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PROGRESS REPORT

ON

METALLURGICAL RESEARCH RELATING TO THE DEVELOPMENT OF METALS AND ALLOYS FOR USE IN THE HIGH-TEMPERATURE COMPONENTS OF JET-ENGINES, GAS TURBINES AND OTHER AIRCRAFT PROPULSION SYSTEMS

June 7, 1948

I - CONTINUATION OF PRIOR WORK

Considerable work has been done on report preparation. All items listed in the March program report will be reported during July.

II - FUNDAMENTAL STUDIES OF THE HIGH-TEMPERATURE PROPERTIES OF METALS

Fundamental studies are in progress to establish the fundamental processes by which treatments and composition control properties of commercial alloys at high temperatures. As yet work has been confined to Low-Carbon N155 alloy and progress has been reported twice previously (see references 1 and 2). The work is divided into two sections: studies of solution treated and aged material and studies of rolled structures. Electron microscopic work has been started as an additional technique for the studies. Brief descriptions of experimental techniques used, results, and interpretation of the data obtained since the last report covering this field are summarized below. Since the work outlined is to a large extent still in progress, the discussion given is to be considered tentative and subject to further modification as additional data becomes available.

Solution-Treated and Aged Studies

Studies of solution-treated and aged Low-Carbon N155 are the most advanced. X-ray line intensity changes upon aging after solution treating 10 hours at 2200° F have been completed at two temperatures of aging, 1400° and 1600° F for time periods ranging from one-half to 1000 hours. Figure II-1 shows the data obtained. Similar studies are in progress on identical material aged at 1200° F. As mentioned in reference (2), the drops in intensity are probably associated with the formation of precipitant nuclei surrounded by strain whose period of occurrence is of the order of 10^{-6} to 10^{-8} cm. The apparently anomalous slowing up of the process at 1600° F when compared with the same process at 1400° F may be due to 1600° F being close to the temperature at which the precipitants dissolve. Becker (see reference 3) has rather thoroughly explored the rates of precipitation for alloys near the border between the single and two phase regions and found that the rate is greatly slowed down due to the difficulty of nucleus formation when precipitation occurs at a temperature close to the single phase region. If one accepts Becker's treatment of the subject, the minima in Figure II-1 can be associated with nuclei formation. Apparently long range diffusion processes do not control because it requires longer aging times at 1600° F to reach a minimum than at 1400° F, and diffusional rates should be greatly accelerated at 1600° F when compared with the same process at 1400° F.

A metallographic study of the 1200° F aging for time periods up to 1000 hours showed secular changes which were first, a gradual outlining of the grains with thin black appearing lines which gradually widened, and then second, precipitation as extremely fine particles along, but not in,

the boundaries. This precipitate appeared to be evolving along certain crystallographic planes. Plate II-5 illustrates this structure. The preferred orientation is in marked contrast to the rather random orientation of the precipitate particles in structures aged at 1400° or 1600° F (see reference 2). This apparently is an example of the temperature dependence of the mechanism of precipitation within the same alloy system.

Figure II-2 shows the lattice parameters of the same samples used to obtain the intensity values of figure II-1. Back reflection technique was used with unfiltered Cu radiation. Chemically precipitated silver with an assumed lattice parameter of 4.0778 Å was used as a standard. Examination of figure II-2 shows that the minimum (111) line intensity upon aging at 1400° F is coincident with the start of a slight lattice expansion. Subsequent removal of the short period lattice distortions as depicted by the increase in line intensity values is associated with a lattice contraction which apparently is not complete after 1000 hours at 1400° F. Aging at 1600° F is associated with an immediate and continued drop in lattice parameter for time periods up to 1000 hours. It is believed that the two processes, nucleus formation and removal of the precipitant atoms both contribute to the character of the curves of figure II-2. Nucleus formation could at both aging temperatures be associated with a slight lattice expansion. Diffusion out of the matrix of the precipitant atoms at both temperatures, and the resultant build-up of precipitant phase particle size will result in a decrease in matrix lattice parameter due to removal of elements with larger atomic radii than the mean radius of the matrix atoms. At 1400° F diffusion is slow enough that the expansion associated with nucleus formation can be clearly seen. Subsequent particle growth by diffusion and

precipitation then acts to reduce the lattice parameter. At 1600° F the reduction of lattice parameter by precipitant particle growth occurs so early in the precipitation process by virtue of more rapid diffusion that the slight lattice expansion is largely masked. In addition, thermodynamic considerations point to the fact that larger (and fewer) nuclei are formed at 1600° F when compared to formation at 1400° F. It is thus possible that a greater degree of lattice depletion of precipitant atoms to form nuclei occurs at 1600° F.

Past investigation in age-hardening systems has relied heavily on hardness measurement. While it is believed that hardness as commonly determined by commercial hardness measuring instruments is not a good basic measure of precipitation processes, such data are included in figure II-3 in order to fit the precipitation mechanism outlined above into more common terms. It can be seen that the period of nucleus formation at 1400° F is not associated with any marked increase in hardening but that growth by diffusion is. Aging at 1600° F is in turn associated with an immediate small increase in hardness and then a gradual decrease.

The Mott and Nabarro elastic strain theory associated with precipitant particles before and after breaking away from the matrix to form a truly separate phase with definite interfaces (see reference 4) can be assumed to hold. With this basis, it can be said that the high elastic strains associated with precipitant particles before breaking away are controlling with aging at 1400° F and that breaking away to form interfaces had not taken place at the time periods up to 1000 hours. On the other hand the larger particles initially formed upon aging at 1600° F and rapidly expanded in size break away quickly, and thus the fact that hardness values observed reached a maximum can be explained. In addition, fewer

particles are formed at 1600° F when compared to the number formed at 1400° F, and thus the maximum hardness achieved was lower. The earlier breaking away to form phase interfaces at 1600° F age is due to the larger nuclei and subsequent particles formed. Mott and Nabarro (see reference 4) found critical sizes of precipitant particles with interface formation and, apparently, the critical size at 1600° F age is reached almost at the same time that the original nuclei is formed.

Some differentiation between the type of lattice strain indicated by intensity measurements shown in figure II-1 and the strain controlling the hardness outlined in figure II-3 should be made. The strains causing drop in line intensity are, as mentioned before, short period disturbances, 10^{-6} to 10^{-8} cm being the order of the period magnitude. In view of the fewer and larger nuclei formed at 1600° F when compared with those formed at 1400° F, the period at 1600° F is probably near the upper limit of the above estimation while the period at 1400° F is near the lower limit. Slowing down of rate of nuclei formation and disruption of the regular period of the disturbances set up initially acts to restore line intensities at both temperatures of aging. Elastic strains surrounding the precipitate particles can, however, still exist as long as they do not have either a quite regular short period spacing or a possibly limited magnitude. This latter condition is probably necessary since the lattice errors or disruptions as postulated by Dehlinger (see reference 5) in analyzing line intensities were of small magnitude. The elastic strains still surrounding the precipitate particles and controlled primarily by diffusion rates then act to control hardness as outlined previously.

From a commercial standpoint the necessity of using a 10-hour solution treatment is important. Figure II-5 shows the first data obtained to bear on this question. It is evident that a small but measurable difference in behavior on aging does take place when solution treating time is increased from 1 to 10 hours at 2200° F. If the cause of the initial slight drops in hardness is due to a relief of internal quenching stresses, then the 1 hour solution treatment curve could be made to lie upon the 10-hour solution curve by a shift along the time axis. This in turn suggests that the effect of the 10-hour solution treatment is to move the whole precipitation process backwards along the time scale through greater removal of nucleation centers, for example. More work must be carried out before any definite conclusions can be drawn, however.

The end result of the investigations on solution treated and aged material is the determination of creep properties on a fundamental basis. Figure II-4 shows the creep data obtained to date on stock identical with that used to obtain the data shown in figures II-1, 2, and 3. The tests shown all broke within 12 hours and the minimum rates obtained all occurred within six hours. No interpretation of the minima observed can be made as yet, since, for example, no measure of the effect of stress field existing during testing on the rates of precipitation has been made. Tests at 30,000 psi are in progress to measure the shift, if any of this minima. In any event, it is evident that an optimum treatment exists for the material aged at 1400° and 1600° F when tested at 1200° F and that this optimum is quite probably associated with some stage in the precipitation process covered above.

Rupture tests at high stresses have been nearly completed for the same series of heat treatments covered above. Stress versus time for rupture is plotted for stock aged at 1400° F in figure II-8 and for material aged at 1600° F in figure II-10. Figure II-9 shows the effect of aging at 1400° F on the time for rupture at both 60,000 psi and 70,000 psi. Figure II-11 shows the same thing at 70,000 psi for stock aged at 1600° F. Further work must be carried out before correlation of the observed rupture behavior with the foregoing structure studies can be made. The following conclusions, however, can be drawn from the data obtained to date:

1. There is a marked change of curvature of the stress versus log-rupture time plots when comparing unaged material with stock aged for long time periods at 1400° F and for shorter periods at 1600° F.
2. An optimum aging time apparently exists for aging at either 1400° or 1600° F. Figure II-9 indicates that this optimum time as a function of the stress to which the material is subjected.
3. The aging time for optimum service at a given stress is probably much shorter at 1600° F than at 1400° F and further the optimum rupture time obtained is greater for aging at 1400° F and for aging at 1600° F

Studies on Rolled Structures

In an attempt to explain the increase in creep and rupture strength when Low-Carbon N155 is cold or hot-cold rolled, at least after solution treatment, three hypotheses have been advanced to act as a working basis for investigation:

1. The increase in strength is due to the induced mode of precipitation. It might be possible, for example, that the precipitates are formed preferentially along slip lines, etc., this position offering the maximum resistance to possible further movement.

2. The increase in strength is due to the preferred orientation produced by rolling.

3. The increase in strength is due to the disruption of the matrix structure as separated from the effect of precipitation.

Work on hypotheses 1 and 3 is well under way, but to date it has been confined to material solution treated 1 hour at 2200° F and water quenched before any rolling or cold working operations. In addition, all rolling has been done at temperature below the range for which recrystallization during rolling will take place.

Plates II-1 through II-4 illustrate typical microstructures of solution treated and cold and hot-cold worked material. From such micrographic examination, the following conclusions have been obtained:

1. Rolling at about 80° F results in no noticeable changes in microstructure up to 15 percent reduction in cross section. At approximately 15 percent reduction slip lines or bands start to appear in specimens etched and polished after reduction. With increasing amount of rolling, slip bands or lines appear with increasing frequency. The period of the lines is of the order of 10^{-4} cm. See plate II-1.

2. Hot-cold rolling at 1400° F of solution treated material results in more pronounced grain boundaries when compared to structures rolled at 80° F, but otherwise the structures appear much the same.

3. Aging after large cold reduction results in precipitant coming out preferentially at the slip lines and practically none generally within the matrix. It also appears that the precipitation process in general is vastly accelerated. After intermediate amounts of cold reduction, the precipitation process appears to be accelerated a lesser amount. Precipitate, however, still comes out relatively quickly at any slip line which may be present. Also since fewer slip lines are present to act to deplete the matrix of precipitant upon aging, more precipitate appears generally distributed throughout the matrix and in the grain boundaries. (Compare the cold-rolled structures of figure II-3 and II-4.)

4. Aging of structures hot-cold rolled at 1400° F to intermediate reductions appears to be more accelerated than in material cold rolled the same amount or solution treated. The particles within the matrix are larger and formed into needles or platelets, which indicates probably that they are well advanced past the interface formation stage. Concentration at the grain boundaries also appears to be quite heavy.

The inference drawn from the above discussion is that an increase of working temperature from 80° to 1400° F results in more working of the matrix in such a manner as to accelerate the general matrix precipitation. Precipitation at any slip lines present is extremely rapid and of the same character in either case.

