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R E P O R T

of the

ORE DRESSING AND METALLURGICAL LABORATORIES.

Investigation No. 1805.

Examination of Welded Tubing from Fuselage of Harvard Aircraft.

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(Copy No. 10.)

Bureau of Mines
Division of Metallic
Minerals

Physical Metallurgy
Research Laboratories

CANADA

DEPARTMENT
OF
MINES AND RESOURCES

Mines and Geology Branch

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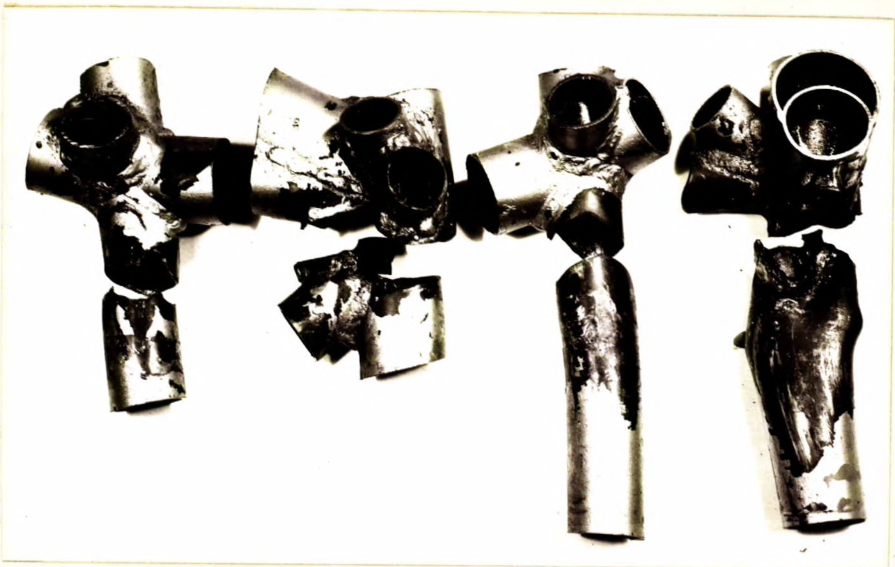
Examination of Welded Tubing from
Fuselage of Harvard Aircraft.

Origin of Material and Object of Investigation:

On December 8, 1944, samples of tubing (see Figure 1) taken from the fuselage of a Harvard aircraft which had broken up in the air were submitted, for metallurgical examination, by the Chief of the Air Staff, Department of National Defence for Air, Ottawa, Ontario. The locations of the samples (arbitrarily called A, B, C and D) on the fuselage frame structure are shown in Figure 2.

The covering letter (File No. 938BG-1-5(AMSO S.4-0-1), dated December 7, 1944, stated that an opinion had been expressed that the initial failure occurred at section A due to the existence of a previous crack followed by secondary failure at the other sections. It was requested that a metallurgical examination be made of the failure in tubing A, as well as the other materials, in order to determine the cause of failure.

Figure 1.

