THE WELDABILITY OF DUCTILE IRON

by

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THE WELDABILITY OF DUCTILE IRON

INTRODUCTION

Ductile iron is a cast ferrous product quite similar to grey cast iron in composition, yet possessing the strength and ductility of steel. This material, developed by the International Nickel Company at Bayonne, New Jersey in 1948, has received wide acclaim as being one of the most important engineering materials developed in recent years.

Greater tonnages of grey cast iron have been produced in cast form than all other metals combined, since its discovery by the Western civilized world in Germany early in the fourteenth century. The first use of this material was for cannon and other implements of war. At the present day, it is not only widely used throughout all industry, but is also found in every household and on every farm in many forms.

Grey cast iron is regarded by the metallurgist as an iron based alloy with 2.0 to 4.0 per cent carbon, having a steel matrix containing from 0.10 to 0.90 per cent carbon in the form of cementite or iron carbide and with the remaining carbon in the form of flake graphite. The tensile strength of grey iron, which ranges from 20,000 to 60,000 psi, is far less than that of steel with a similar matrix. This is attributed to the content of platelets of flake graphite that produce internal weakness by interfering with the grain structure.

Metallurgists have long been well aware of the fact that if the free carbon in cast iron could be caused to form into nodules instead

of flakes, that the mechanical properties of the material would be vastly improved. This would be due to the greatly decreased surface area of the free carbon particles, and the strength and ductility of the matrix would not be impaired by the internal notching effect of the graphite flakes.

The International Nickel Company discovery consisted of a means of nodulizing the free carbon in cast iron by means of the ladle addition of a small amount of magnesium. It is believed that very minute particles of magnesium oxide act as nuclei of crystallization for free carbon, causing it to form as hexagonal crystalline spheroids (15, p.17). A material of great economic importance was thus produced that combined the strength and ductility of a cast steel with the reduced cost, wear resistance and damping characteristics of cast iron. This material, ductile cast iron, is also known as nodular or spheroidal iron.

One of the foremost problems concerning this new engineering material is its integration with steel or other alloys into machinery construction by means of welding. Other problems consist of production of built-up castings from simple cast component parts, that would be difficult or perhaps impossible, to cast as a whole, and the repair of broken castings or of foundry defects by means of welding.

The problems of welding ductile iron are quite complex, for it is actually a family of materials rather than a single material. Its nodules of graphite can exist in ferritic, pearlitic, austenitic, martensitic or acicular matrixes respectively, depending upon heat

treatment and composition. The heat of welding causes a rapid diffusion of carbon from the nodule into the surrounding steel matrix where it remains due to the chilling effect of heat transfer by the base metal. All possible structures found in high carbon and low alloy steels thus exist simultaneously with the nodules of graphite. The problems of welding ductile iron are then beset with the combined difficulties of welding both alloy steel and cast iron.

This investigation was initiated because of the very meager information available regarding the welding of ductile iron. Only one article had been published on this subject and it concerned the welding of a single specialized product, ductile iron pipe, and the use of but one electrode having a nickel-iron core wire. Practically no information was thus available that would permit the engineer to incorporate ductile iron into welded design, or that would provide knowledge regarding industrial repairs of the material. It was, therefore, felt that comparative information concerning all feasible processes of welding ductile iron would be of material economic importance. The object of this study is to provide such information.

REVIEW OF LITERATURE ON WELDING OF DUCTILE IRON

The paucity of information regarding welding of ductile iron was one of the factors that led to the selection of this subject for investigation. The continued increase in production of ductile iron and its constantly improved status as an economic engineering material have rendered its integration into fabrication by welding to be mandatory. There is, therefore, an immediate and urgent demand on the part of the engineer for specific information on this subject that will assist in practical application of design.

At the time of initiation of this project, only one article had been published on the welding of ductile iron. This article (24, pp.45-54) was based on experiences of the Lynchburg Foundry Company in welding ductile iron. All of the work was accomplished with 60 per cent nickel, 40 per cent iron electrodes (Ni-Rod 55). These tests were conducted on 6-inch, 8-inch and 12-inch diameter pipe, which was welded together using single vee butt joints and Ni-Rod 55 electrodes. Bursting tests provided a maximum tensile strength of 58,900 psi and ranged downward to 47,250 psi. Recommendations for welding included the following:

1. Annealing of castings prior to welding.
2. Preheating to 600 F.
3. Referritizing at 1675 F subsequent to welding.
4. Use of intermittent welds if preheating is not used.
5. Remove casting skin to 1/2-inch from the weld area (both face and root sides).

The tests described in the previous paragraph concerned but one type of electrode and one method of welding. They were, furthermore, of specialized rather than of general nature, and might not be valid for other conditions. The recommendations for welding are considered quite practical.

Since the inauguration of this study at Oregon State College, four articles have been published regarding the welding of ductile iron. This would indicate that at least four other groups are actively participating in work of this general nature, which is testimony of the increasing importance of the problem. The area embraced by the scope of this work is indeed large, and much work remains to be accomplished. Although the primary nature of the investigation conducted by each group is different, the combined results fit into a broad pattern. Fortunately, there is sufficient correlation between the research programs here and elsewhere, notwithstanding the variance in objectives, that verification of results can be established in a few instances. Information is also further provided that is of assistance in interpretation of more remotely related data.

The first of the four articles mentioned in the previous paragraph was published in the August 1953 issue of the Welding Journal (16, pp.378S-385S) and a digest of this article appeared in the February 1954 issue of Metal Progress. This article, "Welding Metallurgy of Nodular Cast Iron," was based on work at Massachusetts Institute of Technology. The project included a metallographic study of single pass manual weld deposits on the surface of 6-inch x 8-inch

x 1/2-inch thick ductile iron plates. No groove welds were made and no mechanical tests were conducted in conjunction with the project. In preparation for the study, passes were made with the manual inert gas shielded arc without use of filler rod. Bead welds were also made on ductile iron with both mild steel and 25-20 stainless steel electrodes using alternating current, and with 3.5 per cent nickel electrodes using direct current. Oxy-acetylene weld deposits were made with phosphor-bronze rod (Eutectic rod No. 14) and mild steel filler rod. In addition to these Armco Ingot Iron weld deposits were made with the atomic hydrogen process.

In the metallographic study by Hucke and Udin, four regions of interest were noted in the typical cross section of an arc weld deposit. Region one consisted of the unaffected base metal. Region two consisted of carbon nodules surrounded by complex austenite transformation products. In region three, liquid existed around each nodule but liquid interfaces from adjacent nodules had not met. Complete fusion existed in region four, providing a structure of hypo or hypereutectic white cast iron, depending on the composition of the base material.

Portions of the conclusions of Hucke and Udin (16, p.385S) are quoted as follows:

> The results of the foregoing study have failed to show the way toward a completely satisfactory structural joining method.---
>
> Any method which melts the base plate upon application of filler metal will cause a hard, continuous carbide zone and is, therefore, undesirable.--- The type of filler rod has little effect toward eliminating the carbide region

since part of the base plate becomes liquid after the filler metal has started to freeze. In addition, the problem will never be completely solved by use of preheat, posthead treatment or by juggling other welding variables such as welding speed and energy output.---

From a metallurgical point of view, the solution seems to lie with a brazing process.--- No material exists which fulfills all the qualifications, but aluminum bronze can qualify in every respect except cost.

Preheating does not solve this problem, since, in order to obtain a freezing and subsequent cooling rate slow enough to allow graphitization, the preheat temperature would have to approach the melting temperature. Such preheats would be commercially out of the question. The use of a ferritizing posthead treatment is not satisfactory either, since the carbide region graphitizes in very small nodules which have a deleterous effect.

