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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

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September 24, 1948

NATIONAL ADVISORY COMMITTEE FOR AERONAUTICS

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I - CONTINUATION OF PRIOR WORK

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

II - FUNDAMENTAL STUDIES

INTRODUCTION

Studies are in progress to establish the fundamental processes by which treatment and composition control properties of commercial alloys at high temperatures. Low-Carbon N155 and Inconel-X alloys are being used as indicative of two types of alloys of major interest. Progress has been reported three time previously (see references 1, 2, and 3). The work has been separated into two sections: studies of structures resulting from solution treatment and aging and studies of structures resulting from rolling at various temperatures. Brief descriptions of experimental techniques used, results, and interpretation of the data obtained to date 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.

A. Studies of Solution Treated and Aged Low-Carbon N155

1. Metallographic Examination of Solution Treated and Aged Low-Carbon N155

As an initial investigation into the character of aging in Low-Carbon N155 alloy after solution treatment, it was decided to carry out a metallographic examination of a series of samples aged for time periods ranging from 1/2 to 1000 hours at 1200°, 1400°, and 1600° F after being solution treated 10 hours at 2200° F. Both optical and electronic examination was used. In the case of the electron micrographs, the formvar replica technique as outlined by Williams (see reference 4) was used.

The following results were obtained:

a. Aging at 1200° F resulted in little but the progressive development of a distinct etching resistant grain boundary constituent. At aging periods up to 10 hours, the boundary constituent was incomplete in that it did not surround all the individual grains. At aging periods of 1000 hours, the boundary constituents widened out to an approximately micron-wide band. At 1000 hours slight precipitation was observable in the matrix near the grain boundaries. At 10,000 diameters, this precipitate did not appear to be a distinct phase with an interface but was surrounded by a concentration gradient as revealed by a sloping surface from the center of the precipitate particles to the matrix proper as a result of etching. This would indicate that strains would exist around each such particle due to the probable difference in equilibrium lattice spacing between matrix and precipitate.

b. Aging at 1400° F resulted first in the development of a grain boundary constituent. At 10 hours the boundary band was approximately

one micron wide and changed little in character with further aging. A concentration gradient was present at the boundary, as revealed by a sloping surface towards the grain boundary after etching, throughout aging periods up to 1000 hours. In addition to the grain boundary reaction, general matrix precipitation appeared after aging about 10 hours and increased rapidly in amount up to the longest aging period used, 1000 hours. By 1000 hours, some growth of the individual precipitate particles was noticeable. Average size at this time was approximately 0.1 x 0.4 micron. Here no appreciable concentration gradient appeared around the precipitate particles but rather a definite interface.

c. Aging at 1600° F resulted in almost the same type of reactions as at 1400° F with the exception that the precipitate at the end of 1000 hours age was in platelets averaging 1.5 x 0.2 microns in the plane of polish and spaced an average of 2 microns apart. Definite interfaces were around each precipitate particle.

d. The question of the effect of time at the solution temperature on the character of the subsequent aging is of obvious commercial and theoretical importance. A 10-hour solution time at 2200° F was initially used in the above studies to insure random distribution of the atoms of the precipitating elements in the matrix at the solution temperature and to eliminate effects from prior rolling deformation. It was therefore decided to carry out studies on material solution treated one hour at 2200° F to see if any differences in the character of aging could be detected metallographically. For time periods up to 100 hours little or no difference in the microstructures could be detected between material solution

treated one hour and 10 hours at 2200° F. After 1000 hours at 1200° F the material solution treated 10 hours had the greatest grain boundary constituent widths and less precipitate in the matrix adjacent the boundary.

2. X-ray Studies of Solution Treated and Aged Low-Carbon N155

To obtain additional insight into the character of the matrix lattice of Low-Carbon N155 alloy during aging after solution treatment, it was decided to carry out three types of X-ray investigations: measurement of matrix diffraction line peak intensities, matrix lattice parameter, and matrix diffraction line widths, all as a function of time at the aging temperature. The same heat treated conditions were used for the X-ray studies as for the metallographic examinations.

a. Line intensity studies: Figure II-1 shows the results of line intensity studies on the 10-hour solution treated material when aged at 1200°, 1400°, and 1600° F. Figure II-2 compares the intensity changes, upon aging at 1200° F, of Low-Carbon N155 alloy solution treated 1 and 10 hours at 2200° F. Peak intensities were measured directly from plots obtained with a Norelco Spectrometer and automatic recorder.

As mentioned in reference 2, the drops in peak intensity on the curves of figure II-1 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 nuclei formation process at 1600° F when compared to the same process at 1400° F (compare position of intensity minima on the two curves for these aging temperatures of figure II-1) may be due to 1600° F being close to the temperature at which precipitants dissolve, if one accepts Becker's treatment of the phenomenon. Becker (see reference 5) has rather thoroughly explored the

rates of precipitation for binary alloys near the border between the single and two phase regions and found that the rate of nuclei formation is greatly slowed down due to difficulty of such nucleus formation. Diffusional processes would 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.

Inspection of the curve for material aged at 1200° F on figure II-1 leads to the conclusion that two separate nucleation processes are in progress at 1200° F. In conjunction with the metallographic observations covered above, it could be concluded that the two nucleation processes at 1200° F are the grain boundary reaction and then the general matrix precipitation reaction. The boundary reaction nucleation period could be so speeded up at 1400° and 1600° F as to not appear in the intensity measurements. The significance that the nucleation period of aging plays in regard to physical properties will be discussed later.

From figure II-2 it can be seen that a difference in aging behavior, at least in so far as peak line intensities are concerned, can be obtained by changing the solution treating time from 1 to 10 hours. It is quite clear that the first minimum on the 10-hour solution curve has been suppressed on the one hour solution curve. This could be due to the fact, reported above, that the grain boundary reaction did not occur as rapidly in the one hour solution-treated stock as in the 10-hour solution-treated material.

b. Matrix lattice parameter measurements: For additional information regarding precipitation reactions occurring during aging of Low-Carbon N155

alloy, a study of the matrix lattice parameters was carried out on the same solution treated and aged samples used for the intensity measurements. Figures II-3 and II-4 show the results. Back reflection technique was used with chemically precipitated silver powder as a standard.

The measurement of lattice parameter as a function of time in precipitating alloys can give direct evidence of whether the precipitate particles are growing by agglomeration or by matrix depletion. If the matrix parameter continues to change with increasing aging time, it is evident that the precipitation reaction as such is still in progress and that the precipitate particles are growing by virtue of the reaction and not by combination with smaller, less stable particles. If the latter were true, then the matrix parameter would have reached some steady state value independent of further aging time. From the curves of figure II-3 it is evident that in no case has the aging time been sufficient to complete the precipitation reaction. For this reason, the definite determination of whether the precipitate and matrix compositions are a function of temperature of aging and the character of such a function has not been determined. It appears, however, that the precipitates obtained by aging at 1400° and 1600° F could be slightly different in composition, since the curves for the matrix parameter are approaching steady state values which are probably not quite the same. Aging at 1200° F at the end of 1000 hours had resulted in so little change in lattice parameter that no conclusions can be drawn.

Further significance of the lattice parameter studies will be discussed later in connection with physical properties.

c. Line width measurements: It was early discovered in the series of fundamental investigations covered in this report that the increases in hardness

as a result of aging after solution treatment did not correspond in any way with the evidences of nuclei formation as revealed by X-ray diffraction line studies. Accordingly it was felt that some other change within the matrix lattice was responsible for the hardening. Since Dehlinger had postulated (see reference 8) that long period lattice distortion (of the type which could be associated with each of the small precipitated particles revealed by the metallographic examination of samples after prolonged aging) would result in line broadening, it was decided to measure the line broadening effects. Chromium radiation and (220) line of the matrix (occurring at $\theta = 65°$) was used for the width measurements in order to take advantage of the increased resolution in the back reflection region. A Moll microphotometer was used to measure the photographic recording.

Figure II-5 shows the results to date. It will be noted that the increase in line width occurs only after considerable aging and after line intensities had returned to steady state values. Thus a long period lattice distortion replaces a short period one when the aging time is increased. Further discussion of the significance of such line broadening will be discussed under "Conclusions."

3. Hardness Measurements on Solution-Treated and Aged Low-Carbon N155

Hardness measurements were also made on the solution treated and aged series of samples. The hardness survey was made to provide data which could be used to determine if the technique of hardness measurement was one which gives reliable indications as to resistance to creep at high temperatures. Figure II-6 shows the results obtained on the 10-hour solution-treated stock at 1200°, 1400°, and 1600° F aging temperatures, while figure II-7 shows comparative data between 1 hour and 10 hours solution-treated

stock aged at 1200° F. From figure II-6 it can be seen that conventional aging behavior is followed. The higher the aging temperature, the sooner the approach to a maximum, the maximum however increasing in value with decreases in aging temperature. It will further be noted that the increase in hardness upon aging at 1400° F appears to follow exactly the increase in diffraction line width shown in figure II-5; i.e., the increase in internal strain.

4. Creep Properties of Solution-Treated and Aged Low-Carbon N155

The purpose of the creep and rupture testing carried out on solution treated and aged Low-Carbon N155 alloy was to measure the mechanical behavior of samples used in the physical measurements discussed previously. To achieve this end, all heat treating or processing was carried out with samples for both physical measurement and mechanical testing so that both were subjected to exactly the same conditions.

Figure II-8 and II-9 show the results of creep testing. Several conditions of testing are worthy of note:

a. In an effort to obtain the creep properties corresponding as nearly as possible to the structures developed by solution treating and aging and not those controlled by structural changes during testing, the measurements have been confined to very short time periods. The initial creep rates and the rates as soon as the primary stage of creep had apparently ceased were measured. In the case of the tests at 30,000 psi, the secondary rate measurements were all taken between 20 and 40 hours after the start of testing. In the case of the tests at 60,000 psi, the secondary rate measurements were all taken 3 to 4 hours after the start of testing. In the latter case, the rates taken were the minimum creep rates

as tertiary creep in all cases appeared 1 to 2 hours later followed by eventual rupture. It is obvious then, that complete evaluation of decreasing secondary rates especially after long time periods, was not carried out. The results, it is believed, approximate closely the initial creep properties of the given material. In addition all tests were in the creep unit furnace not longer than 2 hours before the start of testing. Again this was to reduce the alteration of the initial known structure by time at the test temperature.

b. Second, the two stress loads used, it was felt, would bracket the range in which the creep properties are normally measured. The actual results showed that 30,000 psi was just above the proportional limit which was near 26,000 psi.

c. The immediate interpretation that can be given to the results shown in figures II-8 and II-9 that a definite minimum in the secondary creep rate exists when plotted against aging time for either of the two stress loads or aging temperatures used. Further, the optimum aging time at a given temperature is the same when considering either stress level. The higher aging temperature shifts the aging time for minimum creep to shorter aging times when considering the same stress level.

When considering the initial rates measured at 30,000 psi, it is apparent that the optimum age from the standpoint of highest initial creep resistance is the longest aging period for 1400° F and the 100 hour age at 1600° F. (See figure II-10.) Further, when considering creep rates after some time at test temperature (i.e., the secondary rates shown in Figure II-8) the optimum aging time apparently shifts to shorter aging times. This trend, however, has not yet been completely evaluated.

5. Rupture Testing of Solution Treated and Aged Low-Carbon N155 Alloy

Rupture testing was carried out in order to obtain additional information regarding the mechanical behavior of material identical with that used for the physical properties measurements. Again in order to minimize the alteration of the known initial structure, by time at the test temperature, the rupture specimens were in the test furnace a maximum of two hours before being loaded. In addition, only the very short time rupture strengths and deformation characteristics are considered for the same reason. Stated in other terms, what was desired were the mechanical charactersitics of the structures at the instant of the completion of the heat treatment.

Figures II-11 through II-12 show the results of the rupture testing to date. It is quite clear that when aging is carried out at 1400° F a definite maximum in rupture strength occurs with variation in aging time. For rupture times up to 10 hours, before alteration of the initial structure can occur by virtue of reactions at the test temperature, the maximum rupture strength results when the aging time is approximately 100 hours. With increasing time for rupture, the main alteration in the rupture time-aging time relationship appears to be the marked improvement of the unaged material. (See figure II-11). This is most probably due to reactions occurring in the material during testing and will be considered under a separate section later. Inspection of figure II-13 also shows that the aging period at 1400° F for maximum short-time rupture strength is associated with the highest values of deformation prior to fracture. This also held, in fact, for the initial deformation which occurred upon loading. Some alteration of these deformation characteristics is evident with increasing test time, since the curves of figure II-13 are in general sloping downward.

When considering aging at 1600° F a very broad maximum in the short time rupture strength occurs with aging time. A very slight maximum value could possibly exist at the one-half hour age. Again the progressive improvement of unaged stock is evident with increasing rupture time. Figure II-14 shows that maximum deformation values at rupture for material aged at 1600° F follow the same pattern as for material aged at 1400° F, with, however, shorter aging times in equivalent positions (compare one-half hour aged curve of figure II-13 with one hour aged curve of figure II-14).

6. Studies of Structure Alteration During Testing

The previous sections have shown it is possible to measure, at least in part, the characteristics, both physical and mechanical, of the structures induced in Low-Carbon N155 alloy by solution treating and aging. The alteration, however, of such initial conditions by the testing conditions must be evaluated in order to arrive at a proper insight into long time behavior of the alloy. In accordance with this philosophy, two more investigations are at present under way. First, in conjunction with the 1200° F aging of solution-treated Low-Carbon N155 under zero stress several additional series have been aged at 1200° F under varying stresses and are being studied with a view of evaluating the effect that stress has on the precipitation reaction. Second, samples, aged 1 and 100 hours at 1400° F are being or have been reaged for varying time periods at 1200° F and will be studied to find the nature of the precipitation process partially carried out at one temperature and continued at the testing temperature. Studies of this nature are contemplated also with the secondary aging to be carried out under stress fields of varying intensity.

The result of such studies outlined above will give significant information, it is felt, regarding the fundamental reasons for the observed long time behavior in regards to rupture ductility, creep strength, and rupture strength.

7. General Discussion Regarding the Solution Treating and Aging of Low-Carbon N155 Alloy

Several tentative conclusions can now be made in regard to the fundamental behavior of solution treated and aged Low-Carbon N155 alloy:

a. As far as time periods of interest in the application of austenitic high-temperature alloys to aircraft use are concerned, Low-Carbon N155 alloy can be considered extremely stable in aging characteristics. Despite the fact that aging periods up to 1000 hours were used, it was not possible at any of the temperatures of aging used to overage the alloy, in the common sense of the word; i.e., cause the hardness to drop markedly with increased aging. (See figure II-6.) In support of this contention, figure II-3 shows that in no case was the precipitation reaction complete within the aging periods considered by virtue of the fact that no steady state lattice parameter was reached. The effect of stress is still being considered.

b. Material aged to a high internal stress condition (long period lattice distortion) typified by high hardness, has optimum initial or primary creep resistance, but such a condition is extremely short lived. In a matter of a few hours, at stresses near the yield point, and almost instantaneously with high stress, when the secondary rate begins to appear, the maximum resistance to creep is obtained by material prior aged only enough to be in the nucleation period (lattice with short period lattice distortions, no line broadening and low relative hardness). This can be seen

by comparing figures II-1, II-3, II-5, and II-6 with figures II-8 and II-9.

Since the shift in aging time, at either 1400° or 1600° F, for maximum creep resistance is to shorter times with increasing time of test, there is probably no one optimum age for optimum creep resistance. Rather the optimum aging conditions are dependent upon the time of testing considered. However, for all practical times of service, optimum creep resistance is obtained when the precipitate has only been nucleated prior to testing and then is allowed to continue to grow during testing, provided such testing is at a high enough temperature to cause aging to proceed with a finite speed. The intriguing question of the character of aging during testing of unaged material will, it is hoped, be answered by the investigations, covered in sub-section 6. It is clear, however, that the short-time creep resistance of unaged material is not as good as that obtained by nucleation prior to service.

The question of which is the better temperature of aging, 1400° or 1600° F, from a creep standpoint, considering either primary or secondary rates, has little practical significance, since approximately the same minimum rate can be established at either temperature. At 1600° F the optimum aging time for the conditions used to obtain the data in figures II-8 or II-9 is shorter (1 hour) than for aging at 1400° F (10 hours). This is due to the faster growth of the particles at 1600° F by diffusion and hence the passage out of the purely nucleation stage (see sub-section 2).

c. The aging conditions prior to service which give optimum short time rupture strength are in distinct contrast to those for optimum creep resistance. In general, the maximum short-time rupture resistance for material aged at 1400° and 1600° F was obtained by the longer or longest aging times, for which the creep resistance was poorest. Since rupture

strength is dependent upon two characteristics, the creep resistance and the amount of deformation possible before fracture, it is quite clear that the superior rupture strength of the aged materials with poor creep strength is due to the greater possible deformation before fracture. The important conclusion, then, is that Low-Carbon N155 solution treated and aged to high hardness (and high, long period, internal strain as typified by line broadening) is in a comparatively ductile condition under the slow strain rates of rupture tests. This is shown in figures II-13 and II-14. That high internal strain and high relative rupture ductility should go hand in hand is quite surprising. However, apparently the fracture stress is so high above the flow stress, by virtue of the internal strain due to aging, that large amounts of deformation must take place, and hence reduction of cross sectional area, before the true stress reaches the fracture stress. Another equally valid explanation would be that the high internally strained condition resists crack propogation and hence allows the flow deformation to proceed to relatively large values before failure occurs.

d. There seems to be little significant difference in the aging characteristics at 1200° F of Low-Carbon N155 solution treated 1 or 10 hours at 2200° F. In fact, the only difference noted was in the suppression of the first minimum in the diffraction line intensity curves after the one hour solution treatment. This can be related to the fact that general matrix precipitation appeared somewhat sooner and the boundary reaction was not so prominent in this material and hence matrix nucleation would not be differentiated from the boundary reaction so clearly.

