13	1.10	.36	14.95	10.15	56.7	.91	9.82	5.03	.83

---

a - By difference
b - Base analysis of R-27 alloy
c - Standard S-816 alloy
d - Standard X-40 alloy as cast - used for oxidation tests shown in Figure 19
e - First heat melted under present contract

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Contrails

Contrails

TABLE XI (Continued)

Heat	C	Mn	Si	Cr	Ni	Co[a]	Fe	W	Mo	Cb+Ta	B
R-95	.13	1.13	.38	15.20	10.12	61.4	.93	4.63	5.20	.90	--
R-96	.12	1.16	.44	14.88	10.10	60.9	1.17	10.07	--	1.13	--
R-97	.22	1.08	.28	10.10	10.20	61.4	.93	10.05	5.00	.72	--
R-98	.28	1.28	.60	9.98	10.20	61.1	1.19	9.79	5.00	.63	--
R-99	.05	1.26	.44	9.84	10.18	62.1	.80	9.94	5.00	.35	--
R-100	.14	1.23	.37	9.96	10.42	61.1	1.06	9.80	5.00	.92	--
R-101	.14	.89	.27	9.98	9.99	61.9	1.17	10.03	4.80	.86	--
R-109[f]	.10	1.2	2.0	10.0	10.0	Bal	4 max	10.0	5.0	1.0	--
R-110	.09	1.29	.38	9.74	10.08	66.6	1.16	4.17	4.17	1.15	--
R-111	.10	1.25	.40	10.06	10.24	60.1	1.29	10.18	5.10	1.31	--
R-112	.10	1.26	.32	9.94	10.30	59.2	1.45	10.24	5.00	2.15	--
R-113	.09	1.20	.40	14.92	10.24	62.6	1.16	4.19	3.94	1.26	--
R-114	.03	1.26	.42	10.08	10.13	61.1	1.03	10.72	4.24	1.01	--
R-115	.11	1.17	.48	15.20	10.12	62.1	2.08	3.77	4.12	.79	.10
R-116	.13	1.22	.32	15.08	10.14	61.6	2.11	4.04	4.00	1.15	.217
R-117	.12	1.08	.30	14.82	10.05	56.3	1.06	10.19	5.00	1.02	.054
R-118	.11	1.12	.36	15.06	10.07	56.2	1.00	10.27	4.62	1.11	.104
R-119	.11	1.16	.33	15.16	10.04	56.7	1.03	10.13	4.34	.83	.208

---

a - By difference

f - Cracked in forging to 2 inches square scrapped - melting aim analysis listed

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Contrails

TABLE XII

RUPTURE AND HARDNESS PROPERTIES OF R-27 ALLOY MODIFICATIONS

Heat Treatment: 2250°F-1 Hr-W.Q. + 1400°F-16 Hr-A.C. unless otherwise indicated

Heat	Temp (°F)	Stress (psi)	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment: Solution	Brinell Hardness After Heat Treatment: Solution + Age
R-27	1500	25,000	1028	19.3	26.5	255	300
	1600	25,000	24	28.9	29.2		
	1700	10,000	274	7.6	15.4		
	1800	10,000	37	7.0	17.3		
R-55	1500	25,000	195	27.6	54.4	277	311
	1600	18,000	112	43.1	48.6		
	1700	10,000	311	23.2	32.0		
	1800	10,000	35	28.5	51.5		
R-73	1600	20,000	1.3	8.7	14.9	192	192
R-74	1600	20,000	2.8	14.3	15.3	196	192
R-75	1600	20,000	7.8	19.9	25.2	196	207
R-76	1600	20,000	7.3	17.7	21.6	187	207
R-77	1600	20,000	13	27.0	29.0	196	217
R-88	1600	18,000	13	11.2	12.7	228	228
	1700	15,000	3.5	11.5	12.5		
R-89	1600	18,000	91	15.0	23.7	228	235
	1700	15,000	20	18.5	21.6		
R-90	1600	18,000	228	13.3	20.2	228	255
	1700	15,000	61	10.3	21.3		

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Contrails

TABLE XII (Continued)

Heat	Temp (°F)	Stress (psi)	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment: Solution	Brinell Hardness After Heat Treatment: Solution + Age
R-91	1600	18,000	308	10.6	18.2	223	286
	1700	15,000	91	7.1	18.8		
R-92	1600	18,000	250	4.4	12.3	230	277
	1700	15,000	77	13.1[a]	13.4[a]		
	1800	10,000	69	5.0	11.0		
R-93	1600	18,000	425	13.6	24.6	235	293
	1700	15,000	97	24.5[a]	27.1[a]		
	1800	10,000	89	6.4	16.9		
R-94	1500	25,000	468	22.0	25.0	223	277 311
	1600	18,000	464	15.3	16.0		
	1600	25,000	29	15.5	21.6		
	1700	10,000	749	8.6	14.4		
	1700	15,000	77	18.3	22.6		
	1800	10,000	42	26.7	31.0		
R-95	1600	18,000	396	9.2	13.4	228	269
	1600	25,000	33	4.7	12.0		
	1700	15,000	134	5.4	11.5		
R-96	1600	18,000	314	9.1	13.7	241	269
	1600	25,000	24	15.0	32.2		
	1700	15,000	59	5.5	10.8		
R-97	1600	18,000	248	7.6	15.3	248	321
	1700	15,000	54	--	14.5		
	1700	15,000	46	6.6	17.6		
R-98	1600	18,000	148	8.0	15.9	269	321
	1700	15,000	54	9.8	14.5		
	1700	15,000	42	10.4	14.0		

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a - Estimated

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Contrails

TABLE XII (Continued)

Heat	Temp (°F)	Stress (psi)	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment Solution	Brinell Hardness After Heat Treatment Solution + Age
R-99	1600	18,000	34	16.4	19.1	228	223
	1700	15,000	11[b]	17.1	19.6		
R-100	1500	30,000	83	9.2	18.9	223	286
	1600	25,000	37	23.5	24.5	235	286
	1700	15,000	80	7.3	16.3		
	1800	6,000	248	8.0[a]	31.8[a]		
R-101	1500	30,000	131	14.2	16.3	223	286
	1600	18,000	313	10.9	16.7	248	286
	1700	15,000	89	10.5	23.4		
	1800	10,000	58	15.2	23.4		
R-110	1500	25,000	100	20.8	28.6	255	277
	1600	18,000	40	10.5	16.2		
	1700	10,000	69	10.3	15.4		
R-111	1600	18,000	126	12.5	27.2	241	290
	1700	15,000	26	16.9	28.6		
R-112	1600	18,000	177[c]	22.6	25.0	241	321
	1700	15,000	25	32.8	30.4		
R-113	1600	18,000	104	19.6	20.5	241	277
	1700	15,000	22	19.1	19.5		
R-114	1600	18,000	77	23.4	26.0	255	255
	1700	15,000	7[c]	14.1	9.8		
	1800	10,000	11	28.6	26.5		

---

a - Estimated
b - Time accurate to ± 7 hours
c - Center burst discovered in material after fracture

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Contrails

TABLE XII (Continued)

Heat	Temp (°F)	Stress (psi)	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)	Brinell Hardness After Heat Treatment	
						Solution	Solution + Age
R-115[d]	1600	25,000	8	30.6	43.5	---	286
	1700	15,000	33	34.5	48.5		
R-116[d]	1600	25,000	14	40.5	51.5	174	302
	1700	15,000	28	31.0	33.7		
R-117[d]	1600	18,000	144	36.9	70.0	293	321
	1700	15,000	49	63.9	75.3		
R-118[d]	1600	18,000	187	82.2	78.0	300	321
	1700	15,000	36	66.3	77.0		
R-119[d]	1600	18,000	186	59.5	73.0	300	321
	1700	15,000	70	46.5	61.2		

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d - These alloys heat treated at 2150°F-1 Hr-W.Q. + 1400°F-16 Hr-A.C. Boron limited the solution treating temperature of these alloys.

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TABLE XIII

OXIDATION TESTS ON WROUGHT R-27, ITS MODIFICATIONS, AND X-40 ALLOY

Treatment: As-Forged (except X-40 - as-cast)

Test Conditions: 100-hour exposure in still air
Tests run in duplicate.

Heat	Alloy or Modification	Weight Loss (%) at: 1600°F	1700°F
R-27	R-27 Alloy (10 Cr)	.653 .509	2.89 3.12
R-94	R-27 + 15 Cr (Cr replaces Co)	.056 .111	.075 .111
R-95	R-27 + 15 Cr (Cr replaces W)	.168 .207	.079 .076
R-96	R-27 + 15 Cr (Cr replaces Mo)	.028 .032	.054 .053
R-55	R-27 + 20 Cr (Cr replaces Co)	.024 .025	.044 .049
R-26	R-27 + 20 Cr + 20 Ni (replacing Co)	.029 .022	.042 .046
F-489	X-40 Alloy	.032 .027	.052 .044

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TABLE XIV

THERMAL SHOCK PROPERTIES OF WROUGHT R-27 AND MODIFICATIONS

Heat Treatment: 2250°F-1 Hr-W.Q. + 1400°F-16 Hr-A.C.

Heat	Alloy or Modification	Cycles to Failure 800-1800°F Test Temperature
R-17	S-816	6137
		5406
R-27	R-27	4906
		2712
R-101	R-27	3400
		1772
R-94	R-27 + 15 Cr	2832
R-55	R-27 + 20 Cr	5958

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TABLE XV

CHEMICAL ANALYSES OF CAST CO-BASE ALLOYS

(Major modifications underlined)

Chemical Analysis (%)

Heat	C	Mn	Si	Cr	Ni	Mo	W	Fe	Cb+Ta	B
R-17 Modifications										
Base	.40	1.25	.40	20.00	20.00	4.00	4.00	4 max	4.00	-
F-427	.40	1.35	.46	20.37	20.17	4.20	4.11	2.15	3.64	-
F-428	.40	1.43	.50	20.78	20.23	4.18	4.28	1.98	4.00[a]	-
F-456[b]	.40	1.20	.40	20.00	20.00	5.10	4.47	4.00	2.02	-
F-435[a]	.10	1.25	.40	20.00	20.00	4.00	4.00	5.00	4.00	-
F-457[b]	.10	1.20	.40	20.00	20.00	5.05	4.23	4.00	2.41	-
F-436[a]	.70	1.25	.40	20.00	20.00	4.00	4.00	5.00	4.00	-
F-459[b]	.70	1.25	.40	20.00	20.00	5.40	4.22	4.00	2.37	-
F-458[b]	1.00	1.20	.40	20.00	20.00	5.25	3.97	4.00	2.11	-
F-467	1.14	1.26	.52	19.28	20.06	3.70	3.84	2.08	3.93	-
F-468	1.35	1.23	.68	19.56	20.12	3.58	4.29	2.10	4.20	-
F-473	1.40	1.23	.60	19.50	20.50	3.94	3.77	1.91	3.34	-
F-474	.42	1.20	.90	30.09	20.11	3.92	3.90	1.82	3.72	-
R-27 Modifications										
Base	.10	1.20	.40	10.00	10.00	5.00	10.00	4 max	1.00	-
F-488	.12	1.14	.30	9.67	10.48	5.10	10.14	.96	.88	-
F-496	.13	1.09	.41	9.98	9.22	4.93	10.13	.17	.74	-
F-497	.45	1.21	.38	10.16	8.57	5.08	10.29	1.24	.73	-
F-498	.74	1.22	.28	9.63	10.16	5.30	10.37	1.11	1.01	-
F-499[a]	1.00	1.20	.40	10.00	10.00	5.00	10.00	4.00	1.00	-

---

a - Aim analysis
b - Aim analysis except for Mo, W, and Cb+Ta

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Contrails

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TABLE XV (Continued)

Heat	C	Mn	Si	Cr	Ni	Mo	W	Fe	Cb+Ta	B
F-523	.99	1.26	.62	9.86	10.20	4.80	10.18	1.80	1.21	-
F-524	1.33	1.40	.66	9.56	10.32	4.56	9.94	.97	1.26	-
F-500	1.40	.55	.30	10.93	9.87	4.98	10.20	2.60	.85	-
F-477	.10	1.28	.31	9.98	10.30	11.00	4.31	1.00	1.29	-
F-478	.42	1.25	.40	10.06	9.95	10.64	4.99	1.84	1.35	-
F-479	.72	1.12	.25	9.62	10.15	11.08	4.96	.96	1.39	-
F-531	.15	1.17	.36	14.76	10.33	5.70	10.07	1.10	.48	-
F-537	.68	1.13	.36	14.74	10.33	4.68	10.03	1.10	.86	-
F-538	.70	1.10	.40	19.52	10.33	4.80	10.03	.97	.87	-
F-539	.69	1.26	.50	24.72	10.14	4.70	9.97	1.03	1.04	-
F-576	.13	1.17	.45	10.12	10.15	4.76	10.15	1.06	.96	.054
F-577[a]	.10	1.20	.40	10.00	10.00	5.00	10.00	4.00	1.00	.100
F-578	.14	1.18	.29	10.20	10.05	4.82	10.26	1.09	.96	.226
F-579	.14	1.18	.36	10.18	10.00	4.88	10.00	1.12	1.00	.453
F-580	.13	1.19	.37	10.12	10.01	4.84	10.00	4.00	.97	.616
F-581	.14	1.08	.50	10.14	10.09	4.48	9.99	1.09	1.14	.779
F-582	.13	1.05	.55	10.22	9.90	4.66	10.03	1.16	1.15	.870
F-550	.63	.99	.37	15.94	9.88	4.03	3.77	3.98	1.06	-
F-551	.63	1.09	.42	15.75	10.30	4.09	3.91	4.04	.95	.308
F-552	.59	.98	.45	16.14	10.28	4.20	4.02	3.89	1.05	.580
F-553	.59	1.08	.45	16.24	10.17	4.14	3.92	3.89	.96	.924
F-599	.43	1.12	.44	19.98	9.99	--	10.01	1.18	1.08	-
F-598	.40	1.14	.42	19.92	9.95	4.72	--	2.98	.98	-
F-600	.42	1.08	.30	19.82	9.91	4.84	15.04	1.15	.87	-
F-597	.43	1.00	.37	20.06	9.91	9.80	4.71	1.19	.93	-
F-601	.42	1.09	.34	20.00	9.13	10.00	15.09	1.25	.92	-

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a - Aim analysis

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Contrails

TABLE XVI

RUPTURE PROPERTIES OF CAST CO-BASE ALLOYS

Test Conditions: 1600°F - 20,000 psi (unless indicated)

Heat	Analysis Variations	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)
R-17 Modifications				
F-427	R-17	55	22.1	54.5
		20	30.1	61.1
F-428	R-17	52	30.2	45.0
		56	33.7	54.8
F-456	R-17	65	17.2	23.0
		61	14.3	25.1
F-457	.10 C	9	23.8	57.2
		9	34.6	79.2
F-459	.70 C	92	14.3	24.6
		90	20.8	25.9
		12[a]	27.3	37.8
F-458	1.0 C	143	14.9	24.0
		110	13.9	31.0
		28[a]	16.8	25.0
F-467	1.1 C	194	12.2	26.7
F-468	1.3 C	170	11.7	18.3
		36[a]	11.8	23.2
F-473	1.4 C	200	12.0	25.2
		30[a]	12.7	28.1
F-474	15.0 Cr	10	23.7	42.0
		3[a]	30.0	50.4

---

a - Tested at 1600°F - 25,000 psi

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TABLE XVI (Continued)

Heat	Analysis Variations	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)
R-27 Modifications				
F-488	R-27	26	8.1	14.7
		57	19.0	30.0
F-496	R-27	35	15.0	33.7
		63	26.5	63.5
		14[b]	19.0	59.1
F-497	.45 C	49	16.5	22.8
		54	14.5	25.2
F-498	.74 C	78	12.0	19.5
		104	14.0	20.3
		80[b]	20.5	20.6
F-523	.99 C	79	12.0	19.8
		70	13.0	20.2
		91[b]	14.5	21.7
F-524	1.3 C	74	12.0	15.8
		51	10.0	15.4
F-500	1.4 C	34	24.5	66.0
		33	19.5	50.6
F-477	11.0 Mo - 4.3 W	24	16.7	28.7
		32	22.2	40.1
F-478	.42 C - 10.6 Mo - 5.0 W	108	15.5	23.8
		120	38.2	27.6
		80[b]	19.5	21.6
F-479	.72 C - 11.1 Mo - 4.9 W	86	14.2	21.3
		79	3.2	20.2
		89[b]	12.5	29.2
F-531	14.8 Cr	24	15.0	30.0
		25	18.0	34.0
F-537	.68 C - 14.7 Cr	78	12.0	23.0
		120	14.5	22.9
		126[b]	13.5	36.8

---

b - Specimens heat treated as follows: 2150°F-1 Hr-W. Q. + 1400°F-16 Hr-A. C. prior to testing.