The second of the previously-mentioned publications (34, pp.823-833), "The Arc Welding of Ferritic and Austenitic Nodular Cast Iron," by Sohn, Boam and Fisk concerns the research program at the Curtiss-Wright Corporation, Wood Ridge, New Jersey. This problem eminated from the fabrication of the center main bearing support section through which the full thrust of the Wright J-65 turbojet engine is transmitted. Ductile iron was selected as the most economical material that would provide the high temperature properties required for this purpose. An austenitic nodular cast iron containing 20 to 25 per cent nickel was developed for this application, in collaboration with the American Brake Shoe Company, with a coefficient of expansion to match the austenitic stainless steel to which it was to be connected.

The specific application of the austenitic nodular iron required a high strength fillet weld between the casting and 18-8 stainless

steel sheet. The test welds consisted of stainless steel sheet joined to nodular iron with single pass fillet welds using non-grooved joints and also of 5/16-inch thick stainless steel butt welded to 5/16-inch thick nodular iron, using double vee grooved joints. A preliminary study was made with the Aircomatic inert gas-shielded arc, using No. 61 Nickel (97 per cent nickel, 2½ per cent titanium), "A" Nickel (commercially pure nickel), Ni-Rod 55, Inconel No. 62, AISI 307, Airco No. 100 (90 per cent copper, 10 per cent aluminum) and Ambraloy 928 (92 per cent copper, 8 per cent aluminum electrodes). The aluminum bronze welds were the only ones that did not develop cracks during the welding process. Globules of white iron trapped in these welds developed cracks when heated to anticipated service conditions of 1,000 F. All further investigation was, therefore, conducted with the manual flux-coated metallic arc process.

Manual arc welded specimens prepared as previously described with AISI-310, 312, 316, 330, and 347 stainless steel electrodes developed welding cracks and were unsatisfactory. Similar tests made with AISI-430, 431, 442, and 446 electrodes provided improved results. The best results obtained were with AISI-431 electrodes (produced by coating AISI-430 wire with flux containing 2.0 per cent nickel). Ultimate strengths up to 62,500 psi obtained on 1/4-inch diameter tensile test bars of butt welds were considered quite satisfactory, and the previously-mentioned bearing support sections are now in production.

The welding study was continued with low alloy sheet steel welded to ferritic nodular iron. The object in view was a possible material modification which would result in a saving of 34 pounds of nickel per finished part. The nodular iron was similar to that studied here and the alloy steel was NAX, which has the following composition and properties:

Composition and Properties of NAX Steel

Carbon	0.10 - 0.17 per cent
Manganese	0.50 - 0.80 per cent
Silicon	0.60 - 0.90 per cent
Chromium	0.50 - 0.75 per cent
Zirconium	0.15 - 0.50 per cent
Molybdenum	0.15 max.
Sulfur	0.04 max.
Phosphorus	0.14 max.
Ultimate tensile strength, psi	79,700
Yield strength 0.2% offset, psi ...	59,500
Per cent elongation in 2 inches ...	25

Welds on the above materials with AISI-431 electrodes, which provided excellent results with austenitic nodular iron, were not satisfactory. As a result, tests were made with AISI-310, AWS-7016 and Ni-Rod 55 electrodes. A post-weld stress-relieving heat treatment had been found necessary on the austenitic nodular iron weldments and for this reason, all the new test specimens were heated from one to four hours at temperatures ranging between 850 F and 1200 F.

In the tests just described, welds with low hydrogen electrodes produced ultimate tensile strengths from 50,500 psi to 69,000 psi. The tests considered most satisfactory were made with AISI-310 electrodes and provided from 57,300 psi to 67,000 psi. Welds with

Ni-Rod 55 ranged from 36,200 psi to 61,000 psi and it was concluded that these electrodes would be satisfactory only for repair of minor casting defects.

The investigation by Sohn, Boam and Fisk is of particular interest as it supplements the study at this school. It indicates the results of welding ductile iron to steel with different types of electrodes and of welding austenitic ductile iron. When ductile iron is welded to itself more complex metallurgical problems are involved than when ductile iron is welded to steel. It is logical to anticipate higher physical values and more uniform results when welding relatively thin material than when welding thick material with multipass joints.

The third article on welding ductile iron (18, pp.947-956), "Joining of Ductile Iron by Several Arc Welding Methods," primarily concerned manual metallic arc welding with Ni-Rod 55 electrodes. Some information was provided about inert gas shielded metallic arc and inert gas shielded tungsten arc welding processes with nickel-iron filler rod, as well as submerged arc welding. Such factors as operating procedure, effect of preheat and postheat, position operability, electrode manipulation, etc. were also discussed.

The project just mentioned is conducted by the International Nickel Company under the direction of Kihlgren and Waugh. Welds with Ni-Rod 55 electrodes were made on both as-cast and annealed base metal and tests conducted both in the as-welded and post-weld heat-treated conditions produced values between 59,000 psi and 65,000 psi.

Both inert gas shielded arc processes produced results comparable with those reported in the previous paragraph, although greater interest was expressed in the consumable electrode type. The reason for this was not stated. No values were given for submerged arc welds. The only welds of acceptable nature were obtained with type AISI-310 electrodes when using this process.

The most recent account of welding ductile iron (26, p.49) was a brief one-page editorial concerning repair of blowholes and shrinkage cavities in defective castings. The welding technique was developed at North American Aviation, Inc., Los Angeles, and consisted of carbon arc welding. The carbon electrode was substituted for a tungsten electrode in a heliarc torch and an argon gas atmosphere was used. They reported hard deposits with use of heliarc welding and tungsten electrodes as the reason for using carbon electrodes. They use ductile iron filler rod with cast iron flux and a high degree of preheat. It was concluded that the carbon arc process is the most promising means of salvaging defective castings.

THE COMPOSITION OF DUCTILE IRON

One of the primary considerations of ductile iron is its composition. Seemingly small changes in composition can result in a wide variation of such properties as strength, structure, shock resistance, machinability, etc. The composition of the base metal was found to have a marked effect on its response to welding by different processes. Since the composition should be taken into account when considering the problems of welding, the discussion that follows is intended to provide specific information in this regard.

The chemical composition limits of ductile cast iron are given in Table 1 (11, p.103). The essential difference in composition between grey cast iron and ductile iron is the small amount of magnesium or cerium that ductile iron contains. Although magnesium is more commonly used, cerium will also cause graphite to form in spheroids rather than in flakes as in grey cast iron.

It is to be noted that Table 1 does not include the austenitic ductile iron castings which are produced for use under conditions of high temperature and corrosion. The problems concerned with welding high chromium-nickel austenitic castings are not necessarily common to those of ferritic-pearlitic structure, and they are, therefore, not included in the scope of this investigation. It would appear likely that in the near future, many alloy grades of ductile cast iron will be developed which will compare with various grades of alloy steel.

In addition to the magnesium content, both carbon and silicon are slightly higher in ductile iron than in grey iron. The carbon

TABLE 1

COMPOSITION RANGE OF DUCTILE IRON

	C	Si	Mn	P(max)	Ni	Mg
Broad-range, per cent	3.2-4.2	1.0-4.0	0.1-0.8	0.10	0 -3.5	0.05-0.10
Ferritic high-ductility, per cent	3.6-4.2	1.25-2.0	0.35	0.08	0 -1.0	0.05-0.08
Ferritic high-strength, per cent	3.4-3.8	2.25-3.25	0.35	0.10	0 -1.0	0.05-0.08
Pearlitic high-strength, per cent	3.2-3.8	2.25-2.75	0.6-0.8	0.10	1.5-3.5	0.05-0.08

equivalent in ductile iron (carbon plus 1/3 silicon) is normally between 4.3 and 4.6 per cent. The silicon content of grey cast iron is limited due to its coarsening effect on flake graphite. On the other hand, no effect of silicon was noted on the size or distribution of carbon nodules in the castings used in this study, which ranged from 2.63 per cent silicon to over 4.22 per cent silicon. Table 1 indicates strengthening of ductile iron by solid solution of silicon in ferrite, which is accompanied by decreased ductility and impact resistance. Tensile tests made during this study of annealed ductile iron from various heats indicated an increase in strength of over 10,000 psi for each additional per cent of silicon content.

