Fifth

PROGRESS REPORT

(Project SR-99)

on

THE FUNDAMENTAL FACTORS INFLUENCING THE BEHAVIOR OF WELDED STRUCTURES:

The Effect of Subcritical Heat Treatment on the Transition Temperature of a Low Carbon Ship Plate Steel

by

E. B. Evans and D. J. Garibotti

CASE INSTITUTE OF TECHNOLOGY

Transmitted through

NATIONAL RESEARCH COUNCIL'S
COMMITTEE ON SHIP STEEL

Advisory to

SHIP STRUCTURE COMMITTEE

Division of Engineering and Industrial Research
National Academy of Sciences - National Research Council
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ADDRESS CORRESPONDENCE TO:

SECRETARY
SHIP STRUCTURE COMMITTEE
U. S. COAST GUARD HEADQUARTERS
WASHINGTON 25, D. C.

October 30, 1953

Dear Sir:

As part of its research program related to the improvement of hull structures of ships, the Ship Structure Committee is sponsoring an investigation on "The Fundamental Factors Influencing the Behavior of Welded Structures under Conditions of Multiaxial Stress and Variations of Temperature" at the Case Institute of Technology. Herewith is a copy of the Fifth Progress Report, SSC-61, of the investigation, entitled "The Fundamental Factors Influencing the Behavior of Welded Structures: The Effect of Subcritical Heat Treatment on the Transition Temperature of a Low Carbon Ship Plate Steel" by E. B. Evans and D. J. Garibotti.

The project is being conducted with the advisory assistance of the Committee on Ship Steel of the National Academy of Sciences-National Research Council.

Any questions, comments, criticism or other matters pertaining to the Report should be addressed to the Secretary, Ship Structure Committee.

This Report is being distributed to those individuals and agencies associated with and interested in the work of the Ship Structure Committee.

Yours sincerely,

K. K. Cowart

K. K. COWART
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

FIFTH
Progress Report
(Project SR-99)

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under

Department of the Navy
Bureau of Ships NOBS-45470
BuShips Project No. NS-011-078[illegible]

for

<u>SHIP STRUCTURE COMMITTEE</u>

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ABSTRACT

An investigation was made to determine the impact transition temperature and hardness changes attendant to the quench-aging of Project Steel "C" a semi-killed ship plate steel. Aging temperatures extended over the range from 35° to 1100°F after water quenching from 1300°F.

Both impact and hardness tests revealed that this steel can be severely embrittled by the quench-aging mechanism. With aging temperatures up to 350°F, characteristic aging curves were obtained, i.e., the peak embrittlement and the time to attain this peak decreased with increasing aging temperature. For room temperature aging this peak amounted to a 90°F increase in transition temperature and 25 points increase in Rockwell B hardness above that of a series air cooled from 1300°F (unembrittled condition). Specimens aged above 350°F 'overaged' so rapidly that no peak in the aging curve could be detected.

Metallographic examination of quench-aged specimens at 2000X showed that a two-stage precipitation reaction was operative. At low aging temperatures the precipitate was detected as a mottling of the ferrite grains; at higher aging temperatures, where an 'overaged' condition was rapidly reached, the precipitate had grown so as to be resolvable.

It is believed that the quench-aging phenomenon is responsible for the brittle zone previously found in the subcritically heated region in weldments of this and similar ship

plate steels. This study suggests that a low temperature post-heat at 650°F (the solution temperature below which quench-aging effects are absent) would lead to rapid overaging in the brittle zone of ship plate weldments and thus largely eliminate the embrittlement.

INTRODUCTION

This is the fifth and last progress report on the study of the zone of brittleness located adjacent to welds in semi-killed steel plates. The project has been sponsored by the Ship Structure Committee under Department of the Navy, Bureau of Ships Contract NObs-45470 and covers the period from September 1, 1952, to December 31, 1952. The earlier report on this phase of the investigation[1] covered the work from January 1, 1950, to September 1, 1952. Three Progress Reports, SSC-24[2], SSC-34[3], and SSC-54[4] summarized the work on eccentric notch tensile testing of ship plate weldments under the same contract over the period from July 1, 1947, to January 1, 1950.

In previous work reported (1), it was shown that sub-critically heated C steel* base plate could be embrittled, as a result of welding, in a zone adjacent to the weld. This evidence was obtained with changes in transition temperature (eccentric notch tensile and Charpy V-notch) and hardness. The degree of embrittlement increased with increasing (1) solution temperature in the 1100°--1200°F range, (2) cooling rate, and (3) room temperature aging time and was decreased by accelerated aging.

In view of the fact that quench-aging is believed responsible for the zone of minimum ductility found in ship plate

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*The designation "C" refers to Steel C in the series of Ship Structure Committee "Project" Steels.

weldments** in a region not heated above the lower critical temperature at any time(1), it was considered advisable to investigate the maximum embrittlement possible in this grade of steel by quench-aging and how to minimize or eliminate it. To this end, Charpy V-notch impact specimens were solution heat treated at the maximum subcritical temperature (1300°F), water quenched, and then aged for various periods of time in the 35° to 1100°F range. The embrittlement was evaluated by transition temperature and hardness changes, supplemented by microscopic examination.

MATERIAL

The semi-killed, ship plate steel (C Steel) used in the present work was the same 3/4-in. thick, as-rolled plate (Plate II) which supplied specimens for the earlier work on subcritical heat treatment. The properties reported for this steel are as follows:(5)

TABLE I

PROPERTIES OF C STEEL PLATE

Chemical Analysis

Carbon	0.24	Copper	0.03
Manganese	0.48	Chromium	0.03
Phosphorous	0.012	Molybdenum	0.005
Sulfur	0.026	Tin	0.003
Silicon	0.05	Nitrogen	0.009
Aluminum	0.016	Vanadium	0.02
Nickel	0.02	Arsenic	0.01

Mechanical Properties

Yield Point Psi	Tensile Strength Psi	Elongation Per Cent
39,000	67,400	25.5 (8" gauge)

**At least for weldments made of A and C steels, the two project steels investigated.

PROCEDURE

Ten impact blanks (0.420" square) were cut from the as-rolled plate so that the long axis of each specimen was perpendicular to the rolling direction. These blanks were used as hardness check specimens after heat treating at 1300°F for 15 minutes in a neutral chloride salt bath*. One specimen was air cooled while the remaining nine were water quenched. Each of the quenched specimens was then aged for various times in an appropriate bath of water, oil, or tempering salt at one of the following temperatures: 35°, room temperature (80°), 125°, 200°, 250°, 350°, 600°, 900°, and 1100°F. All specimens were air cooled from the aging temperatures.

The 'as-quenched' hardness was taken in about five minutes after quenching. In following the progress of aging, specimens were removed from the aging baths for a time only long enough for the hardness tests to be made.

With the hardness checks as a guide, series of Charpy V-notch impact blanks were taken from the plate in the same manner, quenched from 1300°F, and aged for various times at room temperature, 125°, 350°, and 600°F. In addition, one series was air cooled from 1300°F to give the unembrittled condition.**

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*The composition of the neutral chloride salt was as follows:

$BaCl_2$	55%	NaCl	22%
KCl	22%	"Dicy"	1%

**The Transition Temperature of this series was approximately the same as that of the as-received material. In addition, hardness checks of broken specimens showed no change with time at room temperature.

