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EXPERIMENTAL REPORT

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WATERTOWN ARSENAL
WATERTOWN, MASS.

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Watertown Arsenal Laboratory
Report No. WAL 710/605
Problem No. B-1.3

6 April 1944

ARMOR PLATE - CAST

Metallurgical Examination of a Mn-Cr-Ni-Mo-V Steel Used for Cast Gun Shield Armor

OBJECT

To determine the metallurgical characteristics of a manganese-chromium-nickel-molybdenum-vanadium steel used for 4 to 6 inch armor castings by the Continental Roll and Steel Foundry Company.

SUMMARY OF RESULTS

1. The Mn-Ni-Cr-Mo-V type steel examined has some promise of making satisfactory 4 to 6 inch armor provided that proper measures are observed in the homogenization and quenching of the castings.

2. As has been observed in previous studies of 4 to 6 inch armor at this arsenal, the fibre fracture test and V-notch Charpy impact tests were very effective in revealing differences in the tendency of the steel to exhibit brittle properties such as may be encountered in the ballistic test. In this study the following results were obtained:

a. The two 6 inch sections exhibited extremely poor impact properties as heat treated by the manufacturer.

b. Considerable improvement in notched bar impact properties was obtained by employing a higher austenitizing temperature and quenching cold.

3. The results of one of the experimental heat treatments indicate that the following steps assist in improving the notched bar impact properties of this steel:

a. Homogenize at 1950°F.

b. Double quench to obtain carbide solution in conjunction with a fine grain size.

c. Quenching to as low a temperature as possible without cracking.

d. Step temper at 600°F. prior to tempering to the desired hardness.

4. The upper bainitic structures formed during quenching are believed to be responsible for the poor impact properties observed. These structures, after tempering to 240 Brinell, were difficult to distinguish from tempered martensite. The hardnesses of these structures are not appreciably lower than that of martensite; and the end quench hardenability test, which is based upon a hardness survey, is, therefore, not particularly useful for revealing the presence of undesirable bainitic structures.

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INTRODUCTION

In a previous report[1] of heavy cast armor, the tendency of the steel to exhibit poor impact toughness in ballistic tests and in laboratory notched bar tests has been associated with the presence of undesirable transformation constituents which formed upon quenching of steel having insufficient hardenability. A 2.5% Cr, .5% Mo type steel was found to exhibit satisfactory ballistic properties when heat treated to a rather low hardness (200-230 Brinell). Its impact properties were correspondingly higher than those obtained with the lower alloy steels used. The lower alloy steels were found to possess a large amount of ferrite and pearlite which formed upon quenching and to which the poor impact properties were attributed.

Another steel having an end quench hardenability comparable to the 2.5% Cr, .5% Mo type has recently been employed by the Continental Roll & Steel Foundry Company for making heavy armor. This steel has the following type composition:

C	Mn	Si	Ni	Cr	Mo	V
.30	1.0	.30	1.5	1.0	.45	.10

Considerable difficulty has been encountered by the manufacturer in satisfying the ballistic requirements using this steel with its present heat treatment. This investigation was undertaken to determine the causes for the poor ballistic properties and, if possible, to determine methods for improvement of the armor.

It has been found that notched bar impact test results correlate with the properties which are observed in a shock type ballistic test.[2,3] Since the majority of failures encountered in the ballistic tests of six inch thick plates of the above analysis were brittle failures resulting in excessive cracking or breaking of the plates, the impact test was used as the chief measure of comparison.

Materials and Test Procedure

Two test blocks (6"x8"x18" in size) which had been previously heat treated by the producer were used in this investigation. The blocks were identified as Serial No. 18, Heat 3656, and Serial No. 20, Heat 3695-1, made by the Continental Roll & Steel Foundry Company.

The metallurgical properties were obtained on block No. 18 in the as-received condition. Sections 1/2" to 5-1/4" thick were heat treated according to cycles described in the Results and Discussion, and the metallurgical properties were obtained after these treatments.

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[1]Watertown Arsenal Lab. Rpt. 710/500 - Armor, Metallurgical Examination of Cast Gun Shield Armor Four to Six Inches in Thickness.

[2]Watertown Arsenal Lab. Rpt. 710/532 - Armor, Development of a Fracture Test to Indicate the Degree of Hardening of Armor Steels upon Quenching.

[3]Watertown Arsenal Lab. Rpt. 710/534 - Armor Plate, Correlation of Metallurgical Properties with the Low Temperature Ballistic Shock Characteristics of 1" to 2" Low Alloy Cast Armor Tested at Camp Shilo.

Microscopic examinations, V-notched Charpy impact tests, fracture tests, and hardness surveys were used to determine the relative metallurgical properties under the various sets of conditions.

The grain coarsening temperature of the steel and the austenitizing temperature required for carbide solution were determined by fracture grain size tests and microscopic examination of specimens quenched from temperatures of 1550°F. to 1700°F.

A chemical analysis of both test blocks was obtained. Hardenability tests at various austenitizing temperatures were obtained on block No. 18.

Results and Discussion

1. Ballistic Tests

Ballistic test plates from the same heats as the test blocks used in this investigation were subjected to ballistic tests and the results are summarized in Table I.

TABLE I

Results of Ballistic Tests Conducted at

The Proving Center, Aberdeen

Continental Roll & Steel Foundry Company Plate No. 18, Heat No. 3656

A.P.G. Firing Record P-20158
Thickness - 6"
Projectile - 6" A.P. MK27 at 15° obliquity.

Round	Velocity F/S	Depth of Penetration	Effect on Plate
1	1447	5-3/4" partial	1-1/4" bulge with 6" crack.
2	1437	5-1/2" partial	1-1/4" bulge.
3	1586	5-7/8" partial	1-1/2" bulge with 8" crack.
			Plate acceptable.

Continental Roll & Steel Foundry Company Plate No. 20, Heat 5695-1

A.P.G. Firing Record P-22116
Thickness - 6"
Projectile - 6" A.P. MK27 at 15° obliquity.