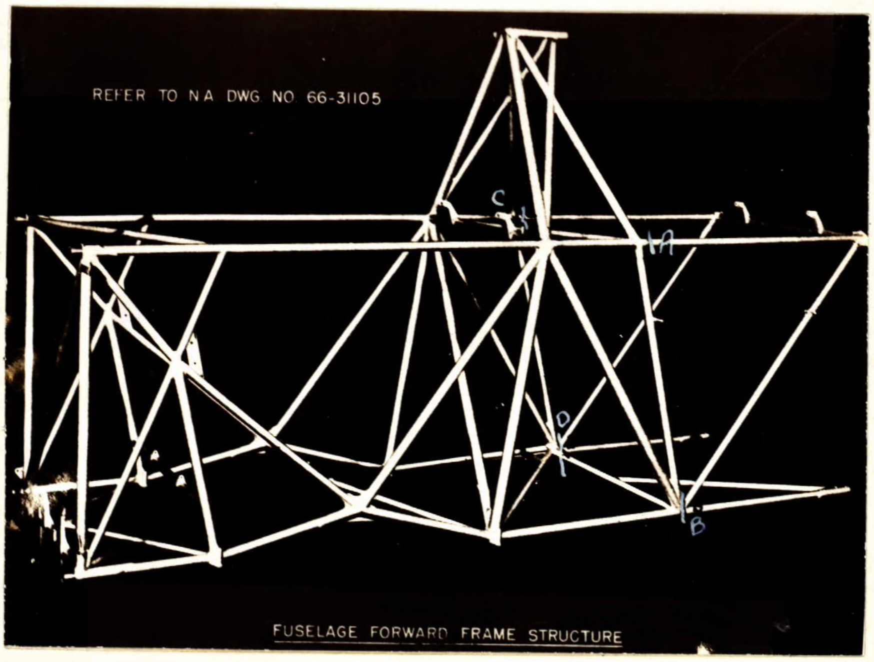


A B C D

FRACTURES IN HARVARD AIRCRAFT FUSELAGE TUBING.

(Approximately 1/3 Normal size).

Figure 2.



POSITION OF FRACTURES IN THE FUSELAGE.

Macro-Examination:

Visual examination revealed the following:

Failure at section A occurred in the tubing proper, a short distance away from a weld (see Figure 1). The nature of the fracture is shown in Figure 3.

Failures at sections B, C and D were located in weld areas, the fractures in B and C occurring at the toe of the weld (see Figures 4 and 5) whereas the fracture at D occurred through the weld metal (see Figure 6).

Chemical Analysis:

Samples of tubing, taken from sections A, C and D, were submitted to chemical analysis. In Table I below, the results are compared with the chemical requirements of SAE X4130 steel.

TABLE I.

	Sample A	Sample C	Sample D	Specification SAE X4130
		- Per Cent -		
Carbon -	0.35	0.31	0.34	0.25-0.35
Manganese -	0.54	0.49	0.44	0.40-0.60
Silicon -	0.26	0.25	0.28	-
Sulphur -	0.016	0.028	0.022	0.05 max.
Phosphorus -	0.014	0.011	0.011	0.04 "
Nickel -	Nil.	Nil.	Nil.	-
Chromium -	0.95	0.77	0.90	0.80-1.10
Molybdenum -	0.14	0.19	0.17	0.15-0.25

McQuaid-Ehn Test:

A sample of the tubing from section A was heated in a carburizing atmosphere for 8 hours at 1700° F. and cooled in the furnace, the object being to determine the grain size.

The grain size was found to be mixed (see Figure 17).

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Normalizing Experiment:

Samples of tubing from sections A and C were heated at 1650° F. and allowed to cool in air. Hardness readings were taken, using the Vickers hardness tester.

The specimens were then drawn at 1250° F. for 2½ hours, in a lead bath, and hardness readings again taken.

Hardness Testing:

A hardness survey was made on all of the sections submitted, using the Vickers hardness tester with a 20-kilogram load. These readings were then converted to Rockwell "C" values.

Hardness readings were made in duplicate at the fractured surface of the tubing of section A. These were found to range from 21 to 52 Rockwell "C".

A break-down of the hardness readings of the different sections at the welds is given in Table II.

TABLE II. - Rockwell "C" Hardness.

Section		Normal Zone	Heat-Affected Zone	Weld	At Fracture
A	-	20-23	36-52	23	21-52
B	-	-	33-40	23-25	33-40
C	-	21-22	39-48	23	39-48
D	-	24-25	31-50	29	31-50

Hardness readings obtained on samples of tubing subjected to various heat treatments are given in Table III.

TABLE III.

Section		Heat Treatment	Hardness, Rockwell "C"
A	-	Normalized, 1650° F.	36-37
C	-	" "	36-38
A	-	Normalized and drawn at 1250° F.	17-20
C	-	Water-quenched, 1550° F.	46-54

-

Microscopic Examination:

Figures 9 and 10, both at X1000 magnification, are photomicrographs of tubing taken from section A at the fracture, away from the heat effect of a weld. The microstructure consists of ferrite, pearlite, and undissolved carbides. The structure shown in Figure 10 appears to be more greatly spheroidized than that of Figure 9.