B. Studies of Cold, Hot-Cold, and Hot-Rolled Low-Carbon N155 Alloy

The very pronounced improvement, by hot-cold working, of the creep and rupture properties of Low-Carbon N155 alloy has been known for some time (see reference 2). The basic reasons for this improvement, however, have not been determined. Hence, the primary objective of studies covered in this section has been to ascertain the fundamental effects of working the metal on the structures of Low-Carbon N155.

Two working hypotheses are being used in the investigation at present:

a. That improved mechanical properties at high temperatures are due to the mode of precipitation induced by the working.

b. That the improved properties are due to the general lattice distortion induced by working below the temperature region of simultaneous recrystallization.

To date the work under section "a" above has been largely metallographic, both optical and electronic, in nature. The results of these examinations are as follows:

a. Cross sectional reductions of solution-treated Low-Carbon N155 up to 15 per cent by rolling at room temperature result in little noticeable change in microstructure. Above 15 per cent, slip lines or bands begin to appear, the greater the number the greater the reduction. Reductions of the order of 40 per cent result in almost complete coverage of the plane of polish with the slip bands. No outlining of grain boundaries is evident at any degree of cold reduction. The principal difference metallographically in increasing the rolling temperature to 1400° F is that outlining of the grains is evident after the reduction.

b. Annealing the above rolled structures results in the almost immediate appearance of precipitate streaks or plates along the former slip lines. No evidence of the former distorted structure of the matrix could be found after even short (1 hour) anneals (i.e., the slip lines were gone or obscured by the precipitate lamella). Continued or long time annealing produces no pronounced change in this short-time annealed precipitate structure. Precipitation thus appears to be microstructurally complete and the precipitate in a stable form after short time anneals. The accelerating effect of the cold working on such reactions as precipitation is, of course, well known and is quite evident from these observations.

Work under hypothesis "a" is continuing, especially along the lines of broadening the metallographic study, to indicate the effect of time at 1200° F on the precipitate structure produced and to include more rolling variables.

Work under hypothesis "b" has been in accordance with the theory that if lattice distortion were controlled, then progressive removal of such distortion would result in progress removal of the known improvements in high temperature properties. This theory was in great measure dictated by the fact that some previous work on metals has been published on the physical changes induced by annealing after cold rolling. All such changes result from removing the lattice distortion induced by the cold work in the first place. (See reference 6 for a fuller discussion of such investigations.) Of the several known lattice properties which change on annealing after rolling, one, diffraction line broadness, was chosen for study. Line broadening as mentioned under sub-section II-A-2 is caused by the presence of long-period lattice distortion, and it was felt that such a characteristic might be a direct measure of the effectiveness of cold or warm reduction on high temperature strength.

Figures II-15 and II-16 show how the broad diffraction lines resulting from cold or warm working are narrowed by annealing. For these data, chromium radiation was used with the photographic recording of the (220) line of the matrix. From the films, relative line widths were measured by use of a microphotometer. It is quite clear that the process of line sharpening is a highly temperature-dependent rate process. Further, the character of line sharpening is dependent upon the exact conditions of rolling. While the amount of cross sectional reduction is the same for both figure II-15 and figure II-16, the temperature of rolling is different as noted. The principal effect, then, of varying the temperature of rolling is seen in the relative stability of the distorted structure at 1200° F. From the two figures the tentative rule can be drawn that to obtain a relatively stable distorted lattice structure at 1200° the metal must be worked above 1200° F. In addition it will be noted that on figure II-16 an apparent maximum exists in line broadness when annealing is carried out at 1200° F. Previous published work (see reference 7) has found such maxima to be most probably associated with precipitation. Here then it is possible that a connection between hypothesis "a" and hypothesis "b" exists. Further work, particularly metallographic, will be done in this connection. Lastly, it will also be noted from figures II-15 and II-16 that the relative stability (resistance to line sharpening) after rolling at room temperature and 1400° F is reversed at 1800° F over that at 1200° F.

Since, as mentioned above, the line sharpening process is a rate process, the Eyring rate reaction theory has been applied to the data shown in figures II-15 and II-16 and the so-called energy of activation of the process determined. Table II-1 summarizes the results. Some extrapolation of the existing data was necessary. The effect of rolling temperature is thus to reduce the activation energy for line sharpening with reduced rolling temperature. This conclusion is very tentative in nature however.

To obtain information on the importance of annealing on creep properties, samples were prepared from material rolled 15 per cent at 1400° F and also as annealed for 1, 10 and 100 hours at 1600° F. These samples were then creep tested at 1200° F with the results shown on figure II-17. The creep tests were all of 50 hours' duration with the rates reported being the secondary rates present between 40 and 50 hours after testing. This short time testing was used to minimize the effects of the test temperature on the known initial structure. It is quite clear that with the removal of the type of lattice distortion typified by line broadening, the creep strength drops. It is by no means proven, however, that some other process than line broadening is not also contributing to the deterioration of the creep properties. Further search, therefore, is in progress to clarify the changes which take place during annealing. In addition, work is being extended to include structures rolled in the simultaneous recrystallization range of temperature and also to include the amount of reduction as a variable.

C. Studies on Solution-Treated and Aged Inconel-X

To a large degree, the investigations being conducted on solution-treated and aged Low-Carbon N155 are to be repeated on Inconel-X. Inconel-X alloy has been introduced into the program because it has very strong age-hardening characteristics in comparison to Low-Carbon N155 alloy. The contrasting precipitation characteristics of the two alloys should result in a wide difference in the structural conditions controlling properties. Data from such differing alloys are needed to check the validity of the tentative theories of structural effects derived from the studies on Low-Carbon N155 alloy.

A part of one heat of Inconel-X in one-inch diameter bars, has been received in the as-rolled condition. As a first step, the solution characteristics of this hot-rolled bar stock were determined. The following results were obtained:

a. For one-hour solution treatments, the grain coarsening temperature for the bar stock was approximately 2050° F.

b. After one hour at any solution-treating temperature above the coarsening range, however, solution of former grain boundary constituents is not complete.

c. Studies of the lattice parameter of a series of solution-treated samples held for one hour at temperatures ranging from 1800° to 2200° F showed little change over the as-rolled bar stock. It is concluded that either the precipitating material was largely in solution in the as-rolled stock or lattice parameters cannot be used to measure the degree of solution.

From the above results, it has been decided that a solution-treating temperature of 2050° F will be used as the standard procedure. Studies of aging in this material are proceeding.

D. Identification of Micro Constituents in Low-Carbon N155 Alloy

Analysis techniques have been fairly well established for analysis of residues obtained by the preferential solution of matrix material over the excess constituents in Low-Carbon N155 alloy. With reasonably reliable procedures of analysis available, work is in progress to establish the efficiency of the separation procedures and to determine the composition of the constituents.

Very tentative results indicate that:

(1) Only about one-half the Columbium Avel carbon can be separated from samples solution treated at 2200° F as separate constituents. X-ray diffractions show that the constituent is CbC and/or CbN.

(2) Quite prolonged aging after solution treating is required before other compounds can be separated and identified by X-ray diffraction patterns. Compounds of the type $M_{23}C_6$ and M_6C have been at least tentatively identified in addition to CbC or CbN.

(3) Indications are that chromium increases in the residues with increasing aging time at 1400° F.

(4) Nearly all of the columbium appears in the residues even after short aging periods at 1400° F.

(5) Nearly all the carbon appears in the residues after short aging time periods at 1400° F.

(6) Completely satisfactory separations for Mo, W and N in the residues have not yet been made.

TABLE II-1

ACTIVATION ENERGY CALCULATIONS FOR DIFFRACTION LINE SHARPENING IN ROLLED LOW-CARBON N155

Rolling temperature (°F)	Rolling reduction (percent)	Annealing temperature (°K)	1/T°K	Time for $W/W_0 = 0.7$ (hr)	Activation energy*
80	15	921	0.001085	10,000	
		1030	.000970	380	
		1144	.000873	24	
		1255	.000797	.50	58,000 cal/mol
1400	15	921	.001085	———	
		1030	.000970	120	
		1144	.000873	5.5	
		1255	.000797	.25	73,000 cal/mol

*Calculated from the relation:

$$t_{0.7} = Ae^{\frac{-E}{RT}}$$

or

$$\ln t_{0.7} = \ln A - \frac{E}{RT}$$

when $t_{0.7}$ = time, hours, to reach W/W_0 of 0.7

T = temperature, °K

R = gas constant

Values given are for annealing in range of 1400° F since in general the $\frac{1}{T}$ vs lnt plots were slightly curved.

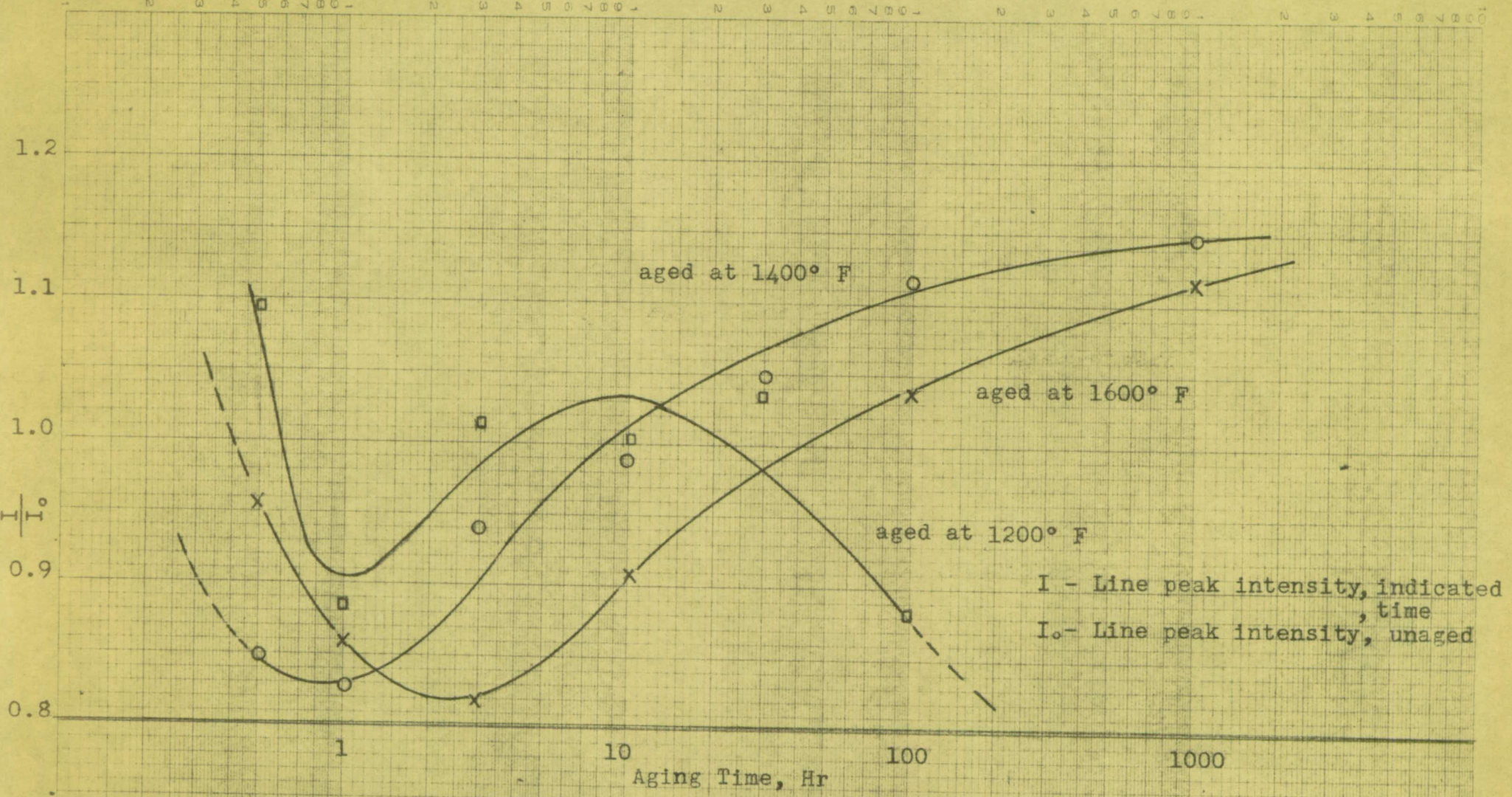

FIGURE II-1.- EFFECT OF AGING ON (111) LINE INTENSITY OF LOW-CARBON ALLOY, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

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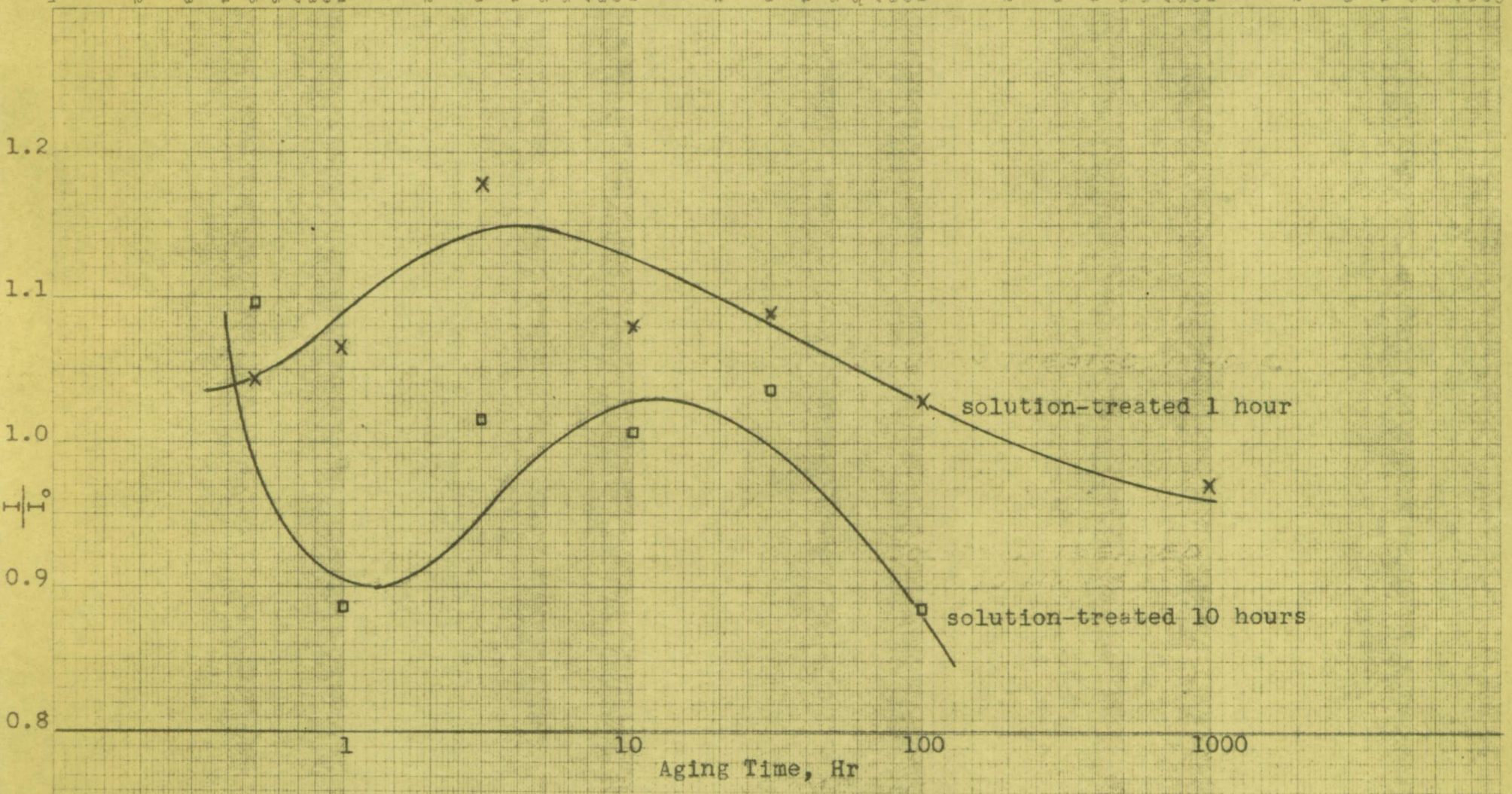

FIGURE II-2.- EFFECT OF AGING AT 1200° F ON (111) LINE INTENSITY OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED AT 2200° F AND WATER-QUENCHED.

