RESEARCH AND DEVELOPMENT OF WROUGHT AND CAST HIGH TEMPERATURE ALLOYS

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FOREWORD

This report was prepared by the Allegheny Ludlum Steel Corporation under Supplemental Agreement S2(53-823) of USAF Contract No. AF 18(600)-149. This contract was originally initiated under Task Nos. 73512 and 73515, (formerly RDO No. 615-13, "High Temperature Alloys" and RDO No. 615-18, "Aircraft Steels"), 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.

ABSTRACT

Study of wrought and cast Co-base and Fe-base alloys was conducted with the object of development of better high-temperature alloys having a minimum strategic alloy content. An alloy containing 10 Ni, 10 Cr, 10 W, 5 Mo, and 1 Cb+Ta was outstanding in rupture properties for the wrought Co-base alloys at 1500° to 1700°F. An 18 Mn, 12 Cr, 3 Mo, .8 V alloy had a good combination of properties for application at 1200°F for the wrought Fe-base alloys. Thermal shock properties were best for the cast alloys containing the highest Co. No correlation was apparent between thermal shock characteristics and other commonly measured properties.

PUBLICATION REVIEW

This report has been reviewed and is approved.

FOR THE COMMANDER:

M. R. WHITMORE
Technical Director
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SUMMARY

Investigations aimed at development of better high temperature alloys containing a minimum strategic alloy content were conducted on wrought and cast Co-and Fe-base alloys.

S-816 Alloys

A promising wrought alloy, R-27, resulted from a study of the influence of chemical composition on the rupture properties at 1500° to 1700°F of Co-base alloys. This alloy contains .10 C, 10 Cr, 10 Ni, 60 Co, 10 W, 5 Mo, 1.0 Cb+Ta, and 2 Fe. It combines a lower critical alloy content with a higher rupture strength than S-816. This modification reflects the two most significant trends which were observed: in general, equivalent rupture properties to those of S-816 were obtained from compositions with less Cb+Ta and C than normally present in S-816; there was also a trend toward improved strength as Co was raised to approximately 60 per cent.

U-912 Alloys

The best modification of U-912 which was obtained contained .12 C, 20 Cr, 20 Ni, 34 Co, 13 Fe, 6 W, 3 Mo, 1.7 Cb+Ta, and .18 N. Several modifications indicated that replacement of Cb+Ta with V was possible without lowering rupture properties. No outstanding modification was obtained.

Fe-Mn-Cr Alloys

Studies in a Fe-Mn-Cr system, based on the German alloy Cromadur which is an austenitic steel of low critical alloy content, led to the following optimum composition: .3 C, .4 Si, 18 Mn, 12 Cr, 3 Mo,

.75 V, .15 N. It has been designated as D-183 alloy. This development included an evaluation of the effect of heat-treatment as well as composition on room temperature tensile, and notch and smooth bar rupture properties at 1200°F. The principle compositional variables studied were those of C, Mo and N. The best treatment found was 1900°F, 1 hour, water quench plus 1300°F, 16 hours, air cool. Stock produced from production size ingots of an 1100 pound arc melted heat gave the same favorable properties as had been found for experimental induction heats.

Cast Alloys

The resistance to cracking during severe thermal cycling, or thermal shock, was evaluated for three cast alloys, Vitallium, F-88 and F-87 and their B modifications. No correlation was found between thermal shock properties and rupture, tensile, or structural characteristics. Vitallium had the best combination of properties. B improved thermal shock life in F-87 only, but resulted in better rupture strength for all the alloys.

INTRODUCTION

This research included a study of materials within three wrought and three cast alloy systems with the object of reducing strategic alloy content as well as raising the operating temperature range for alloys intended for use in aircraft gas turbines. The wrought alloys included Co-base S-816 and U-912 and an Fe-base Mn-Cr austenitic steel. The cast alloys were Vitallium, F-88 and F-87. Evaluation of compositional effects on these alloys was based mainly on rupture properties from 1200° to 1800°F with additional investigations of tensile, hardness, and thermal shock properties being conducted. This is the final report presenting results of these studies under Supplemental Agreement S2 (53-823) of Contract AF 18(600)-149. It covers the period from April 1, 1953 to March 31, 1954.

WROUGHT ALLOYS

Wrought Alloy Procedures

Melting

The experimental alloys in this program were melted as 17 pound heats in a 40 KW Ajax induction furnace using magnesia crucibles. Pouring temperatures were between 2675° and 2800°F as measured by a Pt-Pt + 13 per cent Rh immersion thermocouple. The 8 inch long tapered ingots measured 2-3/4 inches square at the top of the ingot.

Three 1100 pound heats of an Fe-Mn-Cr alloy were melted in a direct carbon-arc furnace. Ingot size was 12 inches square tapered by 26 inches long. A 50 pound pilot ingot from each of these heats was 3-5/8 inches square by 10 inches long. Some difficulty in Mn recovery in the first heat (9X-108), which was attributed to high Mn losses to the slag, was corrected on the second two heats (9X-110 and 9X-129).

Forging

Forging of 17 pound heats was done on a 2000 pound steam hammer between flat dies.

Forging of the S-816 and U-912 type alloys was done from 2150°F. Previous experience with more highly alloyed S-816 modifications indicated that temperatures in excess of this could result in hot-shortness. The reduction of the ingots to bar size required between 8 and 15 heatings. In finishing the bar stock of some heats to 5/8 inch square from 2150°F, some stiffness and minor corner cracking resulted from high heat losses at the reduced size. When this was encountered, the forging was stopped at 3/4 inch or 7/8 inch square.

The standard forging temperature used for the 17 pound heats of the Fe-Mn-Cr alloys was from a furnace temperature of 2050°F. These were forged to 5/8 inch square bar stock and required between four and eight heatings, depending on the plasticity of the modification, to attain the final size. The 12 inch ingots of the latter two of the 1100 pound heats of the Fe-Mn-Cr alloy were cogged on a 1500 ton press and 12,000 pound and 6,000 pound steam hammers to 2-7/8 inch billets from 2150°F. After grinding, the billets were rolled in guide rounds on a 14 inch mill from 2150°F. Final bar stock was 7/8 and 1-1/4 inch round. Considerable cracking was encountered during the initial cogging operations on both ingots. After grinding, subsequent hot working by forging and rolling went quite well. Pilot ingots of both heats worked only fairly during initial forging operations to 5/8 inch square bar stock.

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. The ratio of reduced section to diameter was 4:1.

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 specimen diameter across the base of the notch. The notch geometry was such as to produce a theoretical stress concentration factor of $K_t = 4.2$ for the notched bar in tension. The notches were produced by wet grinding. 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 be at least as long as the standard or smooth bar rupture life.

A lever arm, constant load rupture testing unit was used. The specimens were heated in 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 $\pm$ 3° at 1200°F testing temperatures, and $\pm$ 5° at 1500°F. A drop switch on the lever arm shut off the furnace current and controller when the test specimens fractured.

Tensile Testing

Room temperature tensile testing was done on the Fe-Mn-Cr alloys on specimens having a .505 inch diameter and a 2-1/4 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 extensiometer.

Wrought Alloy Results and Discussions

S-816 Type Alloys

Table I lists the 25 modifications of S-816 which were melted under this contract. Three of these, R-7, R-12 and R-56, cracked severely in forging and no stock was available for testing purposes. This indicated poor forgeability for heats containing either high Mo plus W and/or high Cr.

The majority of rupture testing on these modifications was done at 1500° and 1700°F. Some additional spot tests were run at 1600°F.

These test results along with hardnesses are listed in Table II. Table III presents the effect of solution treatment on the rupture properties of a limited number of modifications. These indicate a general superiority in rupture life at 1500° to 1700°F resulting from the solution treatment used (2250°F, 1 hour, water quench) in comparison to results after a lower solution temperature (2150°F) treatment. In general this higher solution temperature resulted in a slightly higher hardness and age hardening tendency in the alloys. From results in Table II, none of the modifications were found to be significantly age hardening. This lack of age-hardening is typical for this type of alloy.

Table IV gives rupture strengths of the modifications, the graphical presentation of which will be described.

In the selection of modifications earlier in the program, less emphasis was placed on a systematic evaluation of each element and more on large individual variations of the standard analysis which had shown promise in previous investigations or about which little information was available. Where significant trends became apparent, the later modifications were made to verify and expand upon these trends. Such were found for Cb+Ta, C, Ni, Cr and Co contents.

The effect of Ni content on the rupture strength is shown in Figure 1. It is seen that a 10 to 20 per cent Ni range is optimum at testing temperatures of 1500° and 1700°F. It is believed that below 10 per cent Ni, the strength of the alloy can be improved with an increase in Ni. Above 20 per cent Ni, the strength decrease is attributed to the decrease in Co which the Ni replaces.

These curves are redrawn in Figures 2 and 3 to distinguish the effect of Ni from that of Co. In these figures, all of the S-816 modifications made under the contract, except three, are shown. The exceptions are heats R-10, R-34, and R-36. These have been excluded for two reasons. Heat R-34 had a very high Cb+Ta/C ratio (27:1) which will be shown later to have a detrimental effect on rupture strength. Heats R-10 and R-36, which had very low rupture properties, have been excluded since they had a combined Co and Ni content greater than 70 per cent. The rupture properties of alloys above this total Co and Ni content evidently are influenced markedly by an unknown factor which is also manifested by an appearance of high ferromagnetism.

Replacing 10 per cent of Ni with Co did not appreciably influence properties. However, a replacement of 10 per cent Cr with Co in alloys containing 10 per cent Ni, as was done in heats R-9 and R-27, gave an increase in properties up to 60 per cent Co.

Figures 4 and 5 show the limited data which were obtained on the effect of Cb+Ta and C contents in S-816. For comparison with the low-C low-Cb heats shown in these graphs, the rupture times for the standard S-816 heat (R-17) are also included. These results indicate that, if Cb+Ta is lowered to around 1 per cent at a C level of around 0.10 per cent, the same rupture strength properties are obtained as for standard S-816. This trend is more pronounced at 1700°F than at 1500°F. The Cb+Ta to C ratio in these modifications thus remains the same, around 10 to 1, as in standard S-816. This was established for best rupture strength in previous investigations (a) at high Cb levels (4 per cent). It is noted that the rupture

(a) Allegheny Ludlum-General Electric - Unpublished data ductility and room temperature hardness are lower in the low Cb and C modifications of S-816 than in the standard alloy. These offsetting factors, as well as the promising rupture properties obtained, need verification on additional heats by more extensive and longer time testing and checks on the structural stability. The lower levels of C and Cb+Ta were used in most of the further modification work in this program.

Limited modifications made early in the program involved W and Mo variations within wide ranges. The results in Figure 6 show that increasing W from 4 to 10 per cent improved rupture strength at 1500°F but lowered it at 1700°F. Heats R-5 and R-6, containing 0 and 5.6 per cent Mo respectively, showed a difference in rupture strength, R-5 being superior at testing temperatures of 1500° and 1700°F as plotted in Figure 7. These heats both had high W (10 per cent), however, as compared to the low Mo (4 per cent) in the heats demonstrating an improvement at 1500°F from W additions.

Although these data do indicate that W and Mo affect the properties, variations from the standard S-816 analysis apparently have less of an effect than do variations of C, Cb+Ta, Cr, Co and Ni.

A review of the rupture properties of all the modifications indicated that three of these, R-27, R-9 and R-16, appear promising in comparison to standard S-816. Rupture strengths versus temperature for these are presented in Figure 8. R-26, with 20 per cent Cr and 20 per cent Ni, is included for comparison with R-27 which has 10 per cent Cr and 10 per cent Ni. R-17 is standard S-816. R-27, which appears to be outstanding, was made to combine the analysis variations of R-9 and R-16. It was noted that heat R-27, a borderline alloy of the 70 per cent Ni plus Co maximum, was slightly ferromagnetic. Limited oxidation tests at 1800°F indicated that lower Cr (10 per cent) caused an appreciable drop in oxidation resistance. This should be recognized in consideration of the overall properties of this alloy.

The microstructure of the standard S-816 heat, R-17, as compared with three of the modifications, R-27, R-10, and R-36, is shown in Figure 9. Heats R-10 and R-36, which had high Co, were magnetic as mentioned previously. R-27 and R-36, having lower C and Cb contents do not show the excess insoluble globules of Cb carbide that are present in heats R-17 and R-10. That these carbides act as inhibitors of grain growth is shown by the structures. Heats R-27 and R-36 with lower C have a heavier and more continuous grain boundary constituent than standard S-816. The matrix precipitate is heavier in the three modifications than in S-816 and appears to have occurred along definite lattice planes. This is most pronounced in R-27 and is considered unusual for S-816 or any previous modifications of it.

The comparative rupture strengths shown in Figure 10 indicate that R-27 is superior to commercial S-816 and cast X-40 alloy from 1500° to 1700°F. Only the cast Co-base experimental J-alloy has better properties than R-27. Further study of R-27 is to be carried out on a subsequent Air Force alloy development contract.

<u>U-912 Type Alloys</u>

The three objects of work on this material were to verify former properties, to establish better the composition range, and to lower the critical alloy content of the material. The latter object was given primary emphasis in this work. The results described below indicate that no outstanding modification was obtained.

Listed in Table V are chemical analyses of two standard heats and 11 modifications of U-912 alloy which were investigated in this program.

Rupture and hardness results are presented in Table VI. Testing on smooth bar rupture specimens was done at 1350°F - 38,000 psi and 1500°F - 25,000 psi, which are the approximate 100-hour stress levels of S-816 alloy. Testing at 1500°F - 25,000 psi was used for "V-notch" specimens. Notch properties were obtained because notch sensitivity was believed to be more critical in this type alloy compared to the S-816 type. A comparison of the rupture life of the original U-912 induction heat with those of a 2-ton arc heat and two new U-912 induction heats (R-4 and R-49) is shown graphically in Figure 11. It is seen that the rupture life of the 2-ton heat is appreciably below that of the original 17-pound induction heat U-912.

The two standard heats induction melted in the present program of U-912 were R-4 and R-49. Rupture results on these were similar to those of the arc melted heat as shown in Figure 11. This establishes that the replacement of Cb, in the original U-912 alloy with Cb+Ta was not detrimental. It is believed that these heats have better representative rupture properties than the higher properties of the original U-912 induction heat.

Lowering the Cb to .39 (R-13) did not adversely affect standard rupture life, but did markedly lower ductility and notch rupture life. Replacing 5 per cent Ni with Fe (R-14) lowered the life slightly. A replacement of 5 per cent Cr with Co (R-25) did not affect rupture life but did lower ductility.

A review of the original work on U-912 indicated that a further replacement of Co with Fe might be possible. In heats R-22 to R-24 this was done by increasing the Fe from 12 to 25 per cent. The rupture lives of these heats appear to follow the same trend that was observed in Figure 2 for S-816 alloy, i.e. a decrease in Co gives a trend toward lower rupture life. The notch sensitivity and ductility of U-912 were generally adversely affected by lower Co.

Since Cb is a major element affecting the strategic alloy content, replacement with V was considered as a feasible modification of U-912 for further investigations. V was substituted for Cb+Ta in heat R-52. On the basis of one test only rupture properties were markedly lowered. However, a V substitution for all the Cb+Ta in another modification (R-51), in which part of the C was replaced by N, gave properties equivalent to those of the standard heats (R-4 and R-49). This offers the best possibility for lowering the strategic alloy content of U-912 by the elimination of Cb. A V addition of 2 per cent (R-53) to the standard analysis did not affect properties as was also the case for R-54 in which V replaced half of the 6 W in the basic analysis.

Replacing part of the C in the standard alloy with 0.18 N (R-50) was beneficial to properties. This alloy was the strongest of any studied in the U-912 group.

Hardnesses and response to aging, listed in Table V, were similar to those of the standard with a few exceptions which are relatively insignificant.

Fe-Mn-Cr Alloys

In Table VII will be found the analyses of the 40 17-pound induction heats of Fe-Mn-Cr modifications which were melted, forged and tested. Also listed are the three 1100-pound arc furnace heats which were melted to the analysis of the best modification (designated D-183 alloy) as determined in the experimental alloy studies. The last of these three heats, 9X-129, was considered to be closest to this optimum analysis and will be used for more extensive testing later.

The heat treatment used as a base for comparison of the modifications was 2050°F, 1 hour, water quench plus 1375°F, 16 hours, air cool. This treatment which had been used on a previous contract, offered valuable comparative rupture data which is included in this presentation. Investigation of heat treatments, to be described later, has shown that lower solution and lower aging temperatures gave a better combination of properties. The optimum treatment now appears to be 1900°F, 1 hour, water quench plus 1300°F, 16 hours, air cool.

Room temperature test results for the modifications are listed in Table VIII and rupture results are listed in Table IX.

Throughout this investigation a critical balancing of the C and N contents in addition to heat treatment was necessary to attain optimum properties. The property objectives in this development were as follows:

1. A 100 hour rupture strength at 1200°F of at least 50,000 psi.
2. A ratio of notch to smooth bar rupture life ("N/S ratio") of greater than 1.0.
3. A room temperature .02 per cent offset yield strength of at least 60,000 psi.

4. A stable austenitic structure.

The effects of C and N on the properties of modifications at Mo levels of 0, 1.5 and 3.0 per cent are shown in Figures 12 and 13. The beneficial effects of both C and N on room temperature tensile strength and high temperature rupture strengths are seen. In addition, the detrimental effect on the ratio of notch to smooth bar rupture life with increasing C and N is noted. In both these figures there is an evident increase in strength with higher Mo.

This influence of Mo from 0 to 3 per cent on properties at various C and N levels is further shown in Figure 14. In general Mo up to approximately 1.5 per cent raised the room temperature yield strength and 1200°F rupture strength. Beyond this the effect diminished. The N/S ratio, however, is increased with increasing Mo contents from 1.5 to 3.0 per cent. This latter correlation was the main reason for choosing the 3 Mo level in the optimum final analysis of D-183. Recent heat treatment studies have shown the possibility of increasing N/S ratio by a lower solution temperature. It now appears that further investigation of Mo contents in the range of 1.0 and 2.0 per cent will be worthwhile. A possible reason for Mo contents above 1.5 per cent raising the N/S ratio in the lower C or N alloys is that the added Mo has tended to unbalance the austenite. A similar effect was obtained as the Cr was increased as can be seen by the greater notch ductility of heat D-271 (15 Cr) compared to D-266 (12 Cr). In both these cases, with decreasing austenitic stability a large increase in the N/S ratios was noted. There was, however, no structural evidence of this in the D-183 analysis heats. It is also noted that higher Mo did not appreciably improve the N/S ratio in these higher C and N containing alloys.

The influence of V on properties was determined on three heats, D-333 to D-335.

Previous work (Reference 1) investigating the presence of nitrides in this type alloy indicated that vanadium nitride (VN) was the precipitating phase occurring during aging. Earlier work (Reference 2) done under this contract, indicated from X-ray diffraction studies, that VN was present in both solution treated and solution treated and aged conditions, whereas chromium carbide ($Cr_{23}C_6$) was present only in the aged samples. This suggested that $Cr_{23}C_6$ was the precipitating phase. That V is an essential ingredient for strength, however, is shown by the reduction in rupture strength at 1200°F in the two heats (D-334 and D-335) with low V (.35 per cent) as compared with D-333 with normal V (.78 per cent).

A series of three melts, R-42, R-43 and R-44 were made with 2.0, 4.0 and 6.0 per cent Ni in an effort to produce a more stable austenitic structure. Although apparently Ni did accomplish this, the rupture properties suffered to an extent where this advantage was nullified.

A series of three melts, R-62, R-63 and R-64 were made to investigate the influence on properties and structural stability of Mn from 16 to 22 per cent. This series was prompted by D-218 which had higher than average rupture life, possibly because of its higher Mn content (20.0 per cent). This was not verified as in this new series of melts the 16 Mn had the longest rupture life. Little change occurred in microstructure as seen in Figure 15. There was a tendency toward a larger grain size in the higher Mn heat (R-64) at the solution temperature of 1900°F. This temperature did not effect a complete solution of the phases in the original grain boundaries.

The influence of C on increasing the amount of excess microconstituents is shown in Figure 16, the structure for R-59 being typical for the analysis giving optimum properties.

The results of the study of modifications indicated that an alloy containing .25 to .30 C, .15 to .20 N, 18.0 Mn, 12.0 Cr, 3.0 Mo and .75 V had a good combination of properties. Two 1100 pound heats, 9X-110 and 9X-129, of this analysis, which has been designated D-183 alloy, were melted and processed. Heat 9X-129, the closest to the desired optimum analysis, is to be used for further property evaluations under a subsequent contract and for applications in test turbines possibly as buckets, wheels, and/or sheet.

Processing this heat of D-183 alloy has been completed and material was submitted to Thompson Products for forging to buckets for the Boeing 502 gas turbine.

Microstructures of these two heats are shown in Figure 17. Heat 9X-110, with higher C and N than 9X-129, appears to have a heavier precipitate than does 9X-129.

Limited studies of the effect of treatment on the hardness of the optimum D-183 analysis were made on two 17-pound induction heats, R-45 and R-46. Room temperature hardnesses with varying solution and aging temperatures and times were taken. These results are shown in Figures 18, 19 and 20.

In Figure 18, the effect of solution time and temperature indicates that a solution time of one hour appears sufficient at 1900° and 2050°F. At 1750°F, full solution apparently had not occurred out to a four hour solution time. Hardness drops with higher solution temperatures indicate an increasing solution effect.

Figure 19 shows the effect of aging temperature after solution treatment at three temperatures. The higher the solution temperature, the more rapid the hardness increase during aging. Overaging did not occur until around 1500°F after the two lower solution temperatures. Overaging began around 1400°F after the higher solution temperature. In Figure 20, where the effect of aging time is shown, it is seen that after the 2050°F solution temperature, full hardness is not attained until around 48 hours at aging temperatures of 1200° and 1375°F. After a solution temperature of 1900°F, the 1375°F aging temperature gave full hardness after 16 hours which remains the same out to 48 hours. The hardness during the 1200°F age after a 1900°F solution temperature increased out to 48 hours. These results indicate that the alloy derives at least part of its strength through an aging reaction.

Additional studies on D-183 and some other modifications were conducted in arriving at the treatment giving optimum combination of good room temperature yield strength and a high 1200°F notch ductile rupture strength. Results of these studies on heats having approximately the optimum composition are given in Table X. The hardness data supplement those in Figures 18 through 20. Satisfactory yield strengths coupled with good room temperature properties were obtained after all the treatments investigated. Marked variations with treatment occurred in rupture properties. In general, the lower the solution temperature and/or the higher the aging temperature the more notch ductile the alloy was in rupture. However, solution temperatures either below or above 1900°F generally gave lower smooth bar rupture life. It should be noted that these generalities are based on limited comparative data after only one solution temperature (2050°F) above 1900°F and after three solution temperatures below 1900°F. The optimum treatment appeared to be 1900°F, 1 hour, water quench plus, 1300°F, 16 hours, air cool. This is recommended for the alloy pending further studies to be conducted on a supplementary contract.

Typical properties, including a comparison with those after the 2050°F plus 1375°F treatment used during most of the alloy studies, are summarized in Table XI. It can be seen that the properties outlined as objectives in this work have been obtained from D-183. It offers these properties in combination with relatively low alloy content as a heat treatable Ni-free alloy for service at temperatures around 1200°F. At higher temperatures (1350° and 1500°F) its rupture strength falls off rapidly as indicated by the limited data included in Table X.

CAST ALLOYS

The object of this work was to develop alloys of low strategic content which can be used as nozzle vane and turbine bucket materials in gas turbines. In the search for better alloys for cast nozzles, little work has been done to correlate the high temperature properties with thermal shock.

Eighteen heats were cast of Vitallium, F-88 and F-87 alloys to evaluate their resistance to cracking during rapid temperature cycling between 800°F and some maximum temperature in the order of 1700° to 1900°F. The measure of this resistance is termed "thermal shock life".

F-88 and F-87 were developed previously under this contract (Reference 2) for properties comparable to that of Vitallium but with a lower strategic alloy content. The present study was made to attempt to correlate thermal shock life with composition, rupture properties at 1500° and 1600°F, room temperature properties, impact strength, and microstructure. This work on the cast alloys was done during the period of July 1, 1953 to March 31, 1954.

Cast Alloy Procedures

Melting and Casting

The casting of rupture, room temperature tensile, thermal shock and impact specimens was by the conventional lost wax or investment method of assembled wax patterns encased in a steel mold can which is filled with a hydrolized silicate bonded investment material. The molds are preheated to 1600°F prior to casting and then allowed to cool in air to room temperature prior to shake out.

All melts were made with virgin materials in magnesia crucibles, using a 40 KW Ajax induction furnace. The pouring temperatures were controlled by a high speed electronic recorder using a Pt, Pt-Rh immersion thermocouple protected by a quartz tube. The sand blasted test bars were visually inspected for casting defects.

Rupture Testing

The rupture testing procedure and units were the same as those described in the wrought alloy section. All rupture tests were made on as-cast specimens which have a .252 inch diameter and a 2-inch gage length test section. Most of the tests were conducted at 1500° and 1600°F with a 20,000 psi stress. The fractures were checked after tests to determine if the bars had internal defects which may have affected rupture times and elongations.

Tensile Testing

Room temperature tensile tests were made on the same type of as-cast specimens as those used for rupture testing on a Riehle machine as described under Wrought Alloys.

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 two thermal shock machines are shown in Figures 21 and 22. They consist of a steel framework to which is 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 burner and cooling nozzle back and forth in front of the test piece and turn the air blast off and on. The temperature-time cycle of the test is controlled automatically by the use of a thermocouple clamped on the back of the specimen acting through a high speed Brown Circular Chart Electronik potentiometer. The burner and controller are adjusted to obtain the desired edge temperature which is measured with an optical pyrometer. One of the machines tests one sample at a time, the other two.

Three designs of thermal shock samples as shown in Figure 23 were tested to determine the optimum shape for forming by casting or machining and for giving a reasonable testing time. The curved face sample offered fabricating difficulties while the 60° angle sample did not fracture in a practical time. The 45° angle specimen was decided upon as best.

Thermal shock failures were determined by inspecting the specimens under a 20-power binocular microscope every 500 cycles until cracking began and 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. All efforts to automatically record the failure in the specimen were unsuccessful. For this reason it was only possible to conduct the tests during the working hours of the day, thereby limiting the number of tests which could be run.

The thermal cycle can be varied from room temperature to any temperature up to 2000°F. A typical cycle is from 800° to 1900°F which requires approximately 30 seconds, 25 for heating and 5 seconds for cooling. Tests were conducted from a minimum temperature of 800°F to maximums of 1900°, 1800°, 1750°, and 1700°F.

Impact Testing

The impact specimens were made by cutting six specimens from an investment cast slug. Previous work had indicated that these specimens could not be cast to shape because of a center line segregation. The specimens were machined to a standard "V-notch" charpy test piece. Impact tests were conducted on one heat each of Vitallium and F-87 alloy at room temperature, 1500°, 1700° and 1900°F.