X-ray investigations of the structure of cold and hot-cold rolled materials have been started in an attempt to determine the validity of hypothesis 3. Two immediate objectives are:

1. The effect of various amounts and temperature of deformation on the diffraction lines of the matrix.

2. Effect of annealing after reduction on the diffraction lines of the matrix.

Figure II-6 shows results of 1/2 height line width measurement as a function of amount of reduction at room temperature. Similar determinations are in progress for rolling at 1400° and 1800° F. Unfiltered chromium radiation was used with a circular camera equipped with slit collimator. The specimen was mounted with radiative surfaces normal to the incoming X-ray beam. Photometric measurements were made from the film record on the 220 lines (θ approximately 65°) with an L and N microphotometer. Plotting of the intensity versus θ curves was then carried out with the line widths at half height determined graphically. To correct for the presence of the α_1, α_2 doublet, half of the width of the line was determined on the side away from the α_2 line.

The asymptotic approach to a limiting line width with increasing reduction has been noted before by other investigators. The interpretation of exact mechanism of line widening due to deformation has been subject to much controversy. In the past, line widening has been attributed to fragmenting of the lattice into small particles of the same order of size as the wave length of the incident X-radiation. Another explanation has been that deformation introduces a variation of the lattice parameter from point to point within the lattice, and it is this variation in lattice parameter which causes widening. It appears from published evidence that the latter theory is the more likely with the proviso, however, that the lattice is also broken up into small blocks with slight variation in orientation from block to block. This variation in orientation is necessary to explain, for example, the transition of the diffraction line of a polycrystalline material

from one compound of Laue spots to a continuous band or ring upon deformation. Whether the small blocks are small enough to cause particle size widening cannot be answered as yet. It is thus apparent that studies of diffraction line widths and changes induced by deformation and subsequent treatments or service is at least one way of investigating high temperature properties on a fundamental basis.

Data as shown on figure II-7 outlines the restoration by annealing of a lattice previously cold-worked 15 percent. The type of restoration indicated by narrowing of the diffraction lines took place without microscopic evidence of recrystallization. Such data extended to include other annealing temperatures can lead directly to calculation of activation energies associated with this type of lattice restoration. The effect of temperature of deformation on activation energy can also be evaluated. Presumably the process controlling here is one of diffusion of atoms to undeformed or more perfect crystal centers.

In addition to the line width studies outlined above, preliminary determination of recrystallization conditions as commonly defined from micrographic examination has been made. On the solution treated stock reduced 40 percent at room temperature, recrystallization was first observable after 10 hours at 1800° F and after 1000 hours at 1600° F. Appearance of faint and diffuse Laue spots in the diffraction lines of the 40 percent cold-reduced material after 10 hours at 1600° F indicated "submicroscopic" recrystallization had taken place.

Electron Microscope Techniques

Experimental techniques have been developed to permit the use of the electron microscope in examination of metallographic structures in order to obtain additional data for the fundamental studies in general. Advantage is to be taken of both the increased depth of focus obtainable and the greater usable magnification. Collodion replicas, shadow cast with chromium under vacuum as outlined by Williams (see reference 6) have been used to date. Plate II-7 illustrated typical results. It will be noted that with the use of location lines scribed on the metal surface identical areas may be examined both optically and electronically. This should be an aid in assigning significance to structures revealed by the electron microscope at low magnification in the more familiar terms of optical microscopy. The replica shown was taken directly from the etched surface prepared for the optical photomicrograph. The low magnification used was solely for locating purposes, and higher magnification could be readily used over the desired area. The development of such electron microscope techniques is continuing.

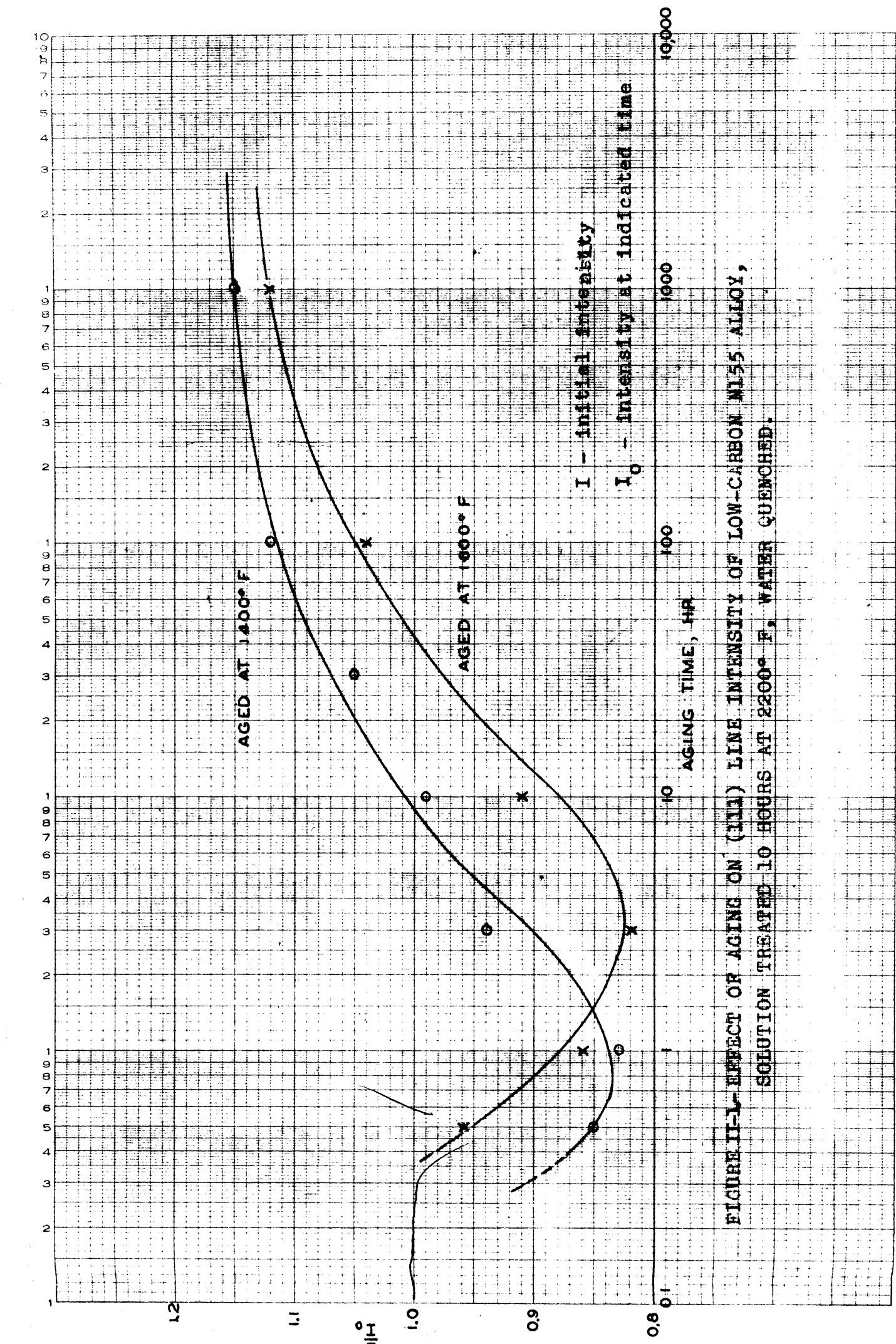


FIGURE II-1-EFFECT OF AGING ON (III) LINE INTENSITY OF LOW-CARBON M155 ALLOY,
 SOLUTION TREATED 10 HOURS AT 2200° F, WATER QUENCHED.

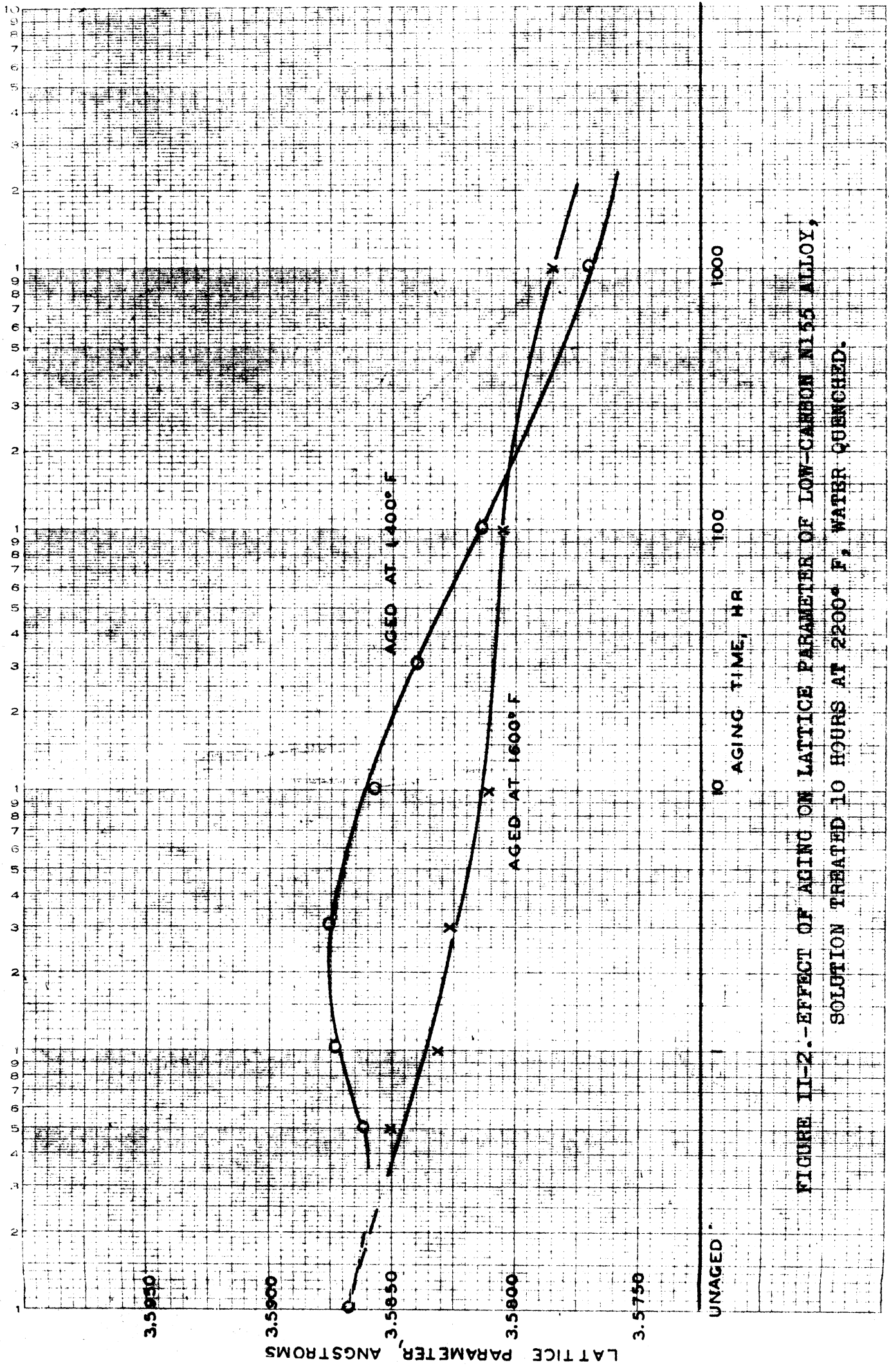


FIGURE II-2.-EFFECT OF AGING ON LATTICE PARAMETER OF LOW-CARBON NICKEL ALLOY,
SOLUTION TREATED 10 HOURS AT 2200° F, WATER QUENCHED.