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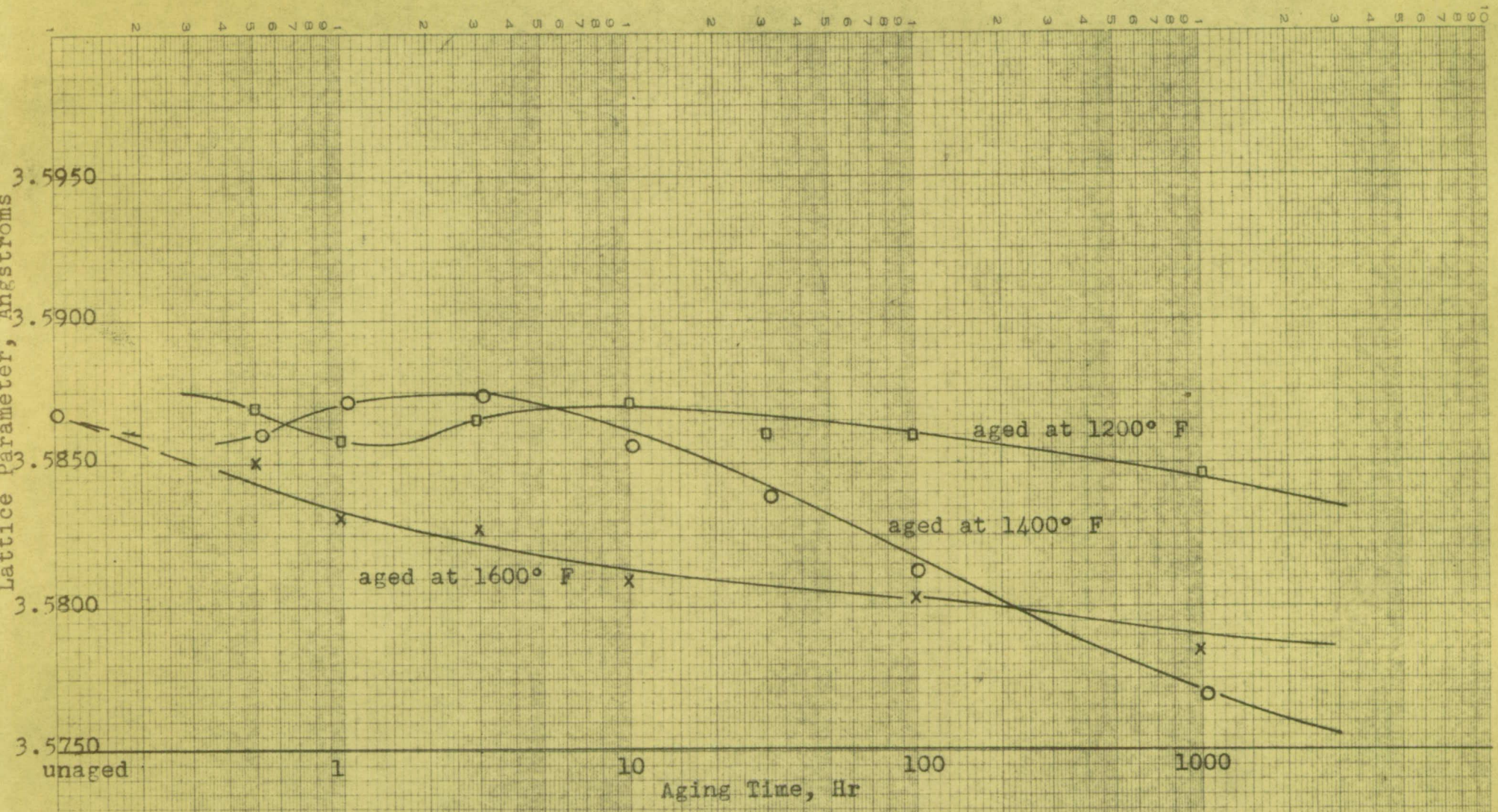

FIGURE II -3.- EFFECT OF AGING ON LATTICE PARAMETER OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

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5 CYCLES X 10 DIVISIONS PER INCH
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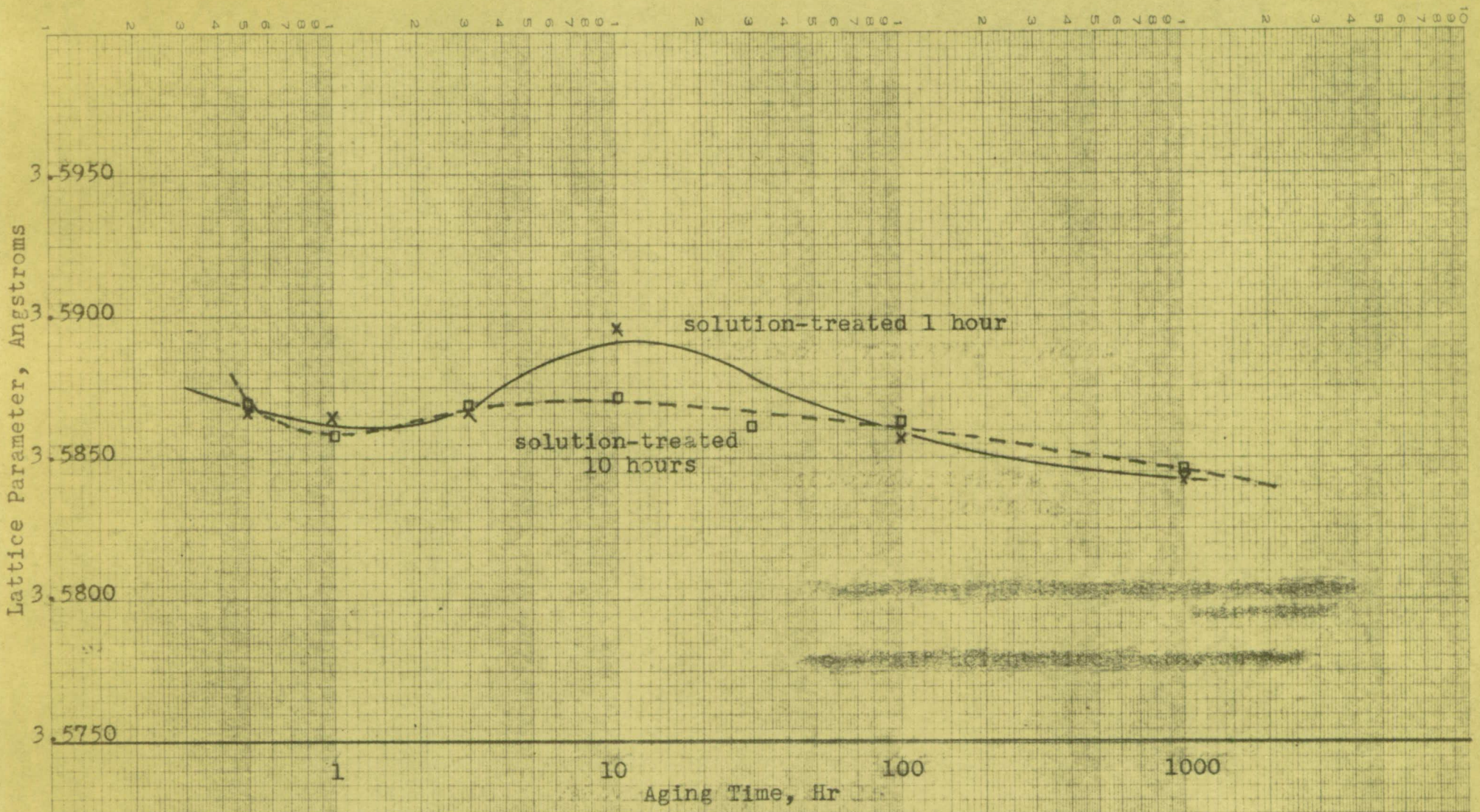

FIGURE II-4.- EFFECT OF AGING AT 1200° F ON LATTICE PARAMETER OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED AT 2200° F AND WATER-QUENCHED.

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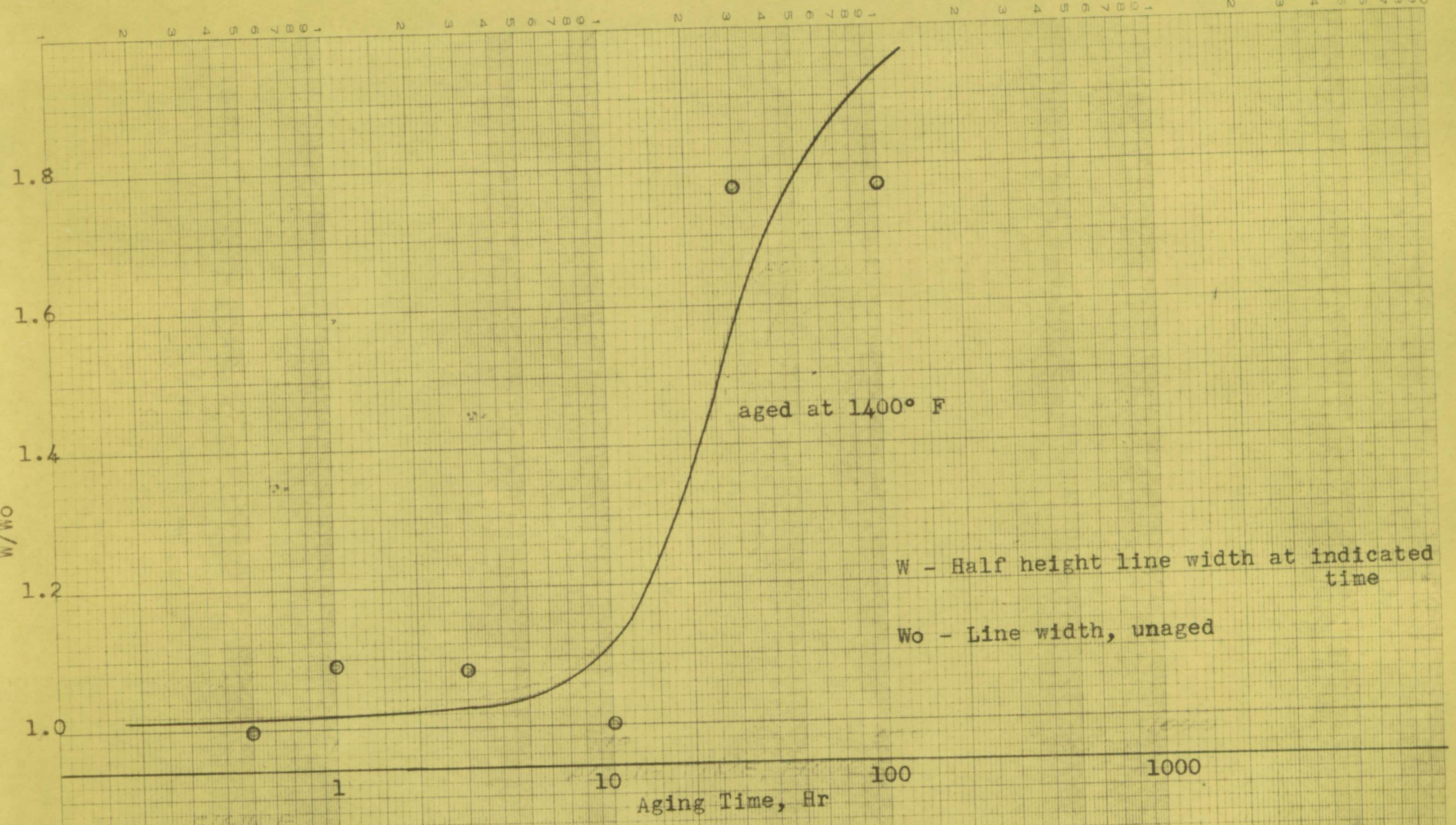

FIGURE II-5.- EFFECT OF AGING ON (220) LINE WIDTH OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

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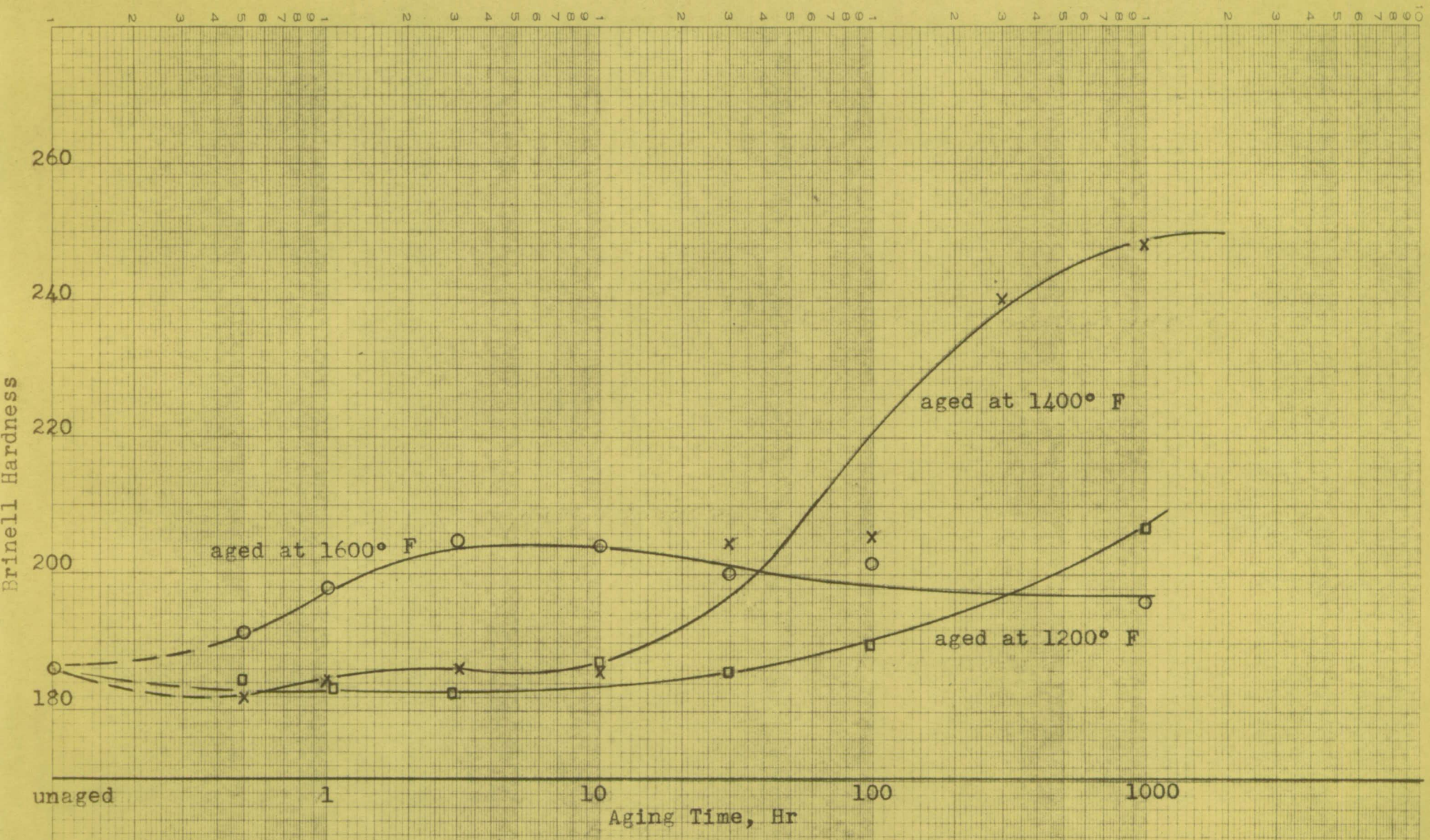

FIGURE II-6.- EFFECT OF AGING ON HARDNESS OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

NO. 340-1510 DIETZGEN GRAPH PAPER
SEMI-LOGARITHMIC
5 CYCLES X 10 DIVISIONS PER INCH
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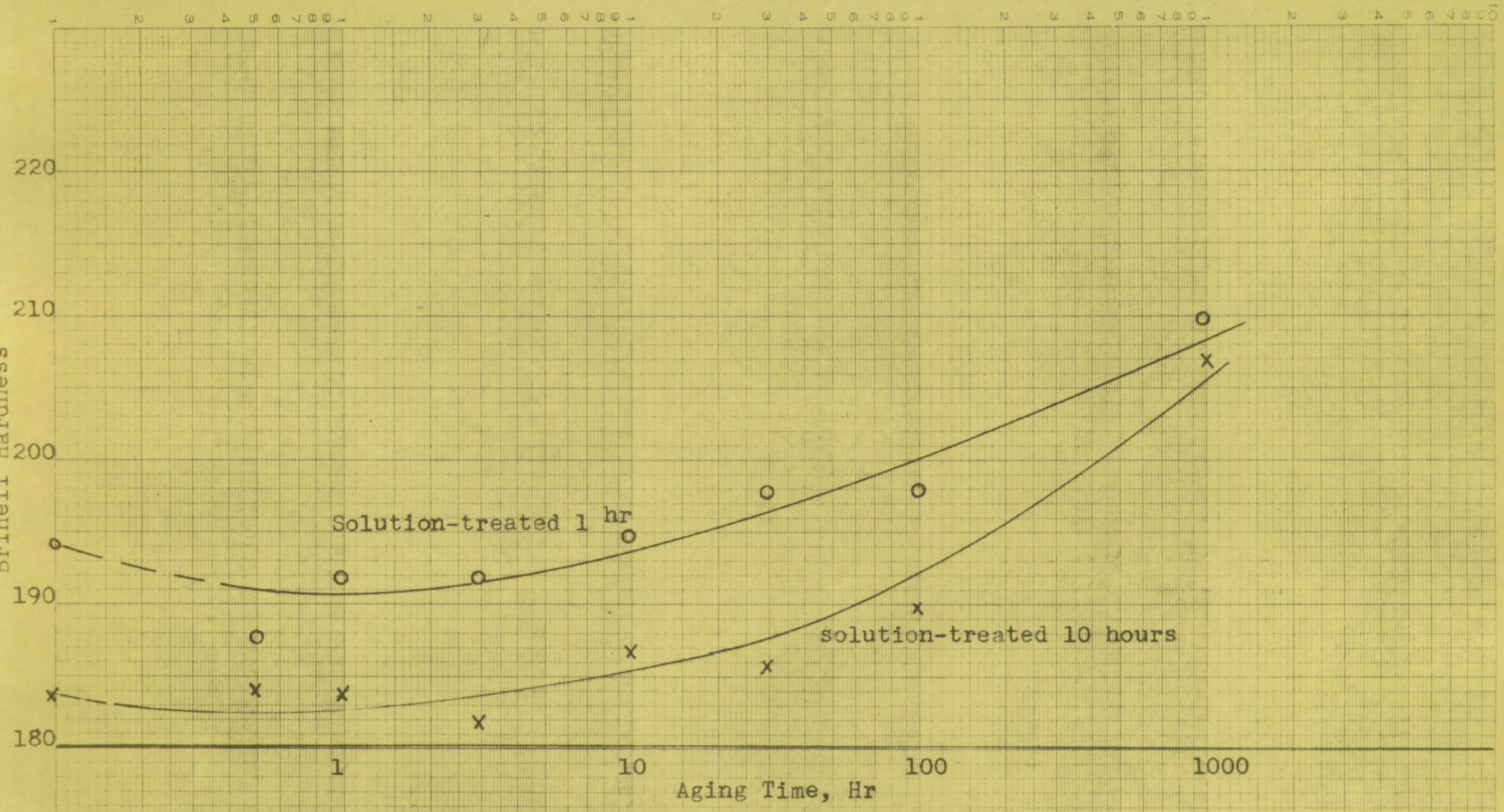

FIGURE II-7.- EFFECT OF AGING AT 1200° F ON HARDNESS OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED AT 2200° F AND WATER-QUENCHED.