Cast Alloy Results and Discussion

The analysis of the heats tested are shown in Table XII. It is noted that only one intentional composition modification was made in each of the alloys, that of a B addition. In F-88 and F-87 the B added was in replacement of N. The stress rupture and thermal shock data for Vitallium, F-88 and F-87 are shown in Table XIII, XIV and XV respectively. The room temperature tensile properties of the three alloys are given in Table XVI and impact data for Vitallium and F-87 in Table XVII.

Comparative cycles to failure in the thermal shock test on Vitallium, F-88 and F-87 and B modifications are summarized in the following table. These figures represent the average of all the tests run under the indicated conditions.

	Cycles to Failure (800°F to Indicated Maximum Cycle Temperature)			
Alloy	1900°F	1800°F	1750°F	1700°F
Vitallium	2380	6610	6456	--
Vitallium + .20 B	2235	3855	--	14,056
F-88	1315	3530	3090	9,815
F-88 + .09 B	1645	3494	2380	--
F-87	1200	1640	1740	--
F-87 + .14 B	1480	3640	2140	--

These properties indicate that Vitallium tended to have the best resistance to cracking during thermal cycling and F-87 the poorest. The main compositional change which correlates with differences in life is for Co, it being highest in Vitallium and lowest in F-87. At 1900°F, the three alloys had approximately the same life, although Vitallium was somewhat the better. At the lower testing temperatures the shock life varied considerably as shown above and in Figure 24.

Most of the testing was conducted at 1900°F in order to produce the severest shock in the specimen and to reduce the testing time so that as many tests as possible could be made in the program. Although shock testing was limited by the time available it is believed that the results are representative at the higher temperatures. The lower temperature tests were made to evaluate the effect of temperature on properties and to carry the temperature range closer to that of actual operation conditions.

There was a considerable spread in the number of cycles to failure for tests made on specimens from the same heat. It is felt that this spread was due in part to the inherent variability in castings and also to variations connected with development of techniques for this new type of test. These variations include surface preparation, testing temperature variations, and the method of inspection for failure.

The addition of B to these alloys did not increase the thermal shock life except in the F-87 alloy with .14 B. The shock life of most of the alloys increased with lowering temperature except between 1800° and 1750°F when the life decreased, as shown in Figure 24. This effect may be due to the carbide precipitation or an overaging effect that weakened the alloys in this temperature range.

A survey of the results indicates that there is no correlation between the thermal shock properties and the rupture, room temperature and/or microstructural characteristics. The B modified alloys with the high rupture strengths have about the same thermal shock properties as the same alloys without B. This increase in rupture strength for the B modified alloys is shown in the following table of 100-hour rupture strengths at 1500° and 1600°F.

Alloy	100 Hour Rupture Strengths (psi) 1500°F	1600°F
Vitallium	19,000	17,000
Vitallium + .20 B	28,000	--[a]
F-88	19,000	16,000
F-88 + .09 B	25,000	--[a]
F-87	18,000	11,000
F-87 + .14 B	25,000	21,000

The limitations on the program did not permit a study of systematic compositional variations other than B on the thermal shock properties. However, one heat of Vitallium, F-357, with a lower C content (.26 per cent) did have a lower shock life than similar heats with the nominal .40 C.

A study of the rupture ductility of the tests at 1600°F under 20,000 psi of each alloy compared to the thermal shock properties indicates that for heats with higher rupture elongation a greater number of cycles to failure were obtained. Further investigation would be necessary to determine the generality of this trend.

Room temperature and high temperature impact tests were made to determine if this property could be correlated to the thermal shock properties. The results of these tests are shown in Table XVII and Figure 25. The impact strength of Vitallium was higher, as was also the thermal shock life, than those of F-87. Further testing is needed for verifying the significance of this trend.

A microstructural examination of as-cast and fractured specimens revealed that all of the fractures were transgranular and per-

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a - Insufficient data available

pendicular to the test edge with the exception of one of several Vitallium fractures which was intergranular. Microstructures are shown in Figures 26, 27 and 28. Considerable aging occurred during the testing cycle with a heavy precipitation of carbides in the matrix. Figure 29 shows the increase in hardness on the test edge in Vitallium and F-87 after the specimen has been cycled to failure. The peaks on the hardness curves are believed to be caused by an overaging effect in the alloys. Most of the cracks occur on or near the low hardness areas in the center of the test piece. No explanation for differences in thermal shock properties could be found in the microstructural studies.

CONCLUSIONS

Given below are the conclusions drawn from the described investigations.

S-816 Alloys

On wrought Co-base alloys of the S-816 type:

1. Lower Cb+Ta and C contents appear to give as high or higher rupture life than that of the standard S-816 alloy at test temperatures above 1500°F. This may, however, be overshadowed by other offsetting factors such as a lower rupture ductility.
2. Co contents up to 60 per cent, above the S-816 level of 43 per cent, appear to increase the rupture strength. A limit to this increase is apparently a combined Ni + Co content of 70 per cent.
3. Variations in Mo and W content from the standard S-816 levels of 4 per cent have less of an effect than the above

two variables.

4. The modification of S-816 which had the highest rupture strength was an alloy, R-27, containing .10 C, 10 Cr, 10 Ni, 60 Co, 10 W, 5.0 Mo, 1.0 Cb+Ta and 2.0 Fe. Low oxidation resistance at this Cr level may offset the advantage of high strength in this alloy.

U-912 Alloys

On wrought Co-base alloys of the U-912 type:

1. No outstanding modification of U-912 was obtained. Apparently the properties of U-912 alloy are better represented by the new experimental heats, than by the original experimental heat U-912 which had somewhat high strength.
2. Substitution of N for part of the C in one heat resulted in the best alloy in this study.
3. Replacement of Cb+Ta with V in a heat containing high N did not alter rupture life and appeared to offer the best opportunity of lowering the critical alloy content of U-912.

Fe-Mn-Cr Alloys

On wrought Fe-base alloys of the Mn-Cr type:

1. A critical balancing of C and N contents was found to be necessary in maintaining a stable austenitic structure without impairing the ductility or notch rupture life of Fe-Mn-Cr alloys.
2. Mo, up to 1.5 per cent gives an increase in rupture strength while additional Mo up to 3 per cent causes little further improvement. The 3 Mo alloy, however, is best because of

its higher notch rupture life.

3. V has an important strengthening influence.
4. The composition giving the optimum combination of properties contains .25 to .30 C, .15 to .20 N, 18.0 Mn, 12.0 Cr, 3.0 Mo and .75 V. This composition has been designated D-183 alloy.
5. An aging reaction is apparently an important factor in obtaining the outstanding strength of this alloy.
6. The optimum heat treatment apparently is solution at 1900°F for 1 hour, water quench, plus aging at 1300°F for 16 hours air cool.
7. Large ingots from arc-furnace heats of D-183 alloy apparently have poor hot workability in the initial stages. Workability in subsequent reductions is good. Wrought material from such heats has equivalent properties to those of induction heats.

Cast Alloys

On cast Vitallium, F-88, F-87 and their B modifications:

1. The shock lives of the three alloys are almost the same when 1900°F is used as the maximum cycle temperature. Vitallium has the best thermal shock life, especially at the lower temperatures. F-88 has better shock life than F-87 especially at temperatures below 1900°F.
2. B additions to the three cast alloys did not improve shock life except in the case of F-87. However, the B additions did improve rupture strengths.
3. An apparent compositional effect was that lowering C decreased both the rupture strength and the thermal shock

life.

4. There appears to be no correlation of the thermal shock properties with rupture, room temperature or microstructural properties of the cast alloys tested.

REFERENCES

(1) W. J. Robinson: Alloys for Use at High Temperatures, - Report on Visit to Germany and Austria. Mapleton House, New York, 1947.

(2) R. K. Pitler and W. W. Dyrkacz: Casting and Forging Turbine Bucket Alloys. WADC Technical Report 53-274, December 1953.

TABLE I

Chemical Analyses of S-816 Alloy Modifications

(Major modification underlined)

Chemical Analysis (weight percent)

Heat No.	C	Si	Mn	Cr	Ni	Co [a]	Fe	W	Mo	Cb+Ta
S-816	.38	.40	1.20	20.00	20.00	43.0	4 max	4.0	4.00	4.00
R-5	.26	.42	1.56	20.24	19.25	43.9	1.81	9.96	--	2.60
R-6	.27	.40	2.32	20.08	19.70	36.8	2.07	9.86	5.65	2.80
R-7[b]	.25	.40	1.20	20.00	20.00	Bal	4 max	10.00	8.00	2.00
R-8	.27	.38	2.22	19.96	19.82	34.1	2.20	6.28	11.72	3.04
R-9	.26	.38	1.15	10.56	9.80	57.3	2.07	10.36	5.05	3.02
R-10	.27	.32	1.21	10.35	10.17	61.4	3.31	10.28	0	3.01
R-11	.30	.54	1.04	24.74	20.02	37.6	1.70	11.18	--	2.84
R-12[b]	.25	.40	1.20	25.00	20.00	Bal	4 max	10.00	4.00	2.00
R-15	.14	.38	1.13	22.77	22.47	41.4	1.23	4.27	5.58	.58
R-16	.13	.36	1.06	20.08	19.94	47.5	1.16	3.76	5.15	.85
R-17	.41	.45	.69	20.08	20.56	43.9	2.00	4.03	4.40	3.42
R-26	.11	.29	1.08	20.26	21.52	38.7	.97	10.24	5.00	.83
R-27	.11	.25	1.20	10.44	9.26	60.6	.97	10.22	5.00	.87
R-33	.34	.46	1.22	20.06	20.39	43.2	1.95	4.49	4.45	3.46
R-34	.13	.47	1.20	20.52	20.15	43.0	2.21	4.09	4.64	3.54
R-35	.13	.34	1.21	20.04	20.15	46.6	1.30	4.18	4.78	1.28
R-36	.13	.38	1.10	9.92	19.74	58.0	1.04	4.45	4.13	1.19
R-37	.13	.39	1.13	19.91	29.86	37.9	1.04	3.99	4.46	1.20
R-38	.12	.35	1.17	19.67	39.81	28.6	.91	4.07	4.30	1.00
R-39	.37	.50	1.11	19.99	39.40	24.3	1.82	3.83	4.74	3.91
R-55	.14	.40	1.16	20.20	10.03	50.4	1.30	10.15	5.40	.82
R-56[b]	.13	.40	1.20	30.00	10.00	Bal	4 max	10.00	5.00	1.00
R-65	.10	.33	1.10	19.94	--	68.2	1.16	3.87	4.28	1.06
R-66	.10	.33	1.17	20.02	10.15	58.2	.97	4.01	4.13	.88
R-67	.42	.33	1.18	19.78	20.05	48.0	.96	4.06	4.23	.96

a - By difference
b - Melting aim analysis - these heats cracked up during forging.

TABLE II

Rupture and Hardness Properties of S-816 Alloy Modifications

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

Heat No.	Temp. (°F)	Stress (psi)	Rupture Time (Hrs.)	Elong. (%)	Red. of Area (%)	Brinell Hardness Solution Treated	Brinell Hardness Solution Treated + Aged
R-5	1500	25,000	246	58.7	81.5	255	286
	1700	10,000	239	12.9	32.8		
	1700	15,000	41	26.5	46.0		
R-6	1500	25,000	90	36.6	72.5	286	293
	1500	25,000	76	85.2	72.0		
	1700	10,000	114	28.5[b]	58.2		
R-8	1500	25,000	56	51.6	73.7	298	321
	1700	10,000	96	75.1	77.0		
	1700	10,000	148	55.9	71.6		
R-9	1500	25,000	108[a]	6.2	5.2	286	321
	1500	25,000	510	22.5	32.1		
	1500	25,000	655	16.0	28.0		
	1500	30,000	200	25.8	28.8		
	1600	20,000	79	23.8	33.4		
	1700	10,000	362	13.4	18.7		
	1700	10,000	298	15.1	19.2		
	1700	15,000	69	21.3	25.8		
R-10	1500	25,000	110	24.9	37.0	277	293
	1700	10,000	108	11.1	17.4		
R-11	1500	20,000	1638	10.2	29.0	255	286
	1500	25,000	160	24.6	39.4		
	1500	30,000	26	32.7	46.0		
	1600	20,000	61	17.3	34.0	255	286
	1700	10,000	335	27.8	33.2		
	1700	15,000	41	16.1	35.2		
R-15	1500	25,000	49	12.9	17.0	196	248
	1500	25,000	65	18.8	20.0		
	1500	25,000	64	18.9	18.8		
	1700	10,000	168	15.5	23.0		
	1700	10,000	218	15.3	17.8		

a - Bad fracture
b - Estimated

TABLE II (Continued)

Heat No.	Temp. (°F)	Stress (psi)	Rupture Time (Hrs.)	Elong. (%)	Red. of Area (%)	Brinell Hardness Solution Treated	Brinell Hardness Solution Treated + Aged
R-16	1500	20,000	668	7.6	14.0	217	228
	1500	25,000	134	15.0	22.0		
	1500	25,000	76	18.0	19.0		
	1600	20,000	20	15.5	20.9		
	1600	20,000	101	6.3	9.3		
	1700	10,000	957	11.5[b]	13.1[b]		
	1700	10,000	525	10.7	19.5		
	1700	15,000	108	6.4	12.7		
R-17	1500	20,000	1089	31.2	60.5	255	302
	1500	25,000	183	32.7	55.9		
	1600	20,000	51	20.8	42.3		
	1600	20,000	55	23.8	45.0		
	1700	7,500	441+[c]	21.8	40.5		
	1700	10,000	249	20.9	39.2		
	1700	15,000	29	19.7	38.6		
R-26	1500	20,000	1439	5.1	11.9	228	248
	1500	25,000	136[a]	23.4	27.2		
	1500	25,000	170	39.3	50.5		
	1500	30,000	37	45.5	56.0		
	1700	10,000	470	32.0	47.2		
	1700	15,000	26	27.7	43.4		
	1700	15,000	32	32.7	46.0		
R-27	1500	25,000	1028	19.3	26.5	255	300
	1500	30,000	130	21.7	23.9		
	1500	35,000	38	18.2	20.3		
	1700	10,000	274	7.6	15.4		
	1700	10,000	273	--	19.2		
	1700	15,000	80	14.9	27.2		
	1700	18,000	49	32.8	35.0		
R-33	1500	30,000	34	40.6	53.5	269	302
R-34	1500	25,000	61	26.1	27.6	217	262
	1700	10,000	73	27.0	24.0		
	1700	15,000	5.3	26.7	26.0		
R-35	1500	25,000	171	19.9	29.2	212	228
	1500	30,000	37	27.5	30.6		
	1700	10,000	287	11.7	15.6		
	1700	15,000	42	6.2	10.4		

a - Bad fracture
b - Estimated
c - Overheated to 2100°F - ruptured

TABLE II (Continued)

Heat No.	Temp. (°F)	Stress (psi)	Rupture Time (Hrs.)	Elong. (%)	Red. of Area (%)	Brinell Hardness Solution Treated	Brinell Hardness Solution Treated + Aged
R-36	1500	25,000	77	10.2	11.5	217	255
	1700	10,000	56	5.3[b]	5.5		
R-37	1500	25,000	99	17.5	26.3	196	223
	1700	10,000	222	15.3	18.2		
	1700	15,000	15	10.4	20.2		
R-38	1500	25,000	29	17.4	23.9	187	228
	1700	10,000	143	16.0	26.0		
	1700	10,000	166	17.2	18.9		
	1700	15,000	12	22.1	24.2		
R-39	1500	25,000	51	22.9	28.7	241	269
	1500	25,000	53	16.2	30.0		
	1700	10,000	171	35.2	49.5		
	1700	15,000	12	28.2	52.3		
R-55	1500	25,000	195	27.6	54.4	277	311
	1500	30,000	33	37.5	58.2		
	1700	10,000	311	23.2[b]	32.0		
	1700	15,000	35	39.4	49.5		
R-65	1500	25,000	38	17.2	16.3	286	332
	1500	30,000	13	17.2	14.9		
	1700	10,000	328	10.4	8.6		
	1700	15,000	16	17.0	18.9		
	1800	10,000	22	7.7	12.3		
R-66	1500	25,000	84	16.7	22.0	255	286
	1500	30,000	37	27.8	29.0		
	1700	10,000	487	4.8	11.2		
	1700	15,000	49	10.8	12.7		
	1800	10,000	17	7.3	11.5		
R-67	1500	25,000	123	2.0	2.0	241	351

b - Estimated

TABLE III

Effect of Treatment on Rupture Properties of S-816 Alloy Modifications

Heat No.	Heat Treatment [a]	Test Temp. (°F)	Stress (psi)	Rupture Time (Hrs.)	Elong. (%)	Red. of Area (%)	Brinell Hardness: Solution Treated	Brinell Hardness: Solution Treated + Aged
R-5	A	1700	10,000	124	37.5	44.8	269	277
	B	1700	10,000	239	12.9	32.8	255	286
R-6	A	1700	10,000	29[b]	19.6	26.3	302	302
	B	1700	10,000	114	28.5[c]	58.2	286	293
R-8	A	1700	10,000	38	83.1	84.0	302	311
	B	1700	10,000	96	75.1	77.0	298	321
	B	1700	10,000	148	55.9	71.6		
R-11	A	1700	10,000	204	45.0	60.0	269	286
	B	1700	10,000	335	27.8	33.2	255	286
R-16	A	1500	25,000	75	22.5	34.0	196	217
	B	1500	25,000	134	15.0	22.2	217	228
	B	1500	25,000	76	18.0	19.0		
	A	1600	20,000	39	25.0	41.4		
	B	1600	20,000	20	15.5	20.9		
	B	1600	20,000	101	6.3	9.3		
R-17	A	1500	25,000	115	73.6	68.0	262	293
	B	1500	25,000	183	32.7	55.9	255	302
	A	1600	20,000	143	33.0	54.0		
	B	1600	20,000	55	23.8	45.0		

a - A - 2150°F-1 hr-W. Q. + 1400°F-16 hrs- A. C.
B - 2250°F-1 hr-W. Q. + 1400°F-16 hrs- A. C.

b - Bad fracture - center burst

c - Estimated

TABLE IV

Rupture Strengths of S-816 Modifications at 1500, 1600 and 1700°F

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

Standard S-816 Analysis:

C	Mn	Si	Cr	Ni	Co	W	Mo	Ta+Cb	Fe
.38	1.20	.40	20.0	20.0	Bal	4.0	4.0	4.0	4 Max

Heat No.	Rupture Strength (1000 psi) 1500°F 100 Hr.	1500°F 1000 Hr.	1600°F 100 Hr.	1700°F 100 Hr.	Analysis Modifications (%)
R-5	28.0	20.5	19.5[a]	12.5	C-.26; W-10.0; Mo-0; Ta+Cb-2.6
R-6	24.5	--	16.0[a]	10.5	C-.27; W-9.9; Mo-5.7; Ta+Cb-2.8
R-8	23.2	--	15.5[a]	10.5	C-.26; W-6.3; Mo-11.7; Ta+Cb-3.0
R-9	33.8	22.5	21.5	13.5	C-.26; Cr-10.6; Ni-9.8; W-10.4; Ta+Cb-3.0
R-10	25.0	--	--	10.5	C-.27; Cr-10.5; Ni-10.2; W-10.4; Mo-0; Ta+Cb-3.0
R-11	26.4	19.7	19.1[a]	13.3	C-.30; Cr-24.7; W-11.2; Mo-0; Ta+Cb-2.8
R-15	24.0	--	17.5[a]	11.5	C-.14; Cb+Ta-.6
R-16	25.1	--	19.4	14.6	C-.13; Cb+Ta-.9
R-17	27.5	19.4	18.8	12.4	Standard S-816
R-26	26.8	19.0	18.8[a]	13.2	C-.11; W-10.2; Ta+Cb-.8
R-27	31.3	23.5	22.9	14.8	C-.11; Cr-10.4; Ni-9.3; W-10.2; Ta+Cb-.9
R-34	23.0	--	14.8[a]	9.3	C-.13

a - Estimated from Larson & Miller parameter curves.

TABLE IV (Continued)

Heat No.	Rupture Strength (1000 psi) 1500°F 100 Hr.	1500°F 1000 Hr.	1600°F 100 Hr.	1700°F 100 Hr.	Analysis Modifications (%)
R-35	26.8	--	19.5[a]	13.0	C-.13; Ta+Cb-1.3
R-36	24.4	--	--	9.2	C-.13; Cr-9.9; Ta+Cb-1.2
R-37	24.8	--	--	11.3	C-.13; Ni-29.9; Ta+Cb-1.2
R-38	21.8	--	--	10.8	C-.12; Ni-39.8; Ta+Cb-1.0
R-39	23.0	--	--	11.0	Ni-39.4
R-55	27.0	20.0	19.3[a]	12.6	C-.14; Ni-10.0; W-10.2; Ta+Cb-.8
R-65	22.9	--	16.2[a]	11.6	C-.10; Ni-0; Ta+Cb-1.1
R-66	26.0	--	19.3[a]	13.3	C-.10; Ni-10.2; Ta+Cb-.9

a - Estimated from Larson & Miller parameter curves

TABLE V

Chemical Analyses of U-912 Alloy Modifications

(Major modification underlined)

Chemical Analysis (weight percent)

Heat	C	Si	Mn	Cr	Ni	Co [a]	Fe	W	Mo	Ta+Cb	Cb	Others
U-912	.31	.40	1.20	20.00	20.00	Bal	12.00	6.00	3.2	--	1.8	--
R-4	.33	.52	1.20	19.80	19.90	32.4	13.57	5.84	3.45	3.02	--	--
R-13	.35	.38	1.10	20.16	21.51	33.3	12.79	6.32	3.75	--	.39	--
R-14	.35	.48	1.06	19.52	<u>15.22</u>	30.3	<u>21.83</u>	5.56	3.70	2.02	--	--
R-22	.32	.18	.96	<u>15.00</u>	20.51	<u>26.5</u>	<u>25.84</u>	5.98	2.93	1.80	--	--
R-23	.34	.14	.99	<u>14.55</u>	20.43	<u>25.1</u>	<u>26.54</u>	<u>10.15</u>	<u>.10</u>	1.67	--	--
R-24	<u>.20</u>	.18	.97	23.61	20.19	<u>18.5</u>	<u>26.52</u>	6.05	2.88	<u>.89</u>	--	--
R-25	.31	.60	1.06	<u>14.51</u>	20.27	36.2	15.60	6.55	2.86	2.03	1.83	--
R-49	.32	.51	1.00	19.76	20.14	34.1	12.09	6.14	2.76	1.82	--	--
R-50	<u>.12</u>	.44	.95	19.51	20.02	33.9	12.87	6.18	2.75	1.70	--	<u>N-.18</u>
R-51	<u>.12</u>	.35	.92	20.36	19.66	37.1	12.66	5.00	2.74	<u>--</u>	--	N-<u>.18</u>, V-<u>1.96</u>
R-52	.30	.31	.95	19.35	20.24	35.8	11.96	6.12	2.76	<u>--</u>	--	V-<u>2.20</u>
R-53	.31	.54	.95	19.76	19.92	31.8	12.68	6.56	2.82	1.42	--	V-<u>2.11</u>
R-54	.29	.52	1.24	19.67	20.20	37.0	13.25	<u>3.12</u>	2.56	1.88	--	V-<u>2.26</u>

a - By difference

TABLE VI

Rupture Properties of U-912 Alloy Modifications

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

	C	Si	Mn	Cr	Ni	Co	Fe	W	Mo	Cb or Ta+Cb
Standard Analysis:	.30	.4	1.2	20	20	35	12	6	3.2	1.8

Heat No.	Analysis Variations	Test Temp (°F)	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness Solution Treated	Solution Treated + Aged
R-4	Standard	1350 1500	38,000 25,000	-- 659	127 62	35.6 37.2	75.0 65.5	---	262
R-13	.39 Cb	1350 1500	38,000 25,000	-- 10/33	105 78	2.9 2.8	6.3 8.9	---	340
R-14	15-Ni, 22-Fe	1350 1500	38,000 25,000	-- 378	58 35	23.0 19.0	45.5 41.1	---	255
R-22	15-Cr, 27-Co, 26-Fe	1500	25,000	25	24	13.1	24.5	217	244
R-23	15-Cr, 25-Co, 26-Fe 10-W, .10-Mo	1500	25,000	68	29	22.9	26.6	207	241
R-24	.20-C, 19-Co, 26-Fe .89-Ta+Cb	1500	25,000	5	40	7.8	10.4	183	259
R-25	15-Cr	1500 1500	25,000 25,000	-- --	48 47	3.9 7.5	10.9 8.0	217	277
R-49	Standard	1500 1500	25,000 25,000	-- 46	60 45	6.5 20.3	7.3 23.8	255	262

TABLE VI (Continued)

Heat No.	Analysis Variations	Test Temp (°F)	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Rupture Time (Hrs.) Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness Solution Treated	Brinell Hardness Solution Treated + Aged
R-50	.12-C, .18-N	1500	25,000	628	94	23.9	31.5	241	286
R-51	.12-C, 0-Ta+Cb, .18-N, 2.0 V	1500	25,000	211	49	23.6	24.7	241	286
R-52	0-Ta+Cb, 2.2 V	1500	25,000		12	6.0	7.5	228	302
R-53	2.1-V	1500	25,000	73	67	32.5	38.0	255	286
R-54	3.1-W, 2.3-V	1500	25,000	145	40	18.7	33.8	241	286

TABLE VII

Chemical Analyses of Iron-Manganese-Chromium Alloy Modifications

(Major modification underlined)

(Chemical Analysis (weight percent)

Heat	C	Si	Mn	Cr	V	N	Mo	Others
D-182	.25	.19	17.18	12.21	.80	.195	2.64	
D-183	.35	.21	17.18	12.25	.86	.189	3.12	
D-218	.24	.28	19.98	12.17	.89	.164	2.84	
D-250	.29	.24	18.10	12.33	.53	.312	--	
D-251	.10	.48	16.80	12.37	1.04	.181	3.45	
D-252	.09	.24	17.16	12.21	.94	.200	1.52	
D-253	.07	.14	17.51	11.77	.15	.173	1.64	
D-254	.04	.26	17.49	12.65	.73	.299	1.54	
D-255	.08	.22	17.29	12.49	.94	.195	--	
D-256	.05	.18	17.46	12.25	.22	.197	--	
D-257	.09	.24	17.07	12.29	1.04	.192	--	
D-258	.06	.24	17.31	12.49	1.57	.176	--	
D-259	.07	.24	17.42	15.26	.23	.207	--	
D-260	.08	.22	17.13	12.25	.90	.215	--	
D-264	.20	.24	17.91	11.81	.80	.033	1.76	
D-265	.28	.25	17.79	12.24	.95	.035	1.74	
D-266	.38	.19	18.10	11.53	.98	.031	1.37	
D-267	.37	.16	18.20	12.18	.95	.156	1.74	
D-271	.38	.24	17.75	15.24	.89	.039	1.74	
D-333	.30	.26	18.04	12.21	.78	.029	3.18	
D-334	.30	.20	17.78	11.85	.30	.015	1.64	
D-335	.28	.23	17.78	12.49	.35	.022	3.22	