Figures 11 and 12, both at X1500 magnification, are photomicrographs of tubing taken from section C, at areas away from the weld. The microstructure of Figure 11 consists of lamellar pearlite, spheroidized carbides, and ferrite, whereas the microstructure of Figure 12 consists of spheroidized carbides and ferrite. This latter structure may be considered typical of this tubing as delivered to the welder. It should be noted that the tubing of section A, containing the fracture, had the same structure away from the weld as that of C.

Figure 13 shows the microstructure, at X1000 magnification, obtained by normalizing a piece of tubing A from 1650° F. The structure consists of a low-temperature transformation product, proeutectoid ferrite, and carbides. This photomicrograph should be compared with Figure 14, which shows the structure obtained by gas-quenching (rapid normalizing) SAE X4130 tubing of 0.35 per cent carbon content. (Transactions A.S.M. Vol. 33, 1944, p. 312).

Figure 15 is a photomicrograph, at X1500 magnification, of the structure obtained by normalizing a sample of tubing from 1650° F., followed by drawing at 1250° F. in a lead bath for 2½ hours. This structure, which consists of spheroidized carbides in a ground mass of ferrite, should be compared with the photomicrographs in Figures 11 and 12.

Figure 16, taken at X1000 magnification, shows the microstructure in the tubing of section A at an area which was affected by the heat of welding. The structure is martensitic, (Microscopic Examination, cont'd) -

which accounts for the unusually high hardness readings obtained in the heat-affected zones (up to 52 Rockwell "C").

DISCUSSION:

Specification Requirements -

The Army-Navy specifications(1) for aeronautical chrome-molybdenum (X4130) welded steel tubing are covered in detail by the AN-T-3 specification. It lists the following requirements:

TABLE IV.

		Per Cent
Carbon	-	0.25-0.35
Manganese	-	0.40-0.60
Phosphorus	-	0.04 (max.)
Sulphur	-	0.05 (max.)
Chromium	-	0.81-1.10
Molybdenum	-	0.15-0.25

"After the last forming operation, the tubing shall be normalized, stress-relieved, or otherwise heat-treated to develop the physical properties specified in Table V."

TABLE V. - Physical Properties.
(For Normalized X4130 Tubing)

Minimum yield strength	-	75,000 p.s.i.
" tensile "	-	95,000 "
Minimum elongation in 2 inches	-	10 to 12 per cent on full tube.
Minimum elongation in 2 inches	-	5 to 7 per cent on strip.

Normalizing is generally defined as a "heat treatment comprising heating iron-base alloys to approximately 100 degrees Fahrenheit above the critical temperature range, followed by cooling to below the range in still air at ordinary temperature."

According to a paper by W. Pestrak and W. W. Ackerman,(2) "if there is to be no further heat treatment after welding, the steel must be in the normalized state before welding."

(Continued on next page)

(Discussion, cont'd) -

Chemical Analysis -

On the basis of the results of the chemical analyses made on tubing taken from sections A, C and D, the chemical content of the tubing satisfactorily complies with the specification requirements. It should be noted, however, that the carbon content of the tubing meets the upper limits of the specification.

Mechanical Properties -

The spheroidized structure of the tubing (see Figures 10, 11 and 12) as delivered to the welder, indicates that the steel in this condition has a minimum of tensile strength and maximum ductility. If sufficient tubing had been provided to perform tensile tests, it would have been possible to determine the strength of the tubing as supplied to the welder. In the absence of the required length of tubing, it is not possible to state any accurate figure for the tubing in this condition. Rough approximations can be made, however, on the basis of hardness tests.

According to a chart in the book, "Molybdenum in Steel" (section 5, page 15), issued by the Climax Molybdenum Co., the hardness corresponding to a tensile strength of 95,000 p.s.i. would be in the neighbourhood of 17 Rockwell "C". Since the minimum hardness encountered in the tubing examined was 20 Rockwell "C", it is obvious that on this basis failure through low tensile strength cannot be attributed to the spheroidized condition of the steel.