Round	Velocity F/S	Depth of Penetration	Effect on Plate
1	1435	Complete penetration	A star crack on rear of penetration and 27" crack extending to the edge of plate.
			Plate rejectable.

Plate No. 20 exhibited a severe cracking tendency under the ballistic attack at a velocity of 1435 f/s.

Plate No. 18 was subjected to the attack of a 6 inch projectile at a considerably higher velocity (1586 f/s) without exhibiting a rejectable amount of cracking. However, the presence of an 8 inch crack on the back of this plate shows that there is a tendency toward brittle properties. A projectile fired at a velocity high enough to cause a complete penetration would probably expose a brittle exit condition in this plate.

The ballistic test which is required at present* reveals the cracking tendency of very brittle armor and the low resistance to penetration of armor heat treated to abnormally low hardnesses. However, this test is considerably milder than that required for lighter gage armor which must withstand complete penetrations of overmatching projectiles at high obliquities without spalling or cracking excessively.

2. Chemical Analyses

The results of chemical analyses made at this arsenal as well as by the manufacturer are as follows:

Plate No.	Tested By	C	Mn	Si	S	P	Ni	Cr	Mo	V
18	Watertown Arsenal	.31	1.05	.31	.026	.024	1.56	.93	.48	.09
18	Manufacturer	.30	.96	.31	-	-	1.48	1.10	.45	.10
20	Watertown Arsenal	.30	1.12	.37	.025	.037	1.73	1.15	.44	.11

It should be noted that the steel was made in an acid open hearth furnace and the heats were killed with silicon and vanadium.

A residual quantity of boron (.0007%) was observed in plate No. 18, but its influence is uncertain since many steels and irons, including some relatively pure ones, contain a residual amount of this element. Plate No. 20 was not analyzed for boron.

3. Grain Coarsening Temperature and Austenitizing Temperature Necessary to Effect Carbide Solution

Small sections (3/4" square x 6" long) were heat treated according to the schedule in Table II. The samples were oil quenched to room temperature after six (6) hours at the austenitizing temperature indicated. Fracture grain size determinations were obtained on the specimens in the as-quenched condition. The persistence of undissolved carbides was investigated by examining the microstructure at 1000X using a picral etch.

*A partial penetration of a 6 inch projectile at an obliquity of 15°.

TABLE II

Grain Size and Carbide Solution as Influenced by Austenitizing Temperature

Sample Number	Austenitizing Temperature	Hardness BHN	Fracture Grain Size	ASTM Grain Size	Carbide Solution
1	1550°F.	477/495	6	6	Incomplete
2	1600°F.	495	6	5	Incomplete
3	1625°F.	477/495	6	5	Almost Complete
4	1650°F.	444/461	5	4-5	Complete
5	1675°F.	477/495	4-5	4	Complete
6	1700°F.	477/495	4-5	4	Complete

The above results indicate that grain coarsening begins at temperatures of 1650°F. and above. Furthermore, it is apparent that an austenitizing temperature of at least 1650°F. is necessary to effect satisfactory solution of carbides. This combination of circumstances indicates the desirability of using aluminum or titanium deoxidation in a steel of this type in order that a fine grain size will be retained in conjunction with complete carbide solution.

4. Hardenability Tests

Jominy type end quench tests were conducted on plate No. 18 and the resulting hardness surveys are shown in Figure 1. The bars were austenitized at 1550°F., 1625°F., and 1675°F. to determine the effect of carbide solution and grain size upon the hardenability of this steel.

A microscopic examination was conducted on the end quench bars austenitized at 1575°F. and 1675°F. and representative structures are shown in Figure 2. The structure was observed after tempering at temperatures of 600°F., 800°F., 1050°F., and 1250°F. Figures 2G, H, I show the structure of the bar tempered at 1050°F., a temperature which showed the greatest contrast of structures.

At the quenched end of the bar a light etching acicular martensite was present. At distances of 4/16" to 32/16" from the quenched end a mixture of light needles of martensite and dark needles of martensite formed at higher temperatures (possibly tempered) or low temperature bainite were present. At the air cooled end of the bar where the hardness showed signs of dropping slightly, an anisotropic distribution of carbides, probably upper bainite, was found associated with the above mentioned structures. Since the nonuniform carbide distributions are believed to be associated with the tendency of steel to exhibit poor impact properties, such as the cracking tendency in a ballistic test, this structure upon tempering should not have as high an impact toughness as tempered martensite.

Since the air cooled end of the Jominy bar develops a cooling rate approximately equivalent to the center of a 4" plate quenched in water (agitated), the suppression of this structure would be impossible in the quenching of 4" to 6" armor.

The persistence of the oriented carbide distribution is maintained to high tempering temperatures. At tempering temperatures of 1250°F. to 1275°F. the carbides spheroidize to a fairly uniform structure although some of the orientations of carbides continue to persist.

The low hardness at the rapidly cooled end of the bar quenched from 1550°F. is not readily explained. There is a possibility of austenite being retained on the quench in appreciable amounts although the mechanism is not apparent since the undissolved carbides are generally believed to promote transformations.

5. Factors Involved in the Heat Treatment of Heavy Sections

Recent work has indicated the necessity for quenching to martensite as completely as possible prior to tempering in order to attain the optimum ballistic properties in armor. To do this in thick sections entails a consideration of many factors, some of which may be clarified by an examination of the S-curve data (time-temperature-transformation characteristics) of Bain and other investigators. A precise evaluation of the steel cannot be obtained since transformation rates occurring on continuous cooling are known to differ considerably from those taking place during isothermal treatments.

A comparison of the S-curve characteristics obtained in a low alloy steel and the Mn-Cr-Ni-Mo-V steel are shown schematically in Figures 3 and 4. At present no S-curve is available for the latter type steel, but the data in the literature point to a curve having the form shown in Figure 4.

a. Prevention of High Temperature Transformation Structures.

The alloy content in the given steel moves the upper nose (pearlite and ferrite) to the right sufficiently so that little if any transformation takes place in this temperature range when a 6" section is water quenched.

b. Transformation to Bainite.