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TABLE XVI (Continued)

Heat	Analysis Variations	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)
F-538	.69 C - 19.5 Cr	204	17.0	33.6
		165	10.5	21.6
		148[b]	12.5	30.0
F-539	.69 C - 24.7 Cr	110	16.0	19.4
		99	9.0	24.4
		68[b]	16.0	23.7
F-576	.05 B	86	21.5	40.0
		69	21.5	39.6
F-577	.10 B	80	34.5	59.0
		49	27.0	46.7
F-578	.23 B	40	38.0	75.0
		92	31.5	40.0
F-579	.45 B	153	36.5	49.1
		71	21.0	56.0
F-580	.62 B	144	21.0	22.7
		126	18.5	28.5
F-581	.80 B	112	11.5	18.7
		130	16.0	26.3
F-582	.87 B	132	17.0	29.2
		94	9.1	26.0
F-550	.63 C - 19.5 Cr - 4.1 Mo - 3.8 W	68	19.5	50.9
		59	20.0	55.0
F-551	.63 C - 15.7 Cr - 4.1 Mo - 3.9 W - .31 B	82	26.0	54.0
		68	31.0	54.0
F-552	.59 C - 16.1 Cr - 4.2 Mo - 4.0 W - .58 B	63	25.2	44.5
		69	25.0	49.7
F-553	.59 C - 16.2 Cr - 4.1 Mo - 3.9 W - .92 B	40	17.0	25.0
		80	31.5	35.2
F-599	.40 C - 20.0 Cr - Mo None	124	10.0	16.6
		62	5.8	16.1

---

b - Specimens heat treated as follows: 2150°F-1 Hr-W.Q. + 1400°F-16 Hr-A.C. prior to testing.

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TABLE XVI (Continued)

Heat	Analysis Variations	Rupture Time (Hr)	Elong. (%)	Red. of Area (%)
F-598	.40 C - 20.0 Cr - W None	52 55	19.0 15.0	37.6 21.7
F-600	.40 C - 20.0 Cr - 15.0 W	68 67	17.0 19.0	25.2 33.2
F-597	.40 C - 20.0 Cr - 10.0 Mo 5.0 W	72 57	24.5 24.5	37.0 47.2
F-601	.40 C - 20.0 Cr - 10.0 Mo 15.0 W	51 58	15.0 21.5	31.5 43.8

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TABLE XVII

ROOM TEMPERATURE TENSILE PROPERTIES OF CAST CO-BASE ALLOYS

Heat	Tensile Strength (psi)	Yield Strength (psi)		Elong. (%)	Red. of Area (%)
		0.2% Offset	.02% Offset		
R-17 Modifications					
F-427	100,500	59,500	42,300	7.0	6.6
F-428	103,100	63,200	46,400	5.5	5.5
F-456	99,300	63,500	45,200	5.5	7.5
F-457	97,300	56,000	45,300	17.0	28.0
F-459	100,000	63,800	40,600	2.0	1.6
F-458	103,000	73,000	47,200	1.5	1.8
F-467	109,000	72,500	41,700	1.5	2.0
F-468	104,000	72,000	35,400	.5	.4
F-473	91,100	71,000	44,800	1.0	.8
F-474	93,800	62,600	43,200	2.0	2.2
R-27 Modifications					
F-496	95,600	62,100	51,100	17.5	28.8
F-497	87,900	74,700	52,000	1.0	3.5
F-498	93,900	80,300	58,000	1.0	1.2
F-523	71,600	--	62,300	1.5	.8
F-524	74,800	--	65,600	1.0	.4
F-500	71,000	--	67,500	--	--
F-477	84,200	61,800	48,600	6.5	14.4
F-478	92,700	81,500	55,800	1.0	.8
F-479	78,600	--	63,500	.5	--
F-531	106,600	63,200	50,300	14.5	16.6
F-537	65,800	--	50,700	--	--
F-538	65,200	--	37,100	--	--
F-539	76,000	--	50,500	.5	.5
F-576	111,500	64,200	50,000	15.0	15.3
F-577	105,700	61,800	47,500	14.5	20.1
F-578	101,000	59,200	40,300	10.0	13.9
F-579	108,900	66,000	37,000	3.5	3.2
F-580	114,500	70,300	43,700	2.5	1.2
F-581	114,500	73,900	42,100	2.0	1.6

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TABLE XVII (Continued)

Heat	Tensile Strength (psi)	Yield Strength (psi)		Elong. (%)	Red. of Area (%)
		0.2% Offset	.02% Offset		
F-582	118,900	75,600	39,900	2.5	1.0
F-550	98,900	67,300	45,500	3.5	2.0
F-551	89,800	68,800	47,100	1.0	.2
F-552	88,700	72,000	42,100	1.5	.8
F-553	113,300	84,300	51,500	1.0	2.0
F-599	100,800	65,900	48,000	5.0	9.5
F-600	98,300	81,300	54,200	1.0	5.8
F-597	115,200	65,700	39,700	8.5	7.7
	88,300	75,600	48,600	0.5	--
F-601	84,200	--	42,400	0.5	--

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TABLE XVIII

THERMAL SHOCK PROPERTIES OF CAST CO-BASE ALLOYS

Heat	Cycles to Failure 800-1800°F Test Temp.			Heat	Cycles to Failure 800-1800°F Test Temp.
R-17 Modifications					
F-456	2747	2062[a]	1337[b]	F-538	1374
	2991	1837[a]			1374
Average	2869	1949	1337	Average	1374
F-457	2009	1337[b]		F-539	2426
	2009	1480[b]			4693
Average	2009	1409		Average	3559
F-459	2746			F-576	1877
	1800				1343
Average	2273			Average	1560
F-458	1132			F-577	2400
	1132				2400
Average	1132			Average	2400
F-473	987			F-578	1396
					1396
				Average	1396
				F-579	1769
					1769
				Average	1769
				F-580	1455
					1843
				Average	1649
R-27 Modifications					
F-496	2364	1583[a]	1397[b]	F-581	2168
	1909	2028[a]	1197[b]		1929
Average	2137	1805	1347	Average	2098
				F-582	2132
					2132
				Average	2132

Specimen surface completely ground unless noted.

[a] - Test edge ground only.

[b] - As-cast.

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TABLE XVIII (Continued)

Heat	Cycles to Failure 800-1800°F Test Temp.
F-497	2675
	1703
	1555
Average	1978
F-498	2870
	1898
	1855
Average	2208
F-499	1219
	1019
Average	1119
F-523	600
	1250
	500
	251
	764
	969
Average	722
F-524	828
	329
	402
	218
Average	444
F-500	443
	500
	253
	1315
	1601
	209
	801
Average	732
F-531	3000
	2000
	2550
	2550
Average	2522
F-550	1743
	3108
	1082
Average	1978
F-551	2230
	2230
Average	2230
F-552	2415
	2415
Average	2415
F-553	1322
	1322
Average	1322
F-599	1699
	1699
Average	1699
F-598	2432
	2487
Average	2459
F-560	760
	4007
	1588
Average	2118
F-597	900
	1190
Average	1045
F-601	2358
	2359
	475
Average	1731
F-537	480
	314
	380
Average	391

Specimen surface completely ground unless noted
a - Test edge ground only
b - As cast

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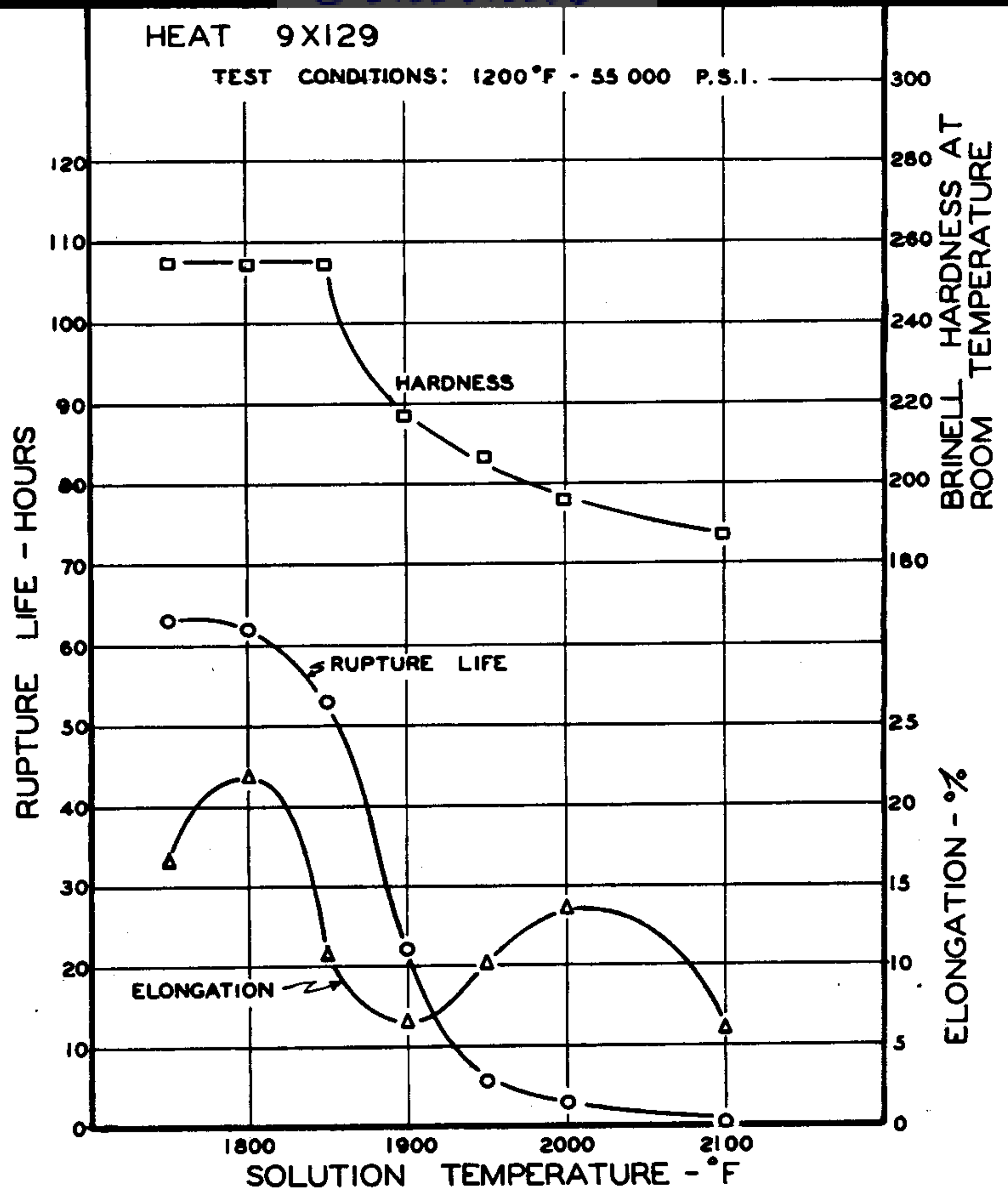

FIGURE 1 - EFFECT OF SOLUTION TEMPERATURE ON PROPERTIES OF D-183 ALLOY

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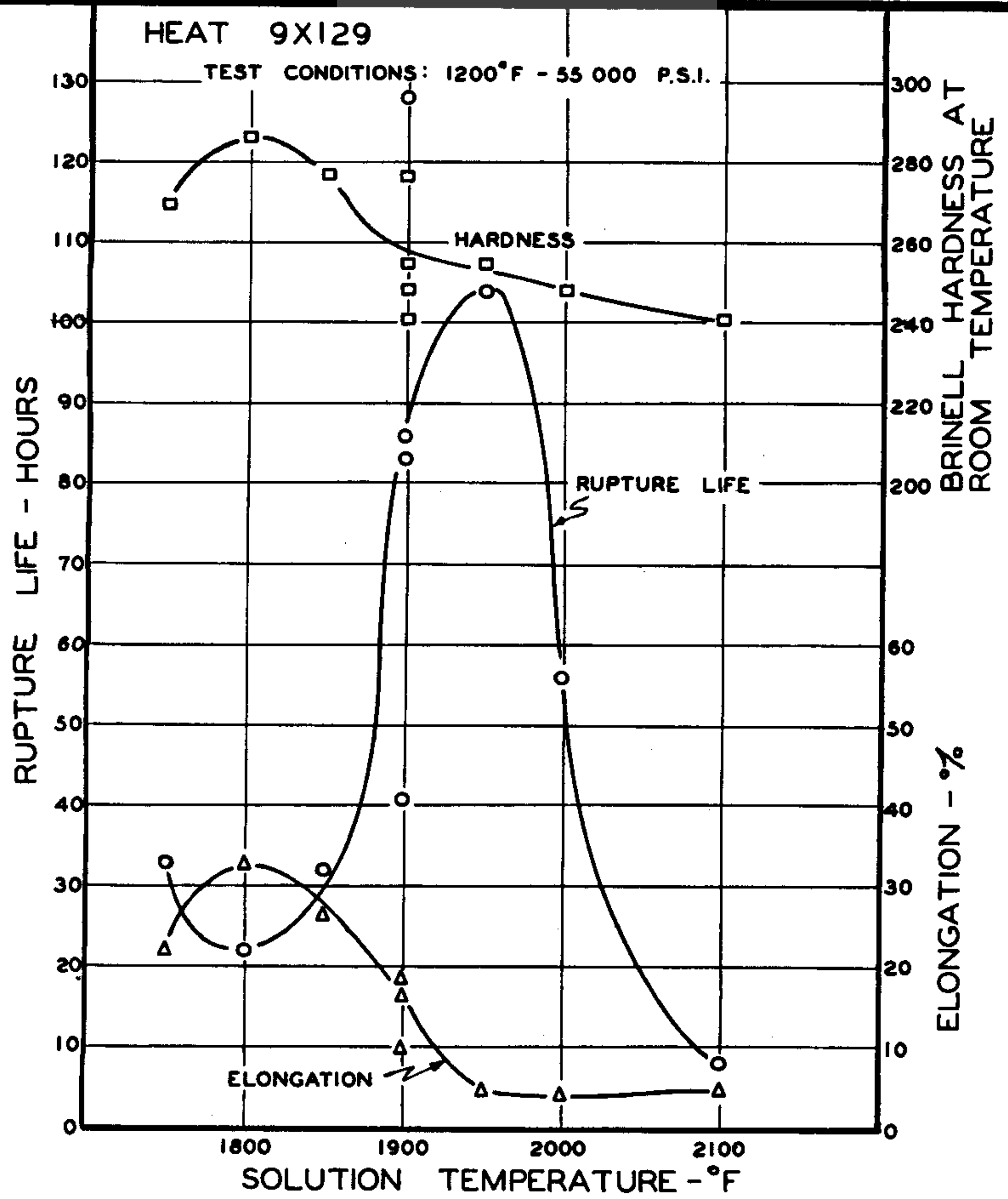

FIGURE 2 - EFFECT OF SOLUTION TEMPERATURE ON PROPERTIES OF D-183 ALLOY - AGED AT 1300°F - 16 HOURS A.C.