Microscopic examination, as well as hardness tests taken along the fracture in section A, shows that the microstructure varies from that of spheroidized carbides "(see Figure 10) to martensite, and the hardness from 21 Rockwell "C" to 52 Rockwell "C" (see Table II). This unusually great range

(Discussion, cont'd) -

in hardness is due to the heat effect of the welding, which was closer to the fracture at one portion of the periphery than at the remaining portion. It also indicates that the steel is capable of an unusually great magnitude of hardness.

To confirm this latter statement, two pieces of tubing, taken from sections A and C respectively, were normalized by heating at 1650° F. and cooled in still air. The resultant hardness was found to range from 36 to 38 Rockwell "C" (see Table III). The mechanical equivalent would be as follows:

T.S.	Y.P.	Elong. (2 in.)
190,000 p.s.i.	155,000 p.s.i.	6 per cent

The physical properties obtained by normalizing SAE X4130 tubing, of dimensions similar to the tubing in this investigation, as reported in the paper[(2)] are as follows:

	T.S.	Y.P.	Elong.
1-1/8" O.D. x 18 gauge	121,800	94,800	23 per cent

The hardness corresponding to these properties would be in the neighbourhood of 23 Rockwell "C".

An explanation to account for the high hardness encountered in the tubing under examination could be found in either of two conditions:

(a) High carbon content.
(b) Large grain size.

(a) <u>High Carbon Content.</u>

It was stated in Paper[(2)] that normalizing by cooling in inert air was not drastic enough to meet the tensile strength required, especially with steels of lower carbon range, and it was necessary to resort to the use of moving gases. From this observation it may be implied that the hardness of SAE X4130 steels is sensitive to the carbon content, and

(Discussion, cont'd) -

therefore steels in the higher carbon range might have a very high hardness.

(b) Large Grain Size.

The McQuaid-Ehn test (repeated in duplicate) showed a mixed grain size. (See Figure 17).

Whatever the reasons may be for the high hardness attained by the steel, even on normalizing, it is very likely that because of this inherent characteristic it was necessary to draw the steel at a high temperature, in order to obtain the ductility required by the specification, since it is obvious that ordinary normalizing would result in a steel much too brittle to meet these requirements.

The microstructure of a sample which had been quenched and drawn, shown in Figure 15, consists of spheroidized carbides in a background of ferrite and is, to some extent, similar to the structure of the tubing "as received". The hardness was found to range from 17 to 20 Rockwell "C", which is slightly below that recorded for the "as received" tubing. It is likely, therefore, that the tubing may have been drawn at 1250° F. for a shorter period of time, or for the same length of time at 1200° F.

Regarding the failure in the tubing of section A, which occurred a small distance away from the weld, two explanations may be offered to account for the failure:

(a) Spheroidized condition of the steel.
(b) Brittleness due to extremely high hardness.

(a) Spheroidized Condition of the Steel.

Because of the inherently high hardness properties of this steel, it was evidently necessary to produce a spheroidized structure to comply with the specification. Although the

(Discussion, cont'd) -

hardness of this steel may be approximately that of the hardness of a normal steel in the normalized condition, the fatigue strength of the former may be considerably less than that of the latter because of the large ferrite matrix of the spheroidized steel. Hence, failure may have resulted because of low fatigue strength. If additional samples of this "abnormal" steel were available in order that a comparison could be made with that of "normal" SAE X4130 steel, the results might prove of general as well as specific importance.

(b) Brittleness Due to Extremely High Hardness.

Hardness readings obtained at the fracture of the tubing in section A, along the periphery, showed a range from 21 to 52 Rockwell "C", because of the heat effect from the welding. It is quite possible that a crack may have been initiated at a location of high hardness and failure have resulted by the tearing of the softer portion of the tubing, resulting in the jagged, ductile fracture apparent in Figure 3.

Welding -

Figures 4 to 8 show typical welding defects found in all sections. The most common defect is that of unfused roots, which, in some cases, have resulted in cracks (see Figures 5 and 6) and, in others, voids (see Figures 7 and 8).

It is also noted that in some cases the welds are improperly placed, that is, although sufficient weld metal has been applied this metal has not been placed in areas of maximum efficiency (see Figures 4 and 6).