It is virtually impossible to prevent some transformation to bainite in the given steel when heat treated in a 6" section. According to Troiano[4] as well as some foreign investigators, this reaction does not go to completion in the times ordinarily encountered during quenching. The amount of transformation which will take place isothermally is zero at the upper temperature limit, and it approaches 100% at the Ms temperature.

[4] O.D.-34-3 - "An Investigation of the Metallographic and Physical Properties of New Types of Gun Steels" - A. R. Troiano. 8 December 1943.

Consequently, if the steel is held at a temperature above the Ms temperature, considerable quantities of austenite may remain undecomposed. This austenite may become stable and resistant to transformation upon subsequent treatment.

Nickel and manganese are believed to be the most potent elements (except carbon) in moving the lower (bainite) nose to the right according to S-curve data available, but it undoubtedly requires an inordinate quantity of alloy to move the bainite transformation curve sufficiently to prevent the formation of bainite with the cooling rates encountered at the center of a 6" thick plate. However, it is desirable to cool through this range as quickly as possible to minimize the amount of transformation which takes place and to cause the transformation which must take place to occur at as low a temperature as possible.

The lower temperature bainite whose carbides are not resolvable with the ordinary microscope may not exhibit markedly inferior properties to those obtained with tempered martensite. On the other hand, the acicular bainites having microscopically visible carbides which transform at the higher temperatures (approx. 900°F.) definitely exhibit inferior properties, and it is necessary to cool the steel below this temperature as rapidly as possible. In order to cool the center below this temperature, the surface will of necessity be considerably colder because of the considerable temperature differential between surface and center of such heavy sections when water quenched.

c. <u>Transformation to Martensite.</u>

Differing from the higher temperature transformation products, it is believed that martensite is not formed isothermally but almost instantaneously on continuous cooling. The amount formed is determined by the temperature to which the steel is cooled. This transformation is represented by a temperature range Ms (start) and Mf (finish) in the diagrams, Figures 3 and 4.

When cooling upon quenching is interrupted above the Mf temperature, the martensite reaction does not go to completion, and the retained austenite will be allowed to form the upper transformation products upon tempering.

One difficulty with the higher alloy steels, especially those rich in nickel, manganese, and chromium, is that the Ms and Mf temperatures are lowered, and thus the steel must be quenched colder in order to transform the austenite to martensite. The effect on the Ms temperature has been investigated by Payson and Savage[5] who observed that the temperature is lowered according to the following formula:

$$\text{Ms } ^\circ\text{F} = 930 - 570\ \text{C} - 60\ \text{Mn} - 50\ \text{Cr} - 30\ \text{Ni} - 20\ \text{Mo}$$

These investigators also observed that the Ms to Mf zone is widened with increases in the above alloys, thus materially lowering the Mf point. For the composition used here, the Ms temperature was calculated to be about 575°F. Since there is a marked tendency for the higher alloy

[5]Payson and Savage, "Martensite Reactions in Alloy Steels", Trans. ASM V33, 1944, page 261.

steels to quench crack, it may be advisable to cool the steel more slowly once the Ms temperature is reached at the center of the section. Interrupting the quench before the Mf point is reached results in the retention of untransformed austenite which will be decomposed on tempering.

d. Transformation of Retained Austenite

Any austenite not transformed during quenching will tend to transform to high temperature transformation products if tempered directly.

It would be expected that the transformation of retained austenite upon tempering (forming pearlite) would tend to produce poor impact properties. This tendency is indicated by the manufacturer's observed experience with this type analysis because the shock properties rapidly go bad if the alloy is increased above a certain value (with respect to manganese and chromium particularly) even though greater hardenability results with increasing amounts of these elements. The manufacturer interrupts the quench at a specific temperature then tempers directly. As the Ms and Mf points are lowered by the increase in alloy, an increasing quanity of austenite is retained to be subsequently transformed upon tempering.

From the investigations of the tempering of steel containing retained austenite, it may be inferred that the austenite retained during the quench may be transformed to bainite in an isothermal treatment at a temperature slightly above the Ms point. The bainite reaction goes to completion more rapidly at this temperature (as mentioned above) than it does at higher temperatures, and its structure is believed to be the most desirable of all the constituents other than martensite.*

e. Summary of Factors Involved in Heat Treatment of Heavy Sections

In the heat treatment of armor plate it is desirable to transform the austenite to martensite during quenching if the armor is to have the optimum ballistic properties after tempering. To do this it is necessary to quench the steel below the Ms temperature (575°F. for the given steel). Once the start of the martensite transformation temperature is reached the steel may be cooled more slowly to room temperature. In the high alloy steel involved in this study there is a possibility of retaining untransformed austenite. This may be transformed to a low temperature bainite by means of an isothermal temper at 600°F. After this treatment the steel is tempered to the desired hardness.

The above concepts will be discussed in greater detail in Watertown Arsenal Laboratory Report WAL 320/28 entitled "Selection of Steel for Fully Quenched Parts", which will be completed in the near future.

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*According to a private communication from Mr. E. S. Davenport of the U. S. Steel Corporation Research Laboratory.

6. Properties of Plates as Heat Treated by Manufacturer

The heat treatment employed by the manufacturer is the following:

Temp. °F.	Time: Hrs. Rise	Time: Hrs. Soak	Coolant
1800	31	28	Air to 600°F.
1250	6	12	Furnace
1625	32	28	Water - Quench to approx. 400°F. on Surface
1175	8	24	Air to 600°F.
1250	7	28	Air

The physical properties reported for plate No. 18 are:

T.S. p.s.i.	Y.P. p.s.i.	El. %	R.A. %	Izod Ft.Lbs.	BHN
112,500	82,750	10.5	19.9	35	212-217

The hardness obtained at this arsenal on the test block No. 18 was 235 to 241 Brinell, based upon accurate measurements through the section.

Fracture tests were obtained on both samples by the manufacturer and the fractures were rated at the arsenal as flat crystalline, indicating poor impact toughness.