A 3/8-inch square bar was removed from a 1/2-inch "Y" block poured from heat No. 66 of the Holt Equipment Company, November 3, 1953. The chemical analysis of this heat was 2.70 carbon, 2.98 silicon, 0.02 sulfur, 0.05 phosphorus and 0.30 manganese. The end of this bar was hand forged at about 1800 F and then curled while hot as shown in Figure 1. The center of the bar was then hot twisted as also shown at about 1600 F. This is evidence of the good hot workability of the material. A bar of similar dimensions was obtained from heat D-4102 produced by the Eagle Foundry Company, which contained 4.22 silicon, 0.032 sulfur, 0.048 per cent phosphorus and 0.39 manganese. Although this bar could be hot twisted to the same extent as the one in Figure 1, it was quite hot short with regard to attempts to hot forge it. This characteristic was attributed to the increase of silicon content.

Castings obtained from the Holt Equipment Company during the summer and early fall of 1953 contained less than 0.05 per cent phosphorus, as Swedish pig iron low in this element was used in the charge. The charge consisted of 10 per cent pig iron and the

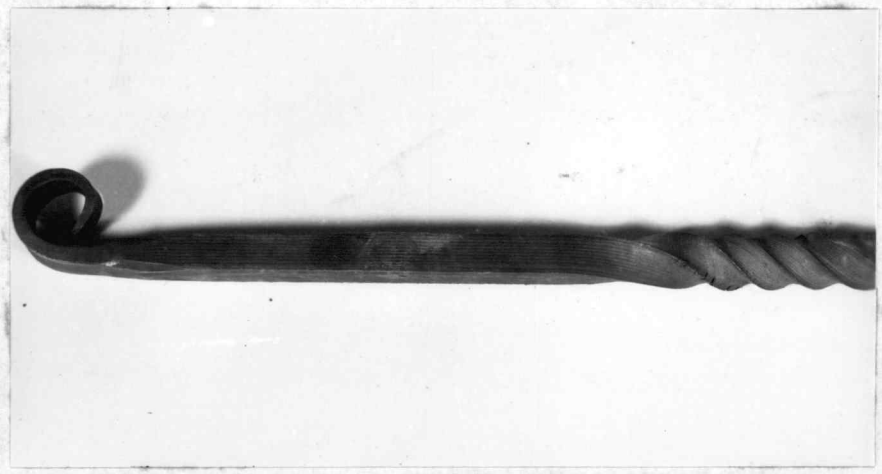

Figure 1. Ductile Cast Iron Bar that has been Forged and also Hot Twisted.

remainder was steel scrap and foundry returns. The use of domestically produced pig iron later in the fall of 1953, which had a high content of phosphorus (0.24 per cent phosphorus), led to an increased amount of this element in the ductile iron castings. Foundry returns caused a gradual increase in phosphorus until it was well over the maximum allowable of 0.10 per cent, and in late February, 1954, it was necessary to lower the bottom of the cupola in order to increase the slag level and provide more opportunity for removal of phosphorus. In the basic process, such as used by the Holt Equipment Company, phosphorus is oxidized (2, p.72), forming P_2O_5 which reacts with the lime slag to form calcium phosphate. The phosphate thus produced

becomes part of the slag and is poured off.

Annealed castings which contained less than 0.05 per cent phosphorus produced an average tensile strength of 66,360 psi with 21.7 per cent elongation. When the phosphorus was increased to 0.086 per cent, with very little change in the remaining composition, strength decreased to 58,900 psi and the ductility to 6.62 per cent elongation. This adverse effect on the mechanical properties was attributed to a very fine network of iron phosphide eutectic and to the tendency of phosphorus to stabilize pearlite.

Castings obtained from the Eagle Foundry Company were all from two heats, No. D-4090 and D-4102 respectively. Heat D-4090, poured February 25, 1954, consisted of 30 "Y" block castings that provided test sections 2½ inches wide, ½ inch thick and 14 inches long. Heat D-4102 consisted of 52 castings of similar dimensions, which were poured March 4, 1954. The analysis of these two heats is given in Table 2.

Sulfur is always very low in ductile iron and the analysis given in Table 2 is quite typical regarding this element. In grey cast iron, sulfur increases carbide stability and inhibits graphitization if the manganese content is too low (2, p.365). Manganese forms the relatively insoluble manganese sulphide and manganese silicate that exist in the form of less harmful inclusions in the casting. Due to the low sulfur content of ductile iron, the manganese is proportionally lower. If the manganese content of ductile iron is above 0.50 per cent, it stabilizes pearlite to such an extent that a completely

ferritic structure cannot be obtained by heat treatment.

TABLE 2

INFORMATION REGARDING EAGLE FOUNDRY COMPANY CASTINGS

Composition	D-4094	D-4102
Magnesium	0.084 per cent	0.056 per cent
Sulfur	0.018	0.032
Manganese	0.41	0.39
Phosphorus	0.048	0.048
Silicon	3.70	4.22
Nickel	0.74	0.79
Chromium	0.05	0.05
Vanadium	0.12	0.16
Mechanical Properties		
Yield strength at 0.1% offset, psi	71,800	71,505
Ultimate tensile strength, psi	84,750	88,995
Per cent elongation in 2 inches	13.0	15.5

The chromium and vanadium content in Table 2 does not appear high. However, these elements are such strong stabilizers that, when lower silicon contents were used, the amounts indicated were sufficient to cause retention of substantial amounts of massive carbides and fine pearlite, even after annealing five hours at 1650 F followed by cooling at a maximum rate of 40 F/hour from 1500 F through 1200 F.

Castings provided by the Holt Equipment Company were samples from numerous heats representing several months' production. This company is not equipped for either physical or chemical analysis of

their material; however, samples are occasionally sent out to commercial testing laboratories. Table 3 is intended to provide information typical of the castings which they provided.

TABLE 3

TYPICAL COMPOSITIONS OF HOLT EQUIPMENT COMPANY CASTINGS

Element	September 15, 1953	November 3, 1953	December 15, 1953
Carbon	2.80	2.70	3.20
Silicon	2.91	2.98	2.63
Sulfur	0.017	0.02	0.02
Phosphorus	0.078	0.05	0.086
Manganese	0.38	0.30	----
Magnesium	0.049	----	0.067

It was previously stated that the carbon equivalent of ductile iron should be between 4.3 and 4.6 per cent. The carbon equivalents of the typical heats in Table 3 are 3.77, 3.69 and 4.08 per cent respectively. Carbon itself has very little effect on the mechanical properties of annealed ductile iron. Foundry characteristics such as chilling propensity, feeding requirements and fluidity are affected, however, and many of the Holt Equipment Company castings were rejected on the basis of radiographic examination as being unfit for use in this investigation.

The as-cast ductility is reduced when the carbon equivalent is low, as evidenced by an average elongation of 1.2 per cent for the castings in Table 3 compared with 3.1 per cent for those in Table 2. It was also noted that the machinability of the castings in Table 2 was much superior to that of the castings in Table 3. Lowering the

furnace bottom at the Holt Equipment Company adjusted the carbon equivalent to its proper condition in addition to reducing the phosphorus content as was previously mentioned.

PRODUCTION OF DUCTILE IRON

The overall foundry procedure is perhaps one of equal importance to composition if satisfactory quality of ductile iron is obtained. The weldability of the material is directly related to its quality. After the composition has been established by means of the charge, the next consideration is the selection of the most suitable foundry technique.