KEUFFEL & ESSER CO., N. Y. NO. 359-125Q
Logarithmic, 5 X 3 Cycles.
V. C. C. Sheet
MADE IN U. S. A.

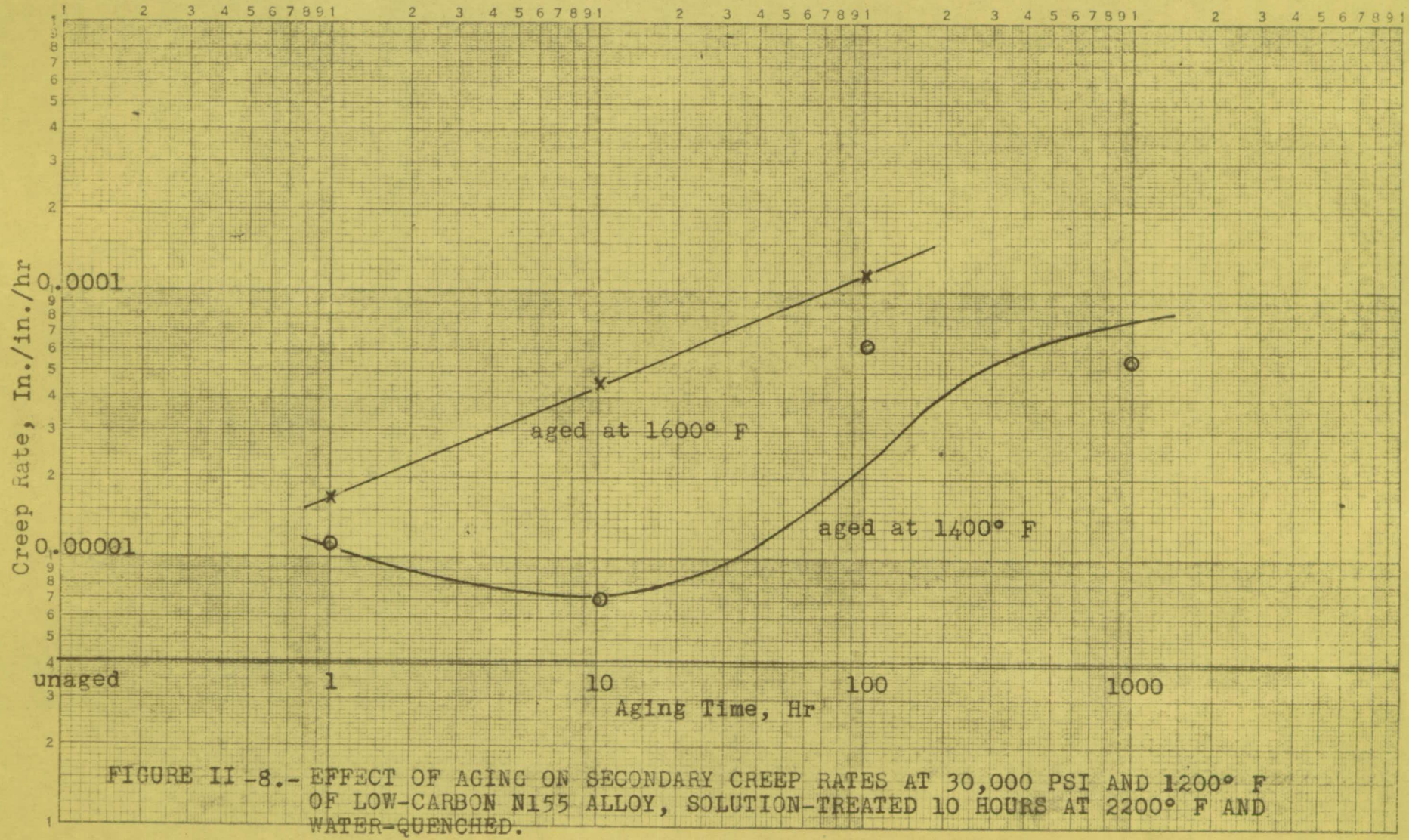

FIGURE II-8.- EFFECT OF AGING ON SECONDARY CREEP RATES AT 30,000 PSI AND 1200° F OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

KEUFFEL & ESSER CO., N. Y. NO. 359-125G
Logarithmic, 5 × 5 Cycles
P. C. C. Sheet
MADE IN U. S. A.

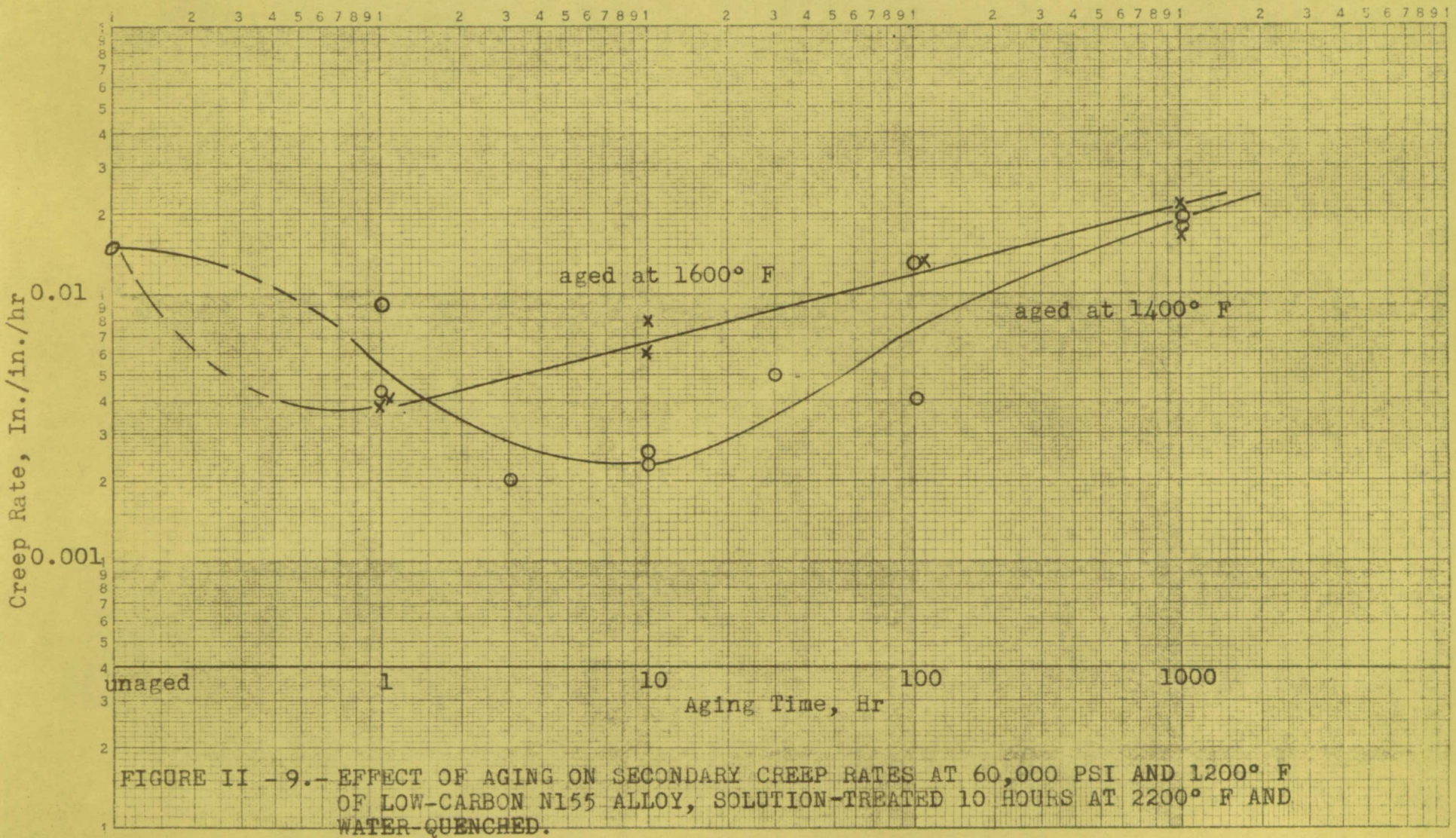

FIGURE II - 9.- EFFECT OF AGING ON SECONDARY CREEP RATES AT 60,000 PSI AND 1200° F OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

KEUFFEL & ESSER CO., N. Y. NO. 359-125G
Logarithmic, 5 X 3 Cycles.
P. C. C. Sheet.
MADE IN U.S.A.

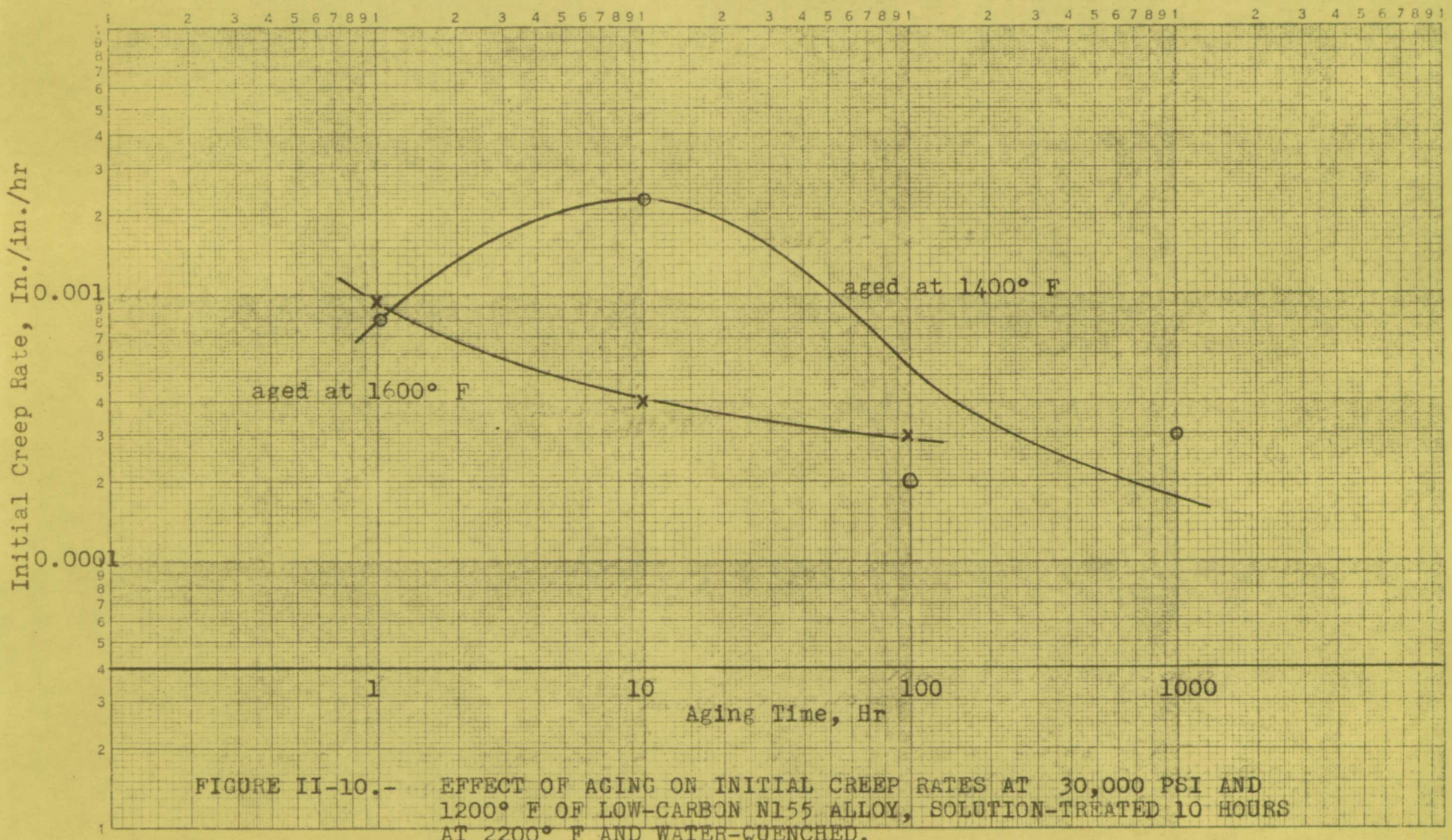

FIGURE II-10.- EFFECT OF AGING ON INITIAL CREEP RATES AT 30,000 PSI AND 1200° F OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

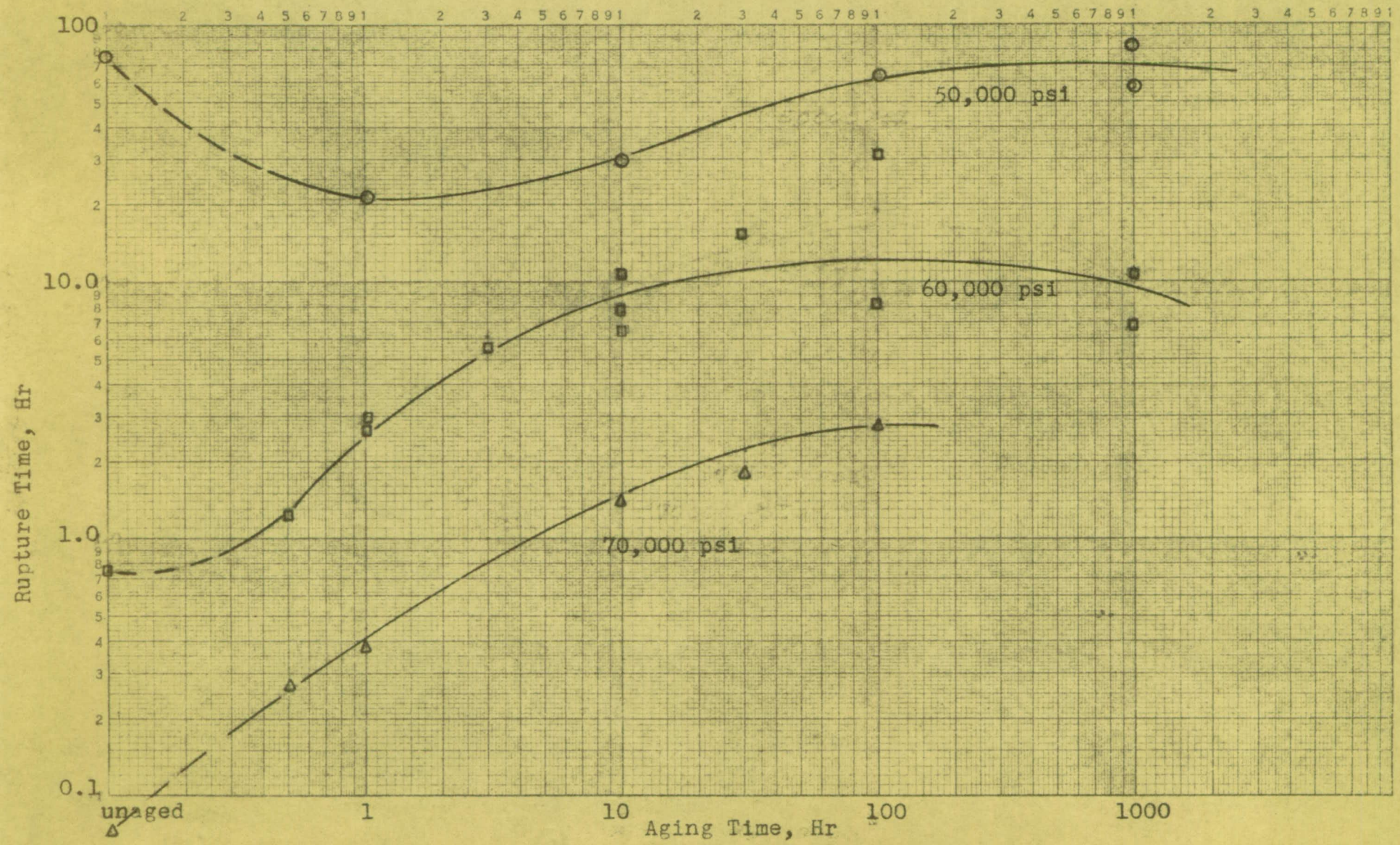

FIGURE II -11- EFFECT OF AGING AT 1400° F ON 1200° F RUPTURE STRENGTH OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

KEUFFEL & ESSER CO., N. Y. NO. 359-125G
Logarithmic, 5 × 3 Cycles.
F. C. C. Sheet.
MADE IN U.S.A.