Other failures have been noted in the toes of the weld (Figures 4 and 5) and are presumably due to high hardness in the heat-affected zone, already discussed.

Cracks and voids at roots of welds, as well as high

(Discussion, cont'd) -

hardnesses in the heat-affected zones, all act as high stress raisers and as such can result in failure by cracking, within short periods of service. These stress raisers are particularly dangerous when welded members are subjected to vibration under stress.

In the failure of these parts, it would appear probable that the primary failure has resulted in overstressing of adjacent welds. These welds subsequently failed through areas containing welding defects or high hardness of the heat-affected zone.

The deleterious effect occasioned by the high hardness resulting from welding heat could be eliminated by subjecting the welded structure to a drawing operation after welding. However, this heat treatment subsequent to welding is most likely unfeasible because of the large frame dimensions.

CONCLUSIONS:

1. The tubing employed in this Harvard aircraft fuselage was made from SAE X4130 steel.

2. The chemical content of the tubing complied satisfactorily with the requirements of Army-Navy Specification AN-T-3, for SAE X4130 steel.

3. The carbon content was found to meet the upper limit of the required specification.

4. Failure at section A occurred in the tubing proper, a short distance away from a weld.

5. Failures at sections B, C and D occurred in weld areas as follows:

Fracture in B and C, at the toe of the weld.
Fracture in D, through the weld.

6. The microstructure of the tubing as received

(Conclusions, cont'd) -

for welding consisted of spheroidized carbides and partially spheroidized pearlite in a background of ferrite, indicating a high drawing temperature. The usual heat treatment for SAE X4130 tubing is normalizing.

7. In the absence of a sufficient length of tubing upon which tensile tests may be performed, it is not possible to determine definitely whether the tubing in the spheroidized condition could meet the requirements of the specification. Hardness readings, which may not be completely reliable, indicate that the steel should satisfy the requirements, since the minimum recorded hardness of 21 Rockwell "C" corresponds to a tensile strength of 110,000 p.s.i. and a yield strength of 90,000 p.s.i.

8. The steel employed, possibly because of high carbon content, has the inherent property of achieving unusually high hardness for an SAE X4130 steel.

9. Because the usual normalizing treatment, if applied to this steel, would result in prohibitive hardness and tensile strength, it was most likely necessary for the manufacturer to spheroidize the steel in order to comply with the specifications.

10. Because of the inherently high hardness characteristics of the steel, those portions of the tubing affected by the welding heat result in a hardness which may be quite dangerous as regards brittleness.

11. Hardness readings at the fracture of the tubing at section A ranged from 21 to 52 Rockwell "C". Failure could have resulted from either the low fatigue property of the spheroidized portion, or the brittleness of the hardened zone, caused by the welding heat.

12. Welding defects found consisted of unfused roots,

(Conclusions, cont'd) -

resulting in cracks and voids, and improperly placed welds. Failures also occurred at toes of welds due to high hardness in the heat-affected zones. However, these failures may have resulted subsequent to the failure in the tubing of section A.

13. No heat treatment after welding was evident because of its impracticability.

14. One cannot state definitely that failure had initiated at section A and then spread to other areas, although it is quite possible that such may have been the case.

RECOMMENDATIONS:

1. It is advisable to procure more of the tubing involved, so that a more complete investigation can be possible. A comparison of this steel with that of normal SAE X4130 tubing might result in information which could be of considerable practical value.

2. The tubing should have been supplied to the welder in a condition slightly harder than that called for in the specification, in order that the spheroidized condition might be avoided.

Acknowledgment:

References in this report were made from the following papers:

(1) "Bright Gas Quenching of SAE X4130 and NE 8630 Welded Aircraft Tubes." - Wm. Lehrer, Trans. A.S.M., Vol. 33, 1944, p. 290.

(2) "Flash Welding SAE 4130 Steel Tubing." - W. Pestrack and W. W. Ackerman, The Iron Age, Jan. 25, 1945, p. 46.

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AF:GHB.

Figure 3.

X32. Section "A".

SHOWING DUCTILE NATURE OF FRACTURE IN TUBING.