The foundry procedure for producing nodular graphite in cast iron involves two steps. The first step consists of a ladle addition of magnesium in such a manner that a small but effective portion is retained in the iron. The second step consists of innoculating with a strong graphitizing agent such as ferro-silicon. The first effect of the magnesium addition is to reduce the sulfur content of the iron to 0.02 per cent or less, and magnesium cannot be retained in the iron until the sulfur is at this level. Magnesium volatilizes at about 2100 F, so a sufficient amount must be added to allow for sulfur removal, volatilization and oxidation, and yet permit retention of 0.045 to 0.10 per cent magnesium in the iron.

Several alloys are available for this purpose (22, p.1), the compositions of which are given in Table 4.

An exceedingly violent reaction takes place when pure magnesium encounters molten iron. For this reason, an alloy containing not over 20 per cent magnesium is used as an additive, for protection of the foundry personnel. A sheet iron shield is used by the Holt Equipment Company, as shown in Figure 2, to surround the ladle during the

Figure 2. Two Views of Magnesium Addition to Ductile Iron.

magnesium addition in order to protect the workmen from the violent reaction that takes place.

TABLE 4

COMPOSITION OF MAGNESIUM ALLOY

	Mg	Ni	Si	Cu	Fe
Inco No. 1	15-20	75-80	---	---	Balance
Inco No. 2	15-20	40-50	25-30	---	"
Inco No. 3	15-20	---	60-65	---	"
Vanadium No. 12	12	---	40	16	"
Electro-Met	8	---	46	---	"

Electro-Met alloy is used by the Holt Equipment Company to provide magnesium addition in the ratio of 15 pounds per 500-pound ladle of molten iron. This three per cent addition provides a magnesium recovery in the iron of 25 to 40 per cent of the amount added at 2500 F. Foundries that tap cupola iron at higher temperatures are inclined to use Inco No. 1 or No. 2 or Vanadium No. 12 as their greater density provides higher recovery of magnesium at more elevated temperatures. The cast iron is first tapped into a holding ladle which is attached to a scale as shown in Figure 3 in order to provide an accurately known weight of molten metal. The magnesium alloy is then placed in a second ladle and the cast iron is poured on top of it to provide bottom introduction of alloy, as shown in Figure 4, which prevents undue volatilization of magnesium. An alloy particle size of about 3/4-inch diameter has been found most suitable for treatment of iron in this amount, while for larger ladle additions of 1,000 pounds or more, particle size of one to three inches provides better magnesium recovery.

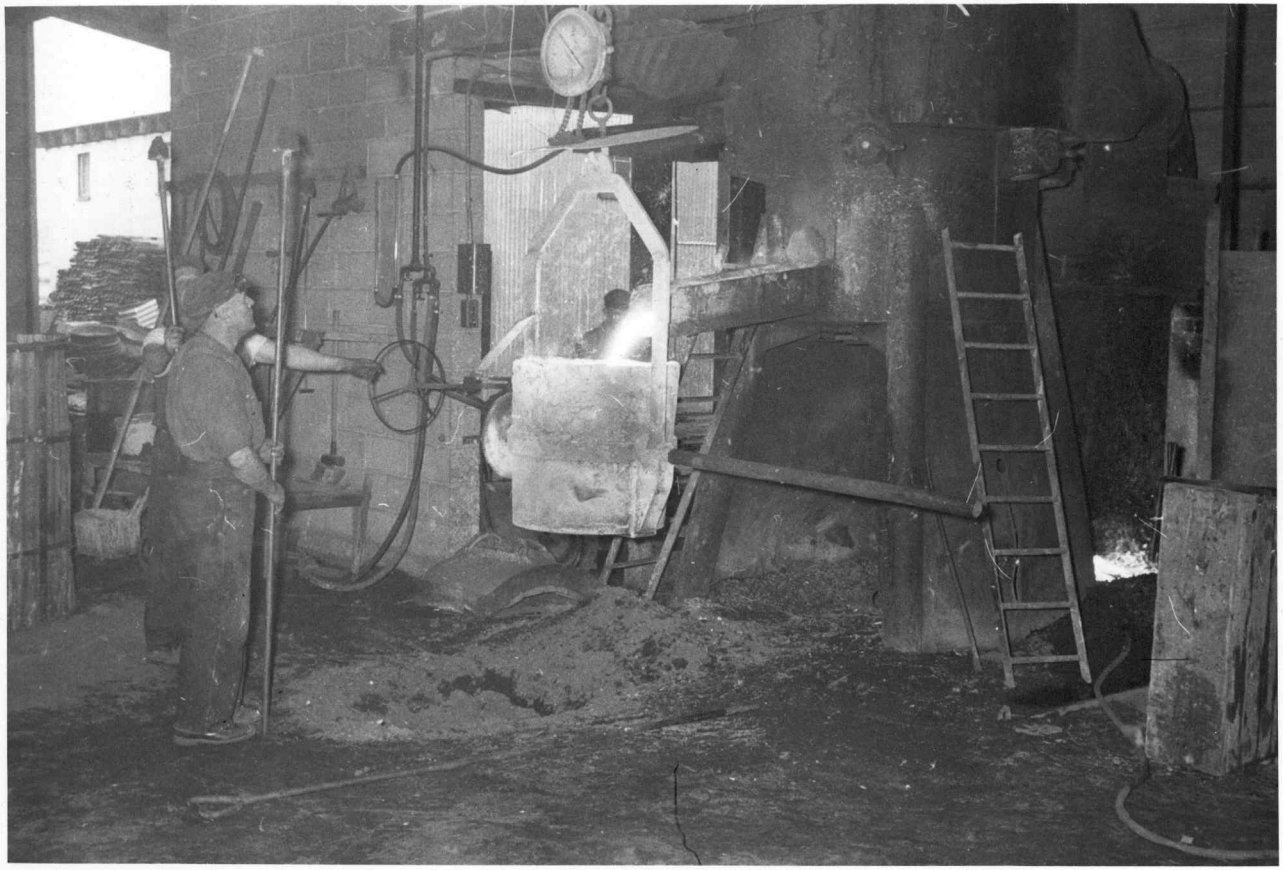

Figure 3. Tapping of Ductile Iron from Cupola.

