UNCLASSIFIED

AD 260 693

Reproduced by the

ARMED SERVICES TECHNICAL INFORMATION AGENCY
ARLINGTON HALL STATION
ARLINGTON 12, VIRGINIA

UNCLASSIFIED

ENGINEERING
B-W
PRODUCTION

Borg-Warner Corporation

INGERSOLL KALAMAZOO DIVISION

1810 North Pitcher Street, Kalamazoo, Michigan

61-4-1
NOX

Project Report SPDIR-25
Contract NOrd 15719 Problem 4

High Strength MBMC #1 Steel Hydrotest Case Evaluation, Metallurgical Properties and Machinability

Copy No. 78
July 14, 1961

INGERSOLL KALAMAZOO DIVISION
Borg-Warner Corporation

Prepared by:

E. S. Mills
E. S. Mills
Asst. Project Engineer

Approved:

L. E. Hershey
L. E. Hershey, Manager
Special Projects Department

FOREWORD

This report has been prepared by the Special Projects Department of Ingersoll Kalamazoo Division, Borg-Warner Corporation, as a part of a development program being conducted in fulfillment of Bureau of Naval Weapons, Department of the Navy, Contract NOrd 15719 Problem 4.

This report covers the work completed during 1960 and consists of three separate reports; the fabrication and hydrotest of prototype MBMC #1 steel cases fabricated by the roll and weld and power spinning methods, the determination of MBMC #1 metallurgical properties, and a machinability study of MBMC #1 steel.

Appreciation is expressed to Professor W. R. Weeks, Western Michigan University, for his contributions in the area of heat treating and decarburization evaluation, the University of Michigan, Professor L. V. Colwell and Mr. K. N. Sederlund for their work on the machinability and the Roy C. Ingersoll Research Center personnel, particularly Mr. E. J. Klimek, for their work in evaluating the metallurgical properties of MBMC #1 steel.

CONTENTS

SCOPE OF CONTRACT

The 1960 program was directed toward the complete evaluation of MBMC #1 steel consisting of the determination of the physical and mechanical properties of MBMC #1 steel including machinability and toward the fabrication of prototype cases capable of consistent high strength fabricated by the power spinning process and the roll and weld process.

ABSTRACT

In an effort to provide the missile industry with information on an ultra high strength steel suitable for the fabrication of missile cases and components, Ingersoll Kalamazoo Division, Borg-Warner Corporation has conducted a fabrication investigation. This investigation covered the evaluation on fabrication of hydrotest cases by both rolled and welded, and hydrospin processes. The investigation has been limited to the machinability, metallurgy, welding, spinability and hydrotesting of MBMC #1, a high silicon, lean alloy steel. This material was originally proposed by the Missile Booster Materials Committee on May 25, 1956.

Submitted in this report are physical and mechanical properties of MBMC #1 steel for metallurgists and design engineers in applying this material to case fabrication. This information was compiled by Roy C. Ingersoll Research Center, Borg-Warner Corporation, Des Plaines, Illinois.

A machinability study was completed comparing MBMC #1 steel with AISI 4130 steel. The machinability study was conducted by the University of Michigan Research Institute at Ann Arbor, Michigan.

Fifty-five production prototype hydrotest cases were fabricated by Ingersoll Kalamazoo Division. Twenty-three were fabricated by the conventional rolled and welded process and thirty-two by the power spinning process. These units were used to determine the reproducibility of fabricated cases. The test proved that the material, with proper heat treatment, can produce a high strength steel case by the rolled and welded or hydrospun methods of

fabrication. The information given in this report outlines certain precautions to be observed during designing, fabricating and heat treating MBMC #1 steel.

CONCLUSIONS

It was determined from the investigation performed at Ingersoll Kalamazoo Division that:

A. General

1. Limited surface decarburization was necessary before consistent high strength results could be obtained on MBMC #1 steel cases fabricated by either spinning or roll and weld.

2. MBMC #1 steel cases fabricated by both spinning and roll and weld resulted in high burst strengths.

3. The spun cases exhibited the higher burst strengths, however, the development of the roll and weld cases was not completed. Upon completion of the roll and weld development similar results should be obtained.

B. Flash Weld Blank Hydrospun Cases

1. MBMC #1 steel rolled and flash welded into cylindrical blanks spun as readily as forged blanks.

2. The flash weld was faulty having excessive inclusions causing premature case burst results.

3. It was indicated that with a clean flash weld results similar to forged and spun cases would be possible.

C. Forged Blank Hydrospun Cases

1. MBMC #1 steel is readily hydrospun into cylinders.

2. MBMC #1 steel can be cold reduced by spinning up to 80% before stress relieve is required.

3. Dimensional tolerances of roundness after spinning and stress relieve are good.

4. Dimensional tolerances of roundness after heat treat increase by a factor of 3 to 4 without the use of restraining fixtures.

5. Cases spun from forgings and heat treated in a neutral atmosphere burst at a low stress level.

6. Cases spun from forgings and heat treated in a decarburizing atmosphere burst at consistently high stress levels.

7. Cases spun from MBMC #1 steel forgings heat treated to decarburize the surface and tempered at 500°F consistently burst at a stress level above 280,000 psi with a ratio of burst stress to uniaxial tensile strength from 1.10 to 1.21 with an average of 1.15.

D. Roll and Weld Cases

1. MBMC #1 steel can be readily rolled and machined in the annealed condition and welded using either MBMC #1 or AISI 6130 steel wire.

2. Cases heat treated in a neutral atmosphere will burst at a low stress level.

3. Heat treating the cases in a decarburizing atmosphere improved the burst strength levels approximately 30%.

4. Heat treating the case in a decarburizing atmosphere after removal of all scale and grinding the girth welds flush improved the burst strength an additional 14%.

5. Relocating the girth welds with respect to the end rings improved the burst strength level an additional 13%.

6. Burst strengths of 280,000 psi are obtainable in the rolled and welded MBMC #1 cases with a ratio of burst stress to uniaxial tensile strength above 1.0 and it was indicated that higher strengths and ratios are feasible.

7. The burst strength levels obtained throughout the development of the various factors affecting strength became more consistent as refinements were made indicating good reliability.

8. The substitution of spun tube sections girth welded to end rings resulted in burst strengths the equivalent of rolled and longitudinally welded tube sections indicating that the girth welds were the cause of the premature failures that occurred.

It was determined from the investigation performed at the University of Michigan that, although the MBMC #1 steel does exhibit some unique machining properties, it can be machined with no great difficulties, particularly in the as-forged condition. When it is hardened it cannot be cut readily with high-speed tools but it can be cut quite satisfactorily with both sintered carbide and ceramic tools.

The data collected during the investigation at the Roy C. Ingersoll Research Center shows that MBMC #1 can be heat treated to strength levels equal to or greater than those of currently used missile alloys.

These high strengths are accompanied by low tensile ductilities in most strength levels. As discussed in Section 2, these ductility levels will most likely be brought up with higher quality melts of the material.

Project Report SPDIR-25
Contract NOrd 15719 Problem 4

High Strength MBMC #1 Steel Hydrotest Case Evaluation

Prepared by

INGERSOLL KALAMAZOO DIVISION
Borg-Warner Corporation

MBMC #1 Steel Hydrotest Case Development

Abstract

Fifty-five production prototype hydrotest cases were fabricated by Ingersoll Kalamazoo Division. Twenty-three of these were fabricated by the conventional rolled and welded process and thirty-two by the power spinning process. These tests proved that MBMC #1 steel, with proper heat treatment can produce a high strength steel case by either the rolled and welded or hydrospin methods of fabrication. Certain precautions, however, have to be observed during design, fabrication and heat treatment.

TABLE OF CONTENTS

LIST OF ILLUSTRATIONS

List of Illustrations continued

LIST OF TABLES

I. ENGINEERING OF HYDROTEST CASE

A. Design of Hydrotest Cases

Test cases for two basic types of fabrication were designed for this investigation. These were the hydrospun tube with integral end rings, Figure 1 and the rolled and welded tube, girth welded to machined end rings as shown in Figure 2. The hydrospun cases were made from two types of cylindrical blanks, one of forged material and the other of rolled and flash welded material. The design of both type cases was standardized to permit direct correlation and comparison of all test data. The physical dimensions of the cases were standardized to make use of existing tooling, fixtures, and equipment to the greatest extent practicable. The cases are nominally 16.210" inside diameter by 52" in length. The diameters of the end rings were designed to allow machining stock on the inside and outside diameters of the hydrospun end rings after spinning. The end rings for the rolled and welded cases were held to the same configuration. The cases were identical except for the method of fabrication.

B. Material of Hydrotest Cases

The material used to fabricate all cases under this investigation was MBMC #1 steel. The sheet stock, .050" thick, for the rolled and welded cases was available from a previous contract, NOrd 15719. The analysis of the sheet stock was:

Carbon	0.425%
Manganese	0.76

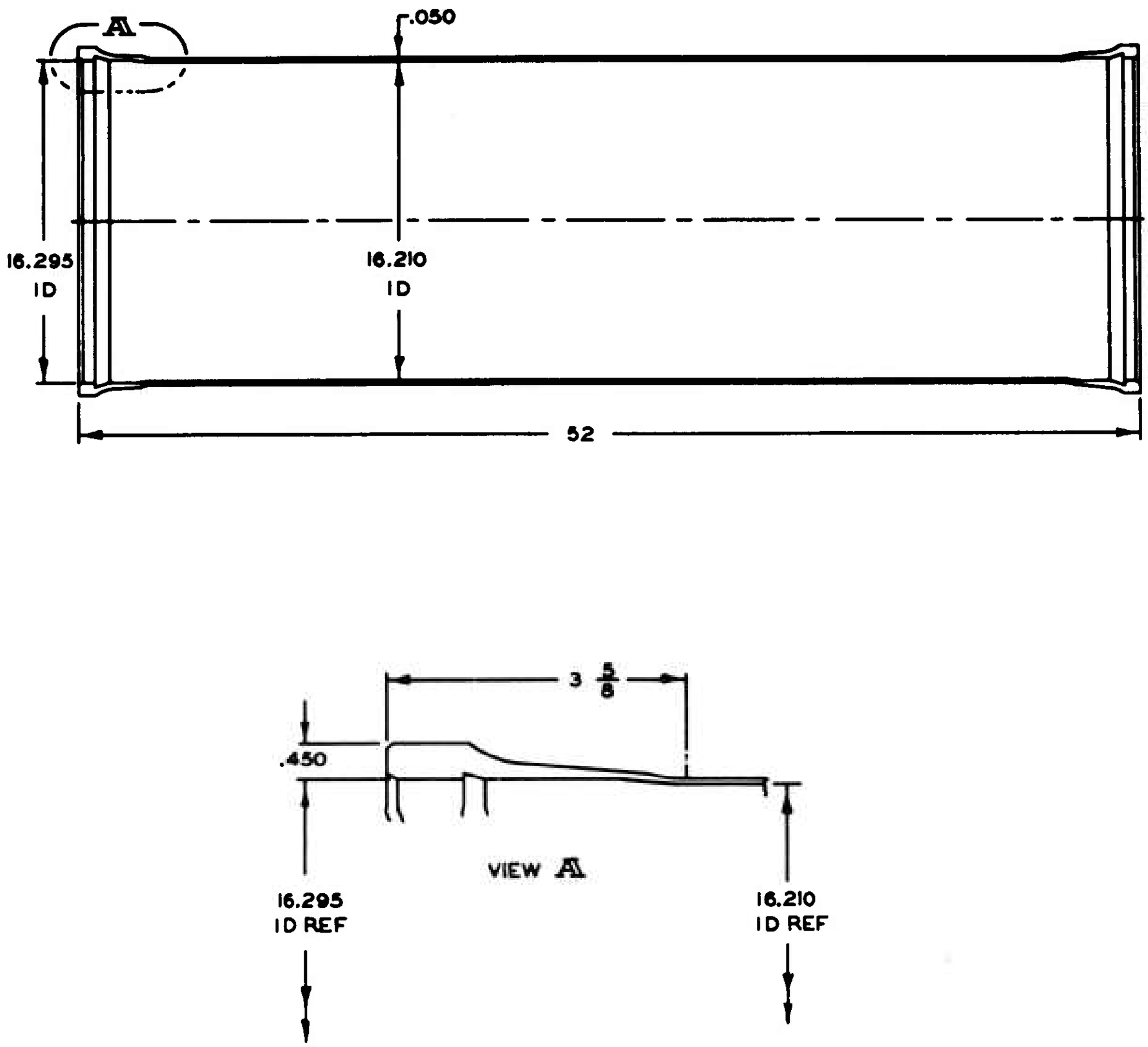


Figure 1. Hydrospun Case Design

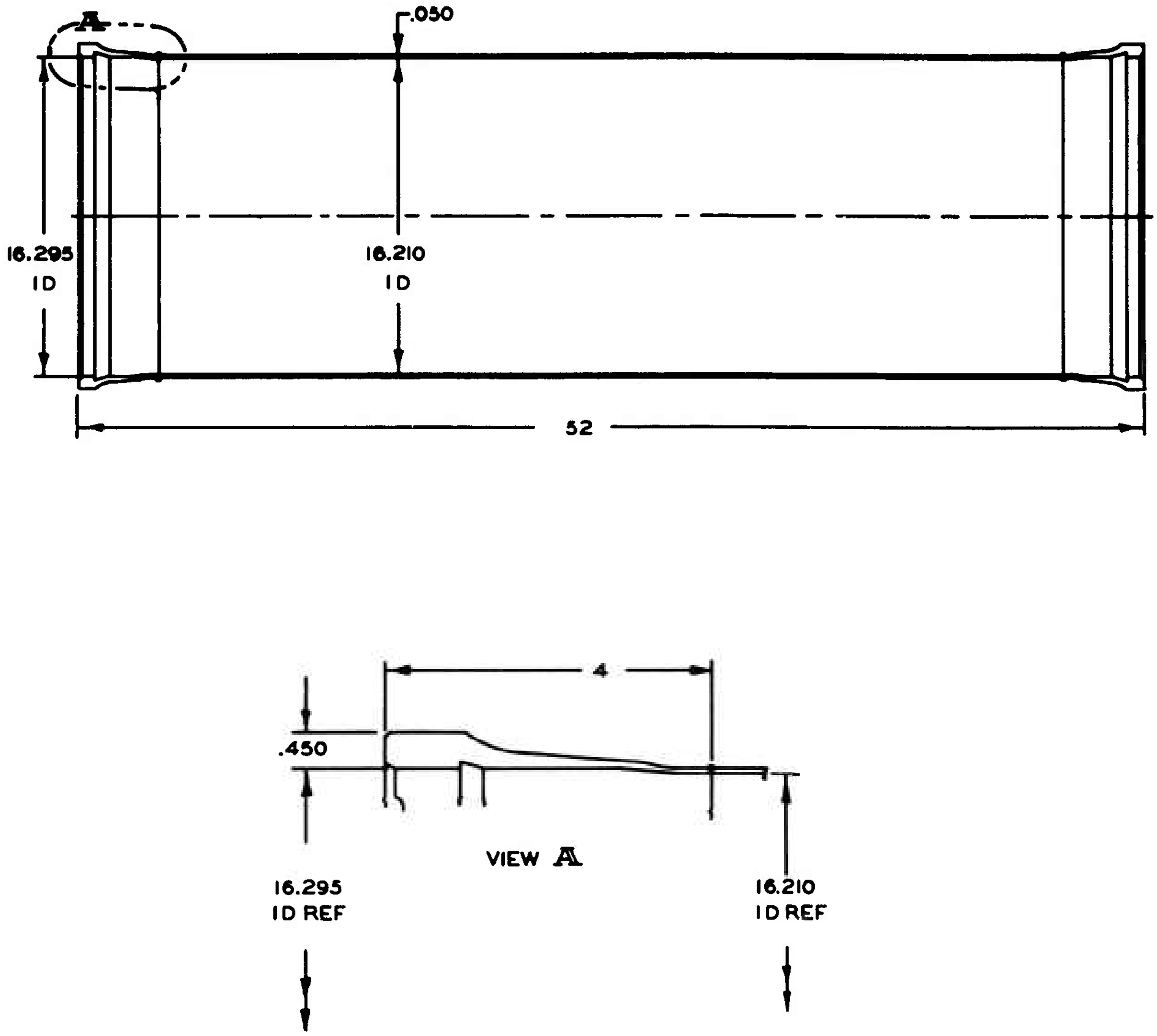


Figure 2. Roll and Weld Case Design

Phosphorus	0.019%
Sulphur	0.020
Silicon	1.42
Chromium	0.76
Vanadium	0.08

The MBMC #1 material used for hydrospinning and laboratory testing was produced by Bethlehem Steel Company. The analysis of this steel melt was:

Carbon	0.42%
Manganese	0.80
Phosphorus	0.015
Sulphur	0.007
Silicon	1.65
Chromium	0.75
Vanadium	0.07

This melt of steel was processed into three lots. One lot was forged into cylindrical tube blanks for the hydrospun cases and to serve as end rings for the rolled and welded cases. The second lot was forged into slabs which were rolled into 3/4" thick plates to be used in fabricating rolled and flash welded blanks for additional hydrospun cases. The third lot was forged into machinability and end quench test bars.

Machinability and end quench test bars of AISI 4130 steel were received from another source. The 4130 steel bars were used as a basis for comparing machinability with MBMC #1 steel.

The forged slabs 3/4" thick were rolled and flash welded into nominally 16" ID by 14" long cylinders by the American Welding and Manufacturing Company of Warren, Ohio.

The forged and flash welded material for case fabrication was spheroidize annealed 90-95%. One forged blank was sectioned to obtain the grain flow pattern, Figure 3.

C. Tooling

Operation process charts were prepared for each of the two methods of fabrication and are presented in Figures 4 and 5. Using these basic operations as a guide, tooling was prepared.

The welding equipment consisted of an Airco Model HMH-E Welding Head, Airco Model HM-C-B Control Panel, Westinghouse Type RA-DC Generator, Airco Heliarc Filler Wire Feeder, Stock No. 2310000, and Berkeley Type EG-3 Electronic Governor controlling welding head travel. An experimental longitudinal welding fixture from previous work was used for the longitudinal welds, after reworking the depth of the relief groove and the diameter contour of the copper back-up bar. The girth weld fixture was reworked from a previous fixture to accommodate the case diameter and was used in conjunction with a Ransome positioner.

The hydrospinning was performed on a 42 x 50 Hydrospin machine manufactured by Cincinnati Milling and Grinding Machine Company. The only special tooling required for spinning the hydrotest cases was a spinning mandrel, clamp ring, and tool rings. Both the mandrel and tool rings were reground from existing tooling. The clamp ring was designed and fabricated to secure the blank on the mandrel during spinning.

A machining mandrel was designed and built for machining the hydrospun cases and for trimming the rolled and welded tubes prior to the girth

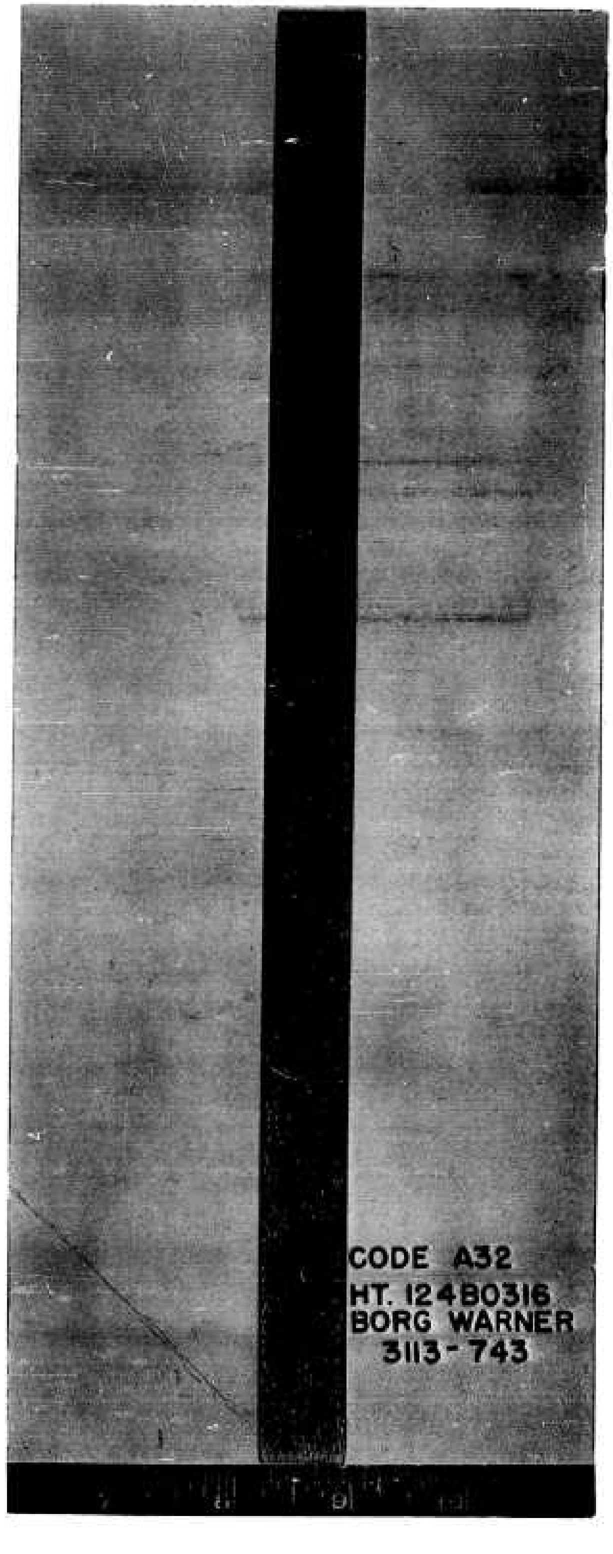


Figure 3. Grain Flow Pattern - Forged Cylindrical Spinning Blank

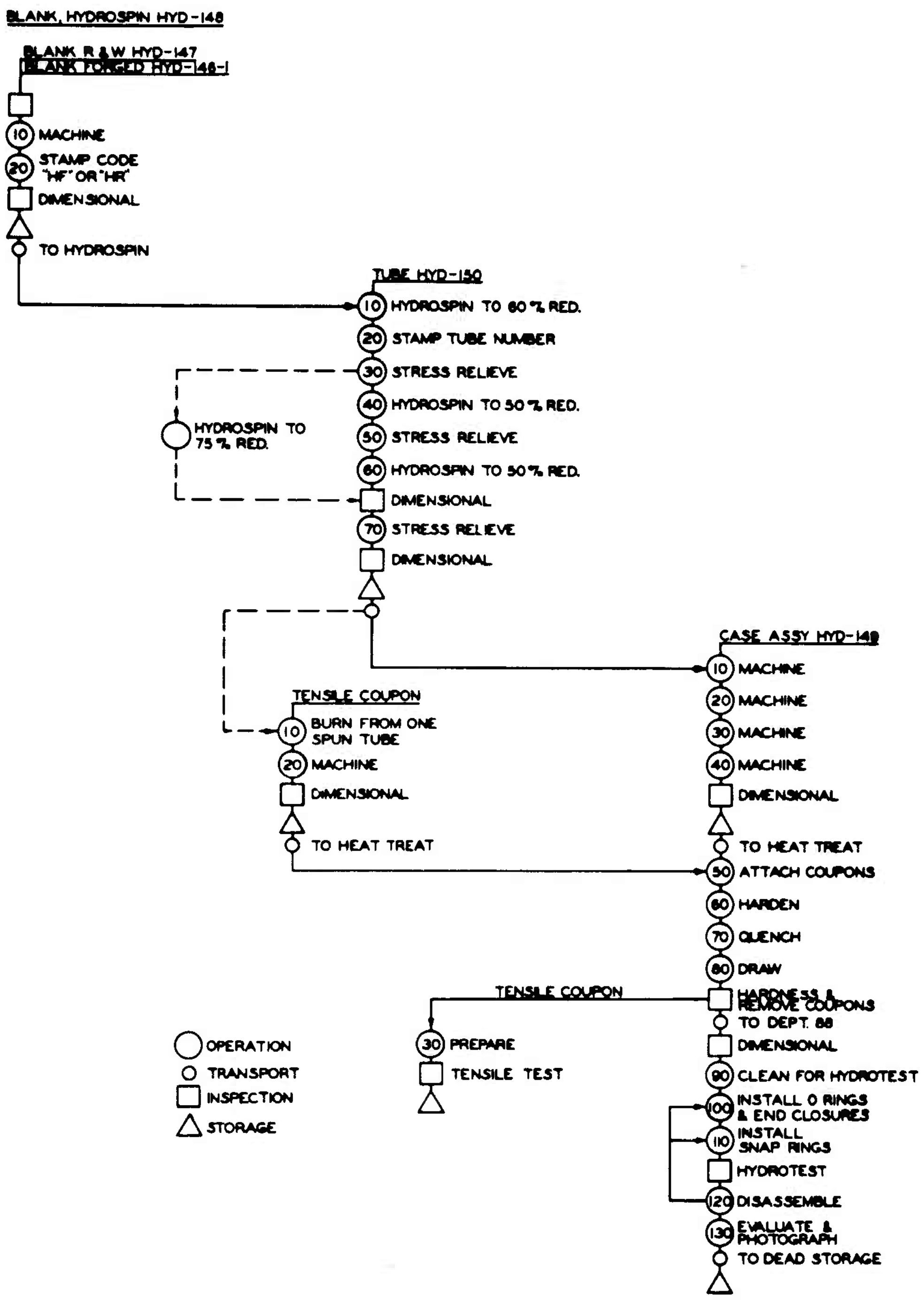


Figure 4. Operation Process Chart - Hydrospun Case

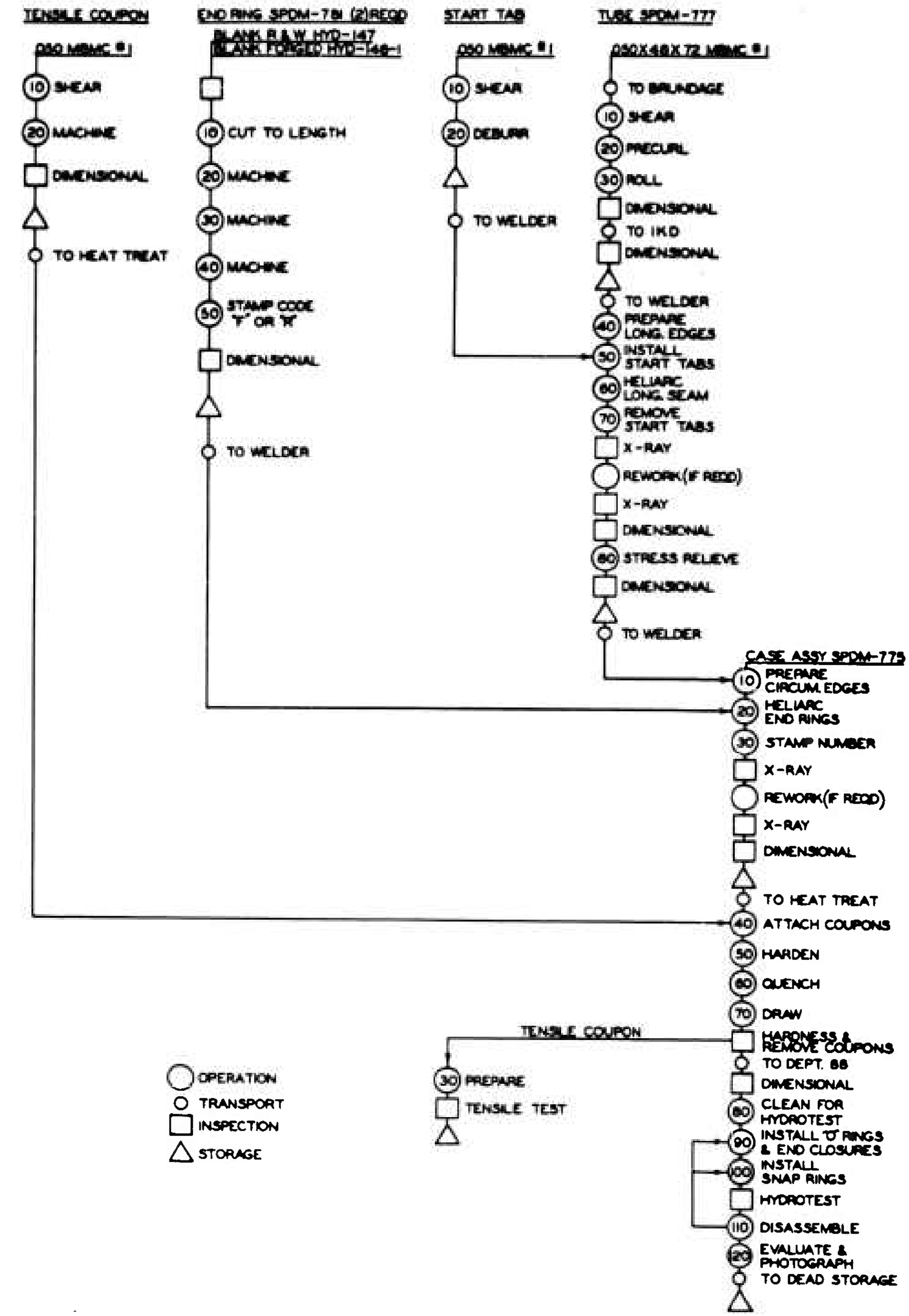


Figure 5. Operation Process Chart - Roll and Weld Case

welding of the end rings. The end rings for the rolled and welded cases were machined to contour with the aid of a contour template and a back-up plate for chucking purposes. No other special tooling was required for machining the hydrospun blanks.

Three heat treat fixtures were reworked to hold the cases in a vertical position during the hardening and quenching operations. No special fixtures were used to minimize out-of-roundness or distortion due to the heat treat operations.

Nike booster heads were modified to serve as end closures for hydro-test purposes. Each head was held by a 4130 steel snap ring.

D. Processing

The processing through fabrication of the spun and rolled and welded cases were identical for all common operations such as machining, heat treat and hydrotesting. The variables were the basic method of assembly fabrication: one group being assembled by roll and weld techniques and the other by power spinning an integral tube and end rings. The physical configuration of the final cases was identical except for the weld seams.

All operations were performed by production personnel supervised by engineering personnel assigned to this project.

1. Machining

All machining was accomplished in a lathe and prior to the heat treat of the part. A single point tool bit was used at medium speeds and feeds. The surface finish was held to 125-150 microfinish. No particular difficulty was encountered in any of the machining operations.

The following procedures were used to accomplish the various operations:

Rough cut .014" feed, 110 rpm, up to 1/4" depth
Finish cut .014" feed, 220 rpm, .010" depth
Tool bit 78B Carboloy
Tool life 3 pieces.

2. Heat Treat

All finished cases were heat treated in a vertical drop bottom hardening furnace, electric, with a Lindberg endothermic atmosphere generator. The quench was oil at 125°F and circulated at 1000 gpm minimum. The tempering was accomplished in a gas fired Lindberg draw furnace. The atmosphere was varied to obtain various amounts of surface decarburization after initial neutral atmosphere treatments did not result in the strength levels desired.

From some previous work done by Ingersoll Kalamazoo Division, and work being done by others, it was felt that partial decarburization of the cases would be beneficial. This condition appears to counteract the destructive effect of minute cracks by increasing the amount of energy that is required to propagate cracks into catastrophic failures. Lack of detailed quantitative and qualitative decarburization information necessitated the collection of data during the work on the program. Samples from each hydrotested tube were microscopically examined to determine the visual depth of decarburization.

In addition to the visual examination, the samples taken from the cases were subjected to microhardness surveys. A Kentron microhardness tester was used for this study using a 100-gram load with Knoop indenter. The microhardness readings were plotted against the depth of the reading from the surface and representative curves drawn for the transition of the hardness from the surface to the core. The microhardness decarburization depth was established as that depth where the hardness was 5% less than the average core hardness. A minimum of five core hardness readings were used to establish the core. This method was not entirely satisfactory due to some instances where a considerable variation in the core hardness readings was encountered.

An analysis of the data presented in this report indicates that many more samples must be studied and reported prior to forming any conclusions regarding the optimum decarb depth or range of decarb depth.

Tensile coupons were heat treated along with each tube to serve as control samples to indicate the strength level of the case. The tensile specimens were made from the sheet stock for the rolled and welded cases and from a spun forged blank and spun flash welded blank for the respective power spun cases. These specimens were wired to the outside of the case and held away from the case surface. On any future cases it would be recommended that the specimens be wired to the holding fixture for the tube to preclude the possibility of non-uniform decarburization at points of contact of the holding wire with the case surface. There was no direct evidence that contact of the wire was a deterrent factor in the strength level obtained on the cases but it is recommended as a precautionary measure in view of the apparent beneficial effect of surface decarburization.

3. Hydrotest

The cases were hydrotested in an identical manner using identical end closure heads. The water pressure was built up with a reciprocating type pump. The pressure was taken on a recorder and checked by a visual pressure gage. Since the end rings were not perfectly round after heat treat some pounding with a mallet was required to drive the head in place. There was no apparent effect on the strength level whether the head was difficult to position or relatively easily positioned. The yield pressure was not evident by visual observation of the pressure gage or recorder and therefore no yield strength record was obtained.

4. Welding

Preliminary welding tests were conducted using .050" thickness sheet stock to determine the exact weld settings and to try out the equipment and fixtures for conducting the final welding. The method of evaluation was X-ray examination for weld soundness and tensile test results for the strength level of the specimen. The weld wires considered were MBMC #1 and AISI 6130 steel.

The longitudinal welds were first tried on a weld fixture normally used for the production welding of .100" thickness sheet. The back-up groove which measured .020-.025" deep by 5/16" wide proved to be unsatisfactory for this thinner material permitting excessive fall through. Also, the hold-down fingers did not function properly with this thinner material permitting misalignment of the parts after welding began. Welding was then transferred to another fixture having screw type hold down clamps and the copper back-up bar was machined to provide a groove .010-.015" deep and 5/32" wide. This fixture proved satisfactory.

Weld tests were conducted on .050" sheet by 8" wide by 48" long formed to the radius of curvature of the final 16.2" ID cases. Heliarc welds were made using .045" diameter AISI 6130 and MBMC #1 steel wires under a variety of wire feeds, welding head travel and current variations.

The settings which gave the best results are shown in Table I.

Twelve tensile specimens free of flaws as determined by X-ray examination were machined from weld test samples of both the AISI 6130 wire and the MBMC #1 wire. All specimens were austenitized at 1600°F and quenched in oil. The specimens were tempered at 600°F for 1 hour.

Table II shows the tensile test results of the coupons welded with AISI 6130 wire. The first six were pulled as welded and the last six had the weld ground flush with the parent metal before test. All specimens broke at the interface between the parent metal and the weld metal.

Table III shows the test results of the coupons welded with MBMC #1 wire. In the as welded condition, the MBMC #1 wire produced a weld 10% stronger than the AISI 6130 wire and 5.5% stronger with the weld ground flush. Nine of the twelve coupons welded with MBMC #1 wire broke at the interface, two in the base metal and one diagonally across the weld.

MBMC #1 weld wire was chosen for use on the first cases and a number of tubes were longitudinally welded using the information obtained from

Table I. Weld Settings and Procedure AISI 6130 and MBMC#1 Steel Wire

	AISI 6130	MBMC#1
Thickness	0.050	0.050
Wire Dia.	0.045	0.045
Weld	Tungston Helium Shielded	Tungston Helium Shielded
Speed	9-3/8"/min.	9-3/8
Amperage	50	55
Voltage	14	14
Sensitivity	5	5
Wire Feed	9-1/4	15-1/2
Gas Rate	30 CFM	30 CFM
Preheat	300^{o}F	300^{o}F
Post heat 10 min.	600^{o}F	600^{o}F
No. Passes	1	1

Table II. Weld Strength .050" MBMC#1 Sheet, AISI 6130 Weld Wire .045" Diameter

Specimen No.	Yield Strength 0.2% Offset psi	Ultimate Strength psi	Elongation % in 2"	Hardness R_c
1		272,500	2.0	51
2		256,900	2.0	51
3		257,800	1.5	51.5
4		267,800	2.0	51.5
5		252,800	2.0	51
6		264,800	2.0	51
Average		262,100		
Highest		272,500		
Lowest		252,800		
7	244,600	269,900	2.0	51
8	245,000	274,100	2.0	50.5
9	248,700	281,000	3.0	51
10	254,000	275,800	1.5	51
11	254,900	273,400	1.5	51.5
12	266,000	273,800	2.0	51
Average		274,700		
Highest		281,000		
Lowest		269,900		

$\frac{\text{Yield}}{\text{Ultimate}}$

Average	0.92
High	0.97
Low	0.89

Table III. Weld Strength .050" MBMC#1 Sheet, MBMC#1 Weld Wire .045" Diameter

Specimen No.	Yield Strength 0.2% Offset psi	Ultimate Strength psi	Elongation % in 2"	Hardness R_c
1		272,400	4.0	50.5
2		272,000	4.0	50
3		291,200	3.5	50.5
4		300,000	3.0	50.5
5		297,500	3.0	51
6		299,200	3.5	51
Average		288,700		
Highest		300,000		
Lowest		272,000		
7	250,000	286,100	3.5	51
8	257,200	288,000	3.0	51
9	254,600	292,400	3.0	51
10	272,700	295,400	2.5	51
11	269,600	286,300	3.5	50.5
12	254,000	292,000	3.5	51
Average		290,000		
Highest		295,400		
Lowest		286,100		

$\frac{\text{Yield}}{\text{Ultimate}}$

Average	0.90
High	0.94
Low	0.87

the preliminary welding tests. X-ray examination, using Eastman type "M" film, of the first tubes welded indicated some porosity along the edge of the weld and some other very minute discontinuities which could not be readily identified. Some cracks were induced by a delay in releasing the hold down clamps on the fixtures after welding. This was corrected by releasing the clamps about 8" behind the weld as the welding head progressed. The welds were stress relieved immediately after the post heat operation.

One tube containing porosity was sectioned, tensile specimens machined and heat treated to determine the effect of porosity on the strength level of the weld joint. All tensile specimens were X-rayed prior to tensile testing. The X-rays were examined with a magnifying optical comparator to determine the type and size of defects. The X-ray film revealed a reasonable number of variations in the size and number of voids and discontinuities in various specimens to permit a rough evaluation of the strength with respect to these potential problem areas.