After aging, the specimens were immediately ground to size and the notch cut perpendicular to the surface of the plate, with each series being machined and tested within seven hours after aging. The impact testing procedure was the same as given previously[1]. As pointed out in the earlier work, the possibility exists that accelerated aging can occur in the testing bath. In establishing the transition curves, it was necessary to test as high as 350°F (specimens were held 10 minutes in the bath to assure temperature uniformity), and thus, accelerated aging in the test bath may occur in those specimens previously aged at room temperature and 125°F. Hardness checks made on specimens from these series before and after breaking indicated that the hardness was unaffected up to testing temperatures of 175°F. In view of the fact that the 15 ft-lb transition temperature is to be used as the criterion of embrittlement and in no case did this exceed 175°F, the results would appear to be unaffected.

RESULTS

Hardness Tests

The Rockwell B hardnesses obtained after aging in the 35° to 1100°F range are shown as a function of the aging time in Fig. 1.

From an as-quenched hardness of R_B87, the hardness of the specimen aged at room temperature (80°F) remained unchanged for about one hour then gradually increased, reaching a maximum

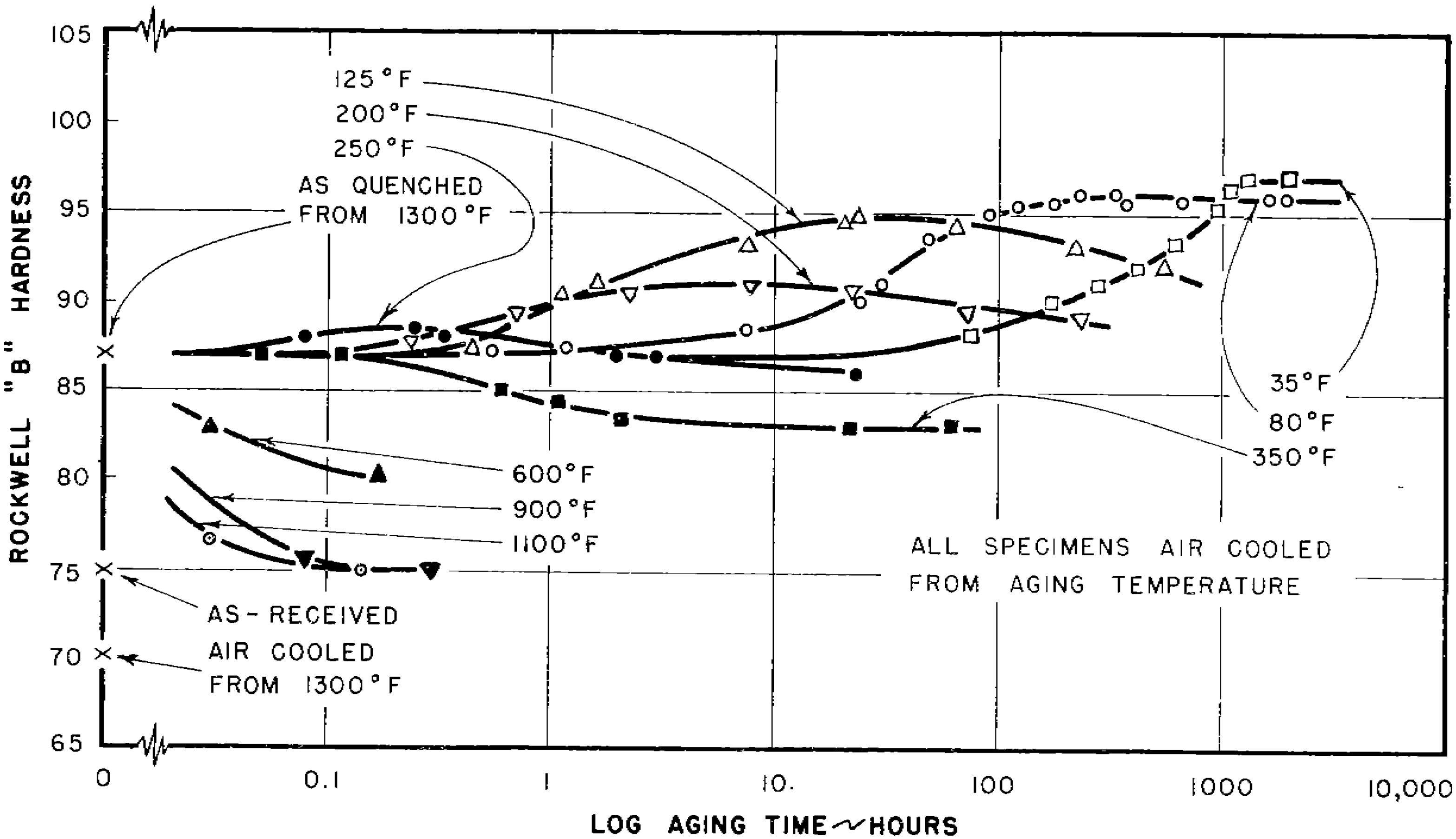


FIG. 1: EFFECT OF VARIOUS AGING TIMES AND TEMPERATURES ON THE HARDNESS OF "C" STEEL AFTER WATER QUENCHING FROM 1300°F.

level of R_B96 after five days. Aging at a lower temperature (35°F) resulted in a slower rate of increase and a longer time (about 50 days) to reach a slightly higher peak level of R_B97.

With increasing aging temperatures above room temperature to 250°F, the hardness increased at an ever increasing rate; but the peak hardness reached and the time to reach this peak decreased. Once the peak had been attained, further aging time caused a decrease in hardness at a rate which increased with aging temperature.

At 350°F and higher the hardness did not rise above that of the as-quenched specimen but fell off at a rate which again increased with temperature.

A summary of the maximum hardness reached and the time to reach this maximum for the various aging temperatures employed are given in Table II.

TABLE II

EFFECT OF AGING TEMPERATURE ON THE PEAK HARDNESS REACHED AND THE TIME TO ATTAIN THIS PEAK

Aging Temperature, °F	Peak Hardness Reached, R_B	Time to Reach Peak Hardness
35	97	50 days
80	96	5 days
125	95	23 hours
200	91	2 hours
250	88	15 minutes
350	Not higher than as-quenched hardness of R_B 87	
600	"	
900	"	
1100	"	

Note: All specimens water quenched from 1300°F prior to aging. Hardness of specimen air cooled from 1300°F was R_B 70.

These results show that by aging the hardness can be raised a maximum of about 10 points above that of the as-quenched hardness (R_B87). In turn the as-quenched hardness was 17 points Rockwell B above that of the air cooled hardness (R_B70). The maximum cumulative hardness increase due to solution and aging then amounts to about 27 points R_B.

Impact Tests

The impact transition temperatures by three criteria obtained of the (1) as-received*, (2) air cooled, and (3) the various quench-aged conditions are summarized in Table III. The individual transition curves for each condition are plotted in Figs. 2--7 with both the energy absorbed and the per cent fibrous fracture plotted as the ordinate.