TABLE VII (Continued)

Heat	C	Si	Mn	Cr	V	N	Mo	Others
R-18	.56	.21	17.10	12.42	.69	.025	1.35	
R-19	.20	.19	17.27	12.01	.66	.118	1.36	
R-20	.44	.13	17.20	12.20	.56	.023	2.92	
R-21	.51	.15	17.78	12.50	.55	.026	--	W-2.82
R-40	.32	.25	18.22	12.11	.77	.094	1.55	
R-41	.38	.21	17.66	12.31	.62	.090	2.99	
R-42	.27	.24	17.38	12.17	.64	.160	2.99	Ni-2.09
R-43	.18	.26	17.69	12.19	.67	.180	3.03	Ni-4.15
R-44	.25	.26	17.60	12.11	.64	.187	3.00	Ni-6.17
R-45	.30	.24	17.36	12.19	.74	.138	3.04	
R-46	.31	.25	17.66	12.04	.70	.165	2.85	
R-47	.30	.26	17.66	11.97	.68	.153	2.88	
R-48	.27	.25	17.60	12.07	.73	.174	2.81	
R-57	.26	.18	17.38	12.18	.72	.227	2.96	Al-.12
R-58	.31	.37	17.30	12.10	.74	.224	2.98	Al-.21
R-59	.28	.13	16.41	13.20	.74	.108	3.08	
R-60	.41	.26	17.98	12.44	.73	.102	3.17	
R-61	.48	.23	17.70	12.18	.76	.094	3.22	
R-62	.24	.19	16.03	12.08	.71	.154	3.20	
R-63	.24	.27	19.89	12.18	.71	.148	3.17	
R-64	.25	.35	21.81	12.14	.74	.152	3.20	
9X108[a]	.34	.20	15.40	12.67	.75	.231	2.63	
9X110[a]	.36	.54	18.88	12.43	.81	.24	3.09	Ni-.78
9X129[a]	.29	.30	18.06	12.48	.70	.156	2.97	Ni-.72

a - 1100-lb. arc furnace heats. All others were 17-lb. induction heats

TABLE VIII

Room Temperature Tensile Properties of Iron-Manganese-Chromium Alloy Modifications

Heat No.	Tensile Strength (psi)	Yield Strength (psi) 0.2% Offset	Yield Strength (psi) .02% Offset	Elong. (%)	Red. of Area (%)
2050°F-1 hr.-W. Q. + 1375°F-16 hrs.-A. C.					
D-182	152,500	100,200	80,900	33.0	30.8
D-183	163,000	104,500	81,200	24.0	19.8
D-218	145,000	98,000	76,500	20.0	17.4
D-250	134,100	71,600	56,400	15.4	18.8
D-251	111,200	62,900	42,000	9.1	6.2
D-252	128,000	76,100	65,800	43.2	47.8
D-253	105,200	44,600	25,200	64.5	66.9
D-254	135,400	76,100	59,300	39.9	38.9
D-255	120,000	52,100	44,400	60.0	58.7
D-256	108,700	43,400	32,200	63.7	64.4
D-257	126,200	60,500	48,000	53.2	51.3
D-258	116,500	40,300	30,000	54.6	56.7
D-259	110,200	53,800	39,400	44.0	44.3
D-260	122,200	56,200	33,600	56.5	51.3
D-264	113,800	46,200	36,000	43.6	47.3
D-265	118,000	54,000	45,600	41.4	38.5
D-266	127,500	60,900	53,400	32.7	32.7
D-267	142,200	100,000	65,600	13.5	10.9
D-271	126,000	59,000	44,900	35.0	21.7
D-333	121,100	59,800	45,400	40.7	36.6
D-334	122,000	48,000	40,300	41.5	30.7
D-335	118,100	49,800	37,600	38.4	27.5
R-18	134,100	--	--	18.5	18.7
R-18	136,800	68,500	55,700	21.0	21.9
R-19	114,700	56,700	48,000	57.4	52.6
R-20	136,000	65,000	44,600	24.0	22.6
R-20	135,300	63,800	48,120	24.5	20.7
R-21	132,000	60,000	43,200	23.0	23.2
R-40	129,000	67,400	56,400	39.5	39.6
R-41	143,900	73,800	53,700	24.4	24.7
R-42	128,500	72,600	60,600	35.5	32.9
R-43	120,000	71,200	59,970	42.0	45.8
R-44	142,500	80,100	66,000	31.5	41.7
R-47	137,600	80,800	65,500	30.0	25.0
R-48	144,200	90,000	75,000	31.0	29.4

TABLE VIII (Continued)

Heat No.	Tensile Strength (psi)	Yield Strength (psi) 0.2% Offset	Yield Strength (psi) .02% Offset	Elong. (%)	Red. of Area (%)
1850°F-2 hrs.-W. Q. + 1300°F-16 hrs.-A. C.					
9X-110[a]	141,200	94,900	61,200	5.5	9.7
1900°F-1 hr.- W. Q. + 1200°F-16 hrs.- A. C.					
D-182	137,600	90,000	77,300	44.0	47.1
D-183	146,100	95,000	77,100	34.0	35.2
D-218	143,000	100,500	86,000	29.0	33.2
R-41	125,000	72,700	61,000	22.0	21.3
R-47	137,600	80,800	65,500	30.0	25.0
R-48	144,200	90,000	75,000	31.0	29.4
1900°F-1 hr.- W. Q. + 1300°F-16 hrs.- A. C.					
9X-110[b]	154,500	97,800	74,700	21.5	18.2
9X-129[b]	136,800	81,600	62,700	33.5	23.4
2050°F-1 hr.-W. Q. + 1325°F-16 hrs.-A. C.					
D-264	111,000	45,600	38,400	60.7	56.5
D-266	124,500	58,100	41,600	48.8	36.0
2050°F-1 hr.- W. Q. + 1400°F-16 hrs.-A. C.					
D-182	142,500	93,400	72,700	16.4	16.7
D-183	146,000	84,700	62,400	12.2	11.5
2150°F-1 hr.-W. Q. + 1325°F-16 hrs.-A. C.					
D-264	112,600	43,200	33,400	52.0	48.3
D-266	122,100	56,300	46,400	48.0	35.3
2150°F-1 hr.-W. Q. + 1375°F-16 hrs.-A. C.					
D-264	118,100	42,300	32,100	48.8	34.6
D-266	129,700	58,600	49,100	64.0	30.0

a - 50-lb. pilot ingot - forged to 5/8 inch square
b - 1100-lb. ingot - rolled to 7/8 inch round

TABLE IX

Rupture Properties at 1200°F of Iron-Manganese-Chromium Alloy Modifications

Heat No.	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Rupture Time (Hrs.) Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness after Heat Treatment
2050°F-1 hr.-W. Q. + 1375°F-16 hrs.-A. C.						
D-182	45,000	--	973	5.2	5.1	311
	50,000	28	263	6.5	11.5	
	60,000	--	a	23.1	62.9	
D-183	45,000	--	886	5.4	5.1	321
	50,000	97	240	4.0	7.8	
	60,000	--	0.5	23.6	58.7	317
D-218	45,000	452	1311	3.7	9.0	311
	50,000	--	116	21.4	47.2	311
	55,000	43	165	5.6	9.0	
D-250	40,000	--	272	11.9	25.7	286
	45,000	33/4	33	10.8	12.8	
D-251	50,000	--	4	25.3	30.8	302
	55,000	14	1.5	21.4	27.1	
D-252	50,000	--	17	5.6	12.4	277
	55,000	5	9	2.5	8.0	
D-253	50,000	5	16	5.0	9.9	196
	55,000	--	a	42.5	70.5	
D-254	50,000	52	18	11.6	27.0	293
	55,000	--	7	18.3	28.7	
D-255	35,000	--	179	4.5	4.7	241
	40,000	--	52	4.7	9.9	
	45,000	21/30	18	10.3	16.4	
D-256	35,000	--	12	26.7	39.5	202
	40,000	--	6	37.8	46.7	196
	45,000	17	1	48.1	57.5	

a - Broke on loading

Contrails

TABLE IX (Continued)

Heat No.	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness after Heat Treatment
2050°F-1 hr.-W. Q. + 1375°F-16 hrs.- A. C. (Continued)						
D-257	40,000	--	141	3.5	9.9	255
	45,000	26	17	6.2	14.8	
D-258	40,000	120	13	23.3	44.5	223
	45,000	63/83	1.3	30.6	54.4	
D-259	40,000	81	57	8.7	10.3	241
	45,000	29	9	14.7	16.2	
	50,000	--	1	24.2	27.1	
D-260	40,000	--	113	4.3	12.3	248
	45,000	65	19	10.3	22.2	
D-264	40,000	--	301	31.2	57.5	196
	45,000	762				
	50,000	--	44	17.5	23.8	
	55,000	48	20	20.0	22.0	
D-265	50,000	--	99	13.7	23.2	207
	55,000	61/81	40	16.7	15.6	
D-266	50,000	70	111	11.6	9.9	217
	55,000	38/48	36	9.7	12.3	
D-267	50,000	--	138	3.0	5.4	311
	55,000	7/5	54	--	5.0	
D-271	45,000	--	186	31.3	58.7	255
	50,000	--	39	3.8	25.7	
	55,000	570/205	23	30.3	49.0	
D-333	45,000	--	112	--	49.0	228
	50,000	765	123	20.2	50.5	
	55,000	--	41	22.3	46.2	
D-334	45,000	424	58	30.5	40.0	196
	50,000	--	3	35.5	40.5	
	55,000	--	0.3	32.5	57.5	

a - Broke on loading

TABLE IX (Continued)

Heat No.	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness after Heat Treatment
2050°F-1 hr.-W. Q. + 1375°F-16 hrs.-A. C. (Continued)						
D-335	45,000	446	2.3	23.8	52.5	217
	50,000	--	15	22.1	31.7	
	55,000	--	2.3	26.8	37.6	
R-18	50,000	128	140	13.6	18.5	269
	55,000	--	53	9.6	12.5	
R-19	50,000	22	a	40.5	69.3	207
	55,000	--	3.5	17.0	35.6	
R-20	45,000	--	196	24.3	55.0	258
	50,000	541	68	27.0	50.0	
R-21	45,000	--	88	10.8	23.9	252
	50,000	19	37	14.1	23.4	
R-40	50,000	21	71	10.2	10.4	248
	55,000	--	32	11.5	12.9	
R-41	50,000	--	59	18.1	42.3	286
	55,000	186	66	15.3	22.6	
R-42	50,000	28	39	4.1	9.9	255
	55,000	--	20	4.9	5.7	
R-43	50,000	23	33	7.7	8.7	241
	55,000	--	6.3	8.2	13.6	
R-44	50,000	66	52	15.8	16.9	255
	55,000	--	11	10.6	17.6	
R-47	55,000	48	55	7.0	13.7	277
R-48	50,000	--	109	3.2	9.0	286
	55,000	19	33	7.1	10.0	
9X-108[b]	55,000	--	39	3.7	8.9	321
9X-110[c]	55,000	47	51	13.5	24.0	311

a - Broke on loading
b - 50-lb. pilot ingot - forged to 5/8 inch square
c - 1100-lb. ingot - rolled to 7/8 inch round

TABLE IX (Continued)

Heat No.	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness after Heat Treatment
As forged + 1375°F-16 hrs.-A. C.						
9X-110[b]	55,000	--	29	9.5	27.3	351
As forged + 1300°F-16 hrs.-A. C.						
9X-129[d]	55,000	--	107	2.5	3.2	332
1750°F-1 hr.- W. Q. + 1200°F-16 hrs.- A. C.						
D-182	55,000	52	19	25.6	43.0	321
D-183	55,000	71	46	30.6	36.7	321
1800°F-2 hrs.-W. Q. + 1200°F-16 hrs.-A. C.						
R-48	55,000	16	32	12.2	22.6	284
1800°F-2 hrs.-W. Q. + 1300°F-16 hrs.-A. C.						
R-48	55,000	57	7.3	25.7	45.5	286
9X-108[b]	55,000	63	19	15.4	30.6	321
9X-110[b]	55,000	16	11	17.6	29.2	332
	55,000	--	27	7.5	16.6	
1850°F-2 hrs.-W. Q. + 1200°F-16 hrs.-A. C.						
R-48	55,000	37	8.8	12.2	17.5	284
9X-110[b]	55,000	28	77	3.7	4.5	321
1850°F-2 hrs.-W. Q. + 1300°F-16 hrs.-A. C.						
R-47	55,000	67	39	14.1	16.2	269
R-48	55,000	32	135	17.1	26.5	286
9X-110[b]	55,000	44	25	20.2	37.1	321
1850°F-2 hrs.-W. Q. + 1375°F-16 hrs.-A. C.						
9X-110[b]	55,000	57	3	38.0	54.5	321

b - 50-lb. pilot ingot - forged to 5/8 inch square
d - 50-lb. pilot ingot - forged to 7/8 inch square

TABLE IX (Continued)

Heat No.	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness after Heat Treatment
1900°F-1 hr.-W. Q. + 1200°F-16 hrs.-A. C.						
D-117	50,000	--	39	4.5	7.1	269
D-182	55,000	7.5	269	3.9	5.5	286
D-183	55,000	8	185	4.6	6.6	311
D-218	50,000	--	471	11.2	30.9	286
	55,000	31	130[f]	6.1	19.6	293
D-251	50,000	--	8	23.5	35.8	302
D-254	50,000	--	28	6.6	14.6	262
D-267	50,000	--	257	8.1	9.0	286
R-18	50,000	--	62	21.8	35.4	262
R-19	50,000	--	9	7.7	8.9	217
R-20	50,000	--	84	17.1	37.6	258
R-41	55,000	36	62	13.5	17.4	255
R-47	55,000	11	194	2.0	5.2	269
R-48	55,000	7	38	2.7	6.9	269
1900°F-1 hr.-W. Q. + 1300°F-16 hrs.-A. C.						
D-218	50,000	--	298	22.3	44.2	302
	55,000	181	--	--	--	286
D-267	50,000	--	69	5.1	10.9	286
R-18	50,000	--	65	26.4	40.0	262
R-41	55,000	--	47	19.4	24.0	269
R-45	55,000	--	97	8.7	14.7	269
R-46	55,000	--	65	10.7	25.6	286
R-48	55,000	36/52	90	9.4	15.2	262
R-57	55,000	347	75	17.0	34.2	269
R-58	55,000	136	128	12.1	23.5	269
R-59	55,000	--	137	14.8	17.4	223
R-60	50,000	--	46	10.0	12.3	277
R-61	50,000	--	56	22.5	34.8	277
R-62	50,000	--	189	10.0	13.7	241
R-63	50,000	--	62	6.2	11.9	241
R-64	50,000	--	73	11.5	24.5	255
9X-108[b]	50,000	99	70	10.9	18.2	311
9X-110[b]	55,000	41	20	28.6	37.6	340
9X-110[c]	50,000	--	80	26.3	30.0	302
	55,000	74	31	19.0	32.6	
9X-110[e]	55,000	66	90	8.3	19.5	

b - 50-lb. pilot ingot - forged to 5/8 inch square
c - 1100-lb. ingot - rolled to 7/8 inch round
e - 1100-lb. ingot - rolled to 1-1/4 inch round
f - Time accurate to ± 12 hrs.

TABLE IX (Continued)

Heat No.	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Rupture Time (Hrs.) Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness after Heat Treatment
1900°F-1 hr.-W. Q. + 1300°F-16 hrs.-A. C. (Continued)						
9X-129[d]	55,000	251	62	19.1	36.6	255
9X-129[c]	50,000	--	265	8.0	20.4	255
	55,000	84	86	10.0	19.5	
1900°F-1 hr.-W. Q. + 1375°F-16 hrs.-A. C.						
R-20	50,000	--	12	11.6	36.3	258
2050°F-1 hr.-W. Q. + 1200°F-16 hrs.-A. C.						
D-267	50,000	--	30	1.9	7.0	302
R-18	50,000	--	58	9.0	10.0	248
2050°F-1 hr.-W. Q. + 1300°F-16 hrs.-A. C.						
D-218	50,000	--	63	2.7	6.0	302
D-267	50,000	--	13	.8	4.6	321
R-18	50,000	--	90	11.0	12.7	241
R-41	50,000	--	57[f]	5.0	8.6	228
2050°F-1 hr.-W. Q. + 1325°F-16 hrs.-A. C.						
D-264	50,000	60	72	17.1	18.3	196
D-266	50,000	52	0.8	15.8	18.1	222
2050°F-1 hr.-W. Q. + 1400°F-16 hrs.-A. C.						
D-182	50,000	57	93	11.8	25.2	269
D-183	50,000	136	126	18.1	31.3	286
D-251	50,000	11	2	21.2	37.8	302
D-252	50,000	30	2	14.1	24.3	248
D-265	50,000	--	31	12.4	24.3	223
2050°F-1 hr.-W. Q. + 1450°F-16 hrs.-A. C.						
D-265	50,000	--	62	22.1	59.0	217

c - 1100-lb. ingot - rolled to 7/8 inch round
d - 50-lb. pilot ingot - forged to 7/8 inch square
f - Time accurate to ± 12 hrs.

TABLE IX (Continued)

Heat No.	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Smooth Bar	Elong. (%)	Red. of Area (%)	Brinell Hardness after Heat Treatment
2150°F-1 hr.-W. Q. + 1300°F-16 hrs.-A. C.						
R-41	55,000	--	28	8.1	10.0	223
2150°F-1 hr.-W. Q. + 1325°F-16 hrs.-A. C.						
D-264	50,000	125	75	23.1	37.0	196
D-266	50,000	31	79	8.9	10.0	210
2150°F-1 hr.-W. Q. + 1375°F-16 hrs.-A. C.						
D-264	50,000	122	47	19.2	37.4	196
D-266	50,000	64	48	7.1	10.0	228

TABLE X

Influence of Treatment on the Properties of D-183 Alloy

(These results summarized from Tables VIII and IX)

A. Chemical Analyses

Heat	Type Heat	Chemical Analyses (weight per cent) C	Si	Mn	Cr	Mo	V	N	Other
D-183	17-lb. induction	.35	.21	17.18	12.25	3.12	.86	.189	
D-218	17-lb. induction	.24	.28	19.97	12.17	2.84	.89	.164	
R-45	17-lb. induction	.30	.24	17.36	12.19	3.04	.74	.138	
R-46	17-lb. induction	.31	.25	17.66	12.04	2.85	.70	.165	
R-47	17-lb. induction	.30	.26	17.66	11.97	2.88	.68	.153	
R-48	17-lb. induction	.27	.25	17.60	12.07	2.81	.73	.174	
R-57	17-lb. induction	.26	.18	17.38	12.18	2.96	.72	.227	.12 Al
R-58	17-lb. induction	.31	.37	17.30	12.10	2.98	.74	.224	.21 Al
9X-110	1100-lb. arc	.36	.54	18.88	12.43	3.09	.81	.240	.78 Ni
9X-129	1100-lb. arc	.29	.30	18.06	12.48	2.97	.70	.156	.72 Ni

B. Hardness at Room Temperature

Heat	Treatment [a] Solution (°F)	Age (°F)	Brinell Hardness Solution Treated	Aged
D-183	1750	1200	311	321
	1900	1200	---	311
	2050	1375	---	316
	2050	1400	---	286
D-218	1900	1200	235	294
	1900	1300	235	294
	2050	1200	---	302
	2050	1300	---	302
	2050	1375	---	311
R-45	1900	1300	255	269
R-46	1900	1300	241	286
R-47	1850	1300	277	269
	1900	1200	---	269
	2050	1375	---	277

a - Treatments - In addition to the temperatures given:
Solution: Cooling - water quench; Time - 1750°, 1900°, 2050°F - 1 hr. - 1800°, 1850°F - 2 hrs.

Aging: Cooling - air cool Time - 16 hrs.

TABLE X (Continued)

Heat	Treatment [a] Solution (°F)	Age (°F)	Brinell Hardness Solution Treated	Aged
R-48	1800	1200	272	284
	1800	1300	272	286
	1850	1200	262	284
	1850	1300	262	286
	1900	1200	---	269
	1900	1300	---	262
	2050	1375	---	286
R-57	1900	1300	---	269
R-58	1900	1300	---	269
9X-110[b]	1800	1300	300	329
	1850	1200	302	321
	1850	1300	302	321
	1850	1375	302	321
	1900	1300	277	340
9X-110[c]	1900	1300	255	302
	2050	1375	228	311
9X-129[d]	1900	1300	223	255

a - Treatments - In addition to the temperatures given:
Solution: Cooling - water quench; Time - 1750°, 1900°, 2050°F - 1 hr. - 1800°, 1850°F - 2 hrs.

Aging: Cooling - air cool Time - 16 hrs.

b - 50-lb. pilot ingot - forged to 5/8 inch square
c - 1100-lb. ingot - rolled to 7/8 inch round
d - 50-lb. pilot ingot - forged to 7/8 inch square

TABLE X (Continued)

C. Tensile Properties at Room Temperature

Heat	Treatment [a] Solution (°F)	Age (°F)	Tensile Strength (psi)	Yield Strength (psi) 0.2% Offset	0.02% Offset	Elong. (%)	Red. of Area (%)
D-183	1900	1200	146,100	95,000	77,100	34.0	35.2
	2050	1375	163,000	104,500	81,200	24.0	19.8
	2050	1400	146,000	84,720	62,400	12.2	11.5
D-218	1900	1200	143,000	100,500	86,000	29.0	33.2
	2050	1375	145,000	98,000	76,500	20.0	17.4
R-47	1900	1200	143,700	88,500	75,000	46.3	40.0
	2050	1375	137,600	80,800	65,500	30.0	25.0
R-48	1900	1200	145,700	93,000	82,500	39.5	41.7
	2050	1375	144,200	90,000	75,000	31.0	29.4
9X-110[b]	1850	1300	141,200	94,900	61,200	5.5	9.7
9X-110[c]	1900	1300	154,500	97,800	74,700	21.5	18.2
9X-129[c]	1900	1300	136,750	81,600	62,700	33.5	23.4

D. Rupture Properties at 1200°F of D-183 Alloy

Heat	Treatment [a] Solution (°F)	Age (°F)	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Smooth Bar	Elong. (%)	Red. of Area (%)
D-183	1750	1200	55,000	71	46	30.6	36.7
	1900	1200	55,000	8	185	4.6	6.6
	2050	1375	60,000	--	0.5	23.6	58.7
	2050	1375	55,000	--	81[f]	--	--
	2050	1375	50,000	97	240	4.0	7.8
	2050	1375	45,000	--	886	5.4	5.1
	2050	1400	50,000	136	126	18.1	31.3

a - Treatments - In addition to the temperatures given:
Solution: Cooling - water quench; Time - 1750°, 1900°, 2050°F - 1 hr. - 1800°, 1850°F - 2 hrs.

Aging: Cooling - air cool Time - 16 hrs.

b - 50-lb. pilot ingot - forged to 5/8 inch square
c - 1100-lb. ingot - rolled to 7/8 inch round
f - Estimated rupture time from log-log stress-rupture time curve.

TABLE X (Continued)

Heat	Treatment[a] Solution (°F)	Age (°F)	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Smooth Bar	Elong. (%)	Red. of Area (%)
D-218	1900	1200	55,000	31	130	6.1	19.6
	1900	1200	50,000	--	471	11.2	30.9
	1900	1300	55,000	181	105	--	--
	1900	1300	50,000	--	298	22.3	44.2
	2050	1200	50,000	--	16	0.4	4.0
	2050	1300	50,000	--	63	2.7	6.0
	2050	1375	55,000	43	165	5.6	9.0
	2050	1375	50,000	--	116	21.4	47.2
	2050	1375	45,000	452	1311	3.7	9.0
R-45	1900	1300	55,000	--	97	8.7	14.7
R-46	1900	1300	55,000	--	65	10.7	25.6
R-47	1850	1300	55,000	67	39	14.1	16.2
	1900	1200	55,000	11	194	2.0	5.2
	2050	1375	55,000	48	55	7.0	13.7
R-48	1800	1200	55,000	16[g]	32	12.2	22.6
	1800	1300	55,000	57[g]	7	25.7	45.5
	1850	1200	55,000	37	9	12.2	17.5
	1850	1300	55,000	32	135	17.1	26.5
	1900	1200	55,000	7	38	2.7	6.9
	1900	1300	55,000	52	90	9.4	15.2
	1900	1300	55,000	36	--	--	--
	2050	1375	55,000	19	33	7.1	10.0
	2050	1375	50,000	--	109	3.2	9.0
R-57	1900	1300	55,000	347	75	17.0	34.2
R-58	1900	1300	55,000	136	128	12.1	23.5

a - Treatments - In addition to the temperatures given:
Solution: Cooling - water quench; Time - 1750°, 1900°, 2050°F - 1 hr. - 1800°, 1850°F - 2 hrs.

Aging: Cooling - air cool Time - 16 hrs.

g - Flaw at fracture may have shortened rupture life.

TABLE X (Continued)

Heat	Treatment [a] Solution (°F)	Age (°F)	Stress (psi)	Rupture Time (Hrs.) Notch Bar	Smooth Bar	Elong. (%)	Red. of Area (%)
9X-110[b]	Forged	1375	55,000	--	29	9.5	27.3
	1800	1300	55,000	16	11	17.6	29.2
	1800	1300	55,000	--	27	7.5	16.6
	1850	1200	55,000	28	77	3.7	4.5
	1850	1300	55,000	44	25	20.2	37.1
	1850	1375	55,000	57	3	38.0	54.5
	1900	1300	55,000	41	20	28.6	37.6
9X-110[c]	1900	1300	55,000	74	31	19.0	32.6
	1900	1300	50,000	--	80	26.3	30.0
	2050	1375	55,000	47	51	13.5	24.0
9X-110[e]	1900	1300	55,000	66	90	8.3	19.5
9X-129[b]	Forged	1300	55,000	--	107	2.5	3.2
	1900	1300	55,000	251	62	19.1	36.6
9X-129[c]	1900	1300	55,000	84	86	10.0	19.5

E. Rupture Properties at 1350°F and 1500°F

Heat	Treatment [a] Solution (°F)	Age (°F)	Test Temp (°F)	Stress (psi)	Rupture Time (Hrs.)	Elong. (%)	Red. of Area (%)
9X-110[c]	1900	1300	1350	25,000	119	24.5	35.2
	1900	1300	1500	12,000	28	21.7	23.9
9X-129[c]	1900	1300	1350	25,000	147	24.2	30.0
	1900	1300	1500	12,000	83	7.5	13.1

---

a - Treatments - In addition to the temperatures given:
Solution: Cooling - water quench; Time - 1750°, 1900°, 2050° - 1 hr. - 1800°, 1850°F - 2 hrs.