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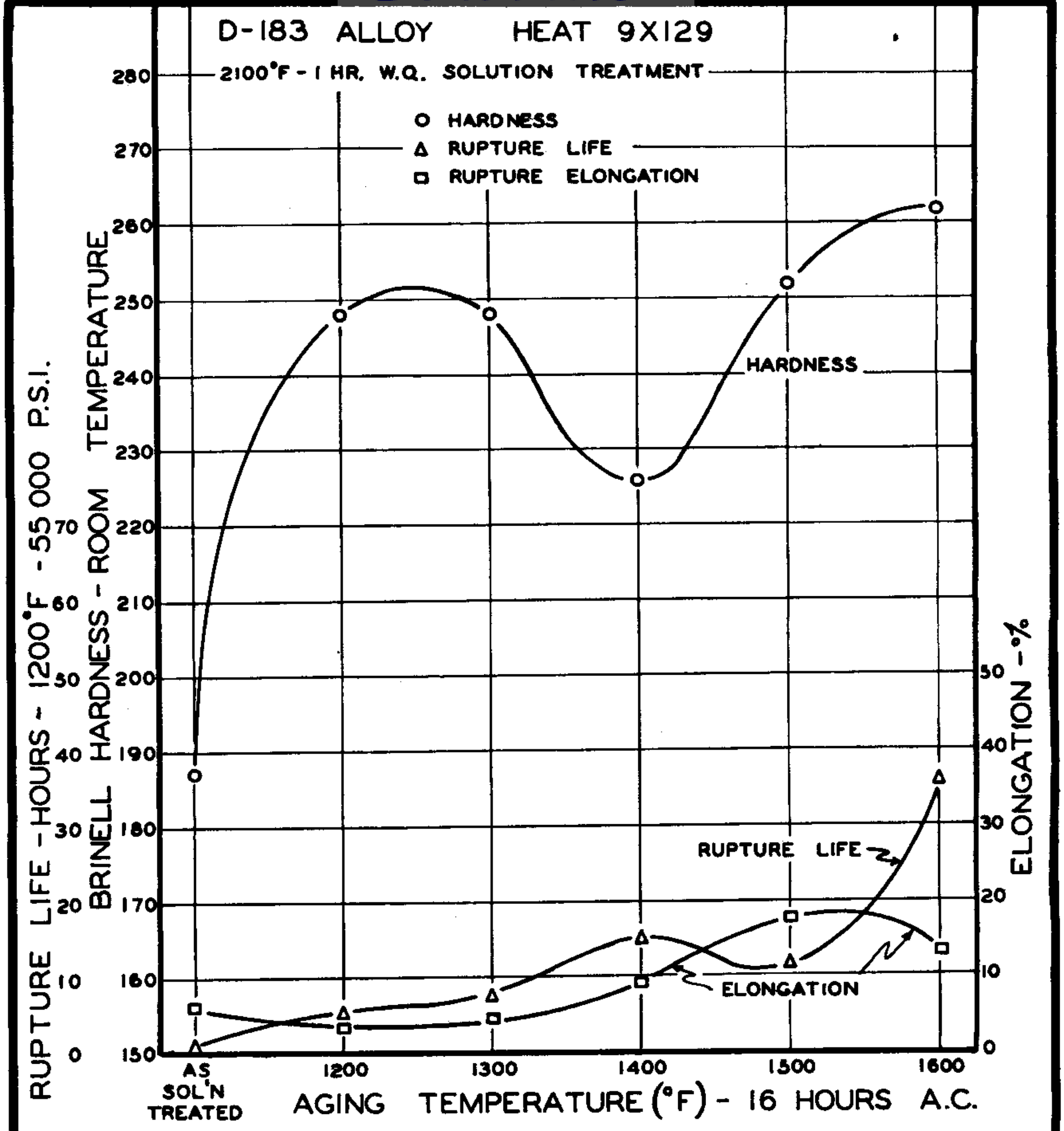

FIGURE 3 - EFFECT OF AGING TEMPERATURE ON PROPERTIES OF D-183 ALLOY - SOLUTION TREATED AT 2100 °F

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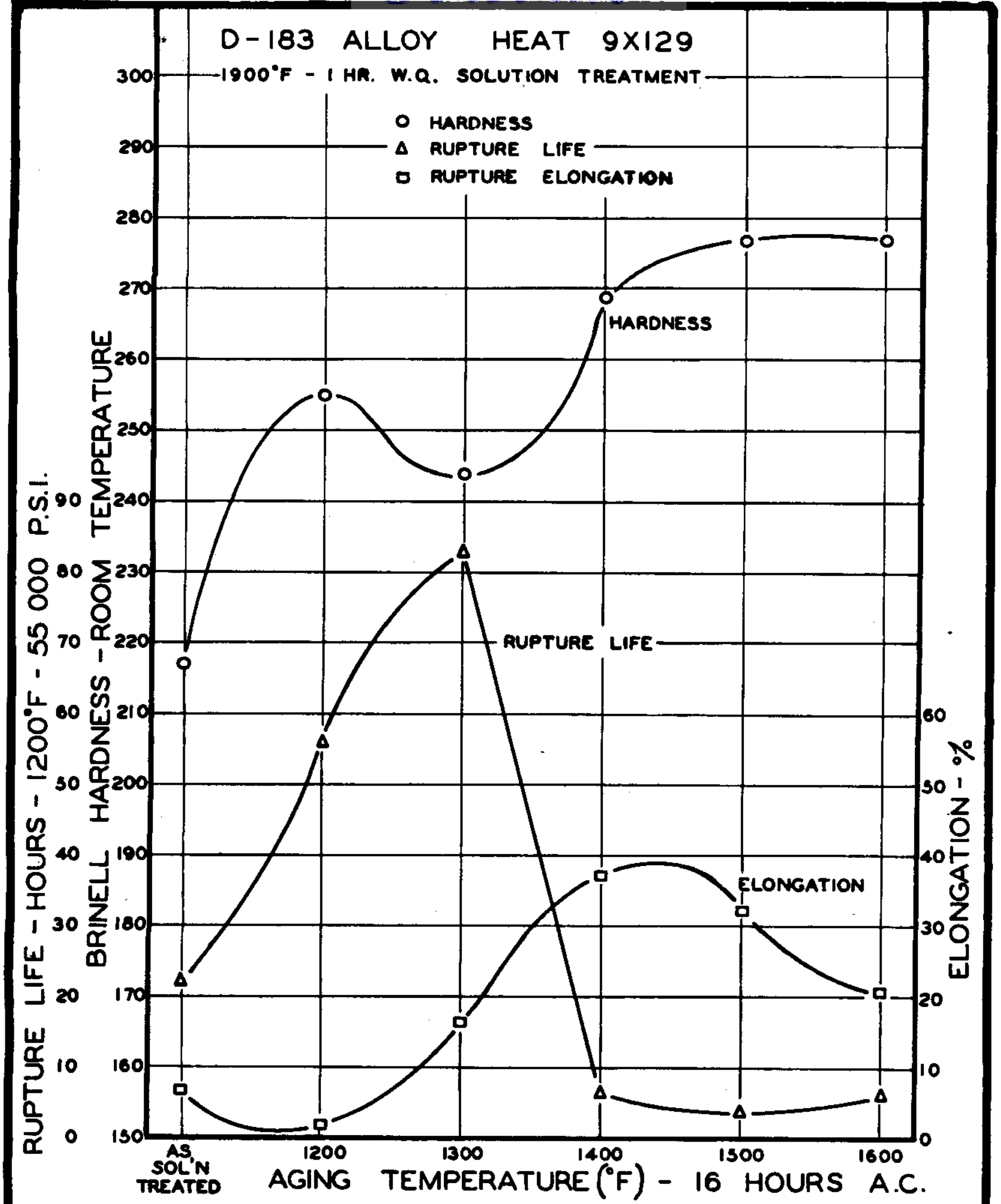

FIGURE 4 - EFFECT OF AGING TEMPERATURE ON PROPERTIES OF D-183 ALLOY - SOLUTION TREATED AT 1900°F

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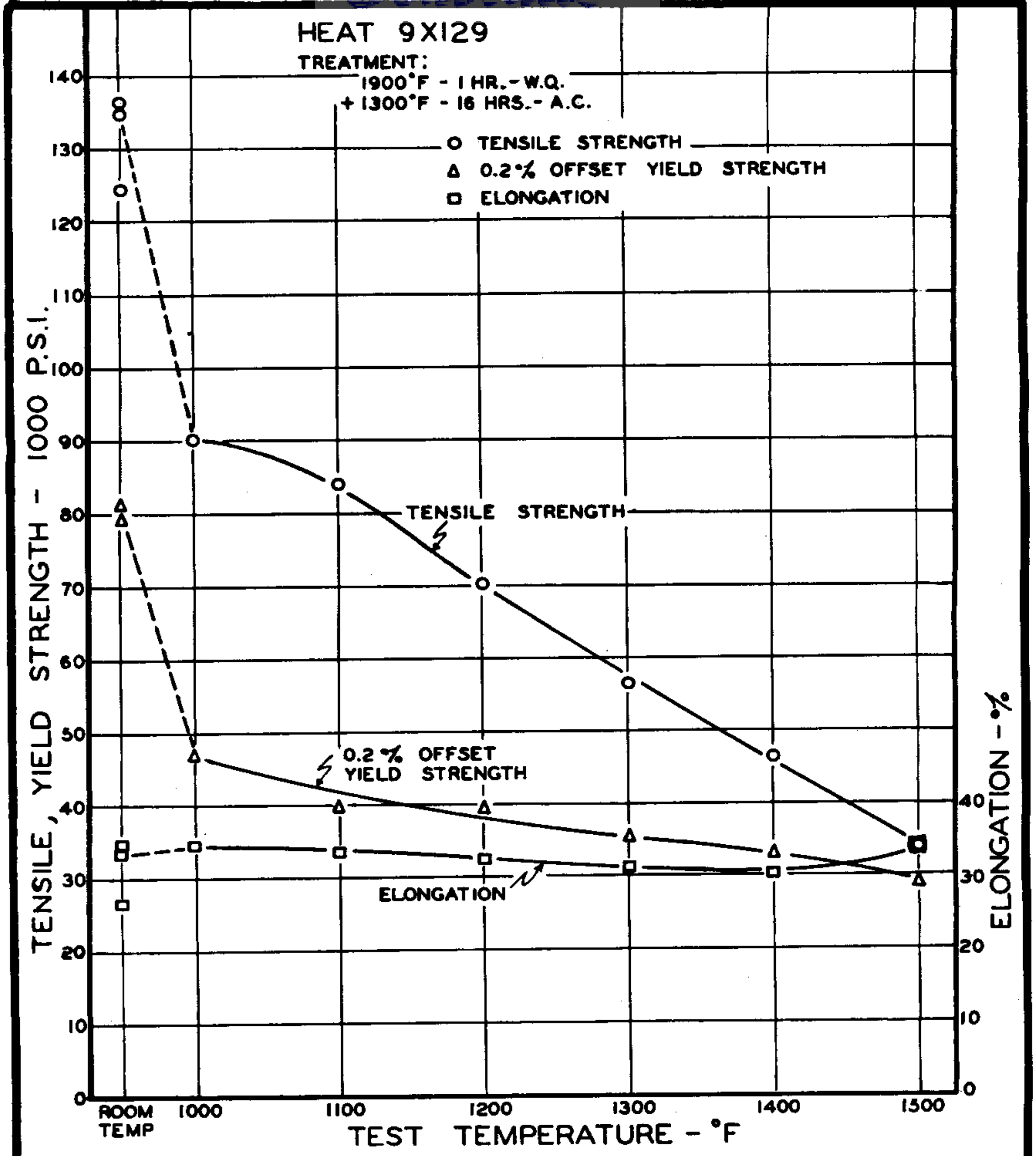

FIGURE 5 - EFFECT OF TEMPERATURE ON TENSILE PROPERTIES OF D-183 ALLOY

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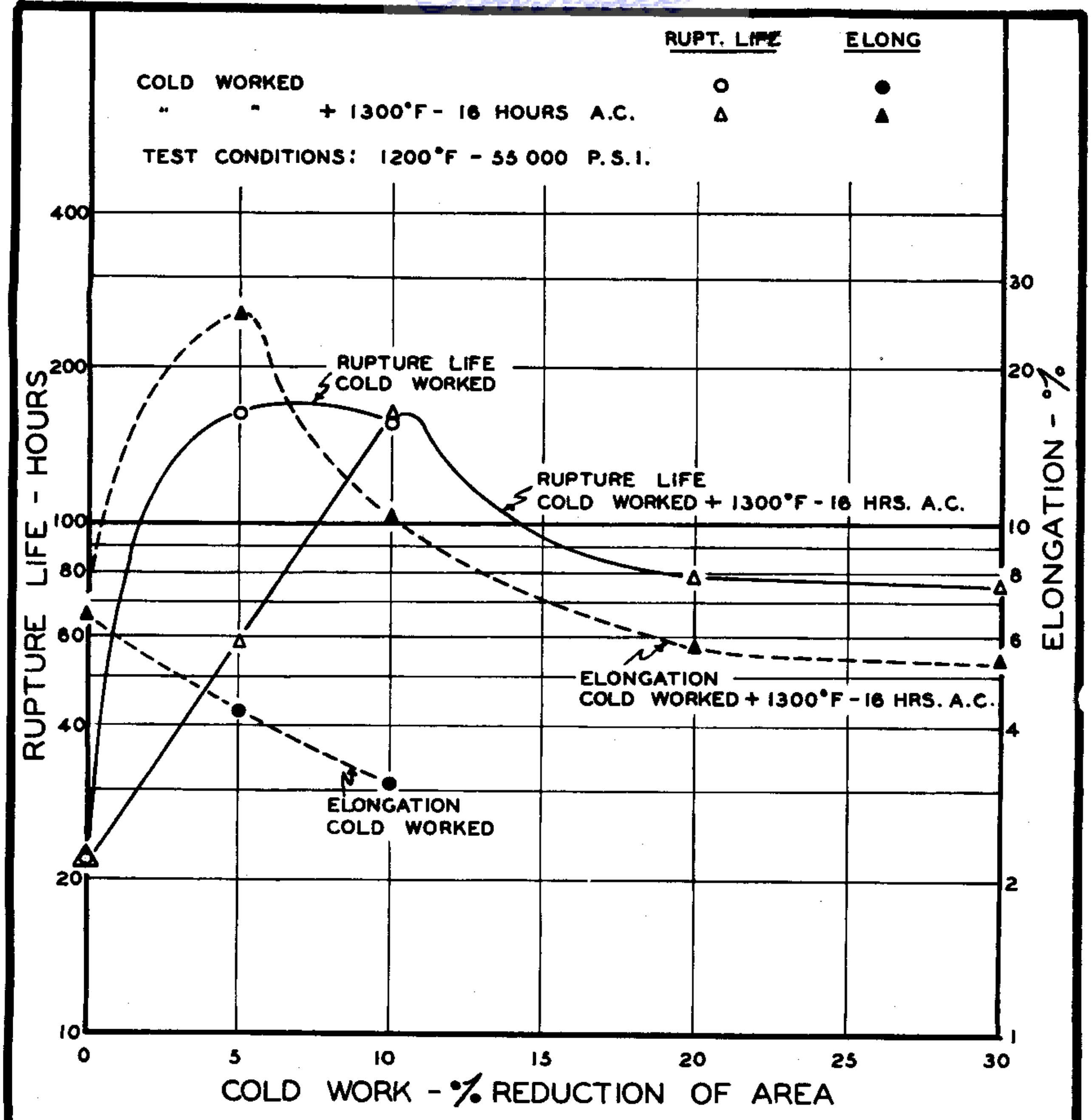

FIGURE 6 - EFFECT OF COLD WORK ON RUPTURE PROPERTIES OF D-183 ALLOY - PRIOR SOLUTION TREATMENT - 1900°F - 1 HOUR WATER QUENCH

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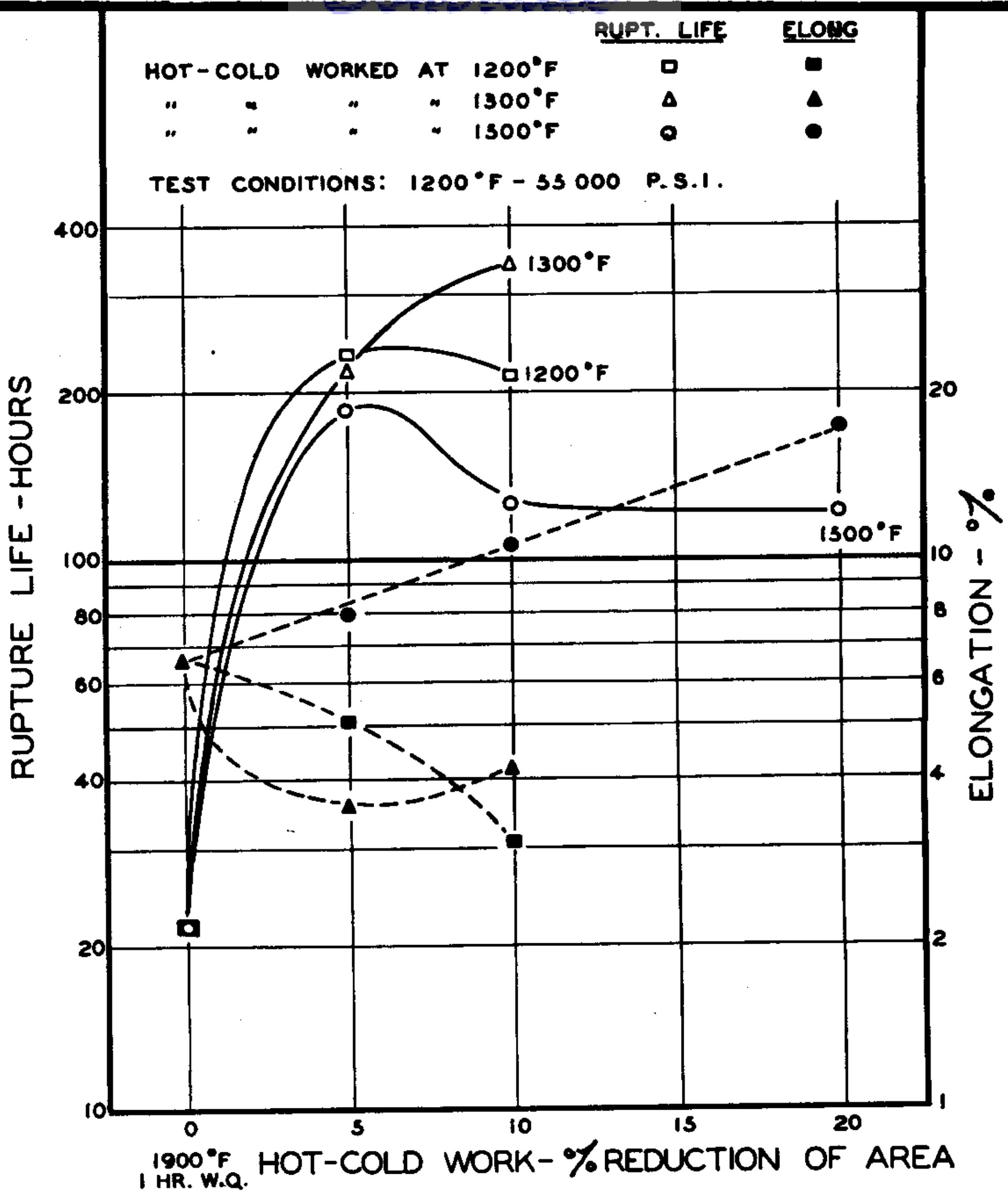

