SSC-93

MECHANICAL PROPERTIES OF HIGH PURITY Fe-C ALLOYS AT LOW TEMPERATURES

by

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and

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

March 31, 1959

Dear Sir:

In order to correlate brittle behavior with the material used in the hull structures of ships, the Ship Structure Committee has sponsored a study of yield point and fracture data in uniaxial tension of high purity iron-carbon alloys. Herewith is a copy of SSC-93, Third Progress Report, entitled "Mechanical Properties of High Purity Iron-Carbon Alloys at Low Temperatures," by G. J. London, G. Spangler and R. M. Brick.

The Committee on Ship Steel of the National Academy of Sciences-National Research Council has provided the advisory assistance to the Ship Structure Committee in this investigation.

This report is being distributed to those individuals and agencies associated with and interested in the work of the Ship Structure Committee. Any questions, comments, criticism or other matters pertaining to the report should be addressed to the Secretary, Ship Structure Committee.

Yours sincerely,

E. H. Thiele
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

Serial No. SSC-93

Third Progress Report
of
Project SR-109

to the

SHIP STRUCTURE COMMITTEE

on

MECHANICAL PROPERTIES OF HIGH PURITY Fe-C ALLOYS AT LOW TEMPERATURES

by

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University of Pennsylvania
Philadelphia, Pennsylvania

under

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Washington, D. C.
National Academy of Sciences-National Research Council
March 31, 1959

ABSTRACT

Results of the second phase of an investigation on several high-purity, low-carbon irons in the α region of the Fe-C diagram which was made in an attempt to correlate brittle behavior, as defined by the Charpy V-notch transition temperature, with yield-point and fracture data from uniaxial tension tests are summarized in this report.

All of the alloys tested showed a sharp rise in impact transition temperature corresponding to the precipitation of Fe_3C. The data also indicate that ferrites with veining have higher transition temperatures than those without veining as quenched from the α or unsaturated region.

Two of the alloys show a decrease in the transition temperature of the ferritic structures as saturation is approached. Two hypotheses are advanced to explain this: 1) as the temperature is lowered in the α region, the concentration of carbon atoms at grain boundaries increases, and 2) carbon segregates at sub-grain as well as at primary boundaries and in some way "strengthens" the sub-grain boundaries so that they assume more of the properties of high-angle grain boundaries.

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INTRODUCTION

Griffith[1] proposed a theory of brittle fracture based on the hypothesis that microscopic cracks of a definite size characteristic of the material existed in brittle material. His computation showed that at a critical value of normal stress (σ_c), a brittle fracture propagated from one of these microcracks. The equation presented in his work is as follows:

$$\sigma_c = \sqrt{\frac{EW}{c}}$$

where E is Young's modulus, W is the energy required to form a unit area of new surface and c is the microcrack radius if the crack is external (bounded by surface) or twice the crack length if the crack is internal.

This equation relates a normal stress and a crack length. This means that a brittle material with microcracks of a certain size fractures when sufficient normal stress is applied. Non-uniform stress distribution can exist so that the fracture does not propagate necessarily from the largest microcrack present, but rather from the largest, most highly stressed one.

Griffith's hypothesis was successfully applied to completely brittle materials (such as glass). It can be used to predict a microcrack radius of several microns in glass and to explain the abrupt rise in the fracture strength of glass fibers approximating this size: a transverse crack of length 3μ cannot exist in a glass fiber of diameter 2μ. However, when applied to the brittle fracture of metals, this theory did not provide a complete explanation for a size effect in metal fibers. The actual size effect in metals occurred at much larger fiber diameters (of the order of 2000μ for mild steel). It seemed as though Griffith's equation were missing some important factor. Orowan[2] suggested that this factor was the small amount of surface deformation that occurred before the propagation of a brittle fracture. The surface deformation factor was estimated to be about 1000 times larger than the interfacial energy term W; thus W can be neglected and the Orowan modification of the Griffith equation becomes

$$\sigma_c = \sqrt{\frac{Ep}{b}}$$

where p is the surface cold work factor and b is the new critical crack radius based on p.

This revised equation described a size effect in mild steel of the order of 2000 to 3000μ in size. This was found to be the case by many investigators.

It followed from this new equation that an analysis of the notched bar impact test was possible. In the notched bar impact test, the metal at the base of the notch is placed under high biaxial and triaxial tension resulting from the restraint imposed by the notch. The tensile stress at the root of the notch can thus exceed the yield stress (i.e. the yield stress in uniaxial tension) because of the biaxial and triaxial stress conditions. It has been shown by Orowan[3], however, that the local normal stress cannot exceed the yield stress of the material by more than a factor of 2 or 3 before yielding occurs. This limits the stress somewhat. Thus the maximum normal stress present in the test specimen near the base of the notch can be some multiple of the yield stress. The yield stress of mild steel can in turn be raised by a factor of 2 or 3 if the loading rate is sufficiently high, which permits a further increase in the maximum local tensile stress. In other words, the maximum possible tensile stress is proportional to the yield stress of the material in simple tension. Thus, if the yield stress is not above a certain minimum value, or if the plastic constraint or loading rate is insufficient, the stress required for cleavage fracture will never be reached in the test and the material will be ductile.

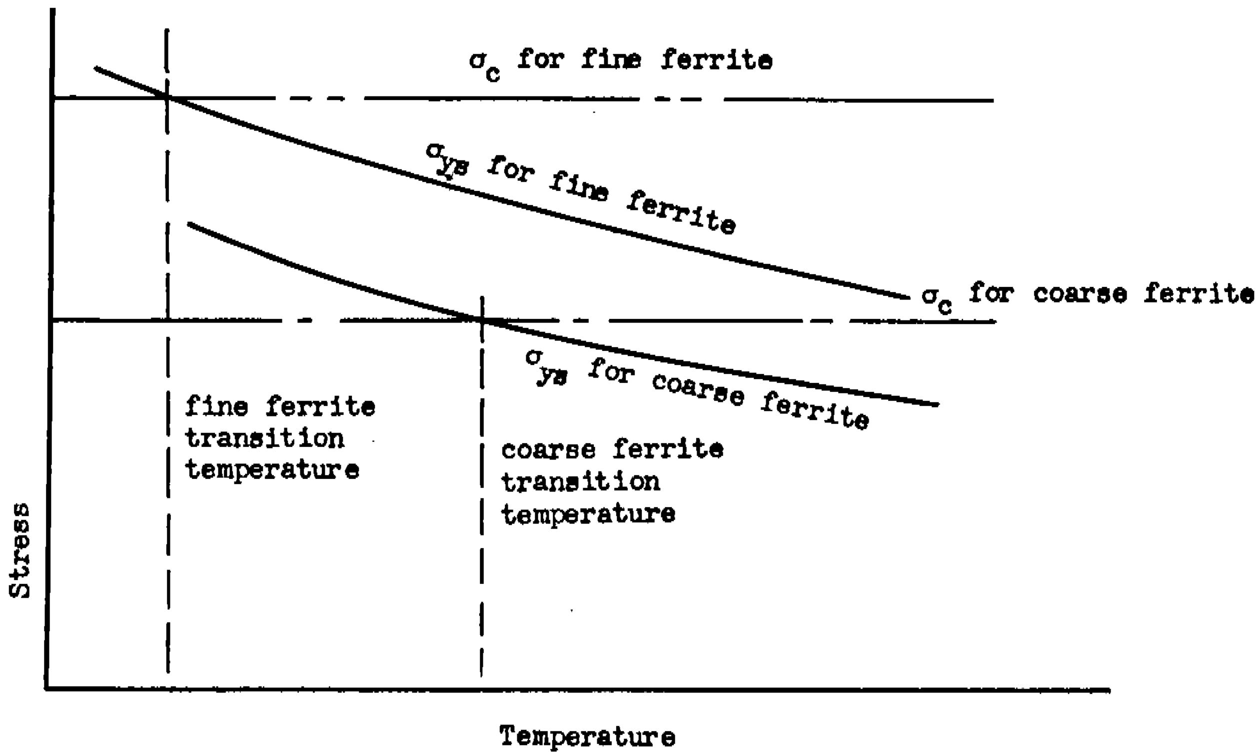
The Griffith-Orowan equation can thus be used to describe ductile-to-brittle transition in mild steel. As the temperature of testing decreases, the yield stress of the material approaches a level which is sufficient to provide the critical stress necessary to propagate a brittle fracture. The temperature at which this transition occurs is termed the "transition temperature" of the metal, since below this temperature the critical normal stress is always attained in testing before yielding can occur. Thus the metal breaks in a brittle manner with a very low energy absorption.

Gradual ductile-to-brittle transitions are explained by the necessity for some plastic deformation to precede the initiation of brittle fracture. This plastic deformation increases the maximum local normal stress to a value at which a brittle fracture can propagate. Thus the metal deforms to some extent and absorbs considerable energy before fracture. With progressively decreasing temperatures less deformation is necessary and therefore less energy is absorbed until a temperature is reached at which the material is completely brittle.