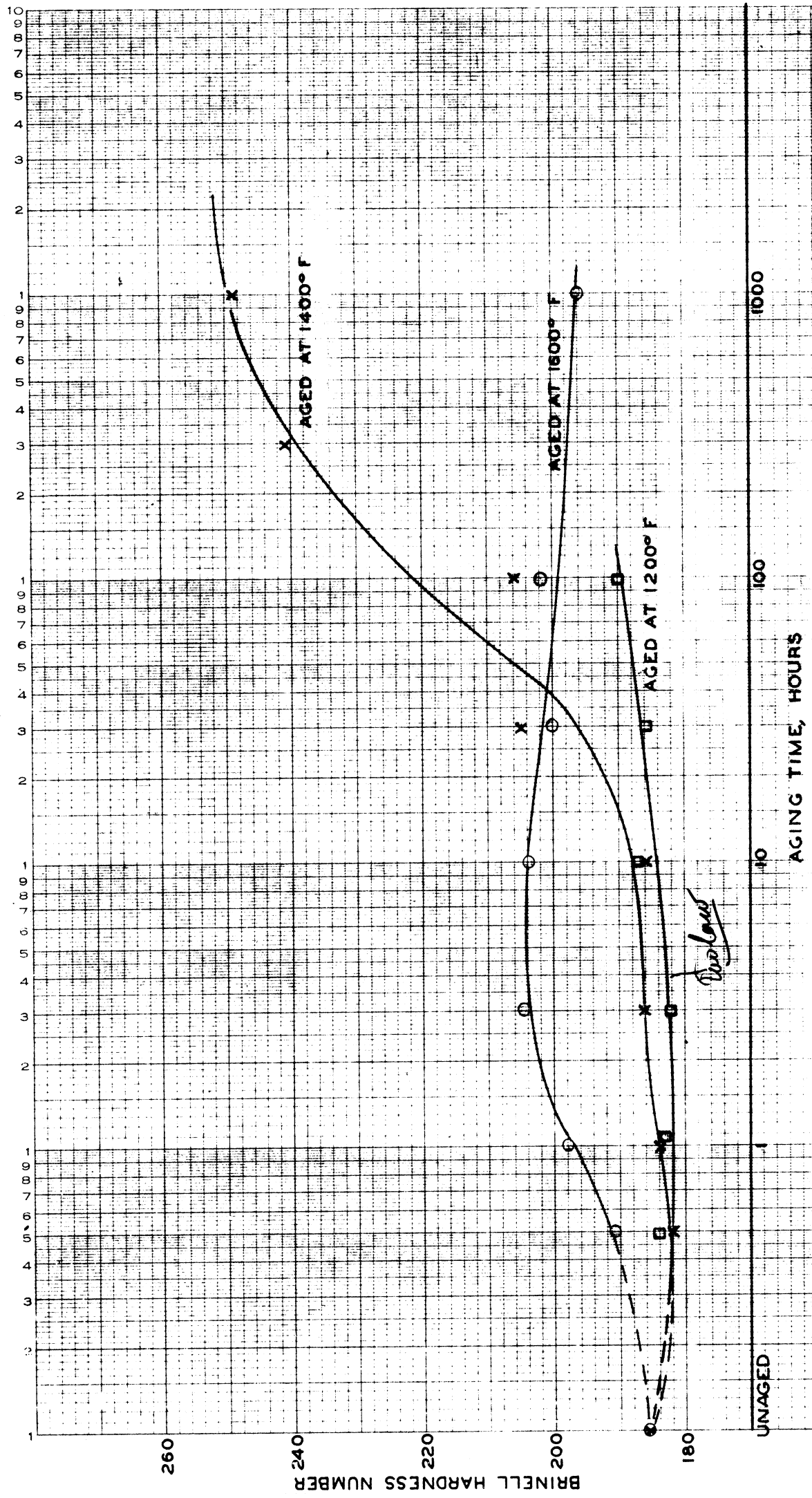


FIGURE 11-3. EFFECT OF AGING ON HARDNESS OF LOW-CARBON NL55 ALLOY SOLUTION TREATED 10 HOURS AT 2200° F, WATER QUENCHED.

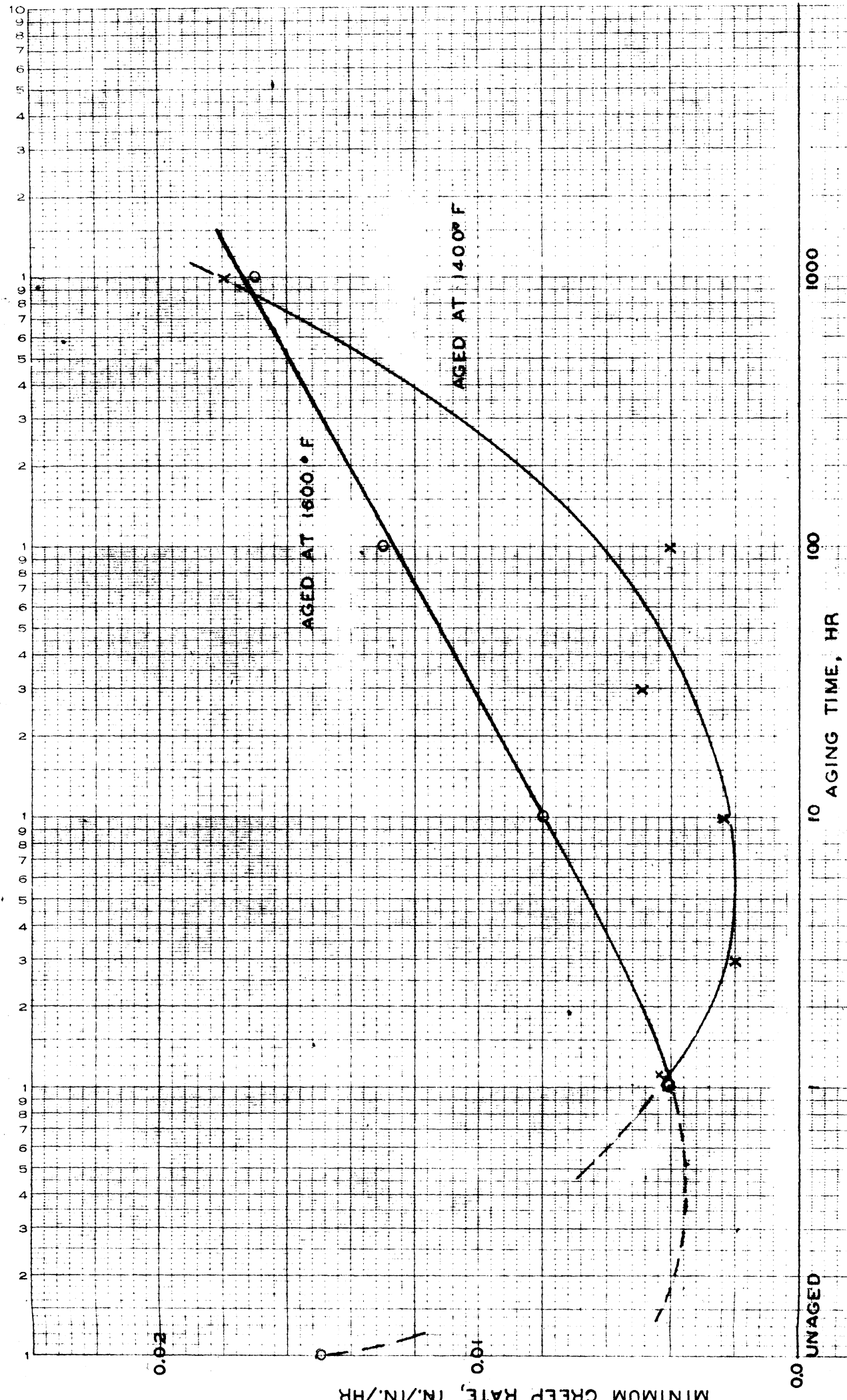


FIGURE II-4.- EFFECT OF AGING ON MINIMUM CREEP RATE AT 60,000 PSI OF LOW-CARBON N155 ALLOY, SOLUTION TREATED 10 HOURS AT 2200° F, WATER QUENCHED.

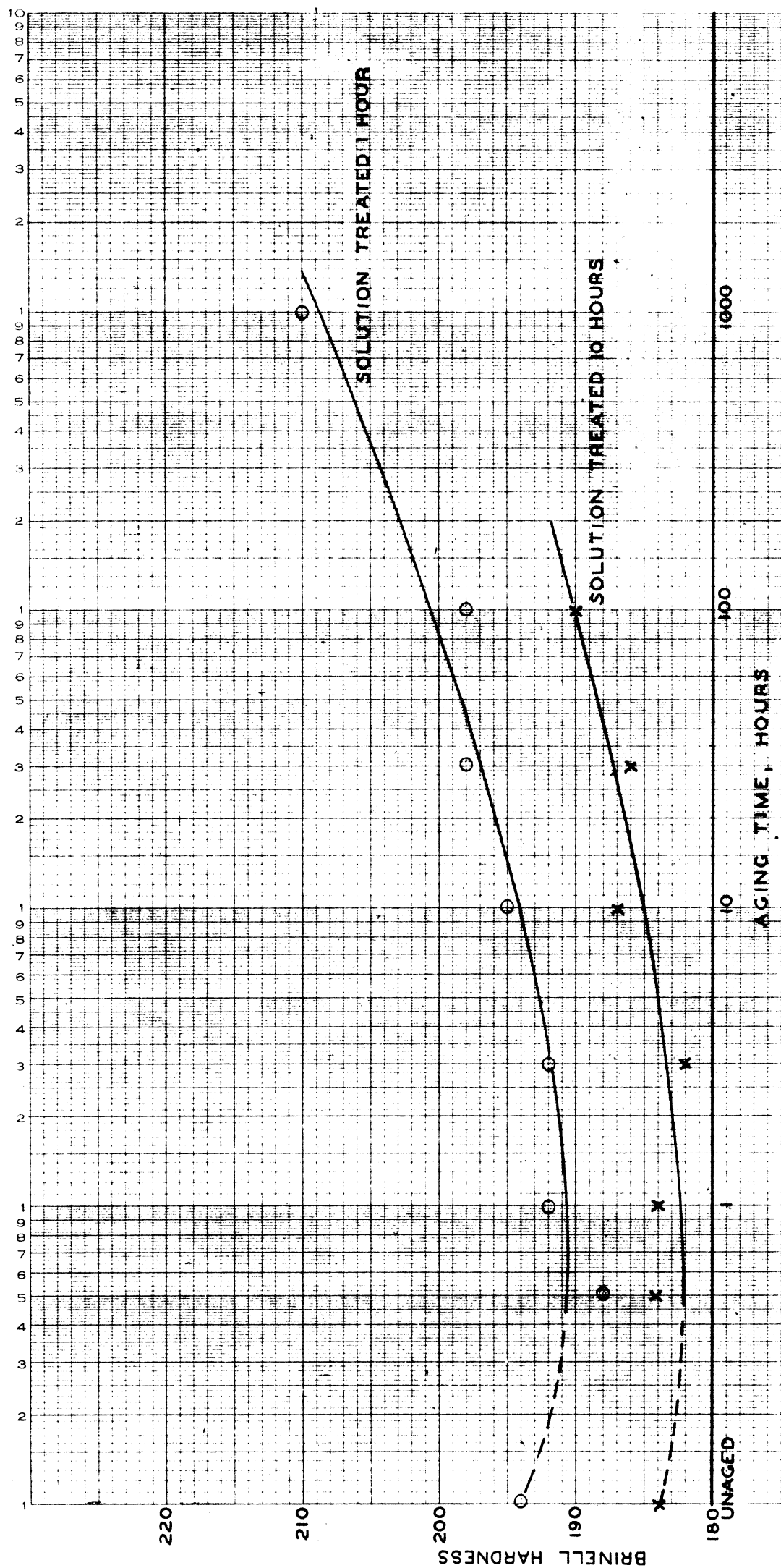


FIGURE 11-5-EFFECT OF AGING AT 1200° F ON HARDNESS OF LOW-CARBON M155 ALLOY, SOLUTION TREATED AT 2200° F, WATER QUENCHED.