To confirm this, V-notch Charpy impact tests were conducted on block No. 18, and values of 24-27 ft.-lbs. were obtained at room temperature. At the given hardness level (235-241 Brinell), the values should have been 60 ft.-lbs., which is the value obtained from bars heat treated in small sections. (See Table IV.)

7. Experimental Heat Treatments

Heat treating experiments were conducted on small and large sections from blocks No. 18 and No. 20 according to the schedule in Table III. V-notched Charpy impact tests were conducted on specimens obtained from the center of the sections, and the results are shown in Table IV.

The treatment No. 1 shows that the steel can be heat treated in small sections to give a desirable impact toughness when tempered to the low hardness (235-241 Brinell) at present employed for heavy cast armor.

The treatments 2 and 3 were used to determine if the small amount of crystallinity observed in all sections and the corresponding inferior impact properties could be associated with the presence of retained austeni[illegible] Consequently, a liquid nitrogen quench was used on a specimen following the water quench, and the results were compared with a sample which was not subjected to the nitrogen treatment. A small amount of speckled crystallinity was observed in the fracture specimen quenched in nitrogen as well as in the control specimen. Subsequent V-notched Charpy impact tests indicated a slight improvement by the nitrogen treatment, although it is within the limit of experimental error. Consequently, it is felt that the effect of retained austenite, if present, in small sections is not appreciable in this steel when tempered to a hardness of 235-241 Brinell. However, it is believed that this condition will not necessarily be true in the larger sections wherein slower cooling rates are involved.

Treatment 4 was used to ascertain the properties of the steel when quenched cold in a large section (5-1/4" thick plate) after the steel had been austenitized at a temperature designed to dissolve the carbides. A considerable improvement was obtained over the steel as originally heat treated by the manufacturer. However, the fracture was still undesirable (mottled - considerable crystallinity) and the low temperature impact properties were considerably less than those obtained in quenched and tempered small sections. (See Treatment 1, Table IV.)

All the presently known factors which might be expected to improve the properties of heavy cast armor through heat treatment were incorporated in treatments No. 5 and No. 6. A high temperature normalize (1950°F.) was used because some producers had observed an improvement in ballistic properties as a result of using a 1950°F.-2000°F. normalizing treatment prior to the quench and temper. The tempering cycle following the homogenize was included because this treatment is necessary to prepare the steel for chipping and welding. The double quench was used to effect complete carbide solution at the higher temperature and then refine the grain size by reheating at the lower temperature. The section was quenched cold to minimize the retention of austenite during quenching. A low temperature tempering treatment followed in order to transform any retained austenite to a lower bainite. Several investigators have observed that the transformation occurs most rapidly at a temperature range of 575°F. to 625°F. for alloy steels of this type. The block was then tempered at 1250°F. to obtain the desired hardness. The fracture test revealed a mixed fracture containing about 50% of the speckled type crystallinity which formed in a dendritic pattern. The notched bar impact values were fairly satisfactory for this hardness (290 Brinell) of cast steel, since the values obtained from small sections containing tempered martensite are approximately 5 to 10 ft.lbs. greater.

The steel was retempered at 1275°F. with a resulting hardness of 229-235 Brinell. The quantity of crystallinity in the fracture was lowered to less than 25% and the notched bar impact values were comparable to those obtained in cast steel heat treated to tempered martensite.

It must be noted that in treatment 5 the plate cracked upon quenching. Consequently, it would be advisable to modify the drastic quench by cooling more slowly once the martensite transformation temperature (about 575°F.) is reached at the center of the section.

The microscopic examination did not reveal any marked differences in structure between the thick sections having satisfactory impact properties and those heat treated to give poor properties. (See Figure 5.) The structure consisted of what appeared to be spheroidized sorbite. There were some lamellarly oriented carbide configurations and divorced ferrite in all sections. The quantity of this oriented structure appeared to be greater in the poorly heat treated sections, although it was difficult to evaluate this condition with any degree of accuracy. Typical structures are shown in Figure 5.

It appears that the oriented carbide distributions are the tempered structures of the intermediate temperature transformation products which formed upon quenching. The tempering treatment tends to spheroidize these products and put them in a more desirable form at the low hardnesses (200 to 240 Brinell). Consequently the high temperature tempering treatment improves considerably the toughness of this type of steel when improperly quenched.

8. General Considerations

The Mn-Cr-Ni-Mo-V type steel has some promise of making satisfactory 6" armor when tempered to a low hardness. In utilizing this composition for armor, there are several factors which should be considered in its heat treatment.

The castings from this steel possess considerable carbon and alloy segregation which produces an appreciable effect upon the hardenability of the steel in the dendritic axes of the steel. The result is that the segregated low alloy areas may not completely avoid the high temperature transformation. Some improvement in homogeneity is possible by employing a high temperature (1950°F.) normalize.

The alloying elements in this steel lower the martensite transformation temperature. Consequently, in order to quench the steel to martensite and the accompanying desirable ballistic properties, it is necessary to quench the steel fairly cold. This steel is also susceptible to quench cracking when a drastic quench is used, and it is advisable not to have any unnecessary variations in section size or sharp angles if cracking is to be minimized. This factor, of course, is very important in all the higher alloyed steels used for heavy armor. During heat treatment, this cracking tendency may be lessened by interrupting the water quench at about 575°F. at the center of the section and cooling more slowly to room temperature. A tempering treatment at 600°F. offers some promise of transforming any retained austenite to a low temperature bainite whose properties after tempering are probably not markedly inferior to those of tempered martensite.

The steel was produced in an acid open hearth furnace without aluminum deoxidation. As a result there is a grain coarsening tendency at 1650°F. However, a high austenitizing temperature is necessary to completely dissolve the alloy carbides. In the present study, these difficulties were overcome by using a double quench: (1) at 1675°F. to dissolve the carbides and (2) at 1575°F. to refine the grain size.

From the tests conducted, it is apparent that a small amount of crystallinity associated with the low alloy dendritic axes persists even in small sections drastically quenched. The scattered crystallinity is under 10% of the fractured area, and it is not associated with any marked reduction in the impact properties when the steel is tempered to a low hardness (235 Brinell). Aside from this condition, the steel may be heat treated to exhibit fairly satisfactory properties at a hardness of 235 Brinell in sections up to 5-1/4 inches in thickness when the proper precautions are taken.