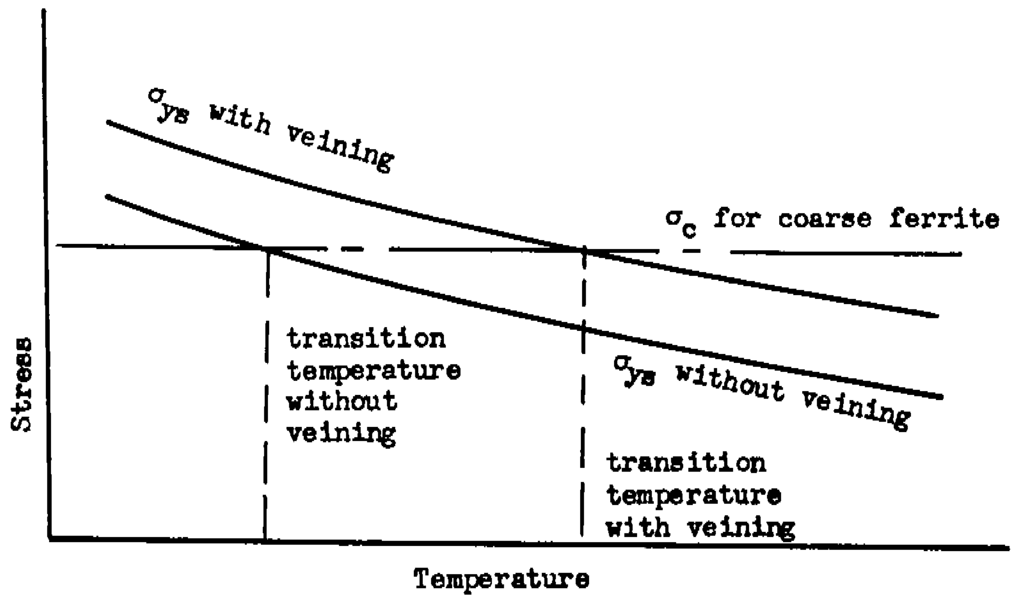
It is known that Charpy V-notch transition temperatures increase with the increase of the ferrite grain size of mild steel. However, both the yield stress and the ductility of mild steel increase with decreasing ferrite grain size. It may be concluded that the influence of increased yield stress in raising transition temperature of steel of smaller grain size is less than the influence of reduced crack length in lowering transition temperature. Since fine-grained ferrites fracture at high normal stresses, they must have inherently smaller crack lengths than coarse-grained ferrites, assuming the Griffith-Orowan theory applies here. Fig. 1 shows a schematic diagram of this behavior.

Work by Wert[4] has shown that the yield strength of iron-carbon alloys in a carbon range of 0 to 0.01% rises very sharply with increasing carbon in solution. Applying the Orowan modification of the Griffith crack equation to this information, it may be presumed that, associated with this large rise in yield stress, there is a sharp increase in impact transition temperature. Hence, increased carbon in solution in ferrite can be considered to raise transition temperatures as a result of this yield strength effect.

This report presents data pertinent to the relationships suggested by the foregoing descriptions of yield strength, fracture and impact transition tempera-



Fine and coarse grained ferrite



Coarse ferrite with and without veining

Figure 1. Schematic graphs showing variation with temperature of stress for cracking and for plastic deformation of different ferrites.

ture. The impact transition behavior of high purity iron-carbon alloys in the α region and in the α plus precipitated carbide region of the Fe-C phase diagrams has been determined. Reference is made to effects of quenching from above, at and below temperatures where iron carbide will precipitate in alloys with and without veining or substructures in the ferrite. Some yield strength and fracture measurements have been made on one of the alloys tested. These data have been correlated with the impact behavior of the alloy.

METHOD AND PROCEDURE

To investigate the problem of low temperature brittleness of mild steel, several high-purity low-carbon alloys were made. Electrolytic iron was melted in air, sulfur was removed with a lime slag, and the material was cast into ingots. Sugar carbon was added to a separate quantity of this air-melted material to produce a 0.2 to 0.6% carbon alloy. Appropriate portions of both the air-melted (oxygen saturated) material and carbon-melted material were then used in the vacuum melting of the final alloys. The iron was held molten until the carbon-oxygen reaction was essentially complete, as evidenced by solidification pressures of 1 micron. Alloys 93V, 98V, 101V and 105V were produced in this manner. Table I shows the composition of the alloys discussed in this paper.

The melts were sectioned into four pieces and hot forged to 7/8 in. square. The alloys were then cold-rolled, with appropriate intermediate re-crystallizations, to 0.450 in. square with grain sizes appropriate for testing. Standard Charpy bars were machined from this stock without, however, being notched. Groups of impact specimens were then heat-treated (as specified in Tables II and III), notched and tested immediately. During the time elapsed between the heat-treating and the testing, the bars were kept at dry-ice temperature except for the 1/2 hour consumed in notching. Slow machining speeds and feeds were used to prevent heating during the notching operation.

Table I - Composition of Iron-Carbon Alloys

		% C	% 0**	% N**	% H**
93V		.0027*	(.0017)	(.0017)	$.0000_6$
98V		.0059*	.0006	.0017	$.0001_3$
101V	Approx.	.0160	(.0056)	(.0012)	$.0000_4$
86V		.0182*	.0010	.0006	$.0000_2$
105V	Approx.	.0190	(.0013)	(.0016)	nil

*Average of three analyses by Bell Telephone Laboratories

**Average of two analyses by National Bureau of Standards

( )Single analysis by National Bureau of Standards run at one time.

Individual specimens were held at test temperature for 10 minutes prior to testing. They were then broken in a standard 120 ft-lb capacity Sonntag testing machine within 5 seconds of removal from the bath.

A sub-size Charpy V-notch test specimen was fabricated by cold rolling the broken halves of the standard Charpy specimens to 0.270 in. square, recrystallizing, and machining a specimen of standard length, 0.250 in. square with a notch 0.50 in. deep. These dimensions produce in the sub-size specimens the same ratio of depth of notch to depth of metal under the notch as exists in the standard Charpy test bar. Even though transition temperatures for the sub-size specimens were not identical to those for the standard Charpy specimens, the response to heat-treatment was identical, and the amount of information that could be obtained from a given quantity of material was tripled.

Tensile specimens were made of Alloy 101V by cold-rolling halves of standard Charpy bars into 0.275-in. diameter bars, recrystallizing, and machining tensile

TABLE II

Processing and Transition Temperature Data
for Alloys Tested as Full-Size Charpy V-notch Bars

Final Heat Treatment	% C Precipitated	Rockwell Hardness	ASTM Grain Size	Transition[1] Temperature	Corrected[2] Transition Temperature	Remarks
A - Alloy 93V; .0027 C						
1/2 hr. 700 C, quench	0	B22	4.3	-56 C		(All specimens of 93V above cold rolled 47%, annealed 1 hour in salt at 700 C, cold rolled 47% additional, annealed 1 hour in salt at 700 C and air cooled; then machined and treated in neutral salt or vacuum as above.)
2/3 hr. 538 C, quench	0	16	4.3	-48 C		
6 1/2 hr. 400 C, quench	.0009	21	4.8	-52 C		
1 hr. 700 C, furn. cool	.002	10	4.8	-35 C		
B - Alloy 98V; .0059 C						
a. 1 hr. 700 C, air cool	.001	B15	2.2	-25 C	-30 C	(Specimens of 98V, a & c, 3/4 hour 930 C furnace cooled to 700 C, hold 1 hour (in vacuum) then air cooled. Specimens b, d & e; same, reheated to 1000 C in vacuum for 1 hour, furnace cooled to 700 C, hold 1 hour, then air cooled.)
b. 1 hr. 700 C, air cool	.001	14	1.2	-31 C	-50 C	
c. 1/3 hr. 756 C, quench	0	30	2.7	+56 C	+58 C	
d. 1 hr. 700 C, quench	0	27	1.2	+46 C	+26 C	
e. 1 hr. 700 C, furn. cool	.005	5	1.4	+13 C	- 9 C	
C - Alloy 86V; .0182 C						
75% c.r., 1 hr 600 C, air cool	.013	B36	6.9	-12 C	+53 C	(All specimens heated to 1000 C in vacuum and furnace cooled prior to cold work.)
28% c.r., 24 hr 700 C, air cool	.013	30	3.9	+36 C	+56 C	
28% c.r., 24 hr 700 C, quench	.001	56	3.9	-24 C	- 5 C	
2/3 hr. 930 C, furn. cool	.018	20	4.4	+47 C	+76 C	
1/2 hr. 930 C, furn. cool to 840 C, quench	.018	56	4.1	-29 C	- 6 C	
D - Alloy 101V; Approx. .016 C						
a. 1/2 hr. 723 C quench		B48	3.6	-16 C		a thru d 47% RA, heated 900 C 1/2 hr., furnace cooled to 723 C, hold 1 hr. (in vacuum) air cooled. e thru g 47% RA, heated 723 C 4 hours in salt, air cooled.
b. 1/2 hr. 690 C quench		44	4.0	-33 C		
c. 1/2 hr. 651 C quench		41	3.7	+ 7 C		
d. 1/2 hr. 600 C quench		38	3.2	+13 C		
e. 1/2 hr. 723 C quench		55	6.8	-62 C		
f. 1/2 hr. 690 C quench		53	7.4	-63 C		
g. 1/2 hr. 650 C quench		52	7.2	-33 C		
E - Alloy 105V; Approx. .019 C						
a. 1/2 hr. 723 C quench		53	3.2	+44 C		a thru e 47% RA, heated 930 C 3/4 hr., furnace cooled to 723 C, hold 1 hr. (in vacuum), air cooled. f thru i 47% RA, heated 723 C 2 hrs. in salt, air cooled. j - Given subcritical treatment then supercritical treatment.
b. 1/2 hr. 723 C quench		53	3.0	+31 C		
c. 1/2 hr. 689 C quench		51	3.1	+33 C		
d. 1/2 hr. 620 C quench		42	3.2	+58 C		
e. 1/2 hr. 650 C quench		50	3.1	+62 C		
f. 1/2 hr. 723 C quench		63	7.3	-63 C		
g. 1/2 hr. 691 C quench		62	Approx. 7.3	-31 C		
h. 1/2 hr. 648 C quench		56	" 7.3	-12 C		
i. 1/2 hr. 599 C quench		47	" 7.3	- 2 C		
j. 1/2 hr. 723 C quench		39	" 2.5	+10 C		