Table IV is an analysis of the specimens examined and Table V shows the physical properties results obtained on these specimens. It was evident that porosity adversely affects the elongation and raises the yield and tensile strength. It was also evident that a limited amount of porosity can be tolerated. Based on this rough analysis maximum limits were set for the final cases as follows:

1. No discontinuity permitted; that is, no cracks or series of voids permitted in such proximity that the overall effect would be a continuous void greater than 0.015" in length and/or diameter.

Table IV. X-ray Analysis of Weld Tensile Specimens MBMC#1 Weld Wire

1.	O.K.	
2.	Porosity:	Scattered porosity which is dense and where three voids are approximately 0.015" daiamter. Porosity on one side of weld.
3.	Porosity:	Scattered porosity which is light and where the largest voids approach 0.010" diameter. Porosity on one side of weld.
4.	Porosity:	Scattered porosity which is dense and with numerous voids approximately 0.020" diameter. Porosity on both sides of weld.
5.	Porosity:	Scattered porosity which is light and where the largest voids are less than 0.010" diameter. Porosity on both sides of weld.
6.	Porosity:	Scattered porosity which is light and where the largest voids are approximately 0.010" diameter. Porosity on one side of weld.
7.	Porosity:	Scattered porosity which is dense and where the largest voids are 0.030" diameter. Porosity on both sides of weld.
8.	Porosity:	Scattered porosity which is dense and where the voids are approximately 0.010" diameter. Discontinuity in weld. Porosity on one side of weld.
9.	O.K.	
10.	O.K.	
11.	O.K.	
12.	O.K.	
13.	Porosity:	Scattered porosity which is light and where the voids are less than 0.010" diameter. Porosity on one side of weld.
14.	Porosity:	Scattered porosity which is light and where the voids on one side of the weld are approximately 0.010" diameter. Porosity on both sides of weld.
15.	Porosity:	Scattered porosity which is light and where one void has a diameter of approximately 0.015". Porosity on both sides of weld.
16.	O.K.	
17.	Porosity:	Scattered porosity which is dense and where the largest voids have a diameter of 0.020". Porosity on both sides of weld.
18.	Porosity:	Scattered porosity which is dense and where the largest voids have a diameter of 0.015". Porosity on both sides of weld.
19.	Porosity:	Scattered porosity which is dense and where the largest void has a diameter of 0.015". Porosity on both sides of weld.
20.	Porosity:	Scattered porosity which is light and where the largest voids are less than 0.010" diameter. Porosity on both sides of weld.
21.	O.K.	
22.	Porosity:	Scattered porosity which is light and where the diameter of the largest voids is less than 0.010". Porosity on one side of weld.

Table V. Physical Properties Weld Tensile Specimens
MBMC#1 Weld Wire

Specimen No.	Yield Strength 0.2% Offset psi	Ultimate Strength psi	Elongation % in 2 "	Hardness R_c
1	250,800	275,700	3.0	50.9
2	255,500	278,300	2.0	51.0
3	256,600	277,400	0.5	51.4
4(1)	-	241,000	2.0	50.5
5	259,500	281,100	2.0	51.5
6	259,700	282,600	1.5	51.0
7(1)	-	187,300	1.0	51.0
8(1)	-	227,500	1.0	51.0
9	252,800	275,100	3.5	51.0
10	255,100	269,300	2.0	51.0
11	248,400	273,400	3.5	51.0
12	251,100	275,100	4.5	51.0
13	256,200	277,100	2.0	51.0
14	260,500	280,500	2.0	51.0
15	259,600	281,200	2.5	51.0
16	257,300	277,200	3.0	51.0
17(1)	-	206,600	0.5	51.0
18(1)	-	252,100	0.5	51.0
19	255,800	264,200	1.5	51.0
20	256,500	274,600	2.0	51.0
21	246,200	265,800	2.5	51.0
22	261,300	280,200	3.0	51.0
Avg.	255,500	275,800	2.4	51.0

NOTES:

Heat Treatment

Harden: 1600°F - 1 hr. - oil

Temper: 600°F - 2 hrs.

All fractures in weld with the exception of specimen No.12 which fractured in the base material. (1) not figured in averages.

2. Scattered porosity on both sides of the weld is permitted provided that it is light in density and that the size of any one void does not exceed 0.015" diameter.

3. Dense porosity can be permitted provided that it is limited to one side of the weld and the size of the largest void does not exceed 0.015" diameter.

5. Spinning

A preliminary hydrospin investigation was conducted on six forged blanks. The material was spheroidize annealed and machined into the hydrospin blank shown in Figure 6. All preliminary spinning was done by forward spinning using standard .140" bite tube spinning tool rings.

The first blank spun, F-1, was successfully spun to 60% cold reduction which represents a change in wall thickness from 0.500" to 0.200". This was accomplished in three passes of approximately 0.100" actual bite each. In attempting to further reduce the wall thickness to 80% total reduction in one pass the material failed.

The second blank tested, F-2, was successfully spun to 60% reduction and then a portion of the tube was spun to 80% reduction in one pass to verify the results obtained in blank F-1 and the remainder of the tube spun to 80% reduction in two smaller passes. Spinning from 60 to 80% wall reduction in two passes produced no visible defect in the material. An attempt was made to further cold reduce the wall thickness to 88% total reduction but the material failed.

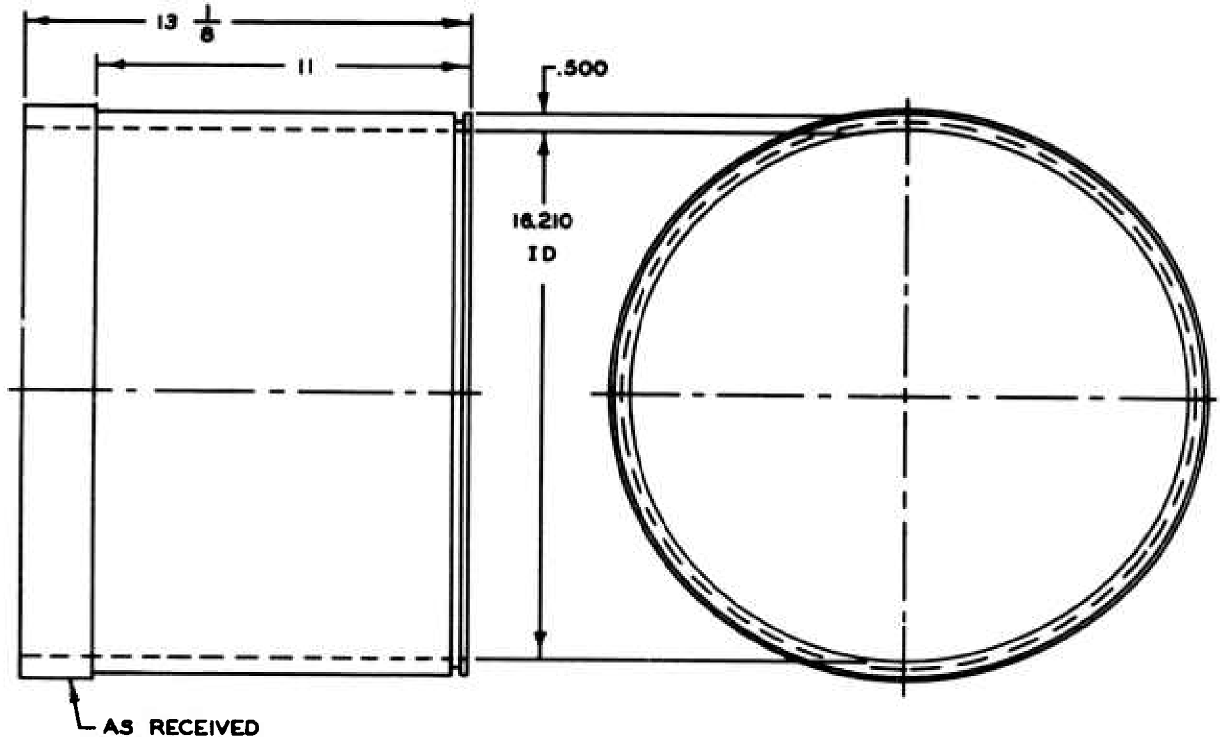


Figure 6. Hydrospin Blank Design

Using the spinning information gained from tubes F-1 and F-2, the remaining four blanks were hydrospun in two series. The first series of blanks, F-3 and F-4, were spun to a 60% cold reduction in passes of 25, 40 and 60% after which they were stress relieved at 1250°F for one hour at heat, and then spun to 80% and the final 90% reduction. The percentage of reduction is calculated on the basis of the original wall thickness of 0.500". The cold reduction after the stress relieve operation was 75%. The final spun blank with material on each end for machining the end rings is shown in Figure 7.

The last two blanks, F-5 and F-6, were spun to 80% cold reduction in passes of 20, 40, 60, 70 and 80%, after which they were stress relieved at 1250°F and then spun to the final 90% reduction. Six passes were required to spin this series whereas the first series required only five.

All tubes spun satisfactorily except F-5 which failed due to a lap caused by a spinning error.

Table VI shows the hardnesses recorded after each series of cold reductions taken on tube F-2. The increased hardness in attempting to reach 88% cold reduction resulted in material failure. The effectiveness of the stress relieve operation onhardness is also recorded for each reduction. All specimens were stress relieved at 1250°F for one hour at heat.

Based on this information it was concluded that MBMC #1 material can be hydrospun to 80% wall reduction without intermediate stress relieving. However, for production reliability it is recommended that the material be

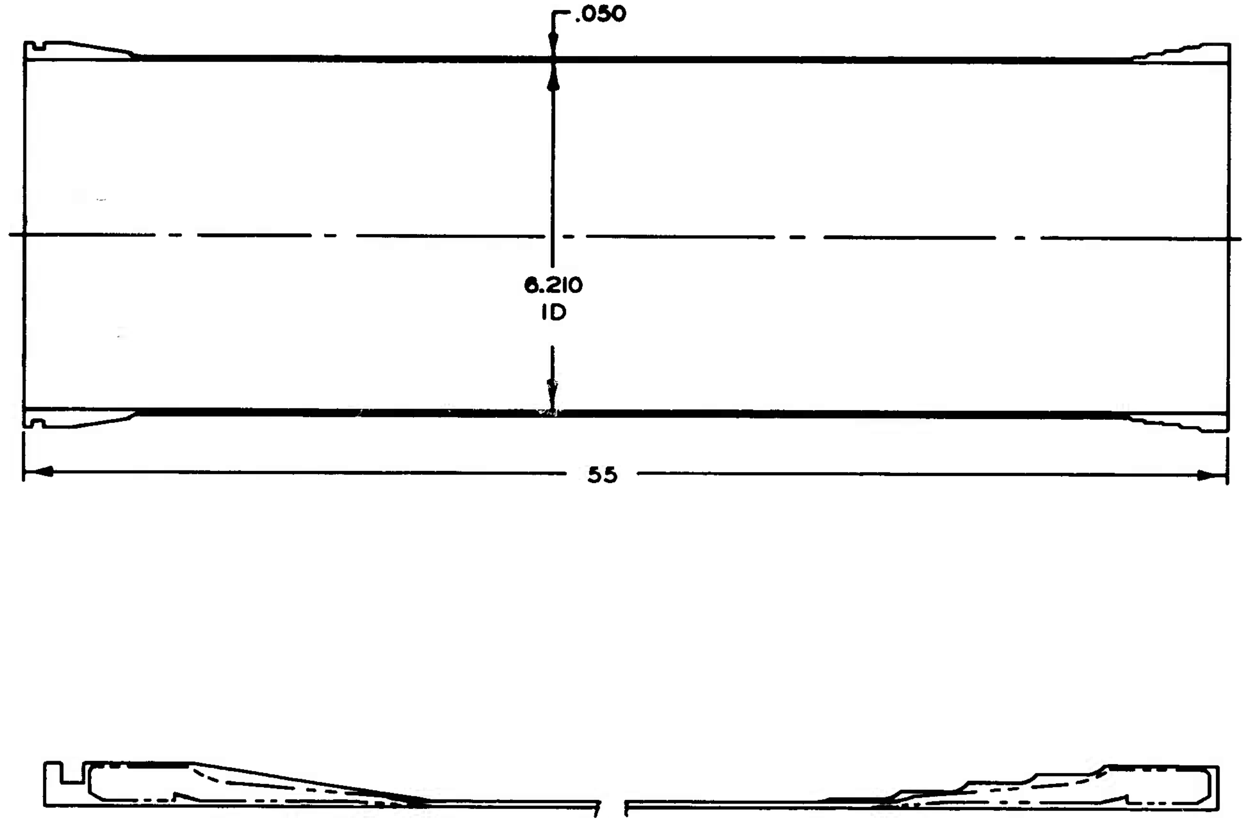


Figure 7. Hydrospun Case - After Spinning

Table VI. Hardnesses Recorded After Various Series of Cold Reduction and After Stress Relieve

Per Cent Wall Thickness Reduction per Pass	Hardness As Spun ID	Hardness As Spun OD	Hardness After Stress Relief ID	Hardness After Stress Relief OD	Number of Passes
Spheroidize Annealed	17	17	17	17	0
0, 20	26	31	13	16	1
0, 20, 40	31	33	15	17	2
20, 40, 60	32.5	36	16	18	3
20, 40, 60, 70	35	37	17	19	4
20, 40, 60, 80	35.5	39.5	18	19	4
20, 40, 60, 70, 80	37	39.5	18	19	5
20, 40, 60, 70, 80, 88	40	41.5	18	18.5	6

All material was forged MBMC#1.

All hardnesses Rockwell C scale.

stress relieved at 1250°F for 1 hour at heat, prior to spinning more than 60% reduction. Therefore, all cases beyond F-6 in this report were hydrospun as follows: 20, 40, 60% reduction, stress relieved, 50% reduction, stress relieved, 50% reduction, stress relieved, for a total of 90% wall reduction.

II. RESULTS OF FABRICATION AND TEST

Fifty-five hydrotest cases were fabricated, 23 by the conventional roll and weld process and 32 by hydrospin. Each case was heat treated and hydrotested. The effects of welding, hydrospinning, heat treating, decarburization, tempering temperature, and case design were evaluated for MBMC #1 steel. The geometry of the cases was standardized so that direct correlation of all test results was feasible. The fabrication of all cases began simultaneously and all spuncases were completed ready for heat treat as required to provide flexibility of processing if necessary.

A. Hydrospin Cases

The MBMC #1 steel hydrospun blanks of both the forged and flash welded material were hydrospun into tubes as previously shown in Figure 7. The method of spinning is described under "Engineering Study" of this report. Five passes were required to spin the 0.500" blank into a 0.050" wall tube. The blanks were stress relieved at 1250°F for 1 hour at heat, after the third, fourth and last passes except for cases F-3, F-4, F-5 and F-6 as noted previously.

All case end rings were machined to the final configuration prior to heat treat and no machining was attempted after heat treat. An expanding type holding fixture was used for machining to round the tube section adjacent to the end ring in order to obtain a smooth blend from the end ring into the tube.

A dimensional inspection consisting of TIR readings was made after the spinning, stress relieve and machining operations and after the heat treat operations. The setup used for this inspection is illustrated in Figure 8. The case was positioned on rollers on each end ring and revolved. The readings were made using a dial indicator positioned along the top side of the case. Table VII shows the readings recorded at the positions shown in Figure 8 for a number of the hydrospun cases. It was noted that a greater out of round occurred at the end of the case that was quenched last in the heat treat operation and that distortion was increased by a factor of 3 to 4 due to heat treat.

Heat treat of the hydrospun cases involved the following operations:

a. Light vacublast of the case and tensile specimens to remove foreign matter and scale.

b. Assemble the case and accompanying tensile specimens to a heat treat fixture. The case was heat treated in a vertical position and the fixture was a simple rod with cross members for support of the bottom end of the case.

c. Preheat the assembly at 800°F.

d. Austenitize at 1600°F for 65 minutes at heat and with the atmosphere variable.

e. Quench in oil at 125°F.

f. Temper for 3 hours at heat. Tempering temperatures were varied from 400°F to 700°F as tabulated.

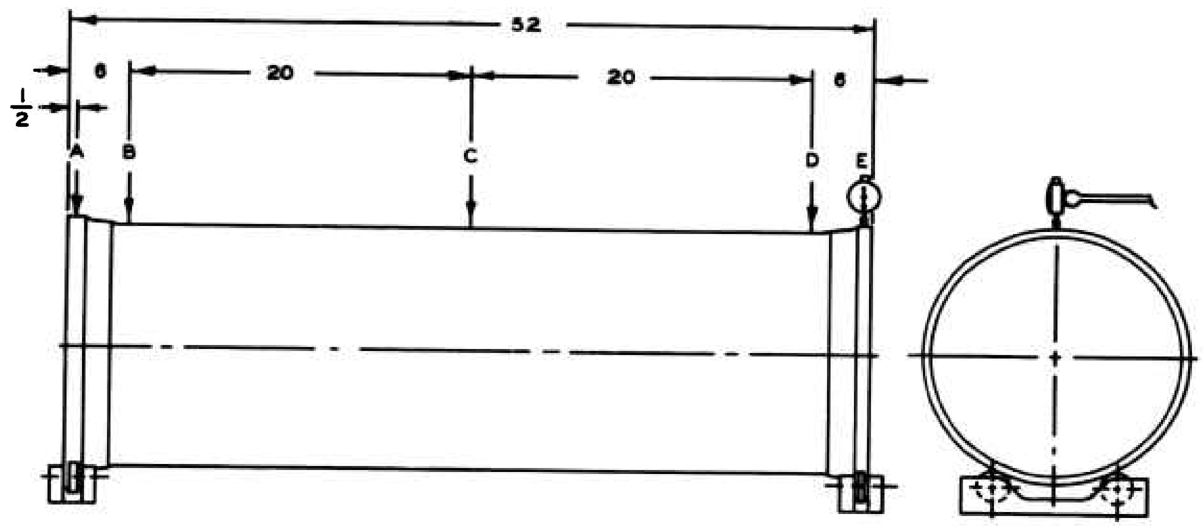


Figure 8. Inspection Setup and Locations for TIR Readings

Table VII. Spun Case TIR Before and After Heat Treat

(TIR Readings - Position - See Figure 8)

Tube No.	A	B	C	D	E
AFTER MACHINING					
F-3	.015	.020	.093	.020	.017
F-4	.031	.070	.028	.038	.034
F-6	.003	.006	.049	.027	.013
F-100	.012	.012	.035	.036	.032
F-101	.012	.015	.038	.021	.021
F-103	.005	.008	.020	.016	.010
F-106	.032	.034	.038	.043	.040
F-108	.012	.011	.020	.022	.022
F-109	.019	.025	.038	.033	.034
F-110	.012	.030	.065	.075	.095
F-112	.007	.013	.020	.028	.027
Avg.	.014	.022	.040	.033	.031
AFTER HEAT TREAT (No restraining fixture used)					
F-3	.048	.050	.135	.130	.105
F-4	.045	.070	.100	.120	.125
F-6	.032	.062	.170	.115	.100
F-100	.025	.103	.195	.125	.100
F-101	.020	.041	.212	.210	.226
F-103	.023	.078	.194	.117	.102
F-106	.046	.072	.160	.102	.083
F-108	.050	.040	.132	.107	.091
F-109	.048	.059	.130	.085	.072
F-110	.055	.100	.250	.195	.079
F-112	.020	.054	.145	.120	.105
Avg.	.037	.066	.166	.130	.108

1. Forged Hydrospun Cases

The test results on the forged hydrospun cases are presented in Table VIII. This table includes the following information:

a. Condition or group number. The cases were processed and tested in groups representing certain conditions of fabrication or heat treat.

b. Tube number and tempering temperature.

c. Results of tensile specimen physical tests. These results were obtained from tensile specimens heat treated with each case.

d. Results of tests conducted on the case. These results tabulate the burst pressures and calculated hoop stress along with a microhardness survey of a sample from the case and present the ratio of the burst stress to the uniaxial tensile specimen results. Hoop stress was calculated from the formula $S = \frac{PD}{2t}$.

The microhardness survey results were included in an attempt to correlate decarburization with burst strength results. The method of measurement utilized was to run a microhardness survey from the surface to the core. The surface hardness is the reading obtained at a .001" depth. The core was at the depth where the hardness readings leveled off. Difficulty was encountered in obtaining the core depth in many instances due to variations in hardness readings within the core. The depth of decarburization shown is based upon an average core hardness from a minimum of 5 readings less 5%. Figure 9 illustrates the determination of decarburization depth for case F-103. No entirely satisfactory method for determining depth of decarburization was

Table VIII. MBMC#1 Forged and Spun Cases - Tensile Specimen and Hydrotest Results

TENSILE SPECIMEN RESULTS - AVG. 2 SPECIMENS												CASE HYDROTEST RESULTS				
Condition	Tube No.	Tempering Temperature	Yield Strength 0.2% Offset PSI	Ultimate Tensile Strength PSI	Elongation % In 2 Inches	Hardness R_c	Ratio YS/UTS	Microhardness R_c Surface	Microhardness R_c Core	Decarb Depth Inches	Tube Inside Dia. Avg. Inches	Tube Wall Thickness Inches	Internal Pressure at Failure Pounds	Hoop Stress at Failure PSI	Ratio Burst Stress UTS Specimen
I	F-I09	500	231,700	268,900	5.5	54.0	0.855	50.0	56.5	.003	16.193	.054	2100	315,900	1.18
I	F-8I	500	231,800	273,500	5.25	55.0	0.845	51.2	56.3	.003	16.216	.047	1590	275,000	1.01
I	F-94	500	259,350	302,I00	5.75	55.0	0.852	55.0	55.0	.000	16.210	.058	1260	177,000	0.59
II	F-3	500	225,750	266,500	5.25	52.0	0.845	31.0	53.0	.007	16.203	.050	1825	290,500	1.10
II	F-6	500	221,050	260,850	4.75	50.0	0.85	23.5	51.6	.007	16.223	.051	1825	290,500	1.11
II	F-I08	500	226,650	266,I50	5.0	53.0	0.85	41.7	58.0	.009	16.200	.053	2050	314,300	1.18
II	F-106	400	237,300	279,800	5.25	54.0	0.85	26.5	55.5	.011	16.I92	.055	2150	317,600	1.14
II	F-I07	400	234,700	280,200	5.5	54.0	0.835	40.0	55.9	.004	16.192	.050	1940	315,000	1.12
II	F-I00	500	229,750	272,300	5.5	50.0	0.84	20.0	54.5	.008	16.186	.049	1880	312,000	1.15
II	F-101	500	226,200	266,000	5.75	52.0	0.845	R_b73.0	57.5	.009	16.200	.052	1875	293,000	1.10
III	F-I03	600	216,450	254,950	6.0	48.0	0.85	38.0	55.5	.007	16.194	.054	1880	283,000	1.11
III	F-1I0	600	211,700	246,200	4.75	50.0	0.87	16.0	50.5	.010	16.214	.051	1640	260,000	1.10
III	F-I02	700	227,000	244,900	5.0	53.0	0.91	38.2	54.9	.005	16.197	.055	2000	296,000	1.21
III	F-I12	700	223,850	243,200	5.25	52.0	0.92	48.0	52.4	.003	16.198	.054	1875	282,000	1.16
IV	F-92	500	230,100	270,950	5.25	54.5	0.827								
IV	F-95	500	232,300	272,550	5.0	55.0	0.85	These tubes hydrotested to 1500 PSI - No failures - Planned for further evaluation by Hercules Powder Co. Allegany Ballistic Laboratory							
IV	F-96	500	234,100	269,600	4.75	55.0	0.87								
IV	F-97	500	230,500	271,300	5.0	55.0	0.85								
IV	F-99	500	232,150	274,250	4.75	55.0	0.845								

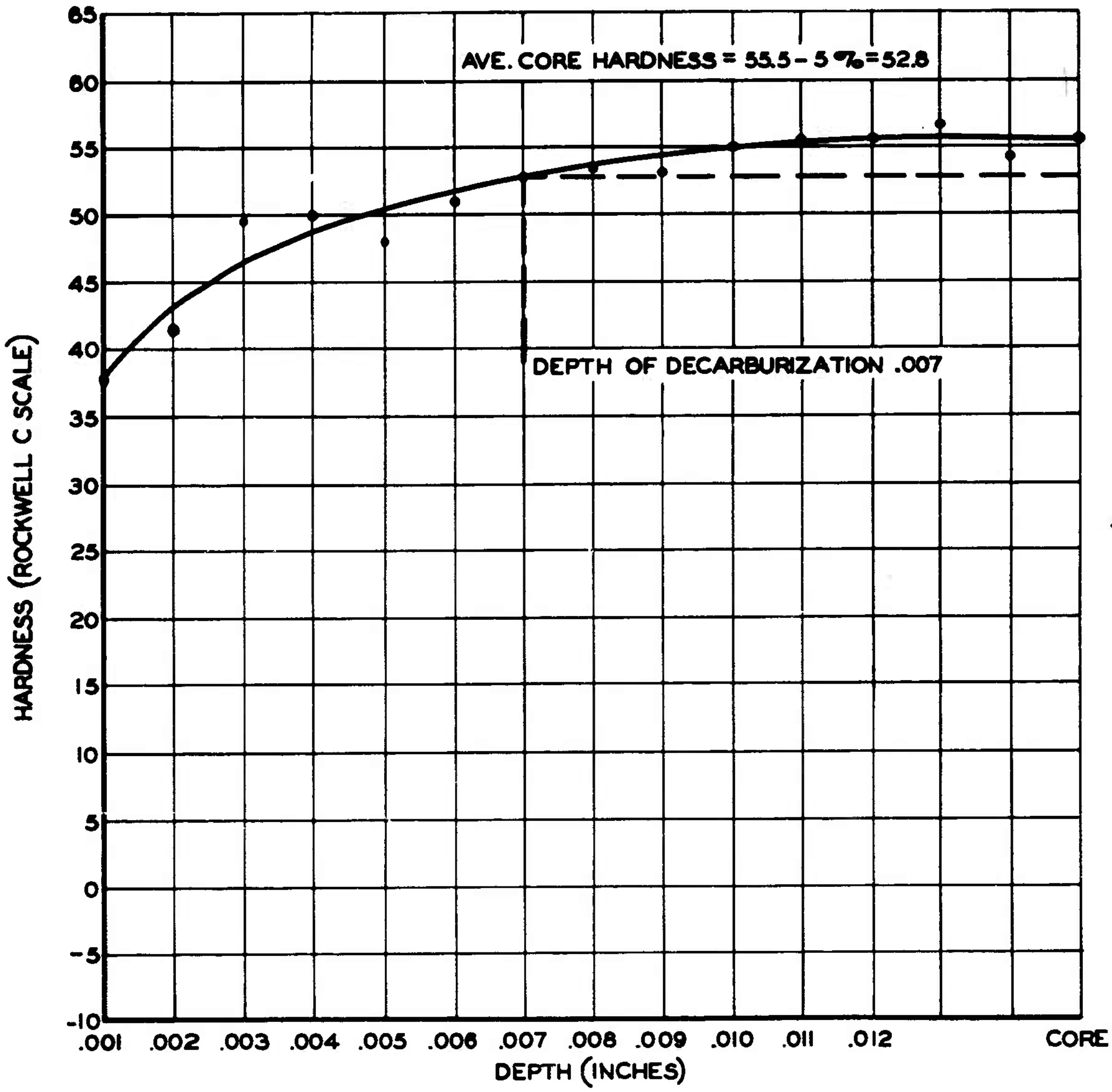


Figure 9. Plot of Microhardness Survey and Depth of Decarburization Determination

found. The surface hardness taken at .001" depth may be a significant factor and was included for reference. The number of cases tested were insufficient to permit any conclusions as to the high and low limits of a surface hardness in conjunction with the depth but it is suspected that an optimum relationship exists.

The wall thickness of the cases was determined by an average of six consecutive one inch spaced readings taken in the thinnest area along the break. The thickness readings were taken using a one inch micrometer at approximately 1/4 inch from the edge of the break.

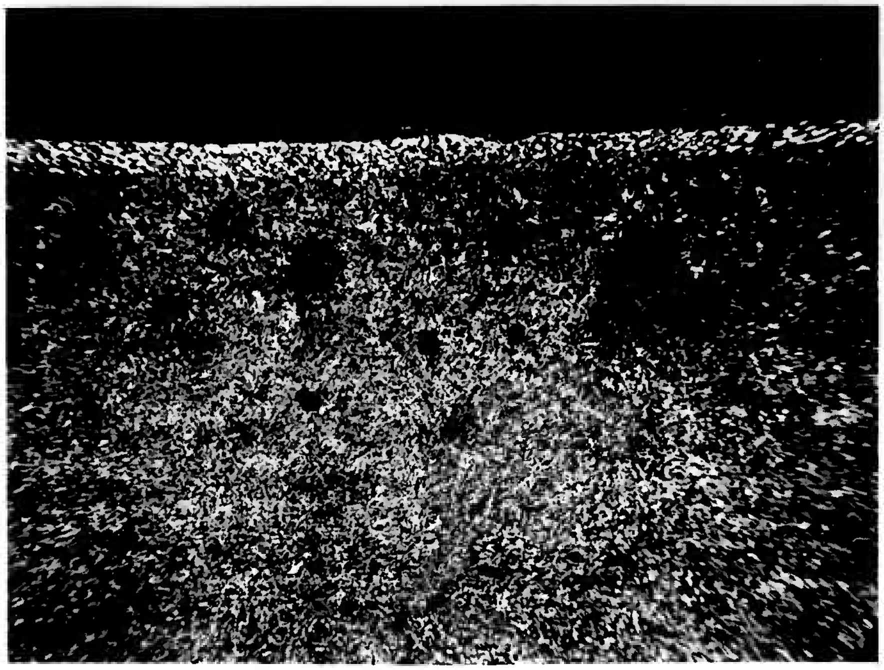
Condition I cases were set up to be heat treated in a neutral atmosphere, 50°F dew point, and tempered at 500°F. The first case tested, F-109, burst at 315,900 psi stress level and at a ratio of burst stress to tensile specimen strength of 1.18. A microhardness survey revealed that the case was slightly decarburized and therefore case F-81 was heat treated with the furnace dew point set 3°F lower. The result was a considerably reduced burst stress of 275,000 psi and ratio 1.01. The microhardness survey again revealed that decarburization had occurred and although the depth of decarburization was the same as case F-109, the surface hardness was 1.2 R_c higher. Case F-94 was then heat treated with the dew point lowered an additional 5°F. The result was a burst stress of 177,000 psi and ratio .59. No decarburization was detected in this case. It was concluded that decarburization was beneficial in obtaining high burst strengths in cases even though the uniaxial tensile strength results were reduced considerably.

Condition II cases were heat treated in a decarburizing atmosphere and tempered at 500°F to determine whether consistent results could be obtained by decarburization and to compare the effect of spinning to various percentages of cold reduction before stress relieve. Cases F-3 and F-6 had been spun with one intermediate stress relieve and cold reductions up to 80% and case F-108 received a maximum cold reduction of 60% before stress relieve.

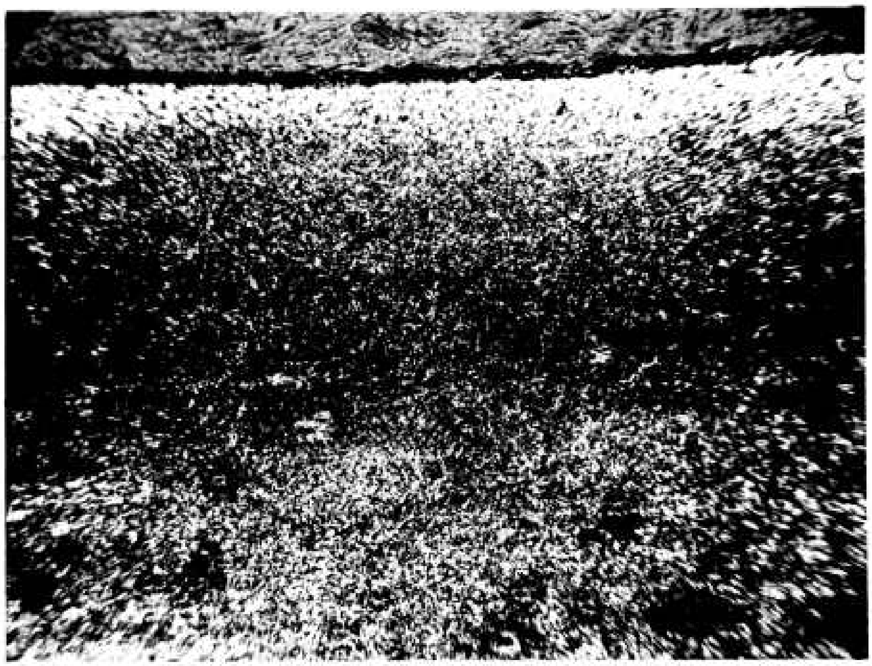
All cases burst at a relatively high level. The decarburization, although variable in depth and surface hardness, was effective on all cases. Case F-108 burst at the highest ratio of burst stress to tensile specimen stress. Since all cases burst at a high level it was indicated that the percentage of cold spinning reduction prior to stress relieve was not necessarily harmful.

It was concluded that consistently high strength results could be obtained with decarburized cases.

Condition III cases were heat treated in a decarburizing atmosphere and the tempering temperatures were varied to determine an optimum tempering temperature for the final evaluation cases. The decarburizing atmosphere was also varied to obtain one case with heavy decarburization and one with medium decarburization for the purpose of selecting a furnace setting for the evaluation cases. Figure 10 shows a photomicrograph illustrating the visual decarburization on cases F-102 (heavy decarb) and F-106 (medium decarb).



Case F-102. Medium Decarburization.
Magnified 100 Times



Case F-106. Heavy Decarburization.
Magnified 100 Times

Figure 10. Photomicrographs Illustrating Visual Decarburization on Forged and Spun Cases

All cases burst at a relatively high stress level. Figure 11 shows photographs of cases F-102, F-103 and F-107 burst patterns which are typical. On a ratio basis a considerable variation occurred which is unexplainable except that all material was air melt and subject to variations. The 500°F tempering temperature was chosen for the evaluation cases based on the high burst stress with severe decarburization, the higher uniaxial physical properties of the specimens and the consistency obtained on the Condition II cases.

There was no correlation of the surface hardness and depth of decarburization with the burst strengths obtained or with the results on the uniaxial tensile specimens. It did not seem reasonable that some rough correlation did not emerge from the wide ranges of decarburization tried in this group. It was apparent that some other factor or factors inherent in the decarburization, the design of the case or the material itself were influencing the strength. However, it was also apparent that consistent high strength results at a ratio of burst stress to uniaxial tensile strength above 1.10 could be readily obtained when the surface hardness was below R_c 48 and the depth of decarburization was at least .003".

Condition IV cases were then heat treated with the furnace atmosphere set at 65°F dew point and the tempering temperature at 500°F. A microhardness survey of the tensile specimen heat treated with one of the cases was made revealing a surface hardness of R_c 37.5, core hardness of R_c 56 and depth of decarburization of .005". This was considered satisfactory based on previous results. All cases were then hydrotested at 1500 psi which was equivalent to a hoop stress of approximately 240,000 psi.

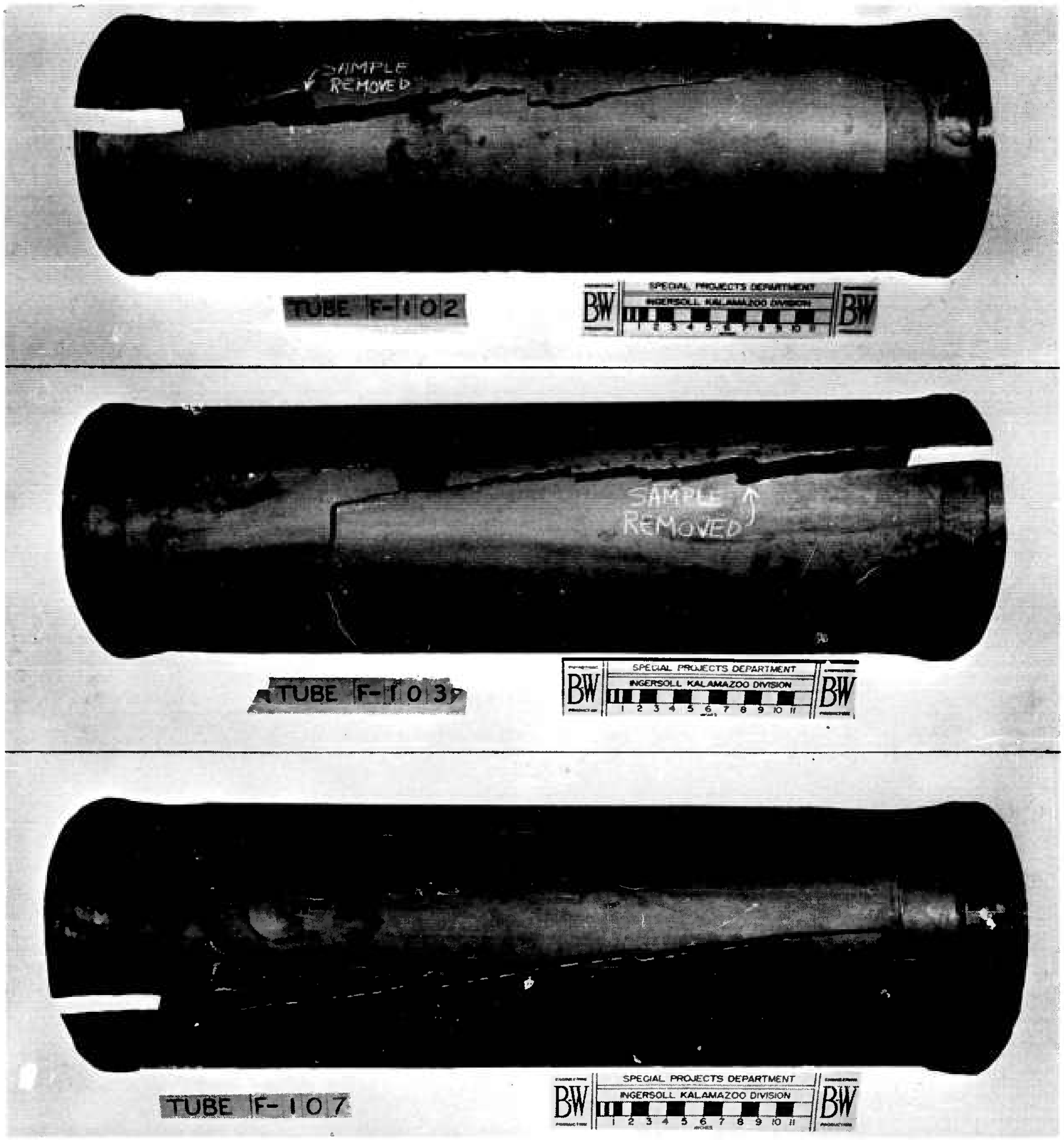


Figure 11. Photographs of Typical Burst Patterns - Forged and Spun Cases

These five cases are planned for evaluation at Hercules Powder Company's Allegany Ballistics Laboratory, Cumberland, Maryland. A shipping date and test schedule have not been received as of the time of publication of this report.