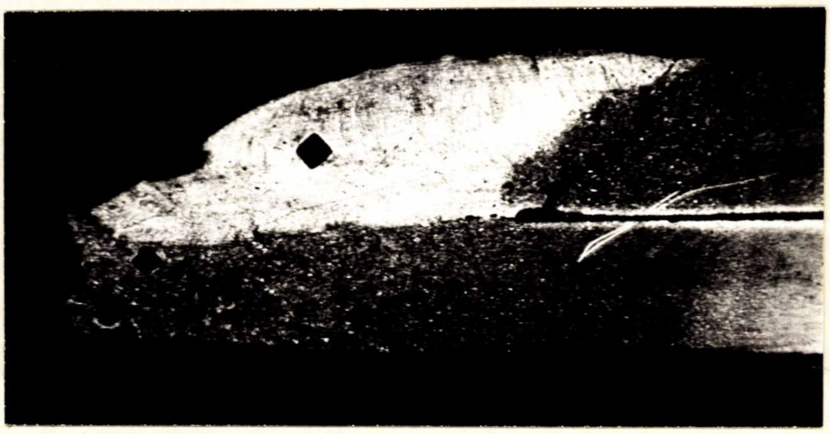
Figure 4.

X11½. Section "B".

FRACTURE AT TOE OF WELD.

Note also unfused root.

Figure 5.



X11½. Section "C".

FRACTURE AT TOE OF WELD.

Note crack originating at toe of weld.

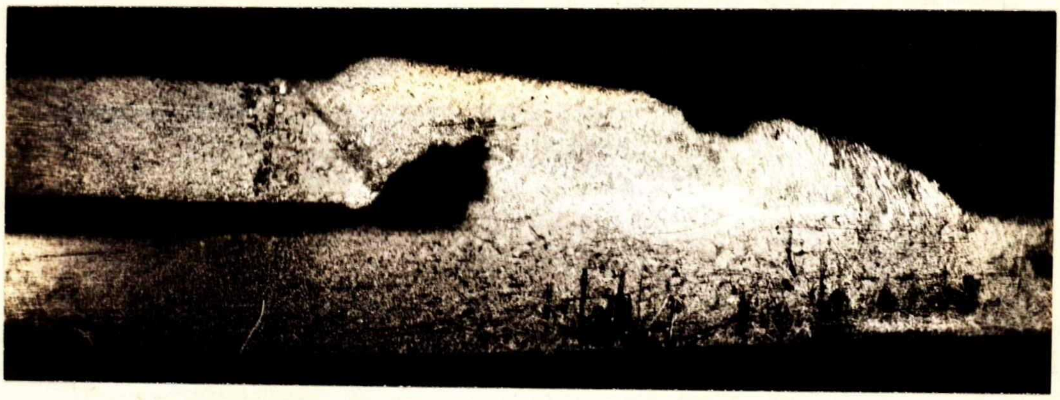
Figure 6.

X11½. Section "D".

FRACTURE THROUGH THROAT OF WELD.

Probably started at unfused root.

Figure 7.



X11½. Section "A".

Note cavity in root of weld, resulting from improper direction of welding flame and subsequently covered with weld metal.

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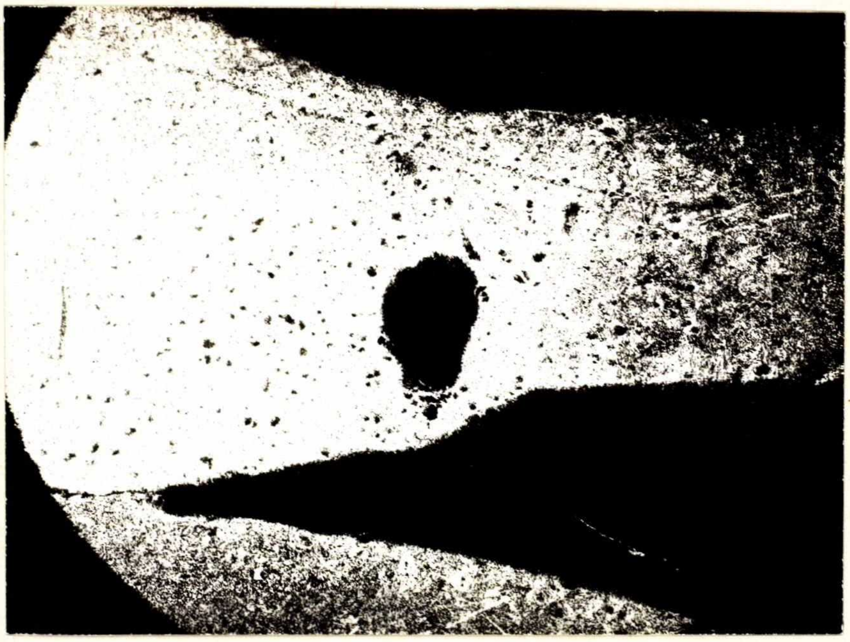
Figure 8.