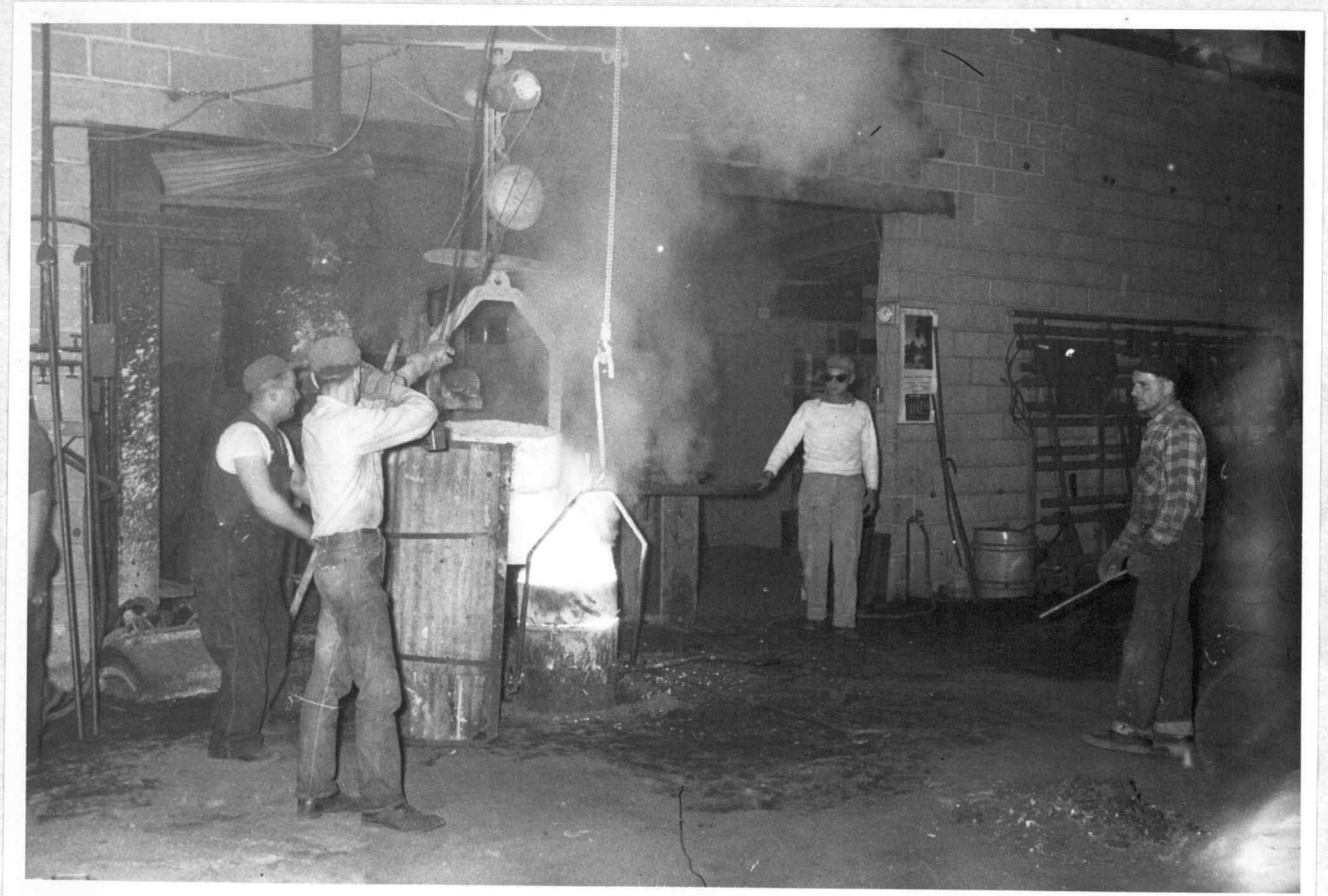

Figure 4. Bottom Introduction of Magnesium Alloy into Ductile Iron.

The magnesium addition provides a strong chilling action on the molten iron which will produce free carbides in the structure of light section unless care and precision are exercised in the subsequent innoculation. Ferro-silicon containing at least 72 per cent silicon with a particle size of about 3/8-inch diameter is added to the molten stream of iron during reladling to provide uniform innoculation. The ferro-silicon which is being dissolved increases the silicon content to such an extent that a super-saturated condition is provided with respect to carbon. An advantage of using the Electro-Met alloy is that reladling innoculation is not required due to its silicon content.

A strong graphitizing condition thus exists which causes precipitation of small nucleating particles of graphite. The work of Gries and Manshake (15, p.16) substantiates the theory that spheroidal nucleation of carbon is initiated about sub-microscopic particles of magnesium oxide and that the carbon nodules grow chiefly above the main stems of, and within, the interstices of primary iron dendrites during solidification.

Spheroidal graphite in ductile iron has a radial polycrystalline appearance, and can be readily distinguished from the flake graphite aggregate intermixed with iron particles that exist as nodules in American blackheart malleable cast iron. The latter are more irregular in shape and a crystalline structure cannot be distinguished under the microscope. According to M. Nacken and E. Piwowarsky (25, p.75), Debye-Scherrer patterns indicate a hexagonal crystal structure for the

carbon spheroids in magnesium treated cast iron.

Certain elements have a detrimental effect on the magnesium addition and tend to prevent formation of nodular graphite. Lead is particularly damaging in this regard, and 0.002 per cent is the maximum amount that can be permitted in ductile iron. Possible sources of lead would be babbit bearings in scrap engine blocks or fluorspar which is frequently used as a flux in the basic cupola or electric furnace. Other elements that produce a similar effect include aluminum, antimony, arsenic, bismuth, selenium, tellurium, tin, titanium and copper. Aluminum and titanium have a stronger effect on producing flake graphite in thick sections than in thin sections. The amount of each of the above elements that can be safely tolerated, of course, varies.

There are a total of 90 companies licensed to produce ductile iron in the United States and Canada, of which six produce complex castings that are radiographically sound. In addition, there are an equivalent number of licensed foundries distributed throughout the rest of the world. The Holt Equipment Company is one of 23 ductile iron foundries in the United States and Canada that operate a basic cupola, and is also one of the 14 basic cupolas that employ the use of rear slagging. The Holt Equipment Company charge consists of 30 per cent steel scrap, 40 per cent foundry returns and 10 per cent high phosphorus pig iron (0.24 per cent P.). A workman is shown in Figure 5 loading this charge into the basic cupola skip hoist at the Holt Equipment Company. Figure 6 shows the pouring of ductile iron

castings, and Figure 7 is a view of the foundry immediately following the pouring of the castings at the Holt Equipment Company.

Figure 5. Charging the Basic Cupola.

The basic cupola is ideal for production of low sulfur, low silicon, high carbon iron. Iron is not oxidized by this process, and the low sulfur content requires use of less magnesium alloy for treatment. The Electro-Met Alloy is ideal for use in basic practice and provides good as-cast properties. The metal is normally about 100 degrees hotter when tapped from the basic cupola than from the acid cupola. It is very important that a good grade of limestone be used to flux out dirt, sand and other impurities and to reduce the sulfur and phosphorus content.

Figure 6. Pouring Ductile Iron Castings.

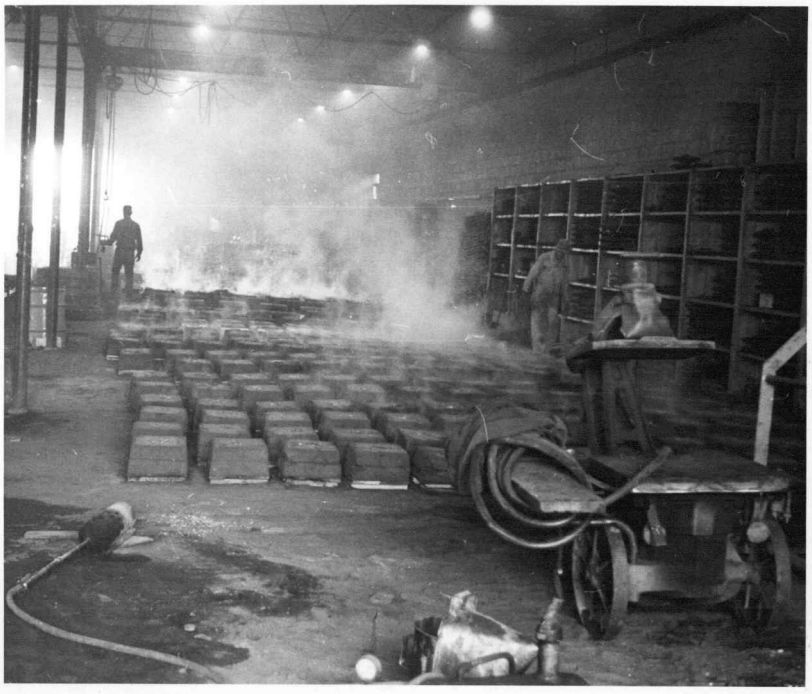

Figure 7. The Holt Equipment Company Foundry.