1. Vertical part of energy-temperature curve.
2. Corrected to an ASTM No. 2 1/2 grain size on basis of 15 C per ASTM No.

TABLE III

Processing and Transition Temperature Data
for Alloys Tested as Sub-Size Charpy V-notch Bars

Final Heat Treatment	% C Precipitated	Rockwell Hardness	ASTM Grain Size	Transition[1] Temperature	Corrected[2] Transition Temperature	Remarks
A - Alloy 98V; .0059 C						
1/2 hr. 700 C, air cool	.001	B12	1.2	-45 C	-63 C	(Broken ends of full size Charpy bars cold rolled 50%, heated 1 hour at 1000 C in vacuum, furnace cooled to 700 C, held 1 hour, then air cooled. Above heat treatments all performed in neutral salt.)
1/2 hr. 700 C, quench	0	26	1.6	+101 C	+87 C	
1/2 hr. 680 C, quench	0	18	1.2	(105 C)	+86 C	
1/2 hr. 650 C, quench	0	15	1.2	+106 C	+87 C	
1/2 hr. 602 C, quench	0	17	0.7	+82 C	+55 C	
40 min. 562 C, quench	(sat. C + N)	14	1.2	-15 C	-34 C	
40 min. 532 C, quench	(sat. C)	14	1.2	-13 C	-32 C	
40 min. 499 C, quench	.0015	15	1.2	+22 C	+ 2 C	
60 min. 403 C, quench	.0041	8	0.7	+25 C	- 2 C	
B - Alloy 86V; .0182 C						
60 min. 700 C, air cool	.013	B27	3.8	+53 C	+72 C	(Broken ends of full size 86V Charpy bars cold rolled 50%, heated in vacuum 45 minutes at 930 C, furnace cooled to 700 C, held 1 hour, then air cooled. Above heat treatments performed in neutral salt.)
20 min. 700 C, quench	.001	45	3.9	-39 C	-29 C	
20 min. 723 C, quench	0	57	3.6	-33 C	-16 C	
30 min. 678 C, quench	.003	54	3.8	+10 C	+29 C	
3 min. 640 C, quench	.007	44	3.6	+42 C	+58 C	
3 min. 640 C, quench aged 8 days room temperature	.007	49	3.6	+48 C	+64 C	

1. Vertical part of energy - temperature curve
2. Corrected to an ASTM No. 2 1/2 grain size on basis of 15 C per ASTM No.

specimens with 0.190 in. gage diameter. The recrystallization treatment involved maintaining the specimen at 930 C for 3/4 hour, furnace cooling to 723 C, holding 1 hour, and then air cooling. Fifteen specimens were then reheated in salt to 723 C for 1/2 hour and quenched, while the other fifteen were reheated to 690 C for 1/2 hour and then quenched. The specimens were kept in a cold bath until tested, except for about twenty minutes during which the gage sections of the bars were wet polished to remove the scale formed during heat-treatment. Five bars from each group were tested at 0°C, -77 C and -194 C in an Instron constant-strain-rate tensile testing machine.

Two basic recrystallization procedures were used in this investigation. The first is termed a supercritical treatment, in that the iron is heated into the γ region during recrystallization. The second is a subcritical treatment in which the α iron is never transformed into γ iron. Cold-rolling and recrystallizing subcritically eliminates all traces of any previous supercritical heat-treatment. The treatments used were as follows:

<u>Subcritical</u>

93V	47% R A followed by 700 C in salt 1 hr A C
98V	None
101V	47% R A followed by 723 C in salt 4 hr A C
105V	47% R A followed by 723 C in salt 2 hr A C

<u>Supercritical</u>

93V	47% R A - 930 C 1/2 hr Q
98V	47% R A (a) 930 C 3/4 hr - F C to 700 C - held 1 hr A C (b) first (a) then 1000 C 1 hr - F C to 700 C - hold 1 hr - A C
101V	47% R A - 900 C 1/2 hr - F C to 700 C - hold 1 hr - A C
105V	47% R A - 930 C 3/4 hr - F C to 723 C - hold 1 hr - A C

The supercritical treatments were performed in a horizontal vacuum annealing furnace. Argon was used to break the vacuum.

RESULTS

<u>Alloy</u> <u>86V</u> - (0.0182% C): Standard Charpy tests were used to determine the effect of the cooling rate on transition temperature. The 700 C quench treatment

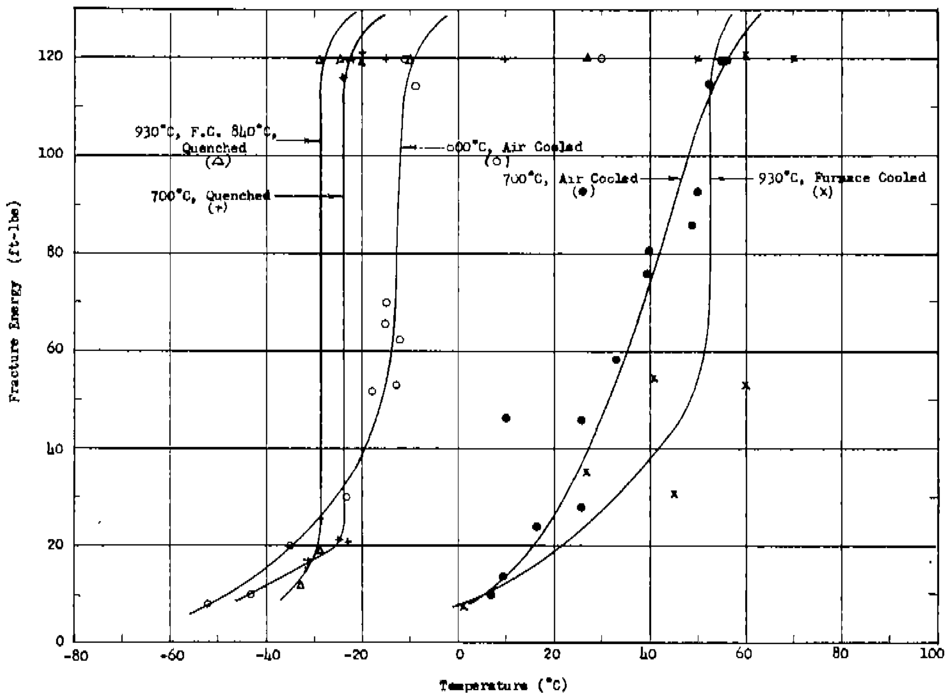


Figure 2. Alloy 86V - .0182% C Energy to Fracture of Charpy V-Notch Bars as a Function of Temperature.

resulted in the lowest transition temperature (this will be discussed later). Results from the 700 C quench tests show a vertical drop in energy with decrease in testing temperature, as can be seen in Fig. 2. However, air-cooled specimens from this temperature show a gradual decrease in energy with decreased temperature. The gradual decreased is thought to result from the Fe_3C precipitate in the air-cooled specimens, which tends to hinder the propagation of a brittle fracture and thus to permit a higher energy absorption.

One group of tests was conducted on material furnace-cooled from 930 C to 840 C. The transition temperature was quite low. The structure at 840 C was ferrite with some austenite at ferrite grain boundaries. On quenching, this austenite was transformed partly to martensite and partly to pearlite.