In the following sections of the report, the effects of the various subcritical heat treatments are evaluated with the 15 ft-lb transition temperature as the criterion of embrittlement and with the properties of the air cooled series reflecting the unembrittled state. The choice of either of the other two criteria listed would reveal the same general effects.

In comparing the impact properties of the as-received plate, Fig. 2, with those of the air cooled series, Fig. 3, it is evident that there is little difference in properties other than about a five ft-lb higher upper level for the air cooled series.

The transition curves after aging for various times at

*Previously reported(1)

TABLE III

CHARPY V-NOTCH TRANSITION TEMPERATURES AND HARDNESSES OF C STEEL AGED AT VARIOUS TEMPERATURES AND TIMES AFTER WATER QUENCHING FROM 1300°F

	Transition Temperature, °F			
Aging Time	Midpoint*	50% Fibrous Fracture**	15 Ft-lbs***	Rockwell B Hardness
As-Received				
---	118	138	85	75
Air Cooled from 1300°F				
---	118	145	85	70
Aged at Room Temperature				
5 minutes	---	---	---	87
7 hours	140	175	115	88
3 days	180	212	158	94
5 days	175	210	160	96
32 days	202	230	175	96
45 days	---	---	175	96
Aged at 125°F				
1 1/4 hours	150	200	127	91
7 1/2 hours	185	215	162	93
21 hours	188	218	168	95
3 days	182	208	168	94
Aged at 350°F				
15 minutes	133	160	115	87
1 3/4 hours	125	160	100	83
19 hours	122	148	98	81
Aged at 600°F				
2 minutes	120	155	93	83
10 minutes	125	152	95	81

*Temperature at midpoint of absorbed energy-test temperature curve

**Temperature at midpoint of per cent fibrous fracture-test temperature curves.

***Temperature at 15 ft-lbs were absorbed.

NOTE: Only the aging times for impact specimens aged at room temperature include the seven hours required for machining and testing.

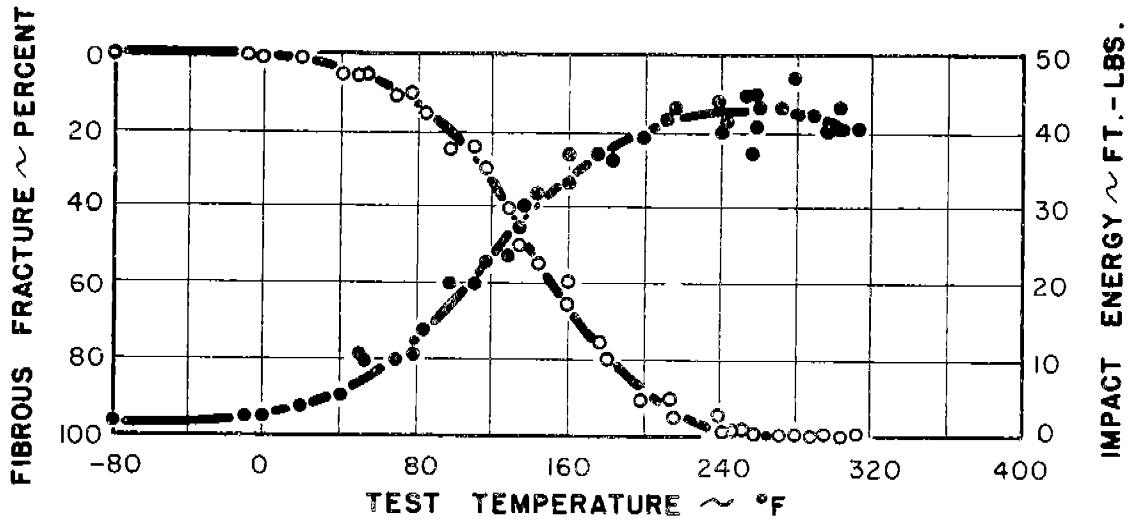


FIG. 2: CHARPY V-NOTCH TRANSITION CURVES FOR AS-RECEIVED "C" STEEL.

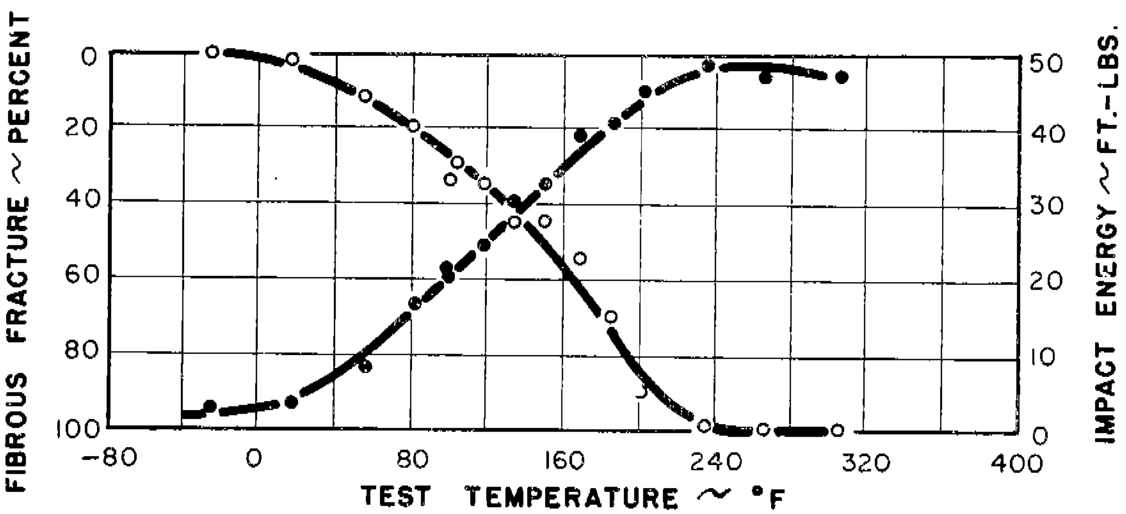


FIG. 3: CHARPY V-NOTCH TRANSITION CURVES FOR "C" STEEL SUBCRITICALLY HEATED AT 1300°F FOR 15 MINUTES AND AIR COOLED.

room temperature, 125°, 350°, and 600°F, are presented in Figs. 4--7, respectively. In all the quench-aged series the upper level of the individual absorbed energy-test temperature curves is lower than that for the air cooled series. In general, a greater difference in levels is associated with a greater degree of embrittlement, with the maximum difference being about 12 ft-lb. The degree of embrittlement can be more readily seen in the summary of the 15 ft-lb transition temperature plotted as a function of the aging time, top of Fig. 8, for the various aging temperatures employed. For comparison purposes the comparable hardness data are plotted in the bottom of this figure.

For room temperature aging (80°F), it can be seen that after seven hours aging the transition temperature (115°F) is 30°F higher than that of the air cooled series (85°F). With increasing aging time, the transition temperature increases, reaching a maximum level of 175°F at some time greater than five days but less than 32 days. This is an increase of 90°F over the unembrittled state. Although no data were obtained for aging times less than seven hours (this was the minimum time for machining and testing), it is not expected that the transition temperature would be appreciably lowered for the shorter aging times because the hardness increased but one point during the first seven hours of aging. The aging curve at the shorter aging times has been interpolated to give an 'as-quenched' transition temperature of 110°F.