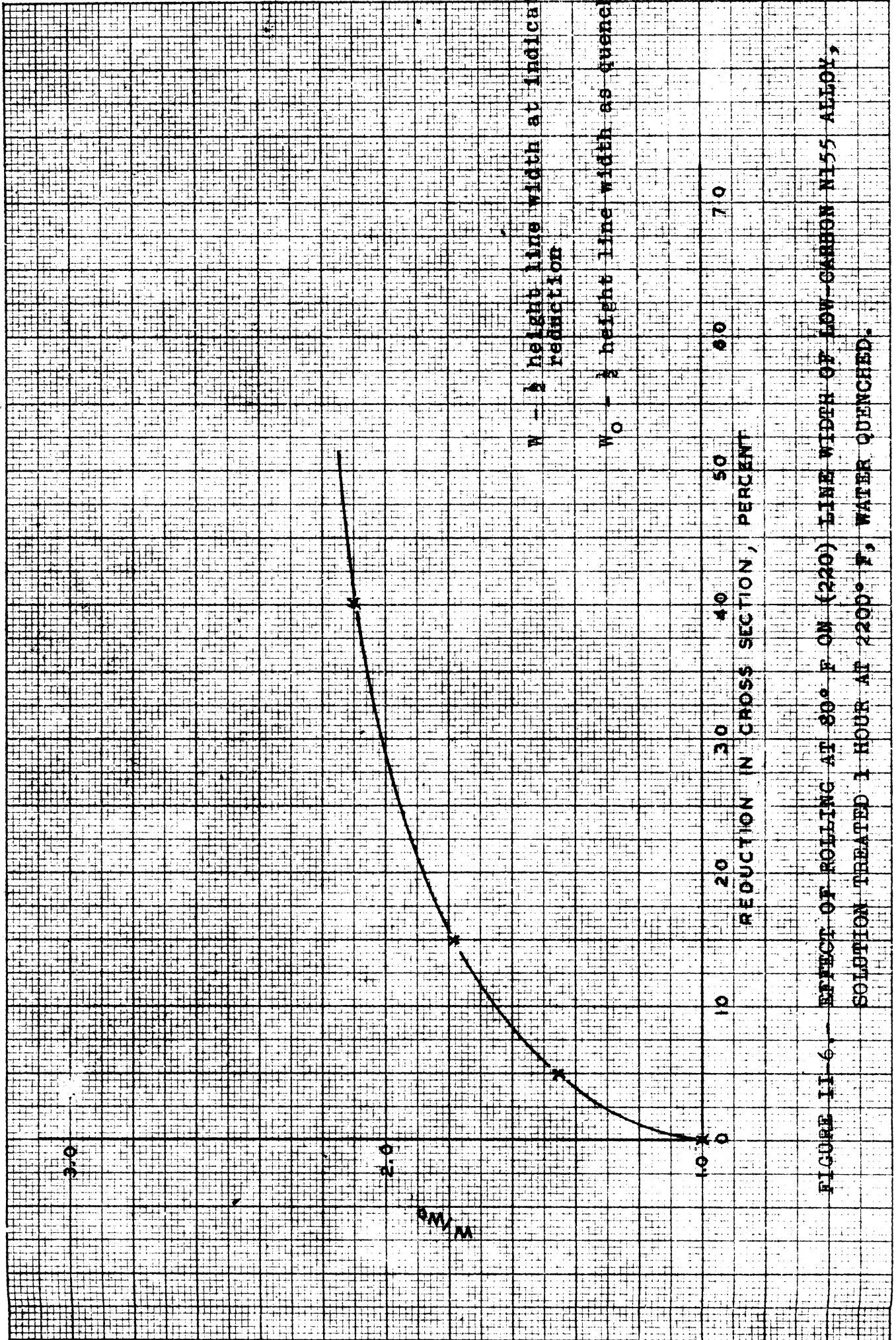


FIGURE 11-6. EFFECT OF ROLLING AT 89° F ON (220) LINE WIDTH OF LOW-CARBON NI55 ALLOY, SOLUTION TREATED 1 HOUR AT 2200° F, WATER QUENCHED.

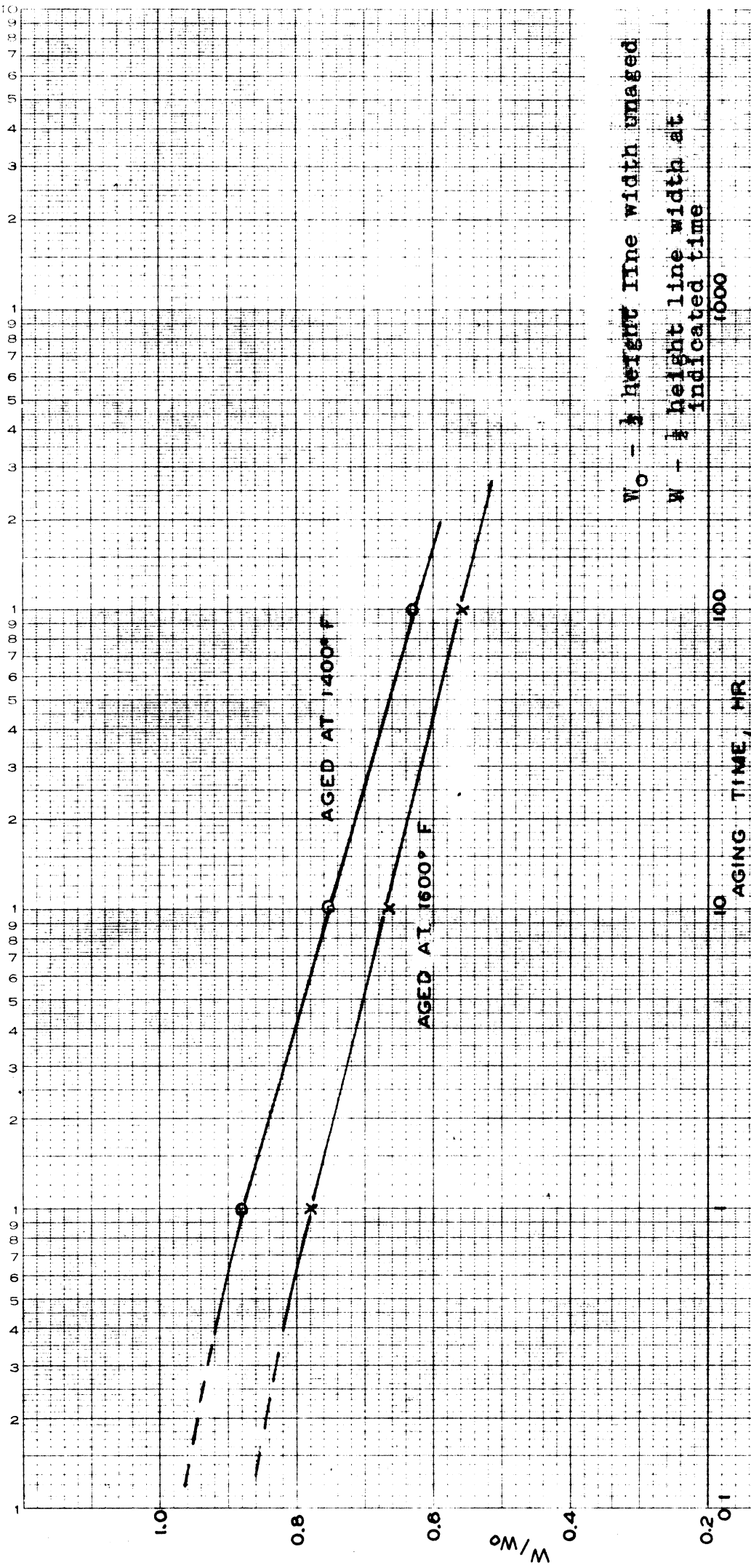


FIGURE II-7.—EFFECT OF AGING ON (220) LINE WIDTH OF LOW-CARBON N155 ALLOY, SOLUTION TREATED 1 HOUR AT 2200° F, WATER QUENCHED AND GOLD REDUCED 15% IN CROSS-SECTION AT 80° F.

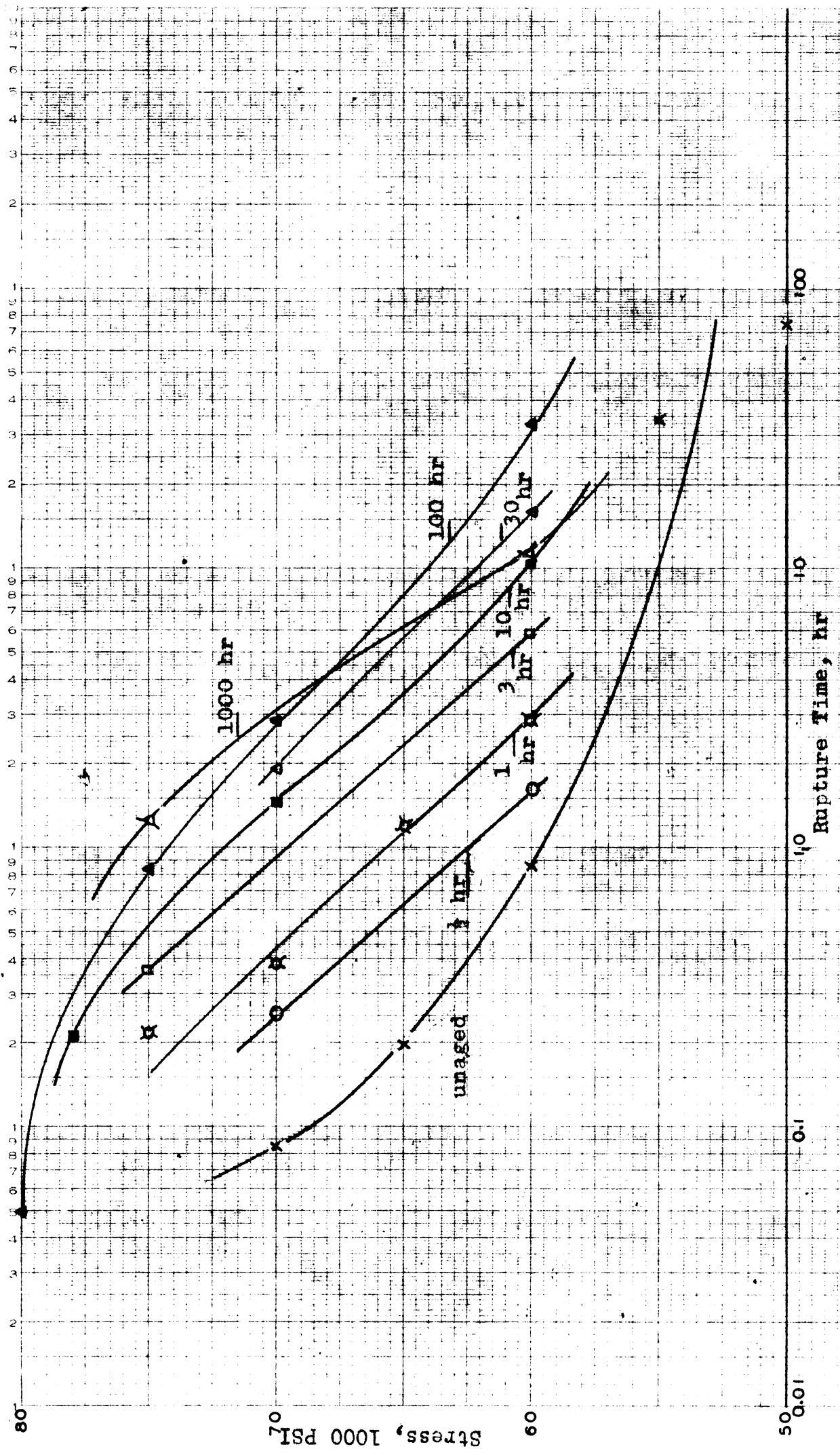


FIGURE II-8. - EFFECT OF AGING AT 1400° F ON RUPTURE STRENGTHS OF LOW-CARBON M155 ALLOY, SOLUTION TREATED 10 HOURS AT 2200° F AND WATER QUENCHED.