FIGURE 7 - EFFECT OF HOT-COLD WORK ON RUPTURE PROPERTIES OF D-183 ALLOY - PRIOR SOLUTION TREATMENT - 1900 °F - 1 HOUR WATER QUENCH

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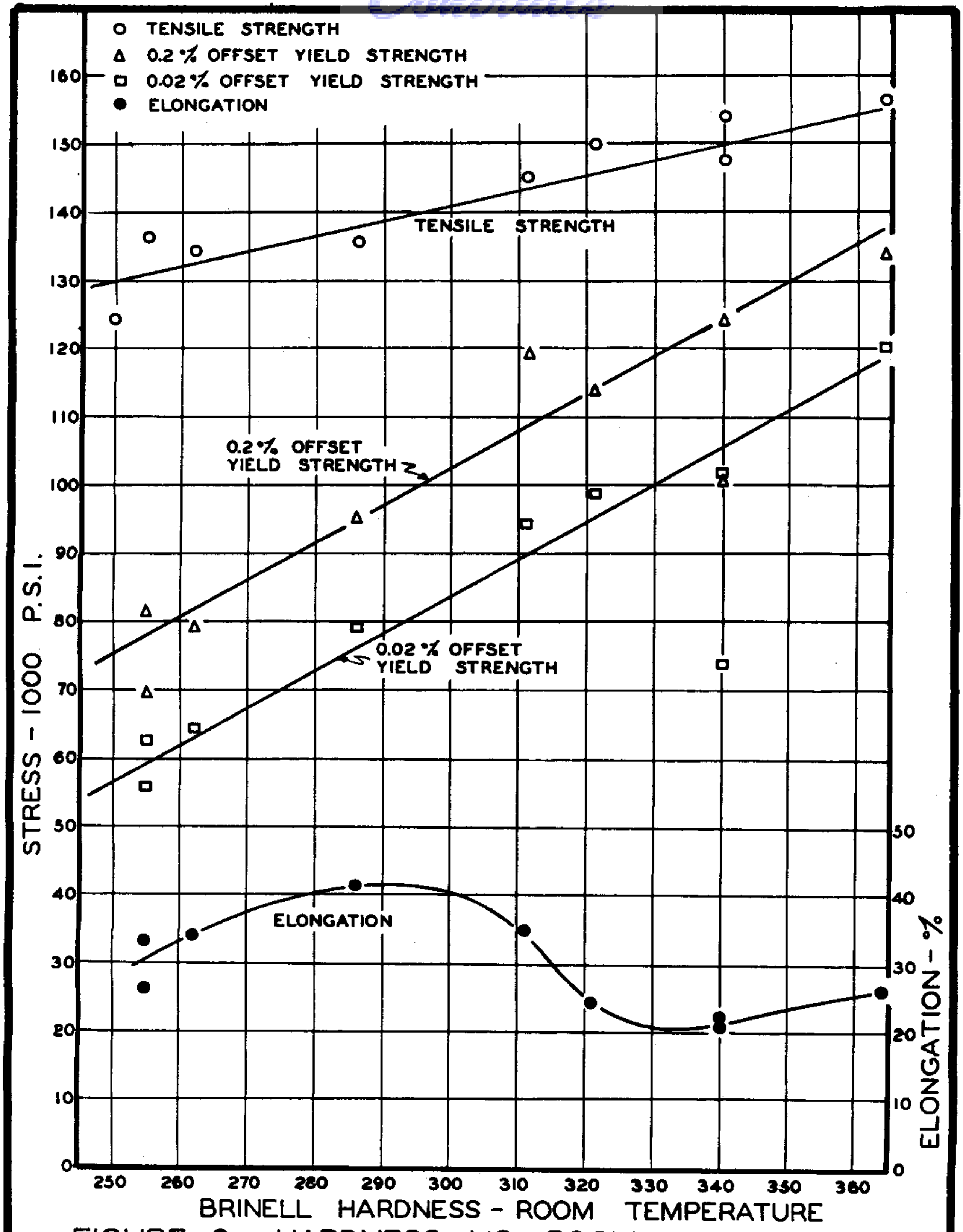

FIGURE 8 - HARDNESS VS. ROOM TEMPERATURE PROPERTIES OF D-183 ALLOY IN THE HEAT TREATED, THE COLD WORKED AND THE HOT - COLD WORKED CONDITIONS

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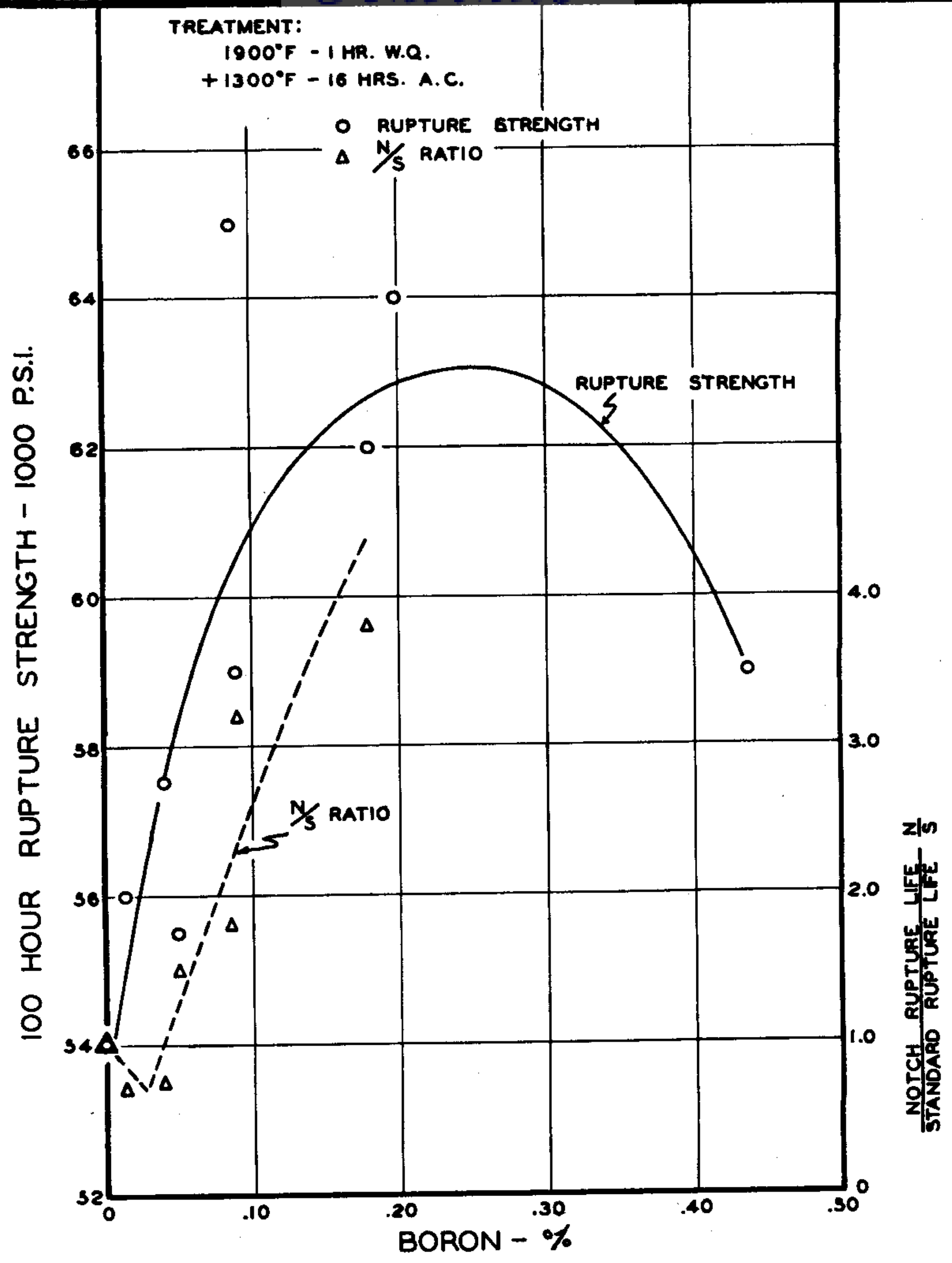

FIGURE 9 - EFFECT OF BORON ON RUPTURE STRENGTH AND N/S RATIO OF D-183 ALLOY AT 1200°F

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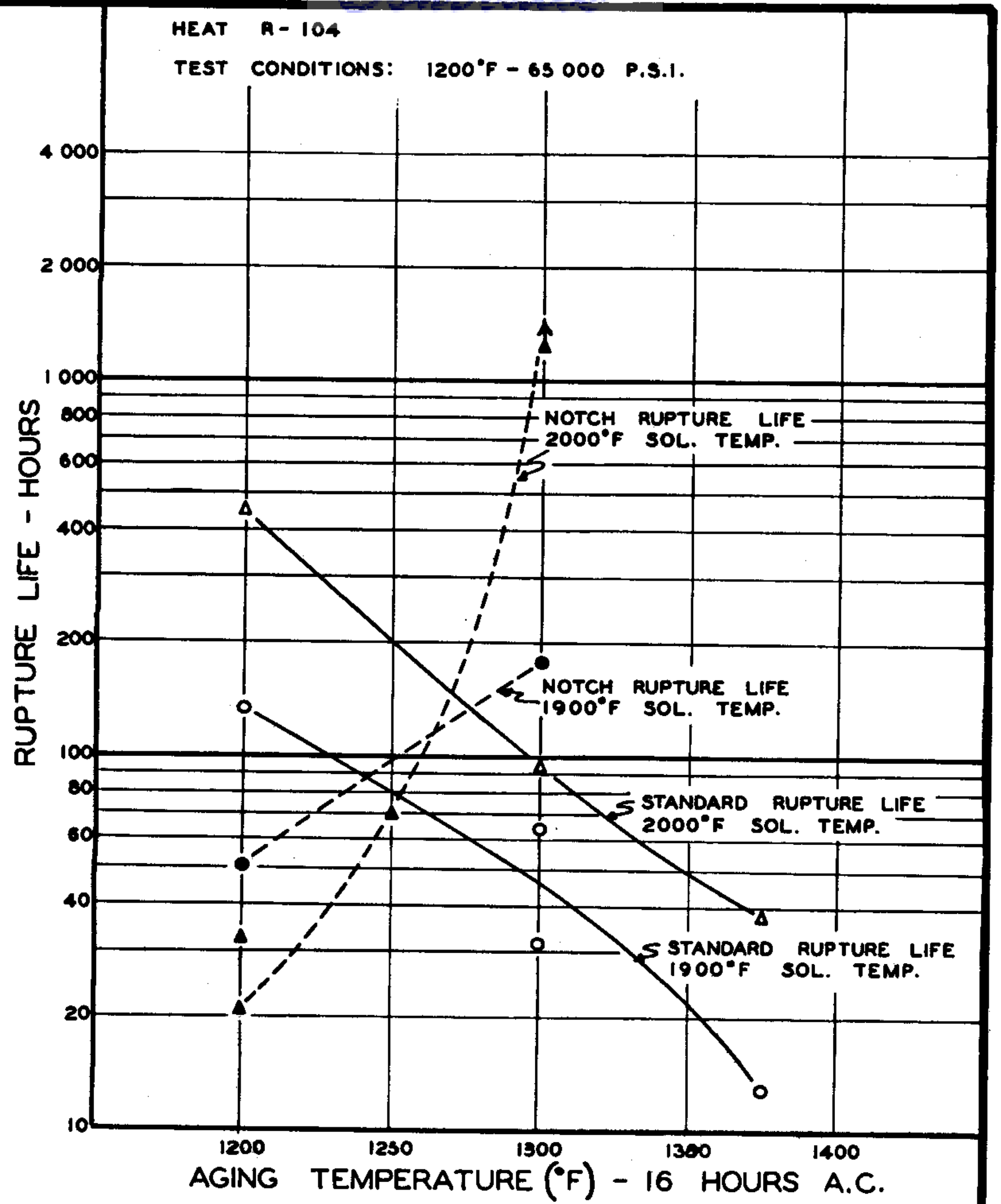

FIGURE 10 - EFFECT OF AGING TEMPERATURE ON STANDARD AND NOTCH RUPTURE LIFE AT 1200°F OF D-183 ALLOY CONTAINING .18% BORON (R-71 ALLOY)

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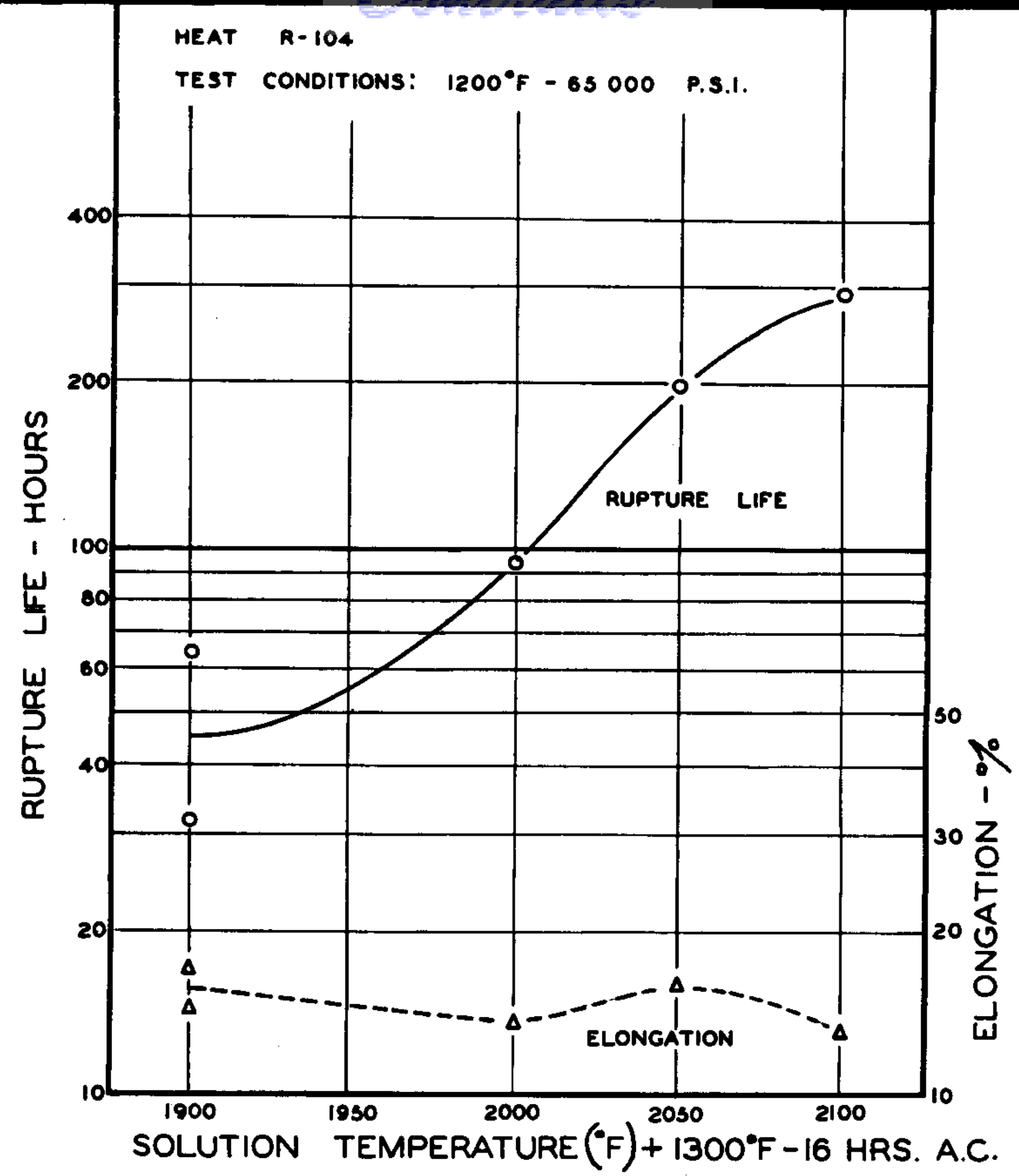