X30. Section "D".

CAVITY IN TOE AREA OF WELD.

Due to improper flame manipulation.

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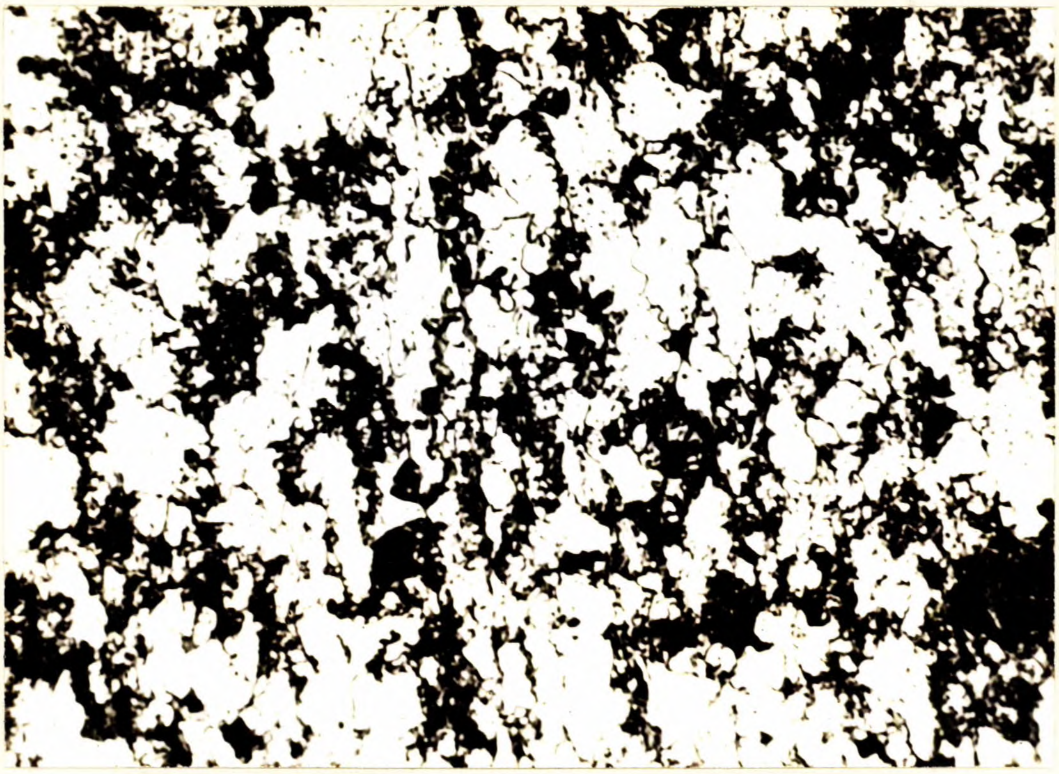
Figure 9.



X1000, nital etch.

SECTION "A" AT FRACTURE.

Ferrite, pearlite, and carbides.

Figure 10.

X1000, nital etch.

SECTION "A" AT BREAK.

Ferrite, pearlite, and carbides.

Figure 11.

X1500, nital etch.

Section "C".

STRUCTURE OF TUBING "AS RECEIVED".

Spheroidized carbides, lamellar partially spheroidized pearlite, and ferrite.

Figure 12.

X1500, nital etch.