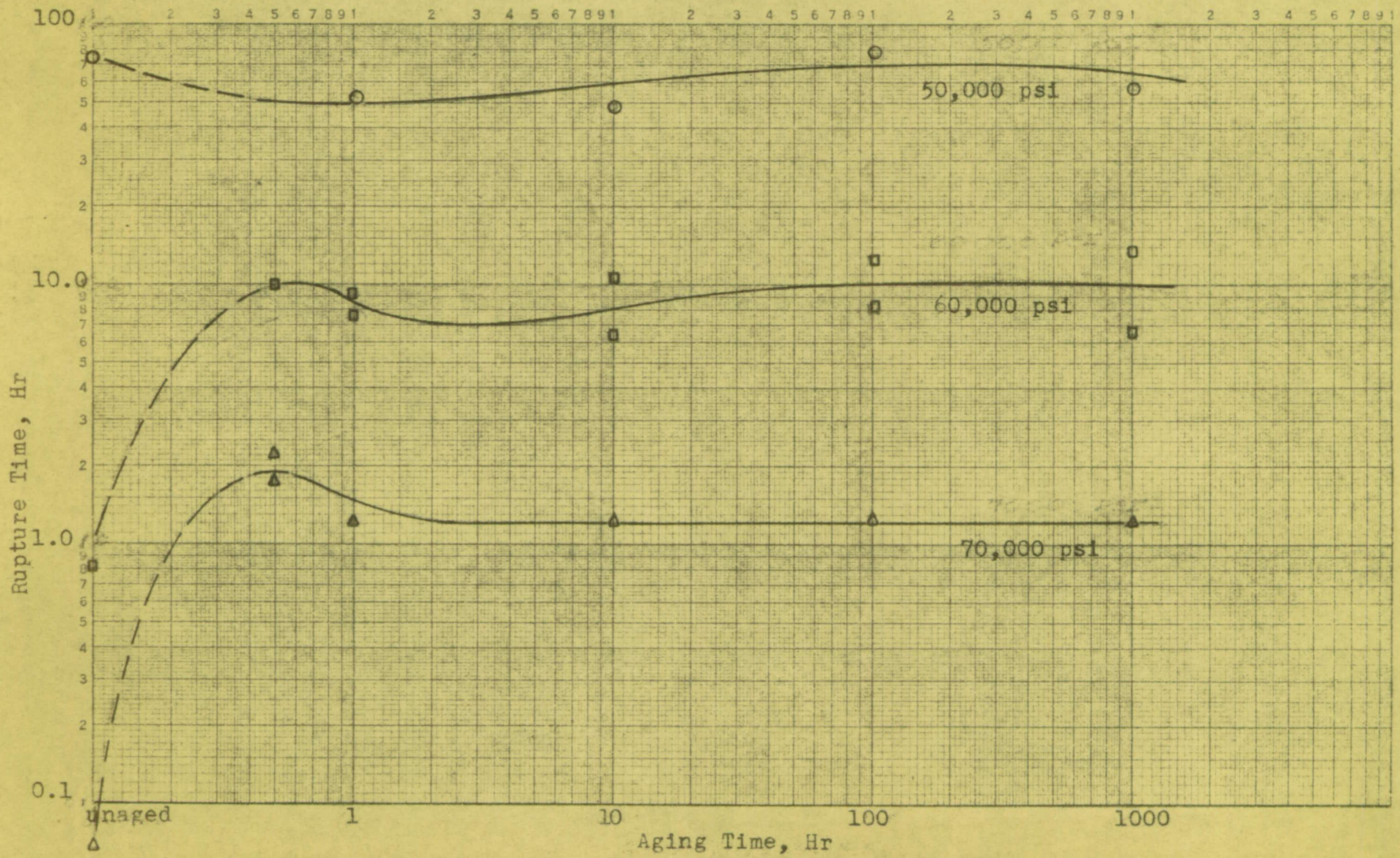

FIGURE II-12 .- EFFECT OF AGING AT 1600° F ON 1200° F RUPTURE STRENGTH OF LOW-CARBON N155 ALLOY, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

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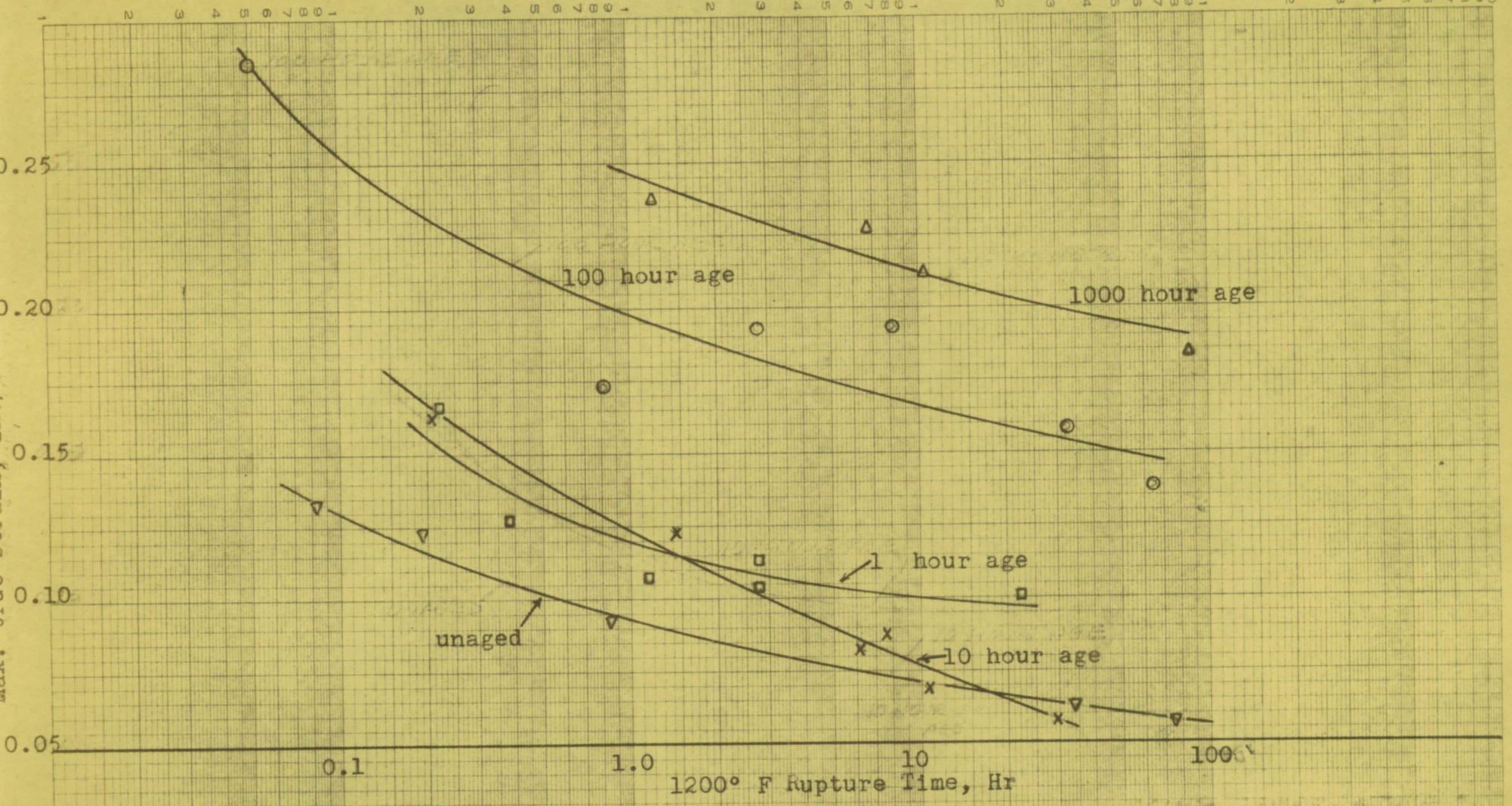

FIGURE II-13.-MAXIMUM TRUE STRAIN AT RUPTURE OF LOW-CARBON N155 ALLOY, AGED AT 1400° F, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

NO. 340-L510 DIETZGEN GRAPH PAPER
SEMI-LOGARITHMIC
5 CYCLES X 10 DIVISIONS PER INCH
EUGENE DIETZGEN CO.
MADE IN U.S.A.

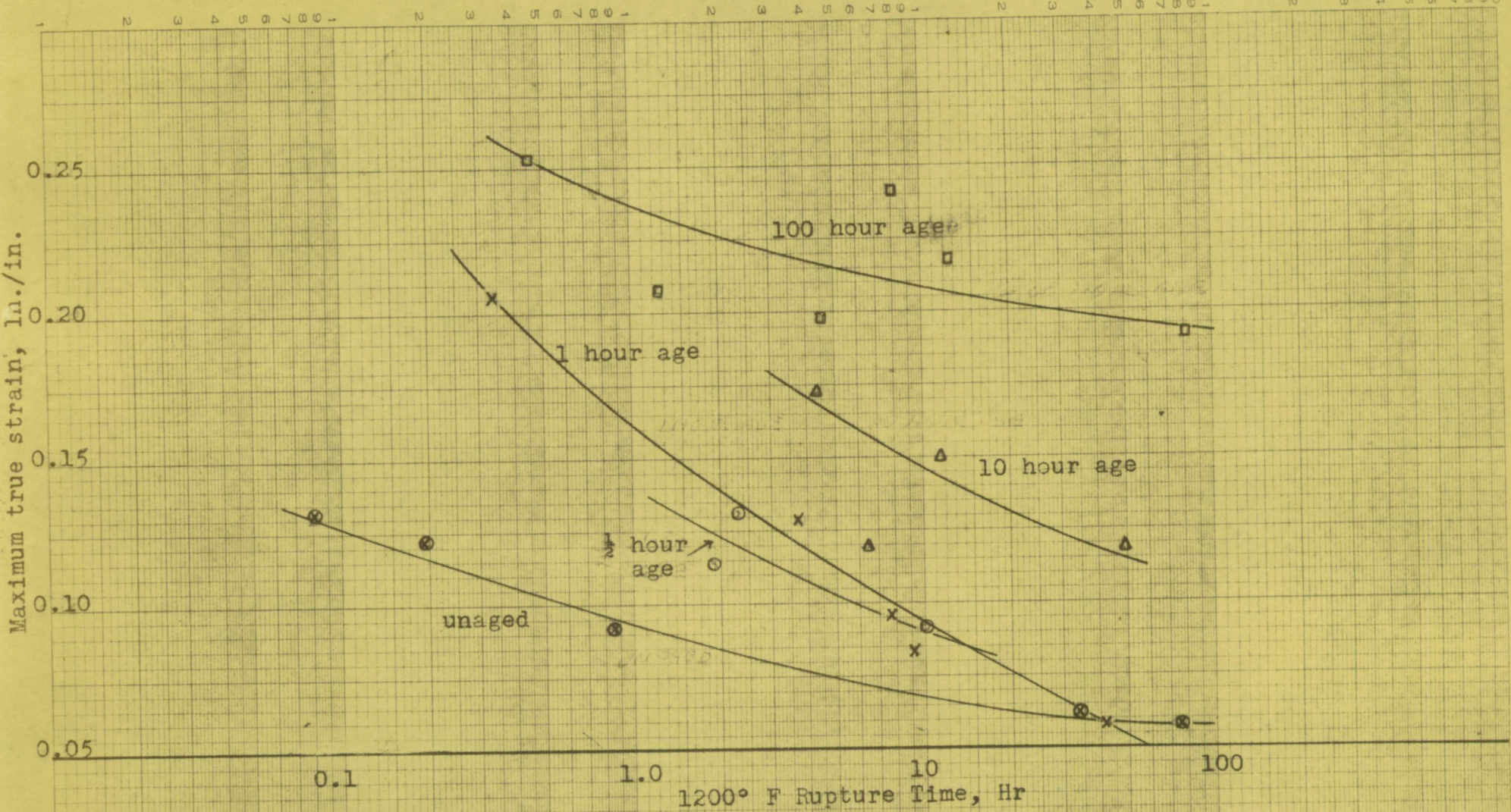

FIGURE II-14.- MAXIMUM TRUE STRAIN AT RUPTURE OF LOW-CARBON N155 ALLOY, AGED AT 1600° F, SOLUTION-TREATED 10 HOURS AT 2200° F AND WATER-QUENCHED.

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SEMI-LOGARITHMIC
5 CYCLES X 10 DIVISIONS PER INCH
EUGENE DIETZGEN CO.
MADE IN U.S.A.

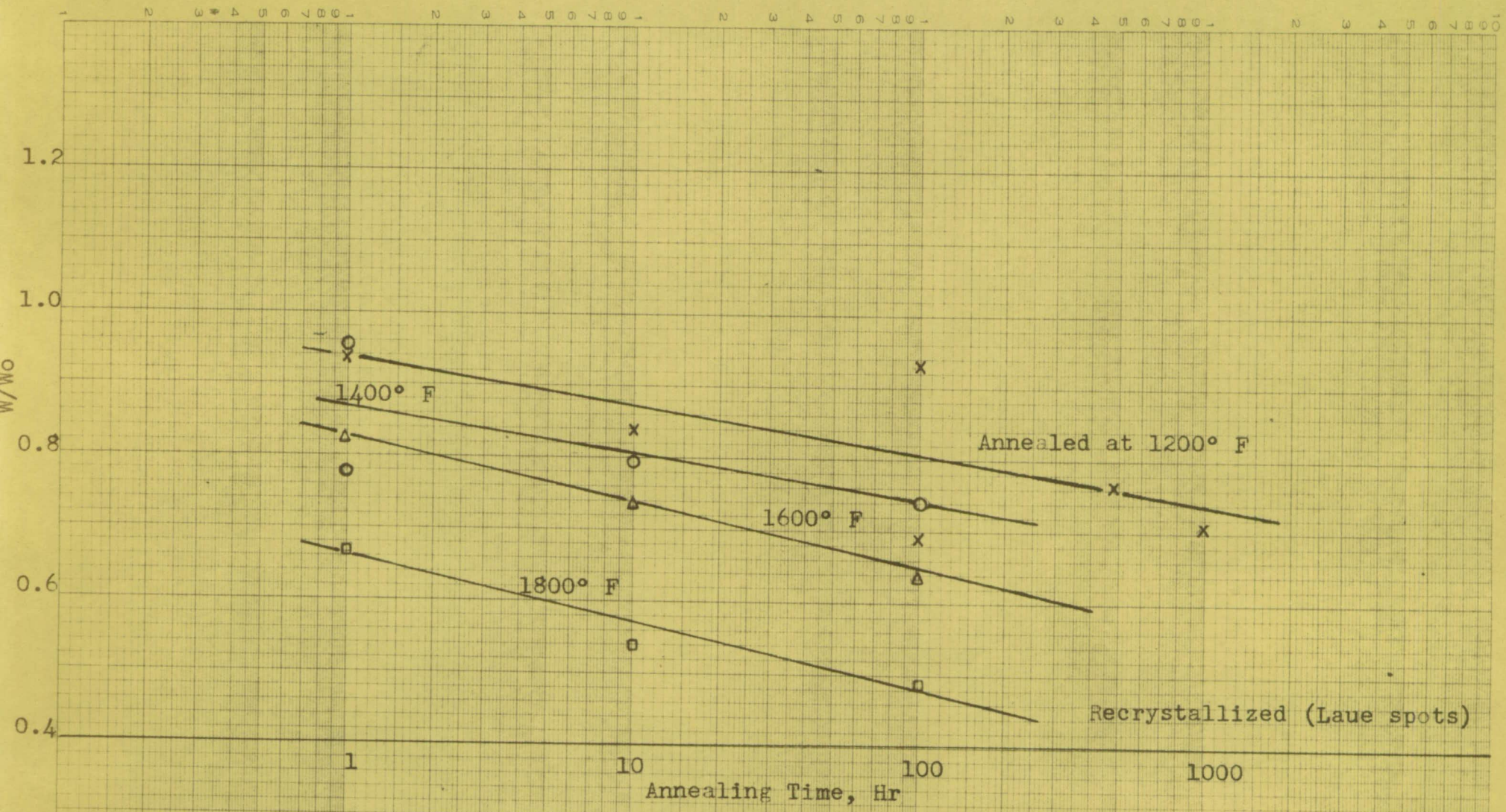


FIGURE II-15.- EFFECT OF ANNEALING ON (220) LINE WIDTH OF LOW-CARBON N155 ALLOY REDUCED IN CROSS SECTION 15% AT 80° F, AFTER BEING SOLUTION-TREATED 1 HOUR AT 2200° F AND WATER-QUENCHED.

NO. 340-L510 DIETZGEN GRAPH PAPER
SEMI-LOGARITHMIC
5 CYCLES X 10 DIVISIONS PER INCH
EUGENE DIETZGEN CO.
MADE IN U.S.A.