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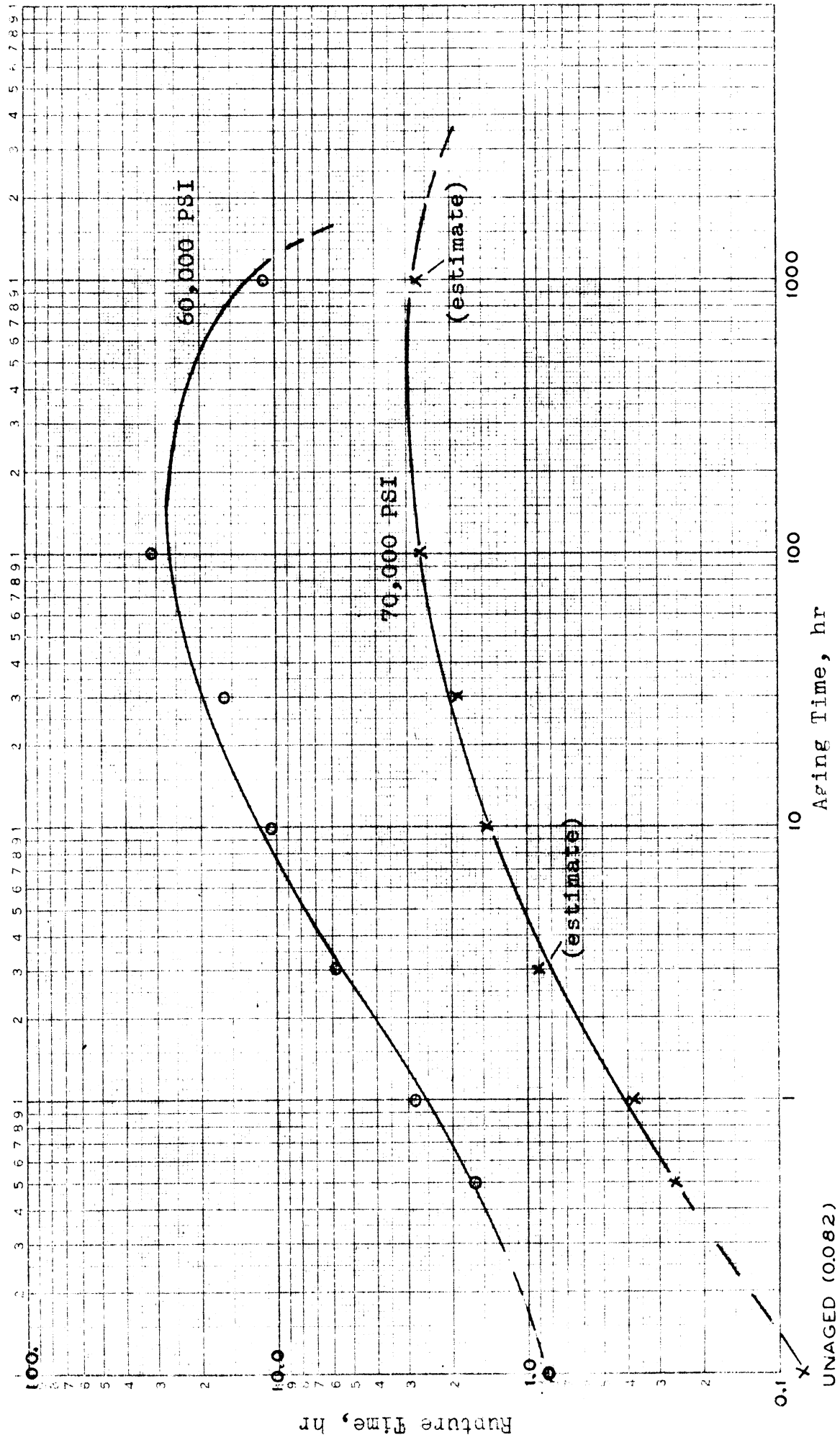


FIGURE II-9.-EFFECT OF AGING AT 1400° F ON 1200° F RUPTURE STRENGTH OF LOW-CARBON NI55 ALLOY, SOLUTION TREATED 10 HOURS AT 2200° F AND WATER QUENCHED.

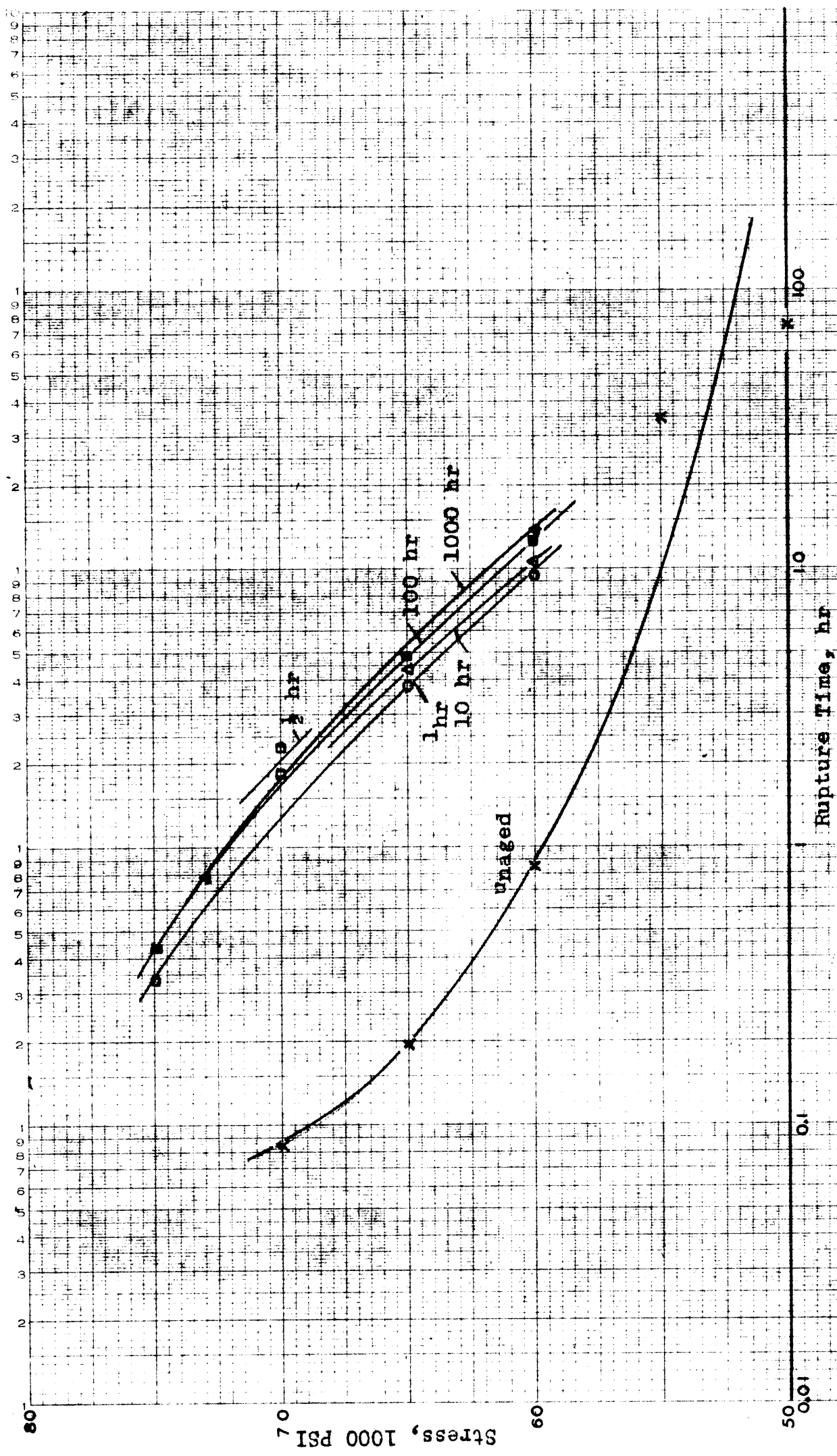


FIGURE 11-10.—EFFECT OF AGING AT 1600° F ON RUPTURE STRENGTH OF LOW-CARBON M155 ALLOY, SOLUTION TREATED 10 HOURS AT 2200° F AND WATER QUENCHED.