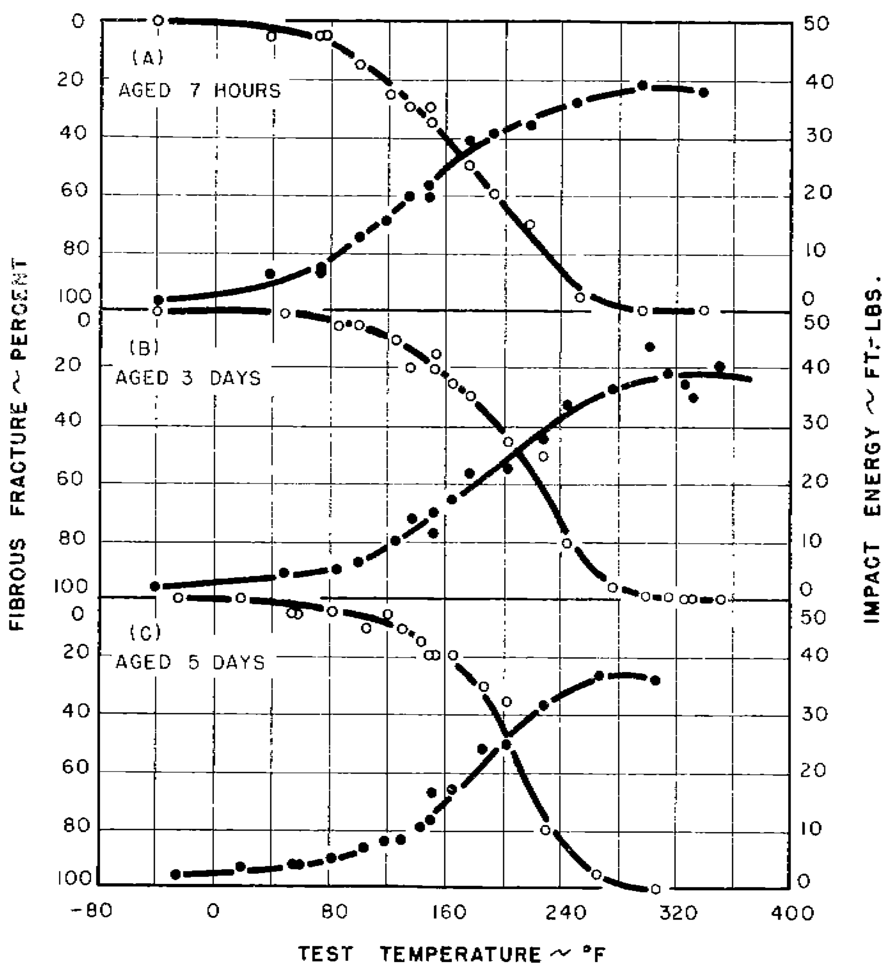


FIG. 4.(A-E): CHARPY V-NOTCHED TRANSITION CURVES FOR "C" STEEL SUBCRITICALLY HEATED AT 1300°F FOR 15 MINUTES AND WATER QUENCHED. AGED AT ROOM TEMPERATURE FOR TIMES INDICATED.

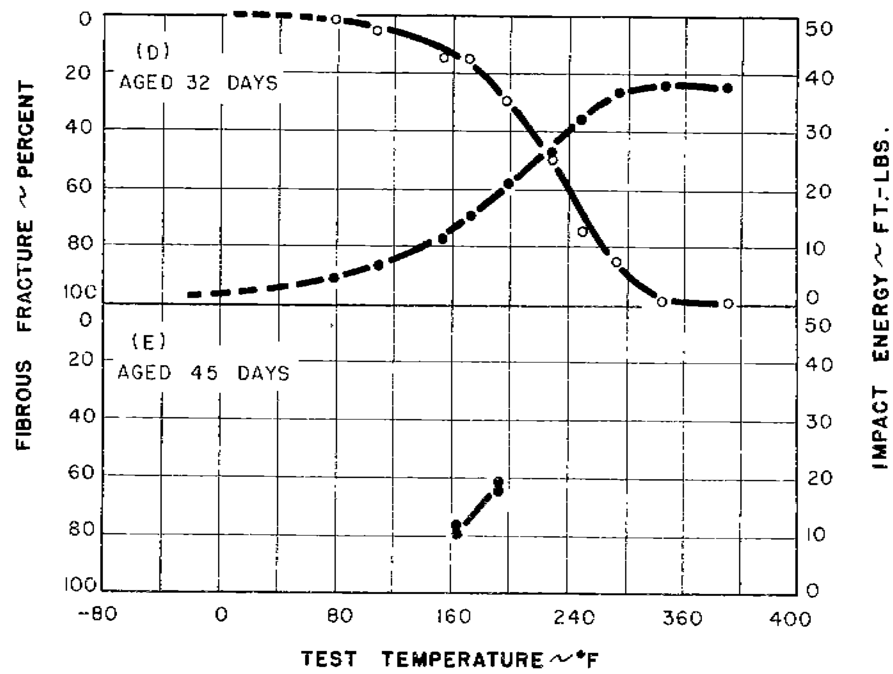


FIG. 4: CONTINUED.