Section "C".

TUBING "AS RECEIVED".

Spheroidized carbides in a groundmass of ferrite.

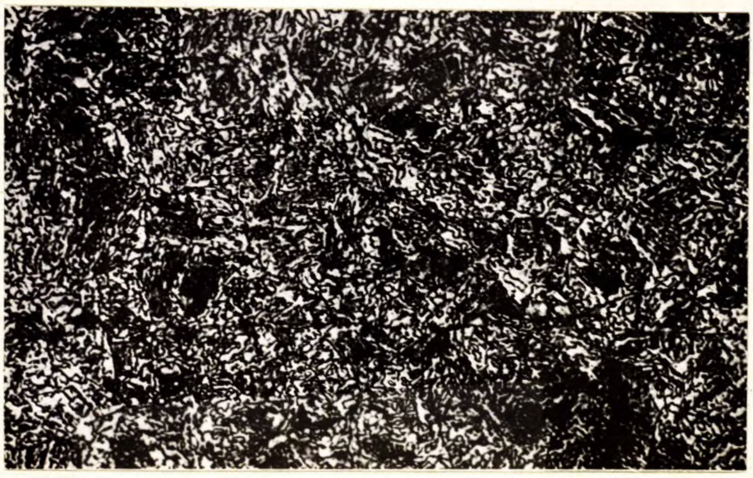
Figure 13.

X1000, nital etch.

SECTION "A", NORMALIZED FROM 1650° F.

Low-temperature transformation product, proeutectoid ferrite, and undissolved carbides.

Figure 14.



X500.

GAS-QUENCHED SAE X4130 STEEL TUBE OF 0.35 PER CENT CARBON CONTENT.

(Trans. A.S.M., 1944, Vol. 33, Page 312).

Figure 15.

X1500, nital etch.

NORMALIZED FROM 1650° F. AND DRAWN AT 1250° F. FOR 2½ HOURS.

Ferrite and spheroidized carbides.

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Figure 16.

X1000, nital etch.

Section "A".

HEAT-AFFECTED ZONE, SHOWING MARTENSITE.

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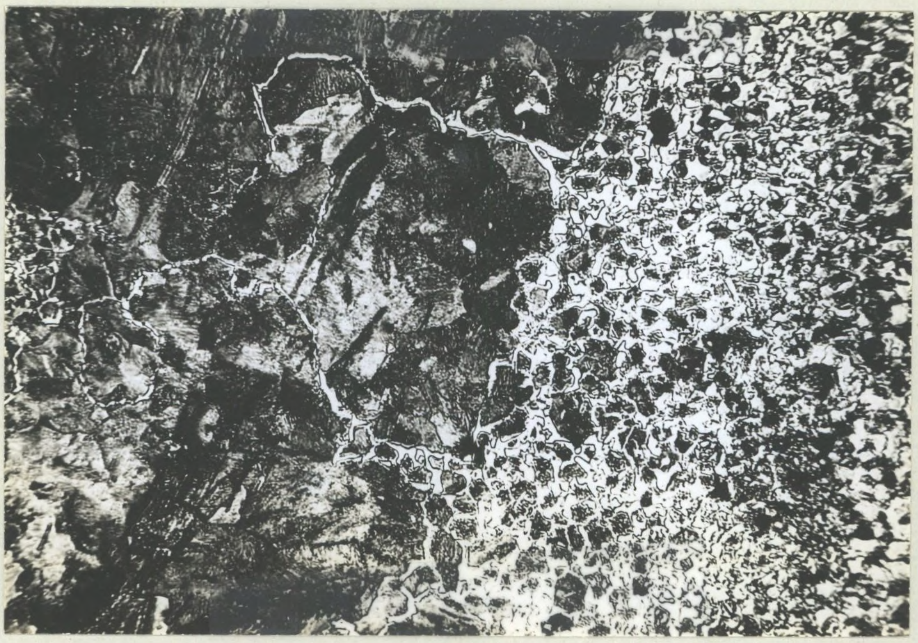
Figure 17.



X200, nital etch.

MCQUAID-EHN TEST: MIXED GRAIN SIZE.

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