The investigation also showed that the poor impact properties associated with the presence of intermediate transformation products are obscured by a high temperature temper which spheroidizes the structure. Consequently a small amount of the bainitic structures may be tolerated in this steel when tempered to a low hardness (under 240 Brinell) without impairing the ballistic shock properties appreciably. However, the ballistic efficiency (resistance to penetration) is decreased by decreasing the hardness.

TABLE III

Heat Treating Cycles

Treatment No.	Plate No.	Sample Size	Equiv. Plate Thickness*	Temp. °F.	Time Hours	Coolant	Fracture**
1	18	3/4x3/4x6"	5/8"	1675 1275	5 8	Water to Room Temp. Air	Fc Trace
2	18	2-1/4x2-1/2x8"	1-3/4"	1675 1275	5 8	Water to Room Temp. Air	Fc 1/4
3	18	2-1/4x2-1/2x8"	1-3/4"	1675 1275	5 8	Water to Room Temp. Nitrogen to -190°C. Air	Fc 1/4
4	18	6x8x18"	5-1/4"	1675 1275	5 8	Water to Room Temp. Air	Fc 7/8
5	20	6x8x18"	5-1/4"	1950 1250 1675 1575 600 1250	10 5 5 5 4 6	Air Air Water to Room Temp. Water to Room Temp. Water to Room Temp. Water	Fc 3/4
6 (Retemper of No. 5)	20			1275	6	Air	Fc 1/4

*The equivalent plate thickness is the thickness of flat plate the center of which has a cooling rate equivalent to that obtained in the sample size used provided that an agitated water quench is employed.

**Impact bar fractures were rated in accordance with a standard procedure, (see Figure 6).

TABLE IV

V-Notched Charpy Impact Strength of

Continental Roll & Steel Foundry Company 6" Armor

	68°F. (20°C.)		+14°F. (-10°C.)		-40°F. (-40°C.)		-94°F. (-70°C.)		
Sample	Value Ft.-Lbs.	Fracture*	Value Ft.-Lbs.	Fracture*	Value Ft.-Lbs.	Fracture*	Value Ft.-Lbs.	Fracture*	BHN
Plate 18	27.0	Fc 1/2	13.0	Fc 7/8	10.0	C			235/241
As-Received	24.0	Fc 3/4	19.0	Fc 7/8	16.5	Fc 7/8			
Plate 18	62.5	F	62.0	F	60.5	F	29.5	Fc 1/2	
Treatment 1			58.5	F	53.0				235/241
Plate 18	61.0	F	31.5	Fc 1/4	46.5	Fc 1/2			235/241
Treatment 2	56.5	F	48.0	Fc 1/4	17.0	Fc 3/4			
Plate 18	62.5	F	58.0	F	32.0	Fc 1/2			235/241
Treatment 3	63.5	F	60.0	F	46.5	Fc 1/2			
Plate 18	51.0	Fc Trace	37.5	Fc 1/2	29.0	Fc 3/4			235/241
Treatment 4	40.0	Fc Trace	37.0	Fc 1/2	23.0	Fc 3/4			
Plate 20	34.5	Fc 1/8	40.0	Fc 1/4	28.5	Fc 1/2			285/293
Treatment 5	39.5	Fc 1/8	36.0	Fc 1/4	34.5	Fc 1/2			
Plate 20	72.0	F			49.0	Fc Trace	37.5	Fc 1/4	235/229
Treatment 6	67.5	F			62.0	Fc Trace	43.0	Fc 1/4	

*Impact bar fractures were rated in accordance with a standard procedure, (See Figure 6.).