Aging: Cooling - air cool Time - 16 hrs.

b - 50-lb. pilot ingot - forged to 5/8 inch square
c - 1100-lb. ingot - rolled to 7/8 inch round
e - 1100-lb. ingot - rolled to 1-1/4 inch round

TABLE XI

Typical Properties of D-183 Alloy

(Summary of results on ten heats as given in Table X)

A. Chemical Analysis

C	Si	Mn	Cr	Mo	V	N
.30	.4	18	12	3	.75	.15

B. Hardness

Treatment	Brinell Hardness Range	Brinell Hardness Average
Hot Worked	293-375	332
1900°F-1 hr.-W. Q.	223-277	245
1900°F-1 hr.-W. Q. + 1300°F-16 hrs.	255-340	290
2050°F-1 hr.-W. Q.	---	225
2050°F-1 hr.-W. Q. + 1375°F-16 hrs.	277-314	305

C. Tensile Properties at Room Temperature

	Tensile Strength (psi)	Yield Strength (psi) 0.2% Offset	Yield Strength (psi) 0.02% Offset	Elong. (%)	Red. of Area (%)
1900°F-1 hr.-W. Q. + 1300°F-16 hrs.-A. C.					
Range	136,750 to 154,500	81,600 to 97,800	62,700 to 74,700	21.5 to 33.5	18.2 to 23.4
Average	145,600	89,700	68,700	28.	21.
2050°F-1 hr.-W. Q. + 1375°F-16 hrs.-A. C.					
Range	137,600 to 163,000	80,800 to 104,500	65,500 to 81,200	20.0 to 31.0	17.4 to 29.4
Average	147,500	93,300	74,600	26.	23.

TABLE XI (Continued)

D. Rupture Properties at 1200°F

	55,000 psi Rupture Time (Hrs.) Notch Bar	55,000 psi Rupture Time (Hrs.) Smooth Bar	Elong. (%)	Red. of Area (%)	100 Hour Rupture Strength (psi)
1900°F-1 hr.-W. Q. + 1300°F-16 hrs.-A. C.					
Range	36 to 347	20 to 128	8.3 to 28.6	14.7 to 37.6	48,000 to 56,000
Average	121	74	14.	25.	52,000
2050°F-1 hr.-W. Q. + 1300°F-16 hrs.-A. C.					
Range	19 to 47	33 to 165	5.6 to 13.5	9.0 to 24.0	50,000 to 58,000
Average	36	77	8	14	54,000

TABLE XII

Chemical Analyses of Cast Alloys

Heat	C	Si	Mn	Cr	Ni	Co	Fe	W	Mo	N	B	Pouring Temp (°F)
	Chemical Analyses (weight percent)											
Vitallium												
Nominal	.39	1X	1X	27.00	2.75	Bal	2X	--	5.50	--	--	--
F-357	.26	.72	.75	25.14	2.58	Bal	.97	--	6.90	--	--	2750
F-360	.40	.28	.09	27.45	2.91	Bal	.58	--	5.60	--	--	2750
F-373	.42	.29	.29	26.20	3.17	Bal	.91	--	4.93	--	--	2750
F-379	.37	.36	.51	26.82	2.11	Bal	1.95	--	5.70	--	--	2760
F-388	.38	.38	.45	27.34	3.11	Bal	1.05	--	5.73	--	--	2750
F-380	.39	.43	.48	26.45	3.03	Bal	1.43	--	5.60	--	.20	2770
F-88												
Nominal	.40	.60X	1.50	25.00	30.00	20.00	Bal	5.00	4.50	.21	--	--
F-359	.33	.02	1.41	23.50	30.03	20.05	Bal	5.71	3.93	.07	--	2750
F-361	.43	.56	1.61	25.62	30.56	20.27	Bal	5.28	4.50	.27	--	2900
F-375	.49	.70	1.63	25.88	30.37	19.21	Bal	6.25	4.70	.20	--	2900
F-378	.31	.40	1.34	23.79	30.49	19.54	Bal	5.10	4.04	.19	--	2850
F-381	.39	.58	1.32	24.31	30.29	19.86	Bal	3.70	4.36	--	.09	2850
F-87												
Nominal	.40	.60X	1.35	25.00	30.00	10.00	Bal	5.25	4.50	.21	--	2750
F-358	.39	.10	1.36	24.38	29.82	10.50	Bal	5.59	4.87	.20	--	2850
F-362	.43	.18	1.38	19.60	30.42	10.65	Bal	5.76	4.70	.19	--	2895
F-374	.42	.40	1.29	19.76	30.13	9.77	Bal	6.18	4.70	.18	--	2895
F-377	.39	.41	.92	25.02	29.16	10.10	Bal	4.73	4.00	.21	--	2850
F-389	.36	.35	1.33	19.96	29.83	11.90	Bal	5.03	4.35	.15	--	2775
F-382	.44	.44	1.25	24.08	30.17	10.72	Bal	4.50	4.67	--	.14	2850

TABLE XIII

Properties of Cast Vitallium

Heat	Rupture Properties: 1500°F - 20,000 psi Rupt. Time (Hrs.)	Elong. (%)	Red. of Area (%)	1600°F - 20,000 psi Rupt. Time (Hrs.)	Elong. (%)	Red. of Area (%)	Others Rupt. Time (Hrs.)	Elong. (%)	Red. of Area (%)	Thermal Shock: Cycles to Failure at Max. Test Temp (°F) of 1900	1800	1750	1700
F-357	35	8.9	37.4	15	In fillet		--	--	--	3000[a]	--	--	--
	40	8.9	21.0	16	10.1	33.5	--	--	--	1193[a] 2096[b]	--	--	--
F-360	54	12.4	15.8	31	10.3	19.8	1500°F - 30,000 psi 8	7.6	10.0	--	--	--	--
	--	--	--	49	16.7	27.1	1600°F - 15,000 psi 237	14.6	17.5	--	--	--	--
F-373	62	9.8	16.4	57	14.6	24.6	1500°F - 30,000 psi 7	9.5	24.2	1660 2015	6933 7232	--	--
	64	4.6	18.8	49	8.6	15.2	1800°F - 10,000 psi 74	73	11.6	-- 1338[b]	5696 6610[b]	6456 6456[b]	--
F-379	62	9.5	23.1	17	7.9	21.0	--	--	--	--	--	--	--
F-388	--	--	--	21	13.6	19.0	1600°F - 30,000 psi 6	17.1	24.6	2562 1842 2202[b]	--	--	--
F-380	697	15.7	28.2	23	25.1	69.2	1500°F - 30,000 psi 55	26.7	40.4	1974	5681	--	--
							1600°F - 30,000 psi 6	30.1	56.5	3596	2028	--	14,056
							1800°F - 10,000 psi 171	20.1	58.5	2235[b]	3854[b]	--	--

a - Test edge polished during test
b - Average of results given

TABLE XIV

Properties of Cast F-88

Heat	Rupture Properties, 1500°F - 20,000 psi: Rupt. Time (Hrs.)	1500°F - 20,000 psi: Elong. (%)	1500°F - 20,000 psi: Red. of Area (%)	1600°F - 20,000 psi: Rupt. Time (Hrs.)	1600°F - 20,000 psi: Elong. (%)	1600°F - 20,000 psi: Red. of Area (%)	Others: Rupt. Time (Hrs.)	Others: Elong. (%)	Others: Red. of Area (%)	Thermal Shock, Cycles to Failure at Max. Test Temp. (°F) of 1900	1800	1750	1700
F-359	92	14.0	22.0	9	11.4	34.5	--	--	--	718[a] 827[a] 1035[a] 708[a] 822[b]	--	--	--
	84	16.0	31.6	12	13.5	37.1	--	--	--		--	--	--
F-361	54	19.6	38.0	11	18.7	34.9	1500°F - 30,000 psi 19	29.6	32.8	--	--	--	--
	88	14.6	29.2	4	12.3	37.5	1500°F - 15,000 psi 323	10.3	35.8	--	--	--	--
F-375	68	15.1	23.0	8	24.2	48.4	497	13.5	30.0	2342	2841 4531	3088	8627
	59	14.6	21.6	9	18.0	49.5	-	--	--	2201 2271[b]	3215 3529[b]	3088 3088[b]	11004 9816[b]
F-378	70 58	17.1 19.3	44.8 64.4	7	21.6	67.7	1600°F - 15,000 psi 171	23.6	72.4	--	--	--	--
F-381 (.07 B)	160	15.5	55.6	25	20.6	67.8	1500°F - 30,000 psi 28	22.6	40.0	1034	3368	2379	--
	328	17.1	62.0	--	--	--	1600°F - 15,000 psi 47	14.1	49.5	2253 1643[b]	3626 3494[b]	2379 2379[b]	--

a - Test edge polished during test
b - Average of results given

TABLE XV

Properties of Cast F-87

Heat	Rupture Properties: 1500°F - 20,000 psi Rupt. Time (Hrs.)	1500°F - 20,000 psi Elong. (%)	1500°F - 20,000 psi Red. of Area (%)	1600°F - 20,000 psi Rupt. Time (Hrs.)	1600°F - 20,000 psi Elong. (%)	1600°F - 20,000 psi Red. of Area (%)	Others Rupt. Time (Hrs.)	Others Elong. (%)	Others Red. of Area (%)	Thermal Shock: Cycles to Failure at Max. Test Temp. (°F) of 1900	1800	1750	1700
F-358	55	16.6	27.8	6	23.6	34.8	--	--	--	610[a] 1481[a] 822[a] 1043[a] 989[b]	--	--	--
	46	22.5	71.1	4	14.5	42.0	--	--	--		--	--	--
F-362	29	19.6	45.0	5	27.0	60.0	1500°F - 30,000 psi 3	20.1	37.0	--	--	--	--
	31	17.1	67.2	3	19.3	72.1	1600°F - 15,000 psi 21	9.6	66.5	--	--	--	--
F-374	44	21.6	51.5	4	29.5	58.0	1500°F - 15,000 psi 488	12.6	39.6	1496	1738	1994	--
	45	24.0	44.9	5	27.0	59.0	1600°F - 15,000 psi 13	23.6	67.0	1040 1268[b]	1538 1638[b]	1486 1740[b]	--
F-377	38	28.6	47.7	6	13.1	46.5	--	--	--	--	--	--	--
F-389	--	--	--	8	33.6	43.0	1500°F - 25,000 psi 12	25.1	56.5	1875 1209 1547[b]	--	--	--
F-382	266	13.6	43.0	23	19.2	50.3	1500°F - 30,000 psi 89	21.6	31.6	1480	3813	3236	--
							1600°F - 15,000 psi 165	27.6	55.6	1480 1480[b]	3489 3641[b]	1045 2140[b]	--

a - Test edge polished during test
b - Average of results given

TABLE XVI

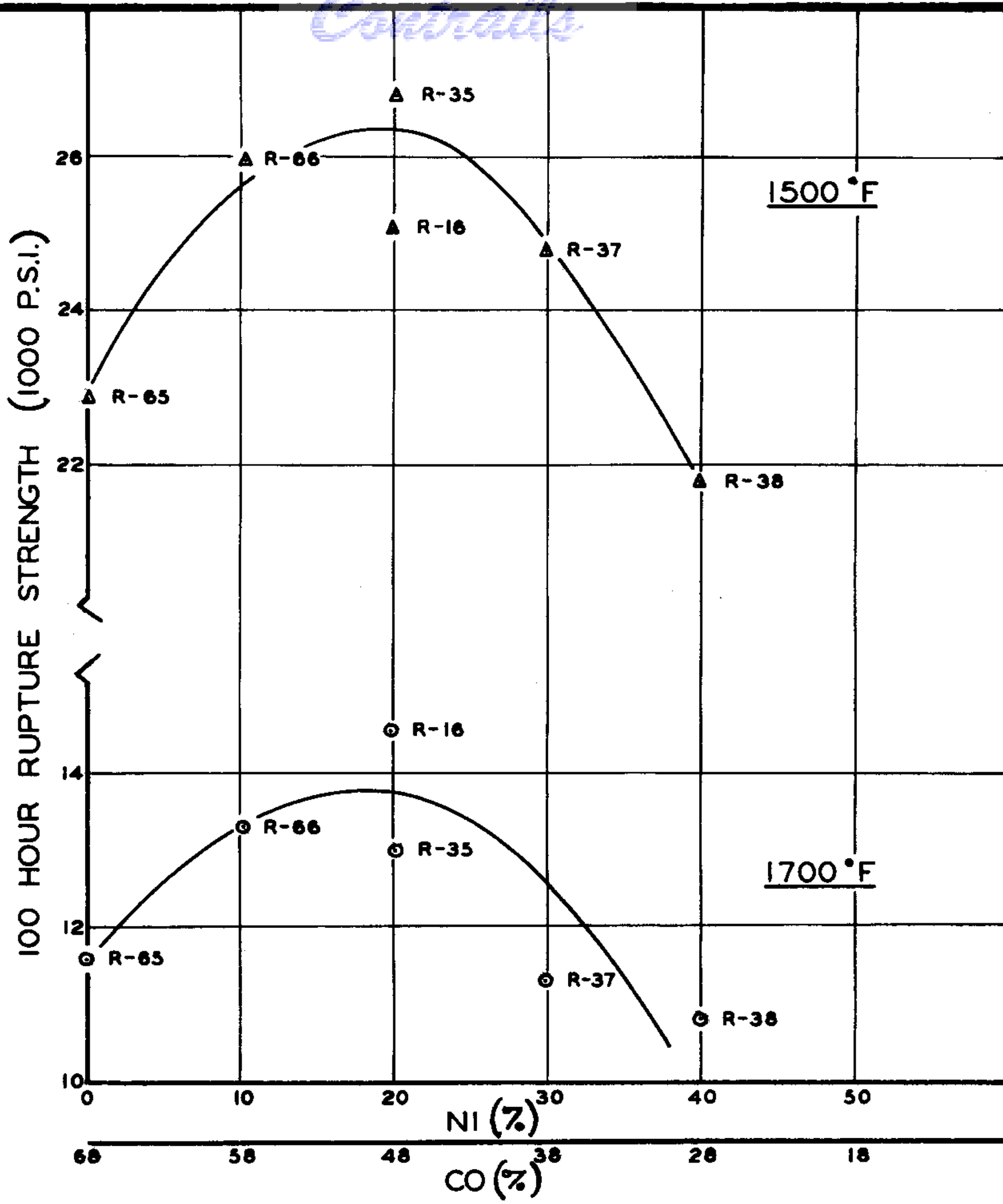
Room Temperature Tensile Properties of Cast Alloys

Heat	Tensile Strength (psi)	Elong. (%)	Red. of Area (%)
Vitallium			
F-357	84,300 92,200	2.9 3.6	33.6 10.1
F-360	100,600 102,000	4.0 5.0	11.6 9.4
F-373	106,500 98,700	3.0 6.0	12.6 11.5
F-379	106,100	2.0	1.2
F-388	101,500	3.4	11.0
F-380	102,200	2.0	5.7
F-88			
F-359	82,300 79,700	3.5 3.5	5.4 5.5
F-361	84,300 81,400	1.0 1.0	1.2 2.4
F-375	83,400 81,300	1.0 1.0	0 2.0
F-378	81,300	3.9	9.2
F-381	74,500 75,100	Broke in fillet 2.0	-- 1.2
F-87			
F-358	81,300 83,200	4.0 5.0	4.7 5.5
F-362	76,450 75,300	2.5 2.5	2.0 2.8
F-374	71,450 74,300	2.5 2.0	6.2 2.0
F-377	81,200	4.0	7.4
F-389	77,400	3.5	3.1
F-382	72,000	Broke in fillet	--

TABLE XVII

Impact Properties of Cast Alloys

Alloy	Heat	Test Temp (°F)	Charpy V-Notch Impact Strength (Ft. - Lbs.)
Vitallium	F-388	Room Temp	6.0
		1500	17.0
		1500	14.5
		1700	18.0
		1900	18.0
F-87	F-389	Room Temp	2.5
		1500	13.5
		1500	14.0
		1700	15.5
		1900	18.0
		1900	14.0

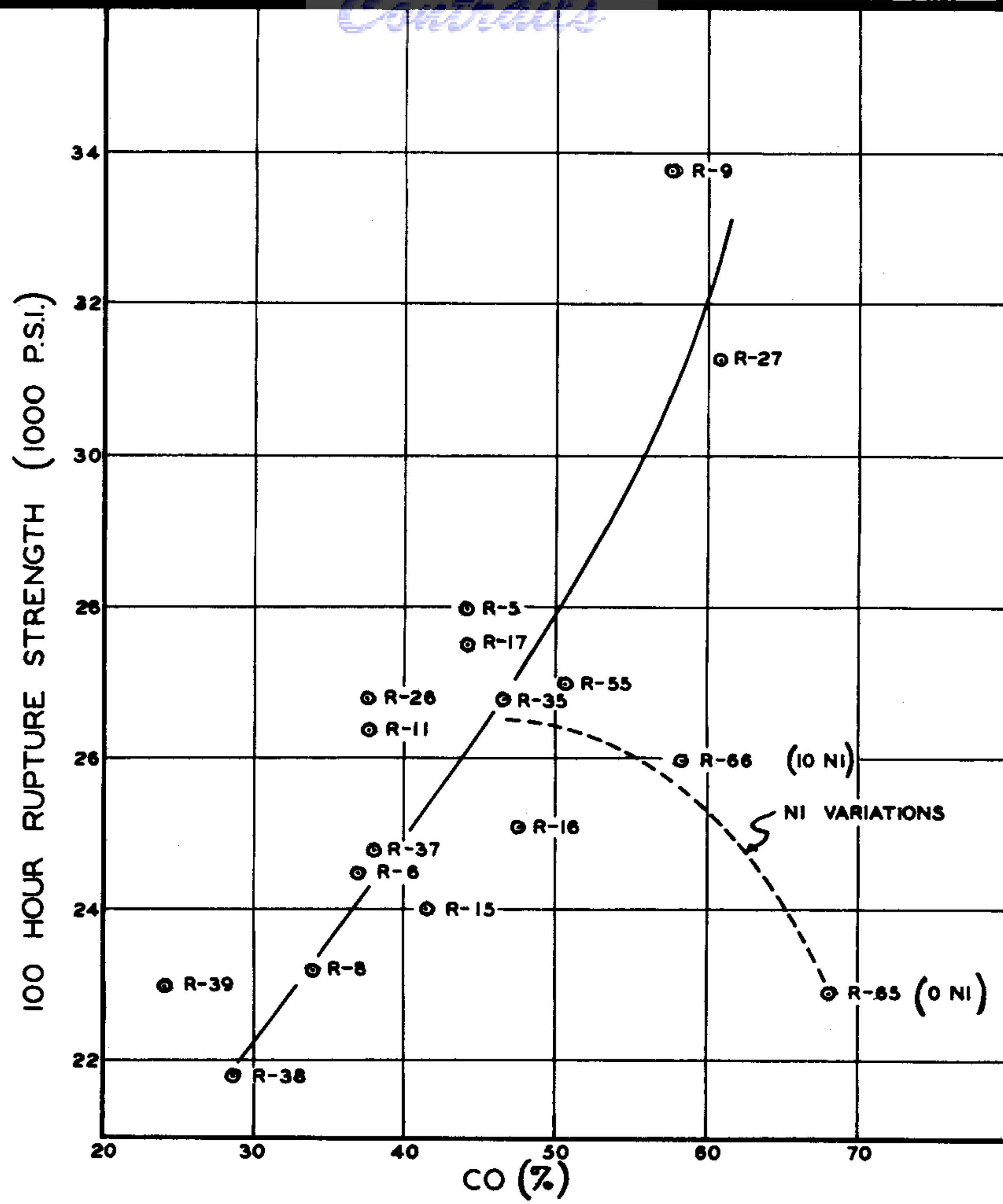


COMPOSITION LIMITS OF MODIFICATIONS SHOWN:

C - .10 / .13   W - 3.8 / 4.2   CB + TA - .85 / 1.3

CR - 19.7 / 20.1   MO - 4.1 / 5.2 $\frac{CB + TA}{C}$ - 6.5 / 10.6

FIGURE - I INFLUENCE OF NICKEL AND COBALT ON RUPTURE STRENGTH OF S-816 MODIFICATIONS AT 1500 AND 1700 °F



COMPOSITION LIMITS OF MODIFICATIONS SHOWN:

C-.10/.42	MO - 0/11.7	$\frac{CB + TA}{C}$ - 2.3/11.5
CR-9.9/24.7	W - 3.8/11.2	
NI-0/39.8	CB+TA-.58/3.9	

FIGURE - 2 INFLUENCE OF COBALT ON RUPTURE STRENGTH OF S-816 (R-17) AND MODIFICATIONS AT 1500°F

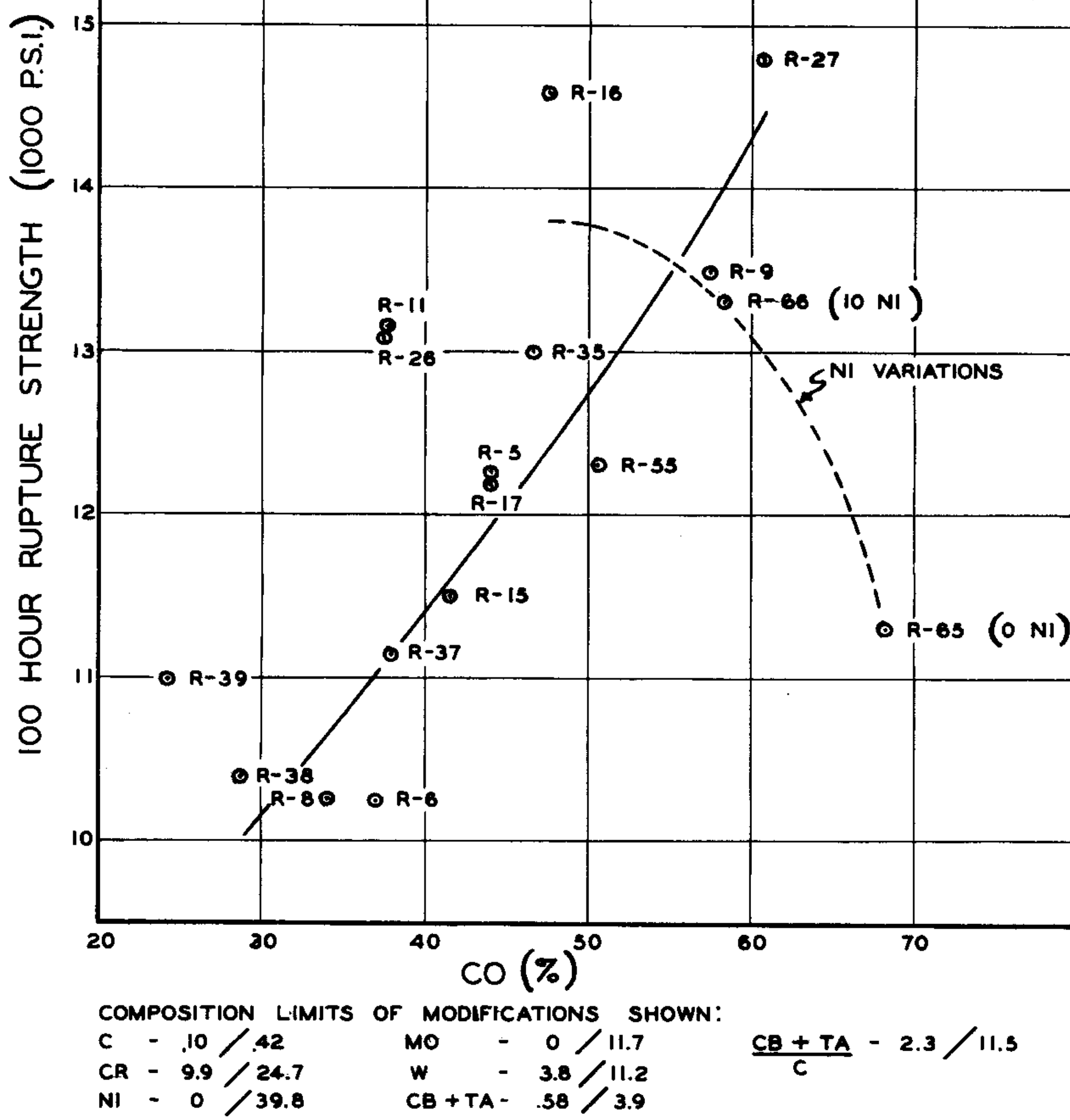


FIGURE - 3 INFLUENCE OF COBALT ON RUPTURE STRENGTH OF S-816 (R-17) AND MODIFICATIONS AT 1700°F

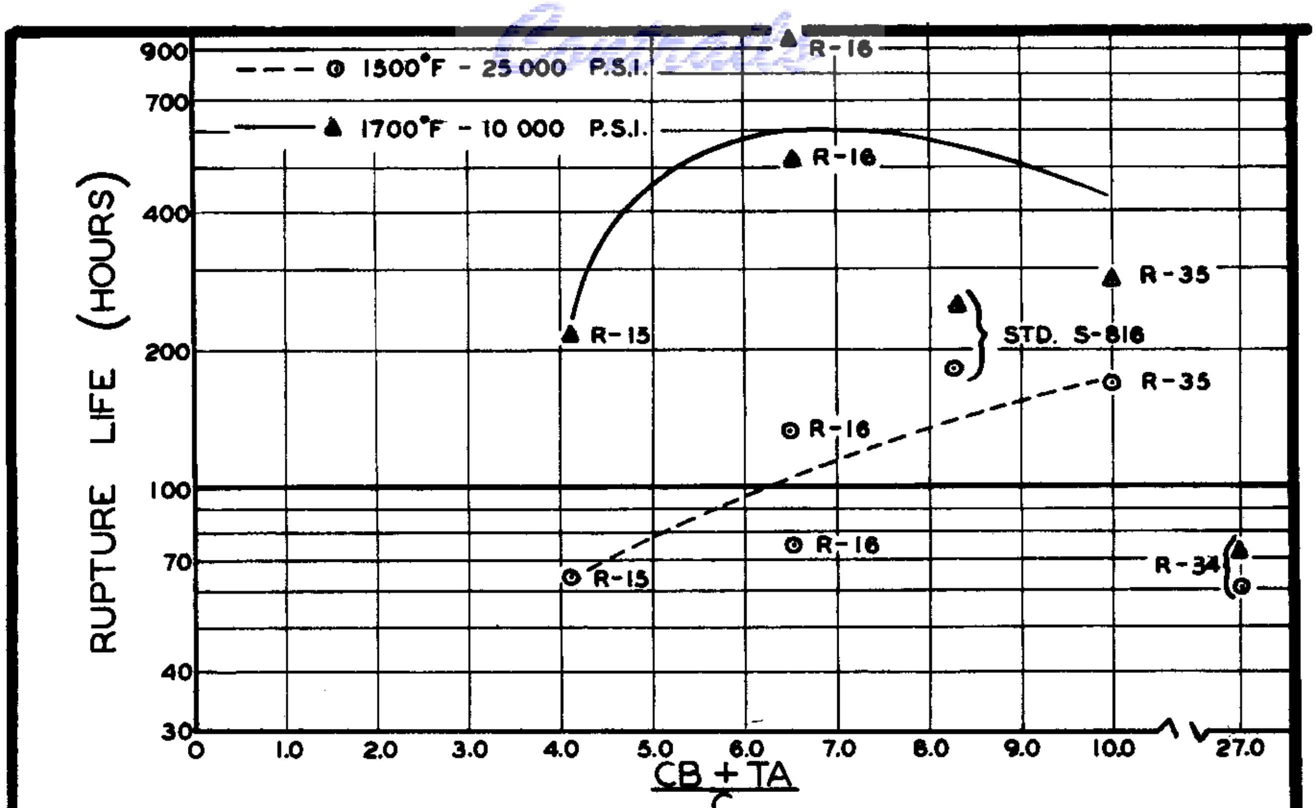


FIGURE - 4 INFLUENCE OF CB + TA / C RATIO ON RUPTURE LIFE OF LOW C S-816

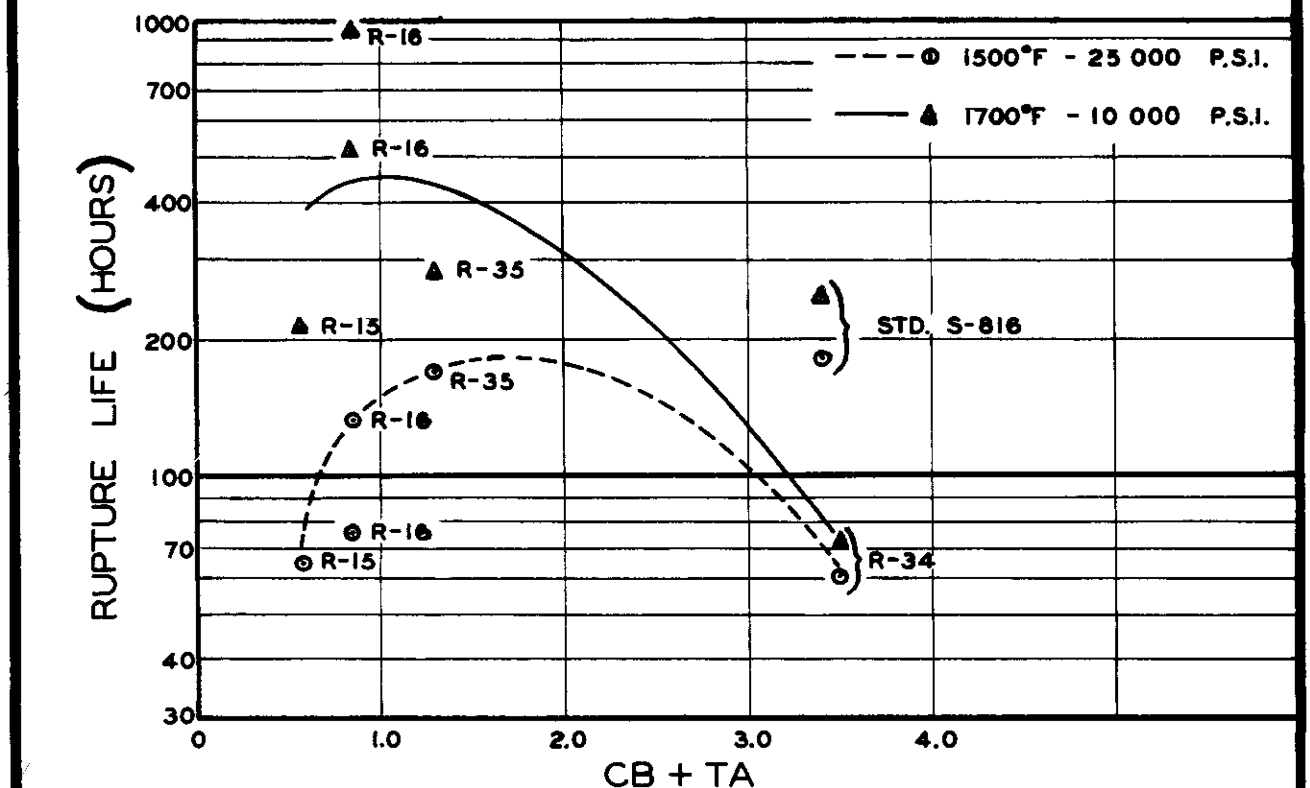


FIGURE - 5 INFLUENCE OF CB + TA ON RUPTURE LIFE OF LOW C S-816

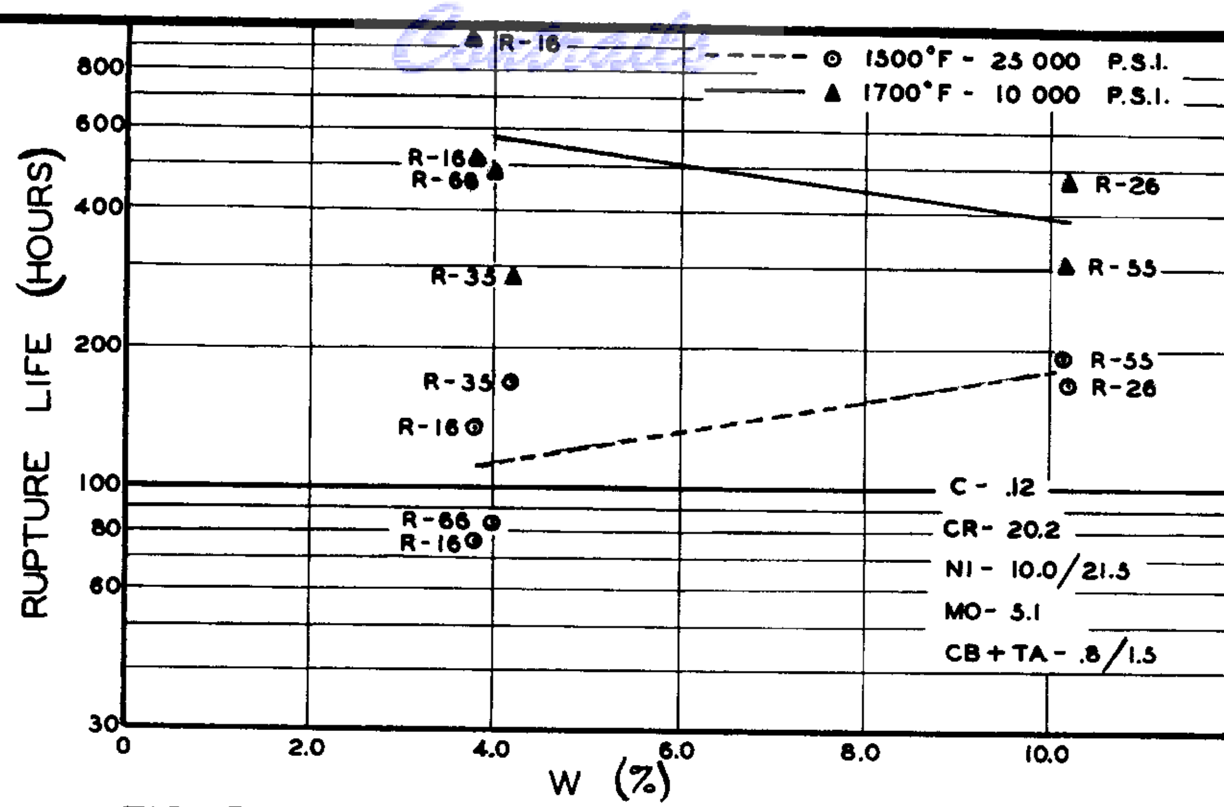


FIGURE - 6 INFLUENCE OF W ON RUPTURE LIFE OF S-816 MODIFICATIONS

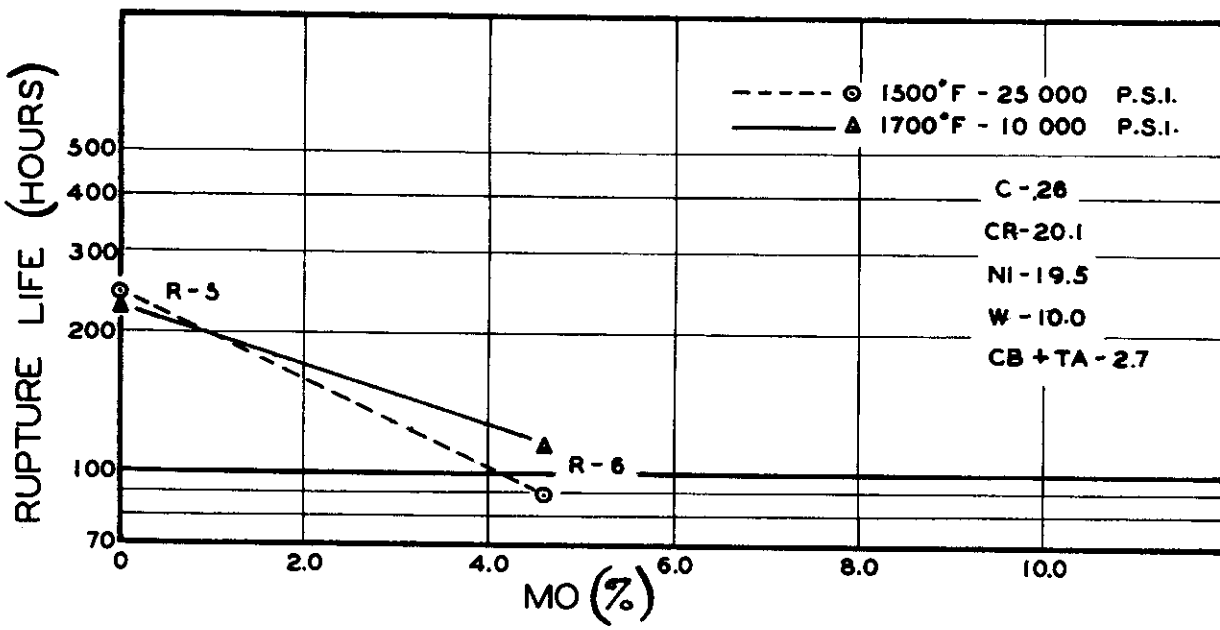
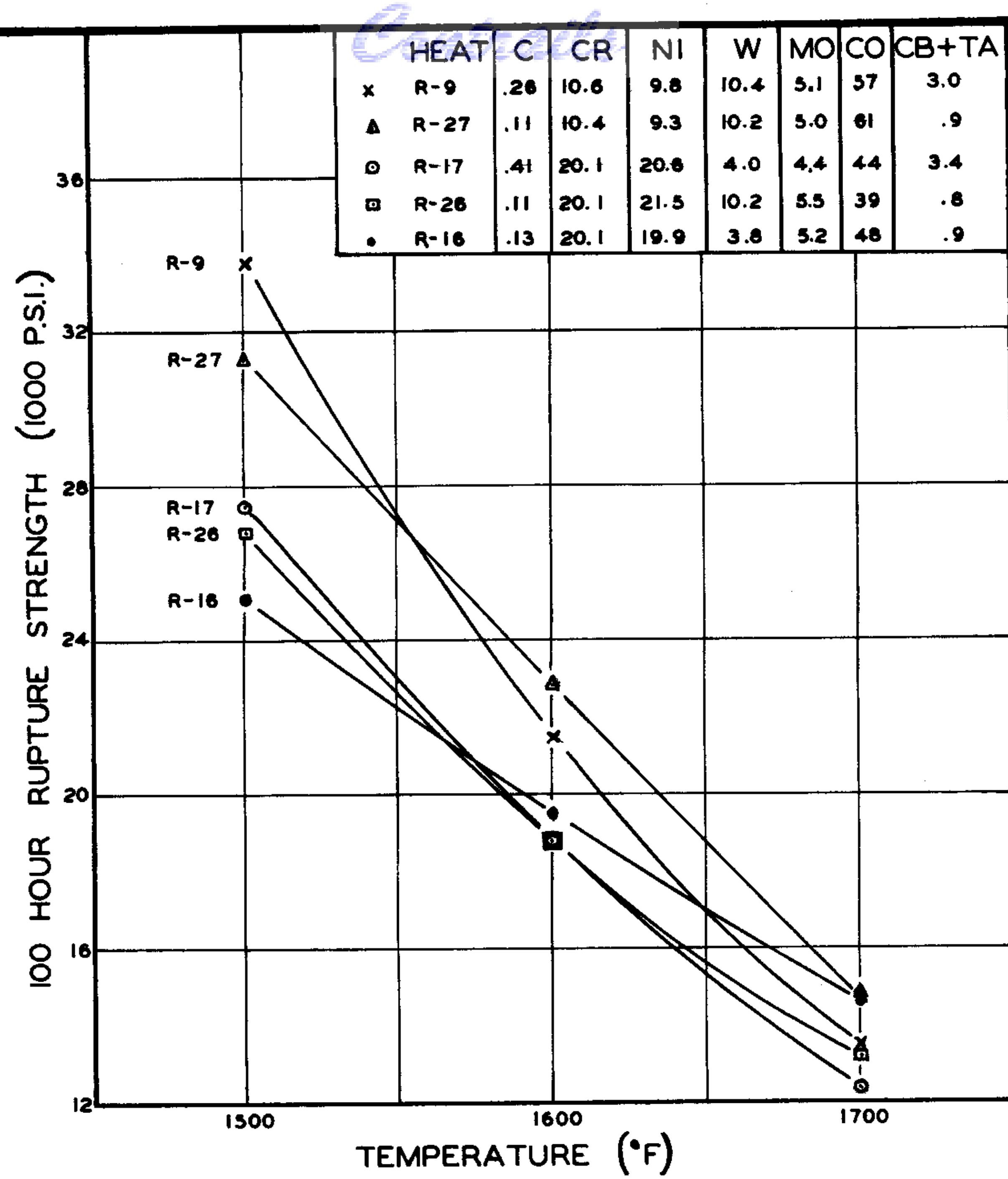
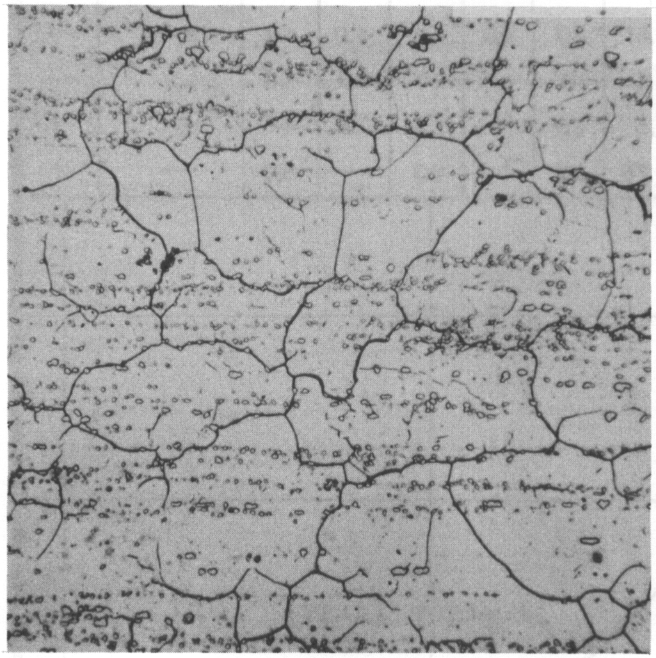


FIGURE - 7 INFLUENCE OF MO ON RUPTURE LIFE OF S-816 MODIFICATIONS

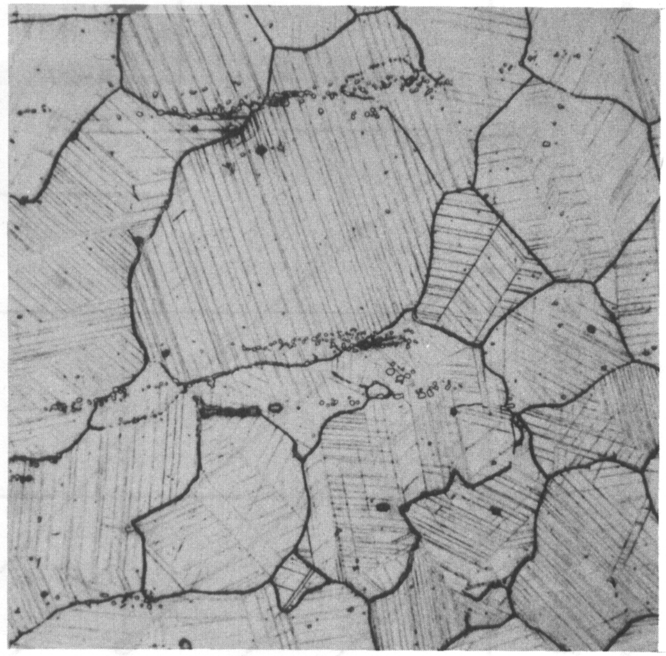


	HEAT	C	CR	NI	W	MO	CO	CB+TA
x	R-9	.26	10.6	9.8	10.4	5.1	57	3.0
▲	R-27	.11	10.4	9.3	10.2	5.0	61	.9
⊙	R-17	.41	20.1	20.6	4.0	4.4	44	3.4
⊡	R-26	.11	20.1	21.5	10.2	5.5	39	.8
•	R-16	.13	20.1	19.9	3.8	5.2	48	.9

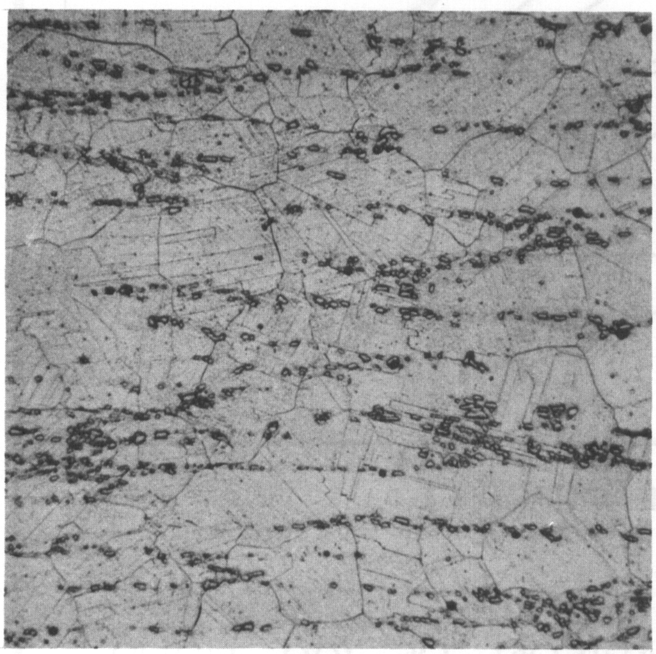
FIGURE – 8 EFFECT OF TEMPERATURE ON THE 100 HOUR RUPTURE STRENGTH OF S-816 (R-17) AND ITS MODIFICATIONS



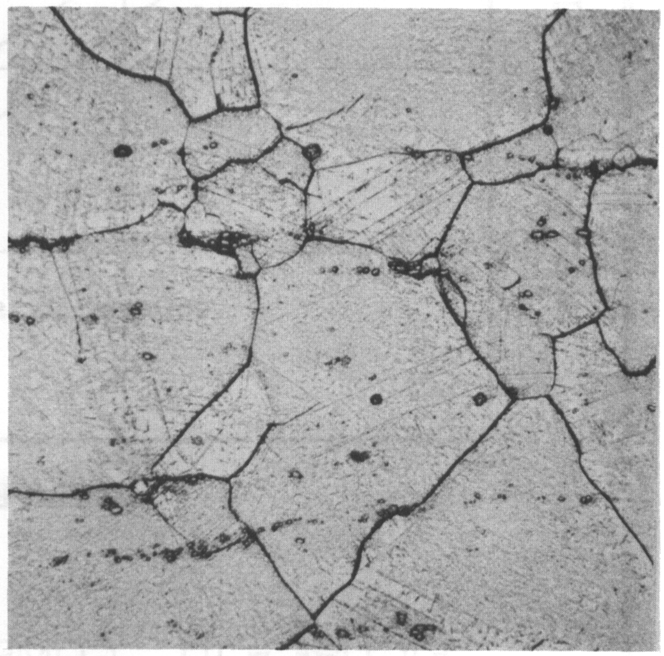
Heat R-17 - Std. S-816



Heat R-27 - Mod. S-816
C-.11, Cr-10.4, Ni-9.3
Co-60.6, W-10.2, Cb+Ta-.9



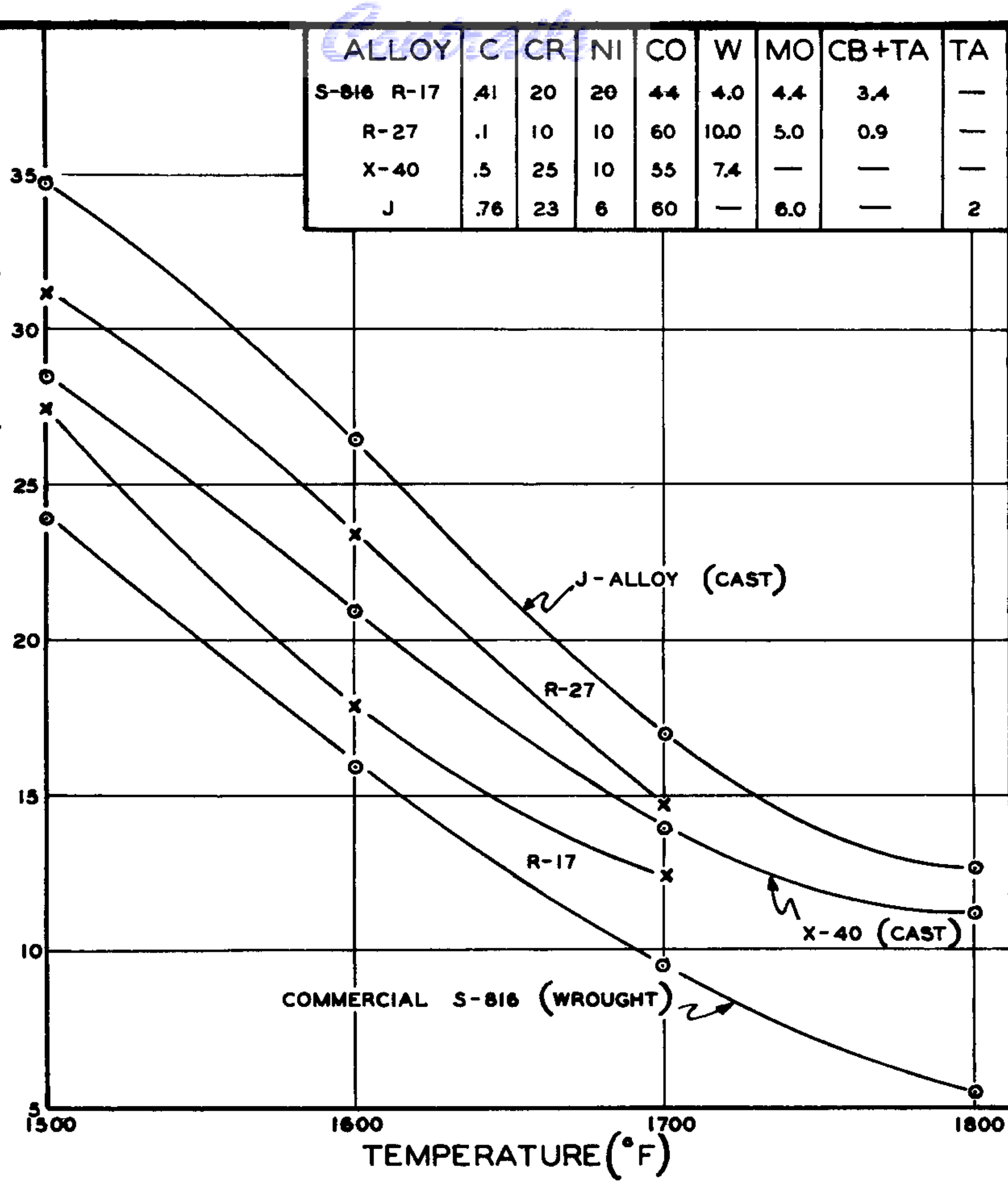
Heat R-10 - Mod. S-816
C-.27, Cr-10.5, Ni-10.2
Co-60.7, W-10.4, Mo-0,
Cb+Ta-3.0



Heat R-36 - Mod. S-816
C-.13, Cr-9.9, Co-58.0
Cb+Ta-1.2

Note: Standard analyses except as indicated
Heat Treatment - 2250°F-1 hour-W.Q. + 1400°F-16 hours-A.C.
Mag. - 500X Etchant: 92-HCl, 5 H_2SO_4, 3-HNO_3

Figure 9 - Microstructures of Standard S-816 and Modifications



ALLOY	C	CR	NI	CO	W	MO	CB+TA	TA
S-816 R-17	.41	20	20	44	4.0	4.4	3.4	—
R-27	.1	10	10	60	10.0	5.0	0.9	—
X-40	.5	25	10	55	7.4	—	—	—
J	.76	23	6	60	—	6.0	—	2