2. Flash Welded Hydrospun Cases

The flash welded blanks were machined and hydrospun by the same methods used for the forged blanks. The spinning results of the flash welded blanks were similar to the forged blanks, except that in three blanks a thin lap in the flash weld seam occurred. The lap extended from .005 to .015" in depth. The cause of this condition was not determined. It is suspected that there must have been a flaw in the weld which was picked up during the spinning operation and rolled over. X-rays on subsequent blanks did not reveal weld defects. The location of the welds were marked for visual observation during spinning, heat treating, and hydrotesting. All cases failed during hydrotest on the weld seam through both end rings. The heat treat and hydrotest results are listed in Table IX.

The first group of cases W-69, W-70 and W-80 tested under Condition I all failed prematurely. These cases were heat treated in a neutral atmosphere. The resulting decarburization of the surface was very slight as indicated by the surface hardness. Figure 12 illustrates the microstructure of the material taken from case W-69. It can be noted that no surface decarburization had occurred.

Table IX. MBMC#1 Flash Welded and Spun Cases - Tensile Specimen and Hydrotest Results

		TENSILE SPECIMEN RESULTS AVG.								CASE RESULTS					
Condition	Tube No.	Tempering Temperature	Yield Strength 0.2% Offset PSI	Ultimate Tensile Strength PSI	Elongation % in 2 Inches	Hardness R_c	Ratio YS/UTS	Microhardness R_c Surface	Microhardness R_c Core	Decarb Depth Inches	Tube Inside Dia. Avg. Inches	Tube Wall Thickness Inches	Internal Pressure at Failure Pounds	Hoop Stress at Failure PSI	Ratio Burst Stress UTS Specimen
I	W-69	500	238800	276750	5.75	53.2	0.86	55.0	52.0	.000	16.192	.050	1350	219300	0.79
I	W-70	500	240900	281800	5.5	52.7	0.85	53.5	58.0	.004	16.208	.057	1200	171200	0.61
I	W-80	500	240900	281500	5.5	52.7	0.86	51.0	56.0	.004	16.214	.057	1475	210500	0.75
II	W-68	500	220900	261350	5.25	51.0	0.85	38.0	59.5	.005	16.200	.050	1525	247800	0.95
II	W-67	700	206350	230200	4.5	49.0	0.90	R_d74.0	51.0	.005	16.210	.050	1300	211400	0.92
II	W-66	700	201000	228750	6.0	47.0	0.88	29.5	56.0	.010	16.190	.054	1350	203100	0.89
II	W-59	600	217000	246150	4.75	51.0	0.88	9.0	51.5	.007	16.194	.050	1115	181200	0.74
II	W-54	600	218650	249400	4.75	52.0	0.88	13.0	51.5	.008	16.205	.054	1575	237100	0.95
II	W-78	600	218300	245000	5.5	52.0	0.89	18.5	53.0	.006	16.201	.053	1500	239200	0.98
III	W-65	500	238000	280000	5.25	53.0	0.85	51.5	57.5	.004	16.207	.046	1450	256200	0.92
III	W-57	500	233250	273150	5.5	52.0	0.85	49.0	54.9	.004	16.204	.050	1980	321800	1.18
III	W-51	500	230600	271550	5.75	53.0	0.85	51.0	53.9	.002	16.219	.045	1350	244000	0.90
III	W-53	500	233100	273850	5.5	52.0	0.85	48.0	54.8	.003	16.201	.053	1600	254500	0.93

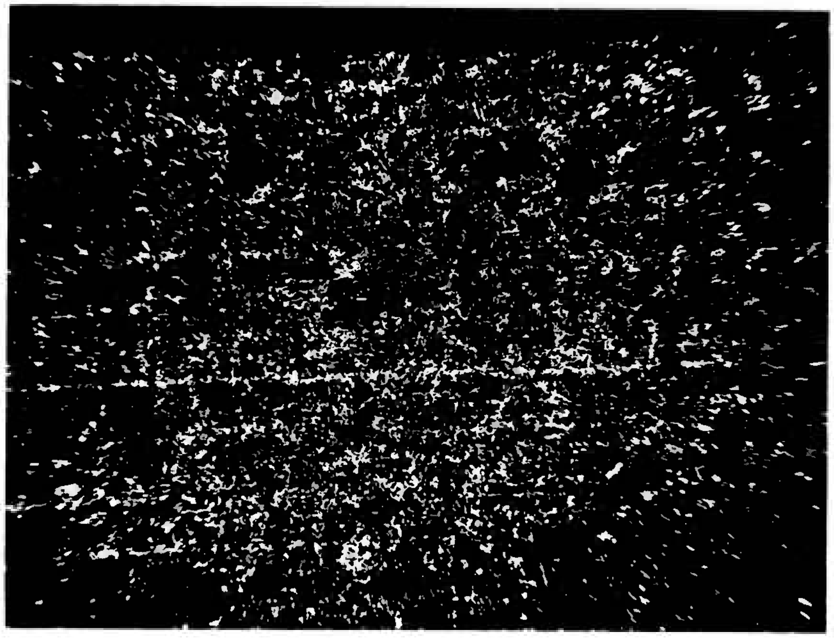


Figure 12. Photomicrograph of Case W-69 Illustrating Lack of Decarburization

Condition II cases were heat treated in a decarburizing atmosphere and at tempering temperatures of 500, 600, and 700^{o}F. All cases failed prematurely although at a higher stress level than the Condition I cases. The decarburization was generally severe with low surface hardness and a depth up to .010". The varied tempering temperature was apparently not a factor. The failures occurred along, and in, the flash welded seam. It was suspected that the fault was in the flash weld. Micrographs of several welds were made revealing a structure having excessive inclusions. Figure 13 is a micrograph of the center of the weld on case W-78. Such inclusions were found in localized areas only.

Four additional cases were heat treated at 500^{o}F and the decarburization was reduced for Condition III. Due to the condition of the welds premature results were expected. Of this group only case W-57 burst at a high level, 321,800 psi, and at a ratio of burst stress to uniaxial tensile strength of 1.18. The remaining cases burst at a level similar to the Condition II cases on a ratio basis. Several micrographs made of the weld on case W-57 revealed no excessive inclusions in the weld. A micrograph of the W-57 weld is shown in Figure 14.

Additional work on flash welded blanks was discontinued since the objective of this project did not permit work in the area of developing flash welding techniques.

It is expected that spun flash welded blanks with clean welds would produce high strength cases similar to forged blanks.

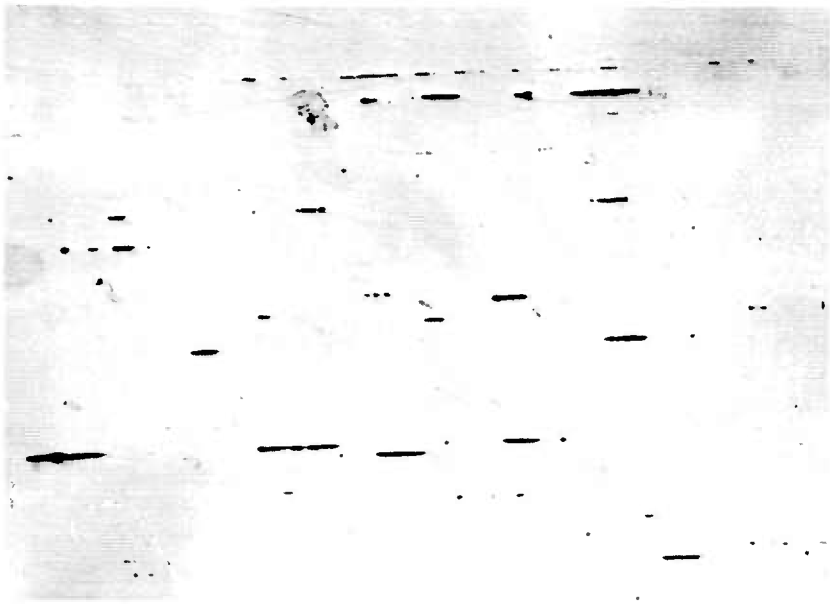


Figure 13. Photomicrograph of Case W-78, Illustrating Excessive Inclusions in Flash Weld. Case Failed at a Low Stress Level (239,200 psi). Magnified 100 Times

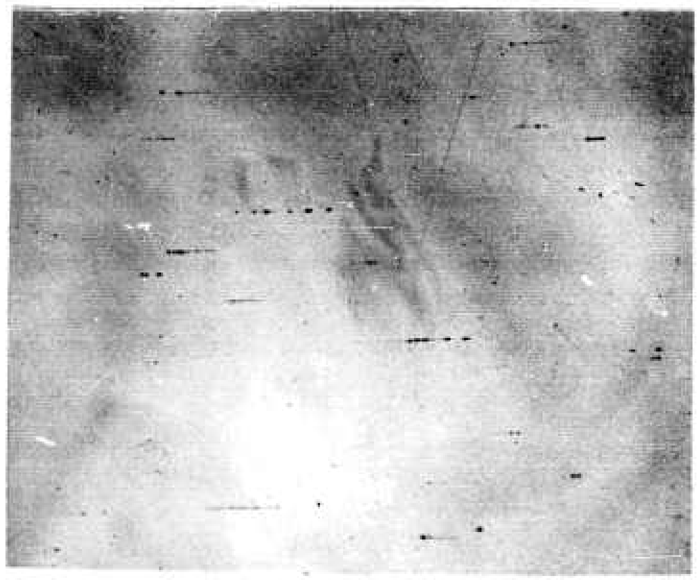


Figure 14. Photomicrograph of Case W-57, Illustrating Relatively Sound Flash Weld. Case Failed at a High Stress Level (321,800 psi). Magnified 100 Times.

B. Rolled and Welded Cases

The cases were welded in groups of 3 to 6 as required in accordance with the welding procedures described previously.

Six conditions or sets of cases were evolved in the development before relatively high strength cases were obtained. Each set of cases indicated a condition which was detrimental and necessitated change before the burst strength level could be increased. Although it would have been possible to anticipate some of these conditions, thereby eliminating some steps, it was considered to be more beneficial in the interest of R & D to change only one variable at a time in order to evaluate its effect on the overall.

Figure 15 illustrates graphically the increased burst strength values obtained with the development of each condition. It was expected that a further increase in burst strength could be obtained by refinement of the welding technique but funding and time did not permit completion of more than this rough phase of the rolled and welded case development.

The following is a summary of the six conditions which lead to high strength cases in the roll and weld program.

I. Insufficient surface decarburization.
II. Surface decarburization incorporated.
III. Surface decarburization and all scale removed prior to heat treat.
IV. Condition III and use of a more ductile weld wire for the girth weld.
V. Condition IV and girth weld ground flush inside and outside.
VI. Condition V and relocate girth weld joint with respect to end ring.

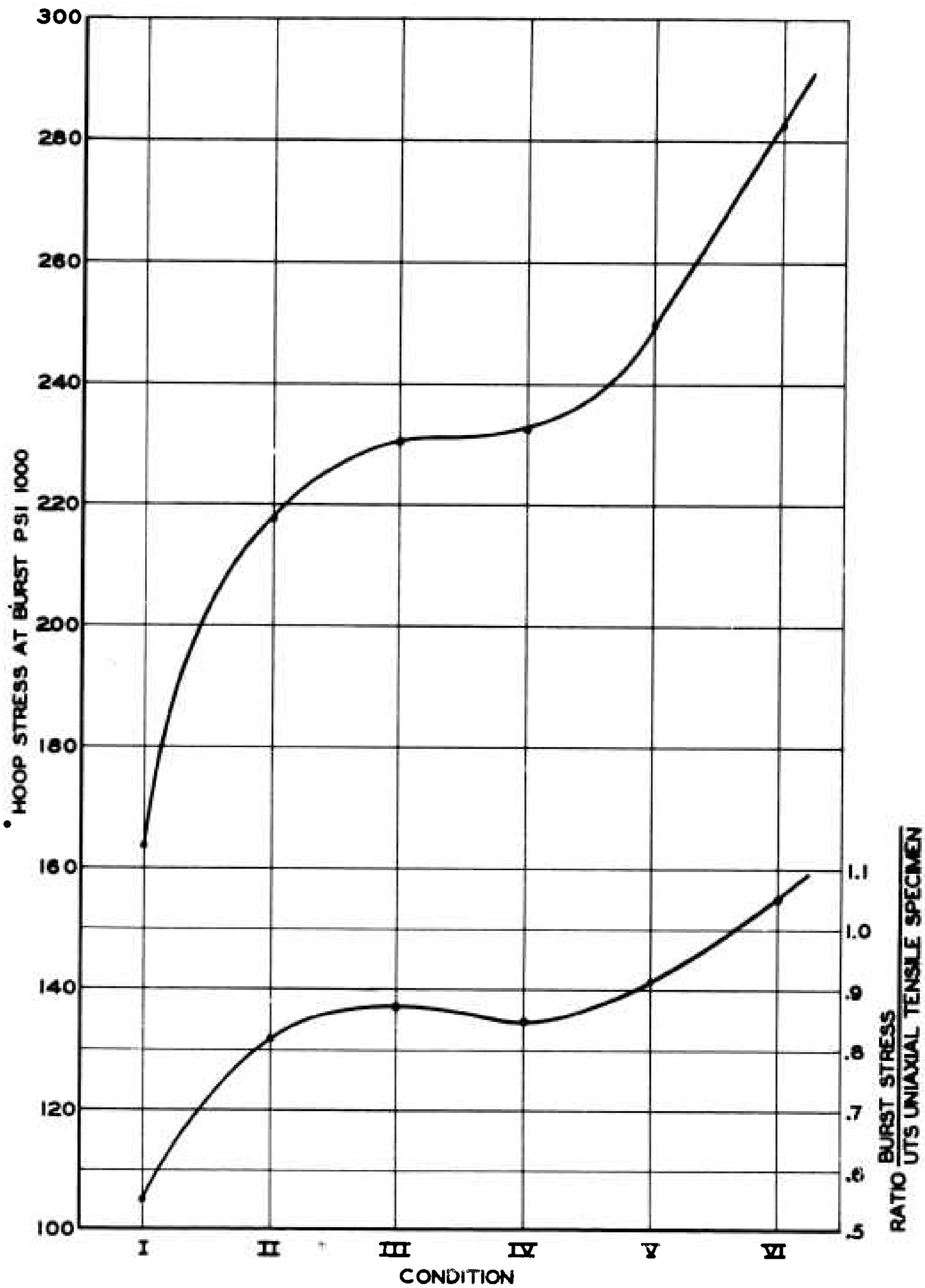


Figure 15. Rolled and Welded Cases - Progress Chart

Table X tabulates the results obtained on the cases tested at each of the six conditions. The austenitizing temperature for all cases was 1600°F. The time at the austenitizing temperature for Condition I was 50 minutes and for all other conditions 65 minutes. The tempering temperature is shown for each case and the time at heat was 3 hours for all cases. The physical properties from the tensile specimens heat treated with each case is shown. The microhardness at the surface (.001" depth was considered the surface hardness) and core of a sample taken after burst from each tube is recorded. The wall thickness of each tube was selectively taken as an average of a length of tube six inches long in the thinnest area along the length of the burst seam. In all cases the wall thickness varied very little along the length. The calculated hoop stress at failure and the ratio of calculated hoop stress to the tensile strength of the uniaxial tensile specimen are shown in Table X.

Photographs of typical burst patterns of cases are shown in Figure 16.

Condition I cases were welded entirely with MBMC #1 weld wire and the welds were not dressed in any way. The cases were heat treated at 1600°F for 50 minutes in a neutral atmosphere without a prior descale operation. Figure 17 is a photomicrograph of case R-23 which was typical of all three cases. There was no visual evidence of surface decarburization, however, a microhardness survey revealed a difference in hardness of as much as 5 points R_c from the surface to the core. In all cases the surface hardness was greater than R_c 52 and it was considered that no effective surface decarburization had occurred. In conducting the microhardness survey it was noted

Table X. MBMC#1 Steel Rolled and Welded Cases - Tensile Specimen and Hydrotest Results

			TENSILE SPECIMEN RESULTS - AVG. 2 SPECIMENS					CASE HYDROTEST RESULTS							
Condition	Tube No.	Tempering Temperature	Yield Strength 0.2% Offset PSI	Ultimate Tensile Strength PSI	Elongation % in 2 Inches	Hardness R_c	Ratio YS/UTS	Microhardness R_c Surface	Microhardness R_c Core	Decarb Depth Inches	Tube Inside Dia. Avg. Inches	Tube Wall Thickness Inches	Internal Pressure at Failure Pounds	Hoop Stress at Failure PSI	Ratio Burst Stress UTS Specimen
I	R-1	500	268,700	303,200	4.75	55.5	0.89	53.5	56.0	.001	16.205	.050	900	138,000	0.45
I	R-19	500	261,300	295,050	5.5	54.0	0.89	-	-	-	16.200	.053	1250	191,700	0.65
I	R-23	500	256,100	291,200	5.0	54.5	0.88	52.0	57.5	.005	16.210	.051	1025	163,400	0.56
II	R-16	500	244,450	283,050	5.25	53.2	0.87	38.5	56.7	.005	16.205	.052	1375	214,900	0.85
II	R-20	500	239,350	274,550	5.25	53.0	0.88	38.0	52.0	.004	16.191	.050	1400	227,400	0.83
II	R-22	500	229,950	269,250	5.25	52.0	0.86	43.0	57.9	.007	16.214	.050	1300	211,400	0.79
II	R-13	600	212,050	234,750	5.5	42.5	0.91	20.2	56.7	.008	16.182	.050	1100	178,600	0.76
II	R-25	600	218,800	241,850	5.0	47.0	0.91	19.0	57.0	.010	16.210	.050	1100	178,900	0.74
II	R-29	600	216,500	242,350	5.25	44.7	0.90	R_b73.0	56.7	.008	16.207	.050	1125	182,900	0.75
II	R-21	700	195,750	221,800	4.75	46.2	0.89	38.0	52.0	.004	6.191	.050	1225	198,800	0.90
II	R-28	700	225,900	242,350	4.5	50.0	0.93	21.0	48.0	.006	16.210	.050	950	154,400	0.64
III	R-2	500	239,200	273,700	4.75	51.7	0.88	41.5	57.5	.006	16.189	.049	1340	222,000	0.81
III	R-14	500	214,050	252,700	5.0	50.0	0.85	35.5	52.5	.009	16.190	.051	1550	246,800	0.98
III	R-17	500	236,200	274,900	4.5	51.7	0.86	31.0	56.6	.006	16.190	.050	1370	222,500	0.81
IV	R-5	500	243,300	276,550	5.25	53.0	0.88	46.7	56.3	.005	16.188	.050	1490	242,000	0.88
IV	R-18	500	240,050	274,200	5.0	52.0	0.88	48.5	54.5	.003	16.186	.050	1350	219,200	0.80
IV	C-90	500	228,700	268,500	4.75	52.5	0.85	44.0	54.8	.005	16.224	.048	1400	237,300	0.88
V	R-11	500	247,750	280,600	5.5	54.0	0.88	45.5	54.5	.003	16.189	.047	1470	253,900	0.90
V	R-30	500	244,350	275,000	5.0	52.0	0.89	38.5	55.0	.005	16.210	.050	1500	243,900	0.89
V	C-98	500	230,350	268,800	5.0	55.0	0.86	45.0	50.2	.003	16.208	.047	1450	250,700	0.93
VI	R-31	500	240,200	270,700	5.5	52.0	0.89	40.2	54.2	.004	16.180	.047	1675	289,100	1.07
VI	C-104	500	230,100	268,100	5.0	52.0	0.86	36.0	54.5	.004	16.205	.050	1700	276,300	1.03

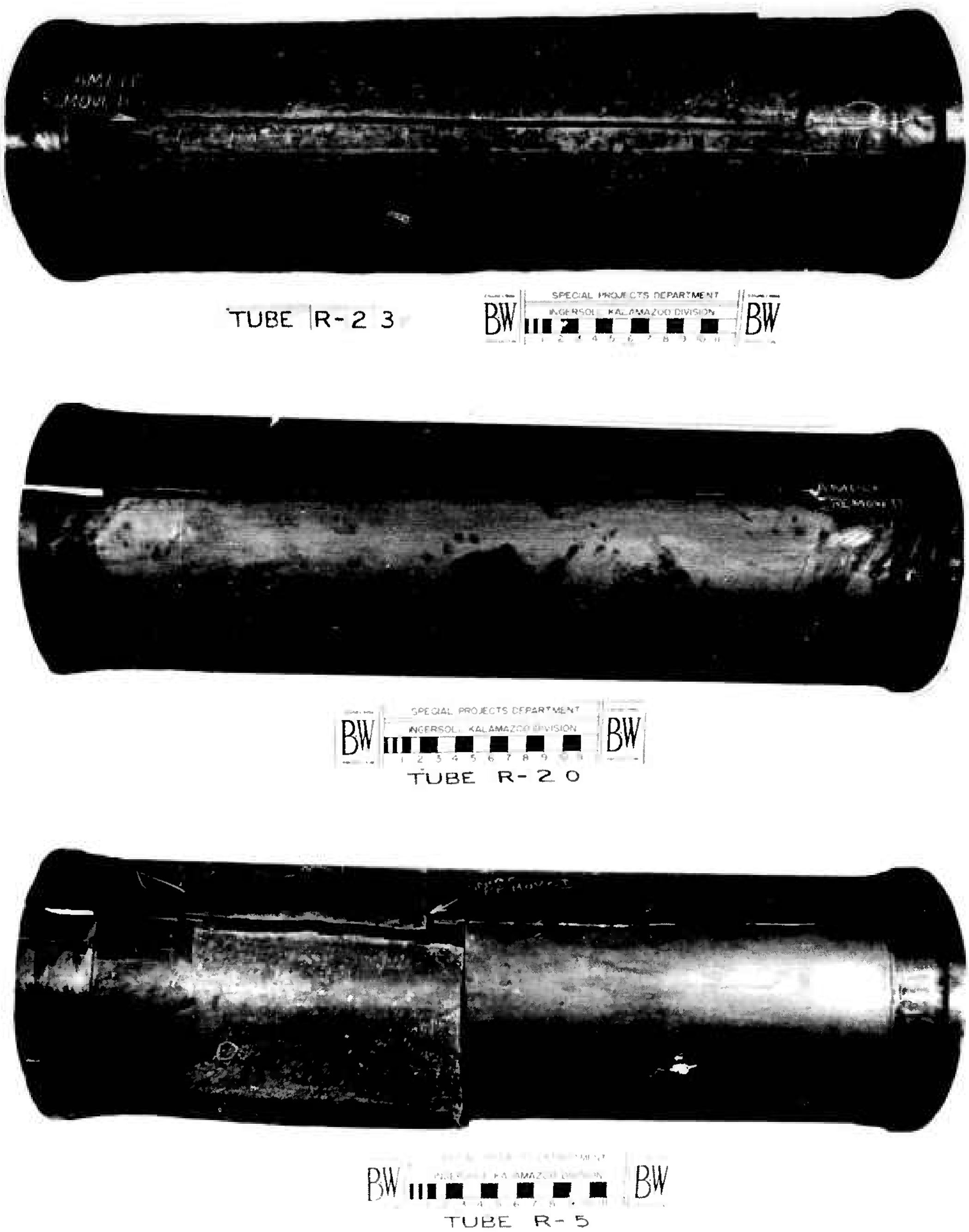


Figure 16. Photographs of Typical Burst Patterns - Rolled and Welded Cases

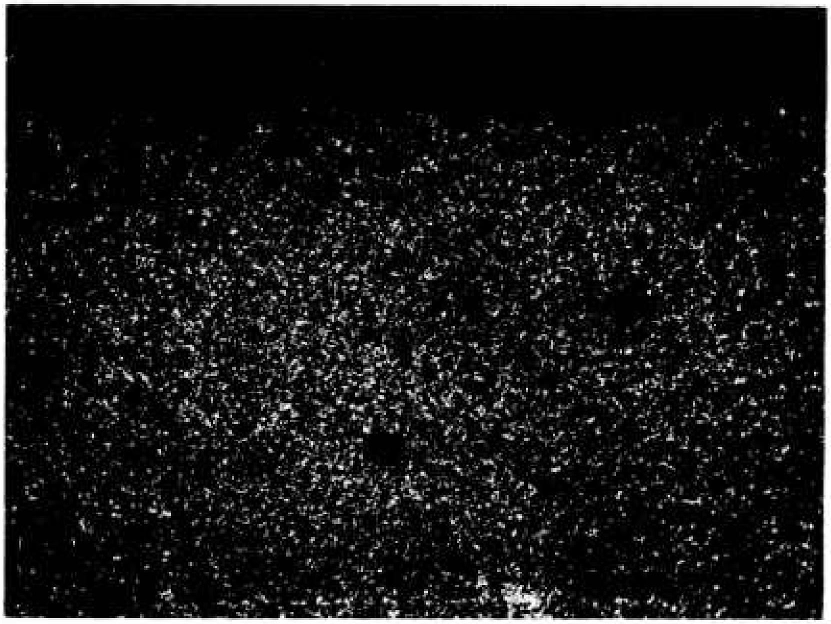


Figure 17. Photomicrograph of Case R-23 Illustrating Lack of Decarburization

that considerable variations in hardness sometimes occurred at a given depth indicating a non-uniform structure. As noted under the metallurgical study the sheet material contained considerable banding and alloy segregation and since this was the only material available it was decided to continue under this handicap. Also, since such a condition could occur to some degree in any material it was considered that the results obtained would be indicative of the minimum results to be expected. The average burst strength of the three Condition I cases was 164,400 psi. The ratio of burst stress to the uniaxial tensile strength averaged .55. The failures were all well below the yield strength indicated by the tensile specimens and it was considered that the failures were due to notch sensitivity of the material. None of the burst patterns were in the longitudinal welds nor were they directly traceable to the girth welds.

It was then decided to try a decarburizing atmosphere and to try several tempering temperatures.

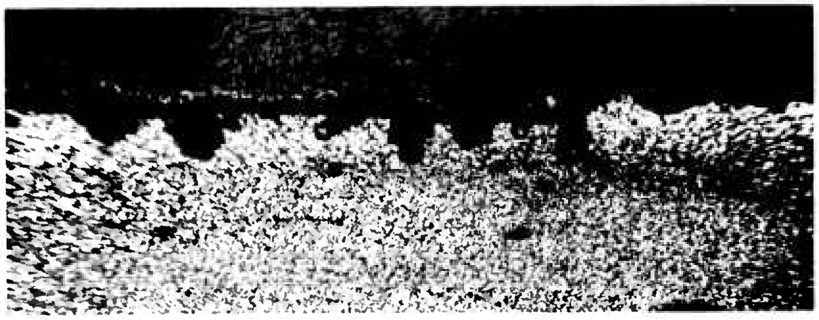
Condition II cases were the same as Condition I cases except that they were heat treated in a decarburizing atmosphere for 65 minutes and tempered in groups at 500, 600, and 700°F. Variations in the furnace atmosphere dew point were made from 60°F to 75°F in an effort to find some satisfactory decarburization cycle or obtain an indication of the minimum amount necessary. No pattern was developed from the results obtained.

The 700° and 600°F temper cases were processed and hydrotested first. The average burst strength for the 700°F temper cases was 176,600 psi and for the 600°F temper cases was 180,130 psi. The ratio of burst strength to uniaxial tensile strength for the combined group of cases was .76. This represented an improvement over the Condition I cases but the results were entirely unsatisfactory.

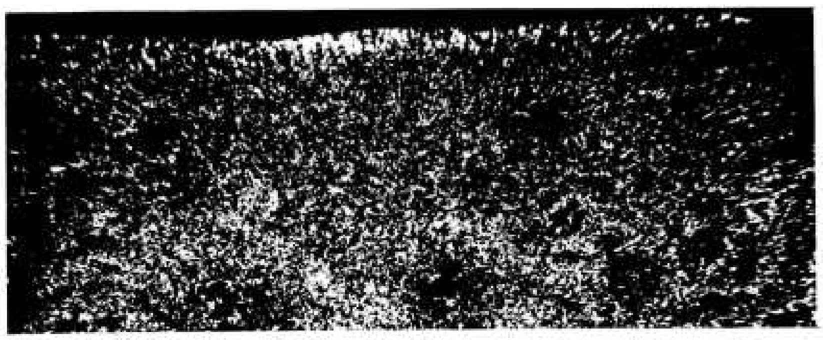
During the course of running the microhardness surveys on these cases it was noted that the depth of decarburization was not uniform and further investigation revealed that tight surface scale in isolated areas was preventing decarburization. Figure 18, case R-28, is a photomicrograph of an area exhibiting this condition. Areas of surface pits that extended up to .003" into the material were also found in the cases. This condition is shown in Figure 18, case R-2. These pits were found in all the case sheet material to some extent. A method of evaluating the effect of these pits was not immediately known but later under Conditions IV, V and VI a method was devised.

The 500°F temper cases were then processed with the addition of a vacublast descale operation prior to heat treat. The result was an average burst strength of 217,900 psi and the ratio of burst stress to uniaxial tensile strength was .82. This was a significant improvement, although the burst strengths were well below those anticipated.

Condition III cases were processed the same as the 500°F temper Condition II cases with the addition of special attention to the descale operation to be absolutely certain that no scale remained on the cases and the wire that was previously wrapped around the outside of the case for holding the tensile specimens was removed to assure that decarburization would be com-



Photomicrograph Illustrating Surface Pits



Photomicrograph Illustrating Uneven Decarburization

Figure 18. Photomicrographs Illustrating Surface Pits and Uneven Decarburization

plete over the entire surface. The dew point of the furnace atmosphere was held at 60°F for R-2, 65°F for R-14 and 70°F for R-17.

The burst strengths averaged 230,430 psi and the burst stress to tensile strength ratio was .85. This was an improvement over the Condition II cases indicating the importance of a clean surface prior to heat treat. The cases burst roughly at the yield strength except for R-14 which apparently had excessive decarburization as evidenced by the low yield strength obtained on the uniaxial tensile specimens. This case burst well above the yield strength but still at a relatively low level.

It was evident at this point that decarburization alone would not produce cases of a higher strength level and that further decarburization would reduce the strength level below any useful level for the purposes of this program.

The examination of burst tubes of Conditions I, II and III revealed evidence of slow crack propagation at the girth weld generally initiating from the weld junction to the parent metal. Examination of the X-ray film revealed no visible flaws at these points. Except for the three tubes none of the failures occurred along the longitudinal weld and all failures except for one case in Condition I extended through at least one end ring.

It was decided to run a stress coat analysis on a case processed to Condition III at a 65°F dew point to determine whether or not a stress concentration existed that would account for the premature failures being obtained. Tube R-26 was sprayed with "Stresscoat", pressurized to various internal pressures and then examined for cracks in the brittle coating. Coating cracks first occurred in the end rings at the radius which blends out the 0.050" thickness. Figure 19 depicts the location of the cracks and their direction. These cracks occurred at a stress level of 45,000 psi, which is the stress level at which the brittle coating was to have cracked. The stress level was raised to 45,500 psi and the coating cracked throughout the entire tube. The longitudinal weld was the only portion of the coating that did not crack. The tube was then progressively pressurized to stress levels of 65,000, 81,500, 98,000 and 163,000 psi. The crack pattern of four locations were photographed at each of these stress levels. The four locations included: (1) the end ring to tube juncture at the longitudinal weld at the top of the case, (2) the longitudinal weld, (3) the parent metal, and (4) the end ring to tube juncture at the bottom of the tube. Figures 20 and 21 are photographs of the longitudinal and girth welds, locations 2 and 4, at the 98,000 psi and 163,000 psi stress loads. The tube was rotated in the hydrotest stand to permit examination of the coating around the entire periphery of the tube after each pressurization. There were no apparent stress concentrations developing in the tube section. The longitudinal weld showed no evidence of stress until the stress reached 98,000 psi. The coating on the longitudinal weld showed additional cracks at the 163,000 psi stress level, but they were not as extensive as in the parent metal in the tube section and their directions were erratic. Cracks which appeared at the end ring radius developed first into a jumble of cracks and ultimately into a pattern resembling a finely divided cross hatching. It is

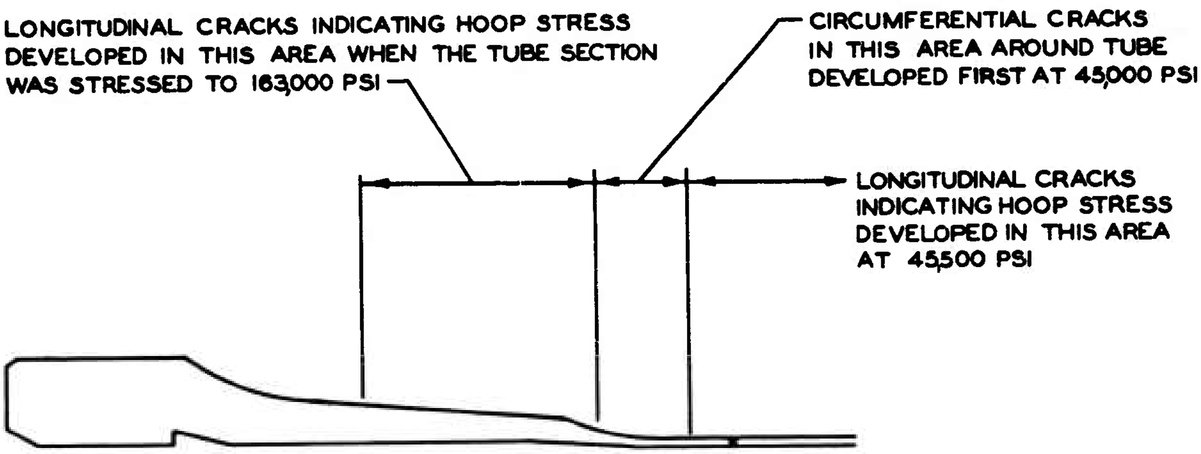
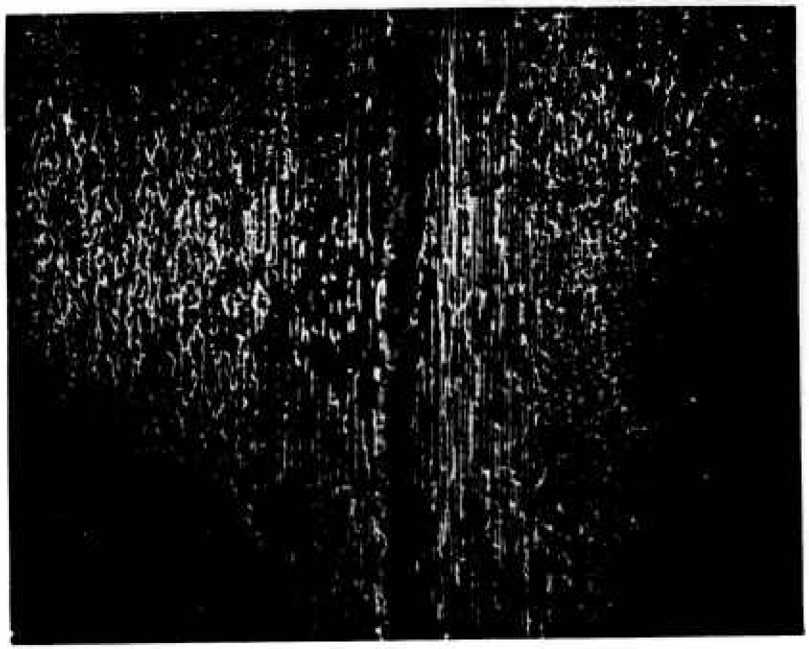
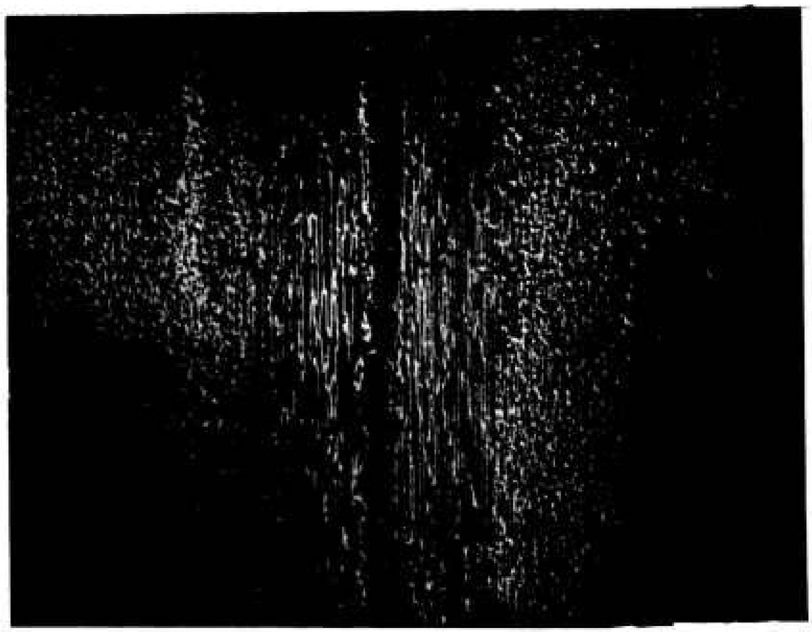


Figure 19. Sketch Illustrating Stress Coat Crack Locations and Types

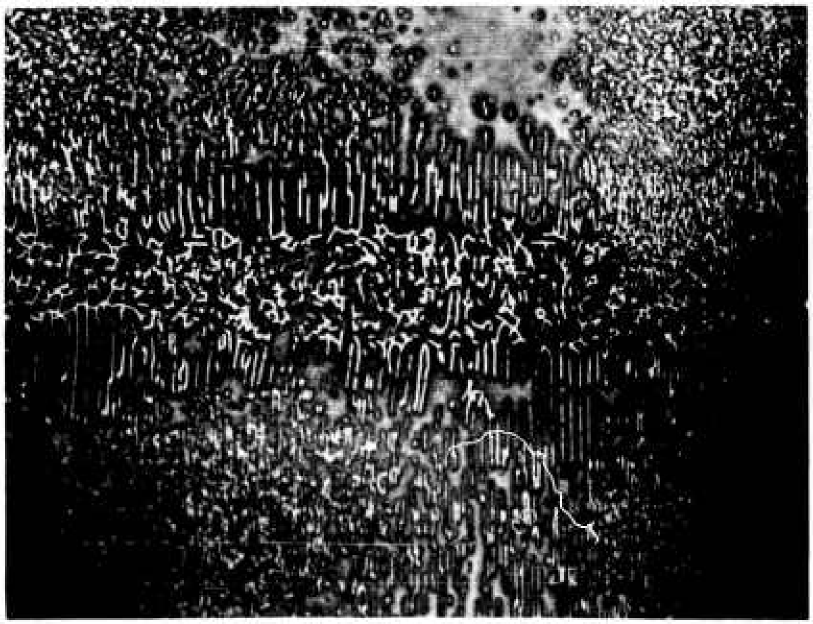


Stress Cracks at 98,000 psi Hoop Stress Along Longitudinal Weld. X 2



Stress Cracks at 163,000 psi Hoop Stress Along Longitudinal Weld. X 2

Figure 20. Stresscoat Crack Pattern along Longitudinal Weld



END RING

GIRTH WELD

Stress Cracks at 98,000 psi Hoop Stress at Girth Weld. X 2

END RING

GIRTH WELD

Stress Cracks at 163,000 psi Hoop Stress at Girth Weld. X 2

Figure 21. Stresscoat Crack Pattern at Girth Weld

significant that the coating on either side of the end ring radius cracked in a hoop stress pattern. It is suspected that the girth weld of MBMC #1 wire forms a narrow reinforcing ring which causes a severe bending moment in the end ring at the radius. The form and pattern of the brittle coating cracks indicated that this end ring area was being put into compression by this bending moment.