FIGURE 11 - EFFECT OF SOLUTION TEMPERATURE ON D-183 ALLOY CONTAINING .18% BORON (R-71 ALLOY)

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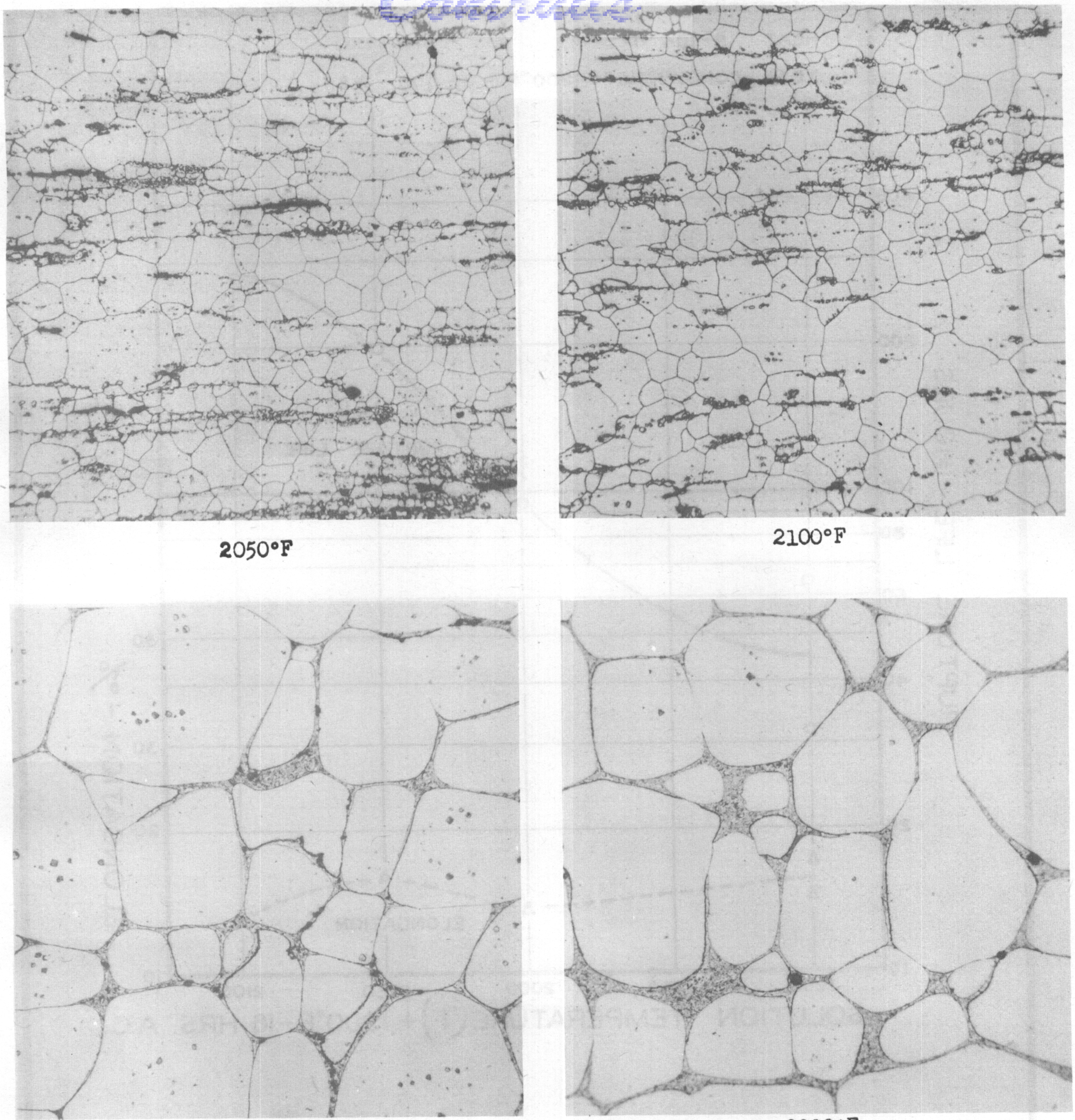

2050°F

2100°F

2150°F

2200°F

Figure 12 All specimens were solution treated for 1 hour, water quenched and aged 1300°F - 16 hours, air cooled.

Mag.: 250X Etchant: 92-HCl, 5-H_2SO_4, 3-HNO_3

Influence of Solution Temperature on D-183 Alloy with .18% B

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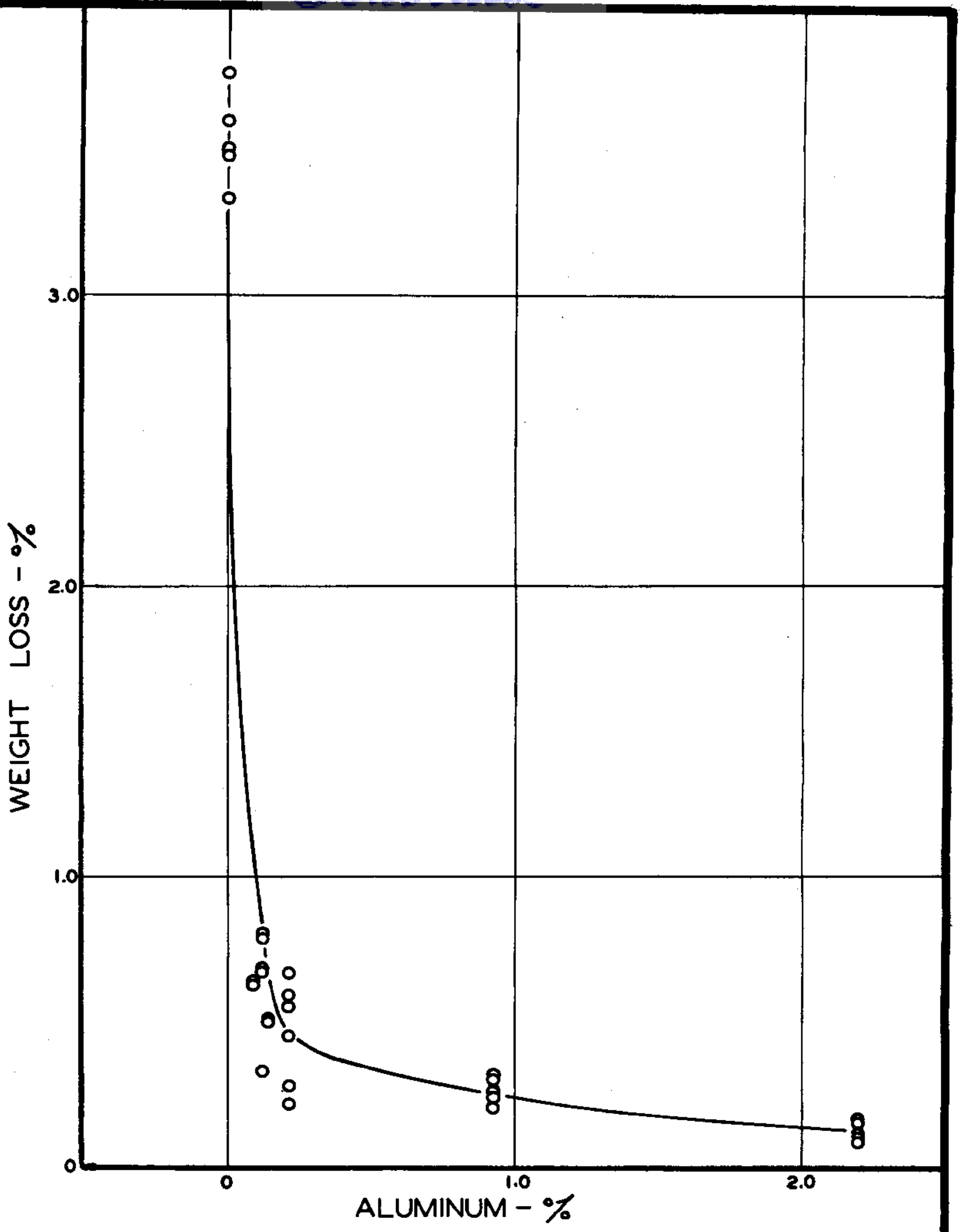

FIGURE 13 - EFFECT OF ALUMINUM ON OXIDATION RESISTANCE OF D-183 ALLOY AFTER EXPOSURE AT 1500 °F FOR 100 HOURS IN STILL AIR

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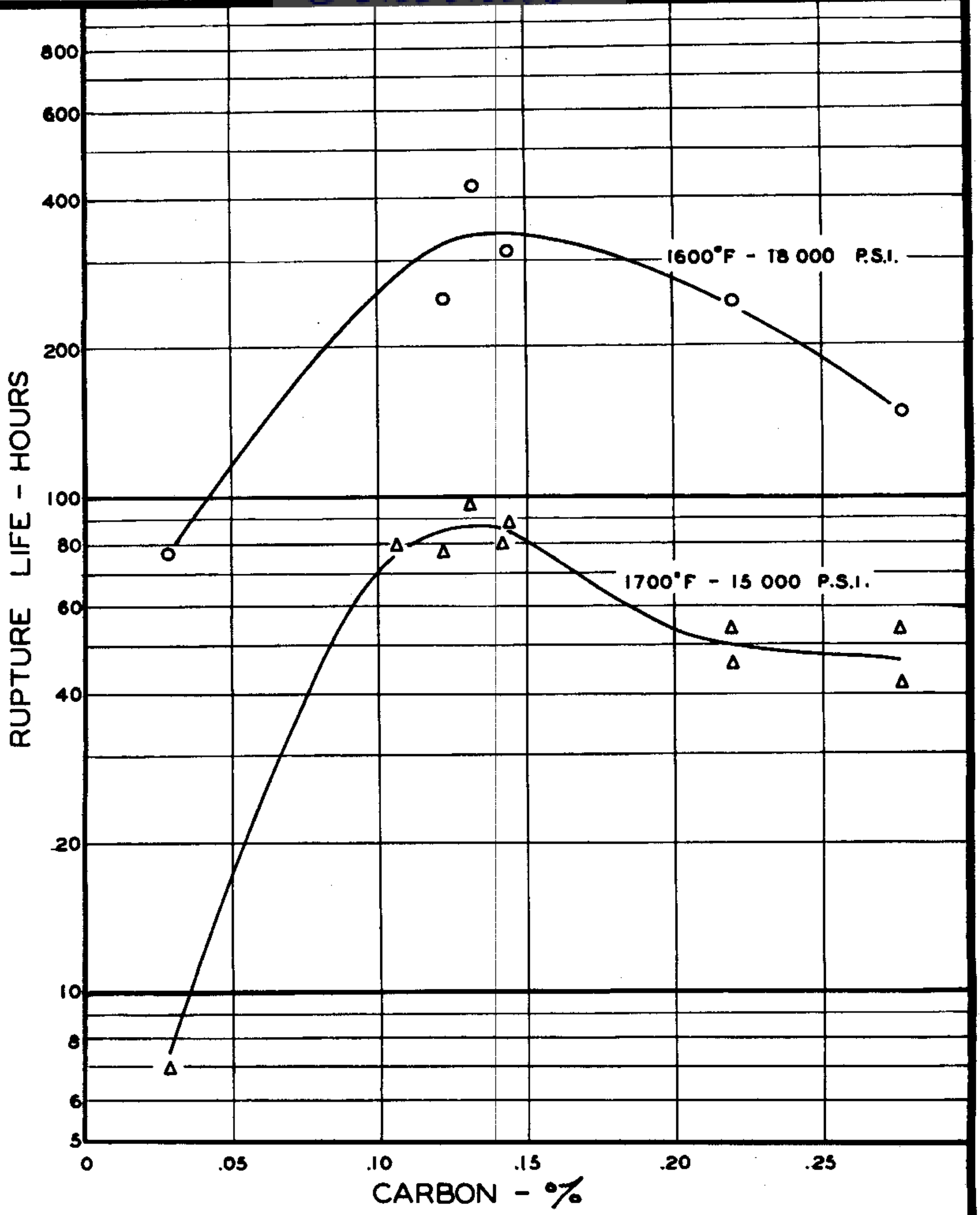

FIGURE 14 - EFFECT OF CARBON ON RUPTURE LIFE OF WROUGHT R-27 ALLOY

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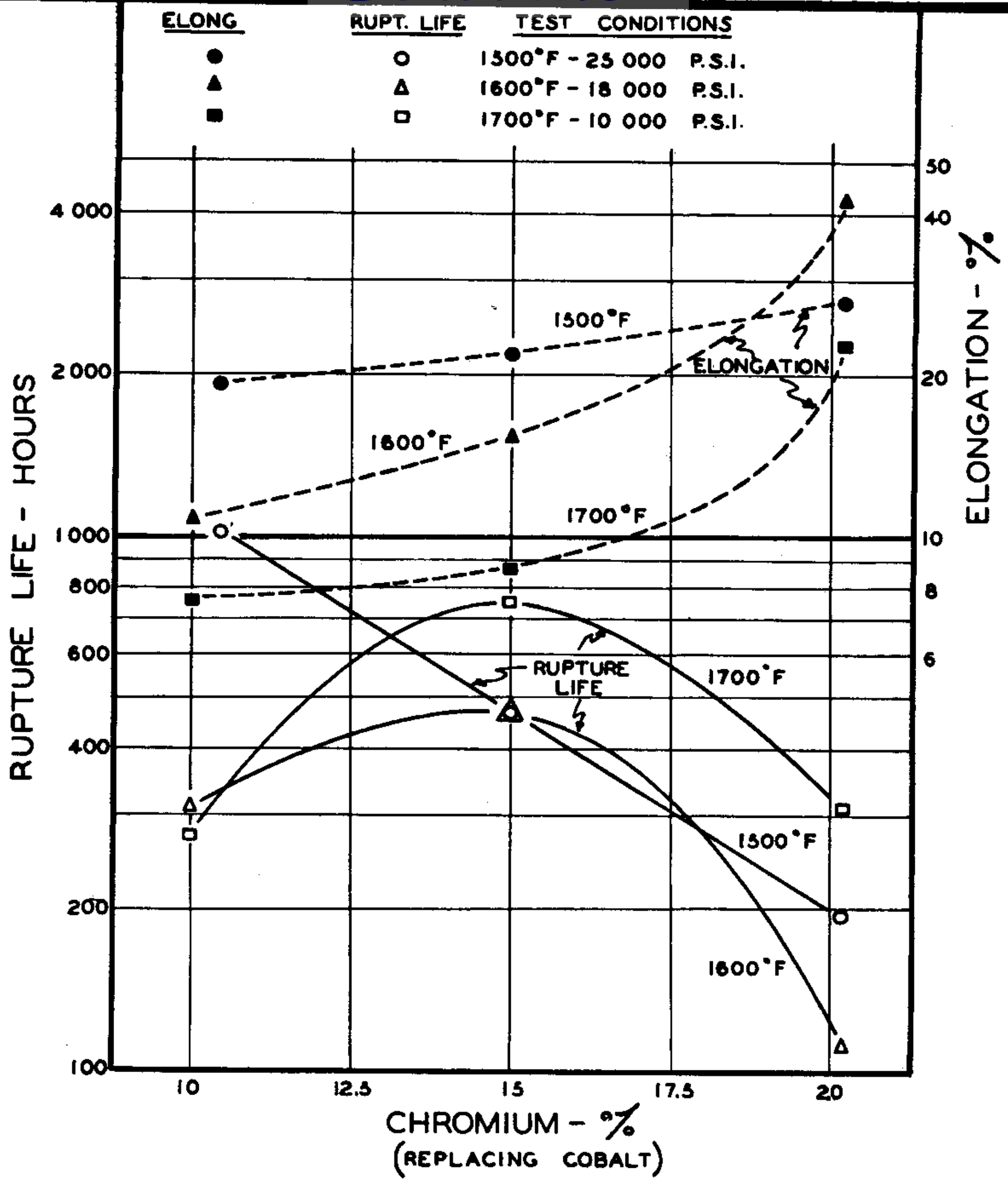