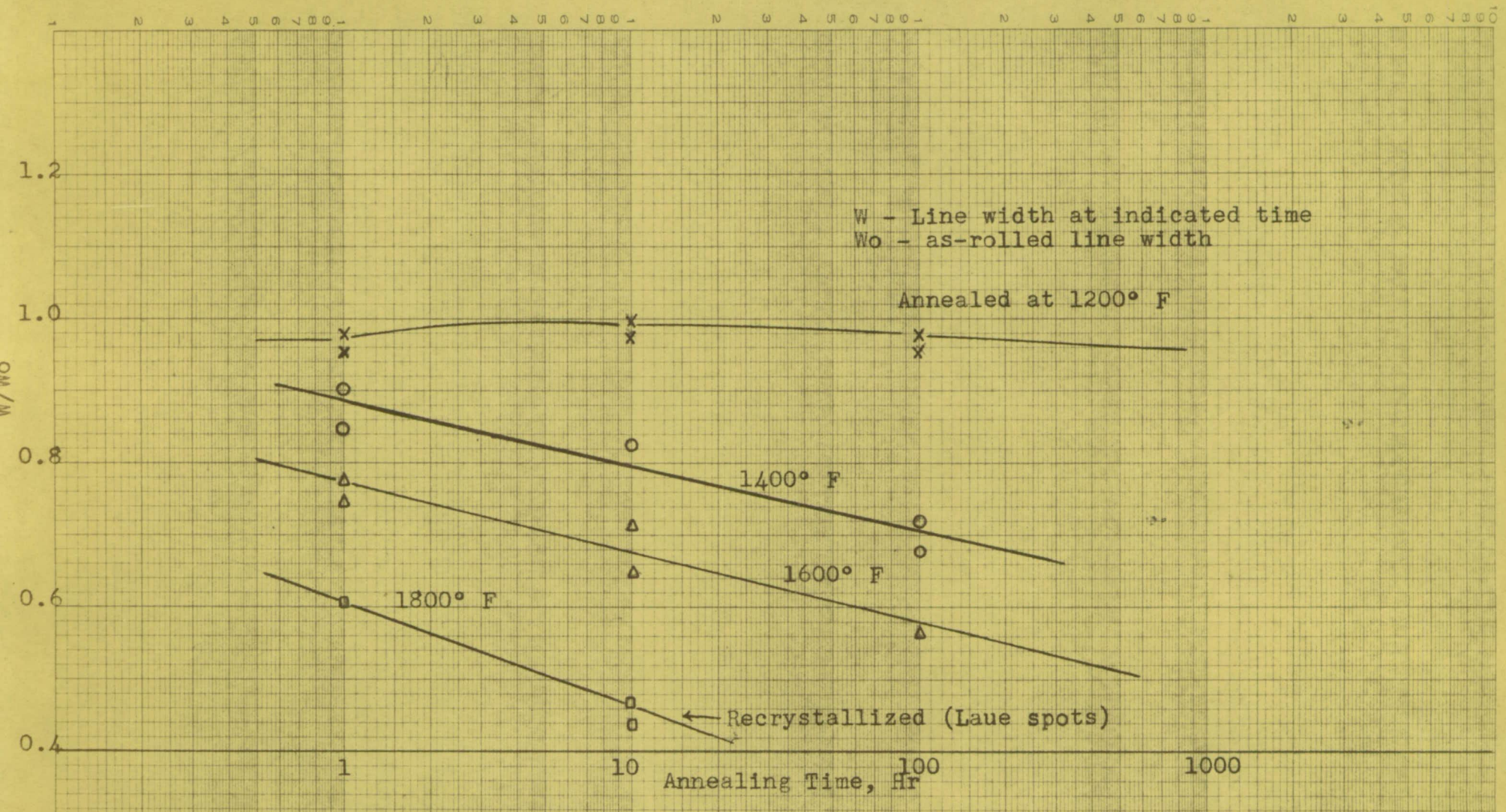


FIGURE II-16.- EFFECT OF ANNEALING ON (220) LINE WIDTH OF LOW-CARBON N155 ALLOY REDUCED IN CROSS SECTION 15% AT 1400° F, AFTER BEING SOLUTION-TREATED 1 HOUR AT 2200° F AND WATER-QUENCHED.

KEUFFEL & ESSER CO., N. Y. NO. 359-125G
Logarithmic, 5 x 5 Cycles.
P. C. C. Sheet.
MADE IN U. S. A.

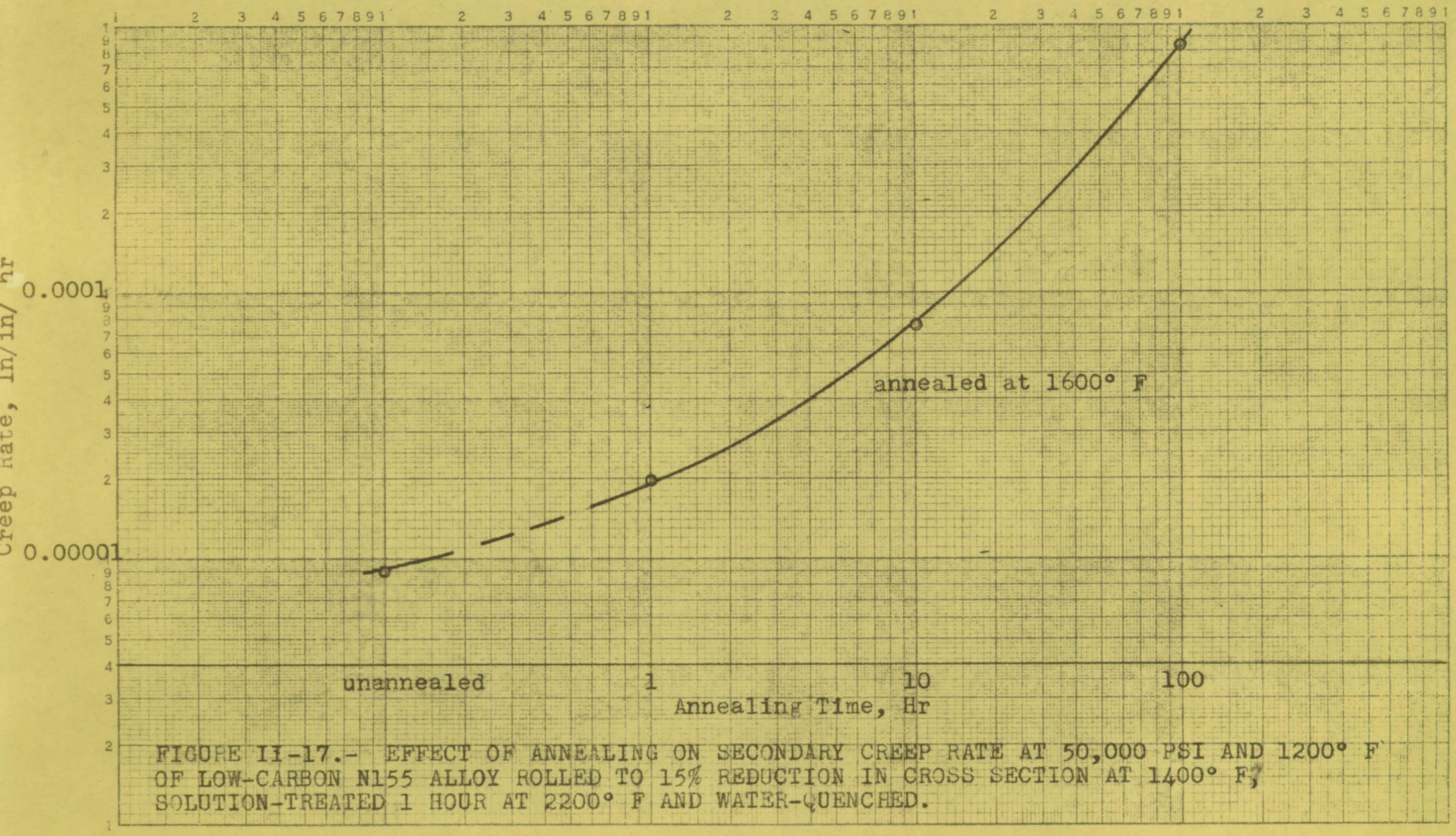


FIGURE II-17.- EFFECT OF ANNEALING ON SECONDARY CREEP RATE AT 50,000 PSI AND 1200° F OF LOW-CARBON N155 ALLOY ROLLED TO 15% REDUCTION IN CROSS SECTION AT 1400° F, SOLUTION-TREATED 1 HOUR AT 2200° F AND WATER-QUENCHED.

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 at high temperatures. At present high-temperature testing has been limited to rupture tests at 1200° F. All alloys tested are being solution treated at 2200° F and aged at 1400° F for 24 hours, a treatment found to best eliminate variations due to forging conditions.

A. Procedure for Production of Test Alloys

As a result of forging experiments a satisfactory forging procedure has been developed. The 400-pound forging hammer has been fitted with dies containing swaging impressions. The tapered ingots (1 1/2 to 1 1/4 inches square) are forged between flat dies to 1-inch squares. They are then reduced to approximately 0.40-inch rounds in a series of four swaging impressions. Total hot-working reduction is approximately 90 percent. The maximum hot-working temperature is 2200° F with minimum temperatures of approximately 1800° F.

Microstructural examination indicates that this forging practice refines the cast structure and produces bar stock having satisfactory uniformity of cross section microstructure. It is the most easily controlled of many procedures tried experimentally, thus reducing the variability of forging conditions from heat to heat.

B. Results

1. Since the June progress report, 51 heats have been prepared. Aim compositions of these heats are given in table III-1. Chemical analyses of at least the major alloy modifications are under way. These heats probably complete the various alloy modifications anticipated at the present time. The series will make possible a study of the separate variation of all the elements present in standard Low-Carbon N155 and of the simultaneous variation of the elements Mo, W, and Cb.

Pouring temperatures and cooling curves were measured for each ingot. Macrostructures of the ingots have been recorded. As the ingots are forged, sufficient data are taken to evaluate the relative hot-working characteristics resulting from the composition modifications.

2. Microstructural examination is being made of all the heats. From figure III-1, showing microstructures of the as-cast condition of four heats with carbon varying from 0.075 percent to 0.57 percent, it is evident that increasing carbon increases the amount and changes considerably the appearance of the excess constituents.

Further similar studies of the effect of chemical composition on the structures of both the ingots and the forged bars are in progress. These studies do show certain characteristic effects from varying the composition which will be of considerable significance in clarifying the effects of the alloys on properties.

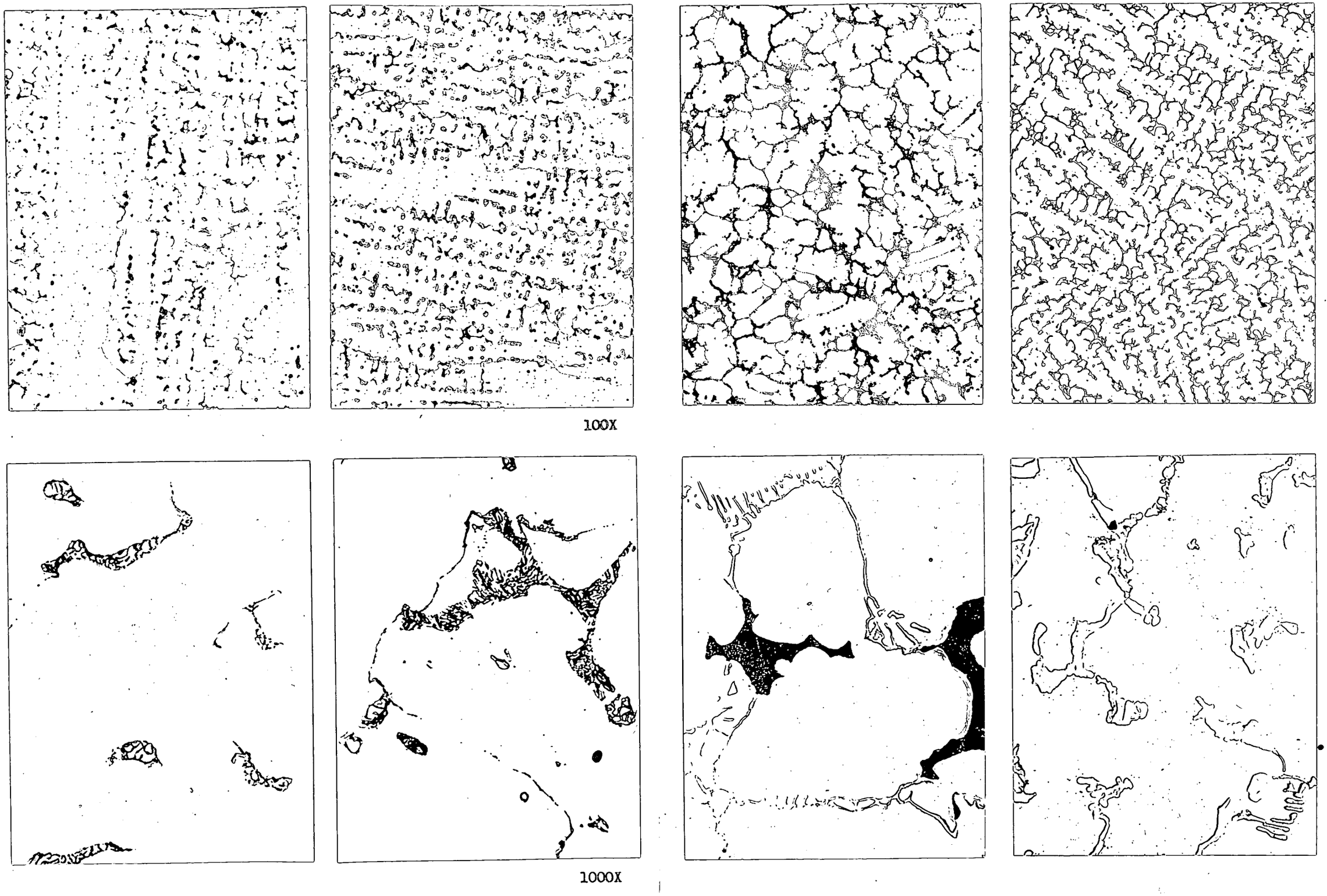
3. Reproducibility of rupture strengths by the procedures developed for making the alloys is indicated by the range in rupture strengths

TABLE III-1

AIM CHEMICAL COMPOSITION OF 51 EXPERIMENTAL HEATS

Note: Standard analysis: 0.15C, 0.5Si, 1.7Mn, 20Cr, 20Ni, 20Co, 3Mo, 2W, 1Cb, 0.12N

Heat number	Elements modified from standard analysis (percent)	Heat number	Elements modified from standard analysis (percent)
EN24	<0.05Mn	EN43	0Mo 0W 0Cb
		EN44	0Mo 0W 0Cb
EN23	0Ni		
EN25	10Ni	EN45	2Mo 0W 0Cb
EN26	30Ni	EN46	4Mo 0W 0Cb
EN27	0.25Si	EN53	0Mo 2W 0Cb
EN28	1.00Si	EN54	0Mo 4W 0Cb
EN29	0Co	EN55	2Mo 2W 0Cb
EN30	10Co	EN56	4Mo 2W 0Cb
EN31	30Cb	EN57	2Mo 4W 0Cb
		EN58	4Mo 4W 0Cb
EN51	10Cr	EN59	2Mo 2W 2Cb
EN52	30Cr	EN60	4Mo 2W 2Cb
		EN61	2Mo 4W 2Cb
EN32	0Mo	EN62	2Mo 2W 4Cb
EN33	1Mo		
EN34	2Mo	EN63	2Mo 0W 2Cb
EN35	4Mo	EN64	2Mo 0W 4Cb
EN36	6Mo	EN65	4Mo 0W 2Cb
		EN66	4Mo 0W 4Cb
EN37	0W	EN67	4Mo 4W 4Cb
EN38	1W		
EN39	4W	EN68	0Mo 0W 2Cb
EN40	6W	EN69	0Mo 0W 4Cb
		EN70	0Mo 2W 2Cb
EN47	0Cb	EN71	0Mo 4W 2Cb
EN48	2Cb	EN72	0Mo 2W 4Cb
EN49	4Cb	EN73	0Mo 4W 4Cb
EN50	6Cb		
EN41	<.03N		
EN42	.07N		



a. 0.075% C
EN13

b. 0.16% C
EN12

c. 0.40% C
EN15

d. 0.57% C
EN16

Electrolytic chromic acid etch

FIGURE III-1.- EFFECT OF CARBON ON THE MICROSTRUCTURES OF LOW-CARBON N155 ALLOY INGOTS.

and ductility for 5 heats of the standard alloy in table III-2. Apparently variations greater than ± 3000 psi in 100-hour rupture strength and ±1250 psi in 1000-hour strength will be required to obtain significant trends from one heat of a given composition.

4. Rupture data obtained to date from heats of variable composition are reported in table III-2, III-3, and figure III-2. The summary of these results in figure III-3 indicates that when Low-Carbon N155 alloy is hot-worked, solution treated at 2200° F and aged at 1400° F for 24 hours:

(a) Carbon contents between 0.07 and 0.60 percent have little significant effect on rupture strength at 1200° F.

(b) Nickel can be varied between 0 and 30 percent with no significant effect.

(c) Cobalt content data are incomplete. The indications are that more than 10 percent is required for a significant effect.

(d) Manganese and silicon contents have shown little effect in the ranges covered by the data.

(e) The most important chemical composition effects will result from variations in Mo, W, and Cb as is indicated by the low strength of Heat En43 in which these elements were omitted:

	Rupture strength (psi)	
	100 hr	1000 hr
Average for 5 heats of standard alloy	48,800	37,200
EN43 (no Mo, W, or Cb)	26,000	19,500

TABLE III-2

RUPTURE TEST PROPERTIES AT 1200° F FOR FIVE EXPERIMENTAL HEATS OF STANDARD LOW-CARBON N155 ALLOY

Treatment: 2200° F 1 hour, water quenched plus 1400° F 24 hours

Heat number	Stock size	Rupture strength (psi)		Estimated 100-hr rupture elongation (percent)
		100 hr	1000 hr	
EN7	3/8 in. sq	48,500	37,000	20
	1/2 in. rd	49,000	37,000[a]	30
EN8	3/8 in. sq	46,500	36,000[a]	18
EN10	3/8 in. sq	52,000	36,500	20
EN11	1/2 in. sq	48,000	38,000	22
EN12	3/8 in. sq	49,000	38,500	22
Average properties for five heats		48,800	37,200	22

[a]Estimated.