Sub-size bar impact tests were used to investigate the effect of variation in quenching temperatures (Fig. 3). These data show a steady

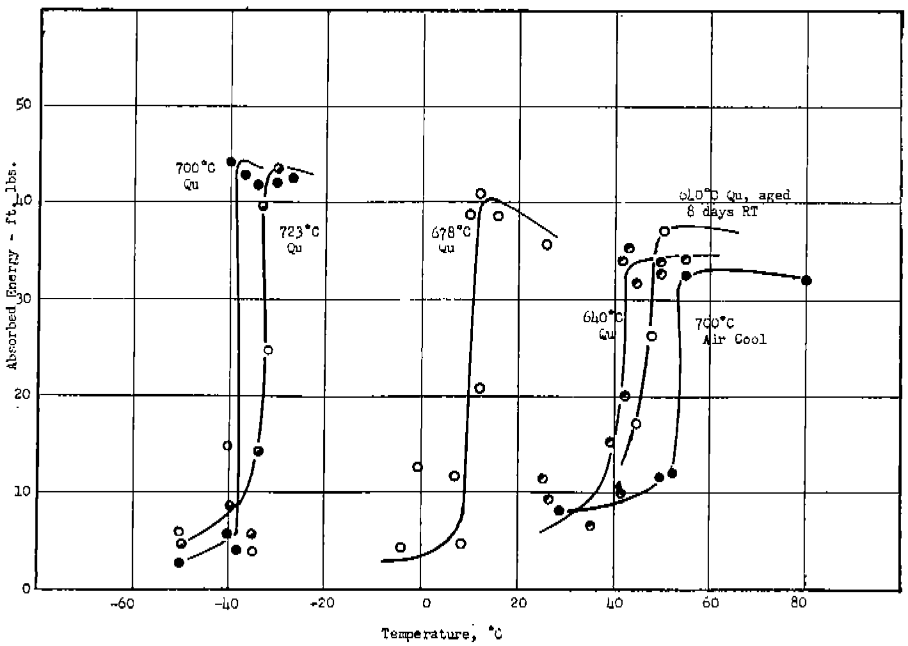


Figure 3. Alloy 86V, .0182% C, Sub-size Charpy V-Notch Bars.

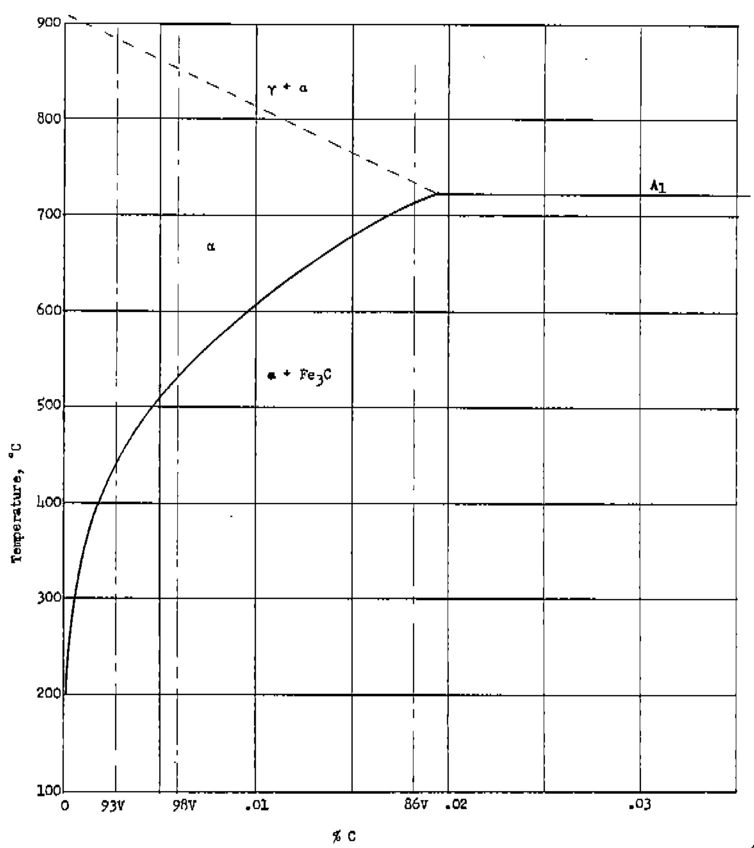


Figure 4. Solubility of Carbon in α Fe from Wert[4]

rise in transition temperature with a decrease in quenching temperature below 700 C. However, quenching from 723 C resulted in no further decrease in transition temperature.

Work done by Wert[4] indicates that the solubility of Fe_3C in ferrite follows the equation

$$C = 2.55\, e^{\frac{-9700}{RT}}$$

where C is the per cent carbon in solution, R is the gas constant, and T is the absolute temperature.

Figure 4 is a plot of this equation. It can be seen from Fig. 4 that the saturation temperature for Alloy 86V (0.0182% C) is about 715 C.

<u>Alloy 93V</u> (0.0027% C): Results from standard Charpy tests indicated no change in the transition temperature for specimens when quenched down to a temperature of 400 C. However, precipitation of carbide resulted in a distinct rise in transition temperature as indicated by the furnace-cooled test results plotted in Fig. 5.

The saturation temperature for this alloy (0.0027% C) is approximately 440 C. It is believed that a prolonged treatment at about 300 C (solubility 0.0005) is equivalent to the furnace-cooled test and can be used for comparative purposes. No carbide precipitate can be detected in this alloy even in the furnace cooled condition (see Fig. 6) but a precipitate is indicated by the rise in transition temperature.

<u>Alloy 98V</u> (0.0059% C): The standard Charpy tests showed the reverse of Alloy 86V in that the 700 C quench resulted in a higher transition temperature than either the air cooled or furnace cooled groups. This result was duplicated with two separate heat treatments and with sub-size test bars.

Results from the sub-size tests indicate a large decrease in transition temperature when the quenching temperature is lowered to about 540 C. Quenching from below this temperature resulted in a sharp increase in transition temperature. Figs. 7, 8 and 9 show the energy temperature curves for this alloy.

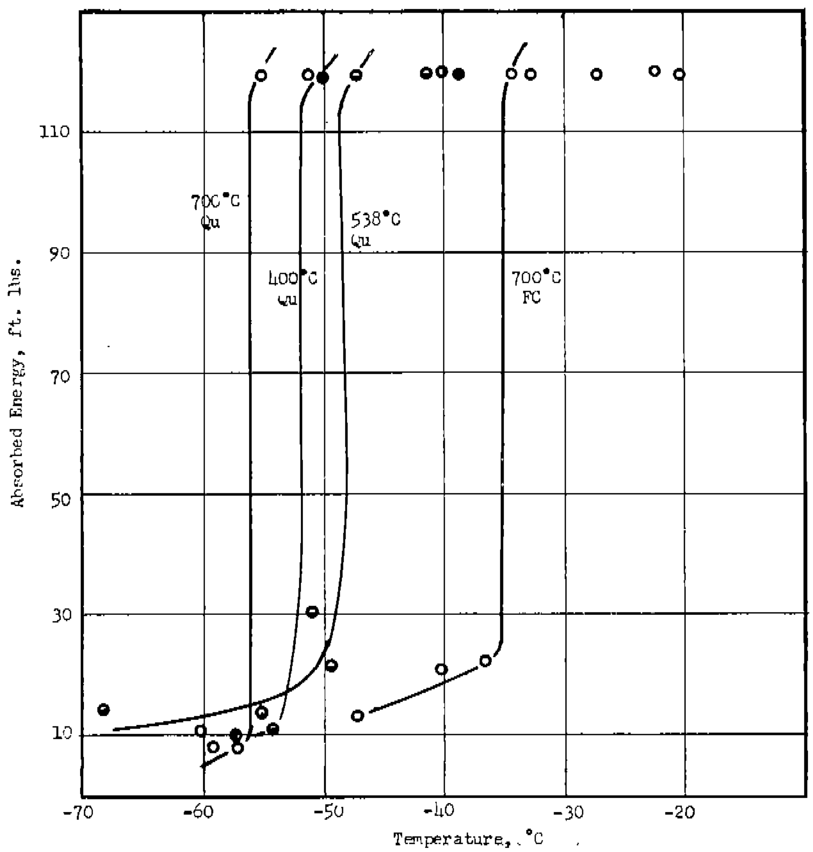


Figure 5. Alloy 93V, .0027% C, Standard Charpy V-Notch Bars

Figure 6. Alloy 93V, .0027% C; cold worked, recrystallized at 700 C and furnace cooled from 700 C. No visible carbides. 2% Nital etch. 200X