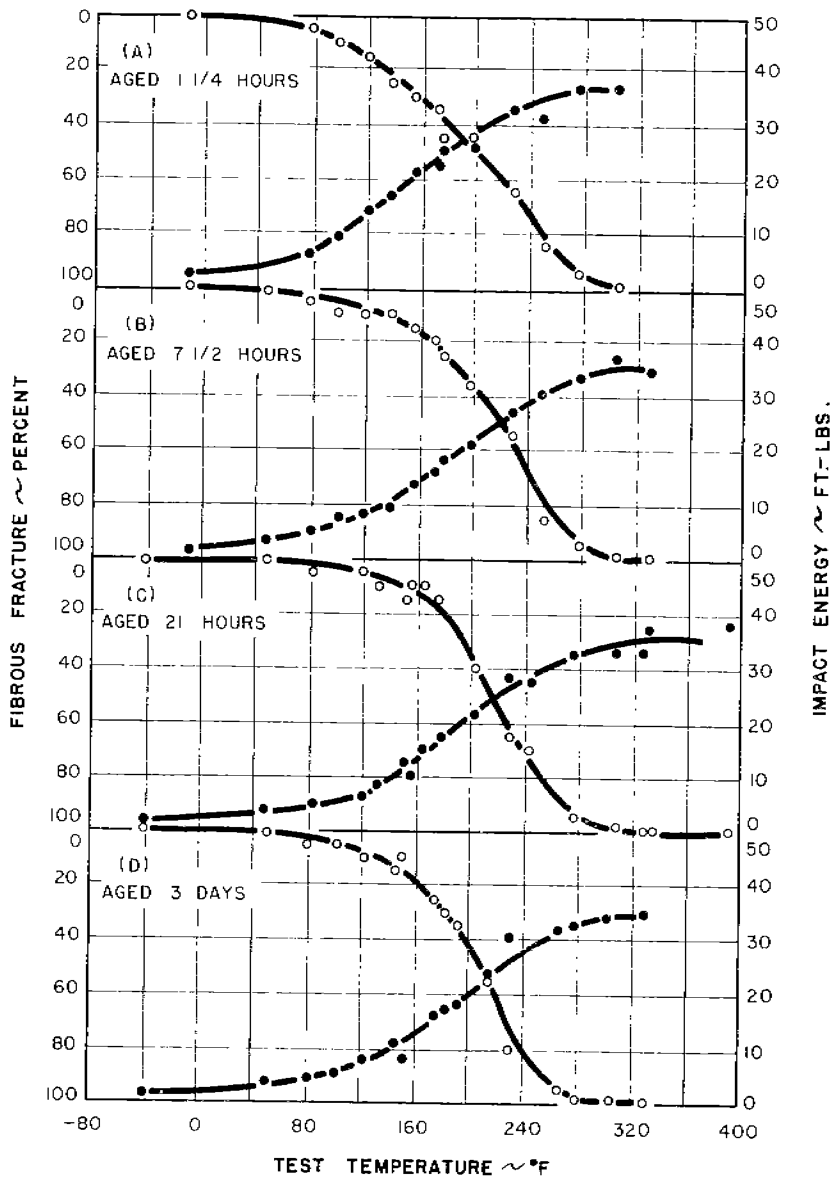


FIG. 5: (A-D) CHARPY V-NOTCHED TRANSITION CURVES FOR "C" STEEL SUBCRITICALLY HEATED AT 1300°F FOR 15 MINUTES AND WATER QUENCHED. AGED AT 125°F FOR TIMES INDICATED.

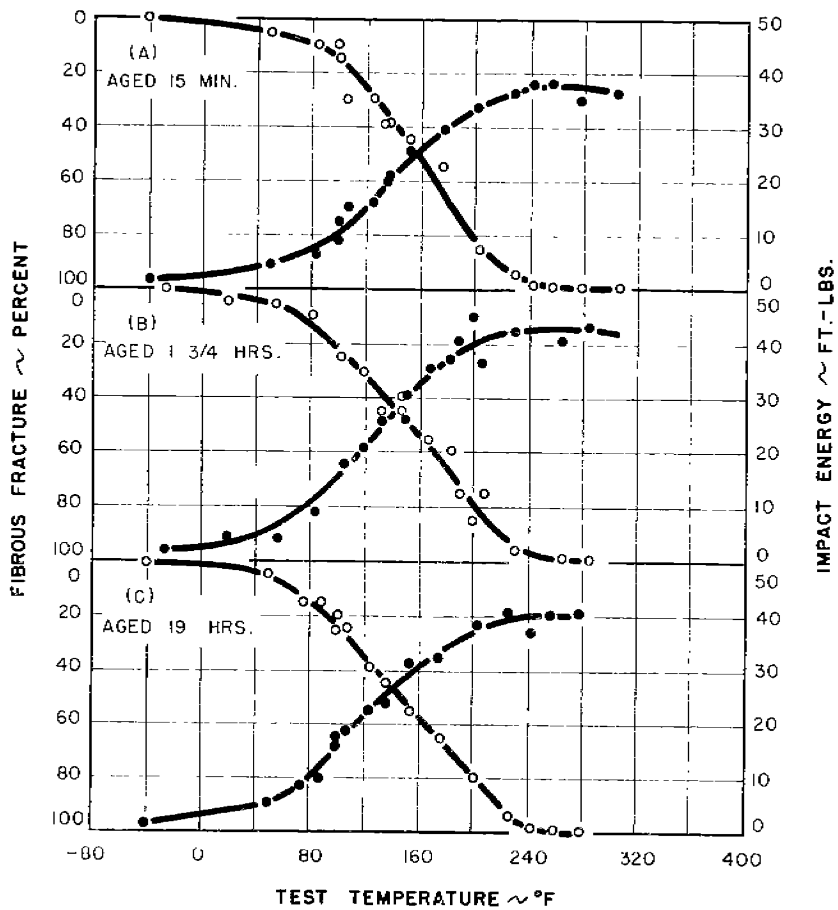


FIG. 6: (A-C) CHARPY V-NOTCHED TRANSITION CURVES FOR "C" STEEL SUBCRITICALLY HEATED AT 1300°F FOR 15 MINUTES AND WATER QUENCHED. AGED AT 350°F FOR TIMES INDICATED.

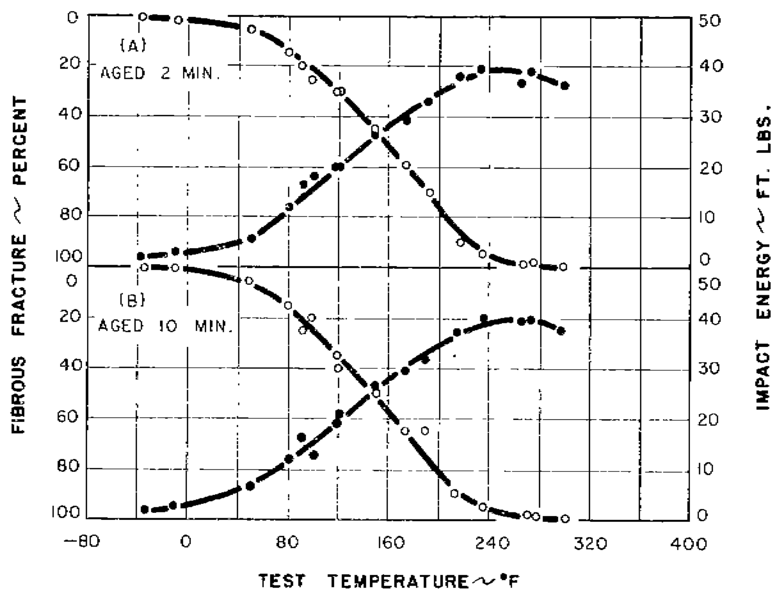


FIG. 7: (A-B) CHARPY V-NOTCHED TRANSITION CURVES FOR "C" STEEL SUBCRITICALLY HEATED AT 1300°F FOR 15 MINUTES AND WATER QUENCHED. AGED AT 600°F FOR TIMES INDICATED.