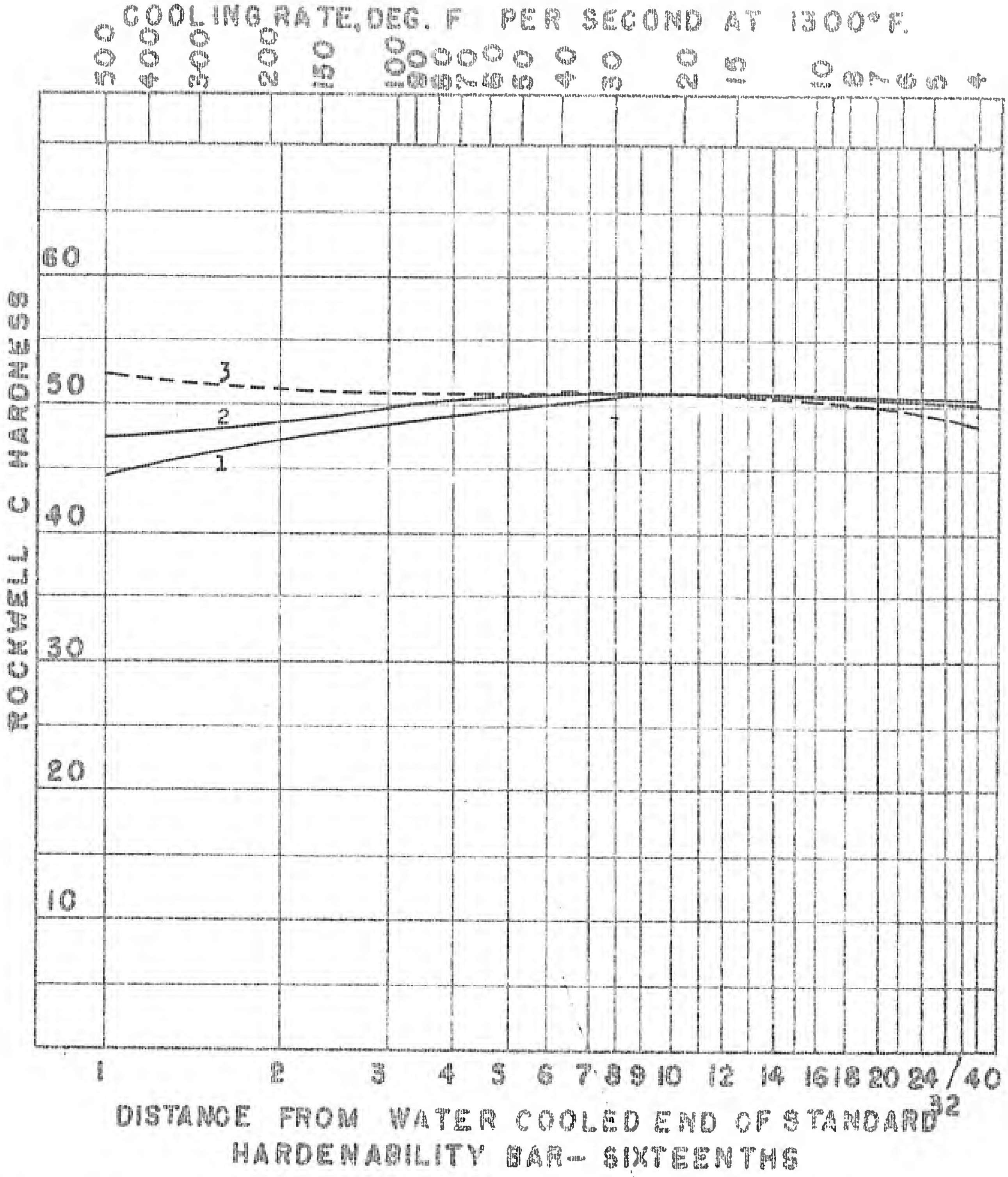


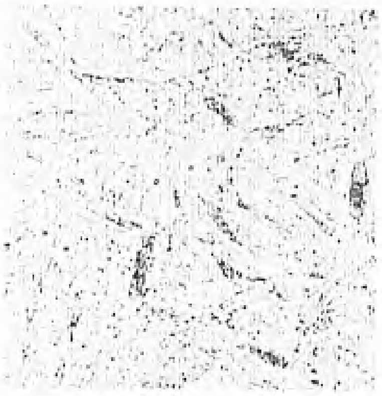
PLATE NO.	HEAT NO.	C	MN	SI	S	P	NI	CR	MO	V	B	QUENCH TEMP	TIME	G.S.
1	3656	.31	1.05	.31	.026	.024	1.56	.93	.48	.09	.0007	1550	6hr.	6
2	3656	.31	1.05	.31	.026	.024	1.56	.93	.48	.09	.0007	1625	" "	5
3	3656	.31	1.05	.31	.026	.024	1.56	.93	.48	.09	.0007	1675	" "	4

FIGURE 4

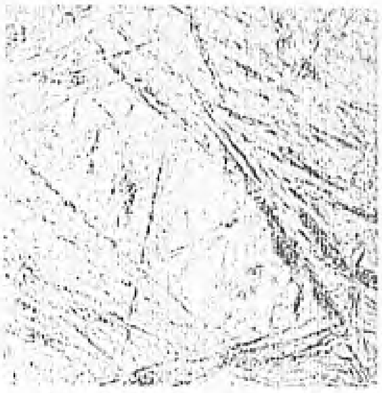
Structures on End Quench Hardenability Bars
Obtained from Test Block 18 Heat 3656

End Quench Bar As Quenched from 1550°F.

A - 1/16" from QE.

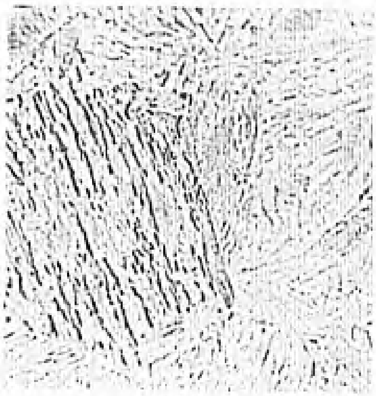


B - 20/16" from QE.



C - 40/16" from QE.

End Quench Bar As Quenched from 1675°F.



D - 1/16" from QE.

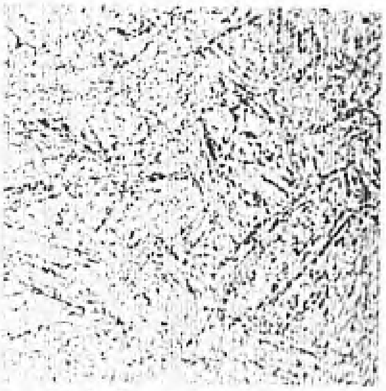
E - 20/16" from QE.

F - 40/16" from QE.

End Quench Bar Quenched from 1675°F., Tempered at 1050°F.