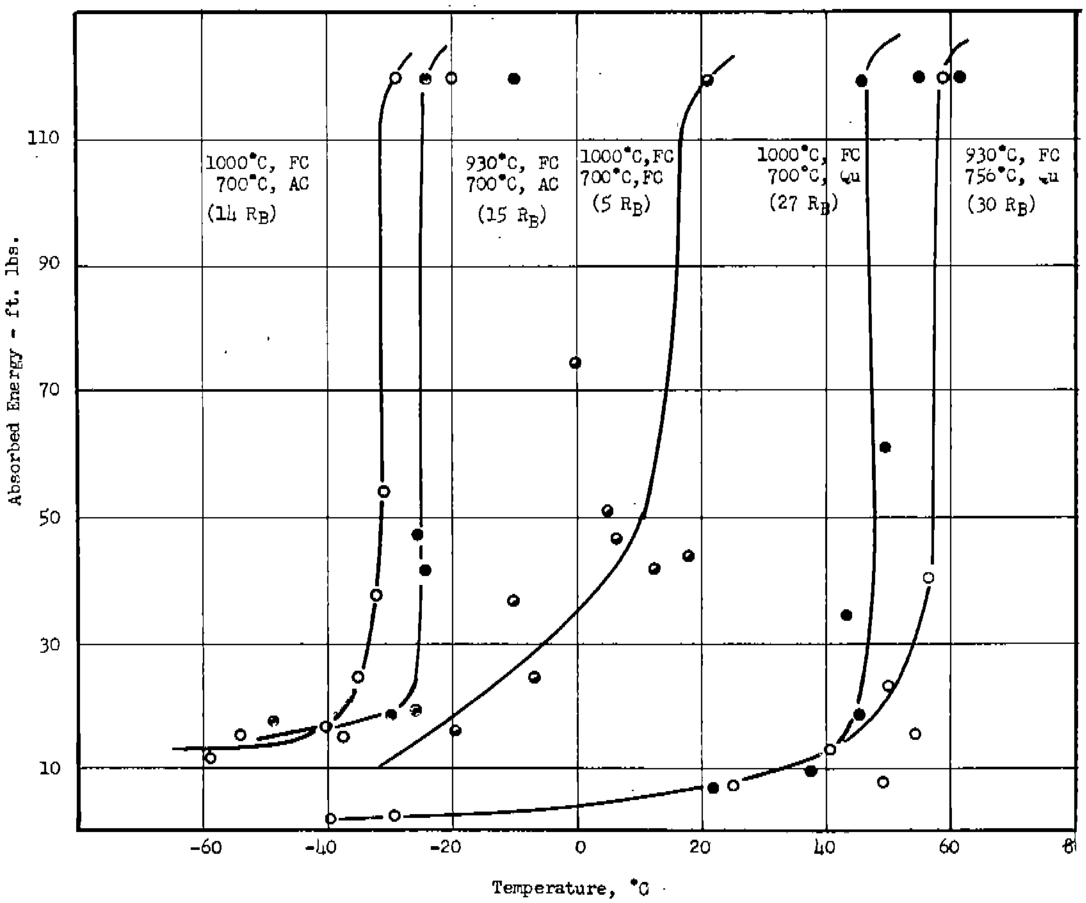


Figure 7. Alloy 98V, .0059%C, Standard Charpy V-Notch Bars

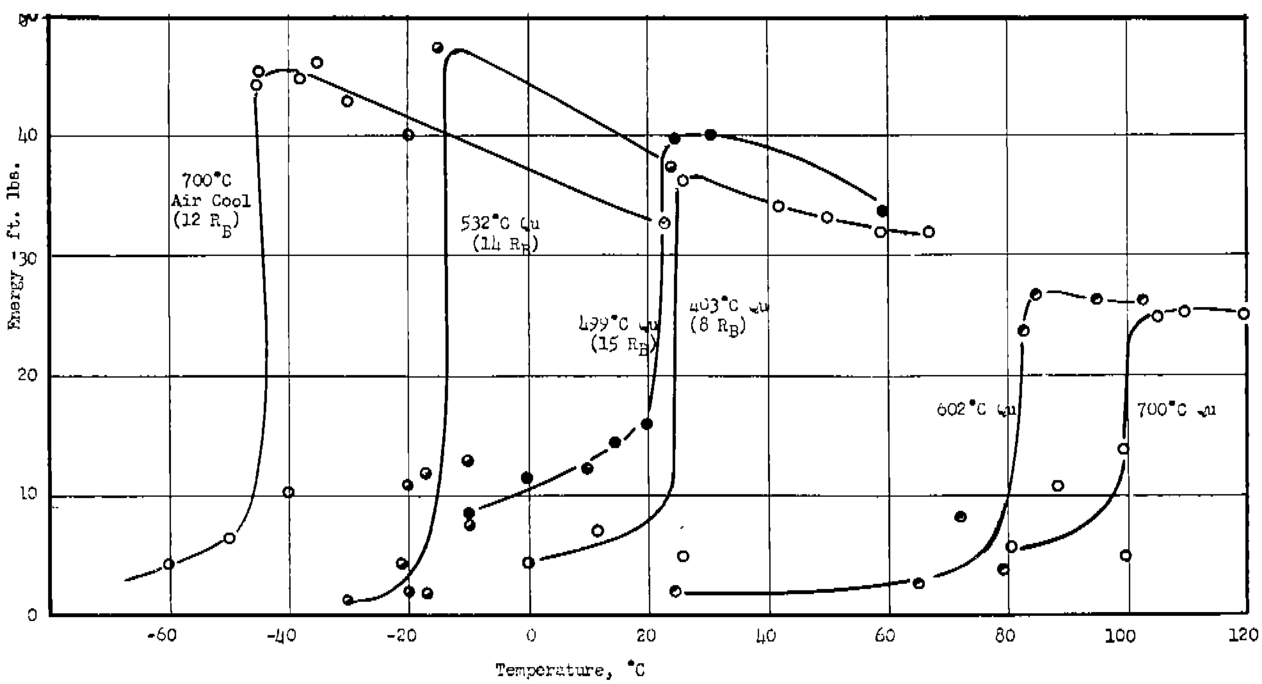


Figure 8. Alloy 98V, .0059% C, Sub-size Charpy V-Notch Bars

Fig. 10 is a plot of quenching temperature vs. sub-size transition temperature for Alloys 86V and 98V. It is seen that both alloys show a minimum transition temperature at their respective saturation temperatures. It is to be noted here, however, that supercritical recrystallization was used for these alloys to obtain structures suitable for testing and collection of data. Also, the sub-size impact data have been plotted against relative carbon saturation for these alloys (Fig. 11).

Alloy 101V (Approx. 0.016% C): The impact behavior of Alloy 101V as shown in Fig. 12 is exactly the same as that of Alloy 98V. The transition temperature vs. quenching temperature curve, has a minimum value at about the 690 C quench for the supercritical recrystallization. However, the subcritically recrystallized material shows no such minimum, although the sharp rise resulting from precipitated carbide is duplicated quite well.

Tensile data for Alloy 101V are tabulated in Table IV. These data were subjected to a statistical analysis that reveals the following:

1. The upper yield point decreased by an average of 600, 1050 and 430 psi for 0°C, -77 C and -192 C respectively, upon reheating to a lower temperature. The difference of 1050 psi is the only one that is statistically significant at greater than 50% probability. This means that for the -77 C tests, the statement that there is a decrease in yield strength of 1050 psi will be proved correct in similar tests more than 50 times out of 100.

2. There is an indication of some slight increase in total strain of the 690 C quenched material over that quenched at 723 C, but no statistical significance is indicated except for the liquid-air-temperature tests, in which case both strains are quite small.

3. Accurate fracture stress measurements were possible only with the -192 C tests. At this temperature, the fracture stress of the 690 C quench showed an increase of 3270 psi over that of the 723 C treatment. This is significant to a probability of about 70%.