This case, R-26, burst at a stress level of 195,000 psi which was expected since it had severe repair work in one area of one girth weld which had rendered it unfit for normal evaluation test purposes.

It was concluded from the stress coat test that a stress concentration was present in the area of the girth welds and that this condition could be overcome by increasing the ductility of the weld and/or relocating the girth welds with respect to the end ring.

Condition IV cases were girth welded with AISI 6130 welding wire, which would represent a more ductile filler material for the girth welds. The weld crown was not ground. The longitudinal seam which did not display evidence of stress concentration was welded with MBMC #1 weld wire.

Further investigating the material problems, discussed under Condition II, a hydrospun tube was used for the tube section of case C-90. The hydrospun forged case would eliminate the pitted condition that was present in the sheet stock and permit evaluation of this condition. End rings were welded to the spun tube section using AISI 6130 welding wire.

The cases were processed through vacublast and heat treat the same as Condition III cases with the furnace dew point at 65°F.

The results were an average burst strength of 232,800 psi and a burst stress to uniaxial tensile strength ratio of .85. This compares with the Condition III cases except that for some reason the decarburization depth was somewhat less. There was no significant change noted due to the change in weld wire.

The case having the hydrospun tube in place of the rolled and welded tube broke at a level equivalent to one of the welded tubes and since the failure initiated at the girth weld it was indicated that the pits observed in the sheet stock were not necessarily detrimental at this strength level.

Condition V cases were processed with the girth welds ground flush to further reduce the stress concentration due to the girth weld locations. Case C-98 was a hydrospun seamless tube girth welded with 6130 weld wire. Case R-11 was a rolled and welded tube using MBMC #1 weld wire for the longitudinal weld and 6130 weld wire for the girth welds. Case R-30 was welded using MBMC #1 weld wire for both the longitudinal and girth seams. The girth welds of all three cases were ground flush both inside and outside before heat treat. The heat treat cycle was set the same as for the Condition IV cases.

The average burst stress of the three cases was 249,500 psi and the burst stress to uniaxial tensile strength ratio was .91. This was a significant increase but it was obvious that the full strength of the material would not be realized with the girth welds in the present location. The burst strengths of the cases were very nearly equivalent on a ratio basis.

The depth of decarburization on cases C-98 and R-11 was less than anticipated for a 65°F dew point setting and erratic, the same as occurred on case R-18. Further investigation revealed that the dew point indicator was not operating properly which could account for the relatively shallow decarburized layer obtained. The results obtained on these two cases indicate that a very shallow decarburized layer may be effective.

Two other factors regarding the decarburization were noted. On case R-18 of Condition IV the surface hardness was high even though the depth of decarburization was the same as that recorded for cases C-98 and R-11. The surface hardness undoubtedly has an upper limit beyond which decarburization is not effective. The surface hardness of R_c 48.5 recorded for case R-18 may be too high for full effectiveness. On cases C-98 and R-11 after polishing and etching a lighter band along the decarburized surfaces was noted. This band could not be readily distinguished under the microscope but was apparent to the eye and at low magnification. The width of this band was approximately .005" on both C-98 and R-18 and it was definitely due to decarburization since it did not appear at the ends of the specimen. This same type of band was present on all decarburized cases and in all cases extended deeper than the hardness measurement used to determine decarburization depth for this program. A method of measurement other than a microhardness survey would be necessary to evaluate this band. It is mentioned here to prevent the conclusion that practically no decarburization effect was present on cases C-98 and R-11. This band and the toughness parameters due to decarburization will be the subject of a carry-on study to this program.

Condition VI cases were fabricated with the girth welds relocated. It was expected that moving the weld location either closer to or further from the end ring would change the stress pattern to an extent that higher strength cases should result. At this point the contract time and funds were nearly expended and only two more cases could be fabricated. Accordingly, it was decided to move the girth weld location for a rolled and welded tube to the end of the radius of the end ring and the location of a hydrospun tube an additional 2" away from the end ring. Figure 22 is a sketch showing the original girth weld location and the two new locations used for the Condition VI cases. The girth welds were made using 6130 weld wire, the welds were ground flush both OD and ID and the furnace atmosphere was set at a 65°F dew point using a re-calibrated dew point indicator.

In the girth welding of both these cases some difficulty with porosity occurred necessitating numerous repairs, particularly on case C-104, where the repairs caused an outward bulge up to .125" high for about a 4" length around the case. Had it not been for this problem it was expected that a considerably higher burst strength would have resulted. The burst strength of these two cases averaged 282,700 psi and the burst stress to uniaxial tensile strength ratio averaged 1.05. The results were in the range that would be considered satisfactory for this material although with refinements of the

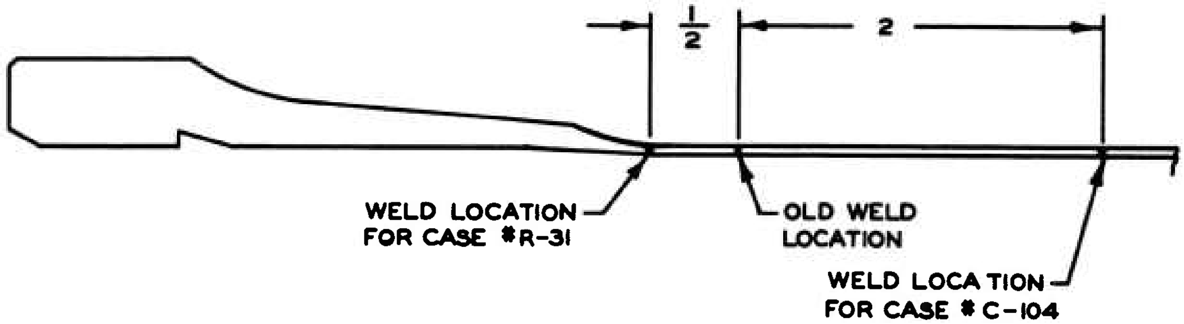


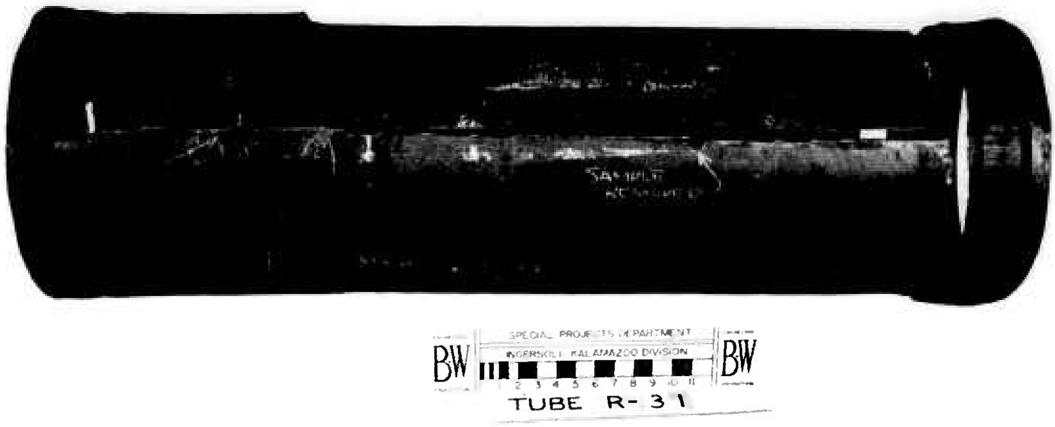
Figure 22. Sketch Showing Relocation of Girth Welds

girth welds it is expected that a higher ratio would be obtainable.

Figure 23 shows photographs of the burst patterns on cases R-31 and C-104. Failure of case R-31 initiated at the junction of the longitudinal and girth weld and progressed along the longitudinal seam in the heat affected zone of the weld jumping from side to side and splitting out into the case on both sides. The end ring where failure initiated was severed. There was no indication of a flaw at the initiation point. Failure of case C-104 initiated at the girth weld in the bulge area. There was no indication of a flaw in the failure initiation area. Both cases burst in a ductile manner indicating a tensile failure.

The performance of the hydrospun case girth welded to end rings as compared to rolled and welded sheet stock welded to the end rings indicated that the pits noted in the sheet stock were not sufficiently detrimental to materially affect the results of the burst strength tests on this design case. This in no way condones the use of pitted stock since many other detrimental and unpredictable results could occur. It does indicate, however, that decarburization can effectively neutralize some defects on a short time basis. The effect of long time storage after heat treat on these pits is not known.

A dimensional inspection consisting of TIR readings was made on a number of the welded cases after the heat treat operation. Table XI tabulates the results. The method and position of the readings are shown in Figure 8 and were the same as used for the hydrospun cases shown in Table VII. The distortion after heat treat was approximately equal for both types of cases.



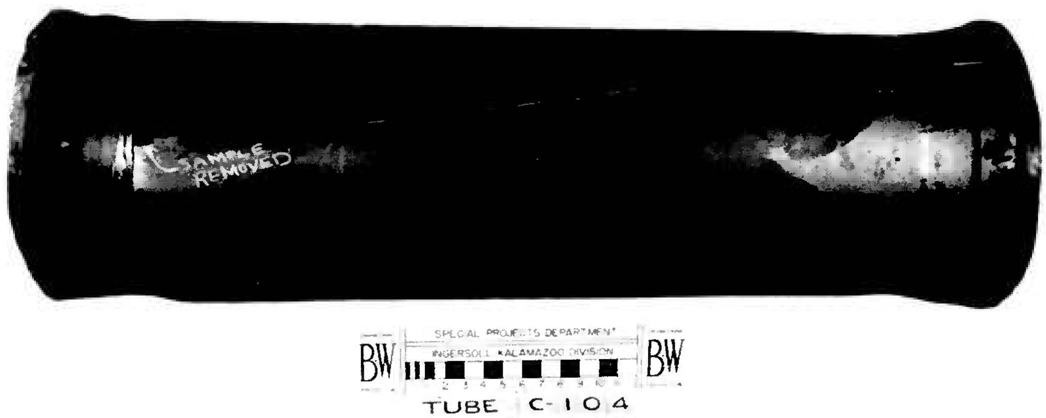


Figure 23. Photographs of Burst Patterns on Cases C-104 and R-31

Table XI. Roll and Weld Cases TIR - After Heat Treat

Tube No.	TIR Readings - Position (See Figure 8)				
R-1	.098	.104	.145	.120	.105
R-2	.028	.038	.163	.108	.102
R-11	.025	.040	.115	.100	.075
R-13	.070	.078	.125	.068	.079
R-14	.108	.080	.215	.145	.120
R-17	.026	.035	.130	.068	.070
R-18	.055	.055	.120	.105	.118
R-20	.087	.095	.137	.104	.119
R-21	.050	.060	.085	.070	.070
R-22	.038	.047	.138	.104	.096
R-26	.070	.095	.145	.127	.122
R-29	.030	.032	.114	.082	.085
Avg.	.057	.063	.136	.100	.097

After Heat Treat (No restraining fixture used)

It was considered that Condition VI completed the preliminary tests necessary prior to the fabrication of high strength evaluation cases by the roll and weld method. Unfortunately, due to funding it was not possible to fabricate and test a number of evaluation cases to complete the study. It is believed, however, that no difficulty would be encountered in producing cases of this design having a burst stress average of 280,000 psi with a ratio of burst stress to uniaxial tensile strength in excess of 1.0. Whether or not the strength levels of the rolled and welded cases would match those of spun seamless cases is not known, however, it was indicated that with refinements of the welding, material, decarburization cycle and location of weld they should become very nearly equal.

III. CONCLUSIONS

A. General

1. Limited surface decarburization was necessary before consistent high strength results could be obtained on MBMC #1 steel cases fabricated by either spinning or roll and weld.

2. MBMC #1 steel cases fabricated by both spinning and roll and weld resulted in high burst strengths.

3. The spun cases exhibited the higher burst strengths, however, the development of the roll and weld cases was not completed. Upon completion of the roll and weld development similar results should be obtained.

B. Flash Weld Blank Hydrospun Cases

1. MBMC #1 steel rolled and flash welded into cylindrical blanks spun as readily as forged blanks.

2. The flash weld was faulty having excessive inclusions causing premature case burst results.

3. It was indicated that with a clean flash weld results similar to forged and spun cases would be possible.

C. Forged Blank Hydrospun Cases

1. MBMC #1 steel is readily hydrospun into cylinders.

2. MBMC #1 steel can be cold reduced by spinning up to 80% before stress relieve is required.

3. Dimensional tolerances of roundness after spinning and stress relieve are good.

4. Dimensional tolerances of roundness after heat treat increase by a factor of 3 to 4 without the use of restraining fixtures.

5. Cases spun from forgings and heat treated in a neutral atmosphere burst at a low stress level.

6. Cases spun from forgings and heat treated in a decarburizing atmosphere burst at consistently high stress levels.

7. Cases spun from MBMC #1 steel forgings heat treated to decarburize the surface and tempered at 500°F consistently burst at a stress level above 280,000 psi with a ratio of burst stress to uniaxial tensile strength from 1.10 to 1.21 with an average of 1.15.

D. Roll and Weld Cases

1. MBMC #1 steel can be readily rolled and machined in the annealed condition and welded using either MBMC #1 or AISI 6130 steel wire.

2. Cases heat treated in a neutral atmosphere will burst at a low stress level.

3. Heat treating the cases in a decarburizing atmosphere improved the burst strength levels approximately 30%.

4. Heat treating the case in a decarburizing atmosphere after removal of all scale and grinding the girth welds flush improved the burst strength an additional 14%.

5. Relocating the girth welds with respect to the end rings improved the burst strength level an additional 13%.

6. Burst strengths of 280,000 psi are obtainable in the rolled and welded MBMC #1 cases with a ratio of burst stress to uniaxial tensile strength above 1.0 and it was indicated that higher strengths and ratios are feasible.

7. The burst strength levels obtained throughout the development of the various factors affecting strength became more consistent as refinements were made indicating good reliability.

8. The substitution of spun tube sections girth welded to end rings resulted in burst strengths the equivalent of rolled and longitudinally welded tube sections indicating that the girth welds were the cause of the premature failures that occurred.

IV. RECOMMENDATIONS

1. An additional ten cases by the roll and weld method should be fabricated to verify the reliability of the final fabrication procedure developed.

2. A study of the effect of surface decarburization should be made on MBMC #1 steel to determine the optimum depth and gradient evaluated by means of notch toughness specimens.

3. The surface decarburization specimen study should be verified by hydrotest prototype cases of both the seamless and rolled and welded types.

4. The effect of surface decarburization on the stress corrosion properties of the material should be investigated by specimen tests.

5. A more satisfactory method for the determination of decarburization depth should be investigated.

6. A detailed study of the microstructure of decarburized specimens should be made to provide a better understanding of the apparent beneficial effects on notch sensitivity.

7. MBMC #1 steel hydrotest cases of large diameter (60") fabricated by roll and weld should be tried utilizing decarburization.

CM-60-100

Roy C. Ingersoll Research Center
Borg-Warner Corporation
Des Plaines, Illinois

Project No. 3225

METALLURGICAL STUDY OF MBMC #1 STEEL

Final Report

by: Edmund J. Klimek

September 18, 1960

TABLE OF CONTENTS

ABSTRACT

Based on tensile data collected in this work and from burst strengths of welded containers fabricated and tested at Ingersoll Kalamazoo Division, MBMC #1 is superior to many materials presently being used for rocket motor cases.

The alloy is based on the strengthening effect of low cost, available silicon. Required amounts of critical and expensive alloying elements (Cr, V) are small.

The material can be machined, joined, and heat treated by conventional techniques and equipment.

MBMC #1 is highly susceptible to decarburization and scaling.

Alloying elements are highly segregated even in sheet stock .050" thick. This segregation has been shown to affect transformation rates on heat treatment and can be expected to affect physical properties. Graphite has been tentatively identified in conjunction with certain transformation products. This graphite is believed due, again, to segregation of alloying elements.

A large amount of non-metallic inclusions of the oxide and silicate type can also be expected to decrease mechanical properties.

Data on metallurgical properties of the alloy are presented. These include: tensile, impact, hardenability, machinability and isothermal transformation information.

I. INTRODUCTION

This report summarizes a study of the metallurgical properties of MBMC #1 alloy. The work was performed at the Roy C. Ingersoll Research Center of the Borg-Warner Corporation as a venture to the parent program.

Details of experimental work and collected data are appended. Discussion of experimental results follows in III.

II. CONCLUSIONS AND RECOMMENDATIONS

The data collected herein shows that MBMC #1 can be heat treated to strength levels equal to or greater than those of currently used missile alloys.

These high strengths are accompanied by low tensile ductilities in most strength levels. As discussed within the report, these ductility levels will most likely be brought up with high quality melts of the material.

If further work were to be performed on the material, initial efforts should be expended on production of clean, homogeneous material.

III. DISCUSSION

A. Microstructure

Prior to any experimental work, specimens were removed from each of the several lots of material to be used in the program.

All materials had a completely spheroidized structure except for the forging bars which contained some coarse pearlite. Generally speaking, all materials had low MnS contents, but the oxide and silicate type inclusion count was high. The inclusions consisted of both continuous and broken up stringers.

The surface on all materials were depleted of carbon. Even the cold finished material contained areas of free ferrite on the surface. Hot finished plate (1/2" thick) contained up to .010" complete decarburization. The hot rolled materials also contained large amounts of an adherent scale on the surface and folded into the material.

B. Tensile Testing

Average tensile data obtained in testing as a function of heat treatment are presented in Figure 1.

The material shows a maximum in yield and ultimate strengths after tempering at 600°F for two hours.

The rise in tensile values with tempering from 400 to 600°F is unaccounted for. The transformation of retained austenite must be ruled out because there is no evidence of its occurrence in quenched specimens.

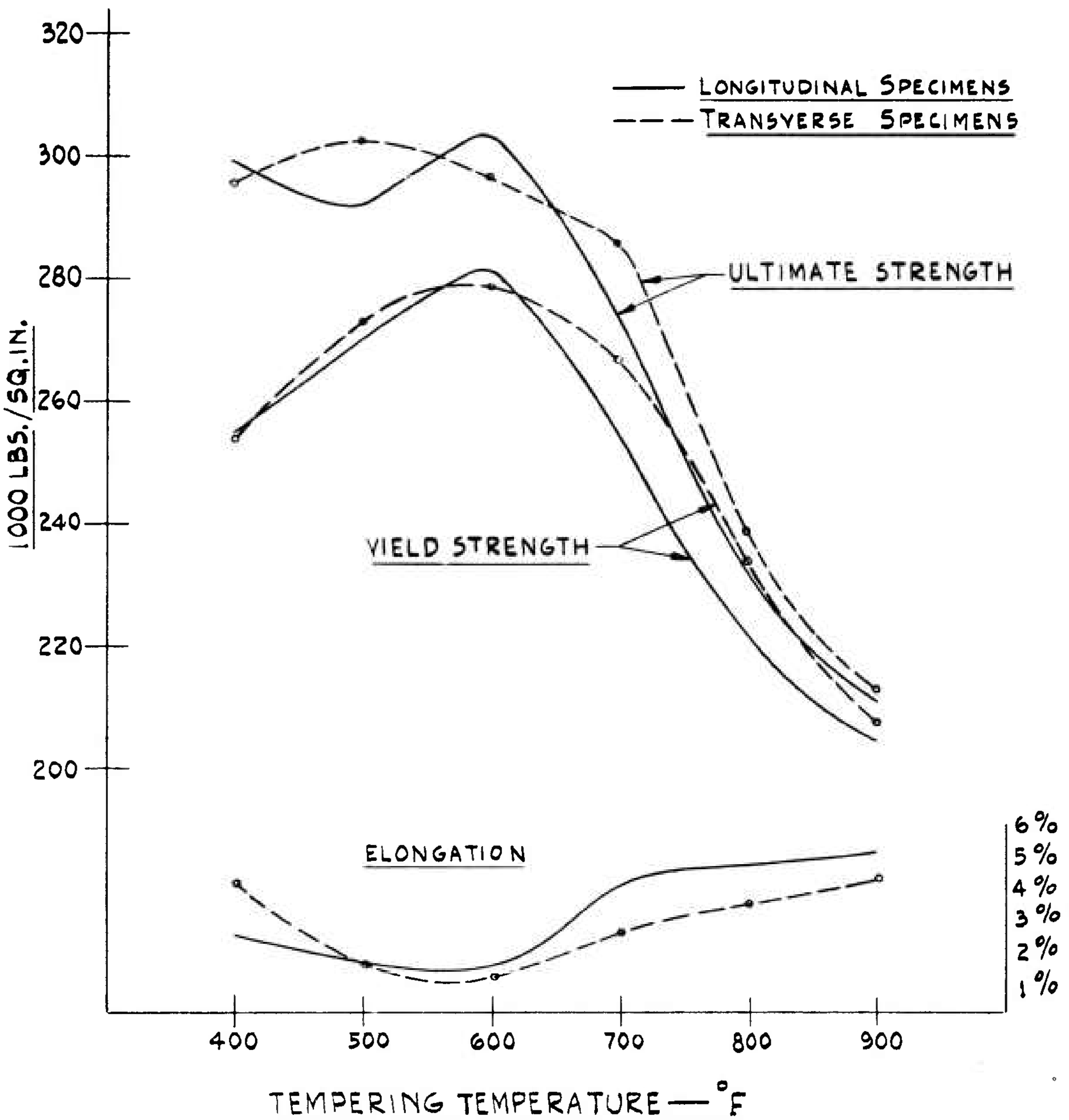
Figure 1

COMPOSITION %

CARBON	MANGANESE	SILICON	CHROMIUM	VANADIUM
0.45	0.76	1.42	0.76	0.08

HEAT TREATMENT : QUENCHED FROM 1600 °F, QUENCHED INTO OIL @125 °F.

PHYSICAL PROPERTIES OF MBMC #1 - 0.050" THICK.



MBMC #1 alloy resists softening (decrease in strength) on tempering until a temperature of about 750°F is employed. At this point, strength values drop off rapidly. The maintenance of strength on tempering is most likely due to the ferrite strengthening effect of silicon.

Tensile ductility is generally low and levels off in material tempered beyond 750°F. This leveling off, rather than the usual increase in ductility, is most likely due to the large amount of non-metallics in the material. It is believed that with a cleaner and less banded material, ductility would generally increase.

C. Impact Testing

Figure 2 illustrates the plotted data obtained from Charpy impact tests on quenched specimens of MBMC #1 steel tempered at 500 and 600°F.

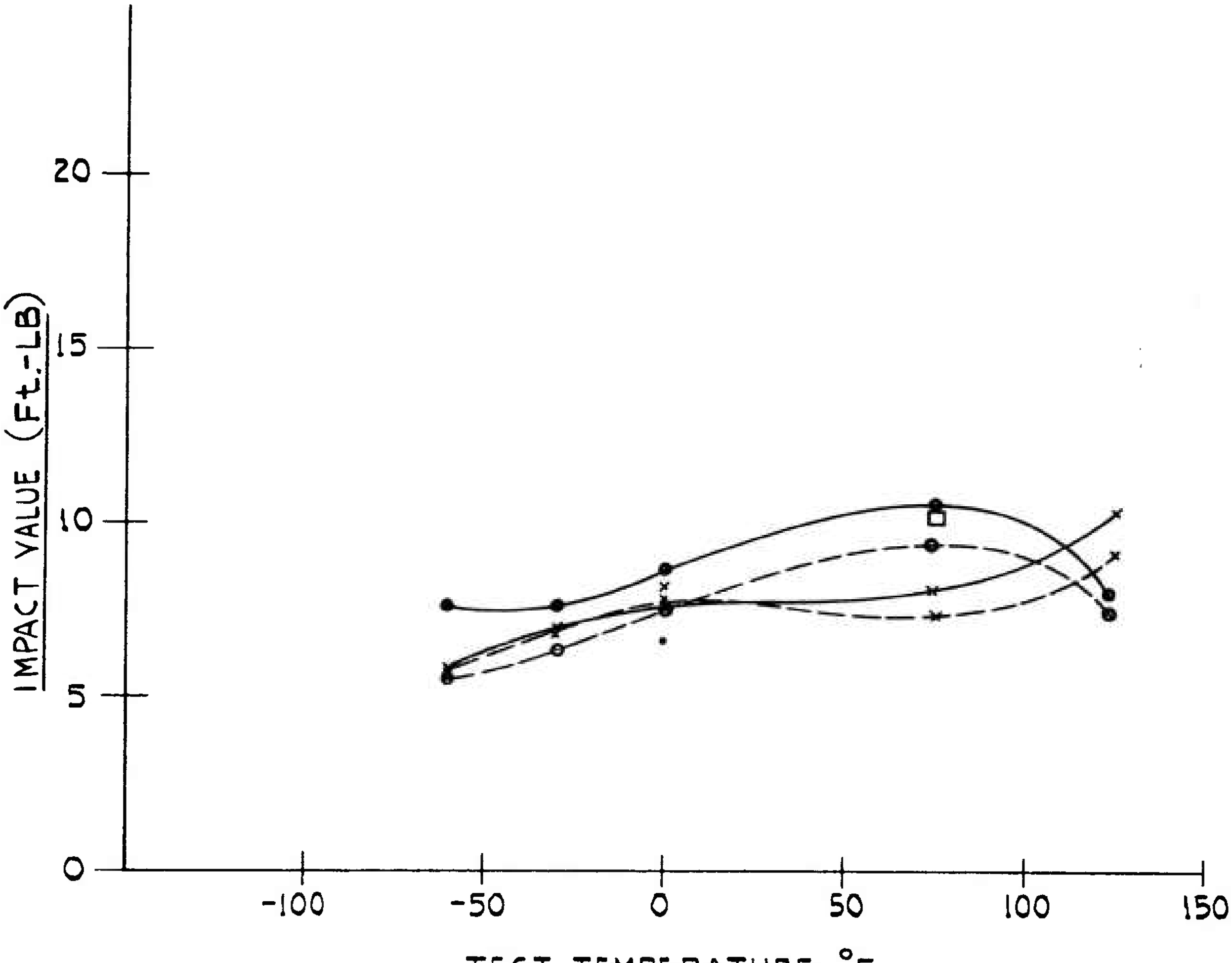
The results as plotted in Figure 2 show no significant difference in impact values for specimens tempered at the two temperatures. Above 0°F, the specimens, both longitudinal and transverse, tempered at 500°F exhibit greater toughness. The 500°F specimens also exhibit a peak toughness value with a rather steep drop-off with increasing temperatures.

The 600°F specimens show a gradual decrease in toughness with lower testing temperature.

In no case can the curves be construed to indicate an abrupt transition from ductile to brittle. The rate of decrease in ductility is that which would be expected in low alloy tempered martensite.

Figure 2

CHARPY V-NOTCH IMPACT STRENGTH AS A FUNCTION OF TEMPERATURE FOR MBMC*1 STEEL.



HEAT TREATMENT: 1600°F FOR 1 HR, OIL QUENCH:

– – – –TRANSVERSE ; ×–TEMPERED 2 HRS AT 600°F.

———LONGITUDINAL; O–TEMPERED 2 HRS AT 500°F.

□ –AUSTEMPERED 35 MIN. AT 500°F.

Examination of fracture surfaces reveals a straited fracture indicative of banding or non-metallic inclusions.

D. Grain Size Measurement

Experiments were conducted to determine if a grain coarsening tendency existed in the material in the normal range of heat treating temperatures, i. e., hardening-annealing.

Except for a slight growth of very fine grains (below ASTM size #8), there was no grain growth in specimens heated at temperatures up to 1800°F.

The materials used in this program were from heats of the alloy produced by three different vendors. Generally, all materials had a fine grain size. The measured austenitic grain size at 1600°F for all the materials used in this work was in the range of ASTM numbers 6-8. All materials had a duplex grain size, the effect becoming more pronounced with the thinner gauge materials. This duplexity is a function of alloy segregation in the materials, and has a large effect on response to transformation as will be discussed in succeeding sections.

E. Critical Temperatures

Critical temperatures for MBMC #1 alloy as required for austenitizing and spheroidize annealing were determined.

Eutectoid temperature (A_1) was determined as 1400°F ± 5°F.

Minimum temperature for complete austenitization was determined as 1575°F ± 5°F.

F. Isothermal Transformation Diagram

An isothermal transformation diagram for MBMC #1 alloy was determined and is presented in Figure 3.

The diagram illustrates two minimums in the start of transformation with the shortest time occurring at about 750°F.

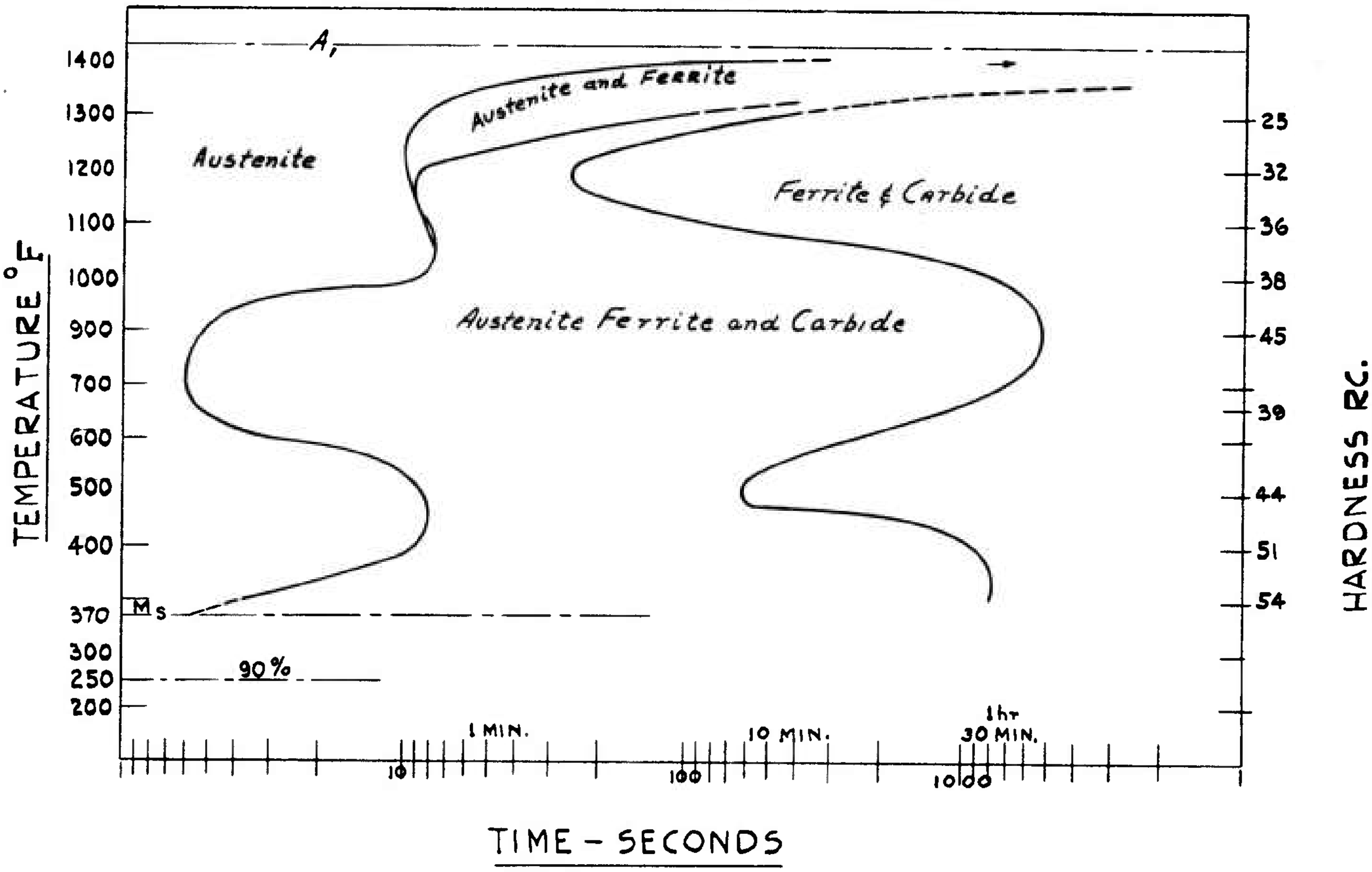
Specimens used in this work were removed from 1/2" plate stock which exhibited a large amount of alloy segregation. As illustrated in Figures 4 - 15, this segregation resulted in selective transformation depending on chemistry of the banded structure. To maintain a constancy of interpretation, any sign of transformation was considered as start of transformation regardless of banding in the specimen.

This condition necessarily resulted in times of transformation that would be premature in a homogeneous alloy. The alloy segregation plus the very fine grain size of segregated areas resulted in a start of transformation curve that can probably be pushed back by at least two seconds at the 750°F knee with a homogeneous material. The finish of transformation would most likely decrease by relative amounts of time.

Another effect of alloy segregation in the material was the tentative identification of graphite in specimens transformed at 1000°F. Within the limits of this venture, only a brief attempt at identification was made. A Formvar replica was pulled from a heavily etched specimen of steel which had been held at 1000°F for 41 minutes after austenitizing. The replica contained particles removed from the surface of the specimen. The

Figure 3

ISOTHERMAL TRANSFORMATION DIAGRAM FOR MBMC #1 STEEL



C - 0.44
MN - 0.80
SI - 1.59
CR - 0.79
V - 0.06

GRAIN SIZE = 50% #8
20% #7
30% #6

Ae_1 TEMPERATURE = 1425 °F

Ae_3 TEMPERATURE = 1575 °F

AUSTENITIZED AT 1600 °F

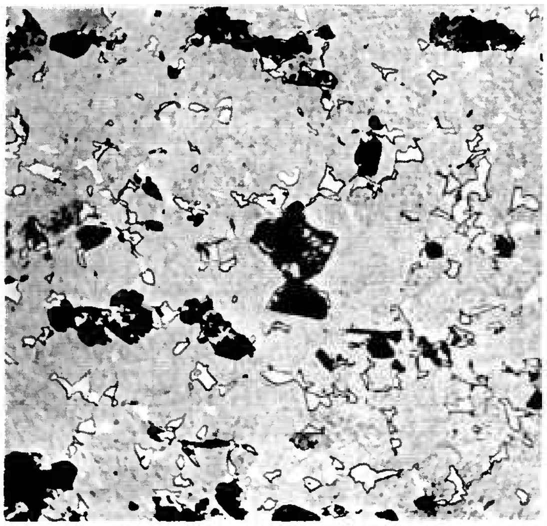


Figure 4 Neg. #767
1300°F - 3-1/2 minutes

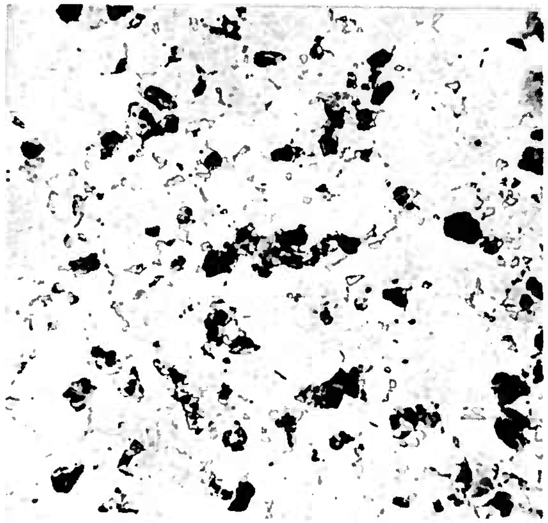


Figure 5 Neg. #768
1205°F - 30 seconds

Microstructures of MBMC #1 alloy. austenitized at 1600°F for one hour. transformed as indicated; then water quenched. All photographs at 250X except where noted. All specimens etched in 2% Nital - 2% Picral solution. The photomicrographs illustrate general characteristics of the transformation product rather than any particular stage of transformation.