The Eagle Foundry Company uses an acid-lined cupola and, therefore, must be quite careful in selecting the materials that compose the charge. The charge consists of 34 per cent steel scrap and the remainder is Chilean pig iron and foundry returns. The Chilean pig iron contains 0.03 to 0.04 per cent phosphorus, 0.023 per cent sulfur, 0.37 per cent manganese, 1.30 per cent silicon and 0.23 per cent chromium. Sulfur and phosphorus cannot be removed in the acid process and, therefore, must be kept low in the charge.

The Oregon State College Foundry poured their initial heat of ductile iron May 16, 1954, which was produced in the Green electric furnace that was recently installed. This furnace, manufactured by the Green Electric Furnace Company of Seattle, Washington, is a three-phase, 300 KVA direct arc furnace with a rated capacity of 400 pounds per hour. The charge for this heat was 100 per cent steel plate scrap, consisting of 400 pounds of steel, 17½ pounds of No. 4 mesh Mexican graphite, 8 pounds of silicon briquettes, 2½ pounds of zirconium-iron-silicon and 8 pounds of limestone. Liquid iron was tapped at 2925 F and the reladling treatment was made with 12 pounds of Electro-Met magnesium alloy. Iron was poured into the molds at 2400 F, producing gear blanks, caps for a hydraulic cylinder and 1/2-inch "Y" blocks. Two of the 7¼-inch diameter caps, showing foundry defects repaired by carbon arc welding are in Figure 8. One of the 8-inch diameter gear blanks is shown in Figure 9.

Other types of furnaces used to produce ductile iron include indirect arc furnaces, induction furnaces, crucibles, air furnaces and

Figure 8. Ductile Iron Caps for Hydraulic Cylinder. Repair of Foundry Defects by Carbon Arc Welding is Shown.

Figure 9. Ductile Iron Gear Blanks, 8 inches in Diameter.

oil and coal fired rotary furnaces. The type of melting unit appears to have little effect on the quality of ductile iron that is produced. Batch-type furnaces such as used at Oregon State College have a tendency to carburize poorly. The fracture of a chill block during the heat indicated that there was not sufficient carbon provided by the 16 pounds of graphite in the initial charge, so the additional $1\frac{1}{2}$ pounds were then added. A subsequent chill test, with a chill about 3/8-inch deep, was considered adequate. The initial carbon and silicon was placed in the bottom of the furnace, underneath the steel, to assure complete solution in the iron. Any undissolved graphite in magnesium-treated iron will develop a mixed graphite structure consisting of flakes and spheroids.

THE PROPERTIES AND USES OF DUCTILE IRON

In order to determine the weldability of any metal or alloy, it is first necessary to be thoroughly versed in a knowledge of the properties of the material itself. While it is always desired to produce a weld with properties that match the parent metal as closely as possible, the end use of the article must also be considered. Cost may be of little consequence for one product that requires the highest physical properties, while the requirements of another product might be such that lower mechanical properties are adequate, and a less expensive welding process would suffice.

Although several specifications have been used for the purchase of ductile iron castings, only two are included in the ASTM Specifications for Nodular Iron Castings. These are contained in ASTM Designation A339-51T (3, p.1276) and are given in Table 5. The intent of these specifications is to subordinate chemical composition to physical properties; however, composition can be also specified if so desired. In addition to tensile properties, the ASTM specifications include both keel block and "Y" block test coupons, type of tensile specimens, number of tests, heat treatment, workmanship and finish, inspection and certification.

In making a study of the machinability of various grades of ductile iron, compared with various grades of grey cast iron, Metcut Research Associate (12, pp.108-114) found the tensile properties of the weakest ductile iron to be far superior to the best grade of grey iron, as shown in Table 6. The grey iron test coupons were taken

TABLE 5

TENSILE REQUIREMENTS FOR DUCTILE IRON

	Grade (80-60-03)	Grade (60-45-10)
Tensile strength, min., psi.	80,000	60,000
Yield strength, min., psi.	60,000	45,000
Elongation in 2 inch, min., %	3.0	10

from 1.2-inch arbitration bars. The ductile iron coupons were one-inch keel bars cast in dry sand. All coupons were machined to standard round ASTM E-8 test specimens with 2-inch gage lengths, and yield strength was measured at 0.2 per cent offset. It is of interest to note that the values for elongation and reduction of area are essentially the same in all cases.

TABLE 6

PROPERTIES OF IRONS TESTED

		Tensile Properties			
Mark	Brinell Hardness	Ultimate Strength (in psi)	Yield Strength (in psi)	Elongation (% in 2 in.)	R. of A. (in %)
		Flake Graphite Cast Iron			
Acicular	263	59,700			
Fine	225	45,000			
Coarse	195	35,000			
Ferrite	100	15,700			
		Spheroidal Graphite Cast Iron			
2%	265	97,250	78,000	2.0	1.5
4%	215	93,000	72,000	4.0	3.5
17%	207	84,700	69,800	17.5	16.5
20%	183	77,100	62,000	20.0	21.0
22%	170	70,000	56,000	22.0	22.0

The compressive strength of ductile iron appears to vary directly with hardness. The yield strength ranges from 57,000 psi at a hardness of 219 Brinell to 95,000 psi at a hardness of 321 Brinell. It is the opinion of the International Nickel Company that the information provided by their tests (12, p.110), Table 7, is low for the specimens bowed after reaching the yield point and eventually failed in shear.

TABLE 7

COMPRESSIVE PROPERTIES OF DUCTILE IRON

Mark	Condition	Yield* Strength psi	Compressive Strength psi	Brinell Hardness	Tensile Strength psi
A	as-cast	94,000	198,000	347	77,600
B	as-cast	88,400	193,000	306	92,300
C	as-cast	76,600	174,000	277	108,700
D	as-cast	57,400	125,600	219	80,500
E	as-cast	95,600	178,800	321	123,000
F	annealed	66,000	122,800	195	72,300

*Offset not stated.

The Cooper-Bessemer Company investigated the impact properties of ductile iron, using the 0.798-inch diameter Izod test bar which is used extensively in England (8, pp.72-82). Annealed ductile iron absorbed over 120 foot-pounds compared with 50 to 90 foot-pounds for the as-cast condition. These values are comparatively quite high as highly alloyed acicular iron absorbed 20-30 foot-pounds and class 30 grey iron absorbed only from 10 to 14 foot-pounds of energy.

Fatigue resistance of ductile iron is similar to that of ordinary grades of steel in both the notched and unnotched condition (10, pp.128-132). This property combined with a relatively high

constant modulus of 25 million psi and excellent wear resistance makes it an excellent material for crank shafts, cam shafts, connecting rods and other parts subject to stress reversal and wear. The good modulus of elasticity and relative low cost makes ductile iron a choice material for rolls for the paper, wrought metal and rubber industries.

One of the outstanding properties of ductile iron is its resistance to wear. A ductile iron rack and a forged manganese steel rack were examined after 1500 hours' service at 50 per cent overload (10, pp.128-131). The steel rack was pitted and badly worn, while the ductile iron rack was only burnished and in excellent condition. In a gear application involving shock loading, the teeth of grey iron gears broke after two weeks' service. After four months, steel gears were worn out, while ductile iron gears were still providing good service after 8 months of usage. Ductile iron bearings provided five times the service life of expensive bronze bearings for the rollout table of a steel mill (10, pp.128-131).

A surface hardness of up to 600 Brinell can be obtained by either flame or induction hardening of ductile iron. In this condition its resistance to wear and seizing are superior to any other material used for certain die applications (11, pp.101-108). One objection to many abrasion materials is their shock sensitivity. A carbidic surface chilled ductile iron with shock resistant core is now extensively used for earth-digging equipment, ore-chute liners and sand and gravel equipment. Scarifier teeth for earth rooting operations are currently being produced by the Holt Equipment Company. Ductile iron is now

used for many parts requiring a high degree of heat resistance such as furnace parts and pot liners. The Holt Equipment Company recently produced a number of flat 3/8-inch thick ductile cast iron plates, 6 feet long and 3 feet wide, that were rolled into cylinders by a Portland firm for use as furnace liners.