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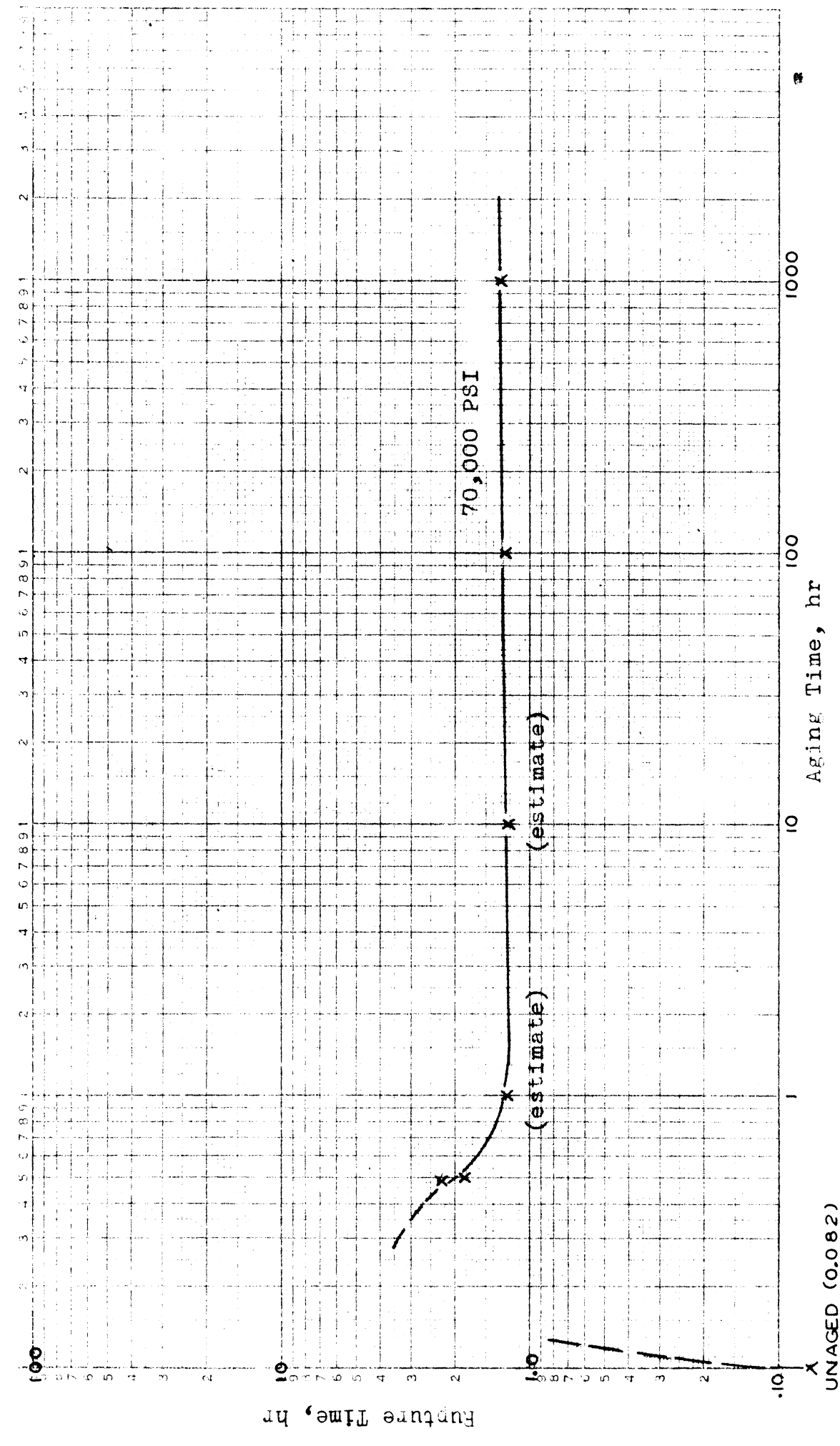
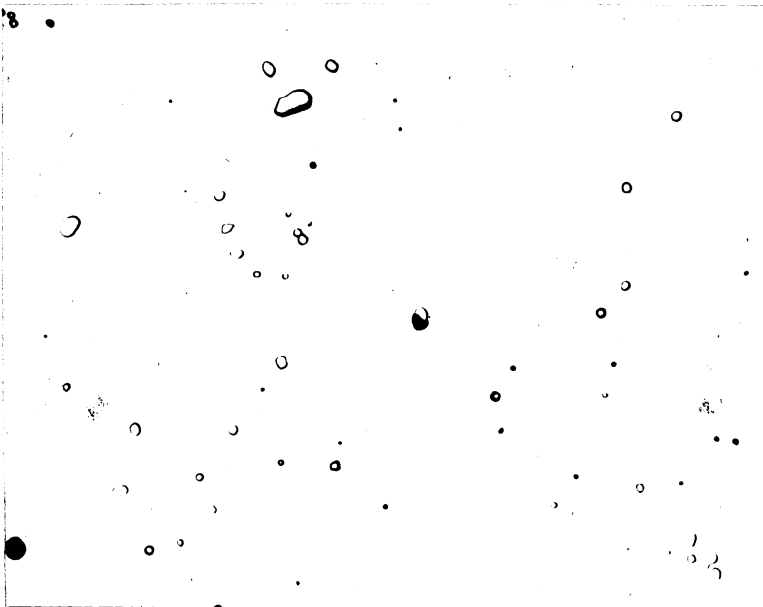
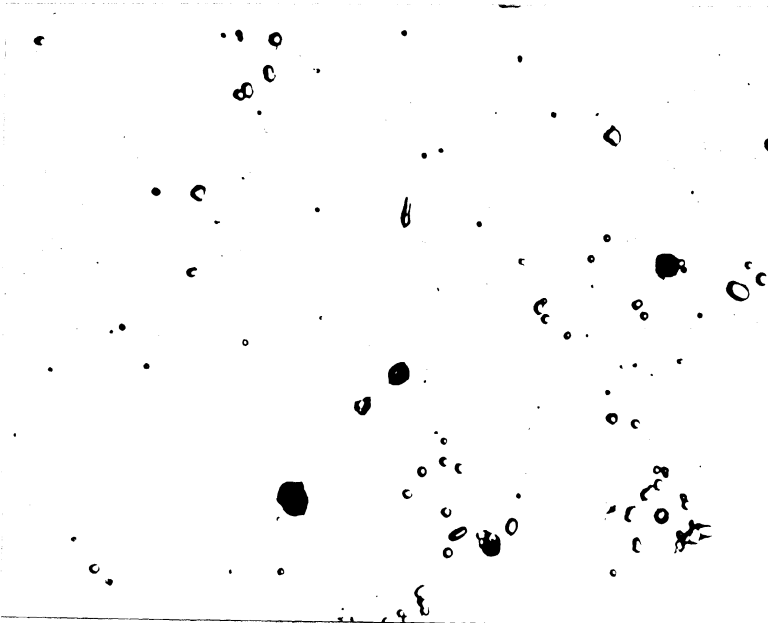


FIGURE II-11.-EFFECT OF AGING AT 1600° F ON 1200° F RUPTURE STRENGTH OF LOW-CARBON NI55 ALLOY, SOLUTION TREATED 10 HOURS AT 2200° F AND WATER QUENCHED.



48-351

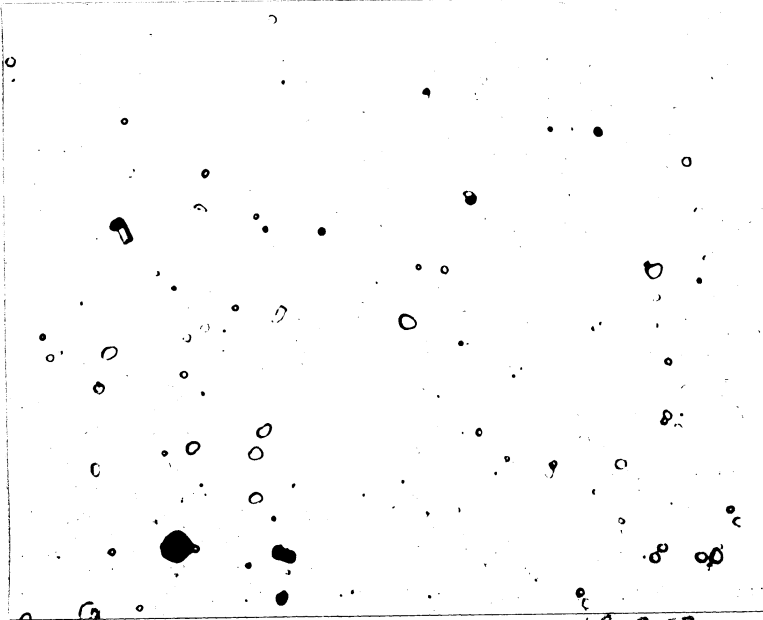
0% reduction
(10 hour solution treatment)



48-352

15% reduction

X1000

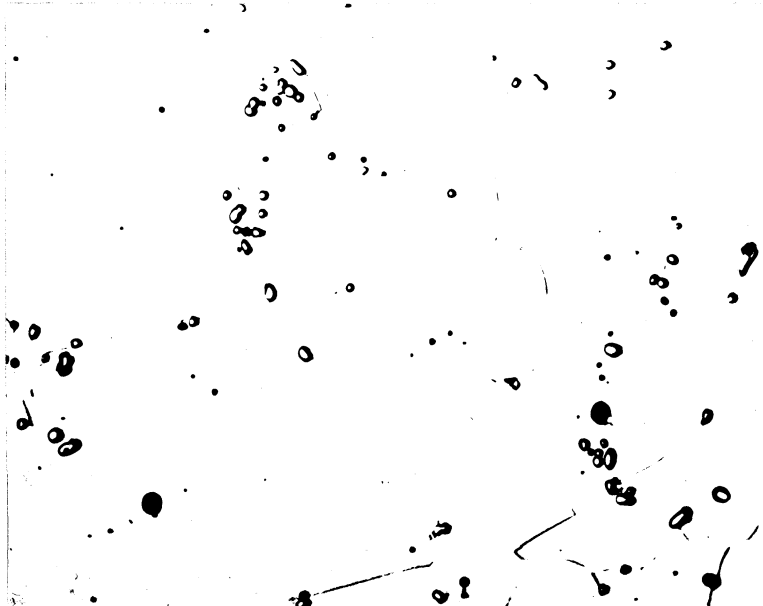


48-353

40% reduction

Electrolytically etched in 10% chromic acid

PLATE II-1.- EFFECT OF ROLLING AT 80° F ON MICROSTRUCTURE OF LOW-CARBON N155 ALLOY, SOLUTION TREATED 1 HOUR AT 2200° F, AND WATER QUENCHED.



48-354

Reduced 15% at 80° F



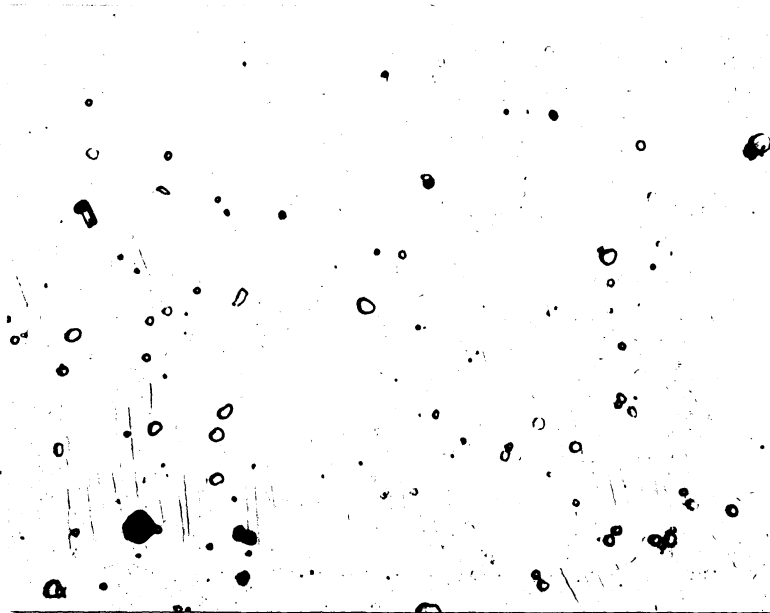
48-355

Reduced 11% at 1400° F

X1000

Electrolytically etched in 10% chromic acid

PLATE II-2.- EFFECT OF ROLLING TEMPERATURE ON MICROSTRUCTURE OF LOW-CARBON NI55 ALLOY SOLUTION TREATED 1 HOUR AT 2200° F AND WATER QUENCHED.



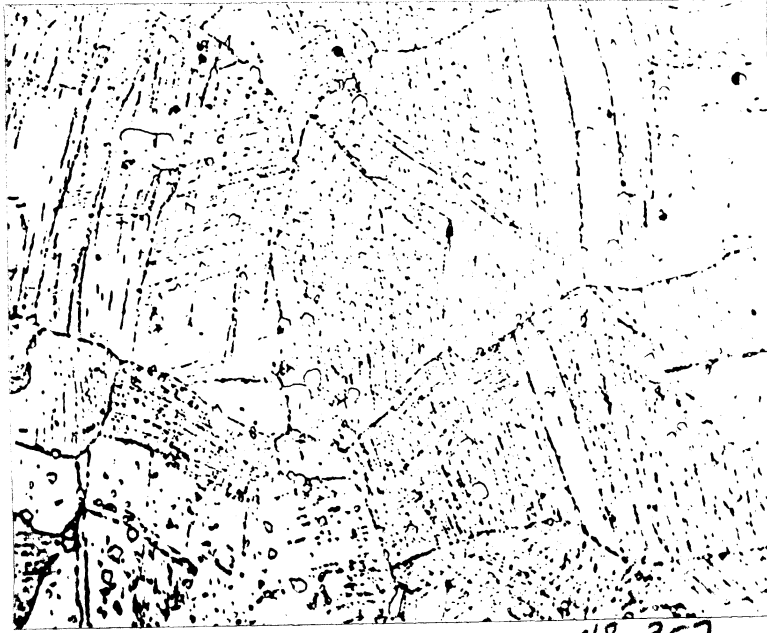
Unaged

48-353



Aged 1 hour 1600° F
X1000

48-356



Aged 100 hours, 1600° F

48-357

Electrolytically etched in 10 percent chromic acid

PLATE II-3.- EFFECT OF AGING AT 1600° F ON MICROSTRUCTURE OF LOW-CARBON NI55 ALLOY, REDUCED 40% AT 80° F AFTER SOLUTION TREATMENT 1 HOUR AT 2200° F AND WATER QUENCHING.