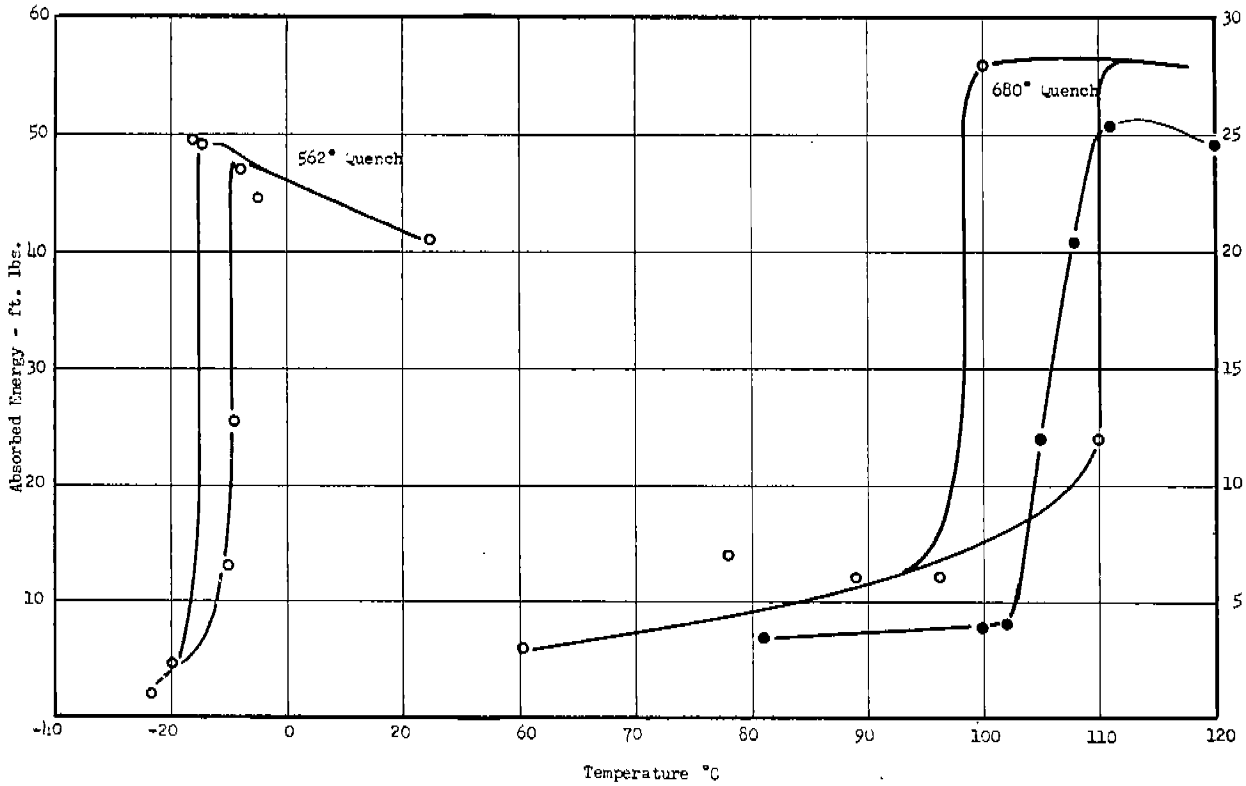


Figure 9. Alloy 98V, .0059% C, Sub-size Charpy V-Notch Bars

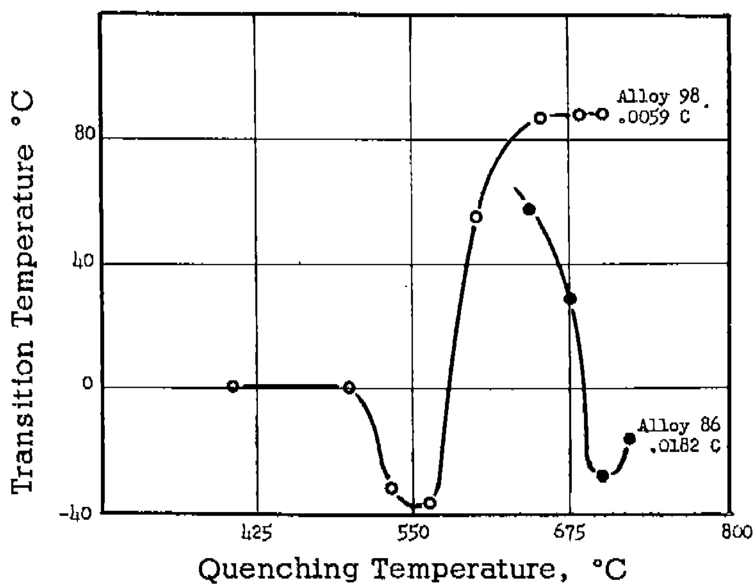


Figure 10. Sub-size Charpy V-notch transition temperatures for two alloys versus temperature of quench for bars previously air cooled from 700 C, then reheated to designated temperature and quenched.

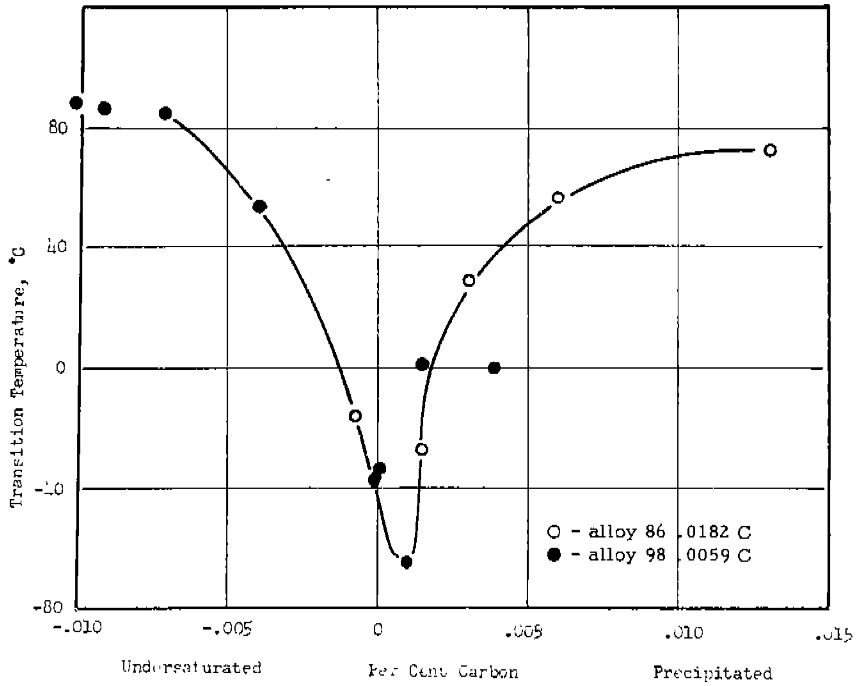


Figure 11. Sub-size Charpy V-notch transition temperature versus amount of carbon precipitated as Fe_3C (right side) or amount present below saturation, (left side).