FIGURE 15 - EFFECT OF CHROMIUM ON RUPTURE PROPERTIES OF WROUGHT R-27 ALLOY

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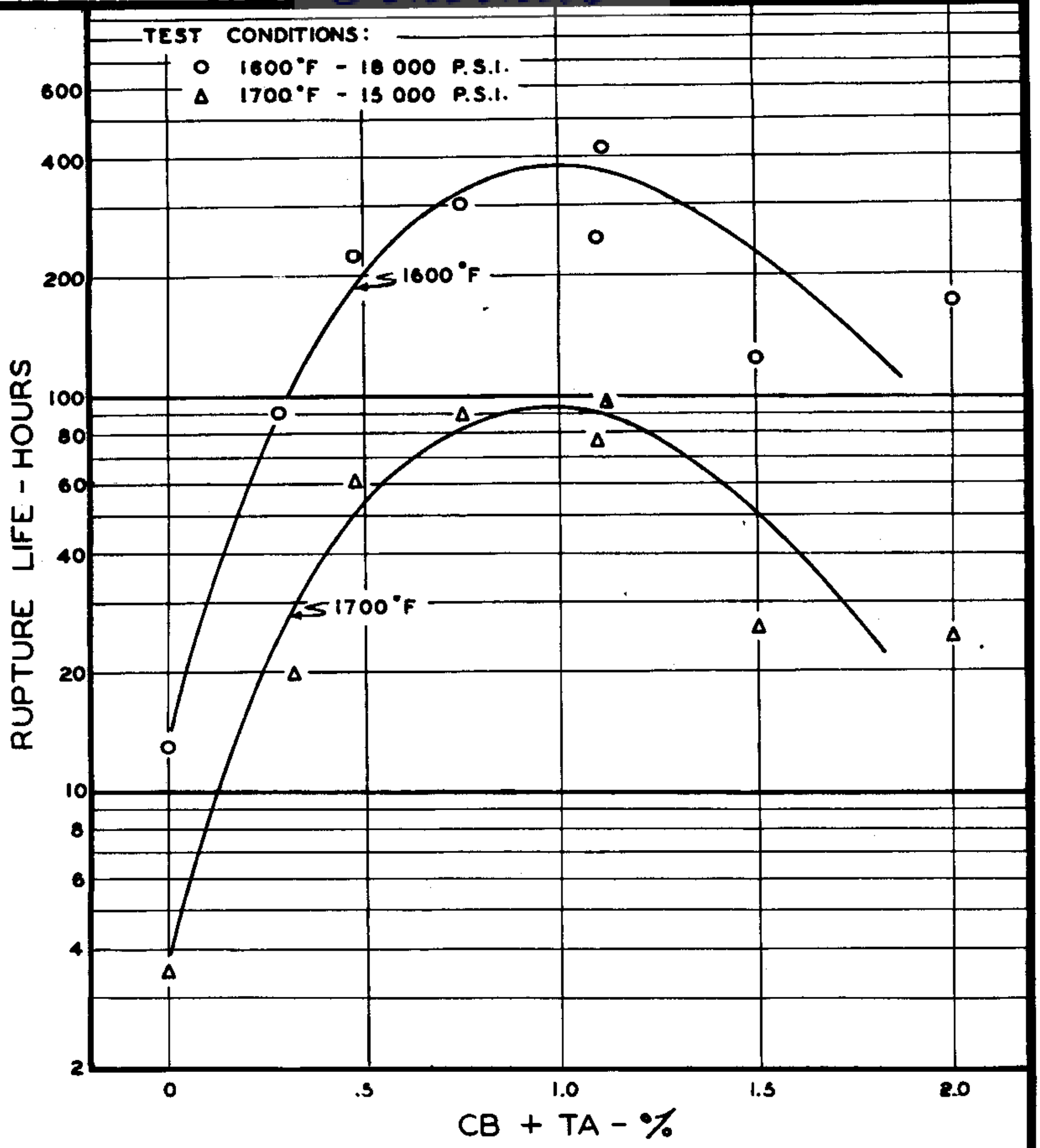

FIGURE 16 - EFFECT OF CB + TA ON RUPTURE LIFE OF WROUGHT R-27 ALLOY

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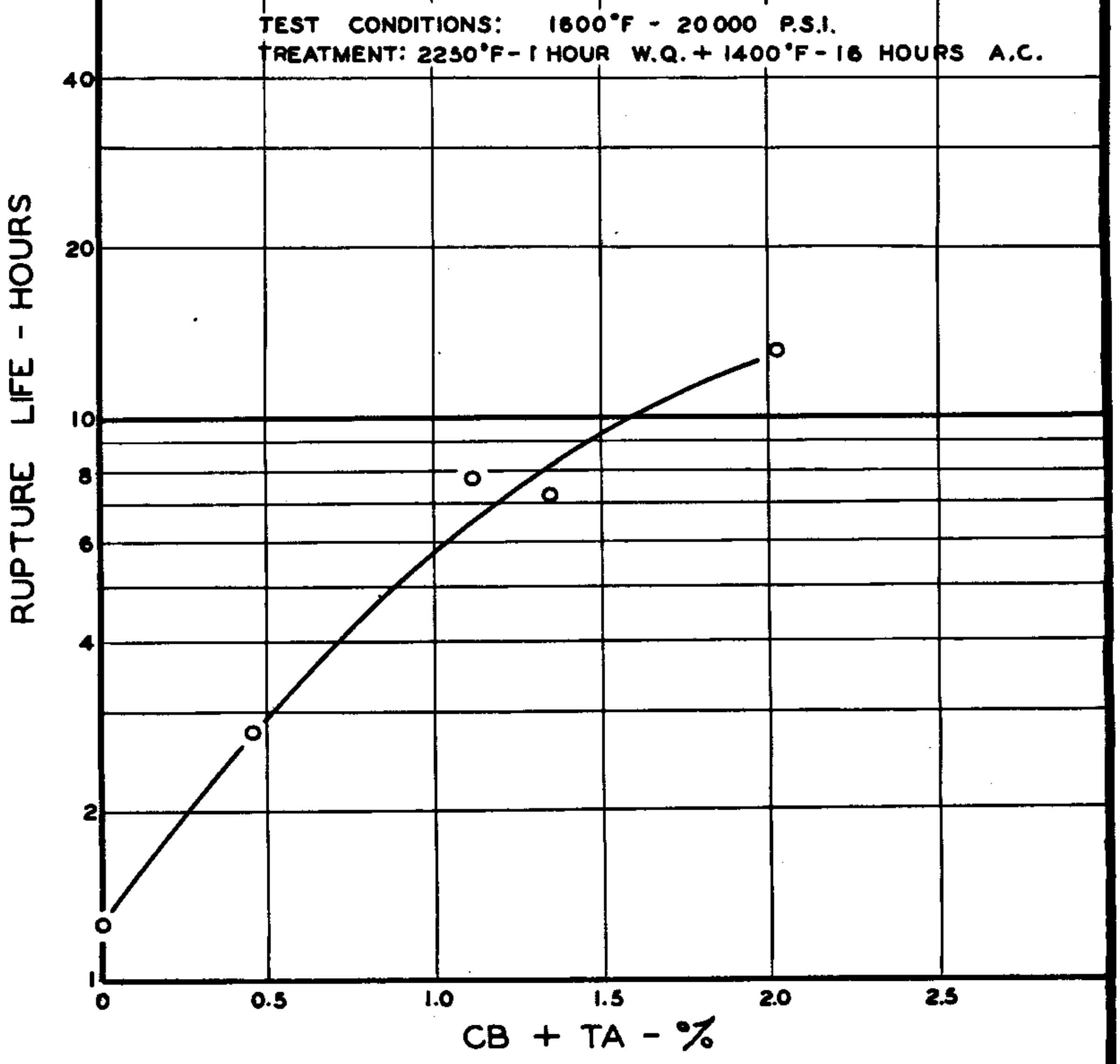

FIGURE 17 - EFFECT OF CB + TA ON RUPTURE LIFE OF WROUGHT LOW C R-17 ALLOY (.04 C)

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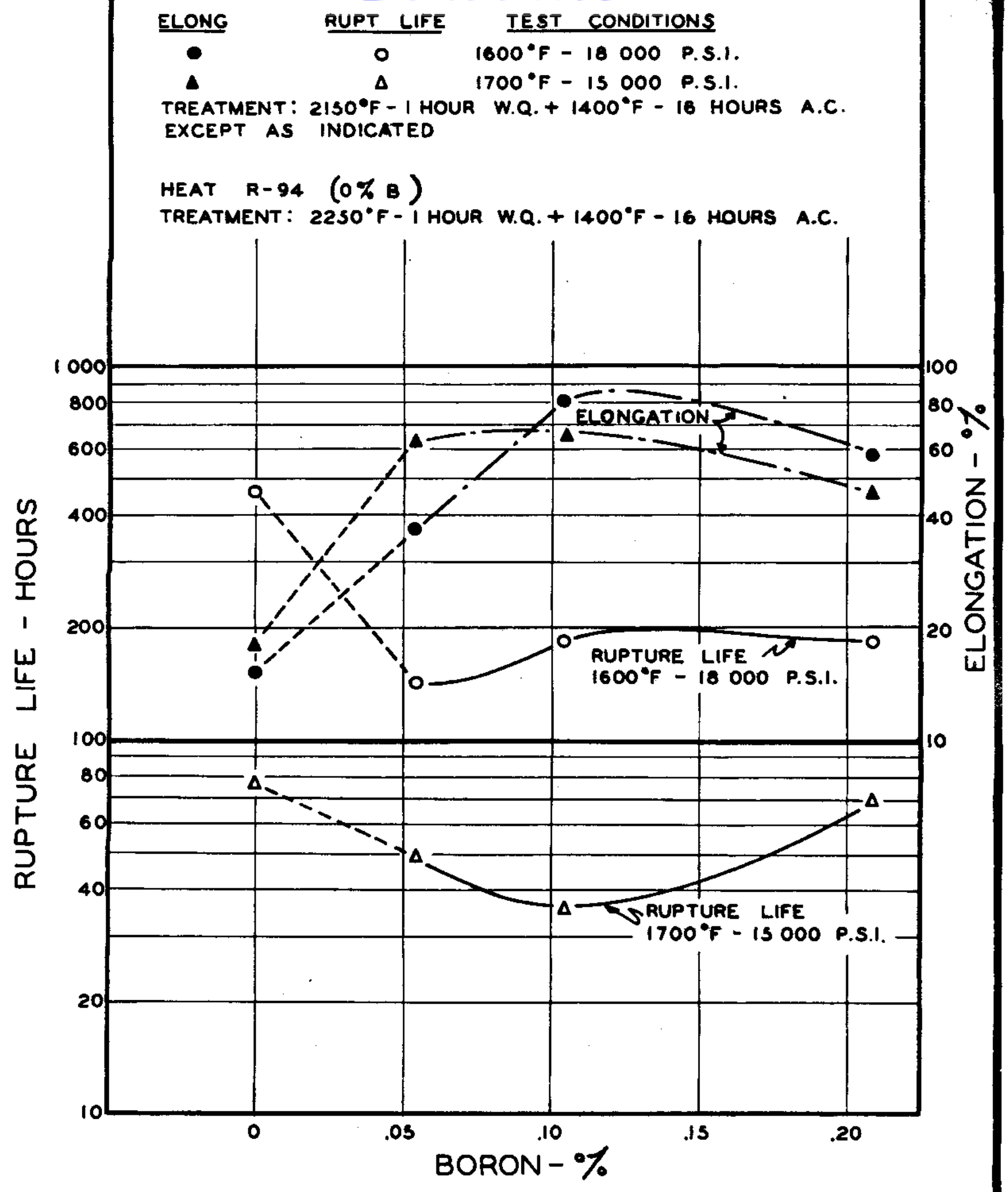

FIGURE 18 - EFFECT OF BORON ON RUPTURE PROPERTIES OF WROUGHT R-94 ALLOY

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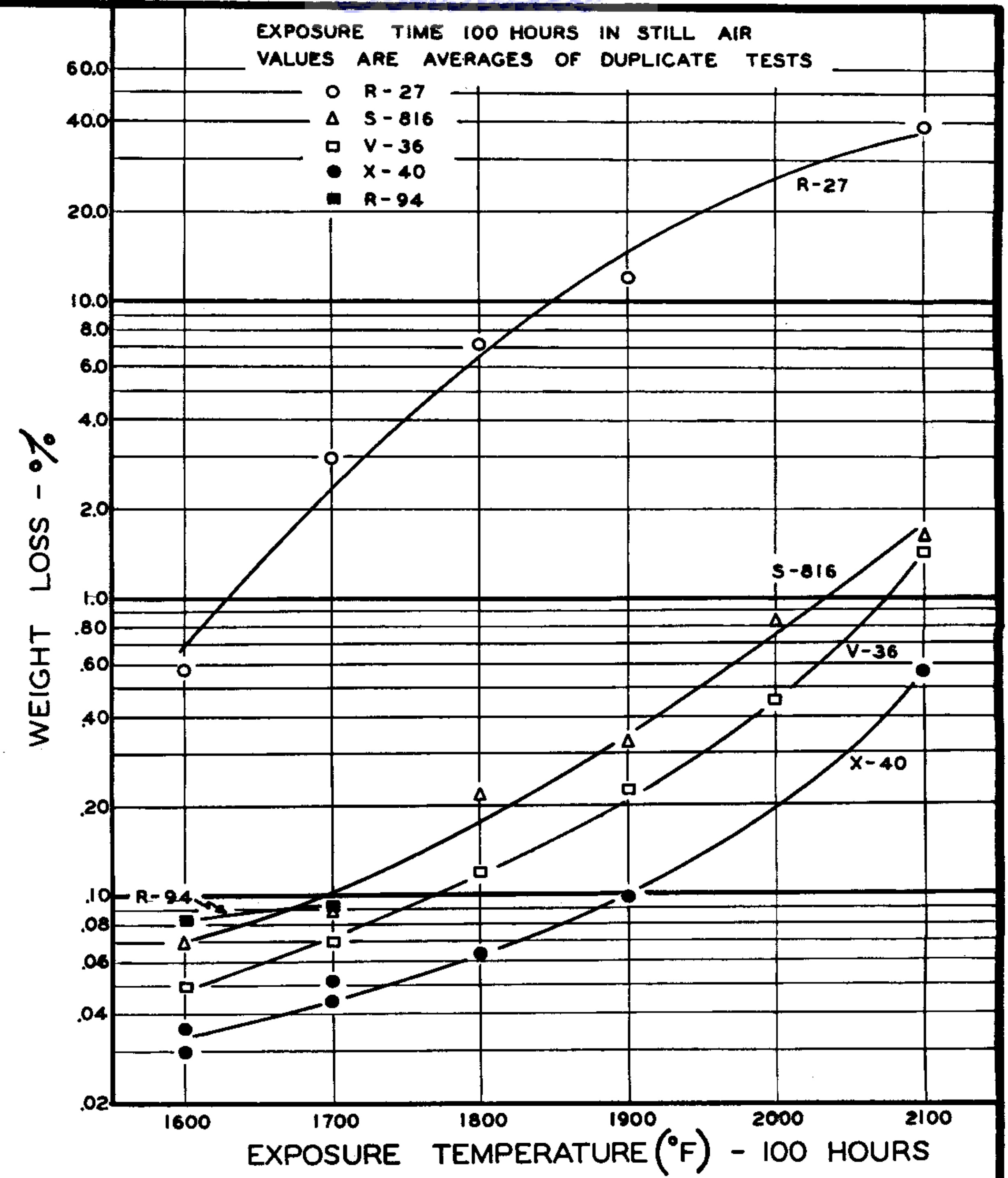

FIGURE 19 - EFFECT OF TEMPERATURE ON OXIDATION RESISTANCE OF WROUGHT R-27 AND R-94 ALLOYS COMPARED WITH WROUGHT S-816 (R-17), V-36, AND CAST X-40 ALLOYS