48-213

0% reduction
(10 hour solution treatment)



48-358

15% reduction at 80° F



48-359

16% reduction at 1400° F

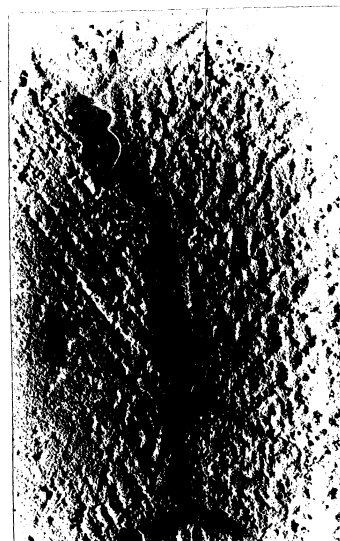
Aged 100 hours at 1600° F - X1000

Electrolytically etched in 10% chromic acid

PLATE II-1.- EFFECT OF HOT-COLD ROLLING ON MODE OF PRECIPITATION AT 1600° F OF LOW-CARBON N155 ALLOY, SOLUTION TREATED AT 2200° F, AND WATER QUENCHED.



Optical photomicrograph
X1000
Electrolytically etched
in 10% chromic acid



Electron photomicrograph
X3000
Collodion replica
Shadow cast with chromium

Original sample
Solution treated 10 hours 2200° F, water quenched, aged 1000 hours at 1200° F

PLATE II-5.- COMPARATIVE PHOTOMICROGRAPHS TAKEN WITH OPTICAL AND ELECTRON
MICROSCOPES.

III - EFFECT OF CHEMICAL COMPOSITION ON PROPERTIES

This investigation is in progress to evaluate the effect of the various alloying elements normally present in Low-Carbon N155 alloy on the properties of the alloy at high temperatures. The work has shown that it is possible, through sufficient control, to minimize the effects of processing variables so that the effect of chemical composition on properties should be evident.

1. Procedure for Production of Test Alloys

Five new heats of varying carbon content and five of varying manganese content have been prepared for testing. Actual chemical analyses were made for the variable elements and for a sufficient number of the fixed elements to insure that compositions were correct.

Further forging work indicated that the higher carbon heats could be forged only with great difficulty to square bar stock between flat dies. The forces present in forging along the diagonal planes of the bars, which resulted in fine grained areas in the low-carbon heats, resulted in bursting on these planes in the high-carbon heats. Forging between a flat top die and a V bottom die also did not prove satisfactory. Forging with swaging dies gave the best results.

As discussed in the March progress report, it appeared that the effects of prior treatment could best be eliminated by solution treating at 2200° F for 1 hour, water quenching and then subsequently aging at 1400° F for 24 hours. This treatment has been carried out on the bar stock from five standard control heats previously prepared and on the stock from the new heats.

2. Results

The chemical compositions, both aim and, where made, actual analyses, of the ten new heats are given in table III-1. Complete analysis, except nitrogen, were run on only two heats. Carbons were run on all heats and manganese on the five heats where this element was varied. Carbon content variations of 0.075, 0.40, and 0.60 percent are included with check duplicate heats at the two higher carbon levels. The other five heats provide manganese variations of 0.03, 0.30, 0.50, 1.0, and 2.6 percent.

Macrostructures of the ten ingots and pouring temperatures are shown in figure III-1. The ingot of the low-manganese heat EN18 had a blow hole through the complete length of the center which rendered the heat useless. One more attempt to produce a low manganese heat is planned. All the other ingots appeared to be sound.

The effect on hardness of a 2200° F solution treatment and 1400° F age for 24 hours on five standard heats and eight modified heats is given in table III-2. This treatment decreased the average Brinell hardness range for the standard heats from 224 to 270 as forged to 210 to 230 as solution treated and aged. Hardness increased with carbon content from 207 Brinell with 0.07 carbon to 291 Brinell with .57 carbon in the solution treated and aged condition. Manganese modifications from 0.30 to 2.58 percent had no apparent effect on hardness. The hardness of all the alloys was decreased by solution treating followed by an increase with subsequent aging.

Rupture test results obtained to date are given in table III-3. Mainly these results show the scatter in strength from five heats made to nearly the same composition. The stress-rupture time curves for the two treatments used are shown in figure III-2 along with the curves obtained for

TABLE III-2

SURFACE BRINELL HARDNESS OF THIRTEEN EXPERIMENTAL HEATS OF LOW-CARBON N155
AND MODIFICATIONS

Note: Hardness for square bar stock except where round bar stock is indicated.

Heat number	Alloy modification	Condition		
		As forged	2200° F 1 hr W.Q.	2200° F 1 hr W.Q. + 1400° F 24 hr
EN7	Standard	224 (2) ^a 257 ^b (3)	190 (2) ^a 208 ^b (1)	210 (6) ^a 220 ^b (1)
EN8	Standard	230 (3)	187 (2)	216 (6)
EN10	Standard	236 (3)	196 (2)	230 (6)
EN11	Standard	270 (6)	-----	224 (4)
EN12	Standard	240 (3)	193 (2)	218 (6)
EN13	0.07C	234 (3)	183 (1)	207 (1)
EN14	.36C	286 (2)	233 (1)	272 (1)
EN15	.40C	276 (3)	238 (1)	277 (1)
EN16	.57C	294 (3) 299 ^b (3)	258 (1) 254 ^b (1)	291 (1) 285 ^b (1)
EN19	.30Mn	259 (3)	203 (1)	221 (1)
EN20	.50Mn	244 (3)	204 (1)	218 (1)
EN21	1.04Mn	241 (3)	191 (1)	218 (1)
EN22	2.58Mn	235 (3)	205 (1)	223 (1)

^aNumber of hardness tests averaged to obtain reported values given in parentheses.

^bRound bar stock. Rockwell "C" hardness converted to Brinell hardness.

a. Heats EN13, 14, 15, 16, 17



Approximate pouring temperature
(°F)

	<u>L & N</u> optical pyrometer	<u>Pt, Pt-Rh</u> immersion thermocouple
EN13	2550	---
EN14	---	---
EN15	---	2535
EN16	---	2585
EN17	---	2585

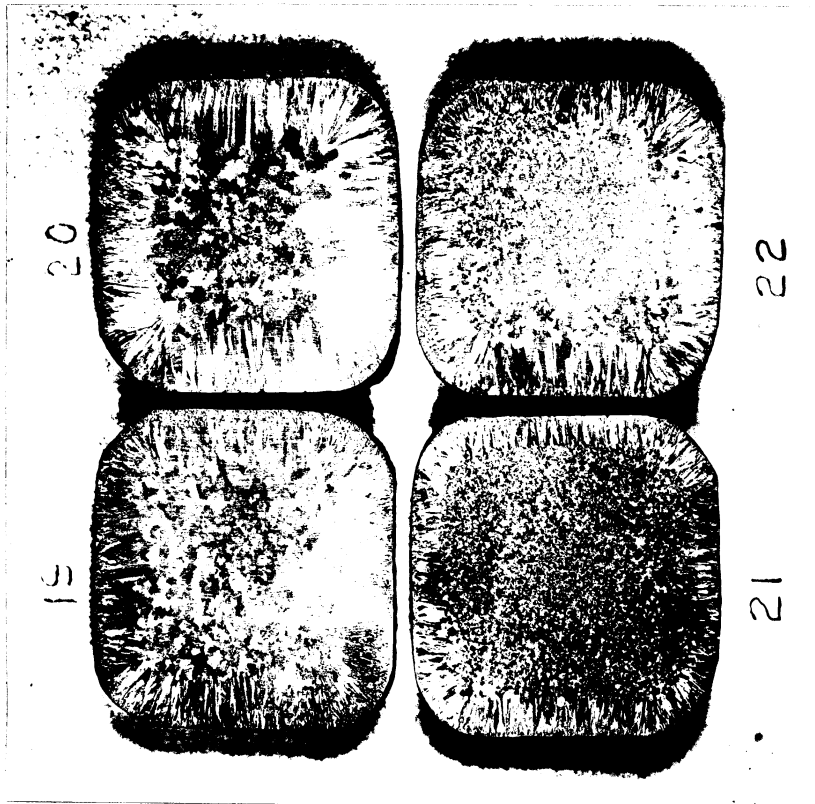
Etchant - Marble's reagent

FIGURE III-1.- MACROSTRUCTURES OF INGOT CROSS SECTIONS OF TEN EXPERIMENTAL HEATS.

b. Heats EN18, 19, 20, 21, 22

Approximate pouring temperature
(°F)

	L & N optical pyrometer	Pt, Pt-Rh immersion thermocouple
EN18	2550	2640
EN19	---	2587
EN20	2640	2557
EN21	---	2585
EN22	---	---



Etchant - Marble's reagent

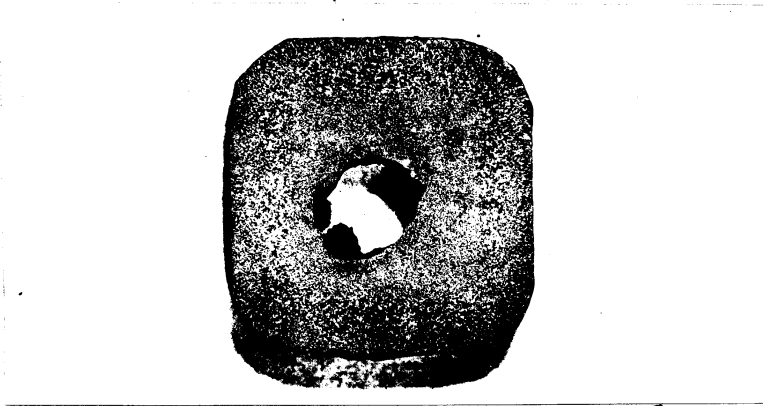
19

20

21

22

48-363



48-362

FIGURE III-1. (CONTINUED) - MACROSTRUCTURES OF INGOT CROSS SECTIONS OF TEN EXPERIMENTAL HEATS.