FIGURE - 10 COMPARISON OF 100 HOUR RUPTURE STRENGTHS OF HEAT R-27 AND S-816 ALLOY (R-17) WITH HIGH STRENGTH CAST ALLOYS

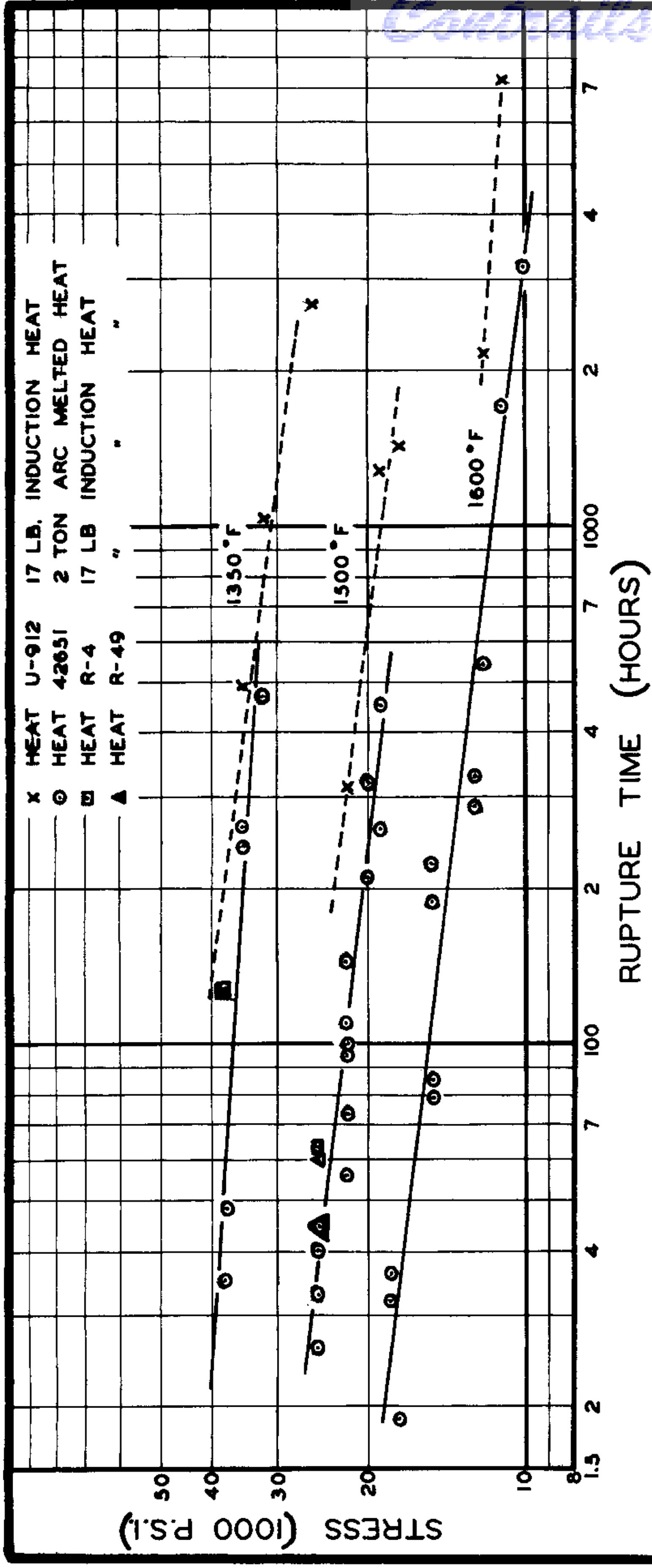


FIGURE - 11 RUPTURE DATA ON U-912 ALLOY

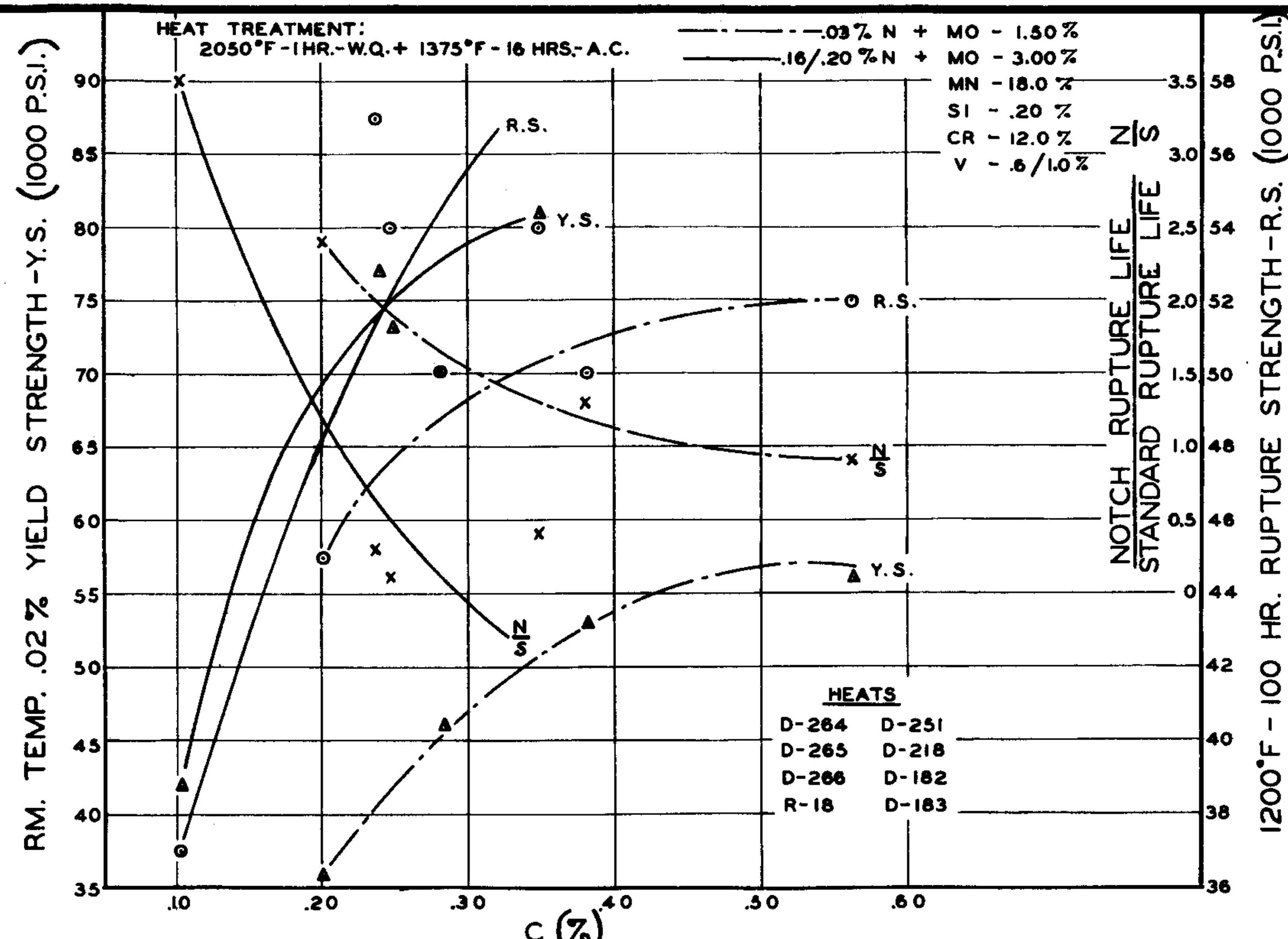


FIGURE - 12 INFLUENCE OF C ON THE PROPERTIES OF FE-MN-CR ALLOYS

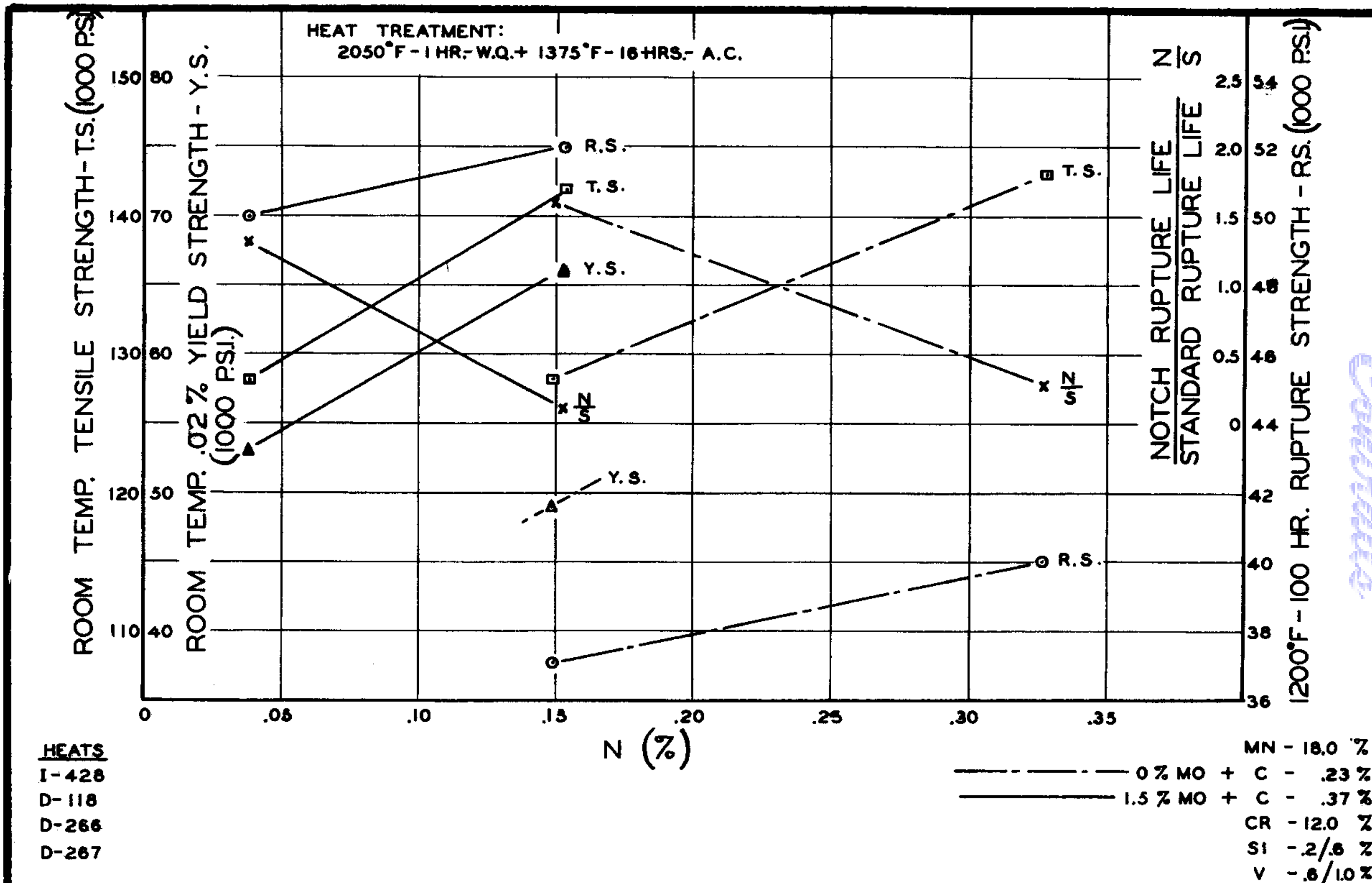
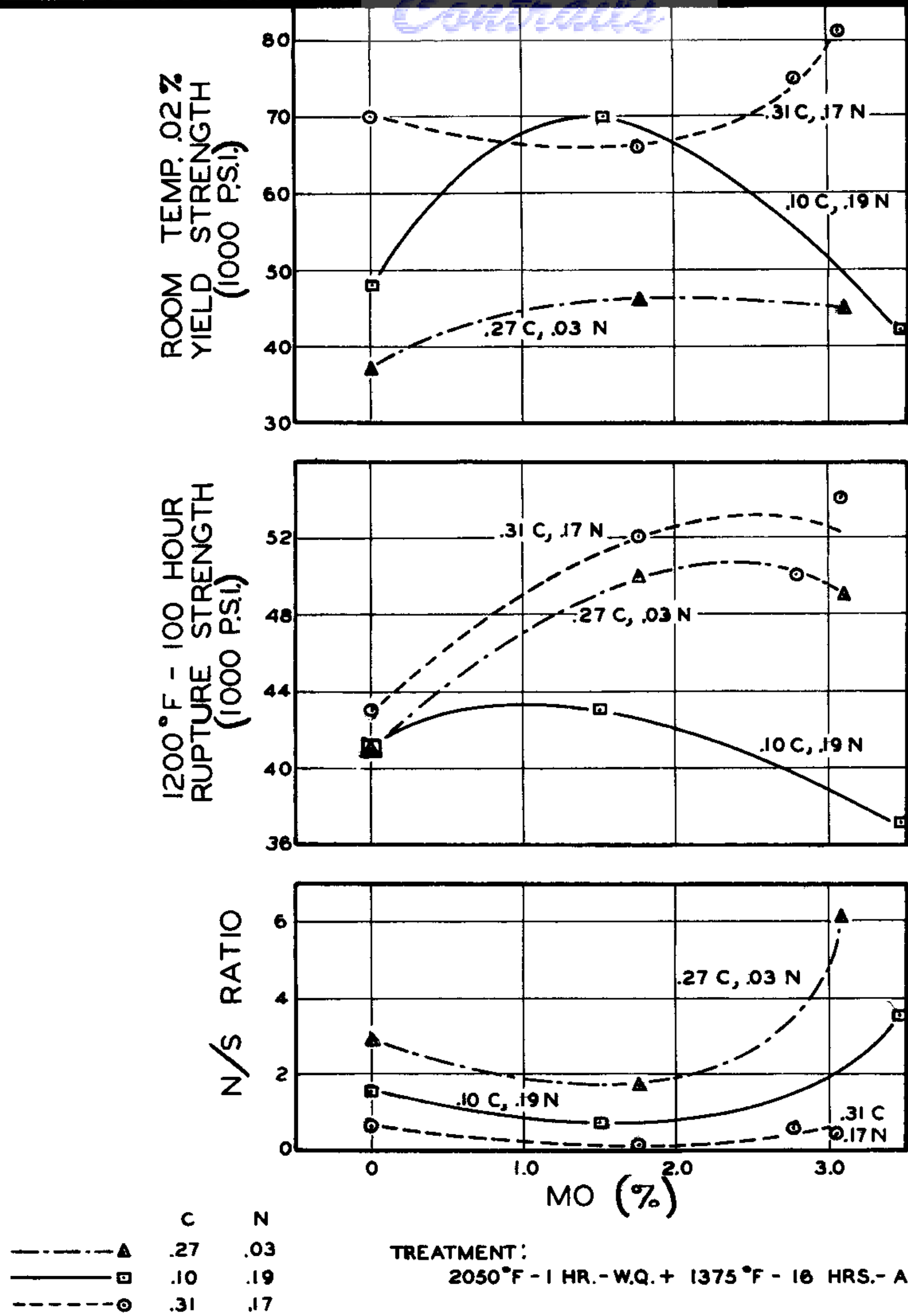
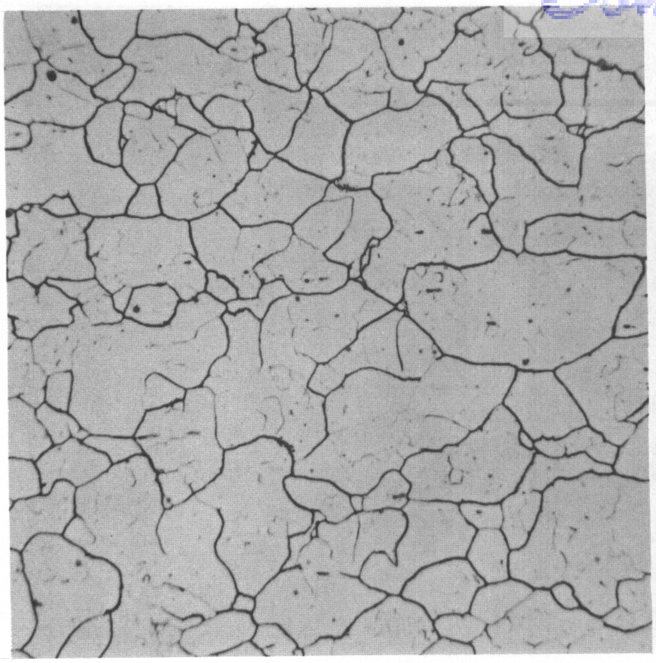


FIGURE - 13 INFLUENCE OF N ON THE PROPERTIES OF FE-MN-CR ALLOYS

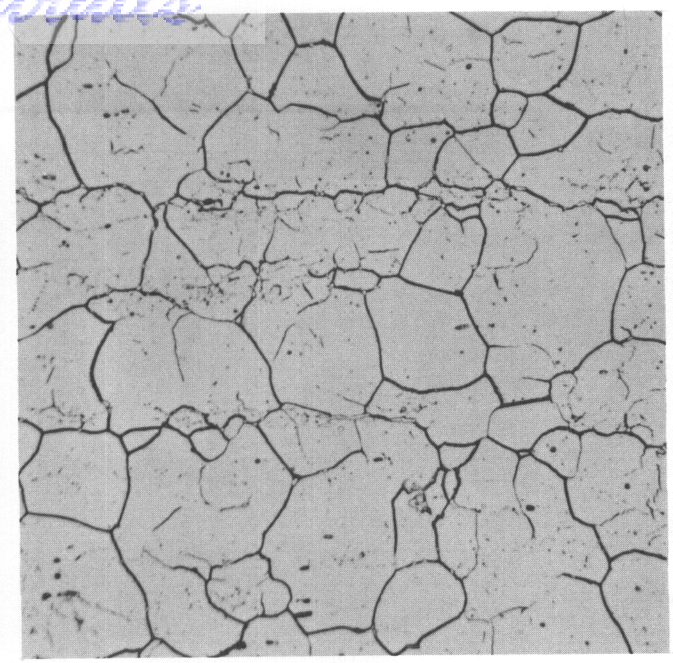


TREATMENT:
2050°F - 1 HR. - W.Q. + 1375°F - 16 HRS. - A.C.

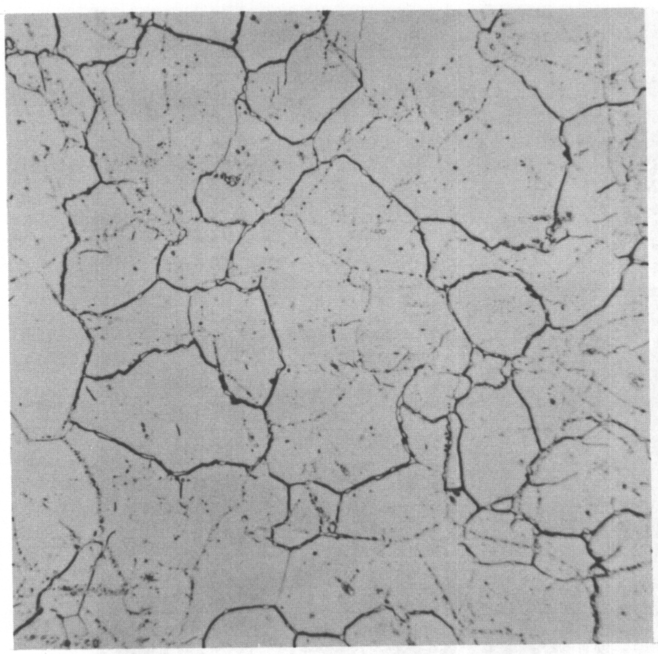
FIGURE - 14 INFLUENCE OF MO ON TENSILE AND RUPTURE PROPERTIES OF FE-MN-CR MODIFICATIONS



Heat R-62
Mn-16.0



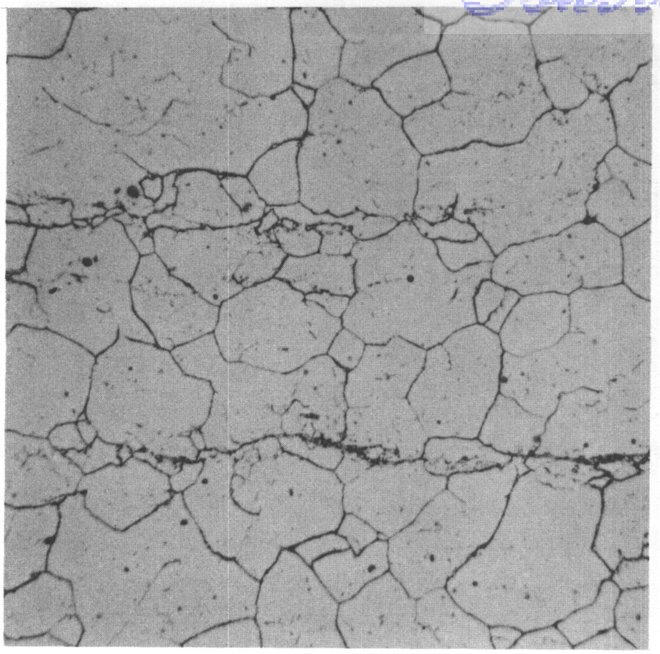
Heat R-63
Mn-19.9



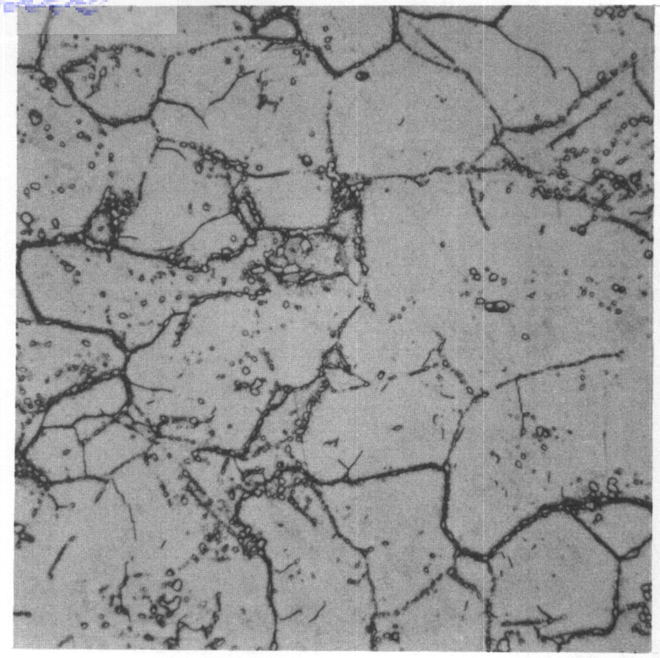
Heat R-64
Mn-21.8

Heat Treatment - 1900°F-1 hour-W.Q. + 1300°F-16 hours-A.C.
Mag. 500X Etchant: 92 HCl, 5-H_2SO_4, 3-HNO_3

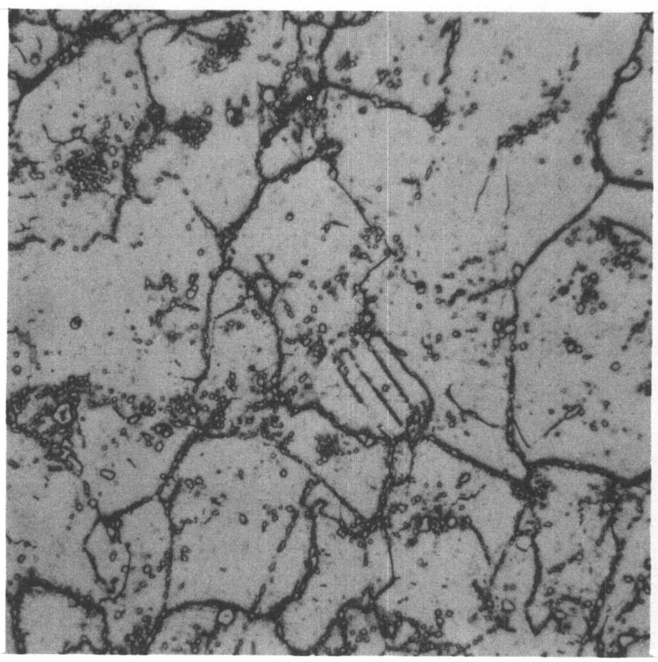
Figure 15 - Microstructures of D-183 Alloy with Increasing Mn Content.



Heat R-59
C- .28, N-.11



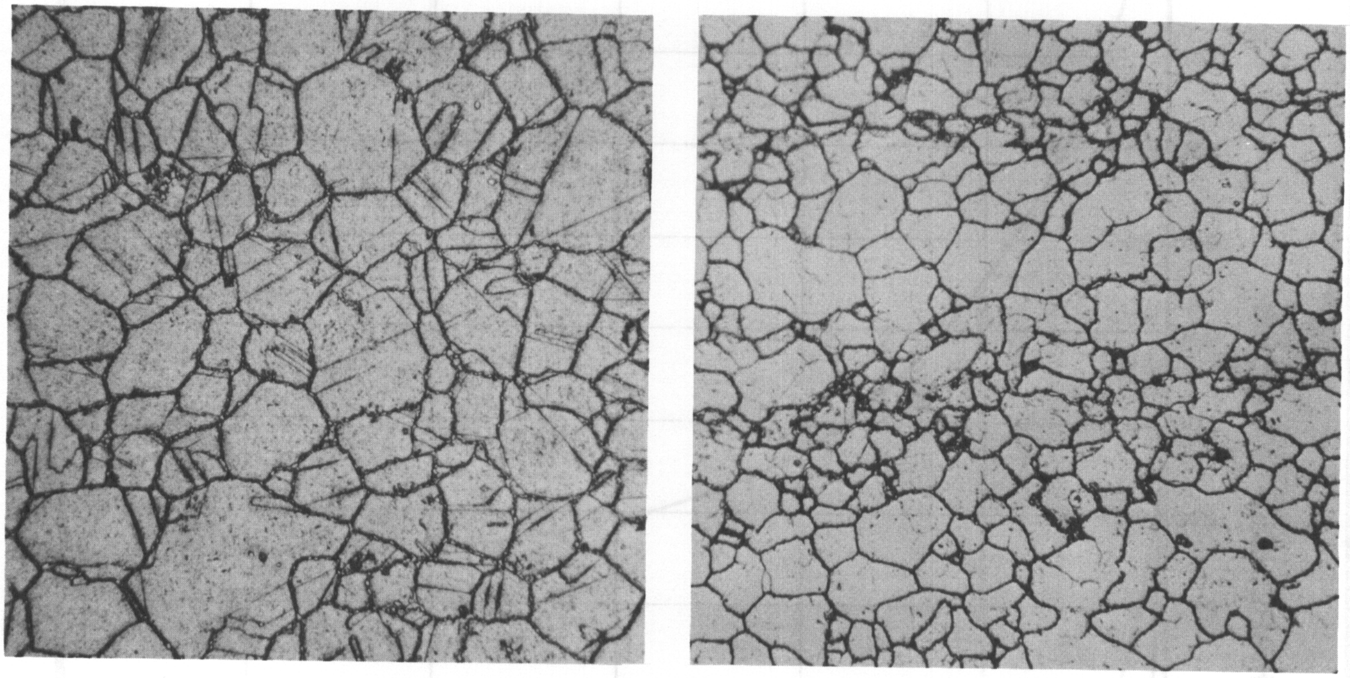
Heat R-60
C-.41, N- .10



Heat R-61
C-.48, N-.09

Heat Treatment - 1900°F-1 hour-W.Q. + 1300°F-16 hours-A.C.
Mag. 500X Etchant: 92-HCl, 5-H_2SO_4, 3 HNO_3

Figure 16 - Microstructures of D-183 Alloy with Increasing C Content.