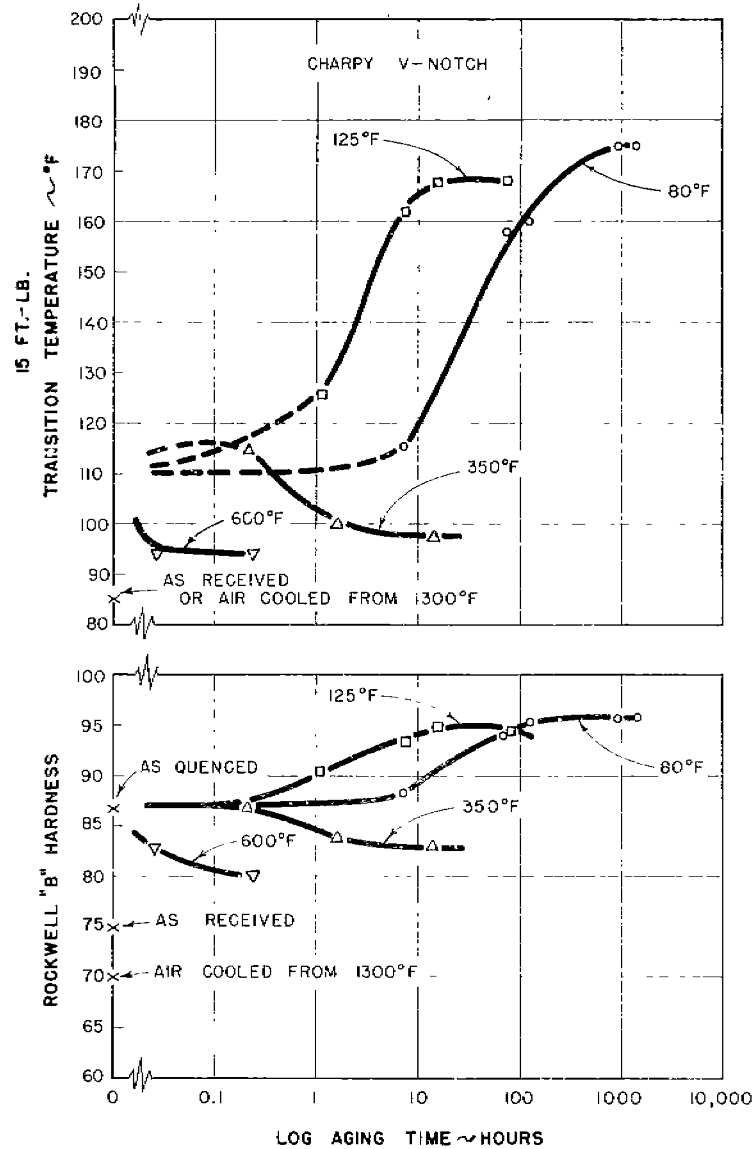


FIG. 8: SUMMARY CURVES SHOWING EFFECT OF VARIOUS AGING TEMPERATURES AND TIMES ON THE TRANSITION TEMPERATURE AND HARDNESS OF "C" STEEL. SPECIMENS WATER QUENCHED FROM 1300°F BEFORE AGING.

The transition temperature--aging time relationship for 125°F aging--shows that the transition temperature starts to increase at much shorter times, reaching a peak (168°F) after about 20 hours and remaining at this value at least up to three days aging time.

For 350°F aging it appears that a slight peak may be present at an aging time less than 15 minutes, but it probably would not be much greater than the 115°F transition temperature obtained after 15 minutes aging because the hardness was unchanged during this time interval. Aging beyond 15 minutes results in a decreasing embrittlement with time.

Aging at 600°F for two and ten minutes, respectively, effects a considerable improvement in the impact properties. For both cases the transition temperature indicated an embrittlement of but 10°F.

From the general trend of these aging curves, it is to be expected that at aging temperatures higher than 600°F the impact properties would approach those of the unembrittled state very rapidly. However, as will be pointed out in the Discussion, this conclusion only holds for specimens cooled relatively slowly from the aging temperature; a fast cool from aging temperatures above about 650° can introduce another quench-aging cycle.

Microstructures

In order to detect any microstructural changes due to quench-aging, a number of specimens representing various quench-aged conditions were examined at 2000X.

The structure of the as-received condition is shown in Fig. 9(a). No difference in structure was noted in the air cooled condition; however, immediately after quenching, a mottled ferrite was evident which did not appear to change with time at room temperature. Fig. 9(b) shows the structure after one month at room temperature. No change in this mottled structure was revealed after aging at 125°F, even after 23 days at this temperature, Fig. 9(c).

After aging 15 minutes at 350°F, the mottled ferrite was still in evidence; after 21 hours at 350°F, Fig. 9(d), the mottled structure appeared to be more intense.

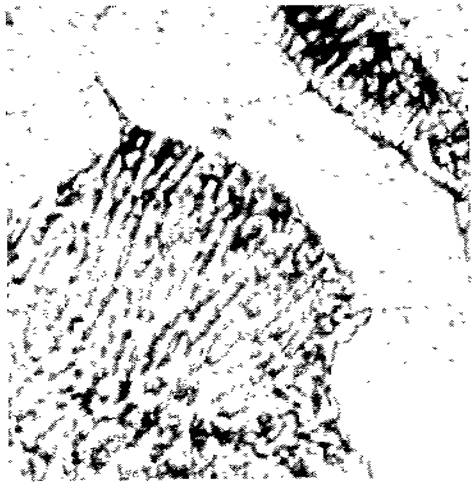
Upon aging at 600°F for two minutes, a precipitate, evenly distributed throughout the mottled ferrite grains, was resolvable. An increase in aging time to ten minutes, Fig. 9(e), resulted in better definition of the precipitated particles, which appeared to be platelets, and an increase in their size.

After aging ten minutes at 1100°F, Fig. 9(f), the precipitate is no longer evenly distributed throughout the ferrite grains but appears to have coalesced along the grain boundaries as spheroids.

DISCUSSION

On the evidence of the hardness properties, Fig. 1, aging curves were obtained which were characteristic of quench-aging

(a) As-Received



(b) Aged 30 Days at Room Temperature

(c) Aged 23 Days at 125°F.

(d) Aged 21 Hours at 350°F

(e) Aged 10 Minutes at 600°F

(f) Aged 10 Minutes at 1100°F

Fig. 9: MICROSTRUCTURES OF AS-RECEIVED AND VARIOUS QUENCH-AGED CONDITIONS. All aging treatments preceded by a water quench from 1300°F. 2000X. 3% NITAL.

systems, i.e., the maximum hardness attained and the time to reach this maximum decreased with increasing aging temperature.

The similar changes observed with the impact transition temperature, Fig. 8, are believed to be new experimental data. The only information found on the impact transition temperature changes due to quench-aging in a similar grade of steel was in the work by Low[6] and was restricted to the room temperature aging effects. In this paper it was shown that after quenching from 700°C (1290°F) the transition temperature increased continually over the aging interval of three years. The total increase amounted to a 110°F rise in Charpy keyhole transition temperature (10 ft-lb value), with the greatest increase occurring in the first ten days aging. This work was done with a 1/2-inch hot rolled semi-killed plate containing 0.17% carbon.

In conjunction with the earlier work under this project, it is now possible to show, Fig. 10, the effect of solution temperature on the magnitude of the peak transition temperature and hardness reached by room temperature aging. As is evident, increasing the solution temperature in the 1100° to 1300°F range results in a greater initial ('as-quenched') hardness and, apparently, a greater initial transition temperature. Upon subsequent room temperature aging, the magnitude of the peak also increases with increasing solution temperature. This is to be expected in line with other quench-aging systems