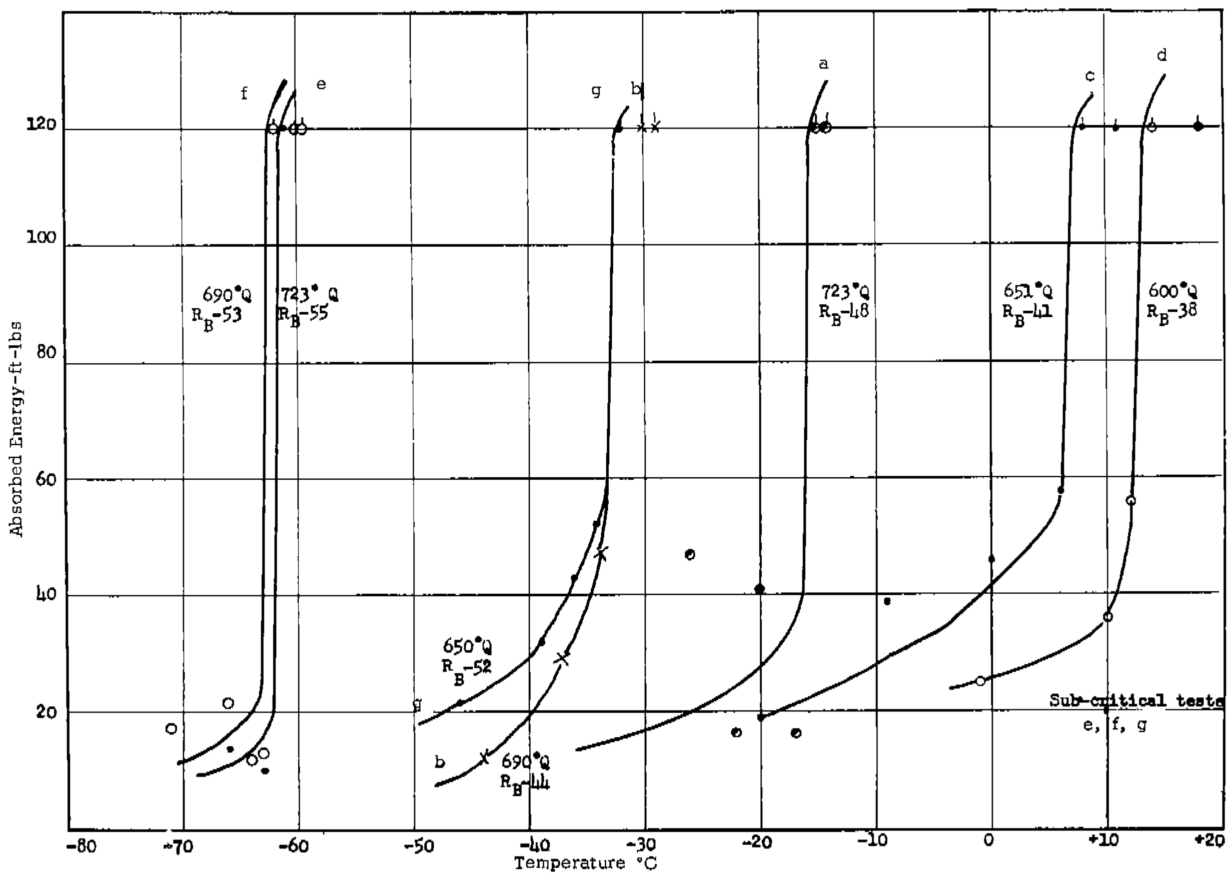


Figure. 12. Standard Charpy V-notch tests on Alloy 101V

Table IV - Tensile Tests on Alloy 101V

50% RA recrystallized 920 C 1/2 hr - FC to 723 C - 1 hr AC

	Reheated to 723 C 1/2 hr Q			Reheated to 690 C 1/2 hr Q		
Temp.	Upper Yield (psi)	Fracture (psi)	Total Strain	Upper Yield (psi)	Fracture (psi)	Total Strain
	19,180		1.41	19,040		1.15
0°C	21,340		1.18	20,170		1.22
	20,710	*	1.12	20,010	*	1.11
	22,130		1.20	20,460		1.28
	18,950		1.11	19,640		1.35
Avg	20,460		1.20	19,860		1.22
	38,380	*	0.919	36,760	83,720	0.813
	38,760	79,650	0.852	36,300	94,340	0.991
-77 C	36,220	85,710	0.868	38,530	*	0.900
	36,800	76,520	0.885	35,830	77,240	0.672
	39,090	88,000	0.896	36,580	84,800	0.816
Avg	37,850	82,470+	0.884	36,800	85,020+	0.838
	71,960	79,000	0.00199	66,910	82,090	0.00199
	67,480	83,570	0.0139	71,000	88,320	0.0276
-192 C	71,880	83,160	0.00796	71,560	84,300	0.0186
	63,760	78,180	0.00298	67,700	85,380	0.0159
	71,200	84,820	0.00995	68,000	---	0.00995
Avg	69,460	81,750	0.00735	69,030	85,020	0.0148

*Fracture load unobtainable due to peculiarity of constant strain rate loading.
+Data not analyzed statistically.

Alloy 105V (Approx. 0.019% C): Alloy 105V conforms to the behavior of the preceding alloys. This alloy has a high carbon content; hence no minimum is observed, since the 723 C quench can be considered as the saturation temperature for the alloy, and no unsaturated tests are possible. Several of the groups of specimens which were recrystallized supercritically behaved in a manner never before encountered. Groups (a) and (e) had a gradually sloping energy-temperature relationship. Also, group (j), whose purpose was to check group (b), both of which were quenched from 723 C, provided a much lower transition temperature (more in the expected position) than group (b), despite its coarser ferrite grain size. It is felt that some peculiarity in heat treatment and possibly the higher oxygen content are responsible for the non-conformity of some of the data.

The subcritically recrystallized material showed no unusual characteristics, as can be seen from the energy-temperature curves in Figs. 13 and 14. An abrupt rise in transition temperature with decreased quenching temperature is apparent for both basic recrystallization procedures. Fig. 15 is a plot of transition temperature vs. quenching temperature for the standard Charpy bars of Alloys 93V, 101V and 105V that were tested.

Because of the differences in grain size existing between groups of test bars, it was necessary to apply a correction to the transition temperatures obtained, in order that comparisons could be made. A correction of 15 C increase in transition temperature per unit increase in ASTM number was used. It is felt that this correction does not in any way affect the validity of the conclusions here presented.

DISCUSSION

All of the alloys tested show a sharp rise in impact transition temperature corresponding to the precipitation of Fe_3C; however, the precipitation is accompanied by a decrease in the amount of carbon in solution in the ferrite matrix. Along with decreased carbon in solution, there should be a decrease in yield strength.

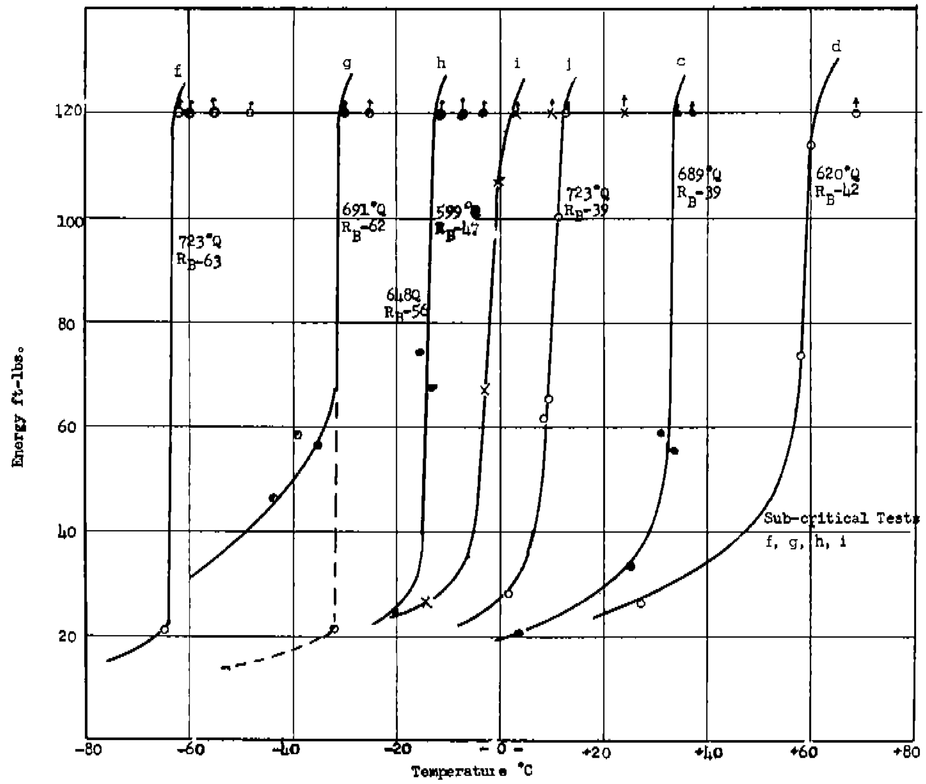


Figure 13. Standard Charpy tests, Alloy 105V (0.019% C)