An interesting application of ductile iron where severe shock conditions are encountered is shown in Figure 10. The ductile iron power arms that connect the hydraulic pistons with the grader blade are now replacing more expensive steel castings. The tractor operator has lifted the front end of the tractor free of the ground by means of the grader blade, which is indicative of the terrific strain imposed on the power arms in service.

Figure 11 shows a ductile cast iron fitting for irrigation equipment produced by the Holt Equipment Company. This casting is used to replace aluminum castings at a reduced cost, and ductile iron was selected because of its property of corrosion resistance and ability to withstand impact.

A grader blade connection bracket is now used by the Holt Equipment Company, Figure 12, to replace a bracket fabricated of welded steel which is also shown. The company records indicate a substantial savings in cost effected by use of the ductile iron bracket, and field service tests show that itis performing in a commendable manner.

There is apparently no limit regarding the size of ductile iron castings that can be produced. The Acme Shear Company of Bridgeport, Connecticut, is producing ductile iron electrical hardware fittings

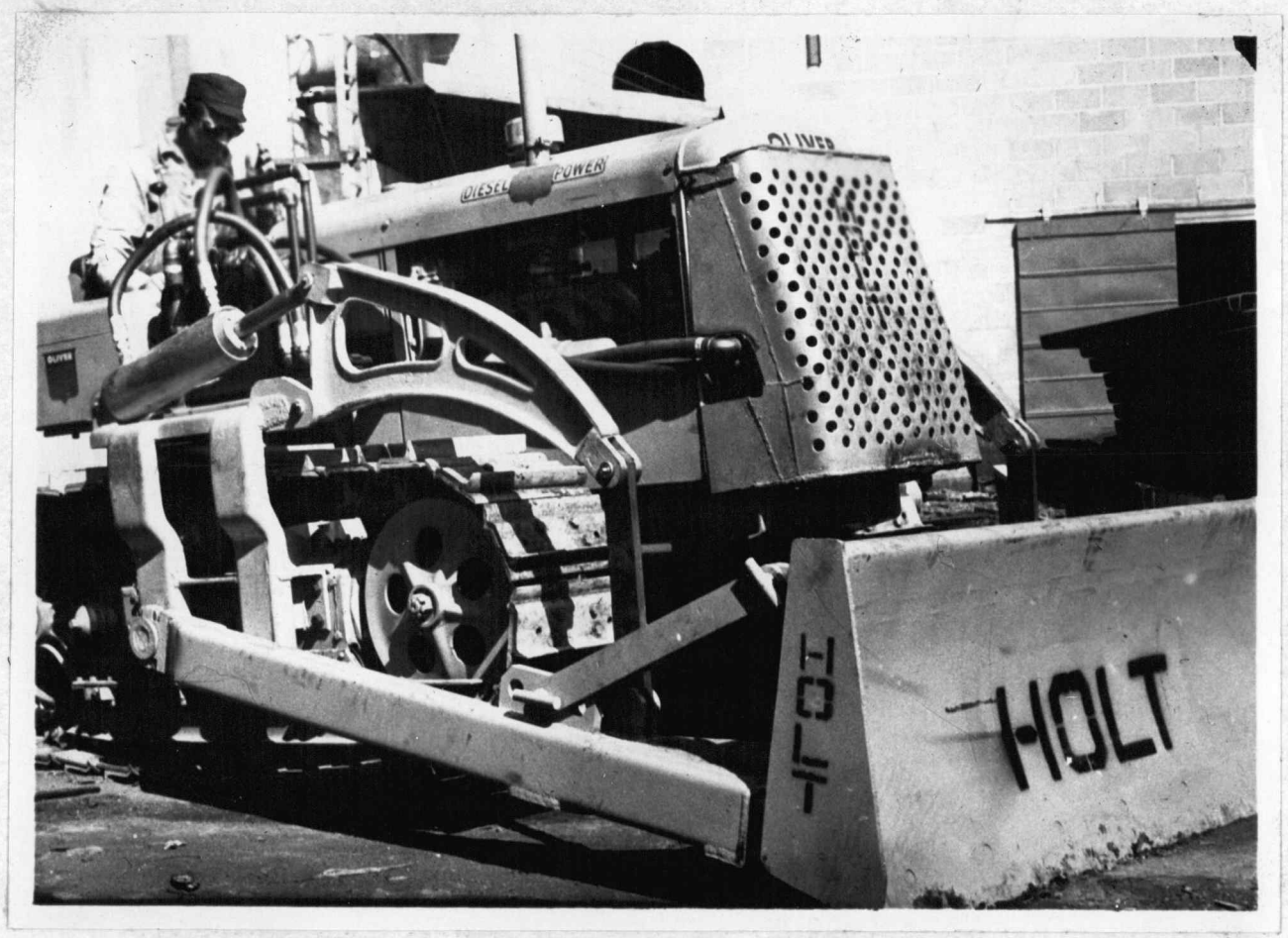

Figure 10. Ductile Cast Iron Power Arm in Operation.

Figure 11. Ductile Iron Irrigation Fitting.

Figure 12. Ductile Iron Grader Blade Connection Brackets, Left and Center.

weighing less than 1/3-ounce each, in production runs of hundreds of thousands of castings (23, pp.79-82). In contrast with these castings, the Chambersburg Engineering Company has produced forging hammer anvils of cast ductile iron, weighing 60,000 pounds each.

MATERIAL USED IN THE STUDY

The preceding discussion was of rather a general nature, intended to provide background information that would be of material assistance in conducting the investigation. The following sections pertain directly to the study itself, and will enable those who are interested in a continuation of this work to proceed in an efficient manner without duplication of effort.

The initial castings used for tensile tests consisted of ductile iron bars 14 inches long, 2 inches wide and 3/8-inch thick. These bars had a reduced section 1-1/2 inches wide, and 3 inches long at the center in order to provide a minimum of machining operations on the welds. The shape of these bars was such that adequate feeding did not take place during solidification, and as a result, extensive microscopic porosity occurred. This porosity, known as center-line or interdendritic shrinkage, is also prevalent if lower pouring temperatures are used.

Although the tensile bars just described were used in a preliminary study of ductile iron welded to steel with nickel-iron electrodes, the test results were of such variable nature, that it was deemed necessary to utilize material of greater uniformity in order to obtain higher physical values, and more consistent results. It was felt that these qualities might be found in flat rectangular plates of larger area, and without any restricted sections. Plates having dimensions of 8 inches by 14 inches were then cast in thicknesses of 1/2 inch, 3/4 inch and 1 inch as shown in Figure 13. These plates