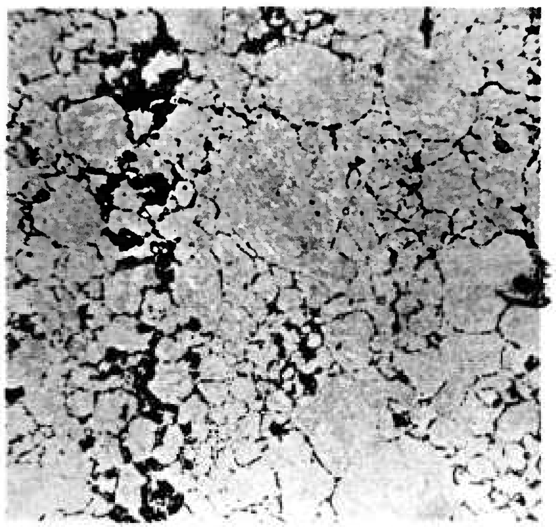


Figure 6 Neg. #769
1100°F - 32 seconds

Figure 7 Neg. #770
1000°F - 41 minutes

Figure 8 Neg. #777
1000°F - 41 minutes

Figure 9 Neg. #771
940°F - 20 seconds

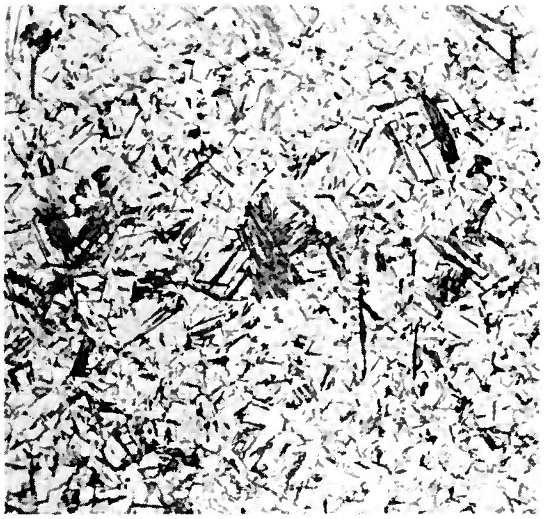


Figure 10 Neg. #772
750°F - 20 seconds

Figure 11 Neg. #773
660°F - 40 seconds

Figure 12 Neg. #774
$600^{\circ}F$ - 2 minutes

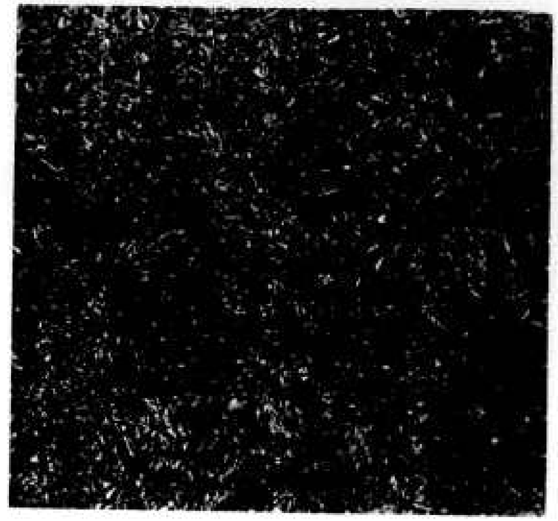


Figure 13 Neg. #775
$500^{\circ}F$ - 15 minutes

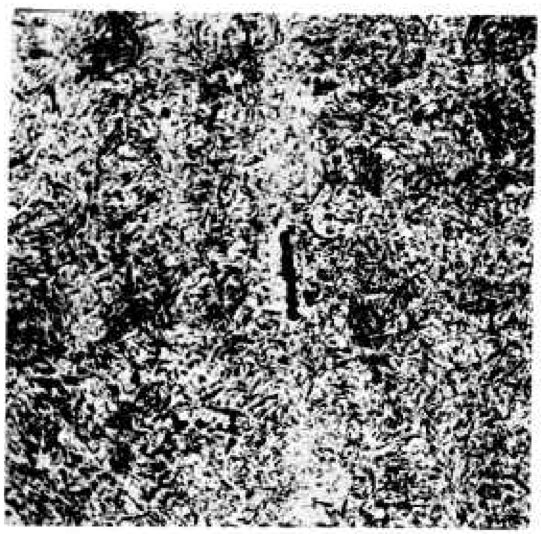


Figure 14 Neg. #776
$400^{\circ}F$ - 14 seconds

Figure 15 Neg. #778
$1000^{\circ}F$ - 20 minutes &
41 minutes
(Mag. X2)

single pattern produced by electron diffraction showed only one strong line for graphite. Details of diffraction studies are contained in Appendix II.

Figure 15 at low magnification illustrates on a macro scale the banding or actually dendritic structure still existing in 1/2" plate of MBMC #1 alloy.

Microstructures of MBMC #1 alloy, austenitized at 1600°F for one hour, transformed as indicated; then water quenched. All photographs at 250X except where noted. All specimens etched in 2% Nital - 2% Picral solution. The photomicrographs illustrate general characteristics of the transformation product rather than any particular stage of transformation.

G. Hardenability

Figure 16 illustrates the hardenability measurement of two bars each of MBMC #1 and 4130 alloy.

The MBMC alloy has a considerably greater "H" value or depth of hardening than 4130 material. Hardness values are also higher as would be expected because of carbon content.

The difference in hardenability values between two bars of MBMC #1 from the same heat and with essentially the same chemistry can again be attributed to segregation in the alloy.

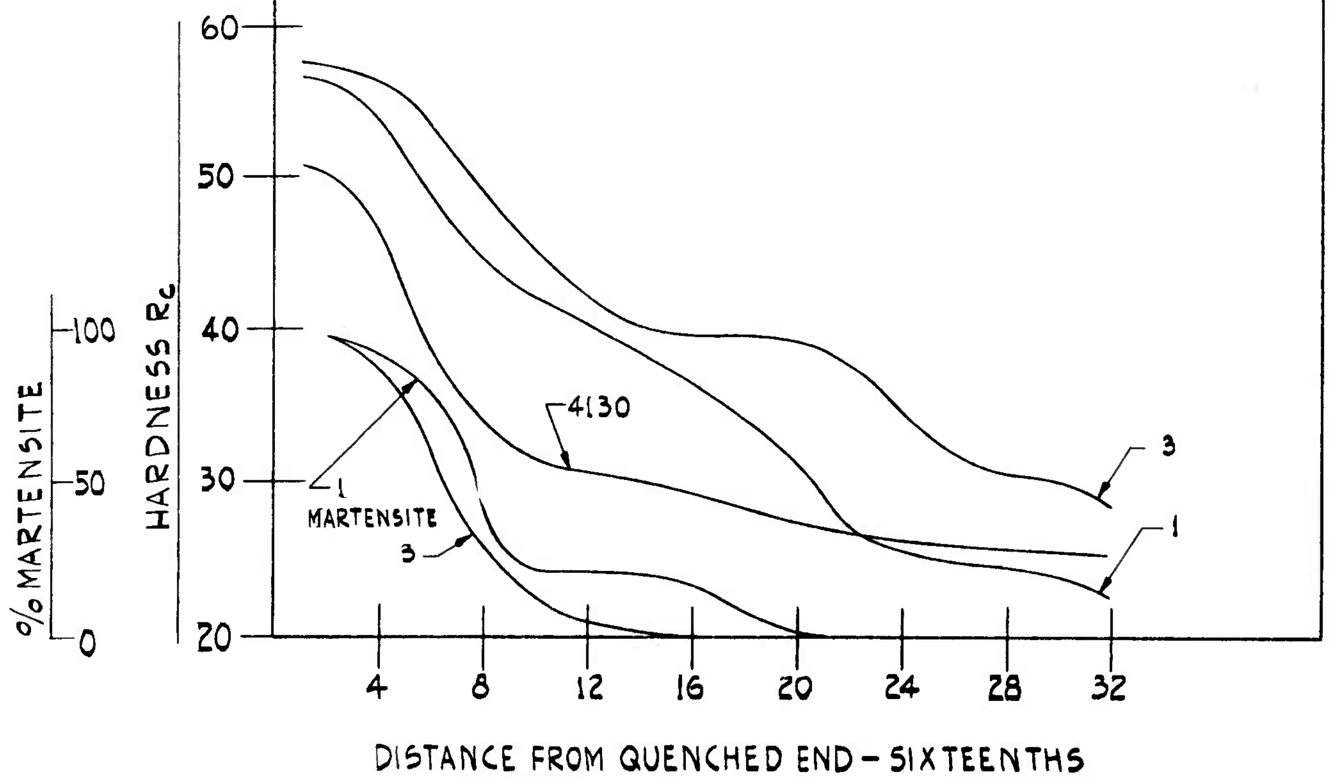
Specimen #1, though having a larger amount of martensite than bar #3, has a generally lower hardenability value.

Examination of microstructure of the two bars finds them to vary considerably in the type and amount of transformation product formed. Bar #1

Figure 16

END QUENCH HARDEN ABILITY CURVES FOR MBMC #1 & 4130 STEELS.

		C	Mn	Si	Cr	V	GRAIN SIZE
MBMC #1	(1)	.41	0.70	1.48	0.71	0.09	15%-7, 85%-8
MBMC #1	(3)	.41	0.68	1.46	0.73	0.07	50%-7, 50%-8
4130		.28/.33	.40/.60	.20/.35	.80/1.10		



has martensite plus ferrite at the quenched end followed by formation of the light etching upper bainite structure which is soft. Bar #3 has a lesser amount of martensite but the transformation products are of the lower bainite type which is a finer and harder structure. There is almost no ferrite formation in Bar #3. Macro examination of the bars after they have been polished and etched, indicates a much greater amount of banding in bar #3 than in bar #1.

To determine grain size of the material used in the hardenability test, one specimen of each material was quenched from 1600°F to 1100°F and held for 30 seconds. The specimen from bar #1 was about 75% transformed to bainite, where the specimen from bar #3 was only about 10% transformed. The transformation again was very dependent on alloy segregation in the specimen.

In examining the structure of bar #1 along its length, the same type of constituent tentatively identified as graphite during T. T. T. diagram work was noticed. The graphite type constituent occurs in areas where the light etching upper bainite phase exists.

Relative to hardenability, the structure of one inch diameter bars when oil quenched from 1600°F compared favorably with the results of the Jominy Tests. These specimens are discussed in detail in the following section.

H. Master Tempering Curve

The effect of tempering temperature on quenched MBMC #1 material of various sizes was determined. Figures 17 and 18 illustrate the effect of tempering temperature and size on material hardness.

The structure of .050" sheet and .250" plate is completely martensitic after oil quenching. The one inch diameter bar exhibited increasing amounts of ferrite and upper bainite when approaching the center of the bar. There was no evidence of the graphite type phase in the oil quenched one inch diameter specimens.

Specimens tempered at 800, 900, and 1000°F were examined to determine if any graphitization may have occured on tempering. There is no evidence of graphite by visual examination. The specimen tempered at 1000°F has a macro structure indicating homogeneity in the alloy as evidenced by different rates of tempering.

I. Annealing Cycle

Determination of a spheroidizing cycle for MBMC #1 alloy has been performed in the past. This work was performed on the same material as used in construction of the transformation diagram.

This work is reported in detail in Appendix III.

The optimum cycle for producing fine spheroidal carbides in the alloy is as follows:

Heat to 1650°F - hold one hour and air cool or oil quench if possible.

Heat to 1350°F - 1400°F maximum - 24 hours and air cool.

Spheroidizing times will be dependent on material size. The 24-hour cycle cited above is for 1/2" thick plate. It is expected that sheet stock will require shorter times.

Figure 17

HARDNESS VS TEMPERING TEMPERATURE ON MBMC #1 STEEL OF DIFFERENT THICKNESS.

QUENCHED FROM 1600 °F. TEMPERED FOR TWO HOURS AT INDICATED TEMPERATURE.

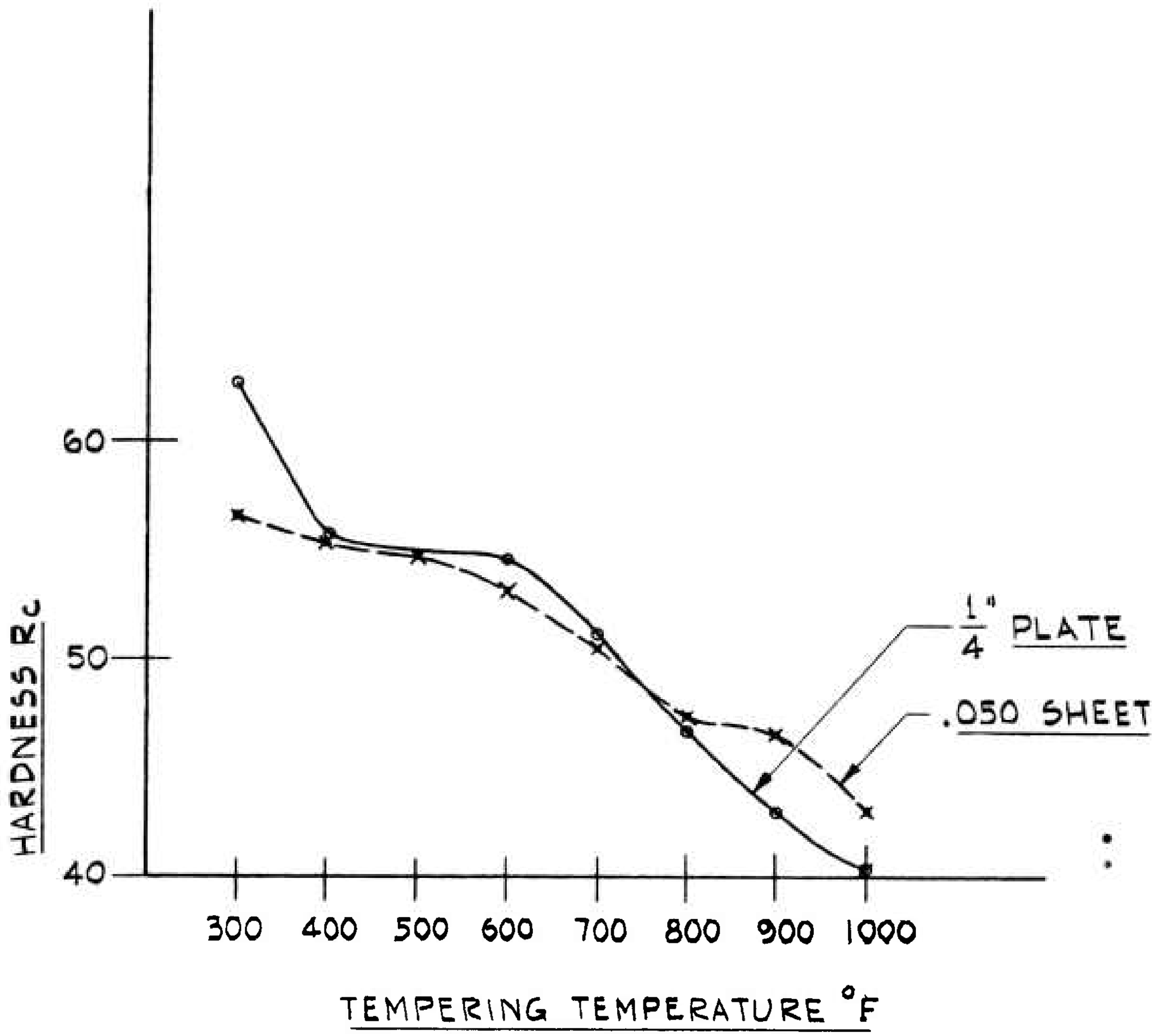
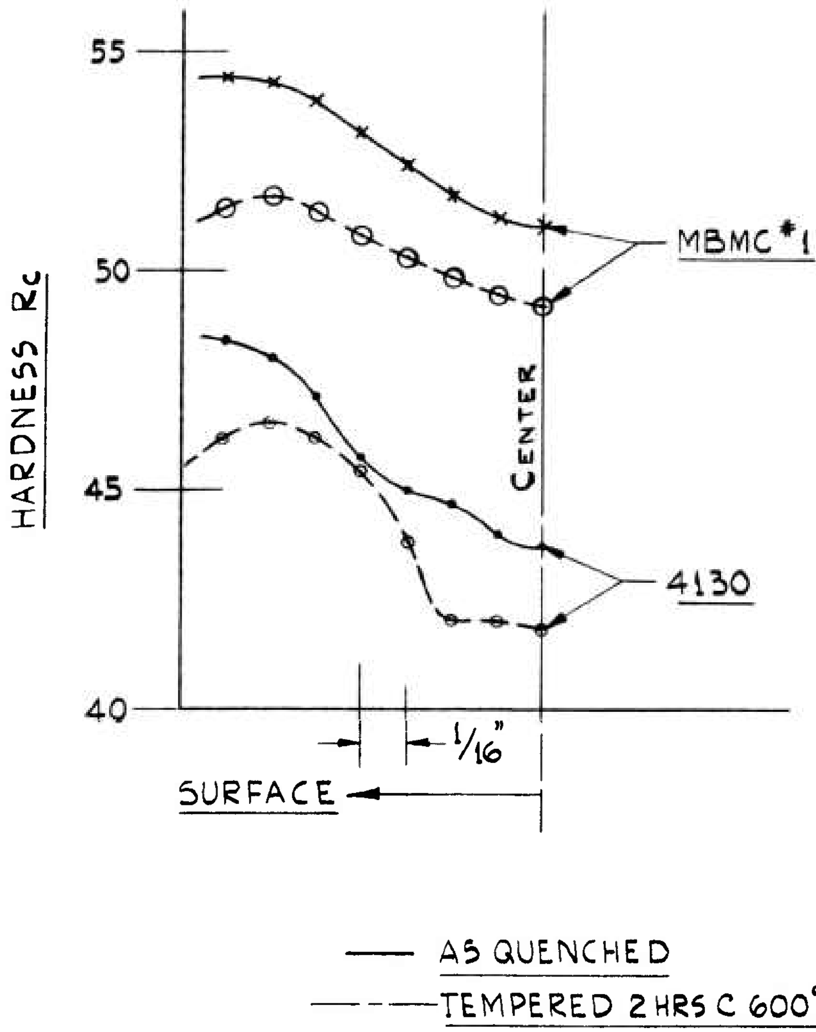


Figure 18

HARDNESS CONTOUR ON 1" DIA. BARS AT MBMC#1 & SAE 4130 STEELS.

SPECIMENS QUENCHED FROM 1600°F INTO OIL AT 125°F.



The air hardening tendency of MBMC #1 alloy will produce a martensite-bainite mixture, especially on sheet stock. This allows for optimum spheroidization.

J. Primary Mechanical Working

Appendix IV contains a brief outline of breakdown and rolling procedures for MBMC #1 steel as practiced at the Ingersoll Steel Division of Borg-Warner.

Their work was performed on small ingots produced from a total heat of five tons.

Of the several vendors who have produced heats of this alloy, Ingersoll Steel had perhaps the greatest initial success in producing sheet stock with some measure of surface control as regards chemistry.

It could be expected that with more work on producing the material, especially in large heats, difficulties like segregation, non-metallics and surface decarburization could be minimized or eliminated.

K. Contributing Personnel

The work reported herein was performed within the limits of time and money allocated.

Experimental work was performed by the following Research Center personnel:

Heat Treatment & Metallograph: E. J. Klimek, A. K. Draeger, Jr.

Electron Diffraction: M. Louise Pierotti

Impact Testing: N. Butzow, E. Cliff, M. Motivala

Analytical Chemistry: B. Olson

Edmund J. Klimek

APPENDIX I

EXPERIMENTAL PROCEDURES AND DATA

A. Tensile Test

Specimens for tensile tests were machined and delivered by Ingersoll Kalamazoo Division. Specimens were machined according to Figure 19. Specimens were heated at 1600°F for one hour in a Heavy Duty Muffle Furnace (type AB 702-90A). A dissociated ammonia atmosphere enriched with propane was employed to restore surface carbon and maintain the 0.43% C of the material. All specimens were identified as to number and direction of rolling prior to heat treatment. A 1/8" diameter hole was drilled in the grip end of the specimens to facilitate fixturing during heat treatment.

Ten specimens were treated at a time, hung vertically from a fixture approximately 4" x 12" square. The average data obtained from the ten specimens heat treated at the same time, constituted one point in Figure 1. The austenitized specimens were quenched vertically in agitated oil at 25°F, and tempered immediately after quenching.

Furnace temperature was maintained by an L. & M. Speedomax, Model H instrument, and checked at least once during each austenitizing cycle by a portable potentiometer with the thermocouple placed in the gauge length area of the specimens.

Tempering was performed in a Heavy Duty Temperite forced air furnace. The specimens were placed horizontally on a wire mesh shelf in the furnace.

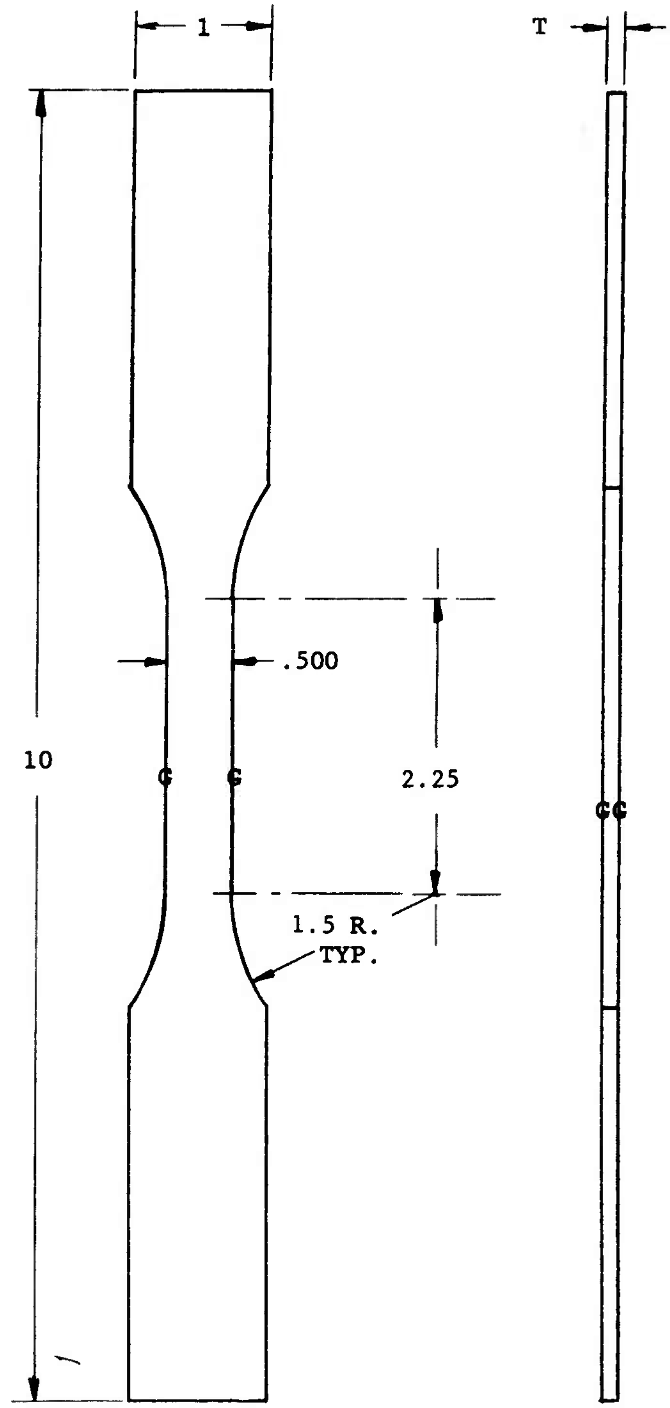


FIGURE 19
TENSILE TEST SPECIMEN

After heat treatment, the specimens were delivered to Ingersoll Kalamazoo Division. All specimens were sanded to remove scale and/or oil contamination, and to remove any notches or scratches. One out of five specimens was tested with a strain gauge to compliment and verify the stress-strain curve produced by the machine.

Average tensile values are included as Table I and plotted as Figure 1.

Detail tensile data and curves are filed in Research Center vaults under Program CM 60-100.

Specimens used for tensile tests had the following chemical analysis:

Carbon	0.425%
Manganese	0.76%
Phosphorous	0.019%
Sulphur	0.020%
Chromium	0.76%
Vanadium	0.08%
Silicon	1.42%

B. Impact Testing

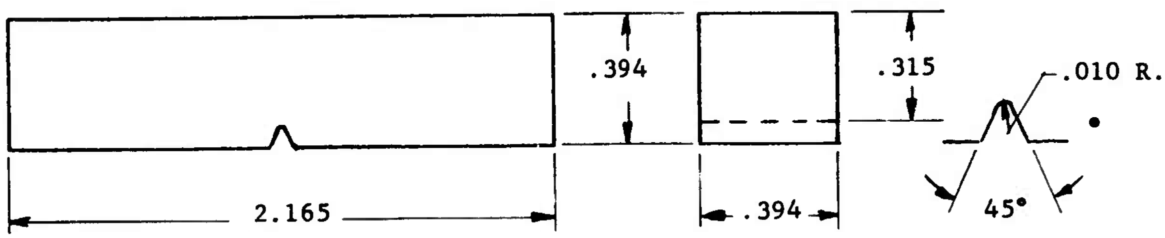
Charpy V-Notch impact specimens were rough machined including the notch, prior to heat treatment. A finish grind allowance of .010" was retained for final grinding after heat treatment to remove any possible surface effects. Final machining was performed according to Figure 20.

The specimens were austenitized in the same furnace as used for tensile specimens. The specimens were hung vertically in the furnace and quenched in the vertical position into agitated oil at 125^{o}F. Specimen temperature was checked at least once during the austenitizing cycle by

Table I

PHYSICAL PROPERTIES OF MBMC #1 - 0.050" THICK

Heat Treatment Harden °F	Temper. °F	Specimen Orientation	Yield Strength 0.2% Offset psi	Ultimate Strength psi	Elong. % in 2"	Hardness Rc	YS/US
1600	400	Longitudinal	255,400	299,700	2.6	54.6	0.85
1600	400	Transverse	248,900	296,100	4.3	52.5	0.84
1600	500	Longitudinal	269,900	292,200	1.7	54.8	0.92
1600	500	Transverse	273,800	302,800	1.6	54.8	0.90
1600	600	Longitudinal	282,000	303,300	1.5	55.0	0.93
1600	600	Transverse	278,200	296,500	1.2	54.4	0.94
1600	700	Longitudinal	255,200	274,100	4.3	52.0	0.93
1600	700	Transverse	266,800	286,300	2.6	52.1	0.93
1600	800	Longitudinal	221,500	231,100	4.8	45.5	0.96
1600	800	Transverse	234,000	243,700	3.5	48.6	0.96
1600	900	Longitudinal	204,200	210,800	5.2	44.1	0.97
1600	900	Transverse	207,400	212,600	4.2	44.4	0.98



PERMISSIBLE VARIATIONS:

CROSS-SECTION DIMENSIONS ------------------± 0.001

LENGTH OF SPECIMEN----------------------- ± 0.010

ANGLE OF NOTCH -------------------------- ± 1°

NOTCH SHALL BE CUT WITH A SPECIAL FORMED MILLING CUTTER OR IN THE CASE OF VERY HARD MATERIAL BY GRINDING.

THE NOTCH IS TO HAVE A RELATIVELY SMOOTH SURFACE.

FIGURE 20

V-NOTCHED CHARPY TYPE IMPACT SPECIMEN

portable potentiometer.

Impact specimens were also tempered in the forced air tempering furnace immediately after cooling.

Specimens were tested on a Reihle Impact Testing Machine.

Specimens tested below room temperature were cooled in dry ice and perchlorethylene to temperatures below testing temperature. The specimens with thermocouples affixed were placed in the machine and held in the fixture until they rose to the desired temperature, then broken.

Detail data on each specimen is included as Table II. Average values are plotted in Figure 2.

Impact specimens were machined from 1/2" thick plate, produced by the United States Steel Company and analyzed by them as follows:

Carbon	0.40%
Manganese	0.79%
Phosphorous	0.03%
Sulphur	0.012%
Chromium	0.82%
Vanadium	0.02%
Silicon	1.53%

C. Transformation Diagram

An isothermal transformation diagram was constructed for MBMC #1 steel of the following analysis:

Carbon	0.44%
Manganese	0.50%
Silicon	1.59%
Chromium	0.79%
Vanadium	0.06%

Table II

IMPACT TESTS ON MBMC #1 STEEL (CHARPY)

SPECIMEN	TEMPERATURE °F	IMPACT STRENGTH FT-LB (1)	IMPACT STRENGTH (2)	IMPACT STRENGTH (3)	IMPACT (4)	STRENGTH (5)	IMPACT STRENGTH AVERAGE
GROUP I	+125°	8.25	7.50	8.25	6.00	7.00	7.40
Transverse	Room (72°)	8.50	11.25	8.50	---	9.50	9.44
Tempered at 500°F	0°	7.50	7.50	8.00	7.75	6.50	7.45
	-30°	----	5.50	7.75	5.75	6.00	6.25
	-60°	6.25	4.75	6.25	5.00	5.50	5.55
GROUP II	+125°	8.00	6.75	7.50	8.25	8.75	7.85
Long. at 500°F	Room (72°)	10.25	9.75	11.50	11.00	----	10.62
	0°	9.00	8.50	6.75	9.50	9.00	8.55
	-30°	8.00	7.00	8.50	7.00	8.50	7.60
	-60°	8.00	6.75	7.25	8.75	7.75	7.70
GROUP III	+125°	10.00	13.00	10.00	9.25	8.75	10.20
LC 600	Room (72°)	8.50	8.50	7.00	8.00	-----	8.00
Long at 600°F	0°	7.50	7.50	8.25	7.25	8.25	7.75
	-30°	8.25	6.50	5.75	7.50	6.75	6.95
	-60°	6.25	5.75	5.25	6.75	5.75	5.95
GROUP IV	+125°	9.50	10.75	8.00	9.00	8.00	9.05
TC 600	Room (72°)	8.00	6.75	8.50	6.75	7.25	7.45
Trans. at 600°F	0°	8.00	8.25	7.50	8.75	----	8.12
	-30°	6.25	7.50	6.75	7.50	7.00	7.00
	-60°	5.75	5.00	6.00	5.50	6.75	5.80
Austempered at 500°F	Room (74°)	9.25	11.25	10.00	9.50	11.75	10.35

Specimens 3/8" square and approximately 1/4" thick were removed from 1/2" thick spheroidize annealed plate. Specimens were ground on all surfaces to remove any decarburization and to maintain constant size.

Specimens were held in open coils of chromel wire during heat treatment. This afforded a sure way of holding the specimen with a maximum cooling rate on quenching.

The specimens were austenitized in a large (8" dia. x 14" deep) salt pot containing Aero Heat 1000 salt as manufactured by American Cyanamid Company. Temperature was maintained by a Wheelco Capacitrol instrument and checked periodically during the austenitizing cyle by a portable potentiometer.

Isothermal transformation was carried out in electrically heated pot furnaces (6" diameter x 10" deep) containing different media depending on transformation temperature.

The following quenching media were used:

For temperatures above 1000°F - Aero Heat 1000

For temperatures between 400°F and 1000°F - Aero Heat 300

For temperatures below 400°F - Cerro Bend, low melting alloy

An arbitrary time of two seconds was allowed for specimens to cool to the transformation temperature. Transformation temperature was measured with a portable potentiometer. After transformation for selected times at each temperature, the specimens were water quenched.

Start and finish of martensite transformation were determined by quenching to a selected temperature from 1600°F and holding for 15 seconds. The specimen was then quenched into liquid salt at 1400°F and held for 15 minutes, then water quenched. In this manner, the martensite formed at the low temperature is tempered at 1400°F. and etches dark. The untransformed austenite quenches to light etching martensite.

After transformation, all specimens were ground to remove a minimum of .015" from one surface and hardness measured on the Rockwell "C" scale. After hardness measurement, the specimens were examined under the microscope to determine degree of transformation. Table III contains hardness data for the T. T. T. specimens.

In all cases, metallographic analysis was relied upon to determine start and finish of transformation. Table IV, below, summaries metallographic analysis of all specimens.

Table IV
METALLOGRAPHIC ANALYSIS OF TRANSFORMATION DIAGRAM SPECIMENS

At 1400°F	Ferrite begins in 10 minutes No pearlite transformation after 45 minutes
At 1300°F	Ferrite starts in 23 seconds Pearlite starts in 185 seconds Pearlite complete in 690 seconds
At 1200°F	Ferrite starts in 15 seconds Pearlite starts in 22 seconds Pearlite complete in 85 seconds
At 1100°F	Ferrite starts in 22 seconds Pearlite starts in 28 seconds Pearlite ends in 180 seconds

Table III

HARDNESS OF TRANSFORMATION DIAGRAM SPECIMENS

Transformation Temperature	Time	Hardness Rc	Transformation Temperature	Time	Hardness Rc
1400°F	60 sec.	61.9	1300°F	25 sec.	59.2
	5 min.	61.0		40 sec.	60.2
	10 min.	61.0		80 sec.	59.2
	15 min.	61.3		2 min.	61.2
	20 min.	61.0		* 3 min.	57.5
	25 min.	61.0		3½ min.	56.5
	35 min.	60.7		7½ min.	28.0
	45 min.	61.0		+11 min.	25.0
1200°F	5 sec.	61.0	1100°F	6 sec.	60.6
	15 sec.	61.3		8 sec.	60.4
	*25 sec.	59.7		16 sec.	60.8
	30 sec.	57.7		24 sec.	60.5
	40 sec.	51.2		*32 sec.	60.1
	60 sec.	33.1		+ 3 min.	36.0
	+120 sec.	30.1		6 min.	32.0
1000°F	20 sec.	61.7	940°F	5 sec.	60.2
	*30 sec.	60.3		*10 sec.	59.1
	40 sec.	60.4		20 sec.	55.6
	5 min.	59.5		40 sec.	57.9
	14 min.	58.4		2 min.	53.5
	21 min.	56.5		4 min.	51.4
	40 min.	55.5		29 min.	50.0
	+60 min.	38.5		50 min.	48.0
				+80 min.	45.0
750°F	2 sec.	60.8	600°F	3 sec.	61.0
	3 sec.	59.7		5 sec.	60.7
	4 sec.	57.0		10 sec.	59.6
	* 5 sec.	60.8		20 sec.	60.7
	10 sec.	59.9		*30 sec.	61.3
	2 min.	49.3		40 sec.	60.9
	4 min.	40.8		2 min.	51.2
	+20 min.	38.1		+ 8 min.	44.4
500°F	*20 sec.	59.6	400°F	5 sec.	57.7
	30 sec.	59.6		*14 sec.	58.4
	40 sec.	59.5		20 sec.	57.3
	60 sec.	59.2		40 sec.	56.9
	90 sec.	58.8		70 sec.	57.2
	3 min.	54.1		90 sec.	57.0
	15 min.	52.6		15 min.	55.4
	+30 min.	51.3		+35 min.	54.5

+ Transformation Complete
* Transformation Starts

Table IV, continued

At 1000°F	No Ferrite visible Bainite (light etching) starts in 25 seconds Bainite complete in 3500 seconds
At 940°F	Bainite starts in 7 seconds Bainite complete in 4800 seconds
At 750°F	Bainite starts in 4-1/2 seconds Bainite 90% complete in 1200 seconds
At 600°F	Bainite starts in 25 seconds Bainite complete in 480 seconds
At 500°F	Bainite starts in 20 seconds Bainite complete in 1800 seconds
At 400°F	Bainite starts in 7 seconds Bainite ends in 2800 seconds

Martensite starts forming at 370°F

Martensite 90% complete at 250°F

The above data and particulars of steel chemistry are plotted as Figure 3.

D. Critical Temperatures

A_1 and A_3 temperatures were determined on .050" thick sheets of 0.43% carbon material as used for tensile specimens.

One inch square specimens were heated in an atmosphere controlled muffle furnace for one hour, then water quenched. After quenching, hardness was measured and specimens prepared for metallographic analysis. Hardness measurement proved inconsistent and inaccurate.

Specimens were heated at succeedingly higher temperatures until critical temperatures were reached.

Specimens quenched from 1400°F and up showed increasing amounts of austenite at these temperatures, as evidenced by white martensite after quenching.

At 1575°F, no ferrite or carbides remained, indicating 100% austenite at that temperature.

A_1 temperature is considered as 1400°F ± 5°F

A_3 temperature is considered at 1575°F ± 5°F

E. Austenitic Grain Size

A series of experiments were performed to study grain growth characteristics of the MBMC #1 alloy.

Specimens as used for transformation diagram work were quenched from a series of selected temperatures into molten salt at 1200°F. Work on the transformation diagram showed that complete transformation at 1200°F produces a fine, dark etching pearlite outlined with grain boundary ferrite.

Specimens were held one hour in a controlled atmosphere muffle furnace, quenched into salt and held 35 minutes, then quenched in water.

Specimens were quenched into salt at 1200°F from the following temperatures: 1600°F, 1700°F, 1750°F and 1800°F.

No significant increase in grain size was noted. A duplex mixture of grains of ASTM size 7 and 8 was present in most specimens. The only noticeable change was the growth of very fine grains (<8) to a size of 7 or 8. The growth was gradual with no great change with slight temperature increase.

G. Hardenability Tests

Jominy hardenability tests were performed on two specimens, each of MBMC #1 and 4130 steel.

One inch diameter by three inch long specimens were machined from 1-1/2" diameter forged bars. Specimens were machined and tests performed according to ASTM specifications as described on page 489 of the 1948 ASM Handbook.

The tests were performed by Professor Earl J. Eckel of the University of Illinois.

Resultant data is plotted as Figure 16. The chart shows a band of hardenability, actually being the spread in values of the two bars tested.

Chemical analysis and grain size of the two materials were determined to find possible cause for difference in values.

	C	Mn	Si	Cr	Va	Grain Size
Bar #1	0.411	0.704	1.48	0.713	0.059	10%-7 85%-8.5%<8
Bar #3	0.413	0.694	1.46	0.728	0.070	50%-7 50%-8

APPENDIX II

ELECTRON DIFFRACTION ANALYSIS OF HEAT TREATED MBMC STEEL

A sample of MBMC steel heated at 1000°F for 40 minutes was prepared as a metallurgical specimen but etched very deeply. A plastic tape, moistened with solvent was put on the etched surface and the solvent allowed to evaporate. When the tape was pulled off a black substance found in the metal came off also.

Carbon was evaporated in a bell jar under high vacuum and the plastic was coated with a thin amorphous film of carbon. The plastic was removed by dissolving in acetone and pieces of the carbon with the embedded black particles were picked up on 200 mesh copper screens.

This screen was put into the microscope and examined as a microscope specimen. When a good area was found, the electron diffraction pattern could be photographed by changing the lenses etc. in the 'scope.

The pattern was read and d_{hkl} spacings calculated. The spacings are listed in Table V with spacings of various compounds in separate columns.

The pattern was a very difficult one to read due to the orientation size of the particles giving the type of pattern shown in Figure 21. The arcs were measured using calipers and then measuring these distances on a good mm scale. This method does not give extremely accurate readings but is adequate.

The particles examined by electron diffraction contain graphite, cementite and epsilon iron carbide.