G - 1/16" from QE. Rc 40

H - 20/16" from QE. Rc 41

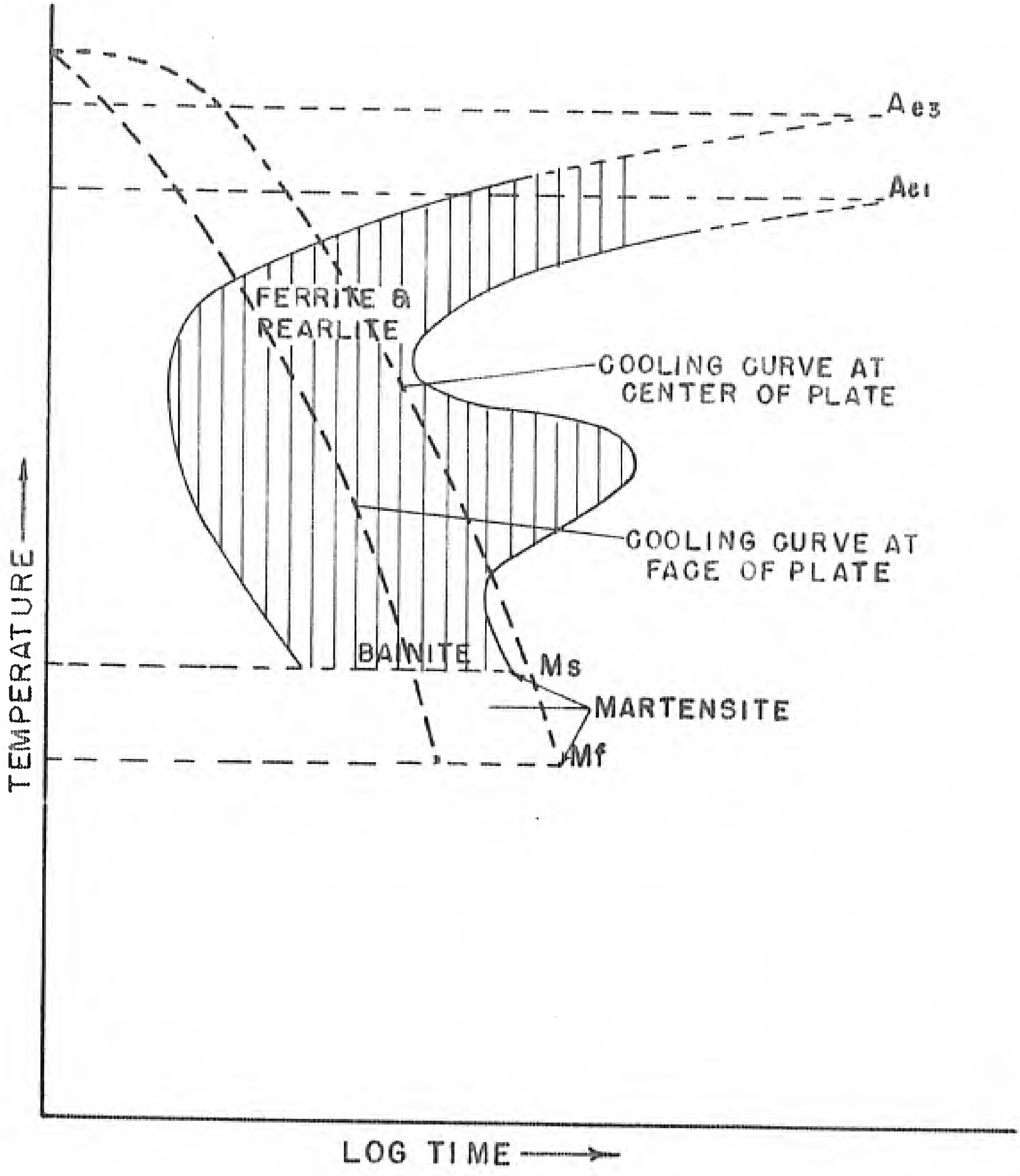


I - 40/16" from QE. Rc 41

Structures all etched with Picral. Mag. 1000X.