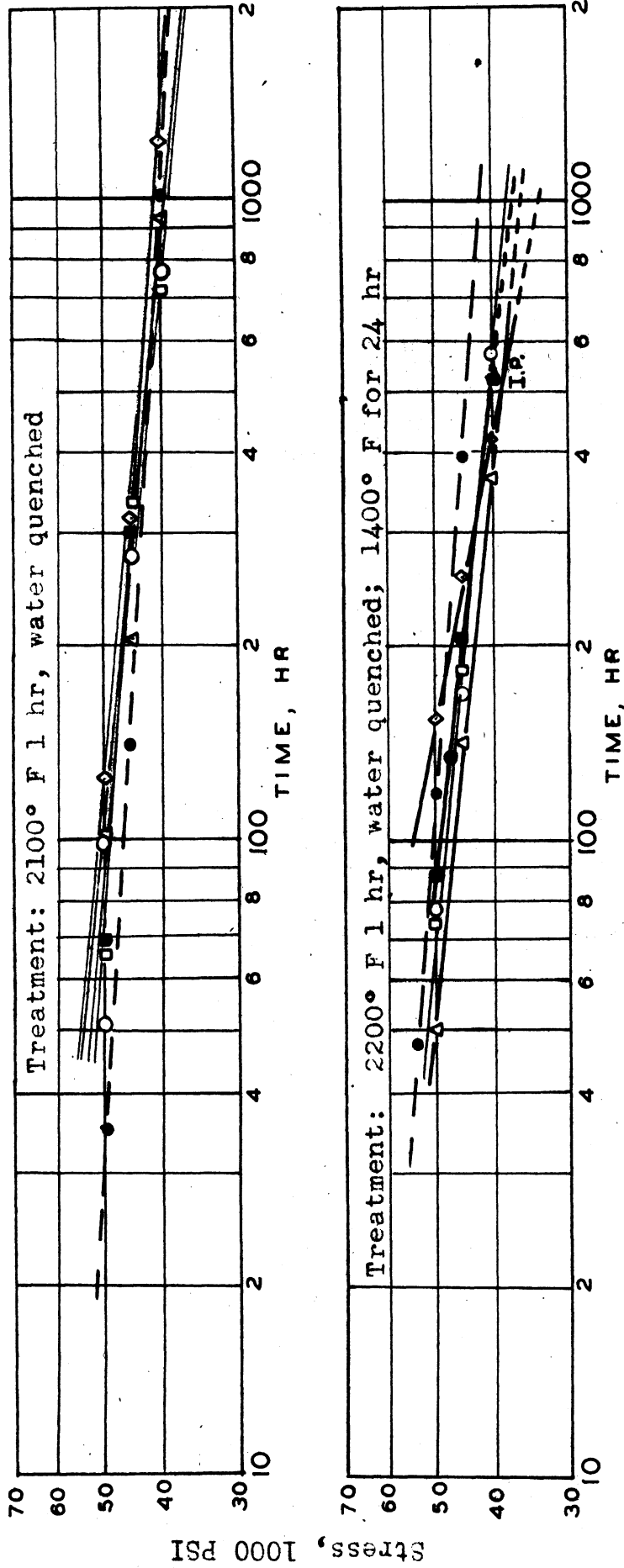


FIGURE III-2.- CURVES OF STRESS AGAINST RUPTURE TIME AT 1200° F FOR FIVE EXPERIMENTAL HEATS OF LOW-CARBON NI55 ALLOY.

Lot 20376 of Low-Carbon N155 alloy in the same conditions. These results indicate that the 2100° F solution treatment gives better uniformity of results from heat to heat than the 2200° F solution and 1400° F aging treatment. However, results on the latter are incomplete.

The limited results obtained on the carbon and manganese modified heats permit no conclusions.

IV - FURTHER WORK ON HEAT-TREATING AND PROCESSING PROCEDURES

1. Rupture Properties

Rupture testing has been continued to establish the effect of test temperature and prior treatment on the rupture properties of the new lot of Low-Carbon N155 alloy (Heat A-1726). The rupture properties established to date at 1200°, 1350°, and 1500° F for five treatments are given in table IV-1 and compared with the same type of data previously obtained from the original lot of Low-Carbon N155 alloy bar stock used in this program (Lot 30276). These data indicate:

(a) The rupture strength of the as-rolled stock, Heat A-1726, is considerably higher with lower elongation than that of the previously tested stock, particularly for 1000 hours at 1350° and 1500° F.

(b) When solution treated at 2200° F and aged at 1400° F the two stocks have similar rupture strength with the new heat having lower elongation.

(c) Stock solution treated at 2100° F does not have quite the high rupture strength of Lot 30276 at 1200° F but compares favorably at 1350° and 1500° F with higher rupture strengths at longer time periods. The new heat has lower elongation at 1200° and 1350° F but higher at 1500° F than the old heat.

(d) The new stock in the as-rolled plus 15 percent reduction by rolling at 1200° F condition has lower short time strength at 1200° F but surpasses the old heat in strength at longer time periods. At 1350° F the new heat has considerably greater strength, especially over long periods of time. Its elongation is much lower at both test temperatures.

TABLE IV-1

COMPARATIVE RUPTURE PROPERTIES OF TWO HEATS OF LOW-CARBON NI55 ALLOY

Treatment	Test temperature (°F)	Rupture-test properties					
		Heat A-1726a			Lot 30276b		
		Strength (psi)		Elongation % in 1 in.	Strength (psi)		Elongation % in 1 in.
		100 hour	1000 hour	100 hour	100 hour	1000 hour	1000 hour
As rolled	1200	48,000	43,000	5	49,500	37,500	17
	1350	34,000	29,000	20	32,000	18,500	42
	1500	15,500	11,500	25	13,500	7,800	40
2200° F 1 hr, W.Q.; 24 hr at 1400° F	1200	47,000	42,000	10	50,000	42,000	14
	1350	32,000	25,500	25	30,500	24,000	47
	1500	21,000	14,500	35	21,000	14,000	50
2100° F 1-hr, W.Q.	1200	44,000	38,500	4	46,500	40,000	7
	1350	30,500	24,500	15	31,000	22,000	35
	1500	18,000	14,500	58	17,500	12,500	40
As rolled; 15% reduction at 1200° F	1200	59,000	54,000	1	63,000	48,000	6
	1350	37,500	27,000	6	35,000	18,000	18
2050° F 2 hr, W.Q.; 15% reduction at 1200° F	1200	(55,000) ^c	(48,000) ^c	(3)	62,000	53,500	1
	1350	(40,000) ^c	(32,000) ^c	—	38,000	28,500	12
	1500	24,000	17,000	14	22,000	12,500	16

^aAll test specimens taken from center bar from the ingot.

^bSee data given in section I-4 of preceding progress report and the report "A Study of the Effects of Heat Treatment and Hot-Cold Work on the Properties of Low-Carbon NI55 Alloy."

^cBased on incomplete tests.

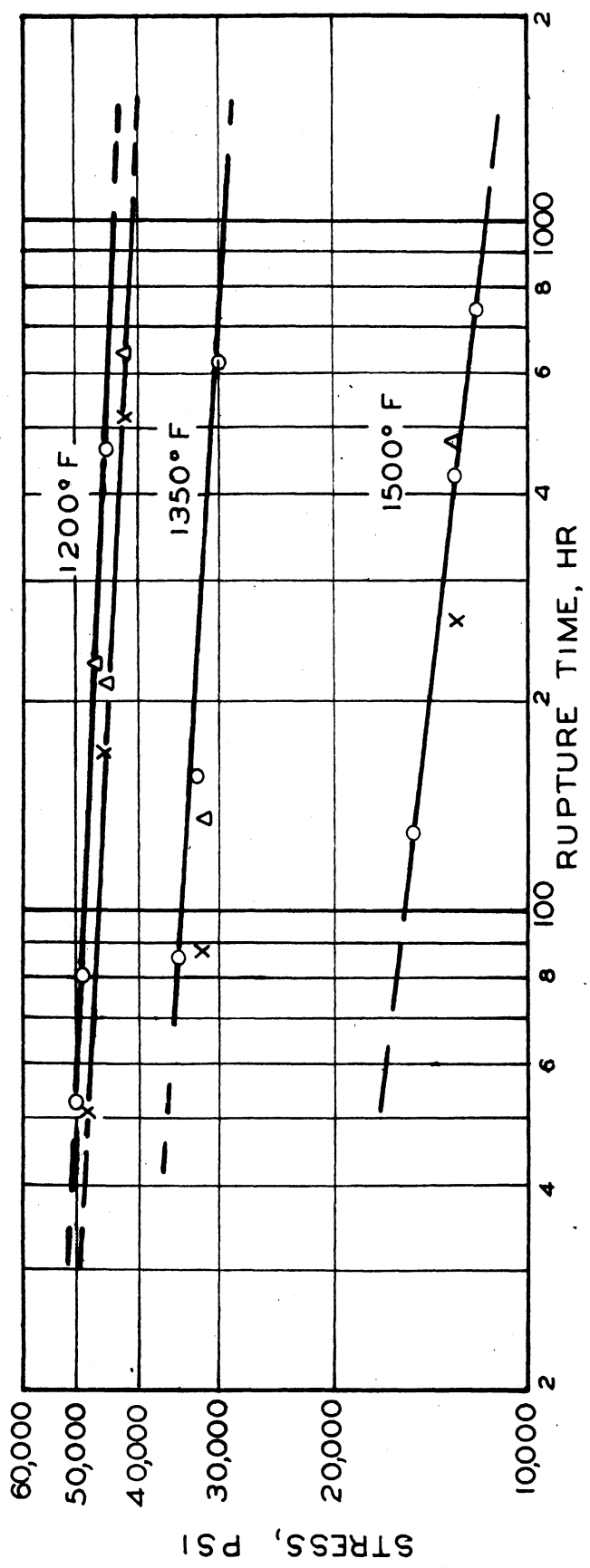
(e) Heat A-1726 solution treated at 2050° F plus 15 percent reduction at 1200° F will evidently have a lower rupture strength at 1200° F with greater ductility than the previously tested stock. At 1350° F the two heats will probably compare well, but at 1500° F the new heat has greater strength with similar elongation.

Thus far no attempt has been made to explain the differences in strength between the two heats of Low-Carbon N155 alloy. It is evident, however, that when solution treated at the high temperatures of 2200° F, differences in properties between the two heats are small and that differences for other conditions could be due to the influence of prior hot work.

Testing is being extended to include lower deformation rates (creep tests) and to extend the testing temperatures to as high as 1800° F.

2. Uniformity of As-Rolled Stock from New Heat

As-rolled stock is being rupture tested at 1200°, 1350°, and 1500° F as one means of determining the uniformity of the new heat. Tests are being made on specimens taken from bars representing the bottom, center, and top of the ingot from which the bar stock was rolled. The results to date, figure IV-1, indicate a slight but consistent difference between bar stock from extreme positions in the ingot.



Δ - Bottom } Position of test stock in original ingot.
 ○ - Center }
 x - Top }

FIGURE IV-1.-STRESS-RUPTURE TIME CURVES SHOWING UNIFORMITY OF AS ROLLED LOW-CARBON N155 ALLOY AT 1200° F, 1350° F, and 1500° F.

V - REPRODUCIBILITY OF DATA IN RUPTURE TESTS

Tests have been in progress for the determination of the effect of holding time at a test temperature of 1200° F and the reproducibility of test data. As previously reported, duplicate tests made under the same conditions check very well as shown in figure V-1. At short time periods for rupture, increasing holding time at test temperature before applying the stress appreciably increased the rupture strength. The difference in strength, however, decreased with testing time so that at 1000 hours very little difference exists.

The test results on the effect of holding time at 1200° F are not as yet understood. It has been concluded, however, that it has had very little influence on prior test work where comparisons have been based on rupture strengths at 100 and 1000 hours. At shorter rupture time, however, an aging effect, similar to that shown in section II, apparently influences rupture strength.

VI - COOPERATIVE FATIGUE TEST PROGRAM

The Low-Carbon N155 alloy (Heat A-1726) to be used in this program was described in the previous progress report. The progress to date is:

1. The necessary bar stock has been cut to specimen length, solution treated at 2200° F for 1 hour and water quenched, and aged sixteen hours at 1400° F.
2. Hardness tests are being taken over the entire lot of heat treated specimens to establish uniformity.

3. Tensile and rupture test specimens have been machined and testing is in progress to procure tensile test data at room temperature, 1000° F, 1200° F, 1350° F, and 1500° F and to establish stress-rupture time and time-deformation curves at 1200°, 1350°, and 1500° F.

4. A machining and polishing procedure has been completed for fatigue test specimens.

5. Half of the fatigue specimens requested by the Battelle Memorial Institute are ready for shipment and the remainder are in the polishing stage.

6. The Westinghouse Company fatigue specimens have all been rough machined and two finished sample specimens submitted for their inspection.

7. Syracuse University fatigue specimens are ready for machining.

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