Table V

"d" SPACINGS FOR BAINITE SPECIMENS

MBMC Steel

I	Diam.	d_{hkl}	C (Graphite)		Fe_3C (Cementite)		EFe$_2$C (Epsilon)	
	1.56	3.33	100	3.37				
	2.075	2.50			5	2.54		
	2.198	2.36			65	2.38	40	2.38
S	2.30	2.26			25	2.26		
MS	2.36	2.20			25	2.20		
	2.39	2.17	2	2.13			60	2.16
S	2.49	2.08			60	2.10	100	2.08
					70	2.06		
	2.535	2.05	3	2.036	60	2.02		
					100	2.01		
S	2.649	1.961			55	1.97		
					30	1.87		
MS	2.83	1.835			40	1.85		
M	2.96	1.755	8	1.682	15	1.76		
					15	1.68		
M	3.20	1.623			7	1.61	60	1.60
S	3.32	1.564	2	1.541	20	1.58		
W	3.40	1.528						
MW	3.46	1.501						
MW	3.935	1.320					60	1.37
VW	4.29	1.240	2	1.232			60	1.24
S	4.31	1.205						
VW	4.37	1.189						
M	4.51	1.152					60	1.16
MW	4.65	1.117						
W	4.72	1.100						
	5.30	0.980						

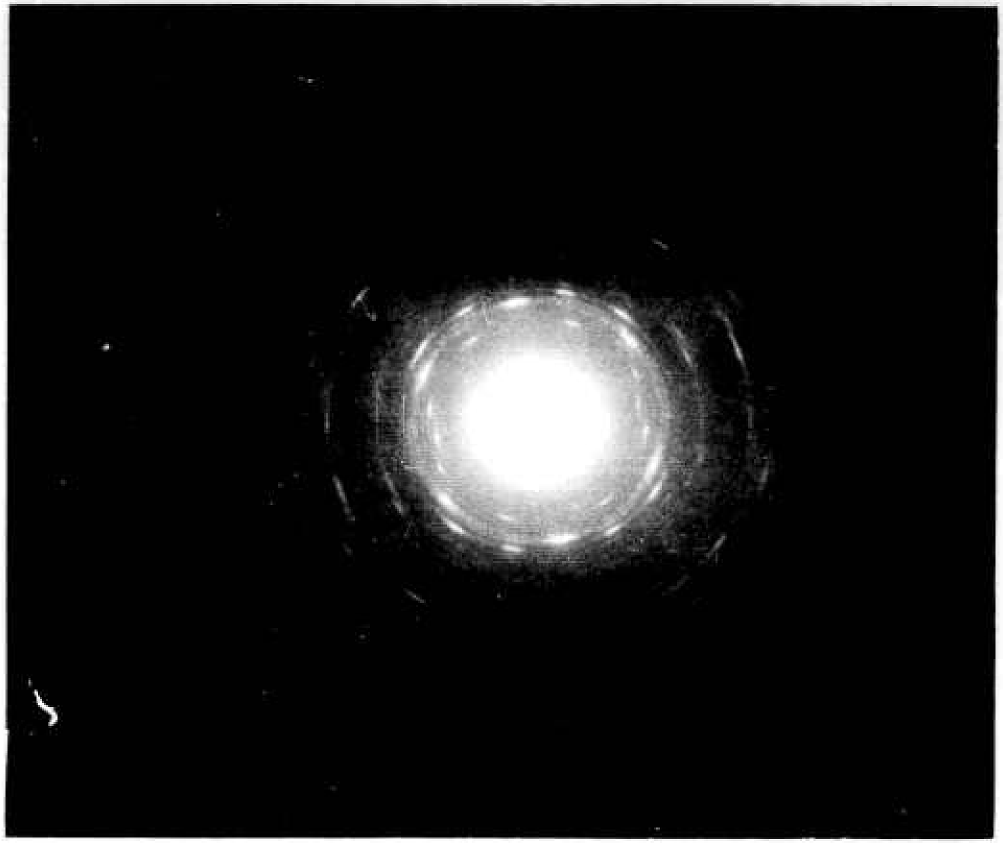


Figure 21

Photographic print of pattern produced by electron diffraction of particles removed from MBMC #1 steel. Specimen was isothermally transformed at 1000°F for 40 minutes.

The graphite pattern is a very difficult one to identify in this type of mixture. The pattern has only one strong line and this one is definitely in the unknown. The other lines are all very weak in comparison but they also appear to be present.

The pattern of cementite agrees very well with the unknown. In a few cases there are many lines in the pattern that do not appear in the unknown. Since these d spacings are very close (2.06-2.02-2.01-1.97, etc.) and only arcs were available in the pattern being read, some of these lines could very easily have been missed.

The pattern of epsilon iron carbon also checks very well.

The intensities marked on the left hand column do not mean too much in this sample identification since orientation will affect this greatly.

M. L. Pierotti

APPENDIX III

ANNEALING CYCLE

I. Summary

A uniform microstructure of spheroidized carbides in a ferritic matrix can be produced in MBMC #1 by holding at 1650°F for one hour, air cooling or oil quenching, and tempering at 1350°F-1400°F for 24-48 hours. Other heat treatments produce a non-uniform structure and/or require several steps which might be impractical.

III. Conclusions and Recommendations

a. A cycle of 1650°F one hour, air cool or oil quench, and then 1350°F for 24-48 hours will produce a uniform, spheroidized structure in MBMC #1.

b. The cycle given in (a) above using an oil quench after the 1650°F treatment and a time of 48 hours after 1350°F is recommended for producing the desired structure. Depending upon the homogeneity of the forgings to be spheroidized, a pretreatment to normalize may be required. This might be of the order of 1650°F for 2-3 hours followed by air cool.

c. An initially pearlitic structure will not produce as uniform a spheroidized structure as will a martensitic structure after heat treatment below A_1.

d. Complete elimination of lamellar carbides upon heat treatment of normalized structure either above or below A_1 and heat treatment of a quenches (oil or air) structure above A_1 is not possible for times less than 24 hours.

e. The A_1 temperature for MBMC #1 is between 1375°F and 1400°F.

f. Air cooling MBMC #1 is sufficiently fast to miss the nose of the isothermal transformation curve and produce bainite and martensite rather than pearlite.

III-A. Experimental Procedures and Equipment

Procedure: A 0.500" plate, 4" x 7" was normalized at 1650°F for one hour in a Hevi-Duty Type 56M muffle furnace using an atmosphere of dissociated ammonia enriched with propane to prevent decarburization. This plate was immersed in a bucket of lime, and after reaching room temperature was cut into pieces approximately 1/2" x 1/2" x 1/2", wrapped in tagged wires for identification and given the thermal treatments listed in Table II.

Equipment: All treatments at 1650°F and 1300°F were given in the Hevi-Duty furnace mentioned above, and treatments at 1500°F were given in a Hevi-Duty Type HD153012-CUM muffle furnace , both using the dissociated ammonia plus propane atmosphere. 1100°F treatments were given in a Hevi-Duty Type 161616A Temperite Muffle furnace, and 1350°F treatments in a Hevi-Duty Type MU 615 salt pot using Park Neutral salt No. K-3.

III-B. Results and Discussion

Metallographic examination and hardness tests were used in determining the spheroidization cycle. Since the as-received material was already spheroidized, it required normalizing. It was also desired to run the trial heat treatments on air cooled and oil quenched material. Results of hardness and metallographic examination are listed below.

Condition	Hardness	Microstructure
As received	97.4 R_B	Spheroidized annealed
Normalized	100.3 R_B	Fine pearlite surrounded by grain boundary ferrite. Grain size 7.
1650°F 1 hour, air-cooled	38.7 R_C	Bainite plus martensite
1650°F 1 hour, oil-quenched	56.8 R_C	Martensite

Five heat treating cycles were tried, and a discussion of the principals governing this choice can be found in "The Annealing of Steel" by P. Payson, a pamphlet printed by the Crucible Steel Company. Results of these cycle are contained in Table VI.

Prior to undertaking these heat treatments it was necessary to determine, within 25°F, the A_1 temperature of the MBMC #1. This was accomplished by heating samples at 1350°F, 1375°F, 1400°F, 1425°F, and 1450°F for one hour and oil quenching. Metallographic examination of these samples revealed martensite only in the samples held at 1400 and above. Traces of a transformation product were found in the samples held at 1375°F and on this basis, the highest feasible temperature for spheroidization was considered to be 1350°F.

Figures 22-25 illustrate some of the microstructures found as a result of the treatments tried. Comparison shows that the 1350°F treatment

Table VI

EFFECT OF VARIOUS HEAT TREATMENTS ON SPHEROIDIZING OF MBMC #1

Heat Treatment	Normalized		Air Cooled		Oil Quench	
(After normalization)	R_B	Microstructure	R_B	Microstructure	R_B	Microstructure
1350°F 4 hours	93.4	Fine pearlite, ferrite in grain boundaries	98.9	Small carbides in temp. martensite	99.3	Small carbides in temp. martensite
1350°F 8 hours	94.5	Spheroidized slightly lamellae still present	96.8	14 hours Carbides in ferrite matrix	98.4	16 hours-Carbides in ferrite matrix
1350°F 24 hours	90.1	Spherodized further lamellae still present See Figure I	96.7	Same as 14 hours coarser carbides	95.5 94.6	24 hrs. - Same as 16 hrs. See Figure 4 48 hrs. - Coarser Carbides
1500°F 2 hrs. → 1300°F 4 hrs	98.6	Fine pearlite, little ferrite	99.0	Fine pearlite, little ferrite	99.8	Fine pearlite, little ferrite
1500°F 2 hrs. → 1300°F 8 hrs	99.2	Spherodized slightly Lamallae still present See Figure 2	98.3	Spherodized slightly Lamellae still present	98.6	Spherodized slightly Lamellae still present
1500°F 2 hrs. → 1300°F 24 hrs	95.9	Spherodized further Lamellae still present	-	-	95.4	Spherodized further
1350°F 4 hrs. → 1500°F 2 hrs. → 1300°F 4 hrs.	96.5	Spherodized slightly Lamellae still present	99.2	Spherodized slightly Lamellae still present	98.3	Spherodized slightly Lamellae still present
8 hrs.	96.9	Slightly coarser carbides Figure 3	95.4	Slightly coarser carbides	95.5	Slightly coarser carbides
24 hrs.	94.7	Same as 8 hour 18 hours	97.5	Slightly coarser carbides	95.4	Slightly coarser carbides
1650°F 1 hr. → 1100°F 4 hrs	105	Fine pearlite, ferrite in grain boundaries				
8 hrs	106	Same as 4 hours.				
24 hrs	106	Same as 8 hours.				
1650°F 1 hr. → 1350°F 4 hrs.	100.5	Fine pearlite little ferrite				
1650°F 1 hr. → 1350°F 8 hrs.	99.3	Spherodized slightly Lamellae still present				
1650°F 1 hr. → 1350°F 23 hrs	93.4	Same as 8 hrs. coarser carbides				

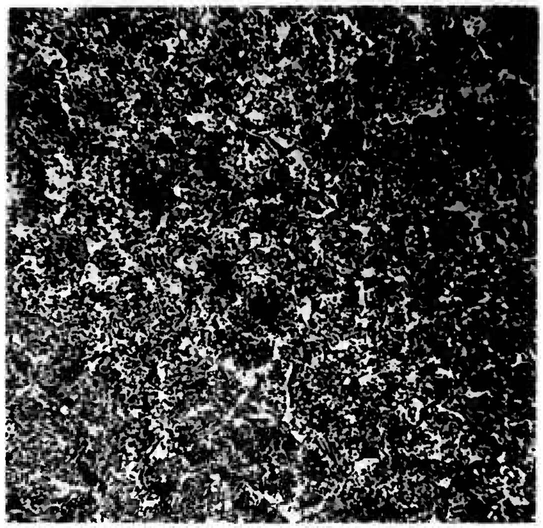


Figure 22 Neg. #595
Normalized (1650°F 1 hour, cooled in lime). 1350°F 24 hours

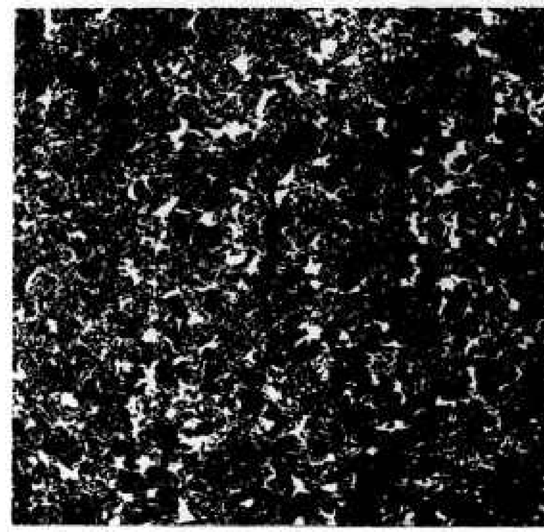


Figure 23 Neg. #594
Normalized, 1500°F 2 hours. directly into furnace at 1300°F. 8 hours

Figure 24 Neg. #596
Normalized, 1350°F 4 hours, directly into furnace at 1500°F 2 hours, directly into furnace at 1300°F, 8 hours

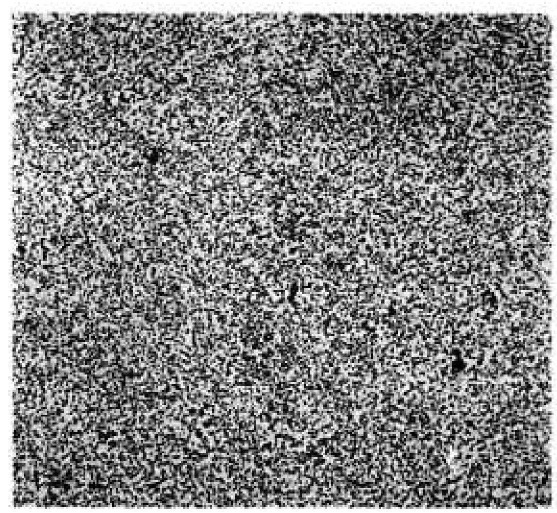


Figure 25 Neg. #597
Normalized 1650°F 1 hours, oil quenched 1350°F 48 hours

after oil quenching produces the most uniform structure.

As mentioned earlier, the samples used in this investigation were already spheroidized, and in addition were small in size and fairly homogeneous. This might not be the case in the forgings under consideration, depending upon the extent and nature of the forging with respect to the starting billet. If not homogeneous, the recommended treatment of 1650°F for one hour may not be sufficient to give a homogeneous austenite and subsequent uniform carbide distribution. Should the forgings not be homogeneous, it is suggested that they be held at 1650°F for 2-3 hours and air cooled prior to the spheroidization treatment.

J. A. Horwath

Data pertaining to this Appendix may be found in notebook #65, p. 164-175.

APPENDIX IV

PRIMARY MECHANICAL WORKINGS:

ROLLING PROCEDURE OF MBMC #1

The ingots are heated to a temperature of between 2200° and 2250°F and held long enough to insure uniform temperature. The furnace atmosphere should be oxidizing and the scale formed removed prior to rolling. The ingot is then rolled to 0.500" thick in about 6 to 9 passes depending on original thickness.

The cogged plate is then reheated in a reducing atmosphere to between 1575° and 1650°F. It usually takes 20 to 30 minutes to heat these plates from room temperature; a walking beam furnace works very well. It takes from 3 to 5 heatings with about 8 passes per heating to reduce the material to 0.125" thick.

If 4 or more heatings are required, the material should be pickled before the last heating to remove the scale building up on the material. This is done prior to the last heating for any thickness of material.

For light gauge sheets any conditioning necessary is done on the 0.125" material before finish rolling. The material is then heated in a reducing atmosphere to between 1550° and 1600°F. Typical reductions are as follows:

Heatings	1st	2nd	3rd	4th
Passes made	5	6	6	6
Gauge obtained in inches	0.090	0.080	0.072	0.063

Andy Matias, Metallurgist
Ingersoll-Steel Division
Borg-Warner Corporation

THE UNIVERSITY OF MICHIGAN

COLLEGE OF ENGINEERING
Department of Mechanical Engineering

Final Report

COMPARATIVE MACHINABILITY OF MBMC-1 AND AISI 4130 STEEL

L. V. Colwell K. N. Soderlund

UMRI Project 03690

under contract with:

INGERSOLL DIVISION
BORG-WARNER CORPORATION
KALAMAZOO, MICHIGAN

administered by:

THE UNIVERSITY OF MICHIGAN RESEARCH INSTITUTE ANN ARBOR

October 1960

TABLE OF CONTENTS

LIST OF TABLES

LIST OF FIGURES

ABSTRACT

Both tool life and cutting force studies were made on the two work materials in both the as-forged and heat-treated or hardened condition. In general, the MBMC-1 steel was more difficult to cut. However, it was feasible to cut it satisfactorily at all test conditions except for turning the hardened material with high-speed steel tools. Quantitative data provide a general guide for shop practice.

I. STATEMENT OF THE PROBLEM

The principal objective was to determine whether the newly developed, high-strength steel MBMC-1 exhibited any unique characteristics which would make it particularly difficult to machine. For this purpose it is necessary to explore tool wear behavior as well as force and energy requirements at a number of different conditions such as turning, drilling, and milling. Since many of the quantitative data cannot be interpreted on an absolute basis, it was decided to obtain comparable information for AISI 4130 steel which is functionally similar to MBMC-1.

II. GENERAL CONCLUSIONS

It was determined that, although the MBMC-1 steel does exhibit some unique machining properties, it can be machined with no great difficulty, particularly in the as-forged condition. When it is hardened it cannot be cut readily with high-speed steel tools but it can be cut quite satisfactorily with both sintered carbide and ceramic tools.

III. DEFINITION OF MACHINABILITY

To avoid misconception, machinability should be defined in specific terms. Two criteria, used to compare the relative machinability of the test materials, have been selected for this investigation. One is the magnitude of cutting force required to machine the materials, using several types of operations and at varying sizes of cut. Low cutting forces are more desirable than high cutting forces and demonstrate better relative machinability from this viewpoint.

Relative machinability may also be based on tool life. A material that exhibits a longer tool life than another is considered to have better machinability from this viewpoint.

Unfortunately, a high machinability rating based upon one criterion does not always indicate corresponding advantages with respect to other relevant machining criteria. Materials that cut with lower forces do not necessarily exhibit better tool life.

Force measurements obtained with sharp cutting tools are not indicative of the nature of tool wear that will result if the cut is continued. A material may show a low force at the start of cut but exhibit a poorer tool life than a material which shows a higher starting force; this could be accounted for by different wear rates not reflected in the forces.

The term "better machinability," if properly applied, should be considered only in terms of the criteria of interest. References such as surface finish and dimensional stability have not been studied <u>per se</u> in this work and specific comments pertaining to these items should not be considered conclusive.

A fundamental study of the type conducted is intended to bring out any pronounced differences that might exist, but the conclusions from any investigation are limited to the actual testing conditions. If, for example, sintered carbide tools are used in a study of tool life, the findings of relative machinability could be of a different order from those obtained with high-speed steel tools.

IV. THE MATERIALS INVESTIGATED

TABLE I

HARDNESS AND SIZE OF TEST SPECIMENS

Work Piece Identification	Size of Work Piece, in.	Hardness		
		Brinell** Reading (mm)	Brinell Reading (mm)*	Average Hardness Number
4130-1-H (heat-treated)	5 OD x 24	2.8	2.85	461
4130-2-H	5 OD x 24	2.6	2.85	461
4130-3-AF (as-forged)	5 OD x 24	4.6	4.7	163
4130-4-AF	5 OD x 24	4.55	4.45	183
4130-5	1 x 2 x 36	4.5	4.65	167
B.W.-1-H (heat-treated) MBMC-1	5 OD x 24	2.95	3.05	401
B.W.-2-H	5 OD x 24	2.7	2.85	461
B.W.-3-AF (as-forged)	5 OD x 24	4.3	4.4	187
B.W.-4-AF	5 OD x 24	4.3	4.0	229
B.W.-5	1 x 2 x 36	4.3	4.4	187

*Obtained at University of Michigan with a 10-mm steel ball and 3000-kg load.
**Provided by sponsor.

V. TEST PROCEDURES AND RESULTS

A. TOOL LIFE WITH HIGH-SPEED STEEL TOOLS

Work specimens used were in the form of cylinders, with an approximate OD of 5-1/4 in. and 24 in. long.

Each test consisted of a lathe turning cut at a constant size of cut and cutting velocity until complete tool breakdown occurred. The total elapsed cutting time, from starting point to failure, constitutes the tool life for those operating conditions. The cutting speed was changed for each tool-life test, thus providing a relationship between velocity and tool life.

Each material was subjected to such tests until enough individual points were obtained to define both level and slope of the velocity, tool-life line on logarithmic coordinates.

Upon the conclusion of each individual test and prior to the subsequent test, a cleanup cut was taken to avoid any possible effects of work-hardening from the previous tool failure.

Test Conditions and Equipment.—Two Monarch engine lathes equipped with variable-speed drives were used for all tests in this series. All machining was done dry, that is, without a cutting fluid. Standard 3/8-in.-square "Mo-Max," high-speed steel tool bits were ground to the signature, 0, 8, 6,6,6, 15, 1/32. The roughing cuts on the as-forged materials were made at a feed rate of 0.015 ipr and a depth of 0.100 in. For the finishing cuts on the same materials the feed was 0.0052 ipr and the depth was 0.020 in. In cutting the 4130 heat-treated material, the feed rate was 0.0075 ipr and the depth was 0.050 in. Attempts at cutting the heat-treated MBMC-1, resulted in an extremely poor cutting action and no consistent data could be obtained at any size of cut.

Test Results.—Figures 1 through 3 contain test results plotted on log-log coordinates. The trends are shown as straight lines which may be described by equations of the general form, $VT^n = C$, where:

V = cutting velocity (at OD) in fpm,
T = tool life in minutes,
C = a proportionality constant, and
n = slope of the tool-life line.

In all cases, the lines represent average performance but dispersion of points around the individual lines is not pronounced.

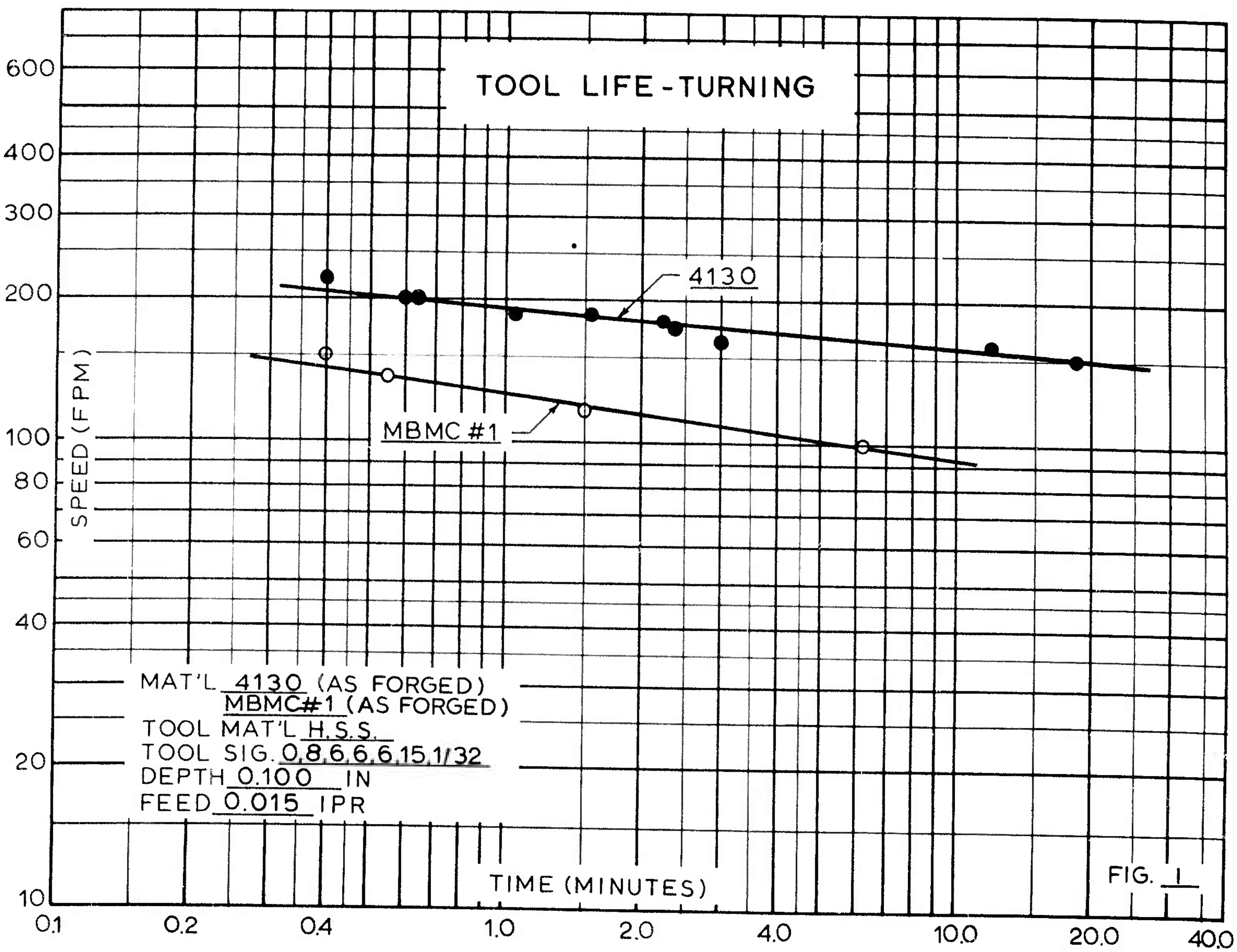


Fig. 1. Tool-life: Rough turning with high-speed steel (as-forged).

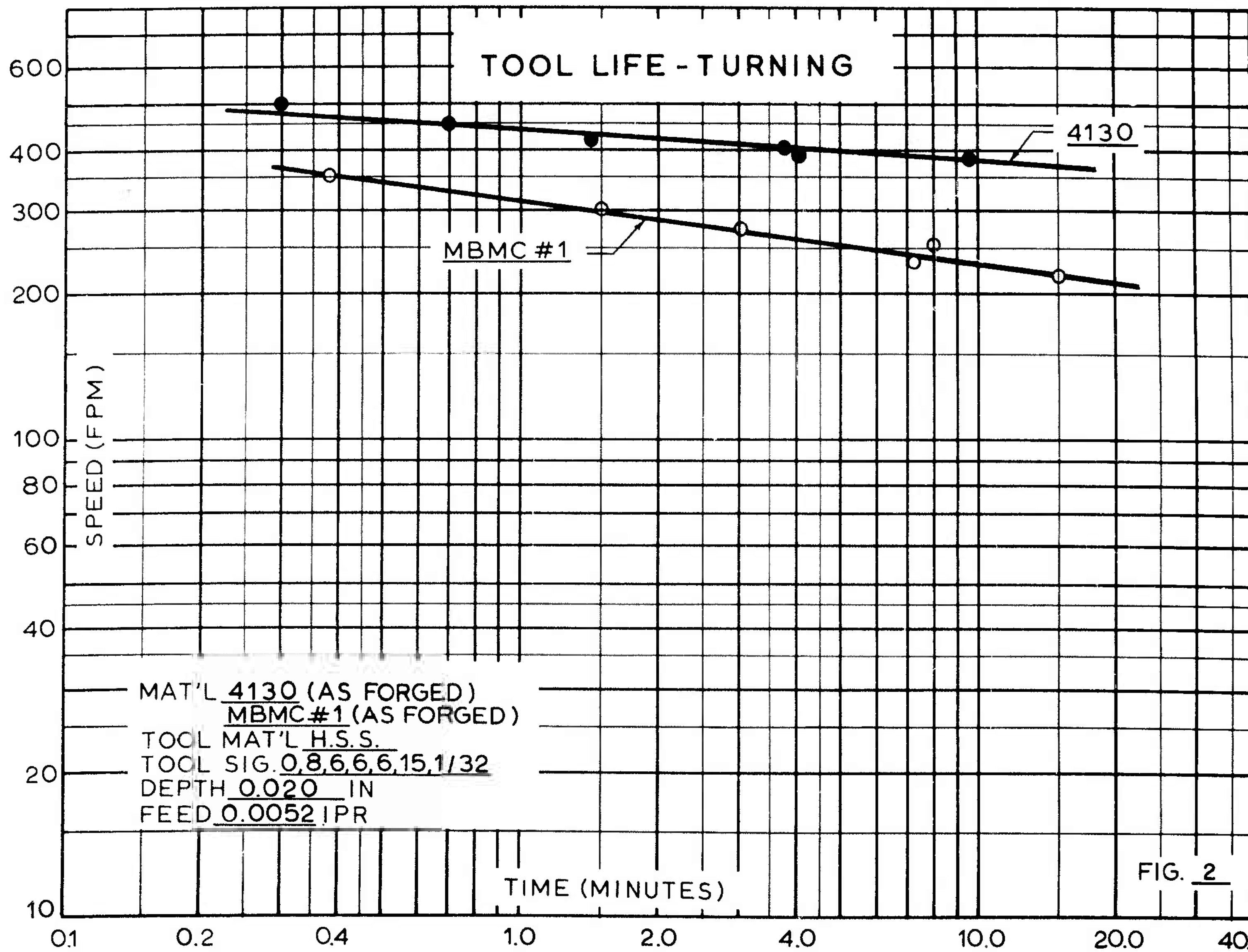


Fig. 2. Tool-life: Finish turning with high-speed steel (as-forged).

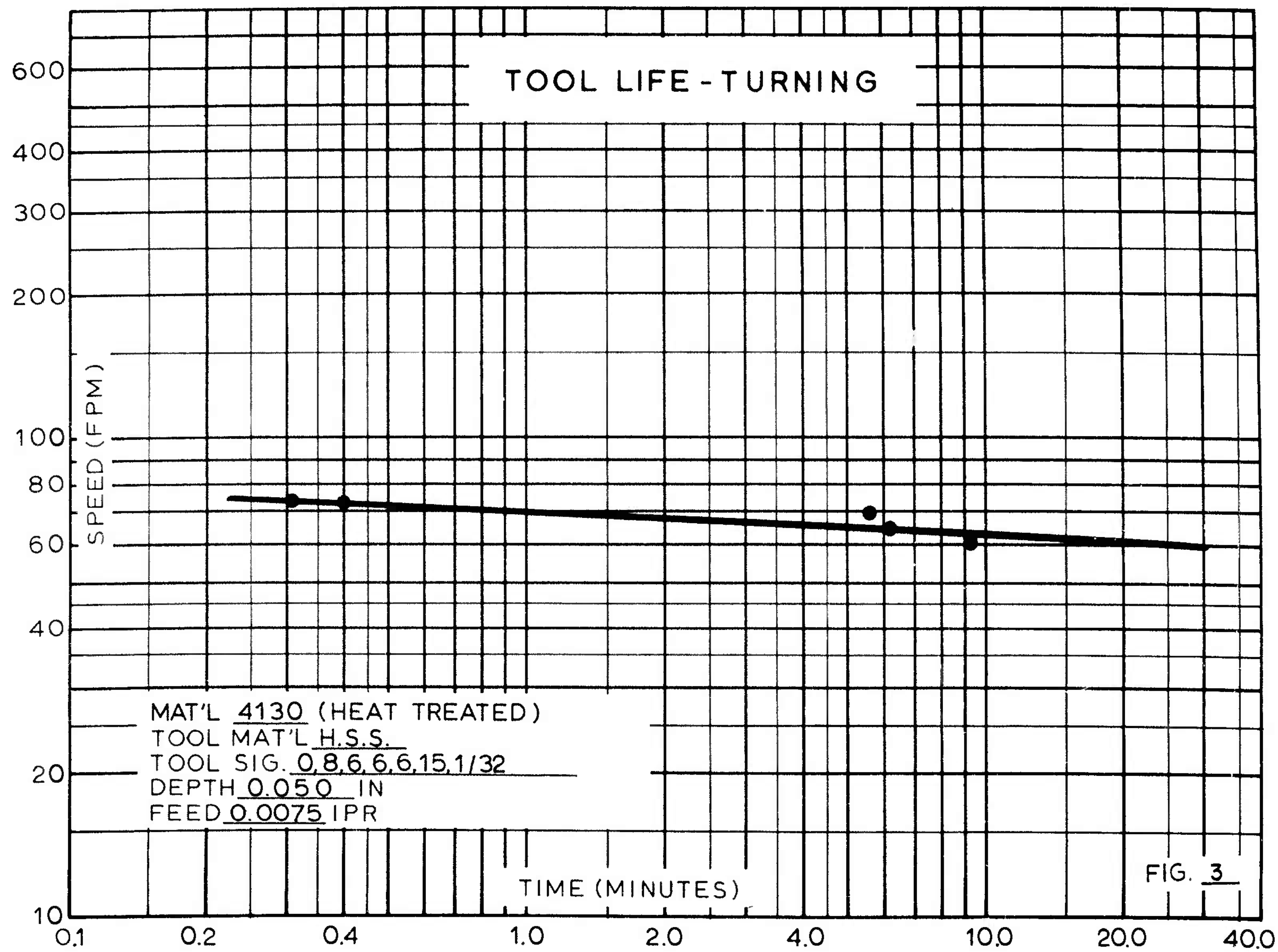


Fig. 3. Tool-life: Finish turning with high-speed steel (heat-treated 4130).

Equations for the resulting tool-life lines are summarized in Tables II and III.

TABLE II

TOOL LIFE WITH HSS TOOLS AND
AS-FORGED WORK MATERIALS

Material	Equations	
	Roughing Cuts*	Finishing Cuts**
4130	$VT^{0.078} = 192$	$VT^{0.0550} = 440$
MBMC-1	$VT^{0.12} = 125$	$VT^{0.12} = 315$

*Depth = 0.100 in.; Feed = 0.015 ipr.
**Depth = 0.020 in.; Feed = 0.0052 ipr.

TABLE III

TOOL LIFE WITH HSS TOOLS AND
HEAT-TREATED WORK MATERIALS

Material	Equations	
	Roughing Cuts*	Finishing Cuts
4130	$VT^{0.0682} = 71.3$	**-----
MBMC-1	**-----	**-----

*Depth = 0.050 in.; Feed = 0.0075 ipr.
**Cutting behavior at these conditions proved to be impractical and the erratic quantitative results are irrelevant.

Conclusions

(1) Both AISI 4130 and MBMC-1 demonstrate orderly behavior in the as-forged condition when cut at both roughing and finishing conditions.

(2) It is impractical to try to remove substantial amounts of MBMC-1 by cutting it in the heat-treated condition with high-speed steel tools.

(3) The steeper slopes of the tool-life lines for the MBMC-1 indicate that tool wear is more abrasive in nature than when cutting AISI 4130 steel.

(4) Cutting speeds for MBMC-1 should be approximately 40% lower than those found to be appropriate for AISI 4130 steel.

B. TOOL LIFE WITH CARBIDE TOOLS

Procedure.—The same specimens used for the HSS tool study in (A) were used for this series of tests. Test procedure consisted of operating a cutting tool at a given velocity, feed rate, and depth of cut. The carbide insert was removed from the toolholder at predetermined time intervals and the flank and rake face were examined under a toolmaker's microscope. The wear on the flank of the cutting tool was measured and recorded. This procedure was continued until a typical carbide wear pattern was established.

Test Conditions.—The machines were the same as those reported in (A). The tools were 1/2-in.-square, mechanically held blanks. The Kennametal toolholders provided a tool signature of -5, -5, 5, 5, 15, 15, 1/32. The sizes of cut for both roughing and finishing of both the as-forged and the heat-treated materials were the same as used for machining the as-forged material with high-speed steel tools. Unlike high-speed steel tools, carbides do not fail abruptly. Instead, the wear is gradual and point of failure is arbitrarily defined by a limiting amount of wear. Consequently, the tools were examined at regular intervals and the wear measured.

Test Results.—The results of tool wear measurements are shown plotted against elapsed cutting time in Figs. 4 to 13 inclusive. These represent a significant range of cutting conditions and carbide grades with both work materials.

Figures 4, 5, and 6 show the behavior for the as-forged materials. Three different grades of carbide were tried with a roughing cut on the MBMC-1. The results in Fig. 4 show that Carboloy-350 was best when compared to grades 370 and 883. This indicates that MBMC-1 is not particularly abrasive in the cutting behavior despite its silicon content.

Figure 5 gives a comparison of MBMC-1 with AISI 4130 for roughing cuts and Fig. 6 gives a similar comparison for finishing cuts. The results indicate that both materials can be cut at substantially the same speeds and sizes of cut while in the as-forged condition.