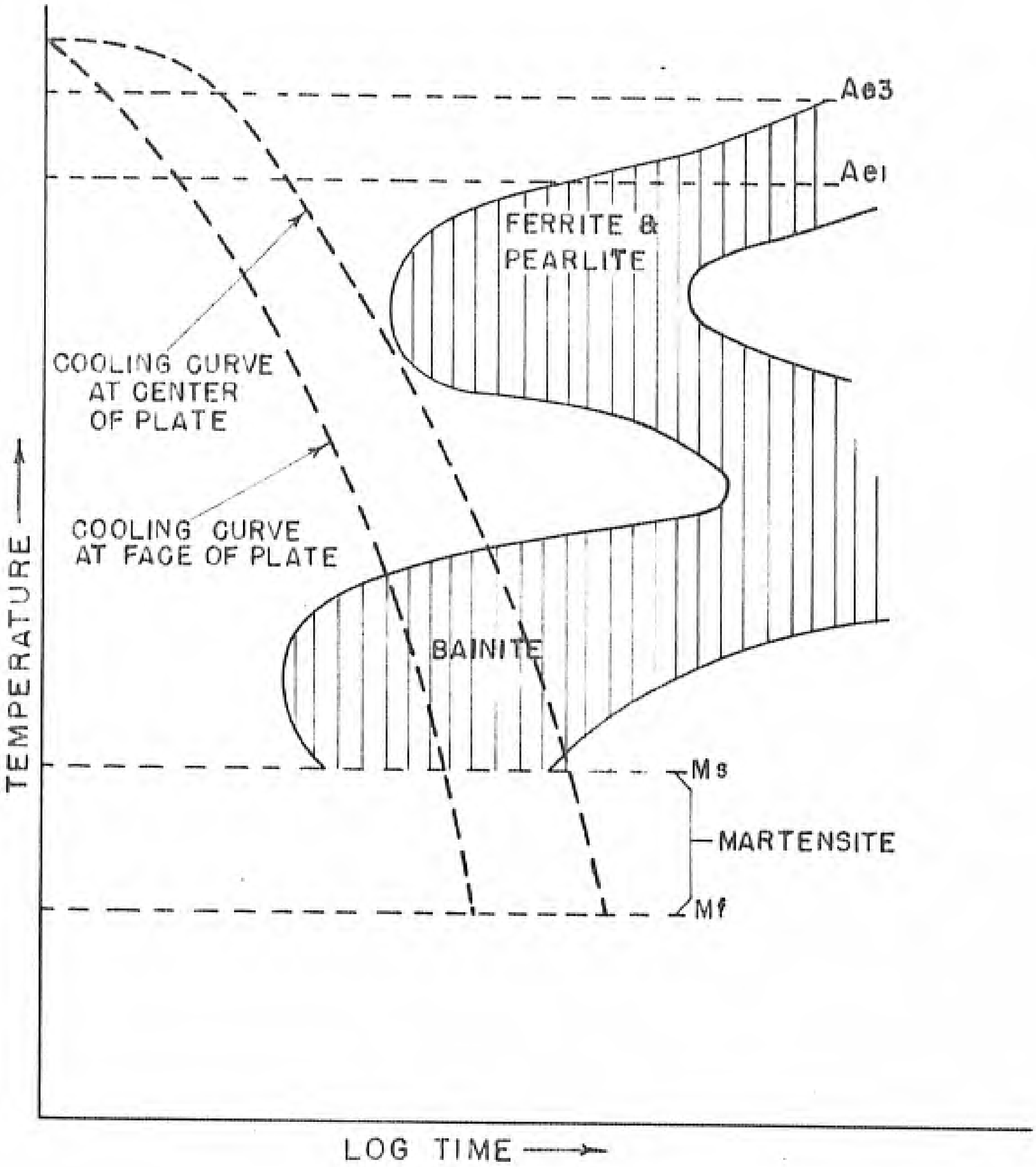
WTN.639-6441

FIGURE 2



SCHEMATIC DIAGRAM OF THE S-CURVE FOR
A LOW ALLOY ARMOR TYPE STEEL

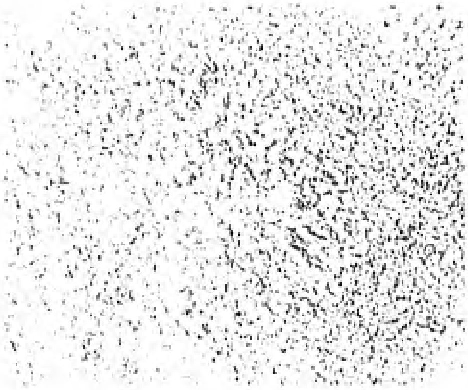
FIGURE 3



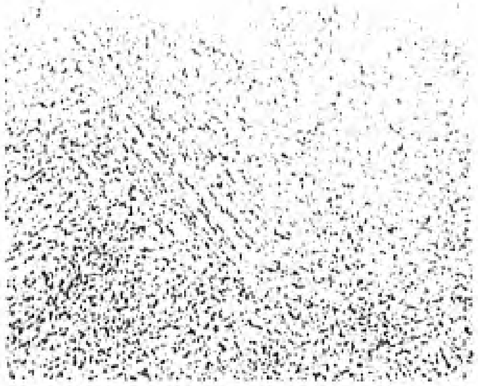
SCHEMATIC DIAGRAM OF THE S-CURVE FOR THE Mn-Ni-Cr-Mo-V TYPE STEEL

FIGURE 4

Microstructure of Heat Treated Mn-Cr-Ni-Mo-V Type Steel

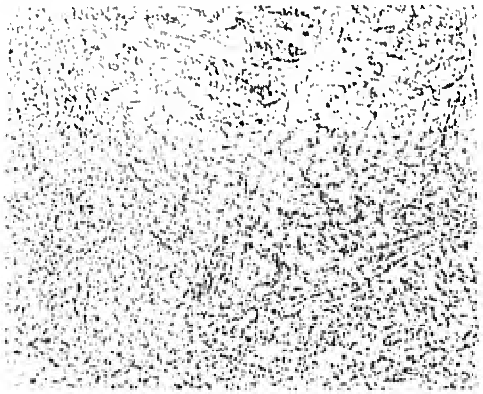


-A-



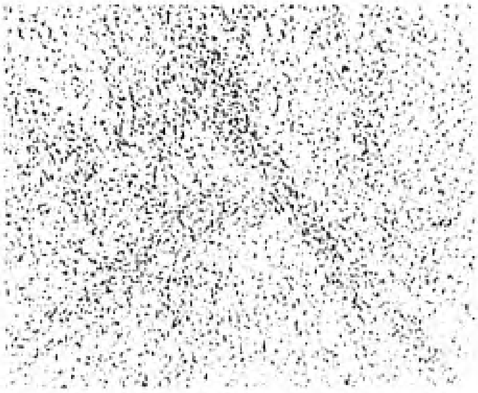
-B-

Structure of Block 18 heat treated at Watertown Arsenal using treatment No. 4. Hardness 235/241 BHN.



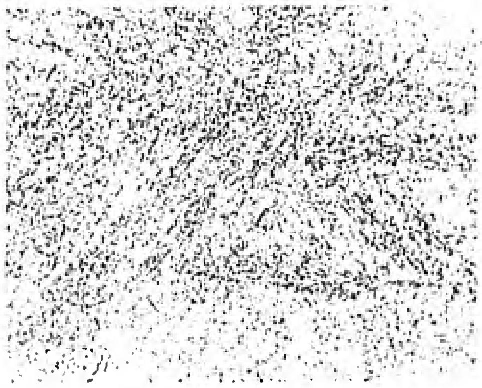
-C-

Structure of Block 18 in as-received condition.
Hardness - 235/241 BHN



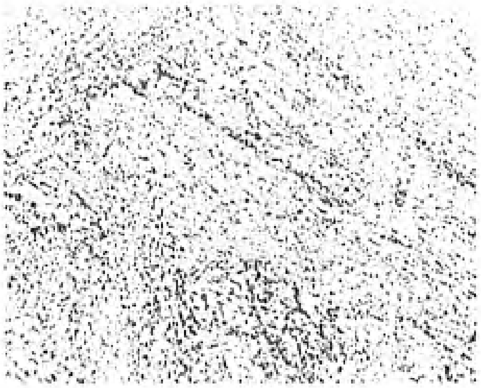
-D-

Structure of Block 20 after treatment No. 6.
Hardness - 229/235 BHN



-E-

Structure of Block 20 after treatment No. 5.



-F-

Structure of Block 20 after treatment No. 5.

Structures all etched with Picral. Mag. 1000X.

STANDARD TYPES OF NOTCHED BAR IMPACT FRACTURES

SYMBOL	DESCRIPTION
F	Fibrous
S	Silky (generally encountered with steels of high hardness)
Fc	Fibrous matrix with spots of crystallinity
Cdf	Dull crystalline patch surrounded by fibrous border
Cbf	Bright crystalline patch surrounded by fibrous border
Cd	Dull crystalline (complete)
Cb	Bright crystalline (complete)

Note 1: Additional terms such as: dendritic, conchoidal, etc., used to describe the fractures, should be written out in full following the fracture type symbol.

Note 2: If it is desired to estimate the relative amounts of fibrous and crystalline surface areas, a fraction will be placed following the fracture symbol. This fraction will refer to the estimated surface area which is crystalline.

FIGURE 6