Heat 9X-110
Forged to 5/8" square
from 50-lb. Pilot Ingot

Heat 9X-129
Rolled to 7/8" round
from 1100-lb. ingot

Heat Treatment - 1900°F-1 hour-W. Q. + 1300°F-16 hours-A. C.
Mag. 500X Etchant: 1 g Picric Acid, 5 cc HCl, 100 cc Alcohol

Figure 17 - Microstructures of D-183 Alloy from 1100-pound Arc-melted Heats

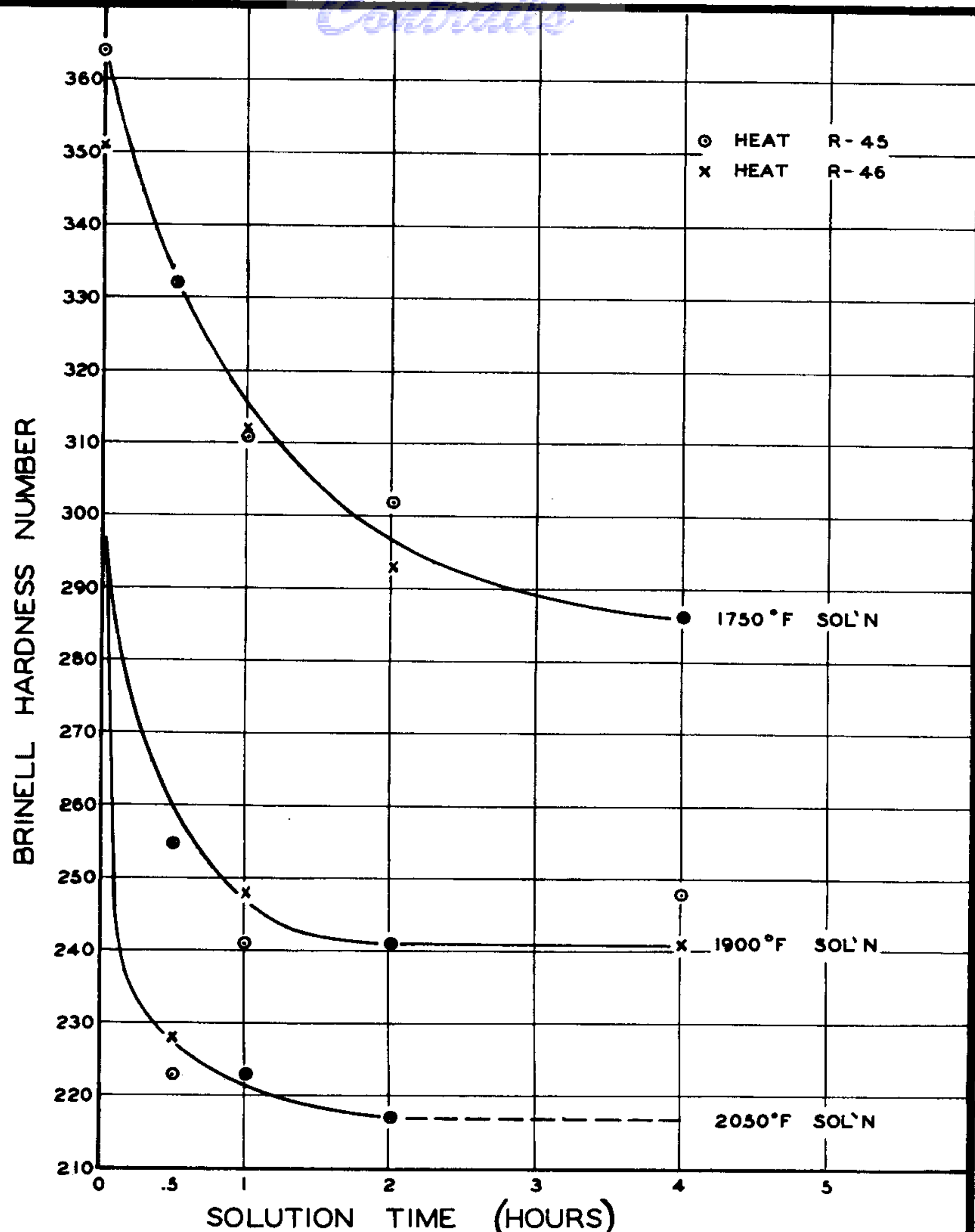


FIGURE - 18 EFFECT OF SOLUTION TIME AND TEMPERATURE ON THE ROOM TEMPERATURE HARDNESS OF D-183 ALLOY

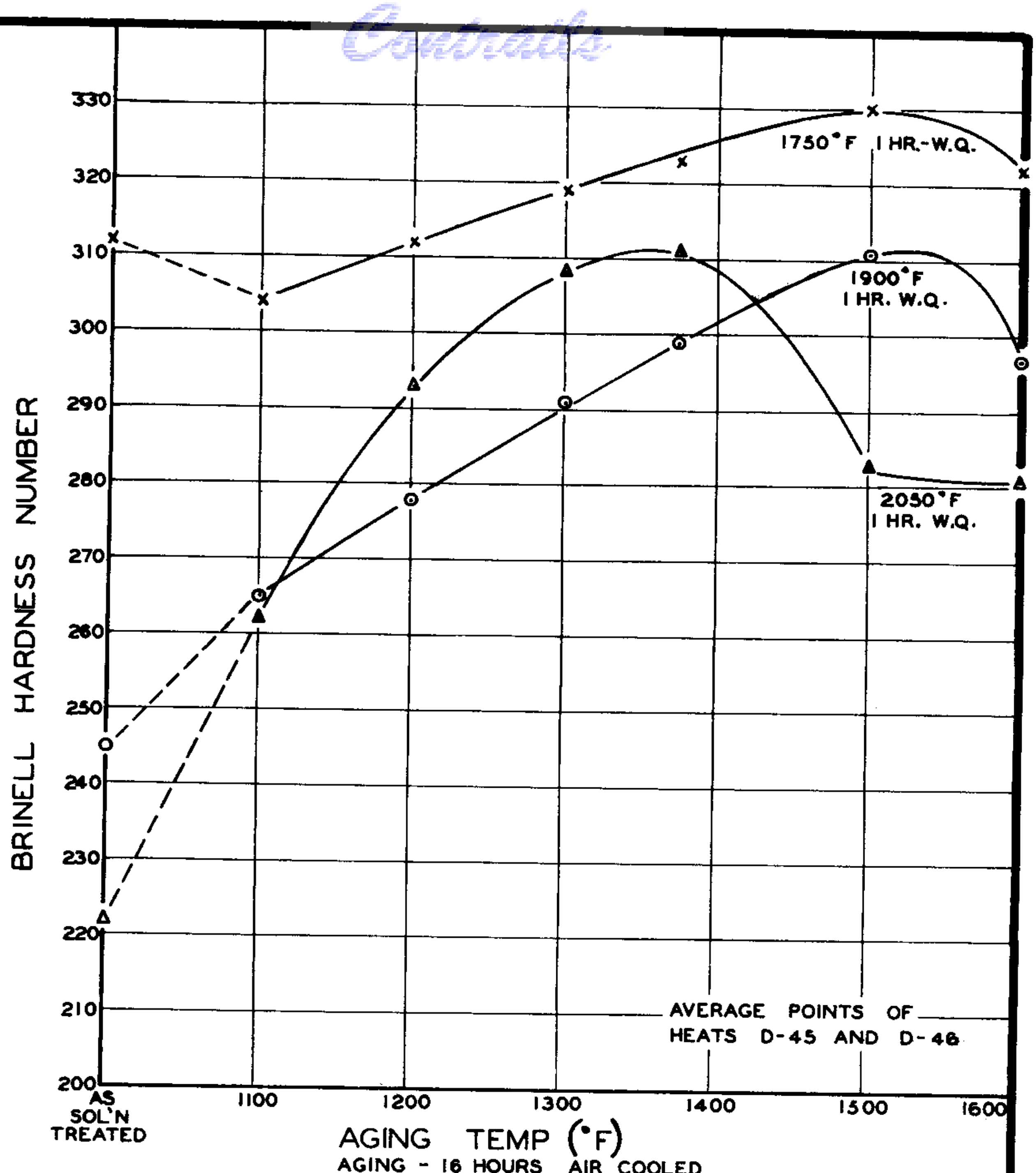


FIGURE - 19 EFFECT OF AGING TEMPERATURE ON THE ROOM TEMPERATURE HARDNESS OF D-183 ALLOY

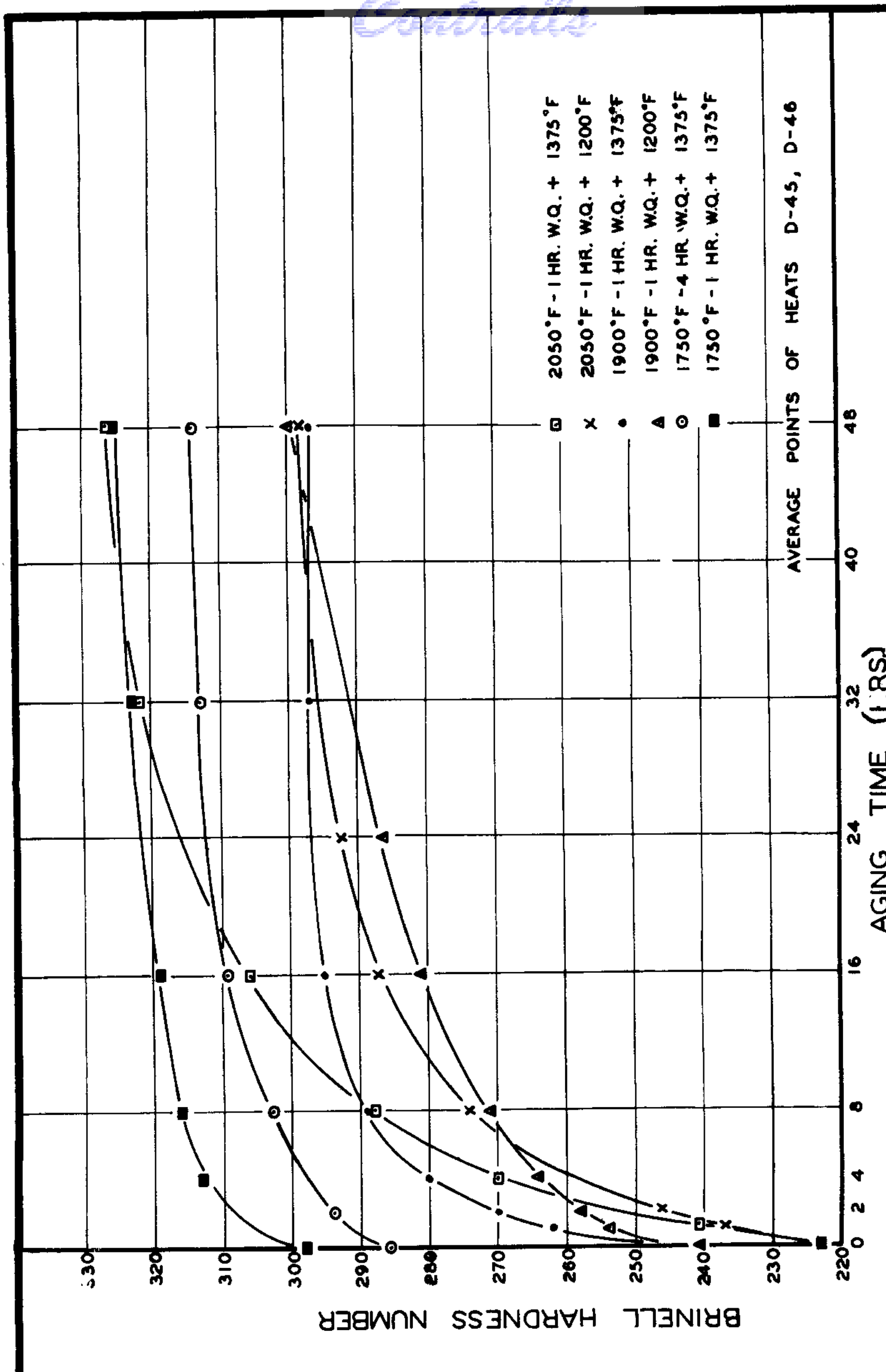


FIGURE - 20 EFFECT OF AGING TIME ON ROOM TEMPERATURE HARDNESS OF D-183 ALLOY

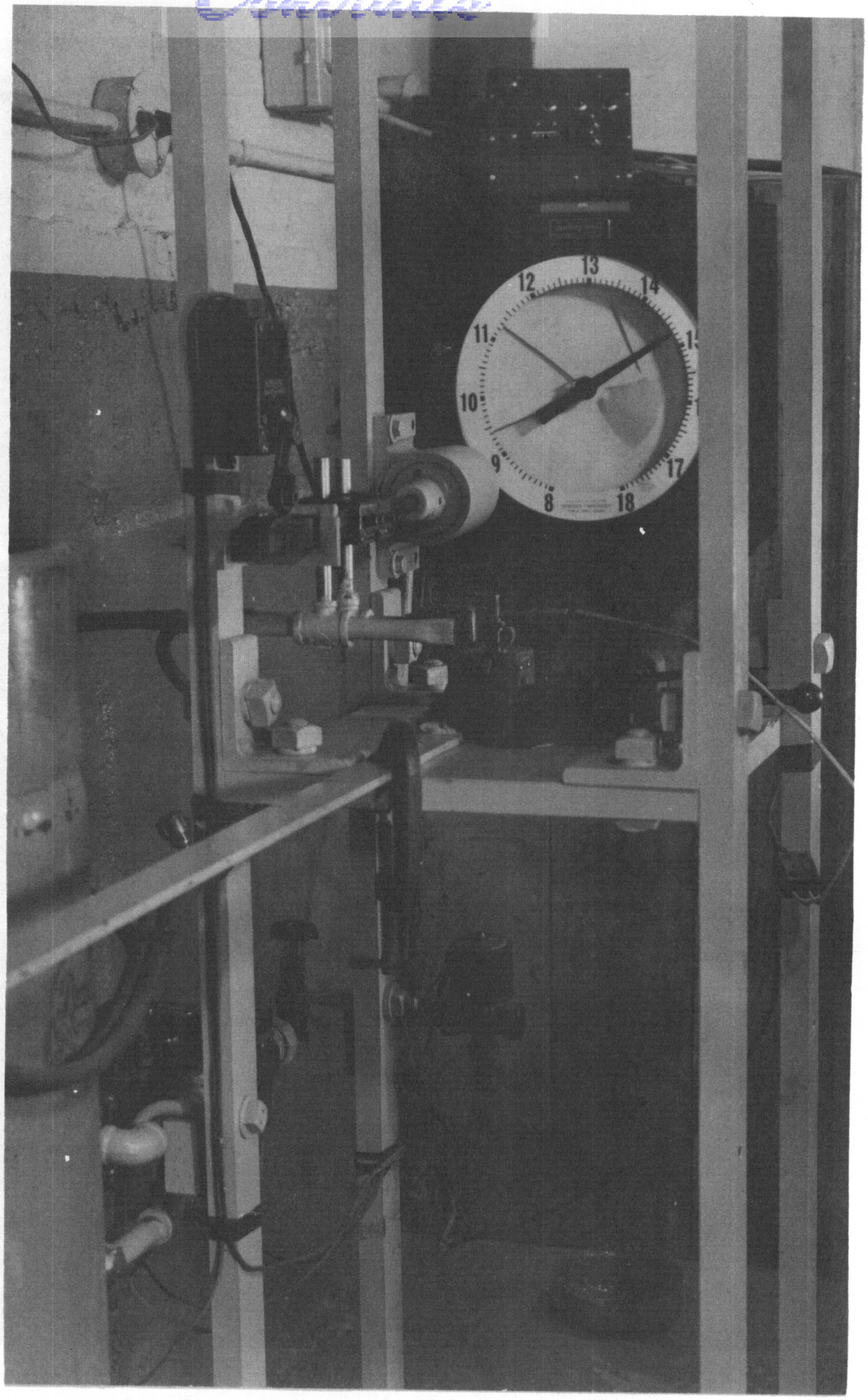


Figure 21. Single Sample Thermal Shock Testing Apparatus

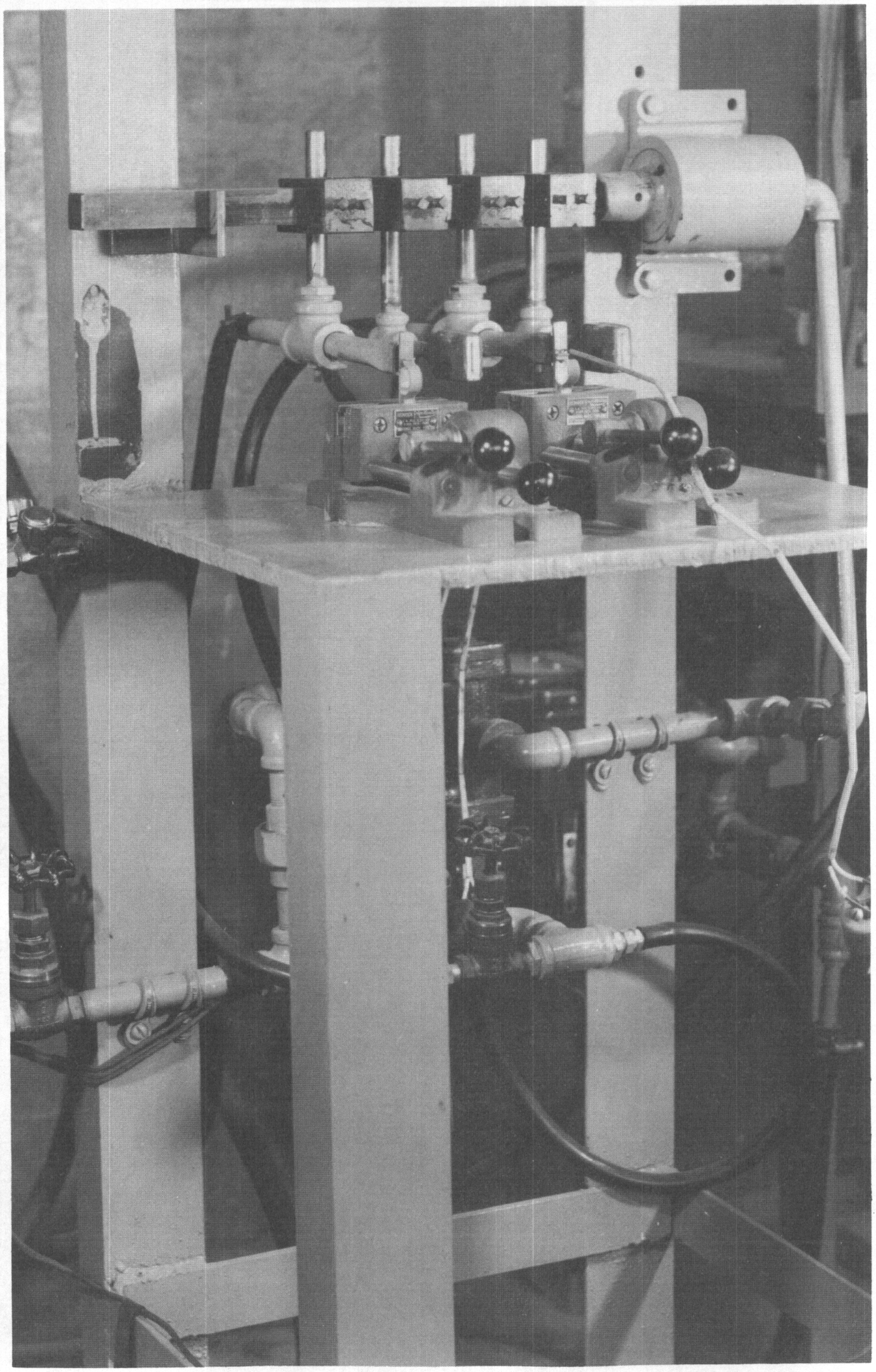
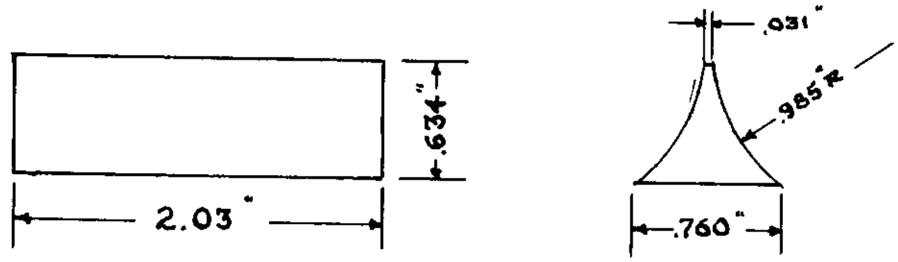
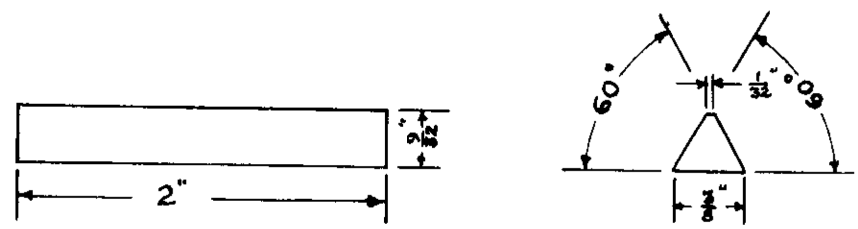


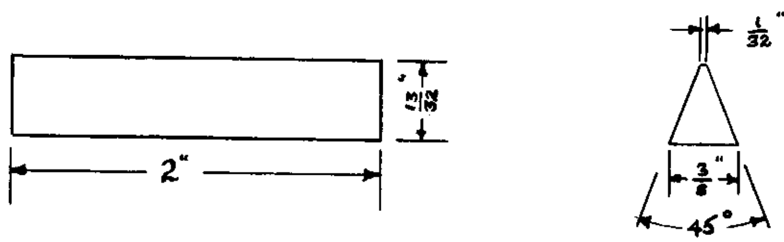
Figure 22. Dual Sample Thermal Shock Testing Apparatus