Test results for the heat-treated condition are summarized in Figs. 7 to 13 inclusive, Figs. 7 to 10 for roughing cuts and Figs. 11 to 13 for finishing cuts. Once more Carboloy-350 was the best grade of carbide as indicated in Figs. 8 and 9. Combining this experience with the variable speed tests of Fig. 7 led to the comparison plotted in Fig. 10 which shows that cutting speed for

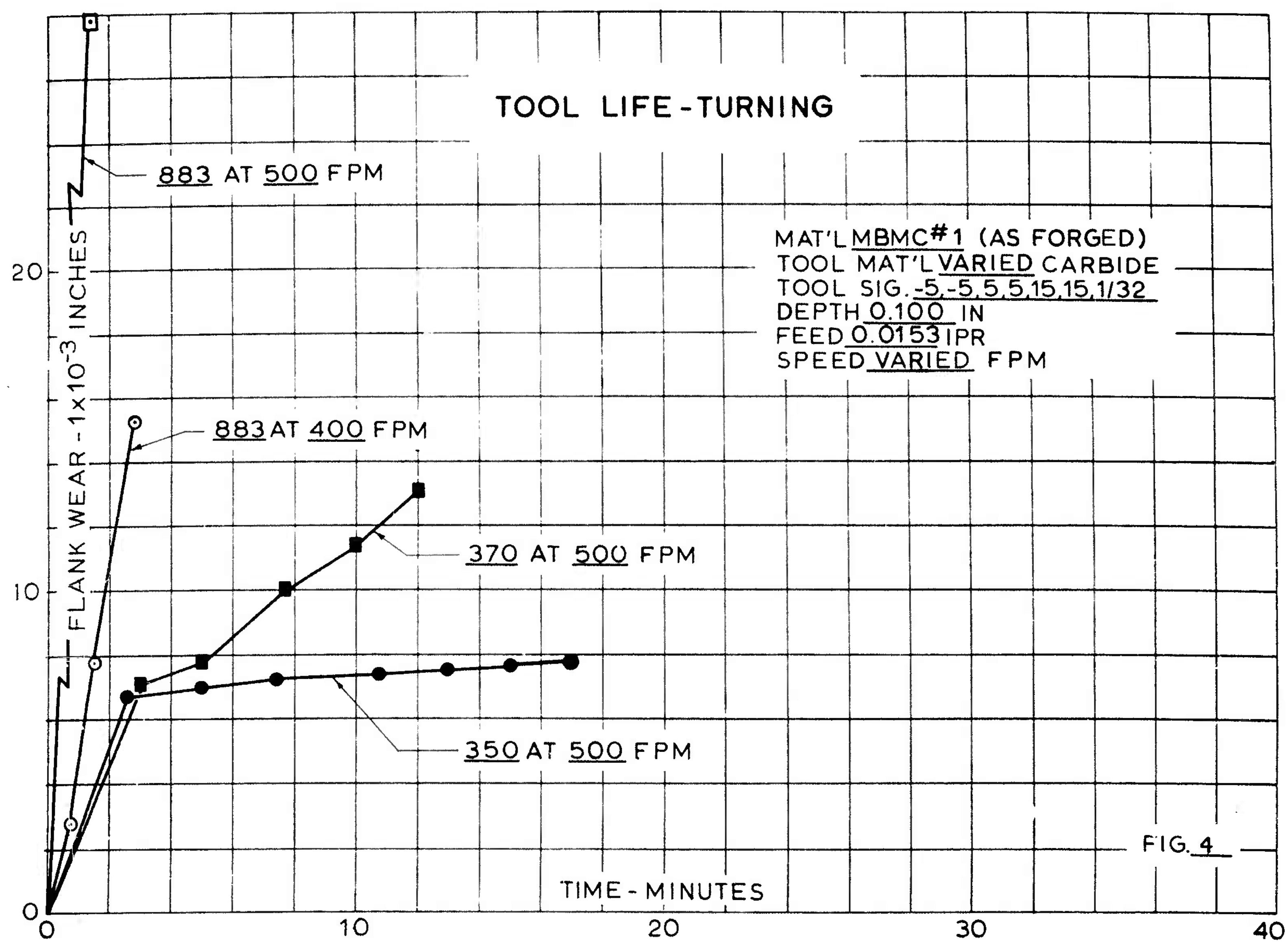


Fig. 4. Tool-life carbides.

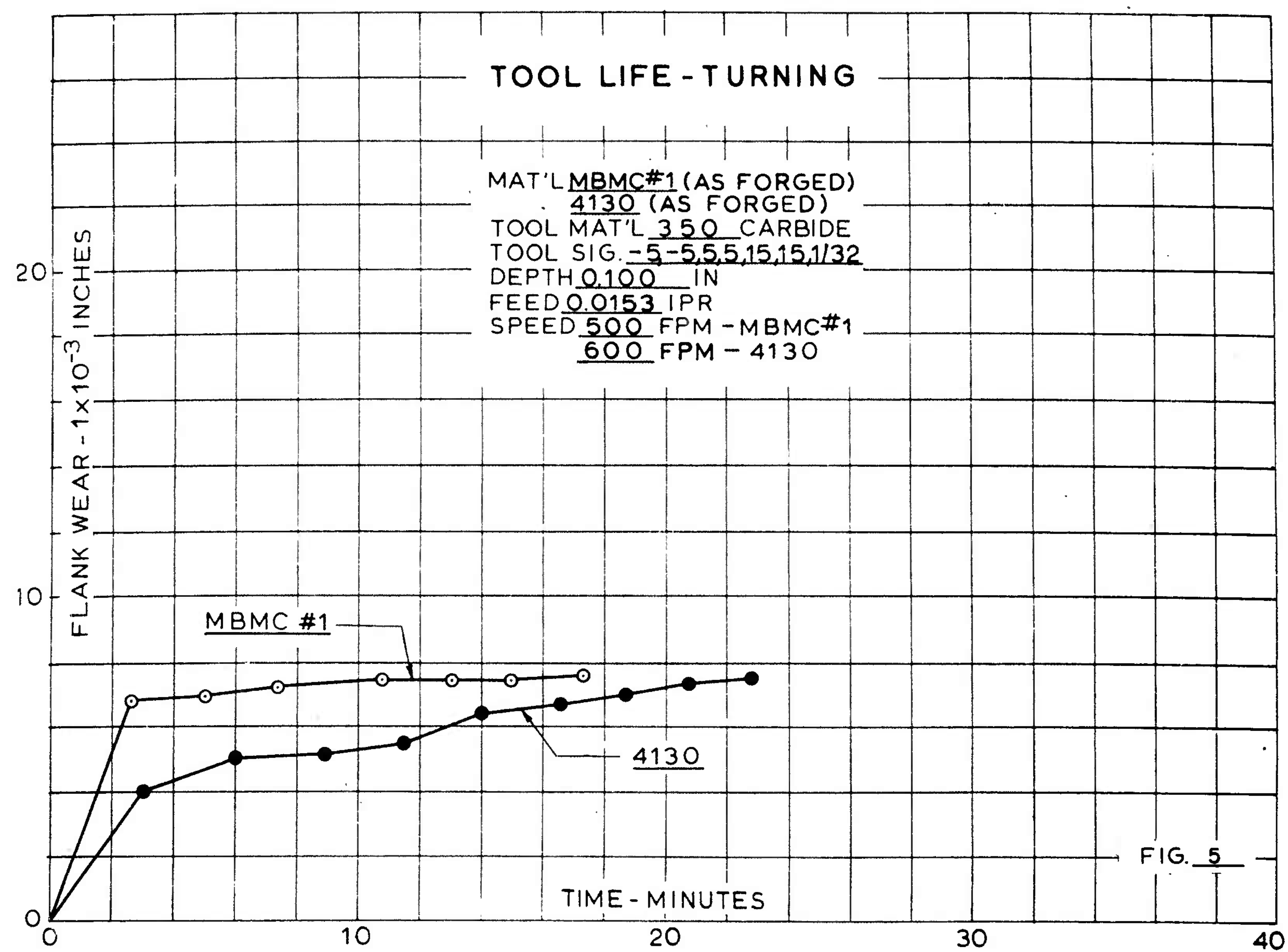


Fig. 5. Tool-life carbides.

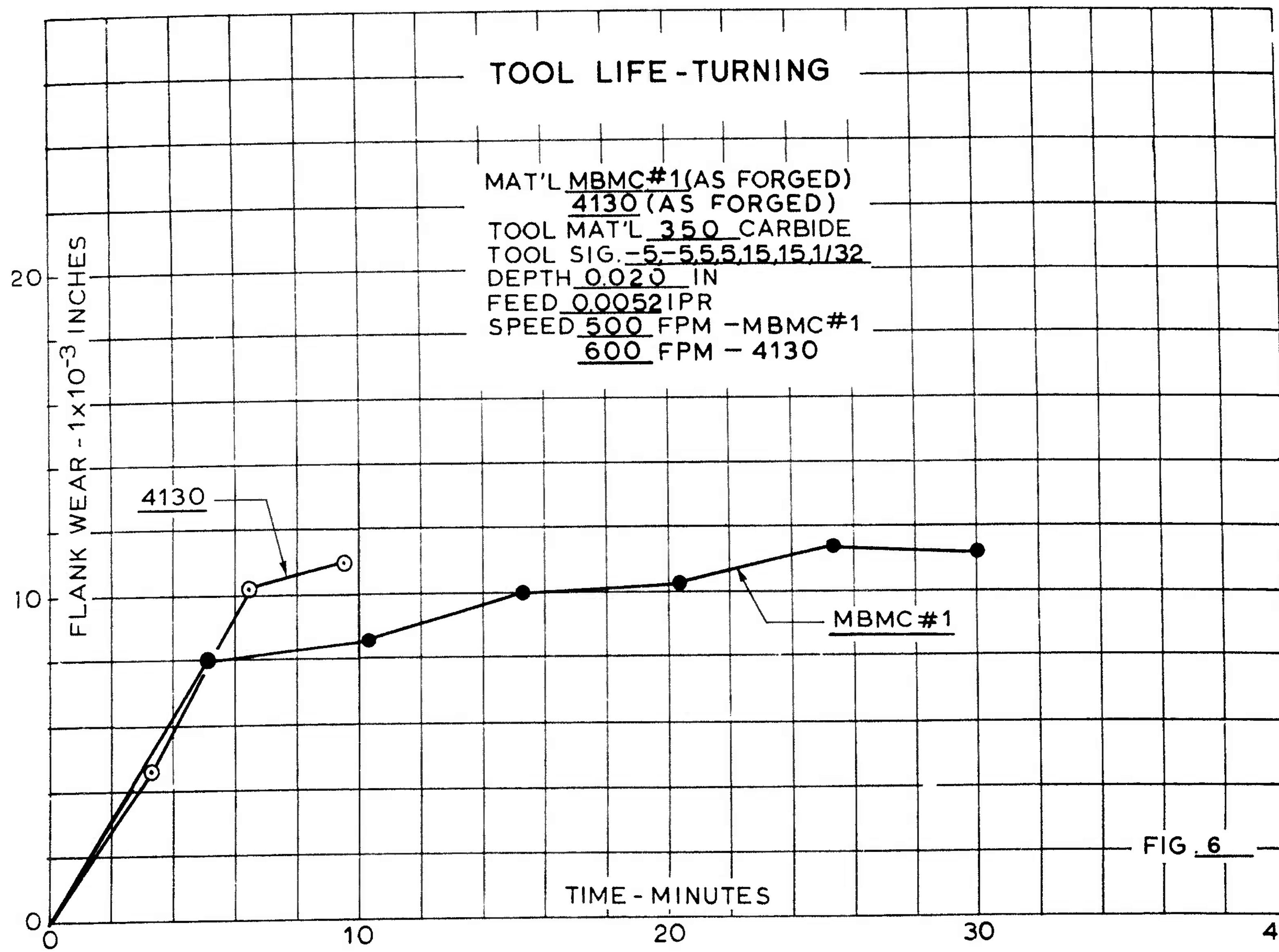


Fig. 6. Tool-life carbides.

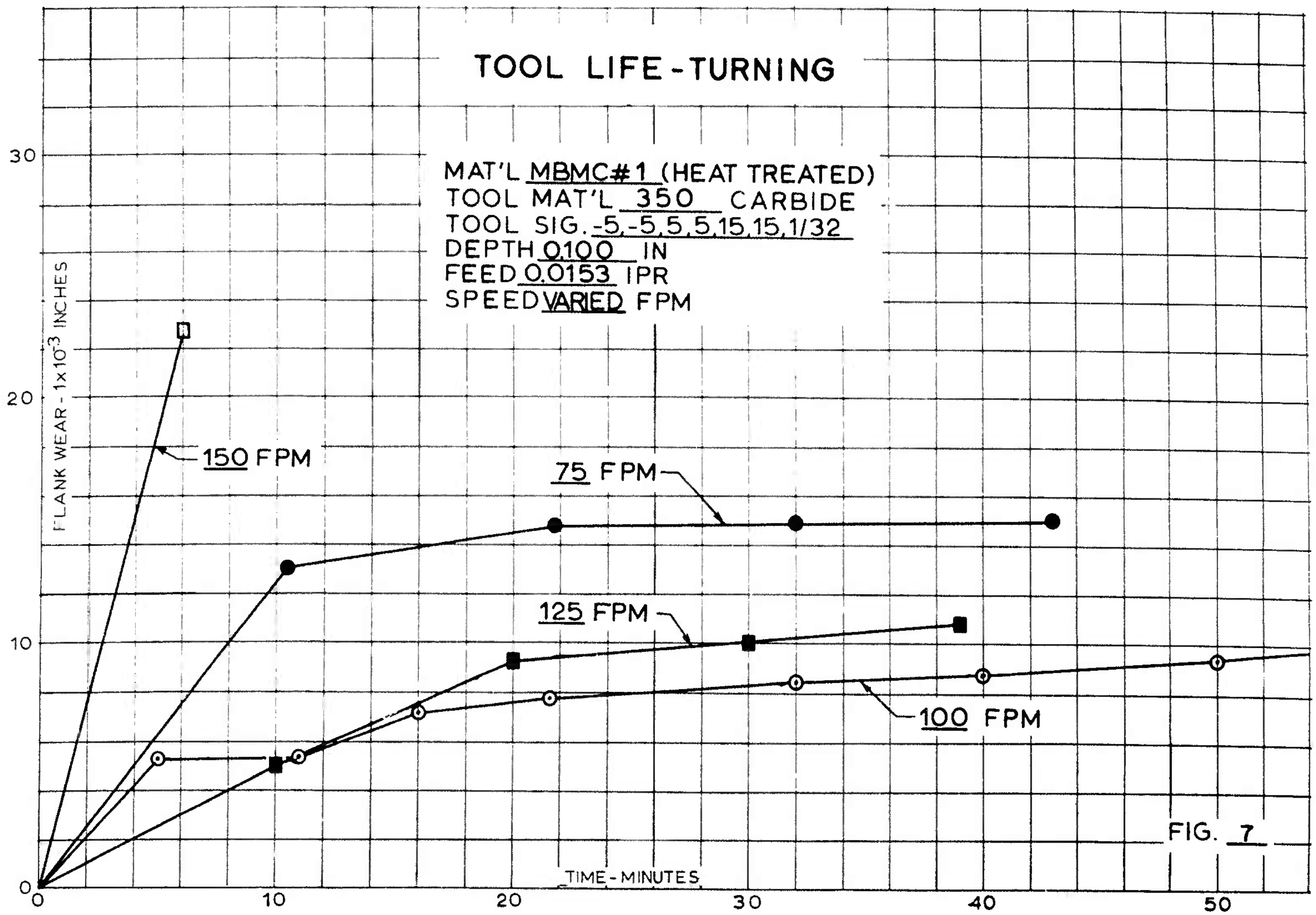


Fig. 7. Tool-life carbides.

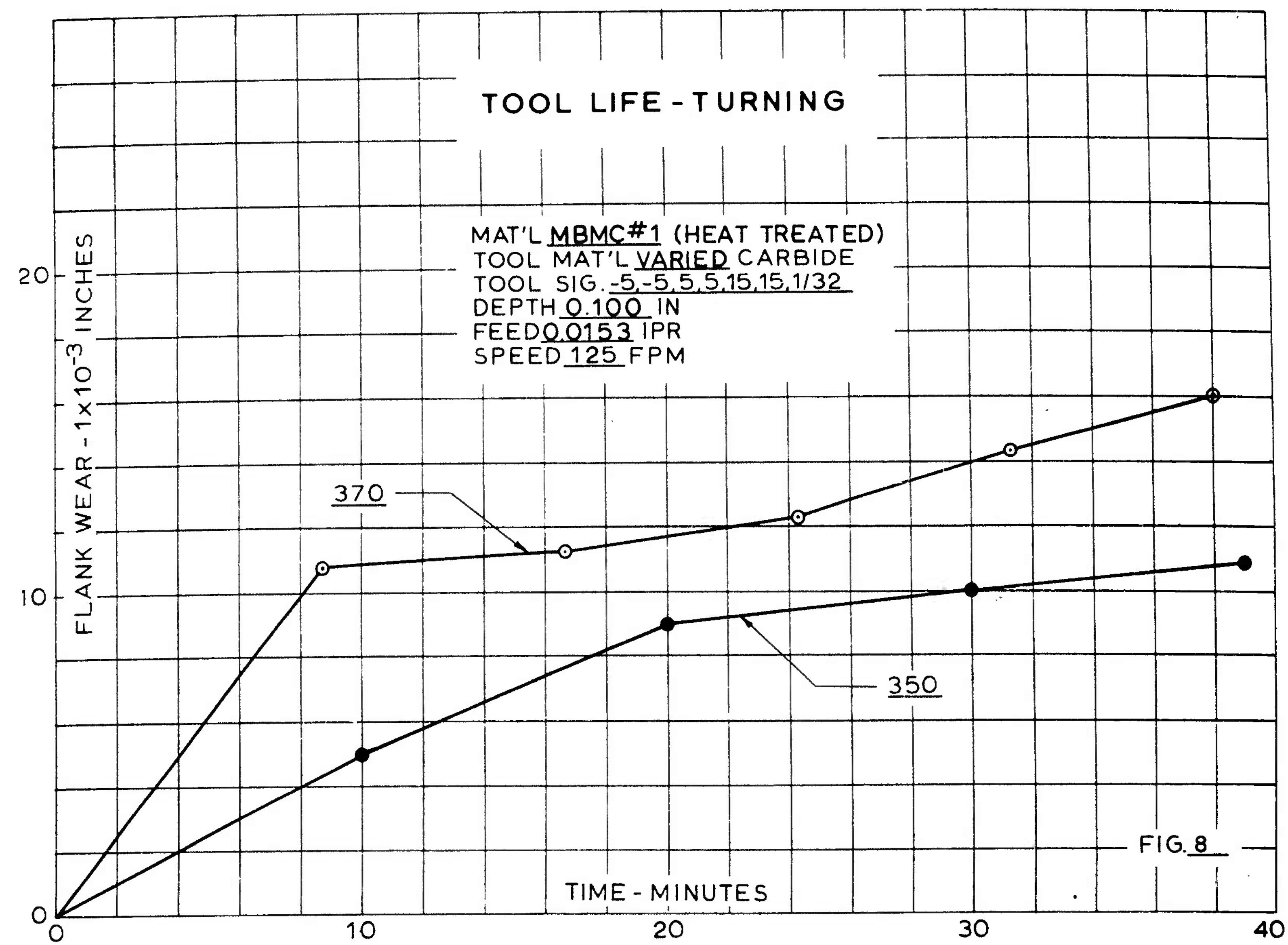


Fig. 8. Tool-life carbides.

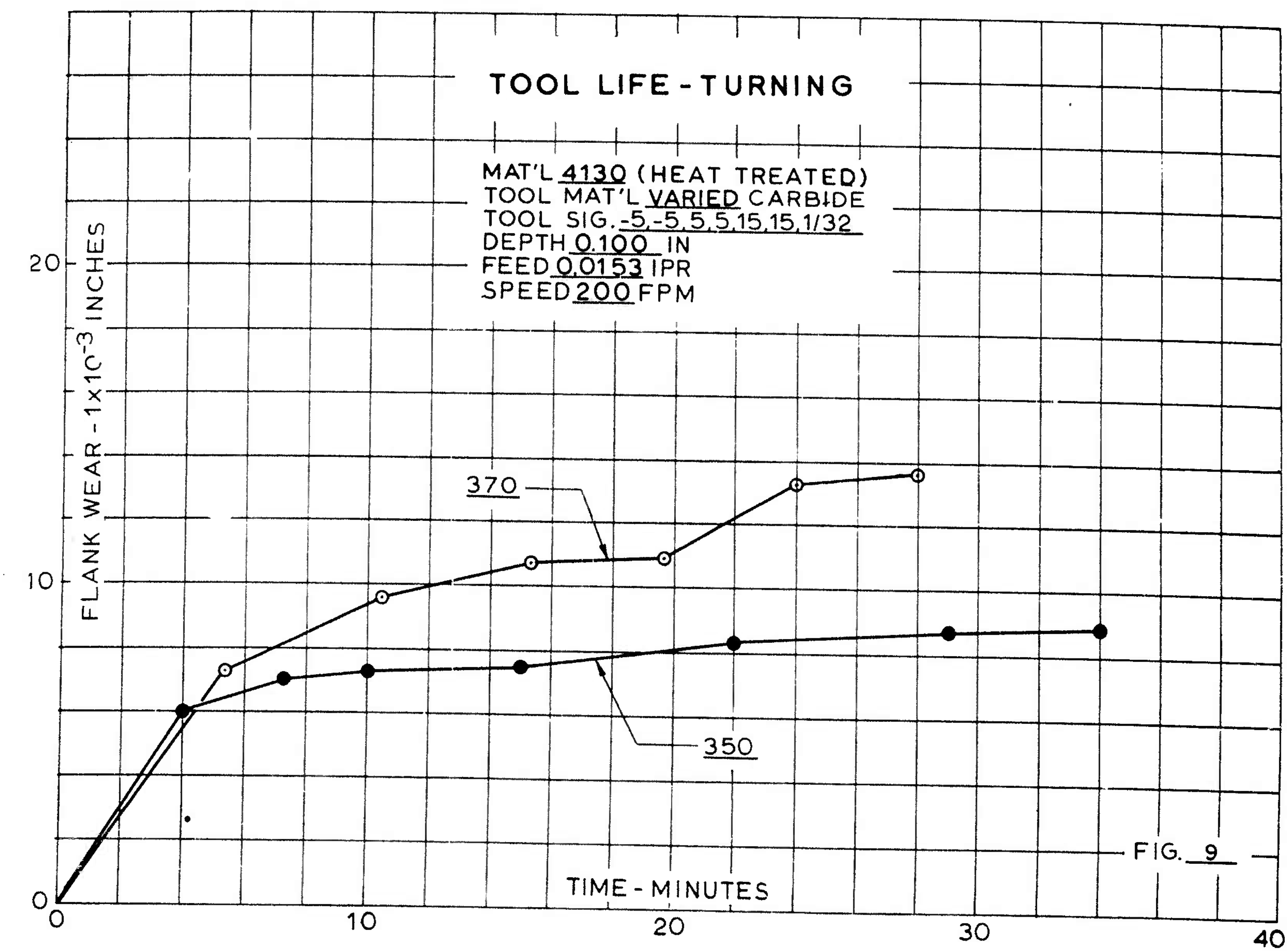


Fig. 9. Tool-life carbides.

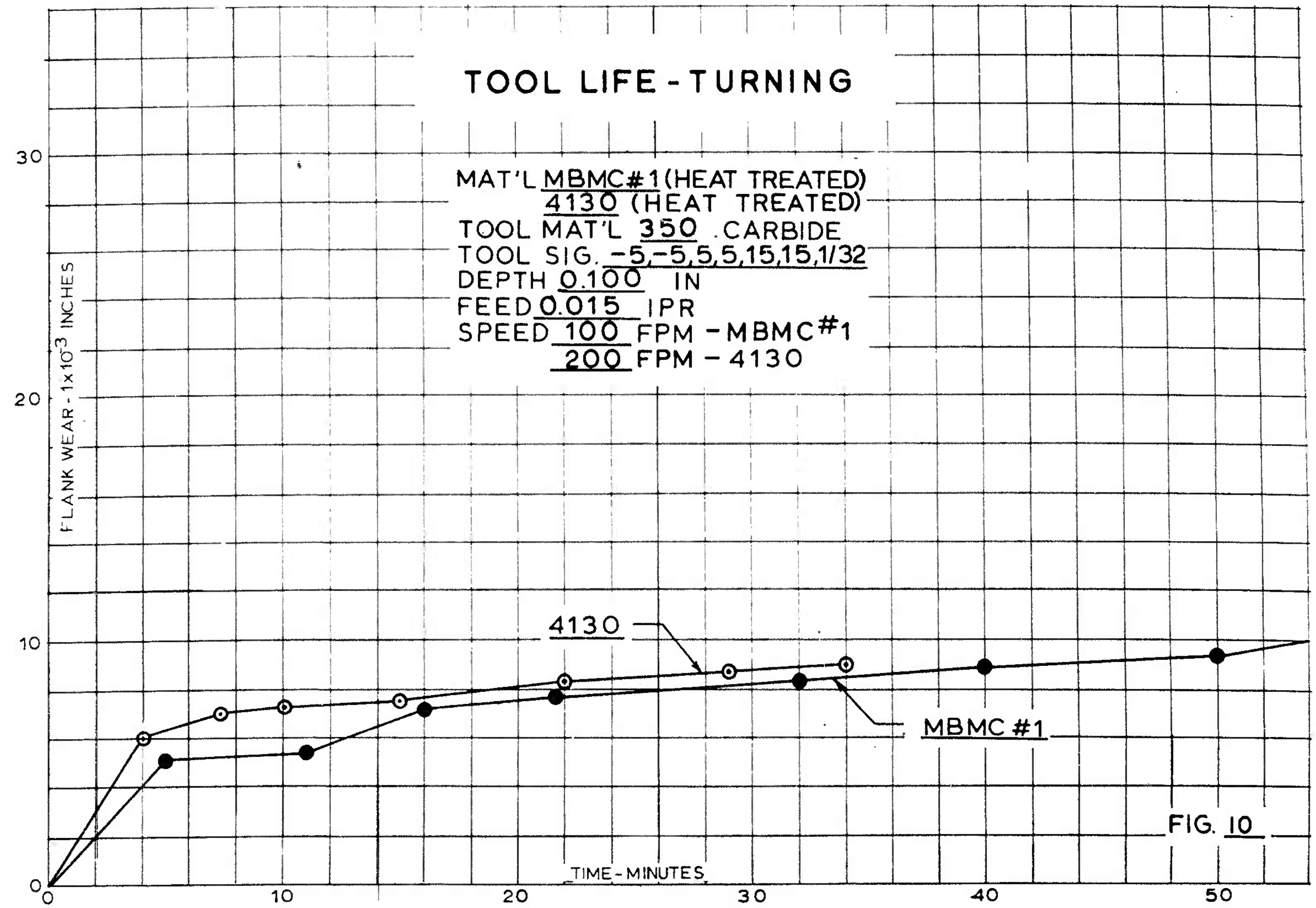


Fig. 10. Tool-life carbides.

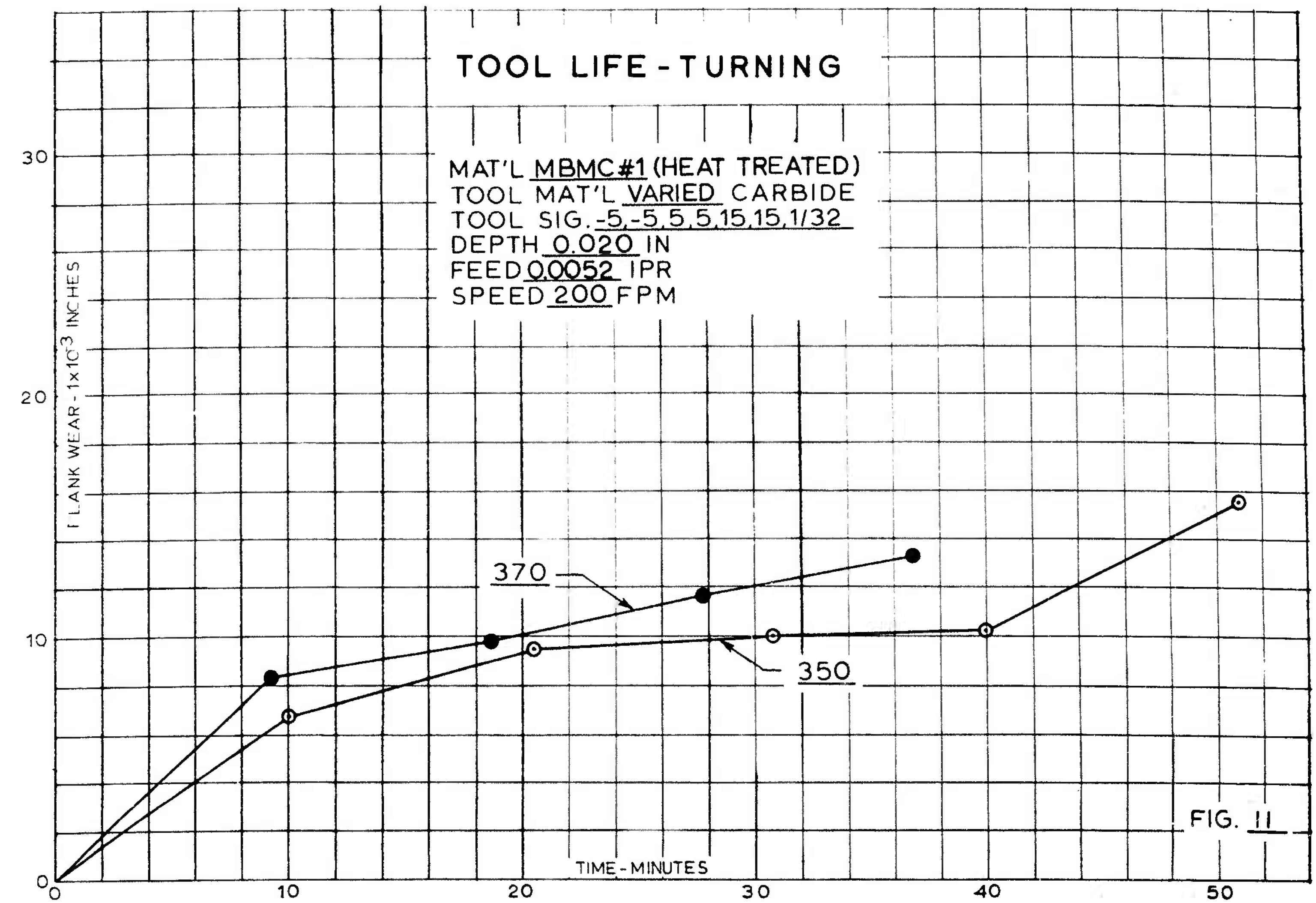


Fig. 11. Tool-life carbides.

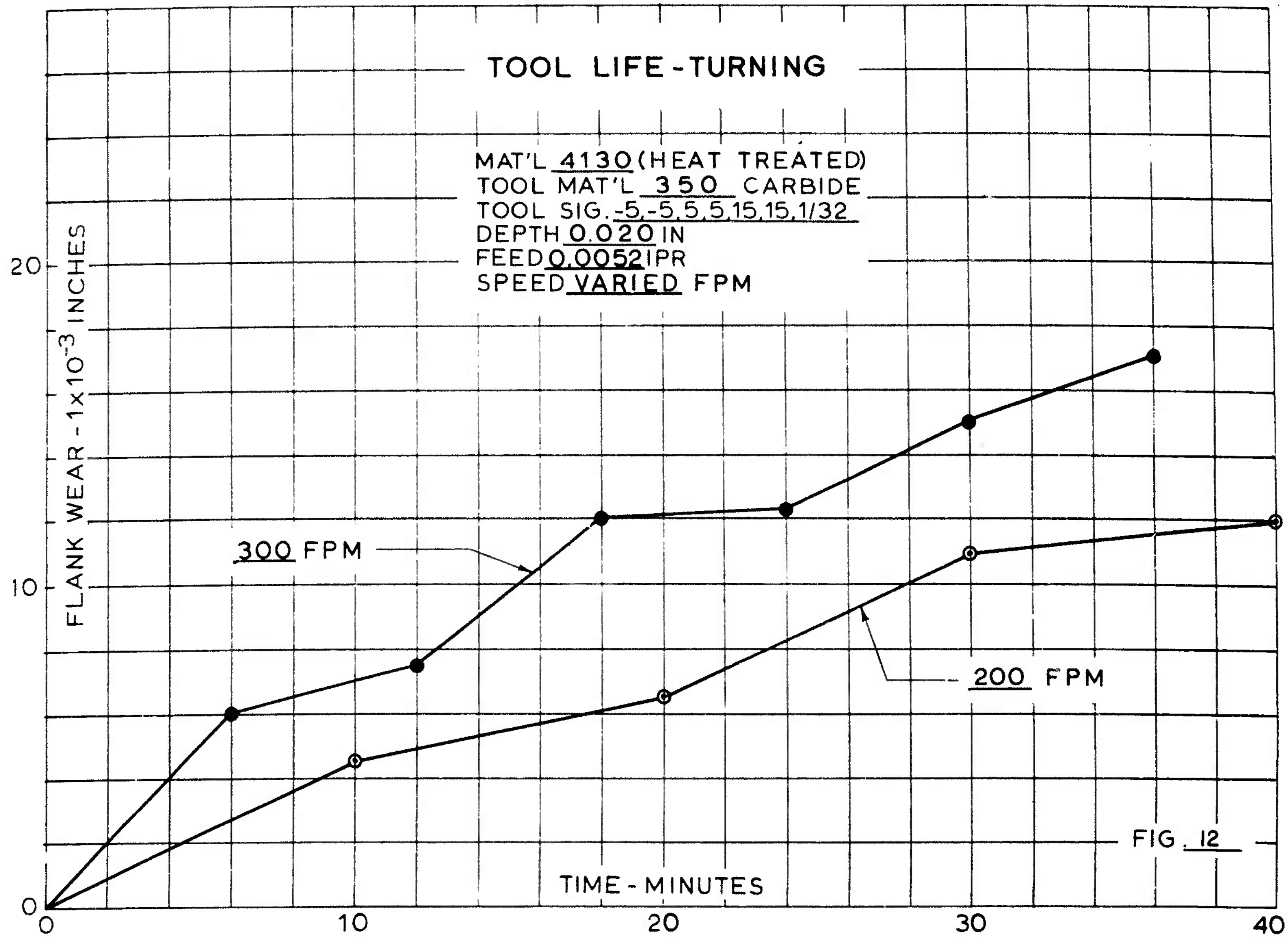


Fig. 12. Tool-life carbides.

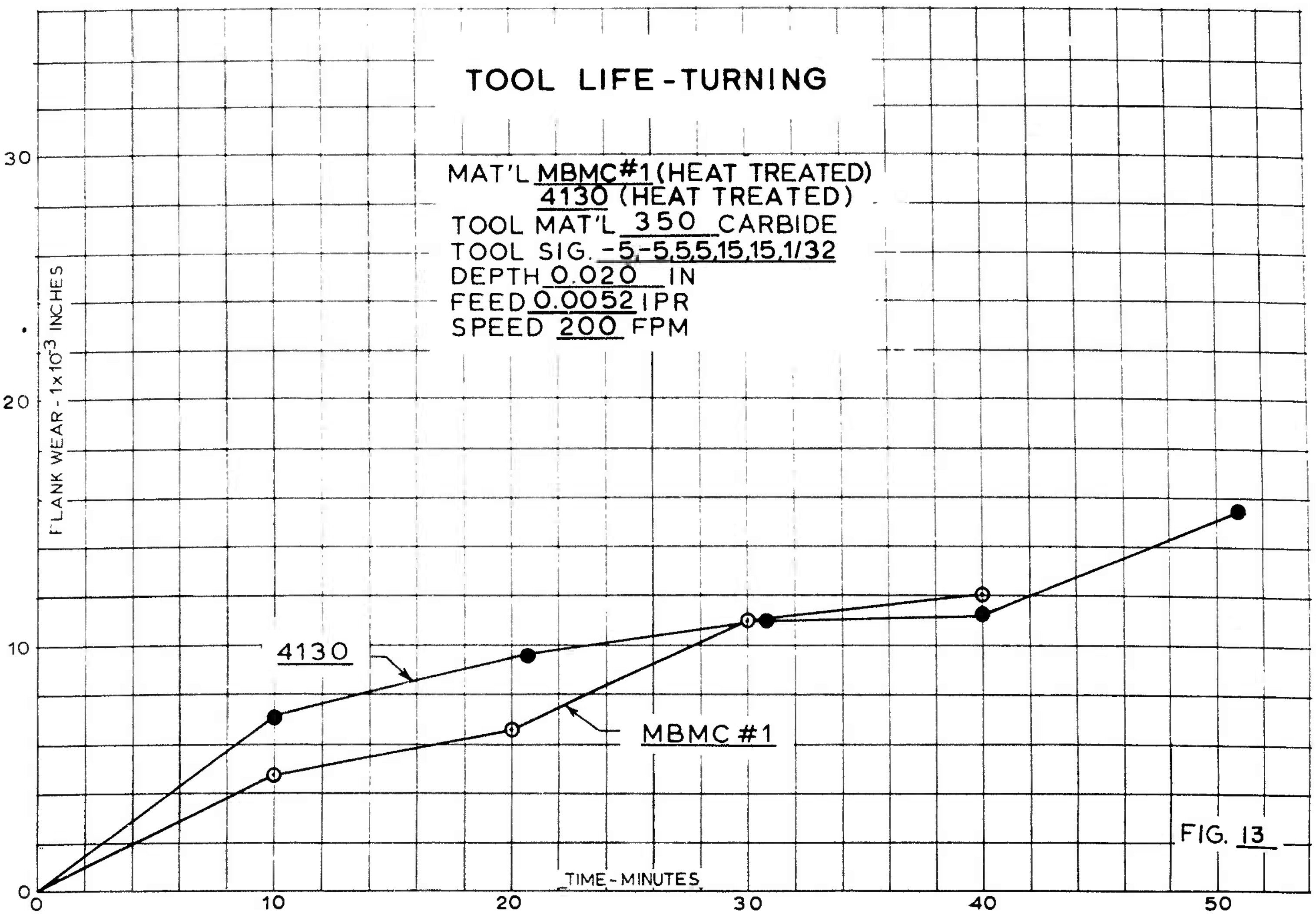


Fig. 13. Tool-life carbides.

the heat-treated MBMC-1 must be reduced 40-50% below that for hardened AISI 4130 for roughing cuts. On the other hand, Figs. 11 to 13 show that MBMC-1 can be cut at approximately the same speed as AISI 4130 for finishing cuts.

Conclusions

(1) Carboloy Grade-350 carbide or equivalent is best for both MBMC-1 and AISI 4130 steels in both the as-forged and the hardened conditions.

(2) MBMC-1 can be cut at the same conditions as AISI 4130 except for roughing cuts on hardened material.

(3) Cutting speeds for hardened MBMC-1 must be reduced 40-50% below those used for rough-turning of hardened AISI 4130 steel.

C. CUTTING FORCES-TURNING

The hardened materials only were subjected to a series of turning tests to determine the magnitude of cutting forces with carbide tools since the tool-life tests indicated that this type of tool material must be used for the heat-treated MBMC-1. Similar information for high-speed steel tools and as-forged work materials was obtained by drilling and milling as reported in (D) and (E).

Test Conditions and Procedure.—The feed rate and the depth of cut were varied in several combinations while the tool shape, tool material, and cutting speed were held constant. All work materials were subjected to the same test procedure wherein readings of cutting force (perpendicular to the plane of tool motion) and feeding force (parallel to axis of work rotation) were obtained for lathe turning cuts.