TABLE III-3

RUPTURE TEST RESULTS AT 1200° F FOR EXPERIMENTAL HEATS OF MODIFICATIONS OF LOW-CARBON N155 ALLOY

Treatment: 2200° F 1 hour, water quenched plus 1400° F 24 hours
Specimen size: 0.250-inch diameter, 1 inch gage length

Heat number	Alloy modification	Stock size	Stress (psi)	Rupture time (hours)	Elongation in 1 in. (percent)	Reduction of area (percent)	Rupture strength (psi) 100 hr	Rupture strength (psi) 1000 hr
EN13	0.075C	0.40 in. sq	50,000	79	14	21.3	48,500	36,000[a]
			45,000	222	14	10.9		
			40,000	325	26	21.8		
EN14	.36C	.45 in. sq	55,000	13	13	16.8		
EN15	.40C	.42 in. sq	50,000	106	19	21.1	50,500	36,500
			45,000	193	19	16.2		
			40,000	518	9	12.4		
EN16	.57C	.43 in. sq	50,000	39	18	18.4	46,000	37,500
			45,000	111	13	12.4		
			40,000	486	12	16.2		
		.51 in. rd	49,734	78	17	16.1	48,000	37,000
			45,000	178	14	16.2		
			40,000	528	13	15.4		
EN24	<.05Mn (aim)	.41 in. rd	50,000	125	13	10.9	51,000[a]	------
			45,000	[b]I.P. 48 hours				
EN19	.30Mn	.44 in. sq	53,000	43	22	23.4	49,000	41,000[a]
			50,000	83	24	22.6		
			45,000	352	19	21.8		
EN20	.50Mn	.40 in. sq	50,000	67	22	21.9	48,000	37,000[a]
			45,000	246	21	22.6		
			40,000	452	23	20.5		
EN21	1.04Mn	.41 in. sq	50,000	53	21	19.9	47,000	38,500
			45,000	199	14	19.7		
			40,000	617	16	17.6		
EN22	2.58Mn	.41 in. sq	50,000	85	20	18.4	49,000	37,500
			45,000	208	19	23.4		
			40,000	563	26	24.0		
EN23	0Ni (aim)	.45 in. rd	50,000	43	13	13.1	46,000	35,000
			45,000	133	25	24.0		
			40,000	305	25	24.8		
EN25	10Ni (aim)	.47 in. rd	50,000	86	17	14.7	49,000	38,000
			45,000	217	15	14.6		
			40,000	642	14	15.4		
EN26	30Ni (aim)	.45 in. rd	50,000	56	24	22.6	47,000	36,000
			45,000	152	27	24.8		
			40,000	369	29	26.2		
EN27	0.59Si (0.25Si aim)	.43 in. rd	50,000	82	26	25.5	49,000	37,000
			45,000	176	18	16.8		
			40,000	525	23	27.5		
EN28	1.0Si (aim)	.44 in. rd	50,000	41[c]	26	20.5	45,000	34,000
			45,000	115	28	30.9		
			40,000	224	31	26.8		
			36,000	[b]I.P. 240 hours				
EN29	0.31Co	.42 in. rd	45,000	89	19	22.6	44,500	35,000[a]
			40,000	278	18	27.5		
			36,000	[b]I.P. 432 hours				
EN30	11.09Co	.41 in. rd	45,000	86	14	24.0	43,000	------
			40,000	142	25	22.6		
			35,000	[b]I.P. 384 hours				
EN31	32.60Co	.41 in. rd	50,000	[b]I.P. 24 hours				
			45,000	[b]I.P. 240 hours				
EN32	0.20Mo (0Mo aim)	.41 in. rd	50,000	9.5	23	24.0		
			40,000	[b]I.P. 24 hours				
EN33	1.00Mo	.42 in. rd	50,000	13.5	13	10.9		
			40,000	[b]I.P. 24 hours				
EN34	2.34Mo (2Mo aim)	.42 in. rd	50,000	[b]I.P. 48 hours				
			45,000	[b]I.P. 72 hours				
EN43	0Mo 0W 0Cb	.45 in. rd	35,000	11	8	13.9	26,000	19,500
			25,000	141	5	7.2		
			20,000	819	13	7.1		

[a]Estimated.
[b]Test in progress 9-24-48.
[c]Broke in gage mark.

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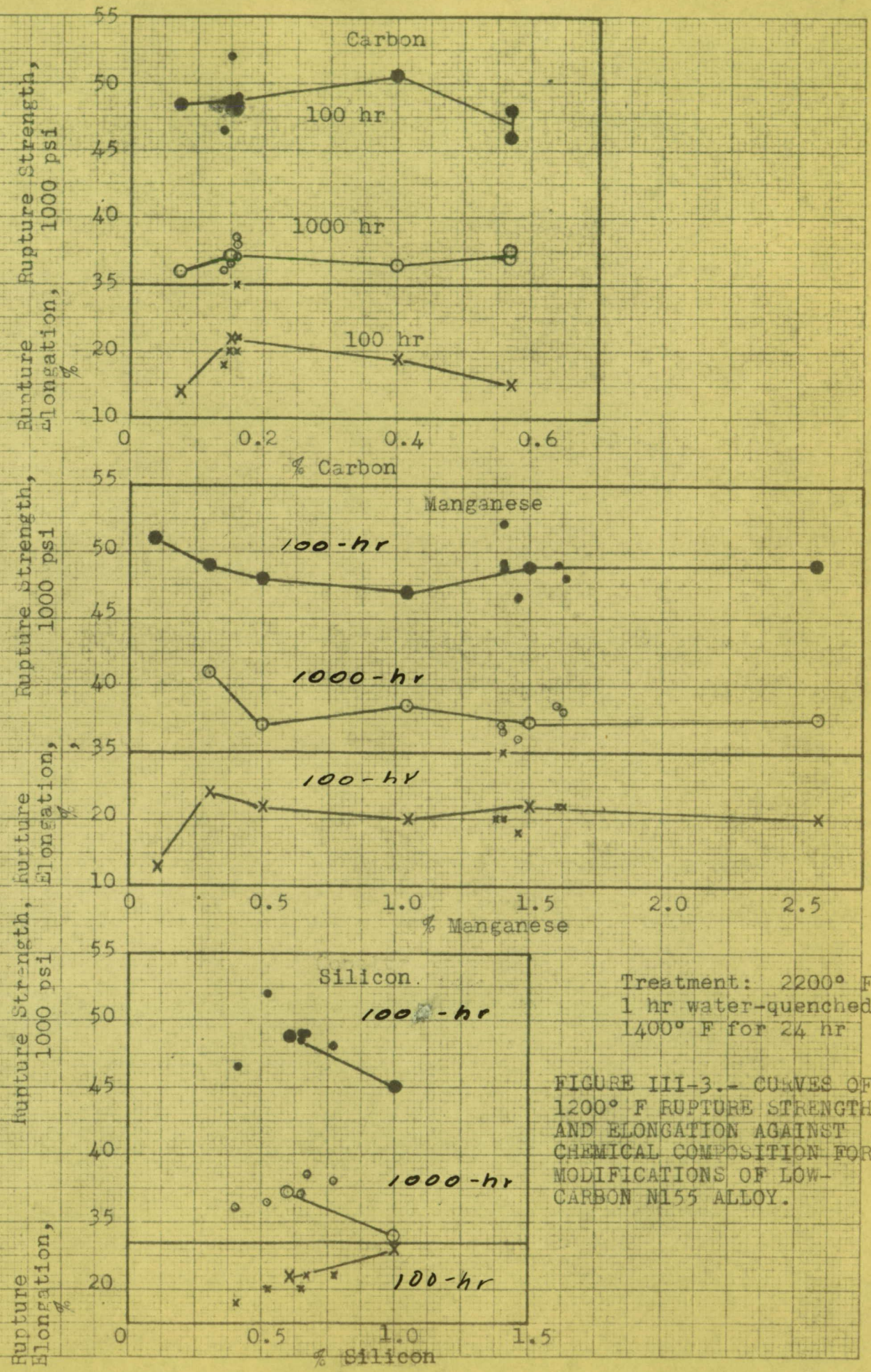
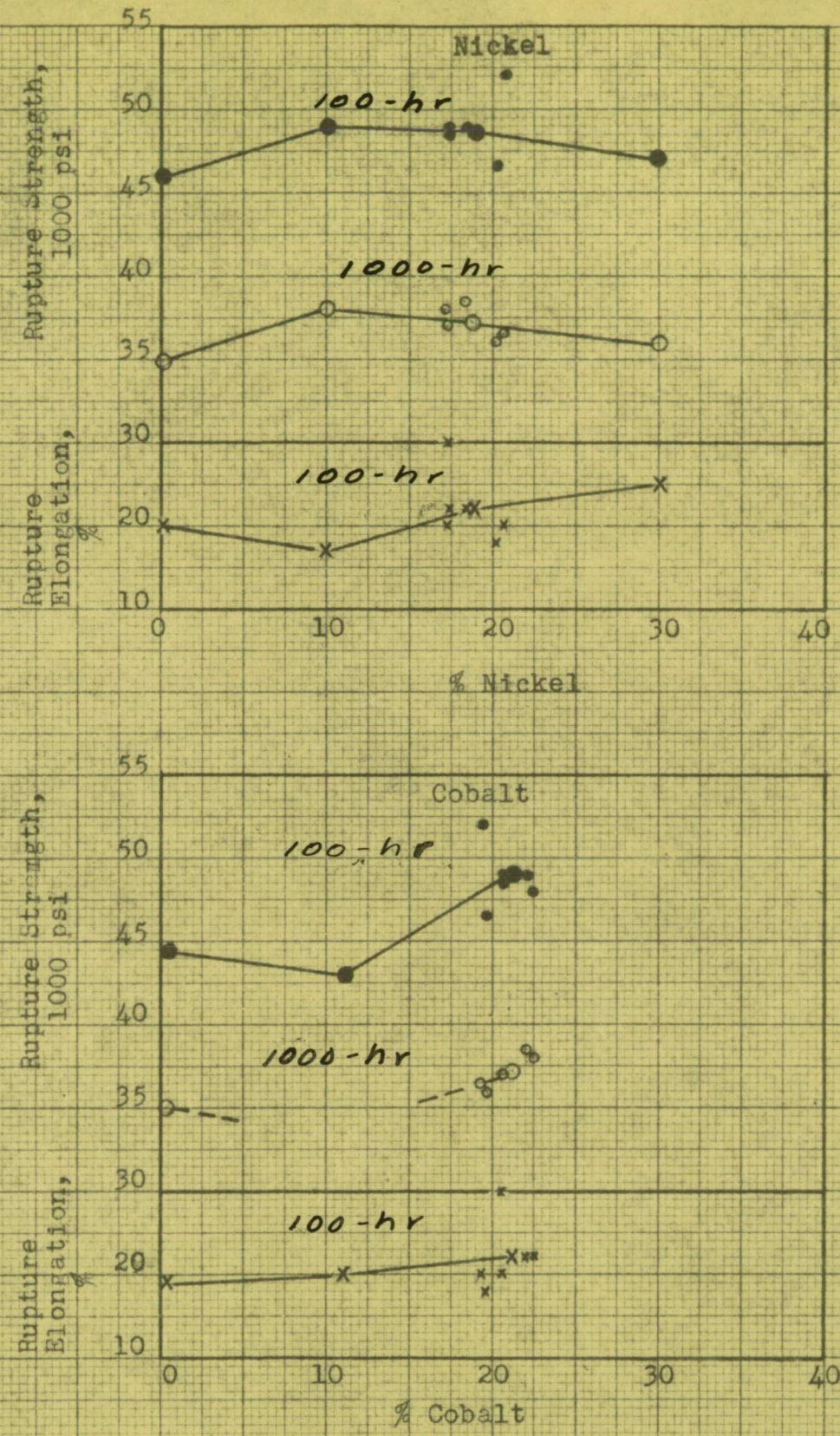
Figure III



Treatment: 2200° F 1 hr water-quenched; 1400° F for 24 hr

FIGURE III-3.- CURVES OF 1200° F RUPTURE STRENGTHS AND ELONGATION AGAINST CHEMICAL COMPOSITION FOR MODIFICATIONS OF LOW-CARBON N155 ALLOY.



Treatment: 2200° F 1 hr water-quenched; 1400° F for 24 hr

FIGURE III-3 (cont) CURVES OF 1200° F RUPTURE STRENGTHS AND ELONGATION AGAINST CHEMICAL COMPOSITION FOR MODIFICATIONS OF LOW-CARBON N155 ALLOY.

ULRICH'S BOOK STORE ANN ARBOR, MICH.
KEUFFEL & ESSER CO., N. Y. NO. 359-11 10 × 10 to the ½ inch, 5th lines accented. Engraving 7 × 10 in. MADE IN U.S.A.

IV - FURTHER WORK ON HEAT-TREATING AND PROCESSING PROCEDURES

A. 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 and at 1650° and 1800° F for four treatments, are given in table IV-1 and compared with the properties of the original lot of Low-Carbon N155 alloy bar stock (Lot 30276) used in this program. The rupture properties of the new heat are plotted against temperature in figure IV-1.

Discussion of comparative properties of the two heats, given in the June progress report, indicated that when Low-Carbon N155 hot-rolled stock is solution treated at 2200° F, differences in properties between heats are small. Differences between lots when heat treated at lower temperatures could be due to the influence of prior hot-work. It is also this prior hot-working history which can govern the effectiveness of subsequent hot-cold working.

Examination of figure IV-1 shows, for heat A-1726, the following trends:

a. Material solution treated at 2050° F and then hot-cold rolled to 15 per cent reduction at 1400° F has the highest rupture strength at 100 hours in the temperature range 1250° to 1700° F and the highest rupture strength at 1000 hours in the range 1250° to 1575° F.

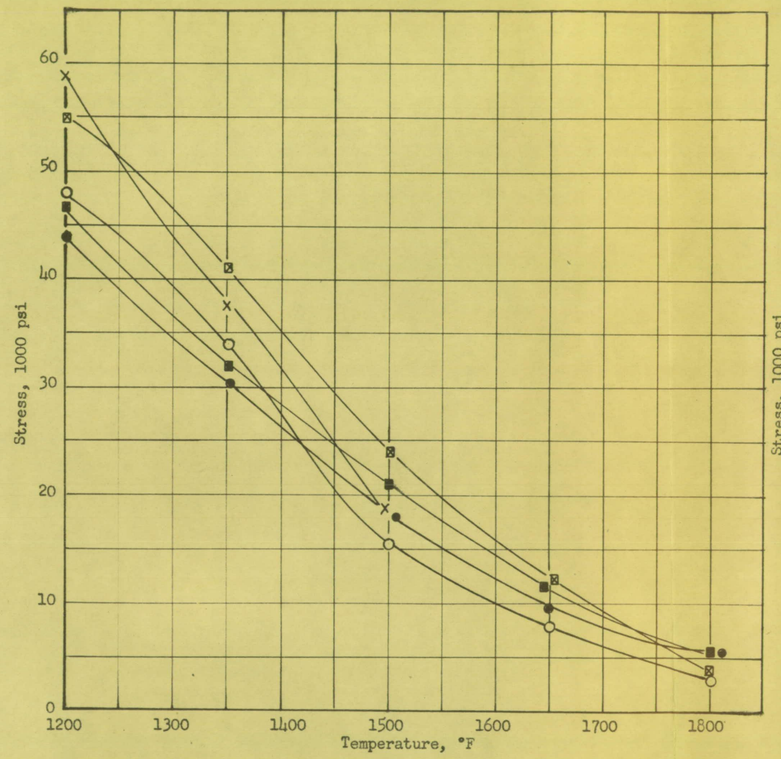
b. At 1200° F the material hot-rolled and then hot-cold worked to 15 per cent reduction at 1400° F had the best 100-hour and 1000-hour rupture strength. The relative strength of this material fell off rapidly with temperature, however, and was at 1500° F nearly the weakest material tested, the

TABLE IV-1

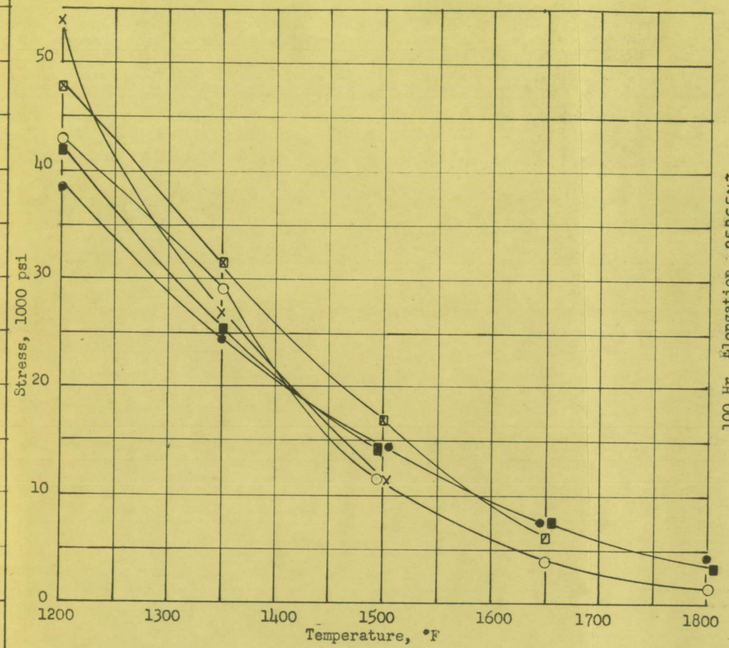
COMPARATIVE RUPTURE PROPERTIES OF TWO HEATS OF LOW-CARBON N155 ALLOY

Treatment	Test temperature (°F)	Rupture-test properties					
		Heat A-1726[a]			Lot 30276[b]		
		Strength (psi)		Elongation % in 1 in.	Strength (psi)		Elongation % in 1 in.
		100 hour	1000 hour	100 hour	100 hour	1000 hour	100 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
	1650	7,800	4,100	19			
	1800	(3,100)[c]	(1,500)[c]	36			
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
	1650	(11,500)[c]	(7,700)[c]	(30)			
	1800	5,700	(3,300)[c]	25			
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
	1650	9,600	7,700	30			
	1800	(5,600)[c]	(4,200)[c]	—			
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
	1500	18,700	11,800	12			
2050° F 2 hr, W.Q.; 15% reduction at 1200° F	1200	55,000	48,000	3	62,000	53,500	1
	1350	41,000	31,500	6	38,000	28,500	12
	1500	24,000	17,000	14	22,000	12,500	16
	1650	(12,000)[c]	6,000	(8)			
	1800	(4,000)[c]	----	(18)			