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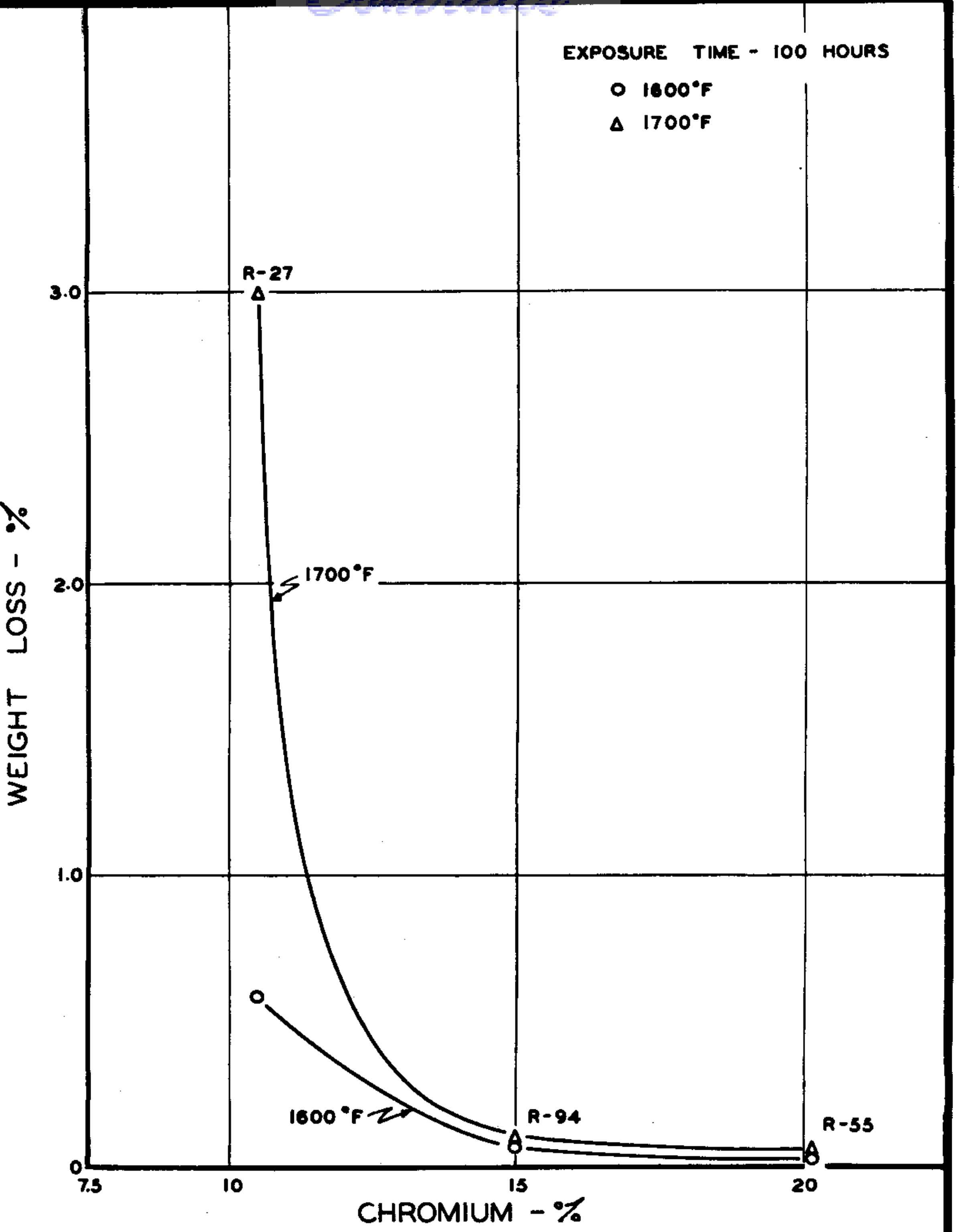

FIGURE 20 - EFFECT OF CHROMIUM ON OXIDATION RESISTANCE OF WROUGHT R-27 ALLOY

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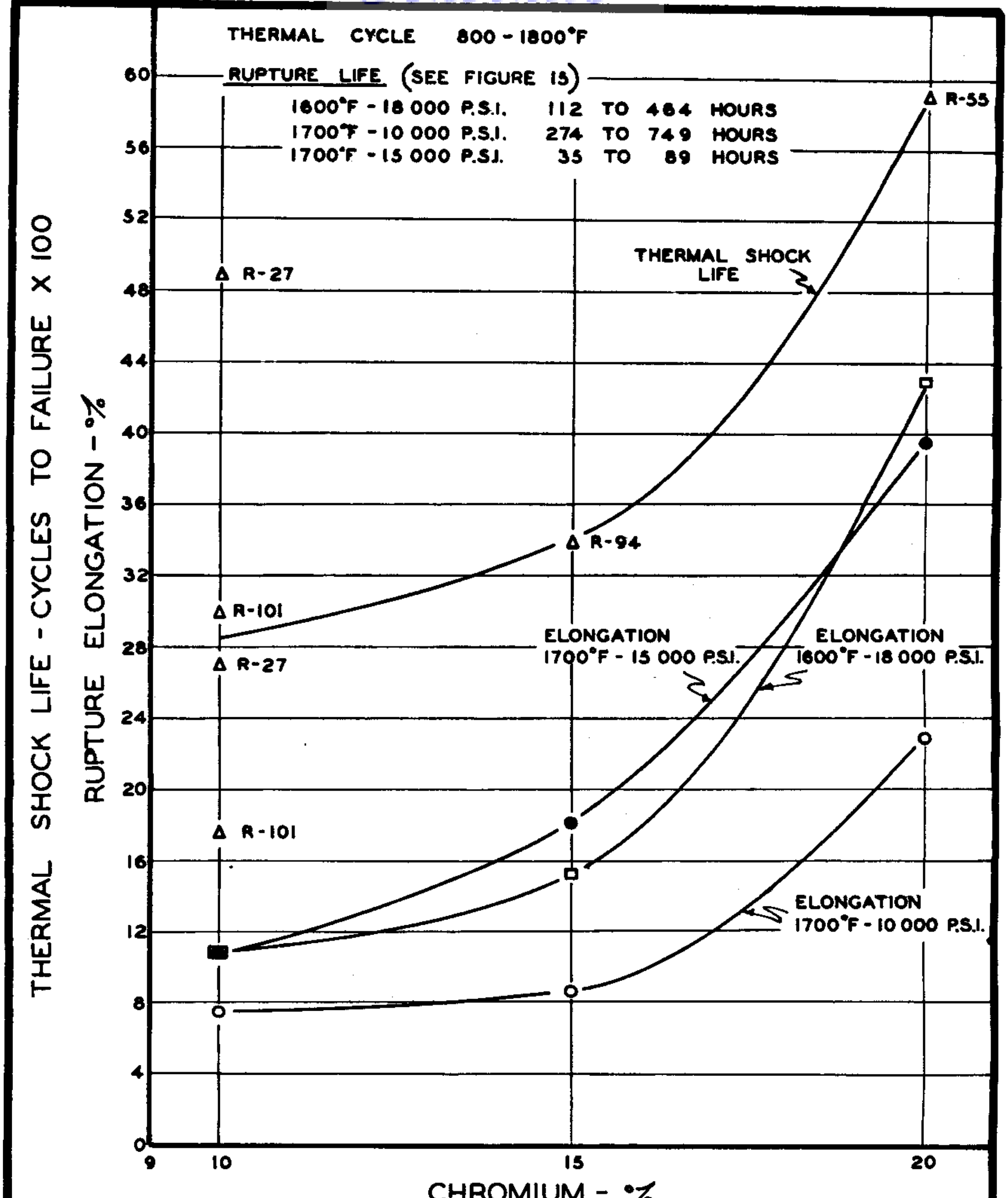

FIGURE 21 - EFFECT OF CHROMIUM ON THERMAL SHOCK LIFE AND RUPTURE ELONGATION OF WROUGHT R-27 ALLOY

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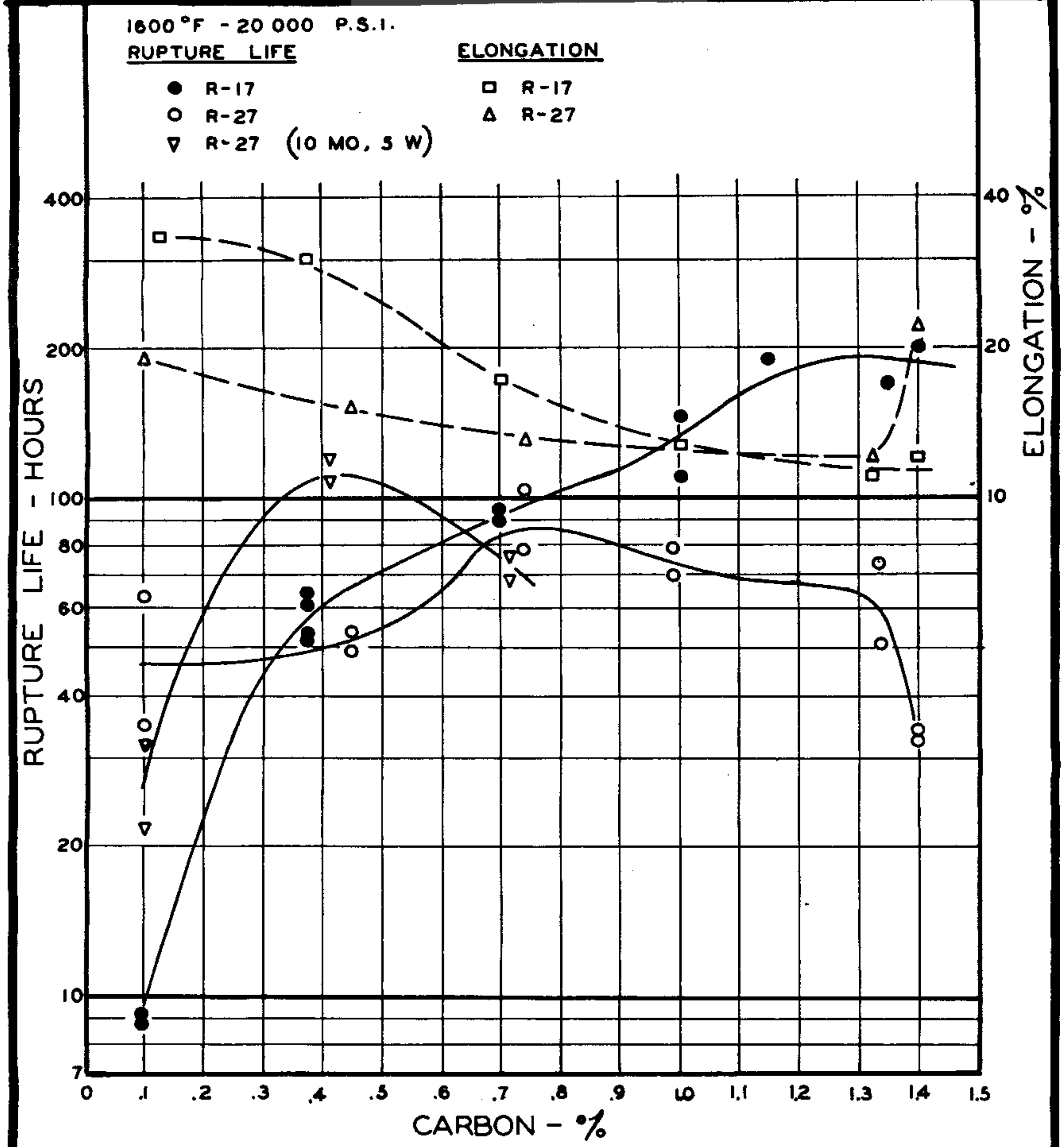

FIGURE 22 - EFFECT OF CARBON ON RUPTURE PROPERTIES OF AS-CAST COBALT BASE ALLOYS

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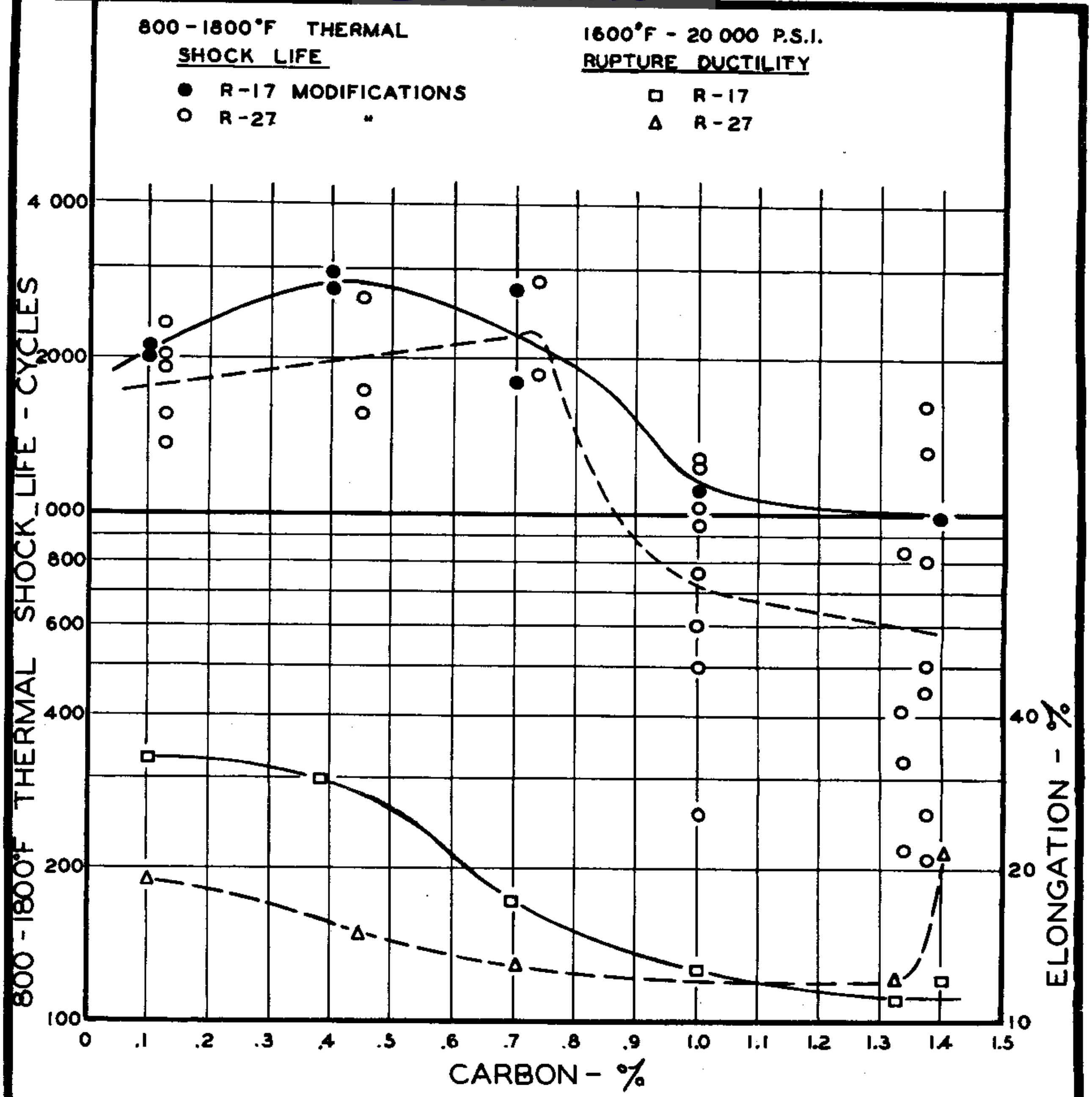

FIGURE 23 - EFFECT OF CARBON ON THERMAL SHOCK LIFE AND RUPTURE DUCTILITY OF AS-CAST COBALT BASE ALLOYS

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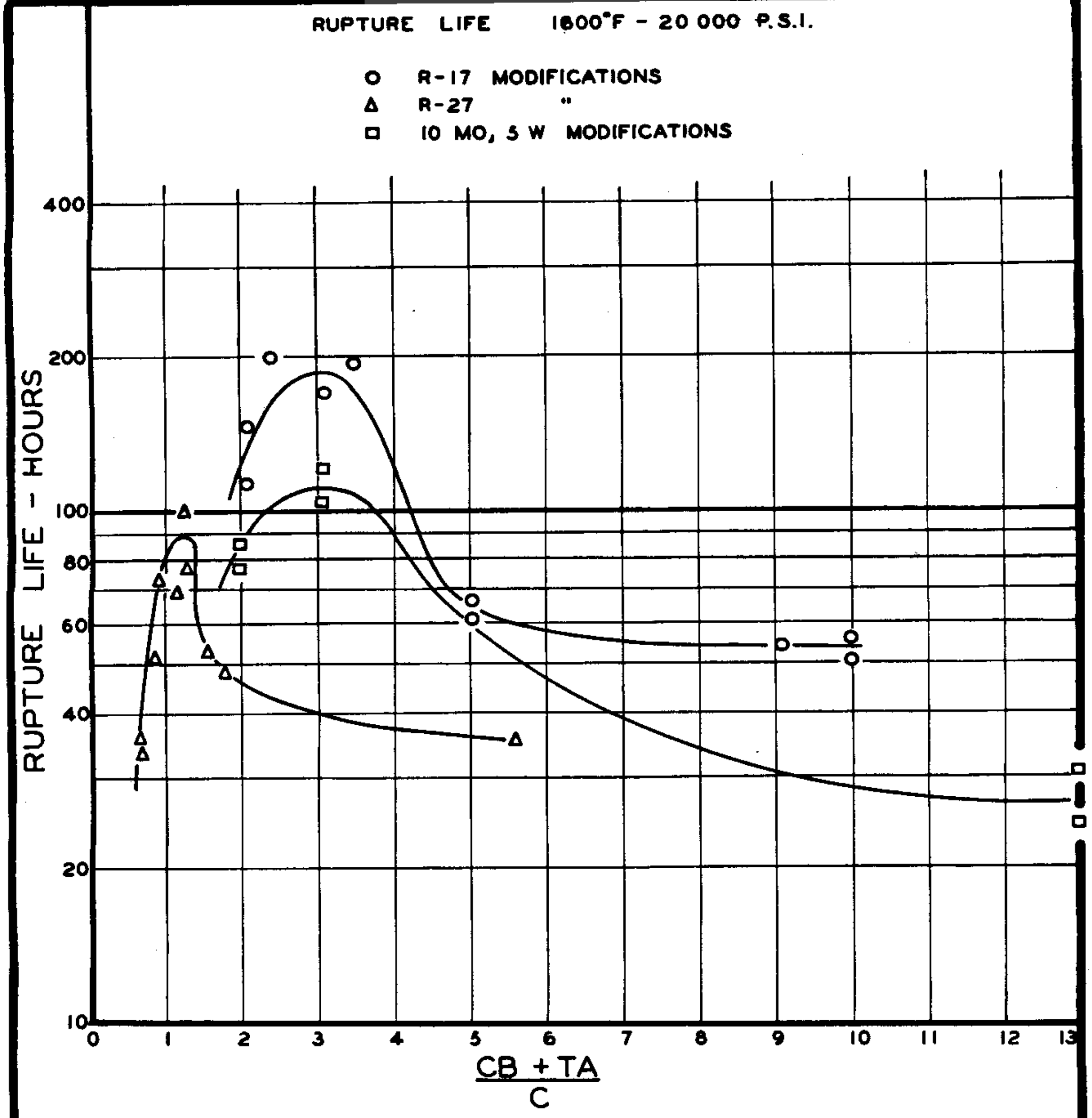