A 14-in. swing Monarch engine lathe was operated at spindle speeds which produced surface velocities of 200 fpm for the heat-treated 4130 and 100 fpm for the heat-treated MBMC-1 steels, respectively. Feed rates for the 4130 were 0.0038, 0.0076, and 0.0153 ipr, while the depth of cut was maintained constant at 0.100 in. Similarly, the effect of depth of cut was evaluated at depths of 0.020, 0.040, 0.060 and 0.100 in. with the feed held constant at 0.0076 ipr. Feed rates for the MBMC-1 were 0.0028, 0.0038, 0.0056, 0.0076, 0.0112 and 0.0153 ipr, while the depth of cut was constant at 0.100 in. Corresponding depths of cut for a constant feed rate of 0.0076 ipr were 0.020, 0.030, 0.040, 0.060, 0.080 and 0.100 in. The tool signature for all tests was -5, -5, 5, 5, 60, 30, 1/32.

Readings of feeding and tangential forces were obtained simultaneously on a dynamometer equipped with electric-resistance strain gages. No cutting fluid was used.

Test Results.—The effects of feed rate and depth of cut on the cutting or tangential force are shown in Fig. 14. The 4130 steel exhibits a true exponential trend. The equation for this line takes the general form $F_c = Kf^x d^y$, where:

F_c = cutting force in lbs.,
K = a proportionality constant,
f = feed rate in ipr,
d = depth of cut in in., and
x and y = exponents.

Table IV contains the equations derived for this material. The MBMC-1 exhibits nonlinearity and discontinuity. Consequently the equations represent only the maximum forces. This phenomenon appears to be sensitive to the rate of energy release and probably to temperature as well. Spot sampling indicates that abrupt increases in forces occurred at lighter cuts when the cutting speed was higher.

The effects of feed and depth on the feeding force are shown in Fig. 15. An exponential trend is clearly shown for the 4130, but again the MBMC-1 exhibits unstable cutting at certain combinations of conditions.

TABLE IV

SUMMARY OF CUTTING FORCE EQUATIONS

Material	Equations	Force Component
4130	$F_c = 88{,}350f^{0.70} d^{0.83}$	Tangential
Inch	$F_f = 11{,}550f^{.40} d^{.84}$	Feeding
MBMC-1	$F_c = 63{,}500f^{.58} d^{.91}$	Tangential
Inch	$F_f = 8{,}100f^{.20} d^{.83}$	Feeding

D. DRILLING TORQUE AND THRUST

The two work materials (as-forged) were drilled at several different cutting conditions to determine the effects of feed rate and drill diameter on

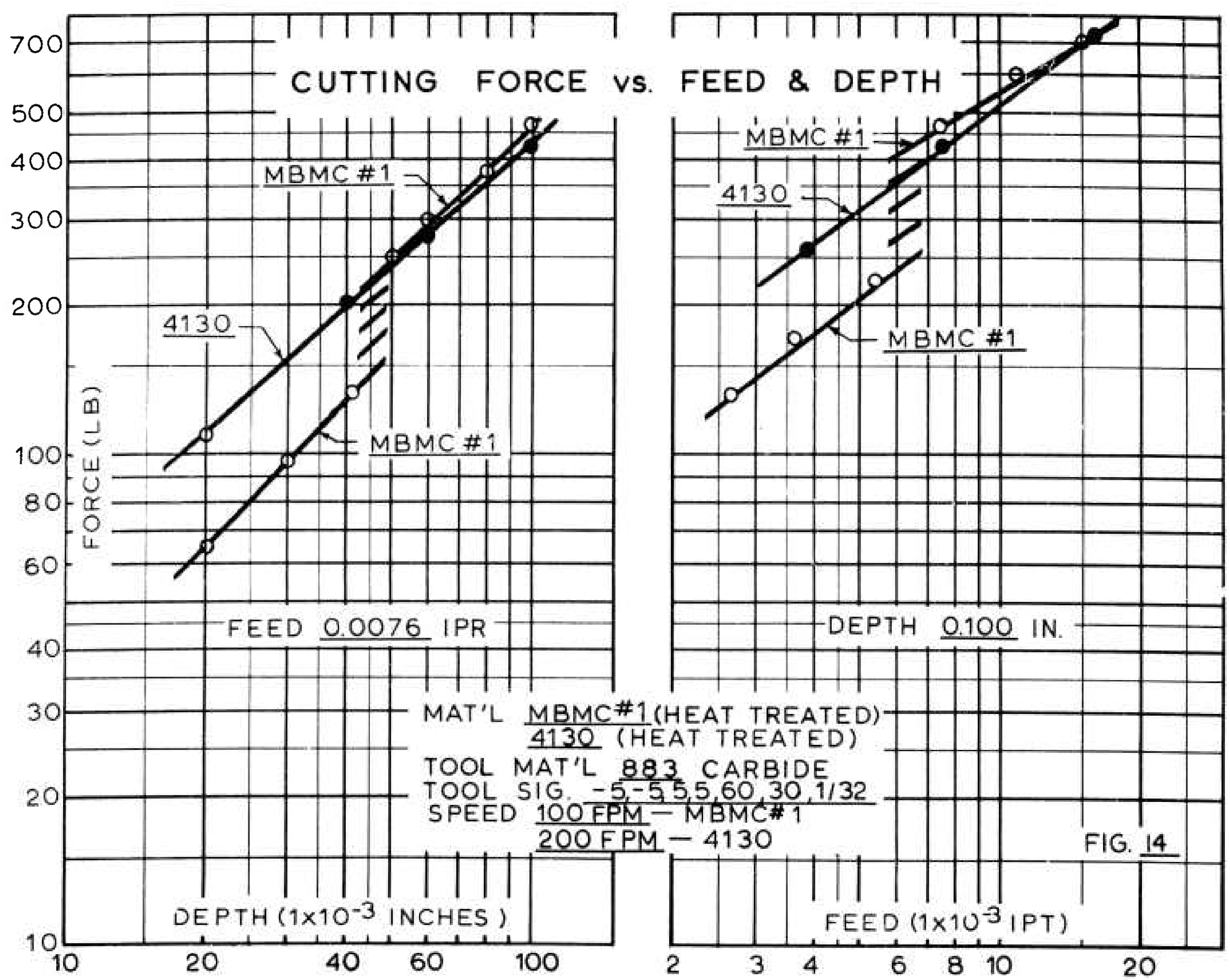


Fig. 14. Cutting force "F_c" vs. size of cut.

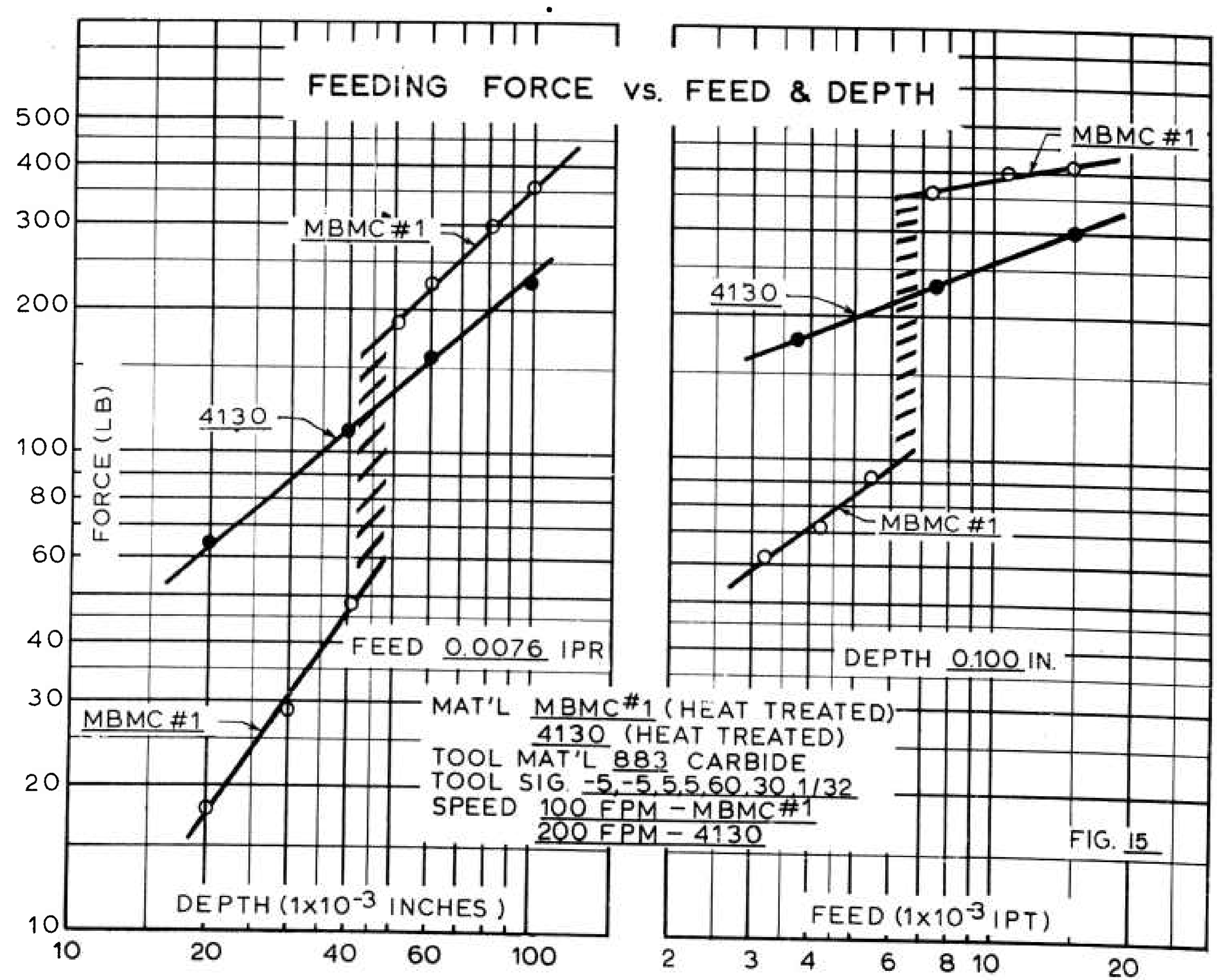


Fig. 15. Feeding force "F_n" vs. size of cut.

torque and thrust requirements.

Procedure.—Test specimens were in the form of rectangular pieces, 1 x 1-1/4 x 4 in. long. The workpieces were held in a vise on a dynamometer and drilled at a particular feed rate and cutting velocity.. When the readings of torque and thrust stabilized, the cut was discontinued. The same procedure was followed for both work materials.

A sharp drill was used for each test. Several tests were duplicated to check the repeatability of the results.

Test Conditions.—All tests were conducted on a Fosdick 25-in., box column, upright drill press at a cutting velocity of 35 fpm. All drilling was done dry.

Drills were standard high-speed steel, two-fluted twist drills. They were ground to a point angle of 118° and a relief angle of 6°. The helix angle was 30°. Web thicknesses are listed below.

Drill Diam., in.	Web Thickness, in.
0.250	0.043
0.375	0.066
0.500	0.082
0.625	0.086

At constant drill diameter of 0.375 in., the feed rates were 0.005, 0.007, 0.009, and 0.011 ipr. At constant feed rate of 0.005 ipr, the drill diameters were 0.250, 0.375, 0.500 and 0.625 in.

A dynamometer containing wire-resistance strain gages was used to obtain readings of torque and thrust.

Test Results.—Figures 16 and 17 show the results plotted on logarithmic coordinates.

The effects of drill diameter and feed rate on cutting torque (lb-in.) and thrust (lb) show a uniform behavior or exponential trend. Such straight lines may be represented by equations of the general forms

$$T = Kf^aD^b \text{ and } B = Cf^xD^y$$

where:

T = cutting torque in lb-in. (sharp tools),
K = a proportionality constant,

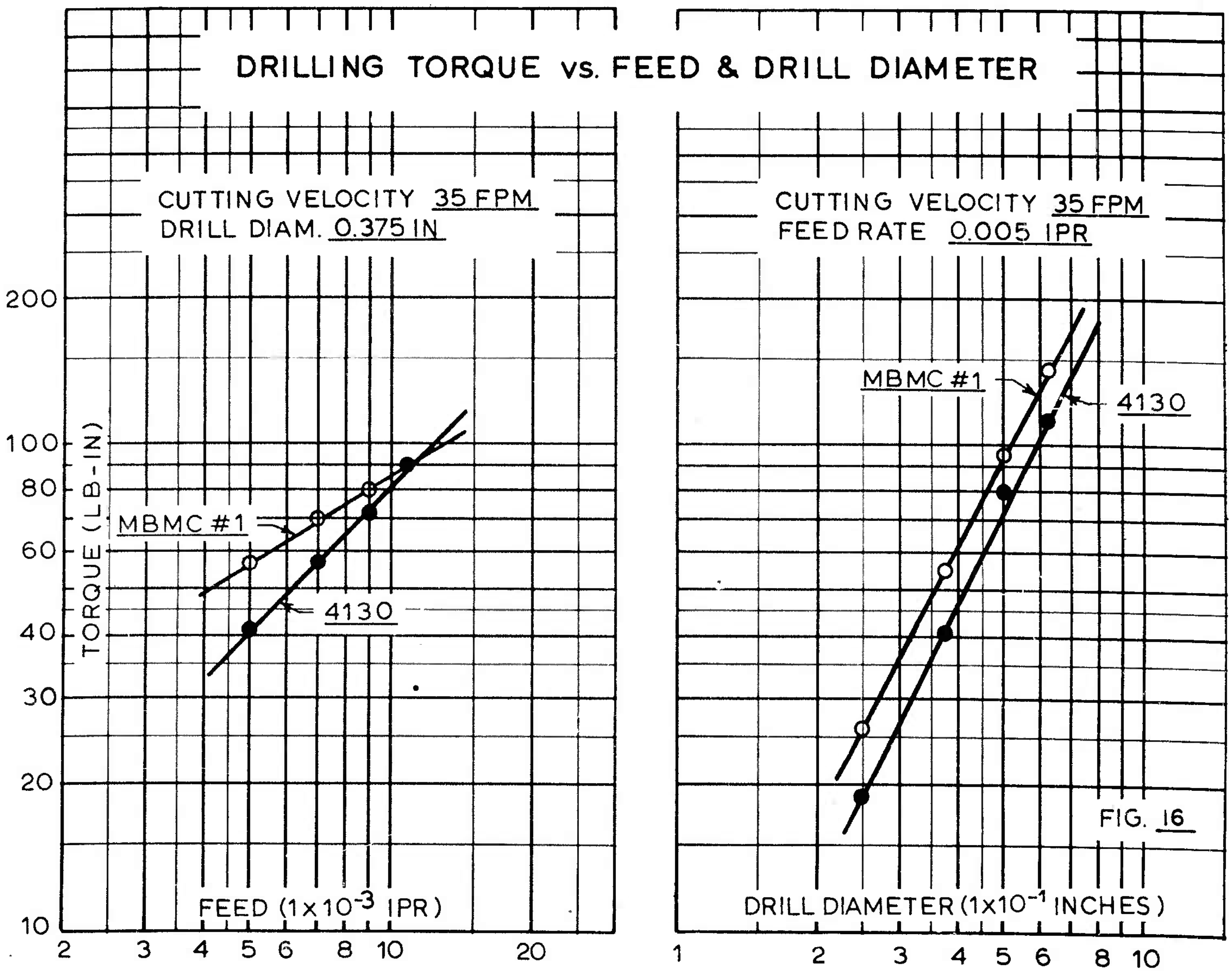


Fig. 16. Drilling torque vs. feed and drill diameter.

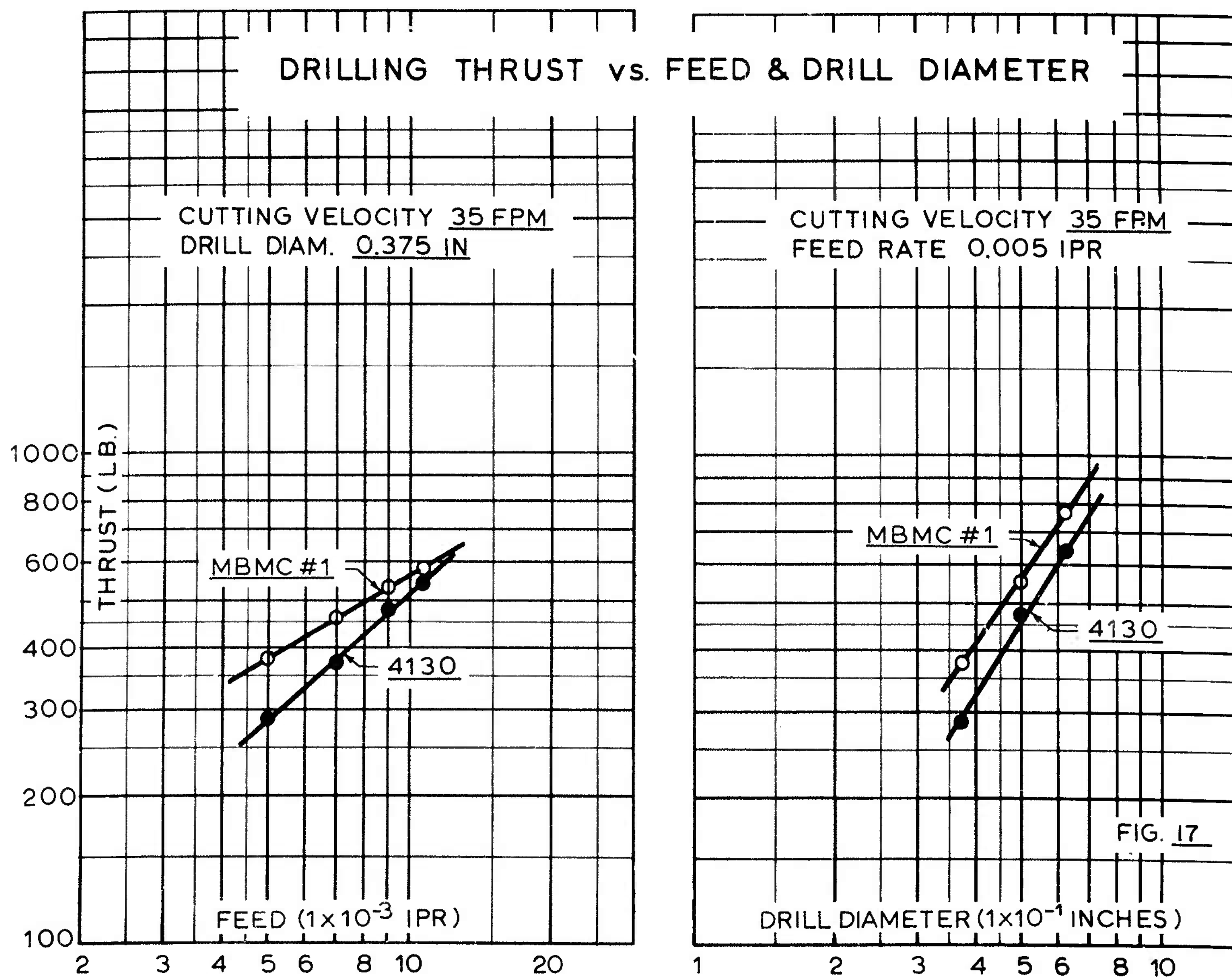


Fig. 17. Drilling thrust vs. feed and drill diameter.

f = feed rate, ipr,
D = diameter of drill, inches,
a and b = slopes of lines showing the effects of feed and drill diameter on torque,
B = thrust force in lb (sharp tools),
C = proportionality constant,
x and y = slopes of lines showing the effects of feed and drill diameter on thrust.

Tables V and VI contain the equations derived from the straight lines. These can be used to calculate the torque and thrust when drilling the two materials (as-forged) and are most applicable within the range of variables actually used in this study.

TABLE V

DRILLING TORQUE AS A FUNCTION OF FEED RATE AND DRILL DIAMETER

Material	Equations
4130	$T = 47{,}650 f^{0.96} D^{2.03}$
MBMC-1	$T = 7{,}230 f^{0.57} D^{1.88}$

TABLE VI

DRILLING THRUST AS A FUNCTION OF FEED RATE AND DRILL DIAMETER

Material	Equations
4130	$B = 76{,}450 f^{0.77} D^{1.54}$
MBMC-1	$B = 30{,}850 f^{0.58} D^{1.38}$

Conclusions

(1) The effect of feed rate on torque is more pronounced for the 4130 material as can be concluded from the slope of the lines.

(2) At the higher feed rate the torque for both the MBMC-1 and 4130 are the same for the given drill diameter.

(3) The MBMC-1 exhibited higher torque readings consistently relative to the effect of drill diameter.

(4) The effect of feed rate on thrust is also more pronounced for the 4130 material.

(5) Effect of drill diameter on drill thrust resulted in consistently higher forces for the MBMC-1 material.

(6) The most important difference between the two work materials is that the MBMC-1 does not cut as well nor as efficiently at small feed rates.

E. ENERGY REQUIREMENTS

Energy requirements were measured directly with a pendulum type dynamometer in a milling cut. High-speed steel tools with a rake angle of 8° simulated average shop conditions.

Test Conditions and Procedure.—The HSS cutting tool had a width of 0.245 in., 8° radial rake angle, 6° side clearance angles, and 12° end clearance angle. Ten cuts were made at a depth of cut of 0.100 in. and a feed setting of 0.003 ipt; the work advanced 0.003 in. for each successive cut. After the above cuts were made, tests were run at a feed setting of 0.006 ipt for ten cuts, then changed to 0.010 ipt for another ten cuts. The above sequence of feed settings was repeated at depths of cut of 0.060 in., and 0.020 in., ten cuts being taken for each combination. The same procedure was used for both work materials.

Test Results.—Figures 18 and 19 show the results plotted on logarithmic coordinates.

The effects of feed (ipt) and depth of cut upon the energy requirements can be expressed by the equation:

$$E = Kf^x d^y$$

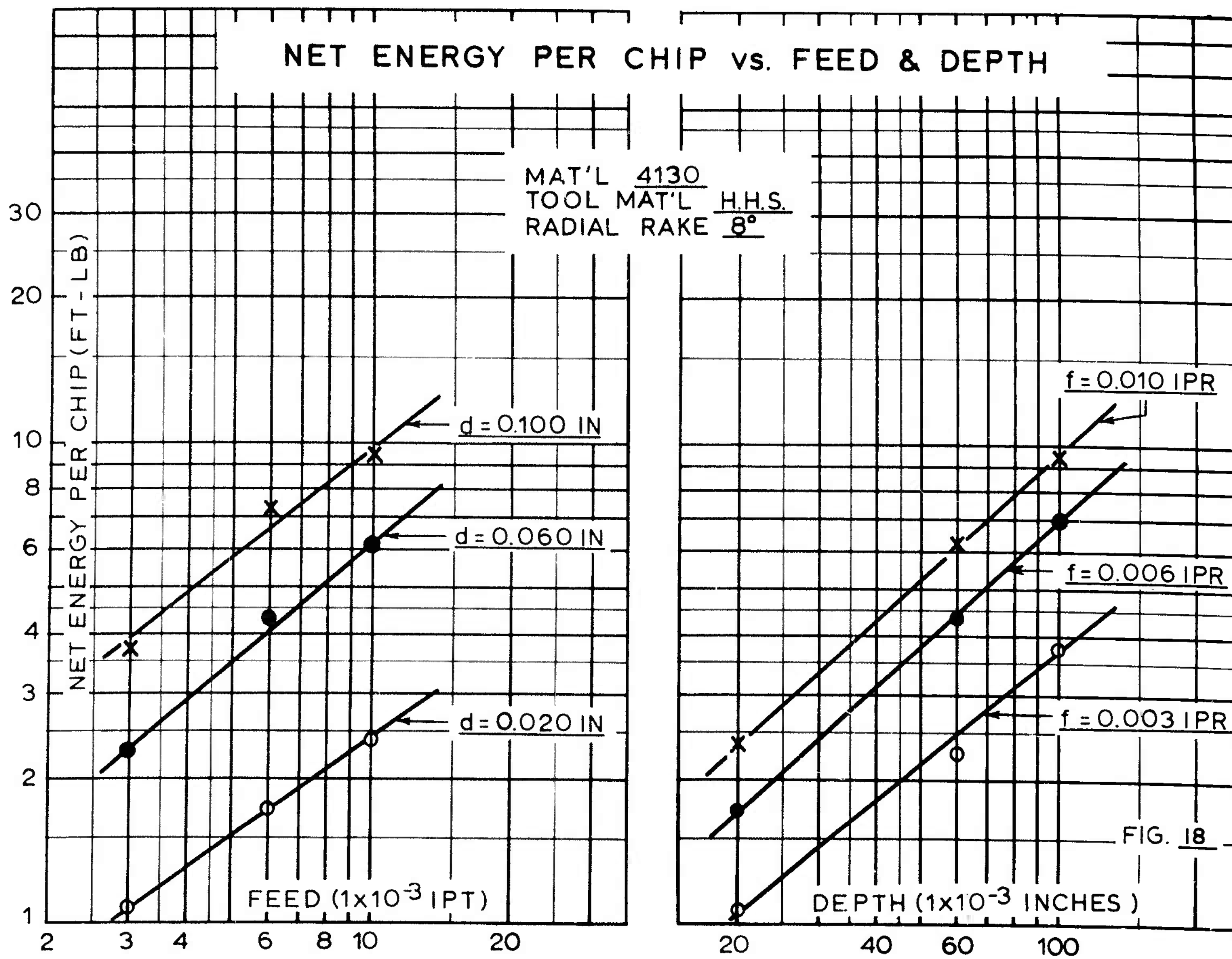


Fig. 18. Cutting energy for AISI 4130 steel.

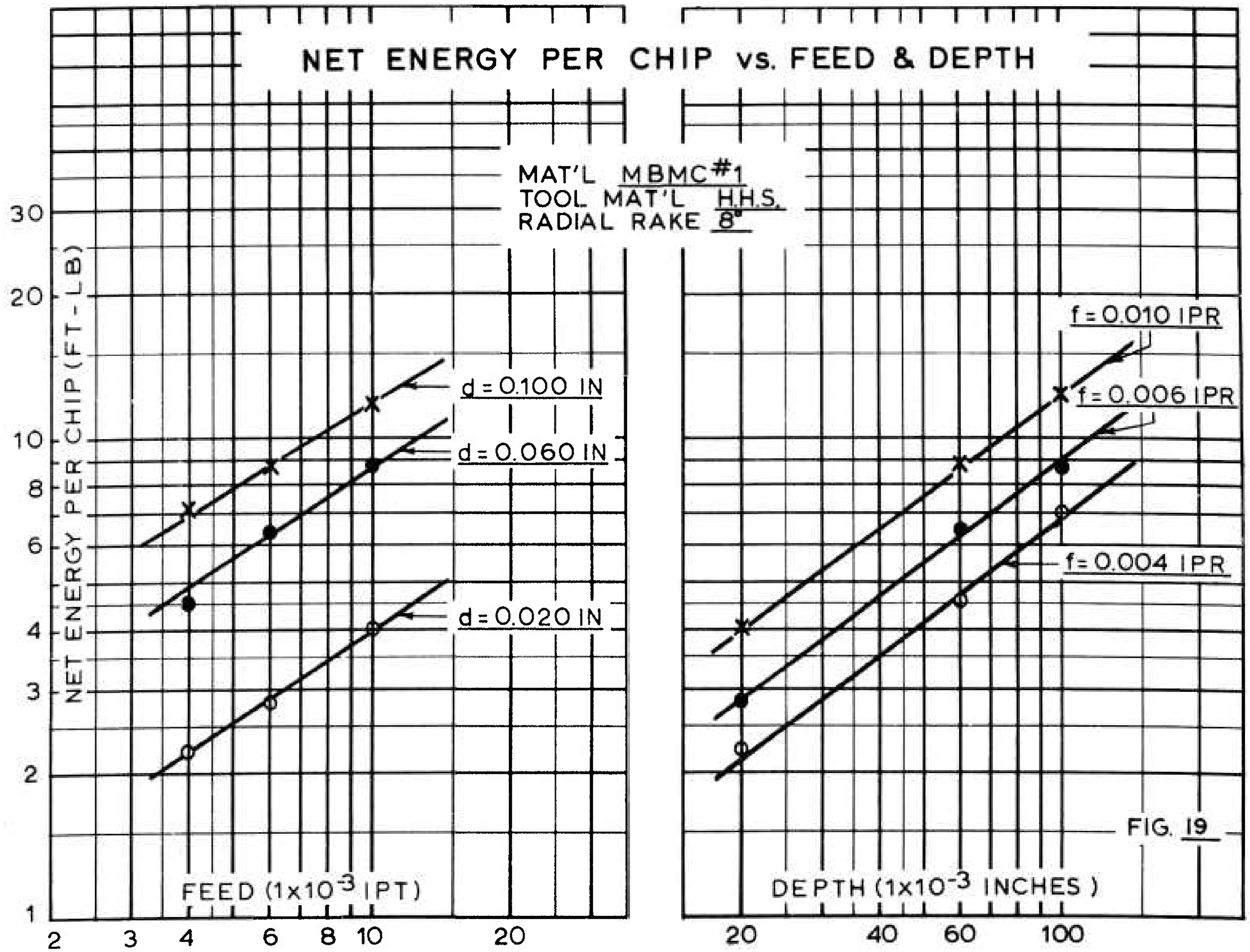


Fig. 19. Cutting energy for MBMC-1 steel.

where:

E = energy in ft-lb per chip,
K = a proportionality constant for calculating energy,
f = feed in inches per tooth,
d = depth of cut in inches,
x and y = slopes of lines showing the effects of feed and depth on the energy required to cut.

Table VII contains the equations derived from the straight lines shown in Figs. 18 and 19.

TABLE VII

CUTTING ENERGY FOR MILLING

Material	Energy Equations	Unit HP_c*
4130	$E = 2{,}200f^{0.76}d^{0.84}$	1.16
MBMC-1	$E = 873f^{0.60}d^{0.70}$	1.48

*Feed, 0.010 ipt; depth, 0.100 in.

Conclusions

(1) The energy required to cut the MBMC-1 is consistently higher for all combinations of cutting conditions.

(2) Both materials exhibit a common characteristic relative to the effect of feed versus the effect of depth. The above equations show that the energy per chip increases with an increase in feed and depth of cut only as the 0.60 power of the feed and the 0.70 power of the depth of cut for the MBMC-1. For the 4130 the same is true, the power of the feed being 0.76 and 0.84 for the depth of cut. This indicates the desirability of taking heavy feeds from a power standpoint. This property is common to most metals.

VI. SPECIFIC CONCLUSIONS

(1) Tool Life with High-Speed Steel Tools

(a) Cutting speeds for the as-forged MBMC-1 must be about 40% lower than for as-forged 4130.

(b) It is impractical to cut heat-treated MBMC-1 with high-speed steel tools.

(2) Tool Life with Carbide Tools

(a) Carboloy Grade 350 or equivalent is satisfactory for both steels in both the as-forged and hardened conditions.

(b) As-forged MBMC-1 can be cut at speeds about 80% as high as for as-forged 4130.

(c) Hardened MBMC-1 must be cut at about 50%-70% of the speed for hardened 4130.

(3) Tool Life with Ceramic Tools

(a) It is feasible to turn both as-forged and hardened MBMC-1 at speeds from 600 to 1000 fpm.

(4) Cutting Forces

MBMC-1 produces substantially higher cutting forces than 4130 in both the as-forged and hardened conditions. Consequently, it is necessary to provide substantial rigidity to obtain satisfactory tool life, surface finish, and size control.

Joint Army-Navy-Air Force Mailing List for the Distribution of Solid Propellant Technical Information

Union Carbide Corporation
270 Park Avenue
New York 17, New York
Attn: B. J. Miller 3

Commander
Aeronautical Research Lab.
Wright-Patterson Air Force Base
Ohio - Attn: RRLC 4

Commander
Air Proving Ground Center
Eglin Air Force Base
Florida
Attn: Technical Library 5

Commander
Air Force Ballistic Missile Division
Hq. Air Res. & Dev. Command
P. O. Box 262 - Attn: WDAT
Inglewood, California 6

United Technology Corp.
P. O. Box 358
Sunnyvale, California
Attn: Librarian 7

Department of the Army
Office, Chief of Ordnance
Washington 25, D. C.
Attn: ORDTB 8

Commanding Officer
Office of Ordnance Research
Box CM - Duke Station 9
Durham, North Carolina

Commanding General
Ordnance Ammunition Command
Joliet, Illinois 10
Attn: ORDLY-AREL, Engr. Library

Commanding Officer
Diamond Ord. Fuze Laboratories
Washington 25, D. C.
Attn: ORDTL (012) 11

Commander
Army Ballistic Missile Agency
Redstone Arsenal, Alabama
Attn: ORDAB-HSI 12

Bureau of Naval Weapons
Department of the Navy
Washington 25, D. C.
Attn: RMMP-2 13

Bureau of Naval Weapons
Department of the Navy
Washington 25, D. C.
Attn: RMMP-331 14

Commanding Officer
Office of Naval Research
1030 East Green Street
Pasadena 1, California 17

Commander
U. S. Naval Ordnance Lab.
White Oak, Silver Spring, Md.
Attn: Library 16

Aerojet-General Corporation
P. O. Box 296
Azusa, California
Attn: Librarian 19

Director
Special Projects Office
Department of the Navy
Washington 25, D. C. 18

Armour Research Foundation
of Ill. Institute of Technology
Technology Center, Dept. D
Chicago 16, Illinois 20
Attn: Fluid Dynamics & System Res.

Arthur D. Little, Inc.
15 Acorn Park
Cambridge, Mass. 22
Attn: W. A. Sawyer

Standard Oil Company
Research Department
P. O. Box 431
Whiting, Indiana 24
Attn: B. H. Shoemaker

Olin Mathieson Chemical Corp.
Attn: Res. Library - Box 508
Marion, Illinois 26

Rocketdyne, A Division of North
American Aviation, Inc.
Solid Propulsion Operations
P. O. Box 548, Attn: Library
McGregor, Texas 28

Aerojet-General Corporation
Box 1168
Sacramento, California 30-32
Attn: R. G. Weitz, Head
Technical Information Center

Hercules Powder Company
Bacchus Works
Magna, Utah 34
Attn: Librarian

Commanding Officer
Picatinny Arsenal
Dover, New Jersey 37-38
Attn: Library

Jet Propulsion Laboratory
4800 Oak Grove Drive
Pasadena, California
Attn: I. E. Newlan 21
Chief, Reports Group

Midwest Research Institute
425 Volker Boulevard
Kansas City, Mo. 23
Attn: Librarian

Thiokol Chemical Corporation
Elkton Division
Elkton, Maryland
Attn: Librarian 25

Reaction Motors Division
Thiokol Chemical Corporation
Denville, New Jersey
Attn: Librarian 27

American Machine & Foundry Co.
Mechanics Research Dept.
7501 North Natchez Avenue
Niles 48, Illinois
Attn: Phil Rosenberg 29

The Martin Company
Baltimore, Maryland
Attn: T. W. Woodard 33

Commanding General
Aberdeen Proving Ground
Maryland 35-36
Attn: Ballistic Res. Lab.

Bureau of Naval Weapons
Department of the Navy
Washington 25, D. C. 39-40
Attn: DLI-3

Commander
U.S. Naval Air Missile Test Center
Point Mugu, California 41-42
Attn: Technical Library

Commander
U.S. Naval Ordnance Test Station
China Lake, California 45-46
Attn: Tech. Library Branch

Atlantic Research Corporation
Shirley Highway & Edsall Road
Alexandria, Virginia 49-50

Rocketdyne
6633 Canoga Avenue
Canoga Park, Cal. 15, 53-54
Attn: Library, Dept. 596-306

Solid Propellant Information Agency
Applied Physics Laboratory
The Johns Hopkins University
Silver Spring, Md. 58-60

Commander
Air Force Flight Test Center
Edwards Air Force Base
California 65
Attn: FTRS

Chief, Bureau of Naval Weapons
Department of the Navy
Washington 25, D. C.
Attn: DSC-3 67-74

Defense Metals Information Center
Battelle Memorial Institute
505 King Avenue 85
Columbus 1, Ohio

Space Technology Lab., Inc.
5730 Arbor-Vitae Street
Los Angeles, California 93
Attn: Mr. R.C. Anderson

Commanding Officer
U.S. Naval Propellant Plant
Indian Head, Maryland 43-44
Attn: Res. & Dev. Dept.

Hercules Powder Company
Allegany Ballistics Laboratory
Cumberland, Maryland 47-48
Attn: Library

Thiokol Chemical Corporation
Redstone Division
Huntsville, Alabama 51-52
Attn: Technical Director

Commanding General
White Sands Missile Range
New Mexico 55-57
Attn: ORDBS-OM-TL

Commander
Army Rocket & Guided Missile Agency
Redstone Arsenal, Ala. 61-64
Attn: Tech. Library, ORDXR-OTL

U.S. Department of the Interior
Bureau of Mines
4800 Forbes Street
Pittsburgh, Pennsylvania 66
Attn: M. P. Benoy, Expl. Res. Lab.

Commander
Armed Services Tech. Inf. Agency
Arlington 12, Virginia 75-84
Attn: TIPCR

Nat'l Aeronautics & Space Adm.
1520 H Street, N. W.
Washington 25, D. C. 86-92
Attn: Chief, Div. of Res. Information

Grand Central Rocket Company
P. O. Box 111
Redlands, California 94-96
Attn: Helen Ashman, Librarian

Borg-Warner Corporation
200 South Michigan Avenue
Chicago, Illinois 97-98
Attn: Mr. J. H. Ingersoll

Bendix Aviation Corporation
South Bend, Indiana
Attn: Tech. Library 100

Hughes Tool Company
Culver City, California
Attn: Materials Dept. 102

Technical Library
U. S. Naval Weapons Laboratory
Dahlgren, Virginia 104
Ref. Doc. Req/Trans. 12/5/59

The Library
United Aircraft Corporation
400 Main Street 107
East Hartford, Connecticut
via
Bureau of Aeronautics Representative
East Hartford, Connecticut

Allison Division - General Motors
P. O. Box 894
Indianapolis, Indiana 99
Attn: Eng. Research Library

General Electric Company
Flight Propulsion Division
Cincinnati 15, Ohio 101
Attn: Tech. Inf. Center

Aerospace Corporation
P. O. Box 95085
Los Angeles, California 103

Thiokol Chemical Corporation
Utah Division
Brigham City, Utah 105-106
Attn: Library Section

In accordance with the terms of Information Exchange Project B-29 this report has been forwarded to the British Joint Services Mission, Ministry of Supply Staff, P. O. Box 680, Benjamin Franklin Station, Washington, D. C., Attn: Mr. E. G. Lewis for transmittal to Imperial Chemical Industries, Ltd. (2) 108-109