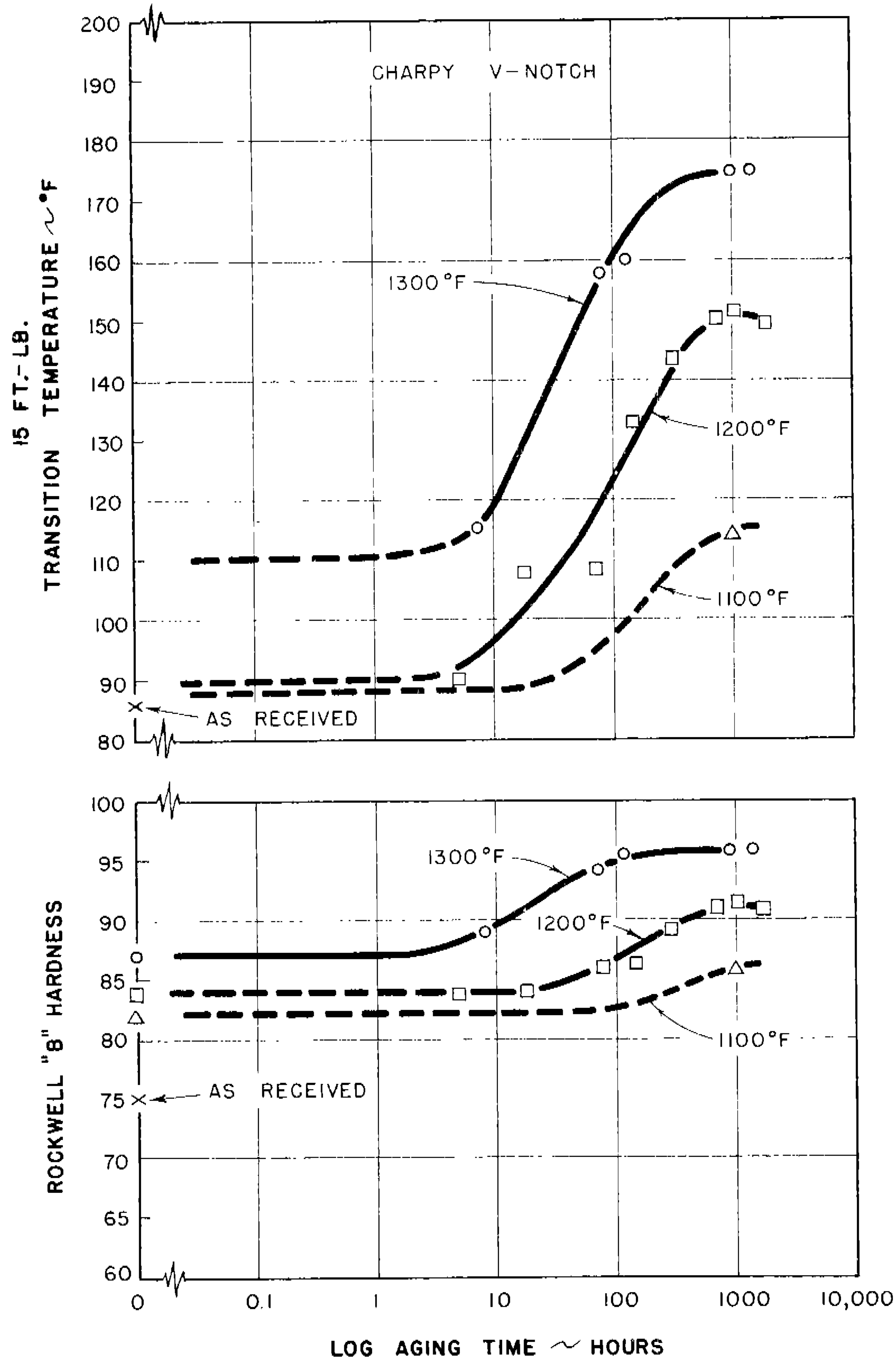


FIG. 10: EFFECT OF ROOM TEMPERATURE AGING TIME ON THE TRANSITION TEMPERATURE AND HARDNESS OF "C" STEEL AFTER WATER QUENCHING FROM TEMPERATURES INDICATED.

which show that a greater change in properties can be expected with increasing solution temperature (greater degree of supersaturation).

To obtain a quantitative measure of the solid solution and the aging effects as a function of solution temperature from room temperature to 1300°F, Fig. 11 has been prepared. Here for each property three heat treated conditions are under consideration: (1) air cooled, (2) as-quenched, and (3) quenched and aged one month at room temperature. The air cooled data sets the base line for evaluating the two effects because in this condition it is assumed that the solid solution and aging effects are absent*. The difference between the 'as-quenched' properties and the base line then yields the effect due to solid solution, whereas the difference between the 'as-quenched' and 'quench-aged' properties gives the effect due to aging. The aging effect may be altered for aging times greater than one month, but it is believed that the change would be slight, if any, and therefore the effect shown may be considered a maximum.

This figure also shows that below a solution temperature of about 650°F the solid solution and aging effects are nil but that above this temperature both effects increase with increasing solution temperature. It is interesting to note that the hardness is affected to a greater extent than the transition temperature by the solid solution effect, whereas the reverse is true

*Hardness checks showed no change with time at room temperature.