A - EXPERIMENTAL SAMPLE



B - EXPERIMENTAL SAMPLE



C - STANDARD SAMPLE

FIGURE - 23 THERMAL SHOCK TEST SPECIMENS

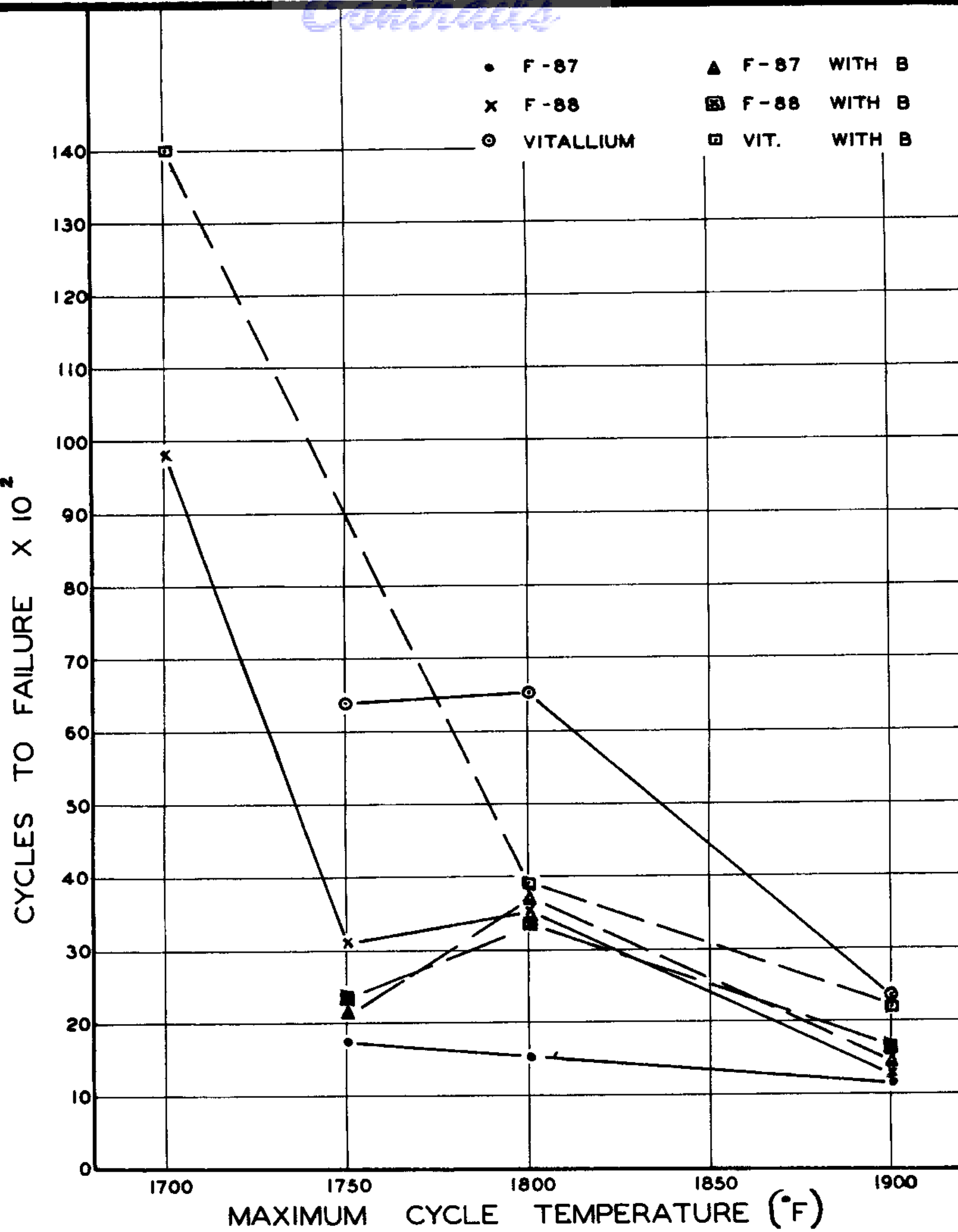


FIGURE - 24 THERMAL SHOCK PROPERTIES OF VITALLIUM, F-88, F-87 AND THEIR B MODIFICATIONS

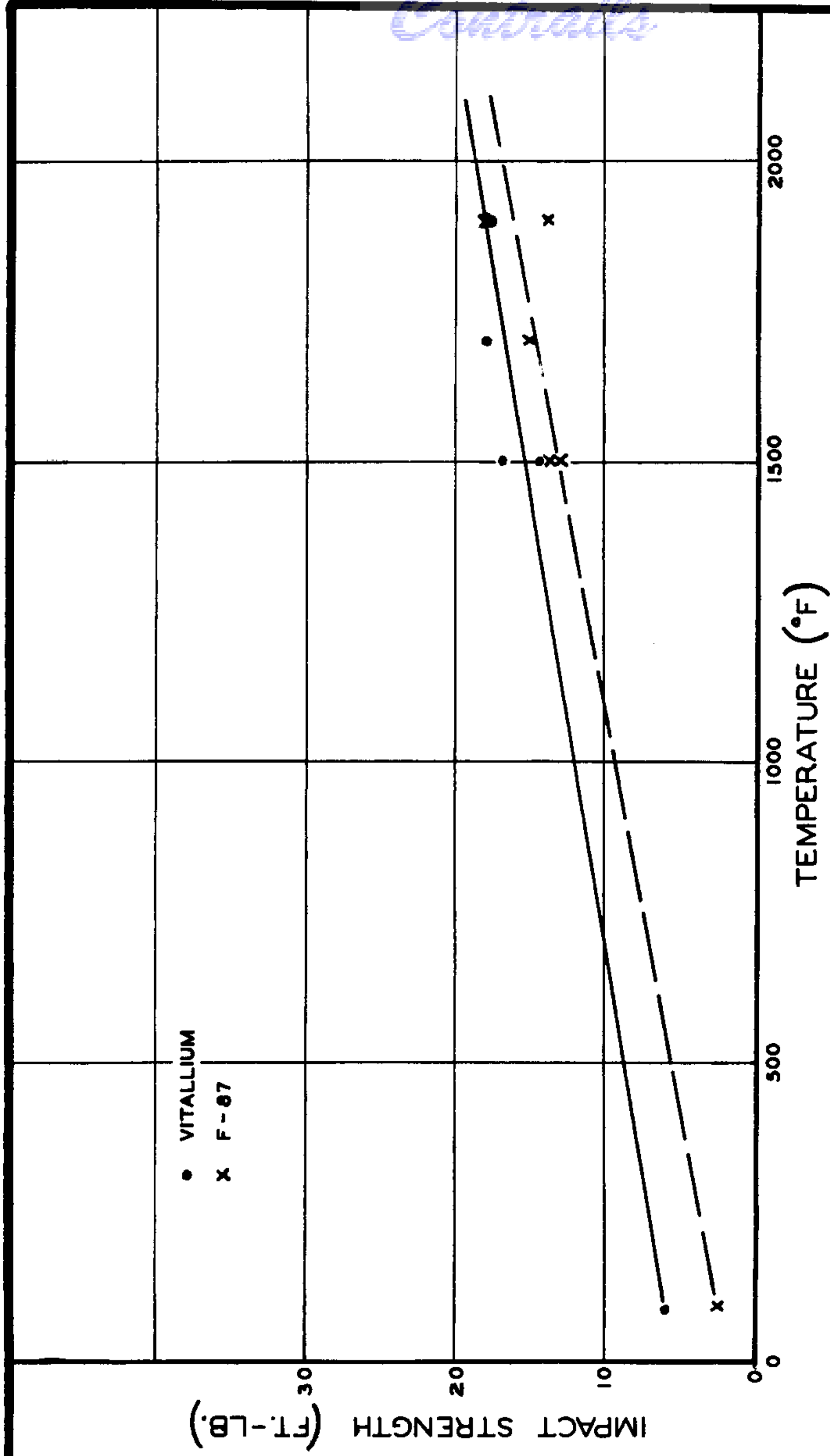
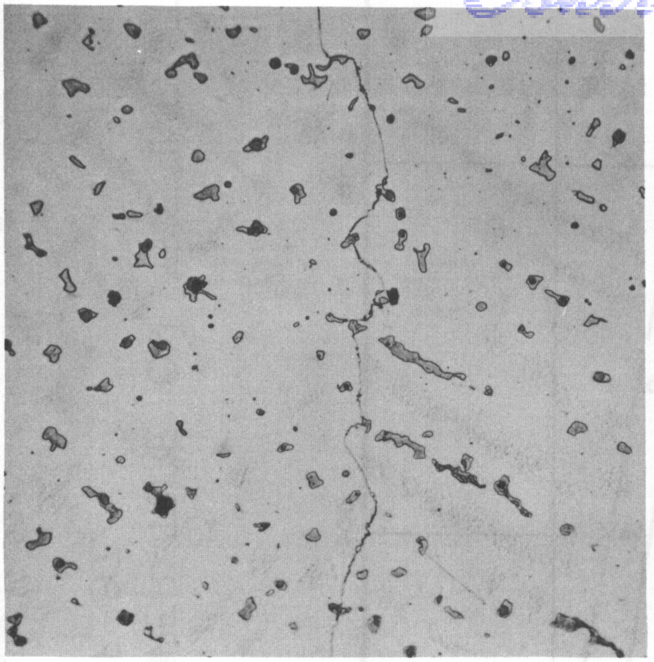
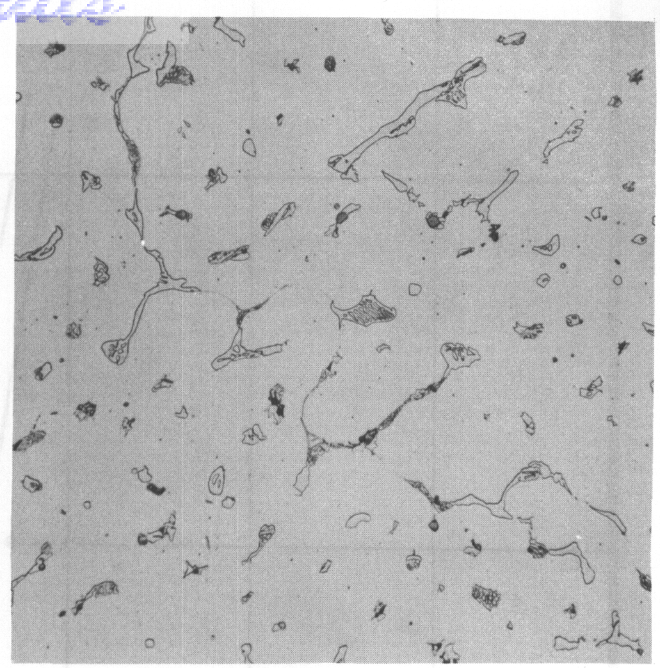


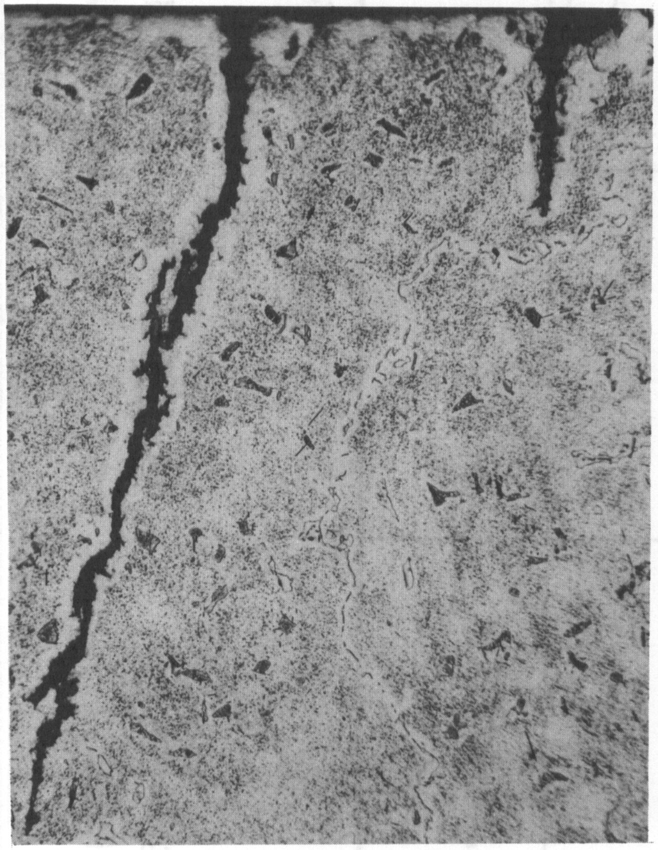
FIGURE -25 INFLUENCE OF TEST TEMPERATURE ON THE IMPACT STRENGTH OF CAST VITALLIUM AND F-87



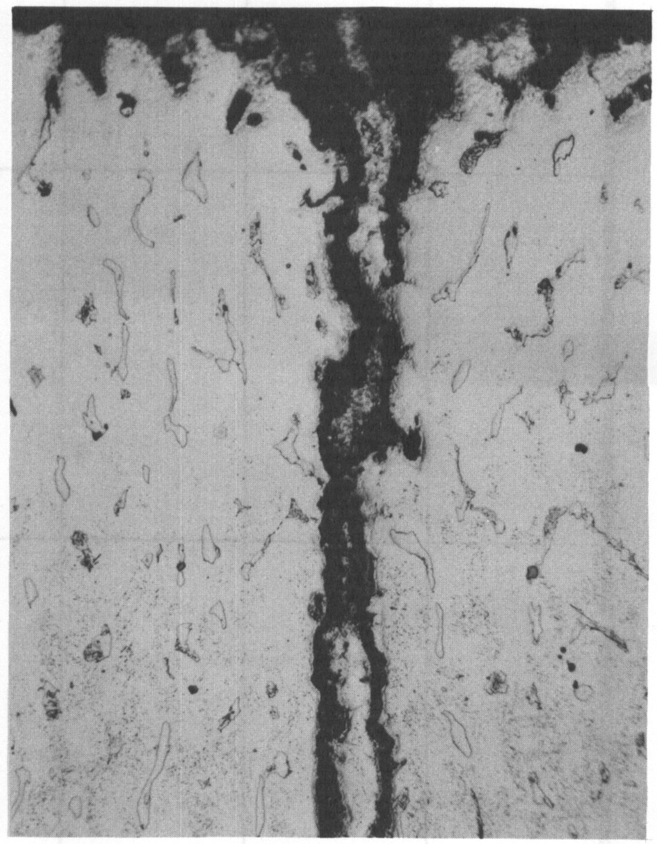
As-Cast Vitallium



As-Cast Vitallium + .20 B



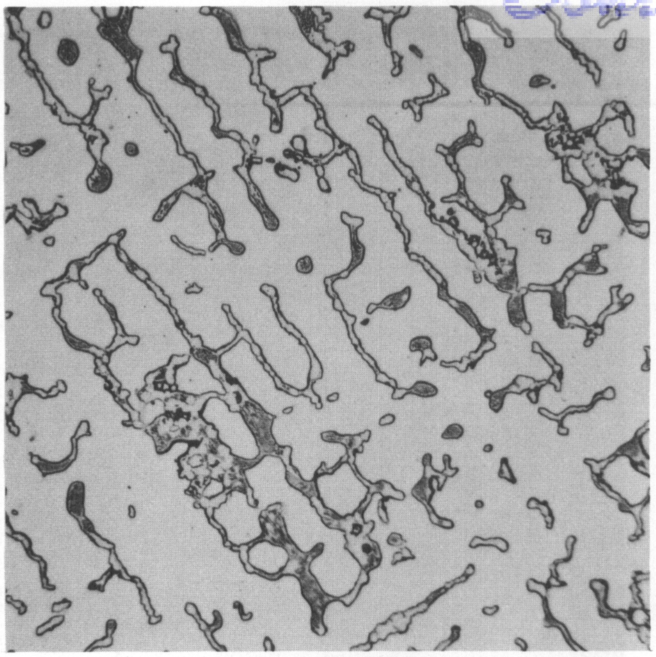
Thermal Shock Fracture
in Vitallium



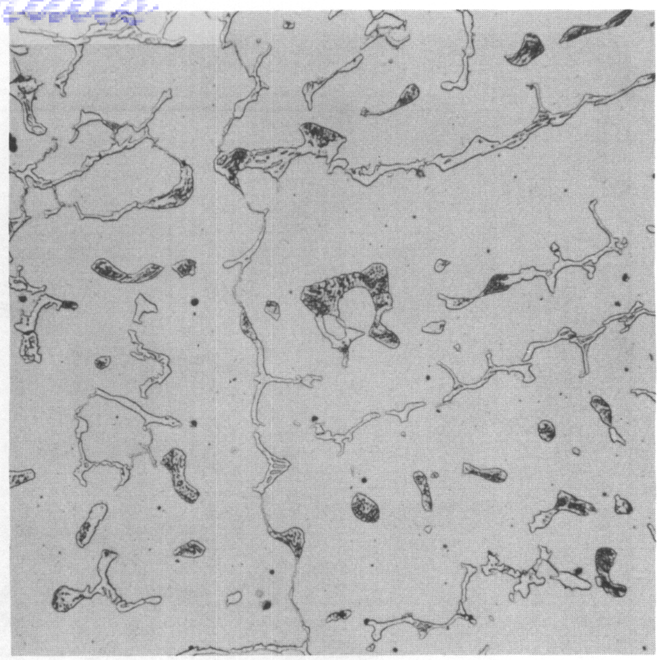
Thermal Shock Failure
in Vitallium + .20 B

Mag. 250X - Etchant: Electrolytic 10% Oxalic Acid

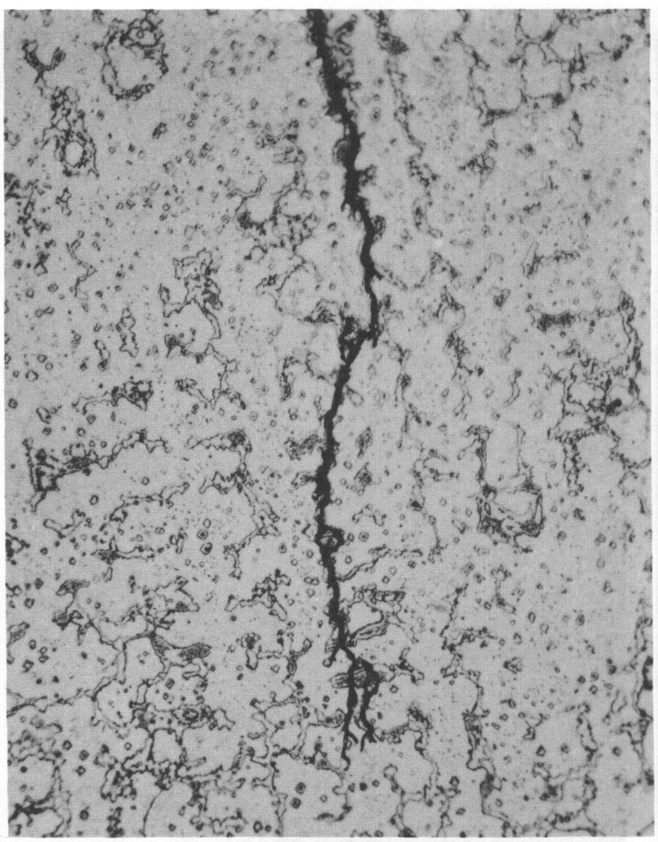
Figure 26 - Typical Microstructures of As-Cast and Failed Thermal Shock Samples of Vitallium and Vitallium + .20 B



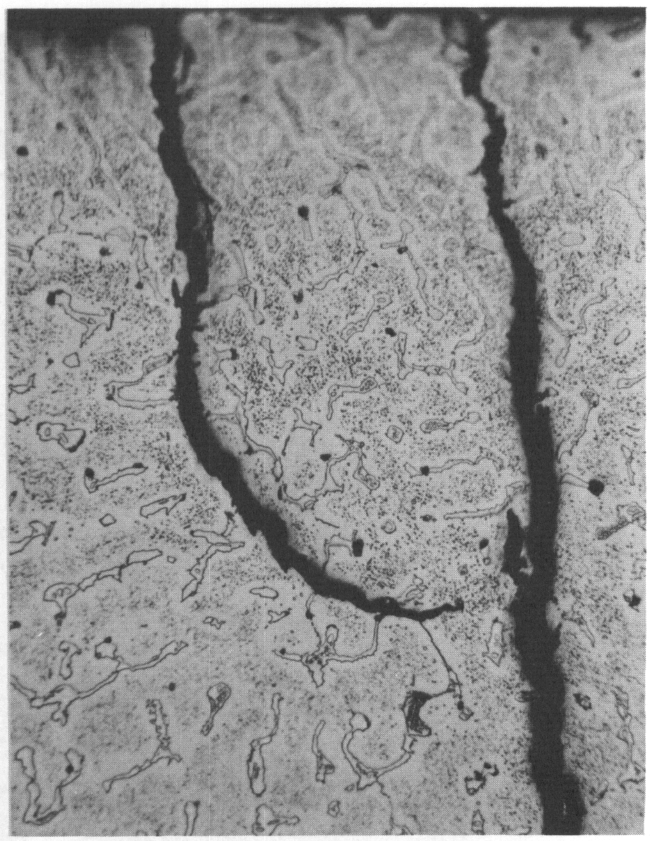
As-Cast F-88



As-Cast F-88 + .09 B



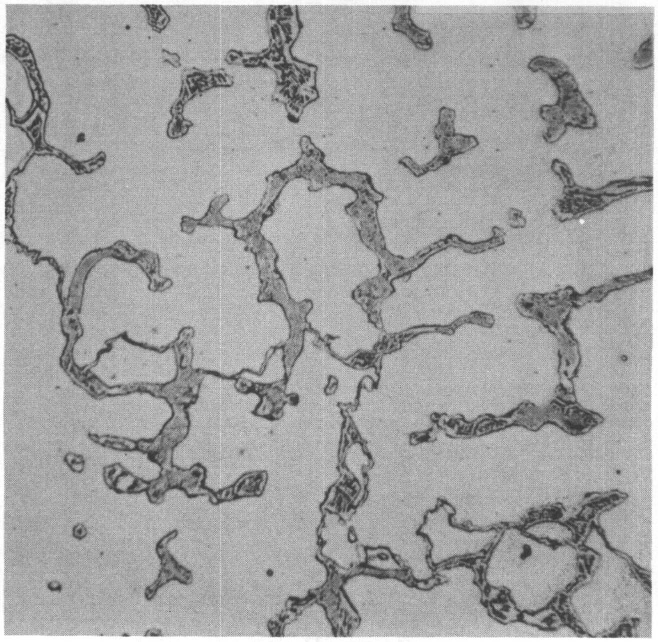
Fracture in F-88



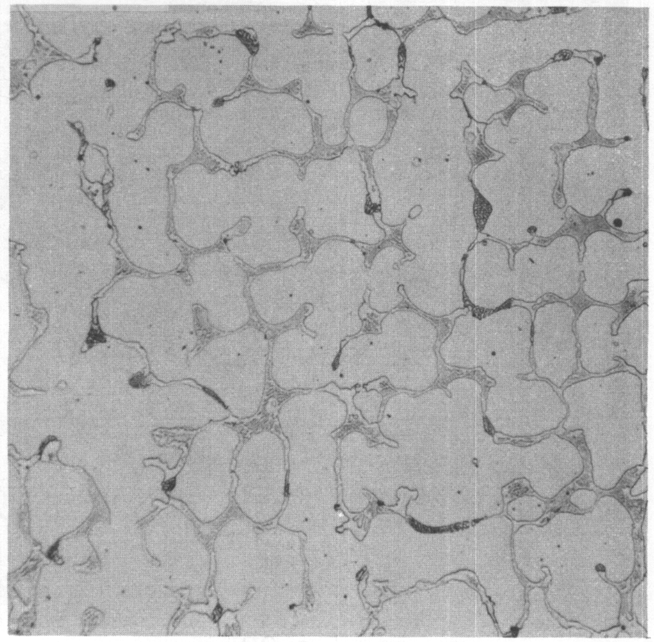
Fracture in F-88 + .09 B

Mag. 250X - Etchant: 10-HNO_3, 10-$HC_2H_3O_2$, 15-HCl

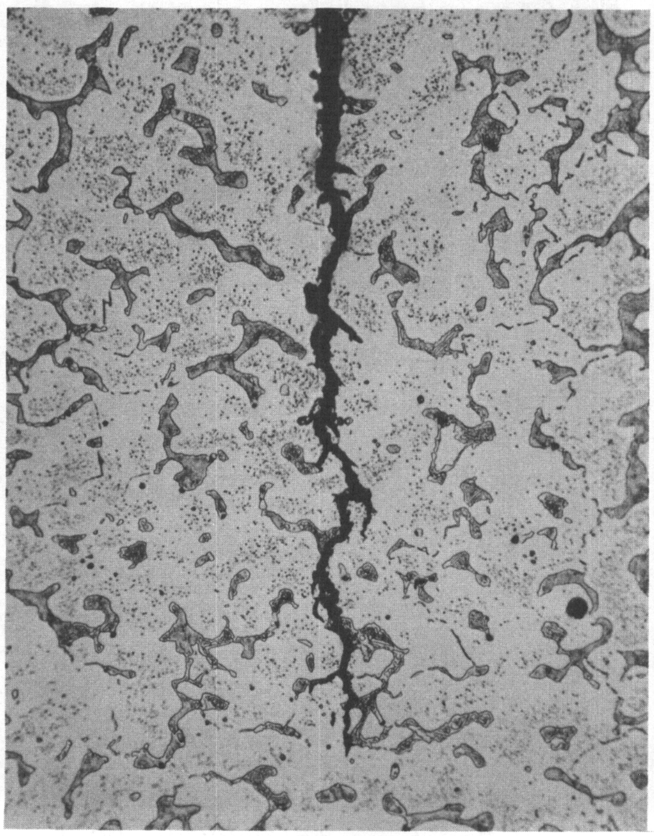
Figure 27 - Typical Microstructures of As-Cast and Failed Thermal Shock Samples of F-88 and F-88 + .09 B



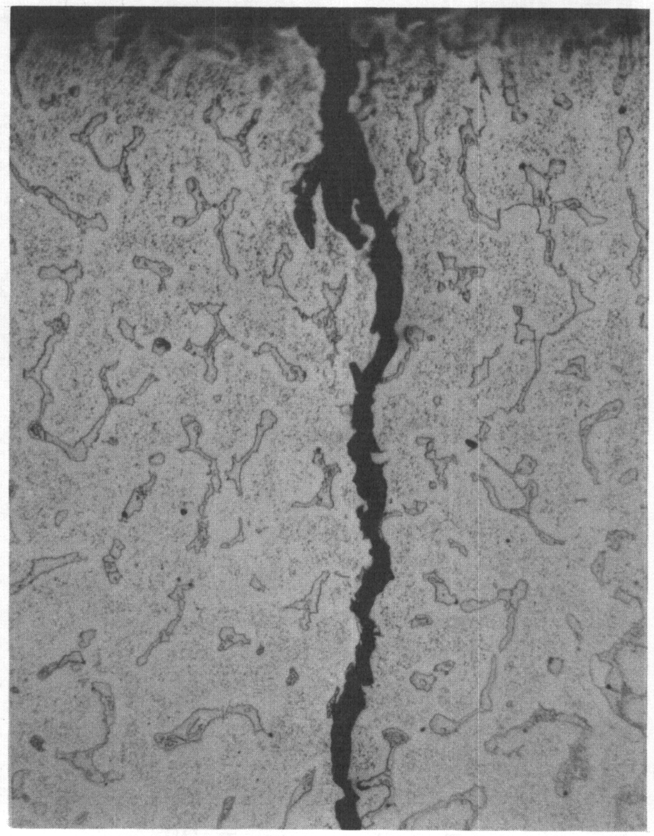
As-Cast F-87



As-Cast F-87 + .14 B



Fracture in F-87



Fracture in F-87 + .14 B

Mag. 250X - Etchant: 10-HNO_3, 10-$HC_2H_3O_2$, 15-HCl

Figure 28 - Typical Microstructures of As-Cast and Failed Thermal Shock Samples of F-87 and F-87 + .14 B.

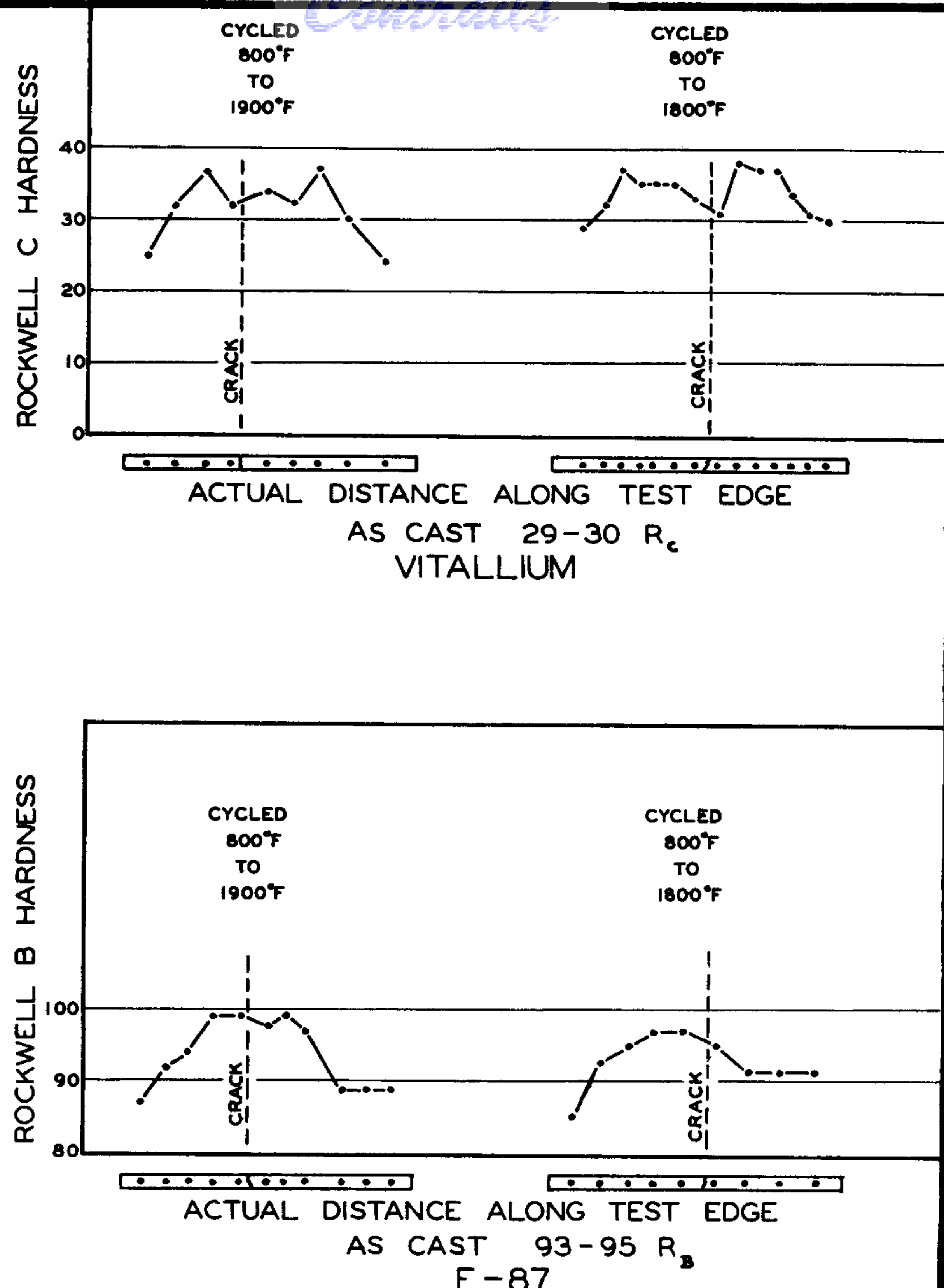


FIGURE - 29 HARDNESS CHANGES RESULTING FROM THERMAL SHOCK TESTING