A [a]All test specimens taken from center bar from the ingot
[b]See data given in section I-4 of preceding progress report.
[c] Based on incomplete tests



100 Hour Rupture Strengths



1000 Hour Rupture Strengths

Low-Carbon N155 Alloy (heat A-1726)

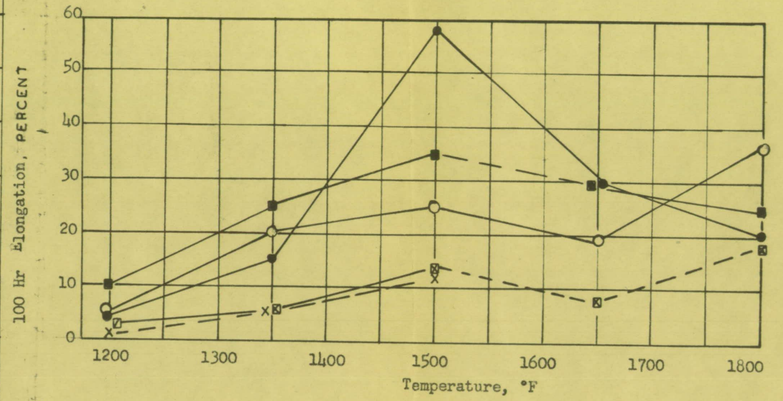
○ Hot-Rolled

x Hot-Rolled; 15% Reduction

● 2100° F 1 hr; Water-quenched

⊠ 2050° F 2 hr; Water-quenched; 15% Reduction

■ 2200° F 1 hr; Water-quenched; 24 hr at 1400° F



Elongation Characteristics of Low-Carbon N155 Alloy (A-1726).

FIGURE IV- 1. - EFFECT OF TEMPERATURE ON THE RUPTURE TEST CHARACTERISTICS OF LOW-CARBON N155 ALLOY (Heat A-1726).

weakest material being the hot-rolled bar stock.

c. At the upper end of the temperature range investigated, material solution treated one hour at 2200° F and aged 24 hours at 1400° F had a relatively good rupture strength. The 100-hour rupture strength was in fact the best found in the range 1750° to 1800° F and the 1000-hour the best in the range 1575° to 1800° F. At the lower temperature range, this material had intermediate strengths.

d. Material simply solution treated one hour at 2100° F and water quenched had rupture strength characteristics closely paralleling the material covered in "c" above. At both 100 and 1000 hours the 2100° F solution treatment gave slightly lower strengths than the material solution treated at 2200° F and aged except in the temperature range 1700° to 1800° F when the strengths of the two materials were nearly identical.

e. As-hot-rolled material from Heat A01726 had intermediate strength at both 100 and 1000 hours in the range 1200° to 1400° F and the poorest strength in range 1400° to 1800° F.

f. Highest ductility was exhibited by the material solution treated at 2200° F and aged except in the temperature ranges 1400° to 1650° F where the 2100° F solution-treated material was superior and in the range 1750° to 1800° F where the hot-rolled material was superior. Poorest ductility was exhibited in all cases by the materials given 15 per cent hot-cold work at 1400° F.

B. Uniformity of As-Rolled Stock

The uniformity of high-temperature properties of the as-rolled bar stock of Low-Carbon N155 alloy used for the studies covered by this program has been checked by means of rupture tests at 1200°, 1350°, and 1500° F. Test specimens were taken from bars representing the bottom, center, and top of the ingot with the results in table IV-2 and figure IV-2.

These tests serve two purposes. They show the degree of variability originally present in the stock used for this investigation. In addition they indicate the degree of variability which might normally be expected in the product of an ingot rolled to bar stock.

The results show some variability. The bar from the center of the ingot was definitely stronger at 1200° and 1350° F than those from the ends. Specimens from one bar may also be erratic. In general, however, the spread in properties was not as large as might be expected in view of the known pronounced influence of hot-working conditions on rupture properties. For this reason they should encourage the control of hot-working conditions as a means of producing high rupture strengths in alloys of this type.

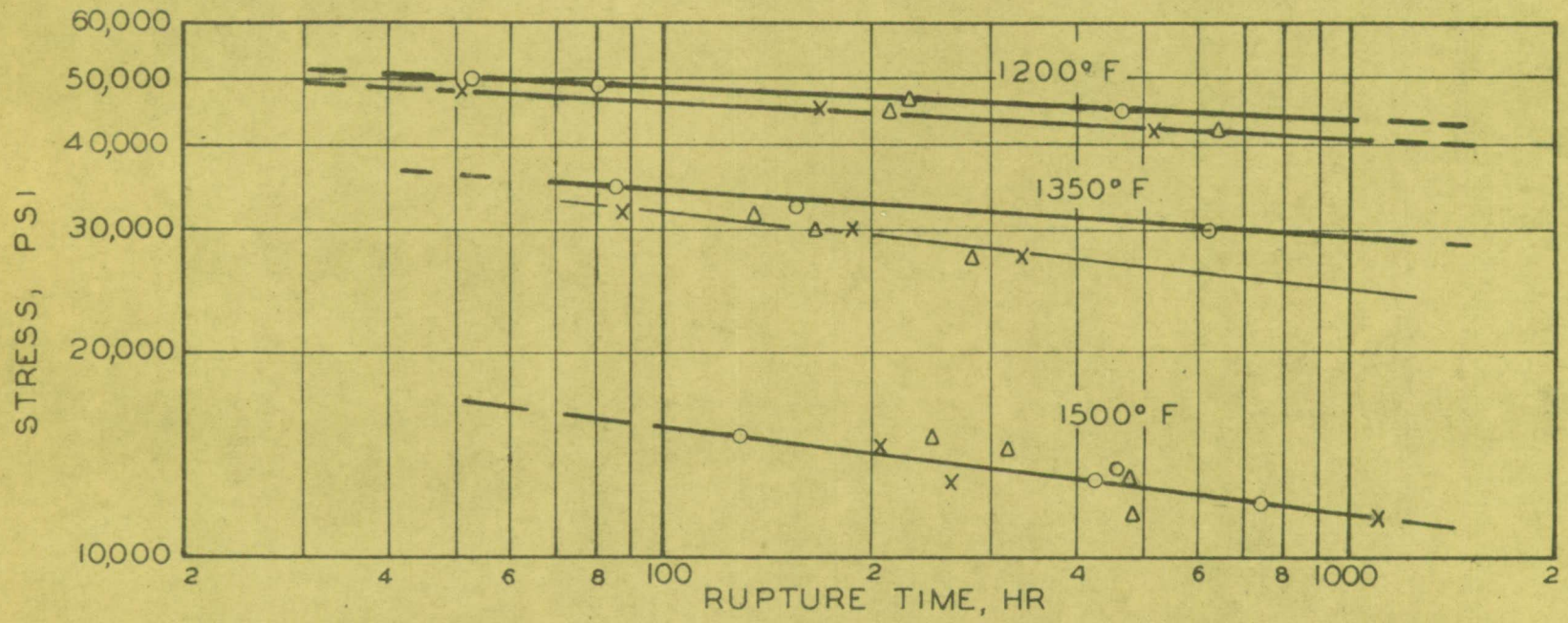
TABLE IV-2

RUPTURE TEST RESULTS AT 1200°, 1350°, AND 1500° F ON HOT-ROLLED BAR STOCK FROM VARIOUS LOCATIONS IN THE INGOT OF HEAT A-1726

Ingot position	Temperature (°F)	Stress (psi)	Rupture time (hours)	Elongation (percent)	Reduction of Area (percent)
Top	1200	48,000	51[a]	4	6.1
		45,000	169[b]	8	10.8
		42,000	520	14	15.3
Center	1200	50,000	52[a]	2	10.2
		49,000	80[a]	5	6.1
		45,000	471	12	11.3
Bottom	1200	47,000	228	11	16.3
		45,000	212[a]	5	8.0
		42,000	654[a]	10	10.2
Top	1350	32,000	87	34	27.3
		30,000	188	31	27.3
		27,500	332	27	32.7
Center	1350	35,000	85	18	21.6
		32,500	156	36	35.5
		30,000	623	23	19.9
Bottom	1350	32,000	134	28	17.1
		30,000	166[b]	23	27.4
		27,500	278	28	35.4
Top	1500	14,500	205	11	16.0
		13,000	264	17	18.0
		11,500	1099	14	11.9
Center	1500	15,000	130	25	25.2
		13,500	458	23	24.0
		13,000	432	22	19.8
		12,000	747	17	20.0
Bottom	1500	15,000	246	22	24.0
		14,500	316[b]	13	25.2
		13,000	478	11	15.0
		11,500	478[b]	10	9.8

[a]Specimen broke in fillet.

[b]Specimen broke in gage mark.



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

FIGURE IV-2.-STRESS-RUPTURE TIME CURVES SHOWING UNIFORMITY OF AS ROLLED LOW-CARBON N155 ALLOY AT 1200° F, 1350° F, and 1500° F.

V - COOPERATIVE FATIGUE TEST PROGRAM

The Low-Carbon N155 alloy (Heat A-1726) being used in this program was described in the progress report of March 13. Prior to machining, the one-inch round bar stock was heat treated at 2200° F for one hour, water quenched, and then aged 16 hours at 1400° F. The following work has been completed:

1. A hardness survey of heat-treated bar stock was made to check the uniformity of the material.

2. The results of short time tensile tests at 1000°, 1200°, 1350°, and 1500° F are given in table V-1.

3. The results of the stress rupture tests which have been completed to date at 1200°, 1350°, and 1500° F are shown in table V-2 and figure V-1.

4. (a) Forty-five Krause fatigue test specimens have been heat treated, machined, and shipped to the Battelle Memorial Institute for tests at 1200°, 1350°, and 1500° F under the sponsorship of the Office of Naval Research.

(b) Fifteen bars have been heat-treated and are being held in reserve for possible additional required fatigue tests at Battelle.

(c) Twenty-four bars have been heat-treated for proposed zero mean stress tests in the Krause machine if a suitable technique can be developed.

5. Thirty-six fatigue bars have been heat-treated, machined, and shipped to the Westinghouse Company for tests at 1000°, 1200°, 1350°, and 1500° F.

6. Fifty fatigue bars have been heat-treated and are now being machined for dynamic creep tests at 1200°, 1350°, and 1500° F by Syracuse University under sponsorship of Wright Field.

TABLE V-1

TENSILE TEST DATA FOR LOW-CARBON N155 ALLOY (HEAT A-1726) FOR COOPERATIVE FATIGUE TESTING PROGRAM

Specimen number*	Test temperature (°F)	Tensile strength (psi)	Proportional limit (psi)	Offset yield strengths (psi)				Elongation in 2 in. (percent)	Reduction of area (percent)
				0.01%	0.02%	0.10%	0.20%		
JM1	Room	119,100	41,000	46,000	48,700	56,100	59,500	45	46
JY1	Room	119,000	40,000	47,600	50,500	58,500	61,500	42	45
JF1	1000	91,250	26,750	31,750	32,500	34,800	35,800	44	49
JW1	1000	93,900	26,000	31,750	34,000	38,750	40,000	39	46
JM4	1000	94,250	26,250	32,500	33,800	37,000	37,750	42	47
JP1	1200	81,200	25,750	29,500	30,500	34,250	35,250	35	38
JG1	1200	79,600	26,000	29,500	31,000	34,900	35,800	33	34
JX1	1350	60,250	22,250	27,500	29,750	34,750	36,500	27	28
JN1	1350	60,125	23,500	28,750	30,750	35,250	37,200	26	28
JE1	1500	45,600	20,000	26,250	28,500	33,800	35,800	19	27
JR1	1500	43,625	20,500	26,800	28,500	33,200	35,800	25	27

*Location of specimen in the ingot.

TABLE V-2

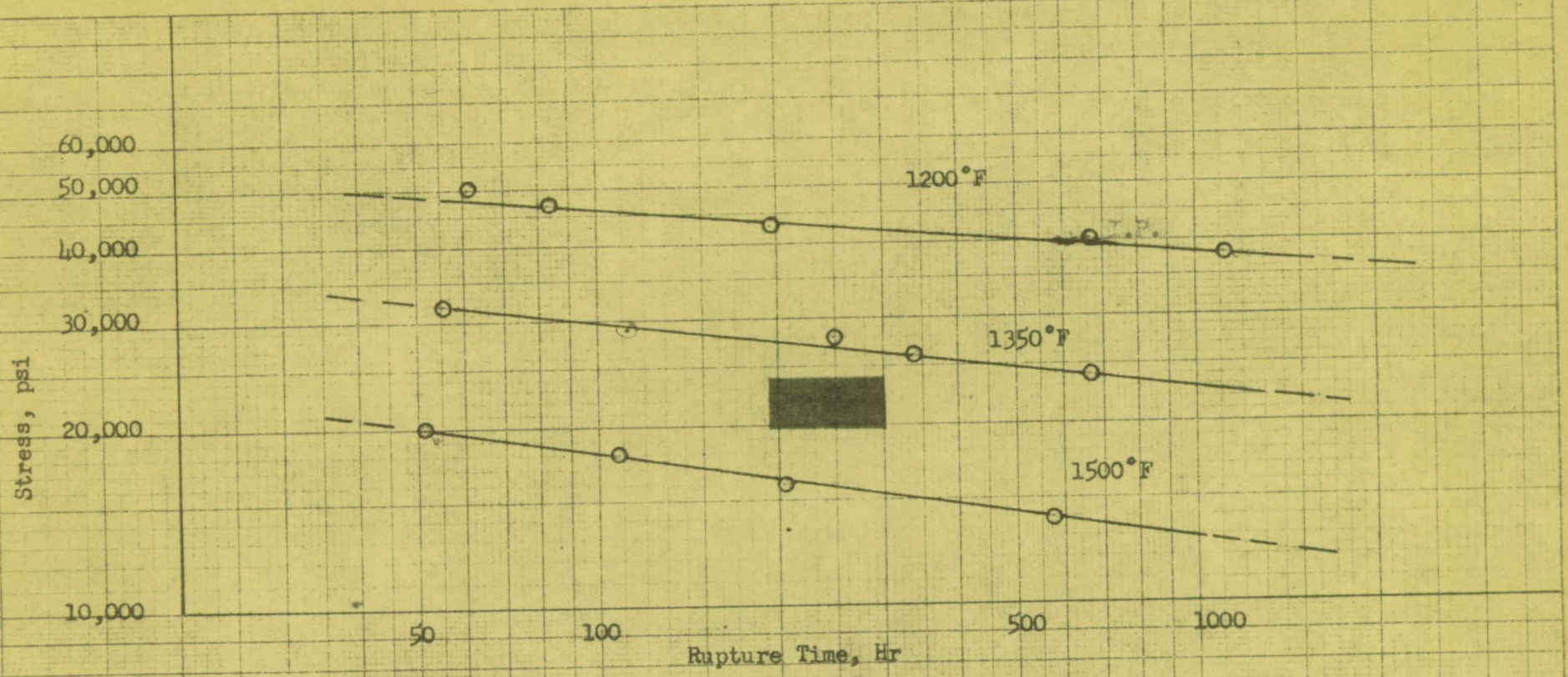
STRESS-RUPTURE PROPERTIES OF LOW-CARBON N155 ALLOY BAR STOCK FOR COOPERATIVE FATIGUE TESTING PROGRAM

Heat A-1726, solution treated 1 hour at 2200° F, water quenched plus 16 hours at 1400° F

Temperature (°F)	Stress (psi)	Rupture time (hours)	Elongation in 2 in. (percent)	Reduction of area (percent)
1200	50,000	61	10	10
	47,000	83	16	10
	43,000	195	15	8.5
	40,000	668	10	16
	38,000	1107	20	18
1350	32,000	55	20	23
	29,000	112	37	40
	28,000	248	25	35
	26,000	336	30	43
	24,000	665	20	30
	22,000	In progress 350 hours.		
1500	20,000	51	34	37
	18,000	108	28	32
	16,000	203	25	37
	14,000	575	26	33

Approximate Rupture Strengths from Available Data

Temperature (°F)	Stress (psi) for rupture in -		
	50 hours	150 hours	500 hours
1200	49,000	44,000	40,000
1350	32,500	28,500	24,500
1500	20,000	17,000	14,000



Heat Treatment: 2200°F 1 hr; water-quenched, and 16 hours at 1400°F

[redacted]

FIGURE V-1. -STRESS-RUPTURE TIME CURVES AT 1200°F, 1350°F, AND 1500°F ON LOW-CARBON N155 ALLOY (Heat A-1726).

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