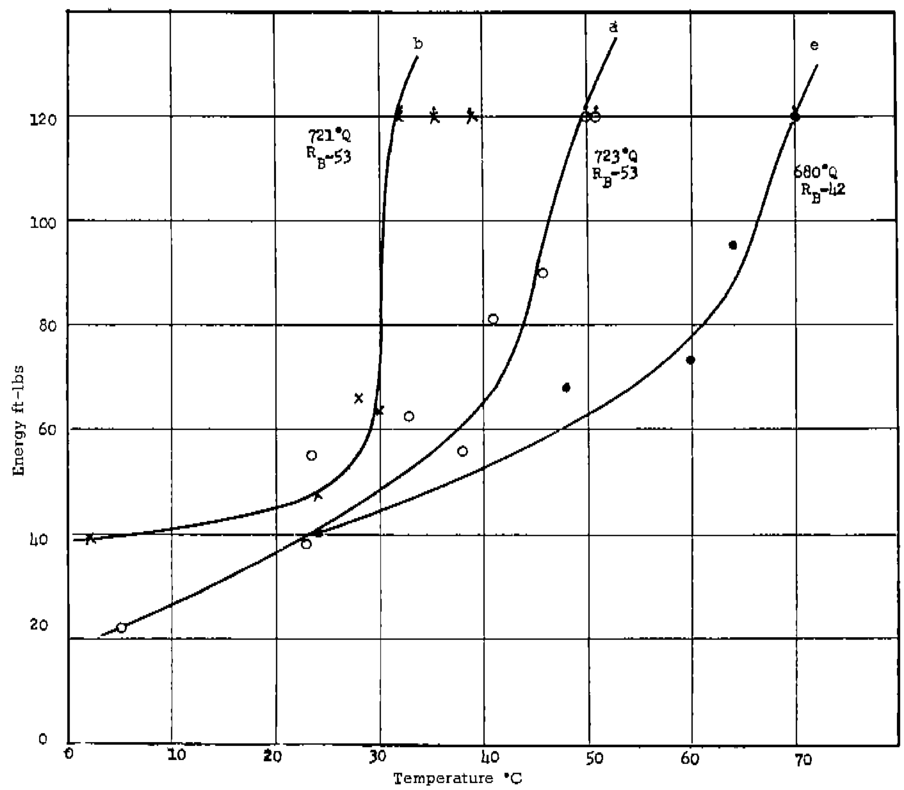


Figure 14. Standard Charpy V-notch tests on Alloy 105V

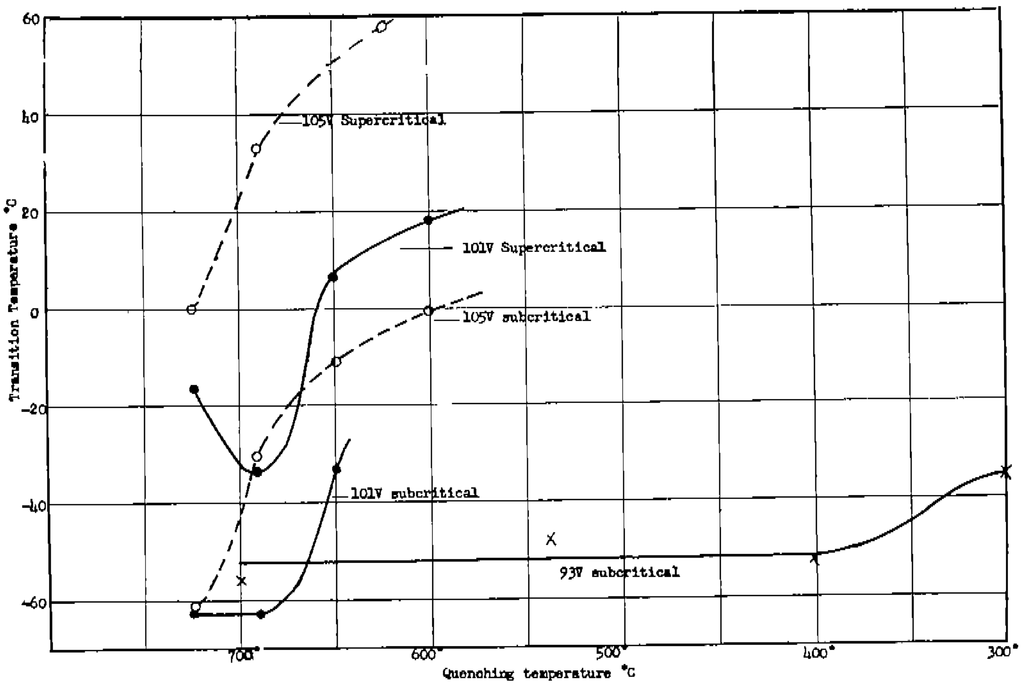


Figure 15. Standard Charpy V-notch transition temperatures (not corrected for grain size) for Alloys 93V, 101V and 105V vs. temp. of quench for bars previously air cooled, then reheated to designated temperature and quenched.

This in turn should result in lowered transition temperature. Thus it can be said that the rise in impact transition is due solely to Fe_3C crystal precipitates. Iron carbide first precipitates as platelets along the ferrite grain boundaries. These platelets act to increase the average crack radius of the ferrite matrix, causing a decrease in fracture stress (brittle fracture at lower critical normal stresses) and higher transition temperatures <u>despite the lowered yield stress</u>.

This description is supported by data obtained recording the continuous drop in Rockwell-B hardness with decreased quenching temperatures. The drop in hardness is an indication of the decreased yield strength of the ferrite. It is necessary that appropriate tensile testing be performed to support this hypothesis. The tensile results should indicate lower yield stresses, much lower fracture stresses and somewhat lower ductility with increased precipitation of iron carbide.

All of the alloys had higher transition temperatures for the supercritically recrystallized condition. This is quite clear from test results in the α or unsaturated region. In this connection it was observed that the supercritically recrystallized ferrites contained sub-grain boundaries within the primary ferrite grain boundaries. (Figs. 16 and 17.) The existence of these sub-grain boundaries can be used to explain the differences in transition temperature that have been found between supercritically and subcritically heat-treated ferrite.

The mechanism of yielding can be considered to be the movement of dislocations to and through ferrite grain boundaries: the initiation of yielding is the dislocation movement within grains (incipient yielding), and the continuation of yielding is the dislodging of dislocations from grain boundaries. The idea that "piled up" dislocations at grain boundaries can cause yielding has been discussed by Smith et al[5] from work by Cottrell. Smith discusses the effect of a sub-structure as that of increasing yield strength without appreciably affecting fracture stress. This is accomplished by the obstruction to dislocation movement offered by the sub-grain boundaries. The large ferrite grains are compartmented (with respect to dislocation movement) by the sub-structure, and fewer dislocations can "pile up" at any one boundary. Hence, larger stresses are necessary to cause the stress magnification needed for yielding. However, no increase in fracture stress as usually associated with finer grained ferrites is predicted. It is postulated that the low angular differences across sub-grain boundaries have little effect on the energy necessary to propagate a brittle fracture through the ferrite. The reasoning behind this is that the change in fracture path is quite small compared to large angular changes at primary ferrite grain boundaries. Note that in all cases the tested ferrites fractured in a transcrystalline manner.

The above can also be stated from the viewpoint that a coarse ferritic structure with veining has the yield strength of a finer ferritic structure, but it has the average crack length (critical normal stress for fracture) of a coarse ferrite. Thus

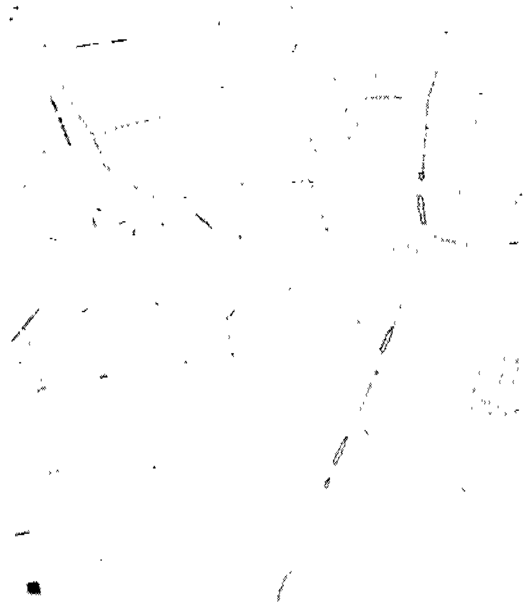


Figure 16. Alloy 98V, .0059% C. Furnace cooled from 1000 C. Grain boundary Fe_3C and Widmanstatten precipitates of Fe_3C. Picral etch. 500 X

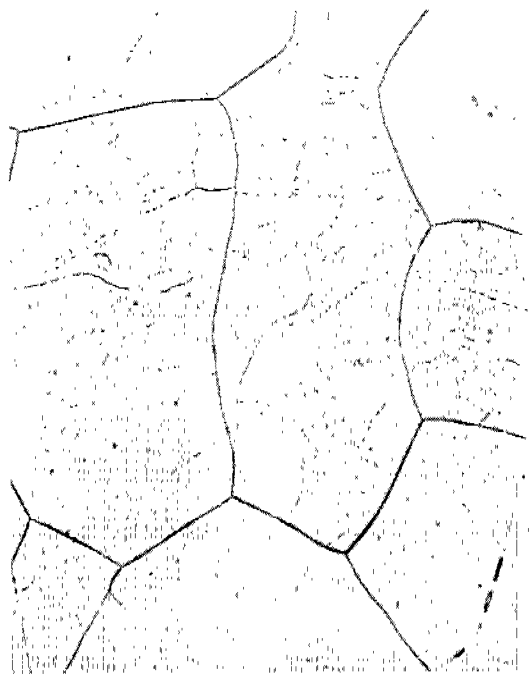


Figure 17. Alloy 98V, .0059% C. Furnace cooled from 1000 C. Different view of specimen of Fig. 16, heavily etched with Picral to bring out veining of ferrite. 200 X

comparing ferrites with identical primary grain sizes, the one with sub-structure will have a higher impact transition due to its higher yield stress. This hypothesis is supported by yield strength data and fracture discussions in Smith's paper.[5] The transition-temperature changes that are associated with the yield-stress and fracture-stress changes of ferrites (with and without sub-structure) have been shown schematically in Fig. 1.

Two possibilities can explain the decrease in transition temperature of the ferritic structures as saturation is approached, as shown by Alloys 98V and 101V. One is that as the temperature is lowered in the α region, the concentration of carbon atoms at grain boundaries increases. This is indicated by the fact that the first carbide precipitate appears on ferrite grain boundaries. At the same time, the concentration of carbon in the ferrite matrix must decrease. This means that, on the average, there are fewer carbon atoms per dislocation, which indicates that the dislocations should become easier to move. The increased mobility of dislocations should then result in a decreased macroscopic yield point. This hypothesis is supported somewhat in one case by the slight but significant decrease in yield point shown in tests on Alloy 101V. It should be noted that the tensile tests corresponded to a drop in impact transition temperature of about 16 C. Had material been available to conduct tests on Alloy 98V (where a difference of about 100 C exists), it is probable that more significant changes in yield data would have been noted.

The other possibility is that the carbon segregating at boundaries does so at sub-grain as well as primary boundaries and in some way "strengthens" the sub-grain boundaries so that they assume more of the properties of high-angle grain boundaries. It is possible that, at saturation, the sub-boundaries have appreciable resistance to the propagation of brittle fracture. This would be manifested in an increased stress necessary for fracture, as is shown in the fracture data for Alloy 101V.

It would seem probable from the data gathered that both of these mechanisms for lowering transition temperature act on the alloys to some extent. The combination of a small decrease in yield strength with a small increase in fracture stress could result in a rather large decrease in impact transition temperature.

CONCLUSIONS

1. Charpy impact transition temperatures of pure irons with from 0.01 to 0.02% C rise abruptly by as much as 100 C upon precipitation of Fe_3C in the ferrite. The mechanism effecting this rise is thought to be the precipitation of iron carbide at temperatures in the range 400 C to 700 C, which results in a decrease in fracture stress.

2. Ferrites with veining have higher transition temperatures than those without veining as quenched from the α or unsaturated region.

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