FIGURE 24 - EFFECT OF $\frac{CB + TA}{C}$ RATIO ON RUPTURE LIFE OF AS-CAST COBALT BASE ALLOYS

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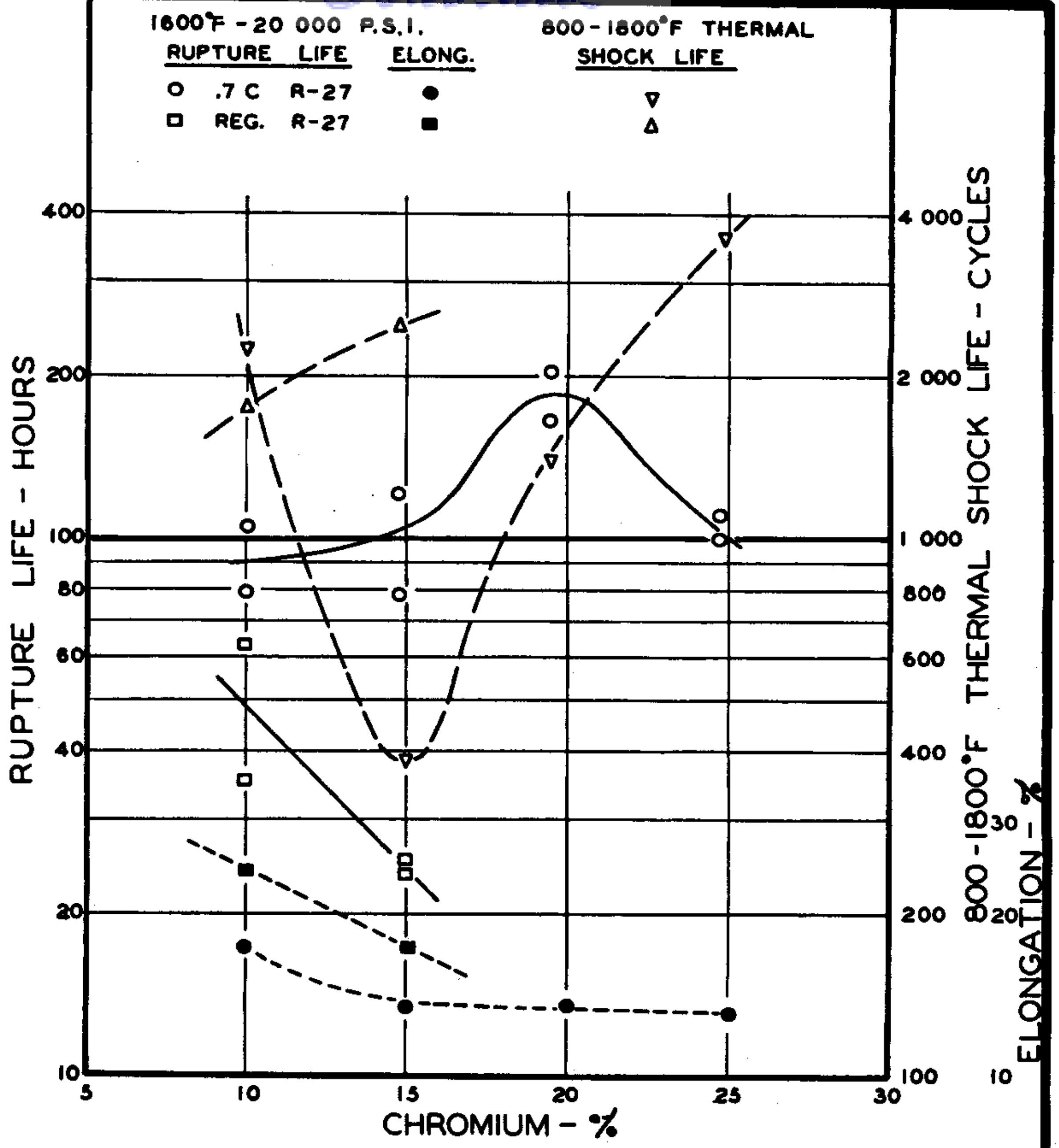

FIGURE 25 - EFFECT OF CHROMIUM ON THERMAL SHOCK LIFE AND RUPTURE PROPERTIES OF AS-CAST COBALT BASE ALLOYS

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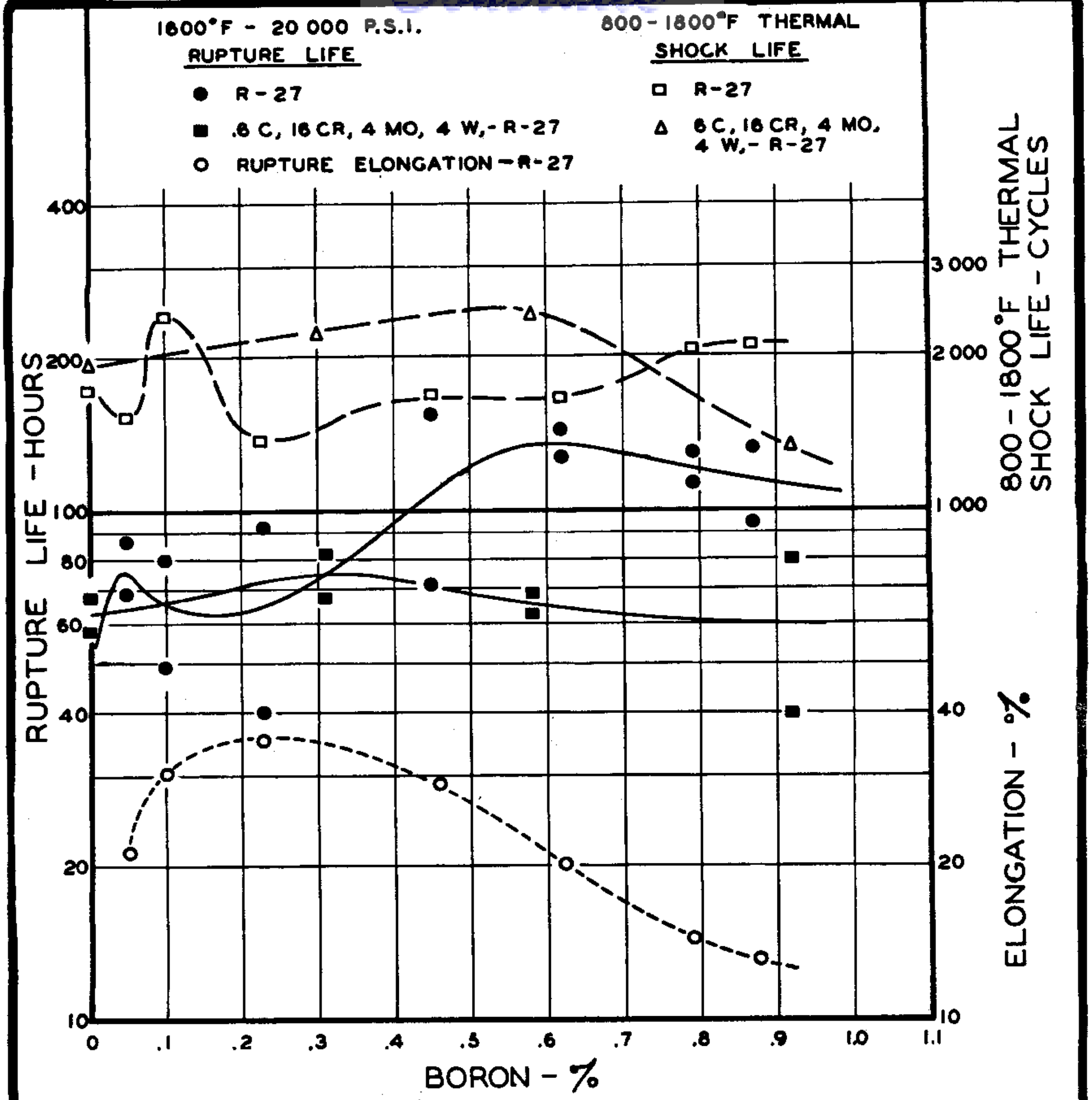

FIGURE 26 - EFFECT OF BORON ON RUPTURE PROPERTIES AND THERMAL SHOCK LIFE OF AS-CAST COBALT BASE ALLOYS

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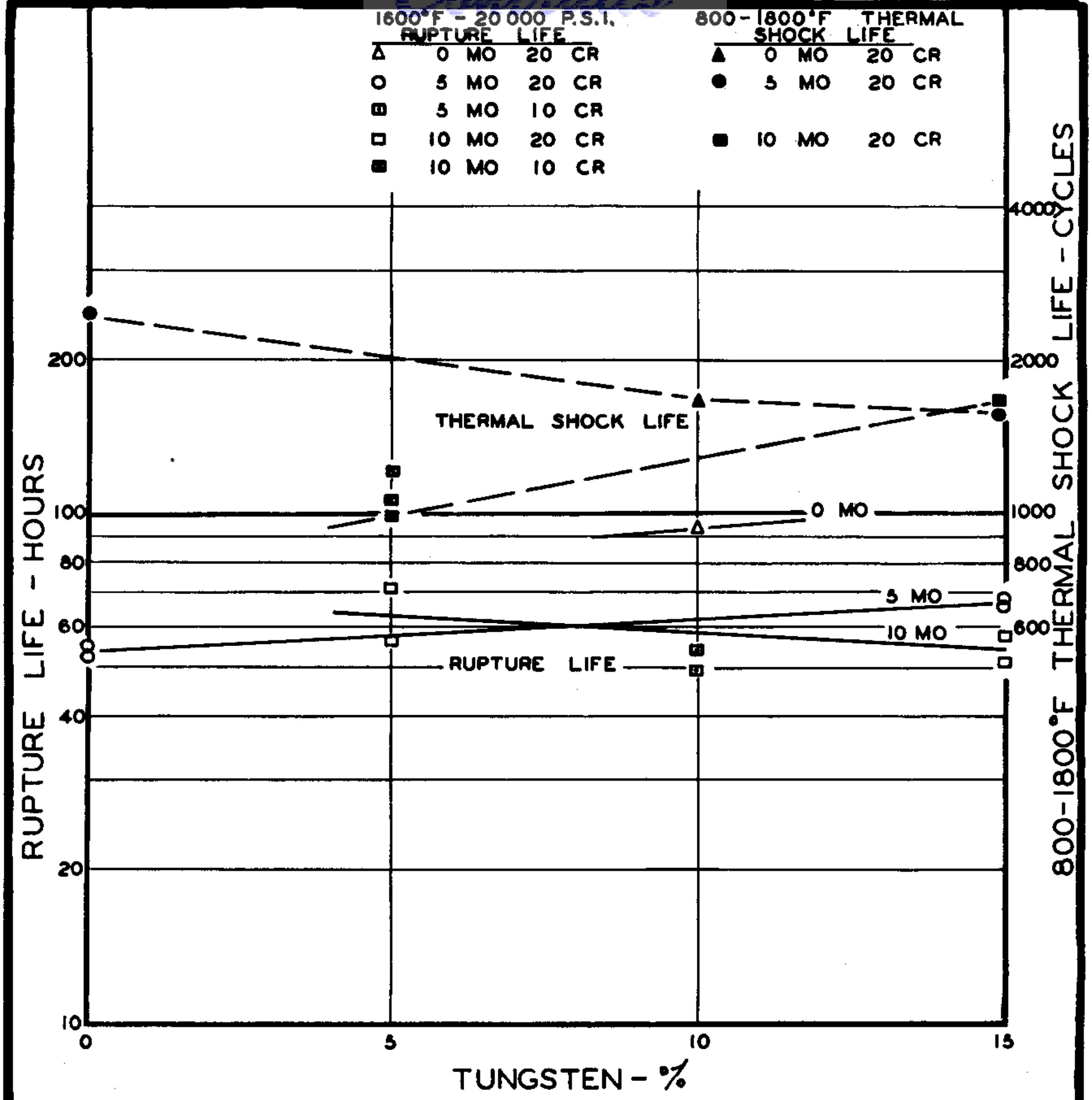

FIGURE 27 - EFFECT OF TUNGSTEN ON RUPTURE AND THERMAL SHOCK LIFE OF AS-CAST COBALT BASE ALLOYS

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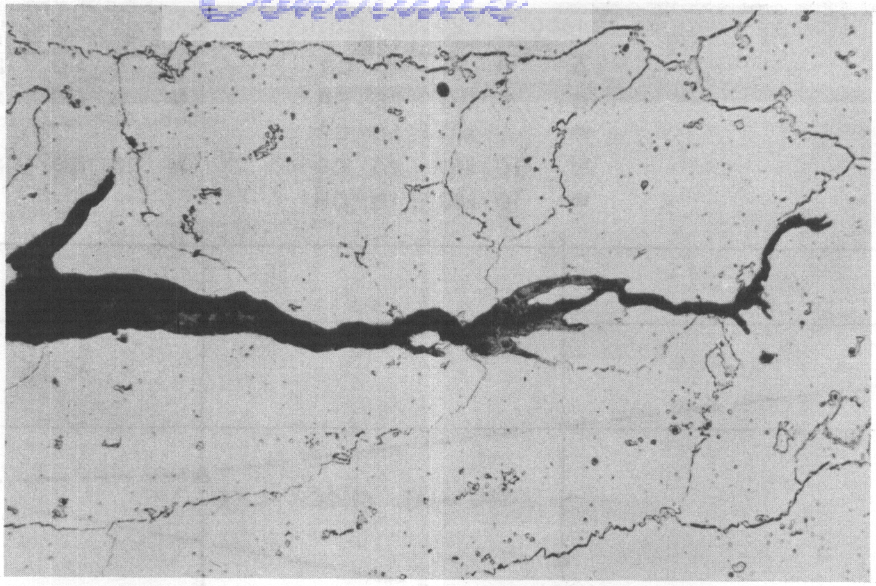
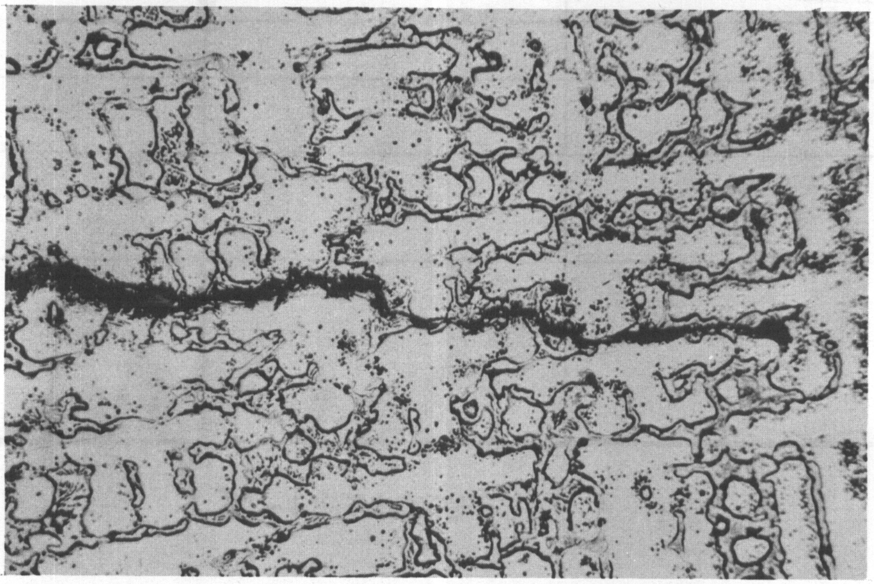


Regular R-27



.7 C, 20 Cr R-27

Figure 28 Thermal Shock Fracture and Microstructure of Regular and .7 C, 20 Cr R-27

Mag. 250X Etchant: 92-HCl, 5-H_2SO_4, 3-HNO_3

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