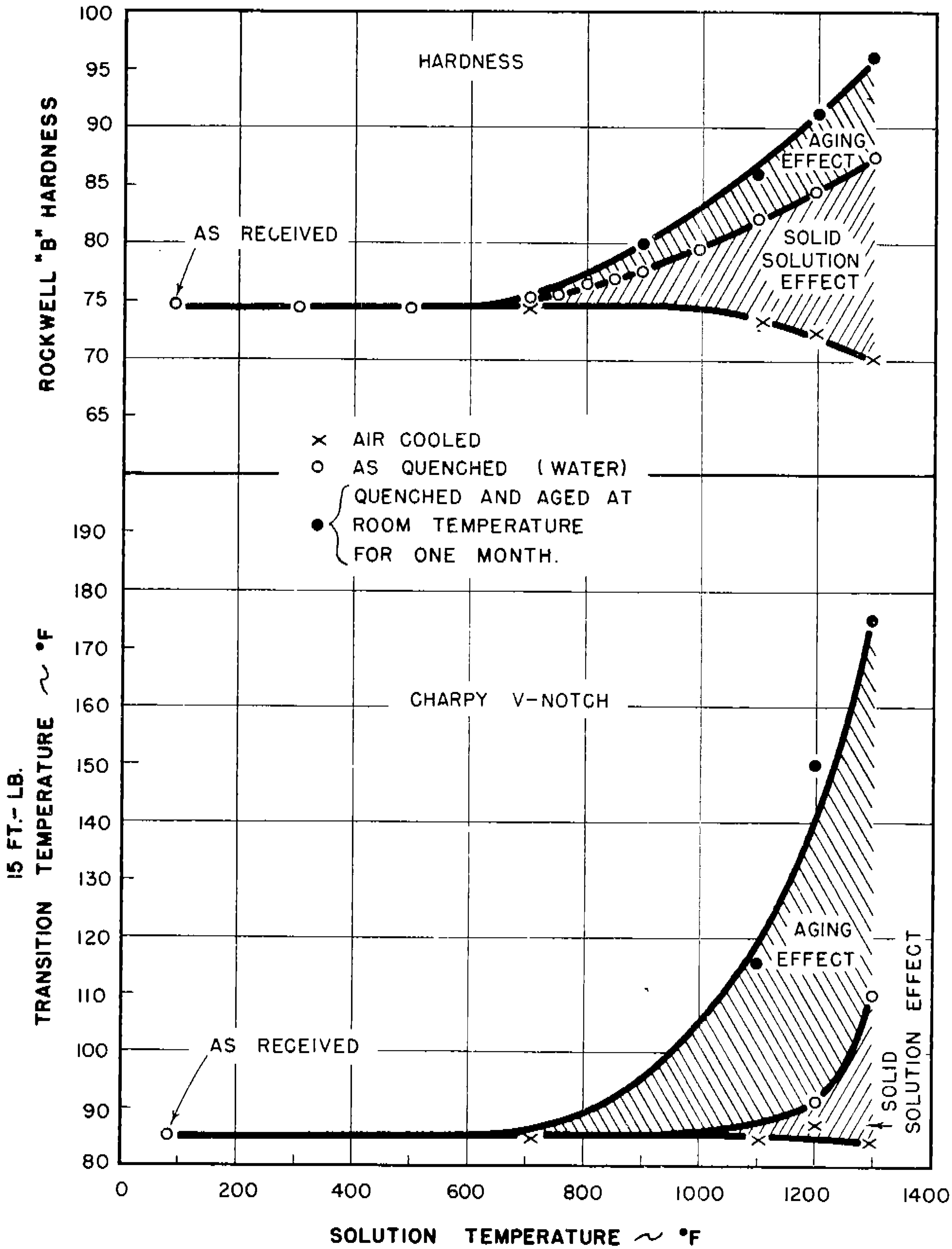


FIG. 11: SOLUTION AND ROOM TEMPERATURE AGING EFFECTS ON THE HARDNESS AND IMPACT PROPERTIES OF "C" STEEL.

with aging. In any case the two effects are interrelated and governed by the solution temperature which in turn controls the amount of carbon available for the precipitation reaction.

With cooling rates intermediate to an air cool and water quench, it is to be expected that these two effects will be minimized due to the lower degree of supersaturation, i.e., less carbon is retained in solution on the quench.

From the metallographic changes occurring during aging, it would appear that a two-stage precipitation reaction is operative. The first stage (evident as a mottling of the ferrite) was responsible for the greater degree of embrittlement, while the second stage, in which a resolvable precipitate was detected, was associated with an improvement in properties corresponding to a rapid 'overaged' condition. Although no attempt was made at identification, the sequence of microstructural changes observed appear to be the same as reported recently (7). In this paper the metallographic changes occurring during the quench-aging of an ingot iron (0.026% carbon) were followed by means of the optical and electron microscopes, and the identification of the two-stage precipitate made with electron diffraction technique. The first stage was identified as the hexagonal ε-iron carbide after 15 hours at 200°C (390°F). The second stage was associated with a change in the crystal lattice to cementite, (orthorhombic Fe_3C), observed after aging 30 minutes at 300°C (570°F).

It can now be speculated that in the aging of C steel the formation of the transition lattice (ϵ-iron carbide) is associated with a high degree of embrittlement, while subsequent formation of the equilibrium phase (cementite) brings about a large improvement in properties, corresponding to the rapid 'overaged' condition.

Insofar as microstructural changes are concerned, the brittle zone previously found in the subcritically heated region in ship plate weldments has not been identified with a precipitation reaction*. However, due to the complexity of the temperature, time at temperature, and cooling rate conditions in multiple pass weldments, it is difficult to determine the exact quench-aging cycle experienced. It has been shown previously (1) that the cooling rate in the subcritically heated region** is such that the ferrite is not supersaturated to the maximum possible as in water quenched test specimens. Consequently, less carbon is available for the subsequent precipitation reaction, and the resultant precipitate may not be detected***.

In regard to the changes in hardness and transition temperature, the results of this investigation give the maximum possible

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*It should be recalled that the demonstrated beneficial effects of increasing preheat (decreasing cooling rate) and postheating ('overaging' certainly point to the quench-aging mechanism.

**The cooling rate at the critical zone was found to be intermediate to water quenched and air cooled impact test specimens.

***It is not expected that the electron microscope would be able to detect any precipitation. In the reference cited in the Discussion (7), the only advantage of the electron over the light microscope was in better resolution of the precipitate once it had formed.

embrittlement of C steel by quench-aging, Fig. 11. Any treatment designed to remove these effects in the base plate should be applicable to weldments made of this and similar grades of steel. As was shown in Figs. 1 and 8, 'overaging' treatments served to minimize the embrittlement, with the degree of embrittlement decreasing with increasing aging temperature. However, this finding must be qualified in view of the fact that the results were based on specimens air cooled from the aging temperatures. A consideration of Fig. 11 shows that if aging temperatures above 650°F were used, then the possibility exists that another quench-aging cycle could be initiated if a fast cool were employed. (Low[6], using a different approach, arrived at the same conclusion). Therefore, the postheating of weldments susceptible to quench-aging should be restricted to a maximum of about 650°F when this danger exists. As this investigation showed, aging in this neighborhood (600°F) effected a rapid and considerable improvement in the ductility by 'overaging'.

CONCLUSIONS

In conjunction with previous work[1] on the subcritical heat treatment of a low carbon ship plate steel (C steel), the following conclusions seem justified:

1. The quench-aging mechanism was responsible for the loss in ductility and the increase in hardness.

2. The severity of the embrittlement increased with increasing solution temperature from 650° to 1300°F; no embrittlement was present when employing solution temperatures less than 650°F.

3. On the evidence of impact transition temperature and hardness changes, characteristic aging curves were obtained, i.e., the peak transition temperature and hardness reached and the time to attain this peak decreased with increasing aging temperature.

4. Metallographic examination showed that a two-stage precipitation reaction was operative.

5. A low temperature postheat at about 650°F would do much to eliminate by overaging the zone of minimum ductility previously found in the subcritically heated region of weldments of this and similar grades of steel. Higher postheat temperatures run the danger of introducing another aging cycle if a fast cool is employed.

FUTURE WORK

This report concludes the experimental work done under this contract. A final report will be prepared summarizing all the work done on (1) the distribution of relative ductility in ship plate weldments and (2) the subcritical heat treatment of base plate.

BIBLIOGRAPHY

1. E. B. Evans, and L. J. Klingler, "The Effect of Subcritical Heat Treatment on the Transition Temperature of a Low Carbon Ship Plate Steel", Fourth Progress Report, Ship Structure Committee Serial No. SSC-60, October 30, 1953.

2. G. Sachs, L. J. Ebert, and A. W. Dana, Jr., "The Fundamental Factors Influencing the Behavior of Welded Structures Under Conditions of Multiaxial Stress and Variations of Temperature, Stress Concentration, and Rates of Strain", Navy Department, Bureau of Ships, Contract NObs-45470, Serial No. SSC-24, May 10, 1949.

3. L. J. Klingler, L. J. Ebert and W. M. Baldwin, Jr., Ibid, 2, Serial No. SSC-34, November 28, 1949.

4. E. B. Evans and L. J. Klingler, Ibid 2, Serial No. SSC-54, October 14, 1952.

5. Technical Progress Report of the Ship Structure Committee, Welding Journal, Vol. 13 (July 1948), pp. 377s-384s.

6. J. R. Low, Jr., "The Effect of Quench-Aging on the Notch Sensitivity of Steel", Welding Research Supplement, Vol. 17 (May 1952), p. 253s-256s.

7. Anna L. Tsou, J. Nutting, and J. W. Menter, "The Quench-Aging of Iron", Journal of the Iron and Steel Institute, Vol. 172, Part 2, October, 1952, pp. 163-171.