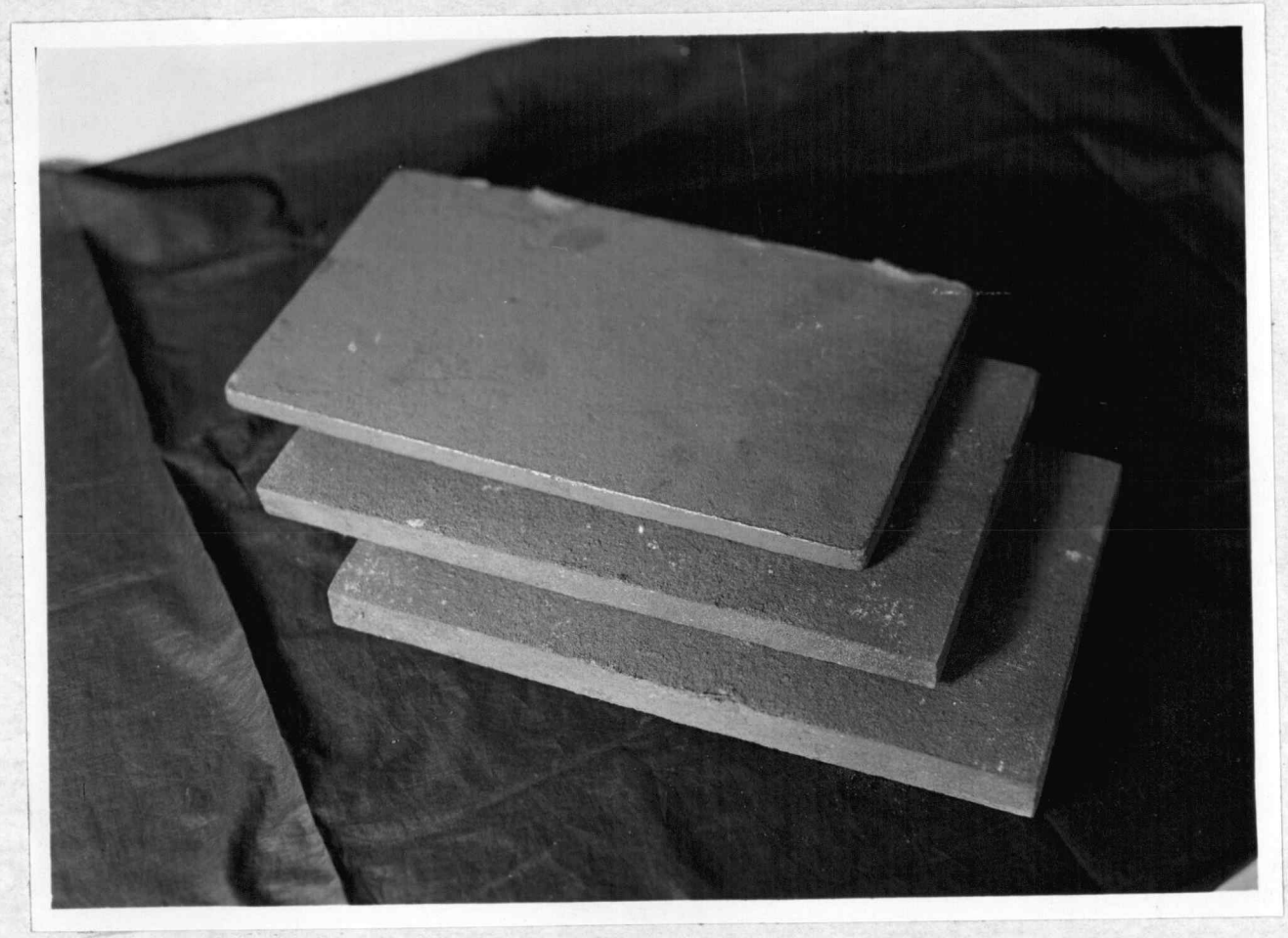

Figure 13. Ductile Iron Plates 14 in. Long and 8 in. Wide.

were radiographed to determine the condition of soundness and all were found to contain extensive areas of interdendritic shrinkage. In addition, the thinner plates contained numerous areas of closely spaced gas pockets close to the surface of the casting, known as pinholes.

A section was removed from one of the 1/2-inch plates in an area of excessive pinholing for microscopic and macroscopic examination. Pinholes can be observed in a portion of this section, Figure 14, which also shows the surface of a ductile iron washer produced by the Holt Equipment Company. The surface of the washer was ground to a sufficient depth adjacent to the hole on one side to disclose pinholes which were detected by radiography. Due to a difference in light reflection, the pinholes in the washer appear white, while those in the section of 1/2-inch thick plate appear dark.

Pinholes all appear to have small openings to the surface. In many instances, they can be detected by careful visual inspection of the surface. As a rule, the pinholes were detected at the most remote distances from the ingates of commercial castings; however, in the case of the flat plates, they were adjacent to the ingates. Moulding sand having less moisture content and greater permeability tends to produce a minimum of pinholes.

If pinholes of as-cast ductile iron are carefully examined, they appear to have a high carbon coating. Oxidation of this carbon during the annealing process produced an enlargement of the holes. It was noted that castings produced by the Eagle Foundry Company contained less pinholes than similar castings of the Holt Equipment Company.

Figure 14. Pinholes in Surface of Ductile Iron Washer (white) and Section of 1/2-inch Plate (dark).

The Eagle Foundry Company uses larger amounts of graphitizing agents and produces a minimum of carbidic structure in the as-cast condition in comparison with the Holt Equipment Company. Although there is no published information regarding the cause of pinholes, it would appear likely that complex iron-magnesium carbides in the molten iron might react with moisture in the sand to produce acetylene gas which is entrapped in the iron as it approaches solidification. The reasons for this belief are as follows:

1. The odor of acetylene gas can be readily detected in all freshly machined or fractured ductile iron castings.

2. Pinholing is less prevalent in sands of low moisture content.

3. Pinholing is less prevalent in molds of more permeable sand that provides for escape of gases.

4. Pinholes are connected with the surface of the castings.

5. An increased amount of free carbides in the as-cast structure appears to be associated with pinholing.

The Holt Equipment Company received information from the International Nickel Company to the effect that additions of either tellurium or aluminum will counteract pinholing. This company has been able to virtually eliminate pinholes by addition of tellurium, which was purchased in pellet form from the American Smelting and Refining Company. The required amount of tellurium is about one gram per 100 pounds of ductile iron. Increased quantities of tellurium promote surface roughness and also retard the formation of spheroidal carbon. There is some industrial hazard involved in the use of tellurium, as the fumes are harmful to the workmen, and it should be used only when adequate ventilation is provided.

From the examination of the cast plates, it appeared very doubtful if their quality was sufficiently good to produce valid test results. This possibility had been foreseen, however, in working with the previous test bars, and as a consequence, patterns had also been made for test "Y" blocks that would provide both 1-inch and 1/2-inch thick test sections. These blocks, shown in Figure 15, provided test sections 14 inches long and 2-1/2 inches wide with a continuous feeding reservoir or riser above the casting that could be removed by cutting with a power band saw.

Both tensile and flexure tests were made of the plate material and of the "Y" block test sections in order to determine the effect of

Figure 15. Ductile Iron "Y" Blocks.

the shape of the casting on physical properties. Specimens numbered B-11 and B-12, Figure 16, were from the 1/2-inch plate and B-3 was from a 1/2-inch "Y" block, all of which were poured from the same heat and were in the annealed condition. Specimen B-11 failed at 42,000 psi ultimate strength with an elongation of 1.5 per cent, while B-12 failed in the grips rather than in the reduced section. Specimen B-3 failed at 68,000 psi ultimate strength with 17 per cent elongation. Figure 17, which shows the fractured ends of these bars indicates a more homogenous structure for specimen B-3.

Flexure tests were made on rectangular bars 1-1/2 inches wide and 3/4 inch thick in the annealed condition. Third point loading was applied between knife edges spaced 12 inches apart. Specimen B-52 was from an area with excessive centerline shrinkage of the 1-inch thick plate, according to the radiograph, while specimen B-51 was selected from the section of plate with maximum soundness. The results of these tests are shown in Figure 18 together with specimen B-6 from a 1-inch "Y" block from the same heat. The fractured ends of the flexure test bars, Figure 19, show the dark area of centerline shrinkage in specimens B-51 and B-52 to be predominately in the cope side of the plate, which is the last portion to solidify. The dark areas in these specimens should not be confused with the dark area in specimen B-6, which is finer grained and lighter in color, and does not represent centerline shrinkage. Metallographic specimens of B-6, Figure 20, disclosed no difference in micro-structure between the dark and light areas. The difference is due to light reflection. The side

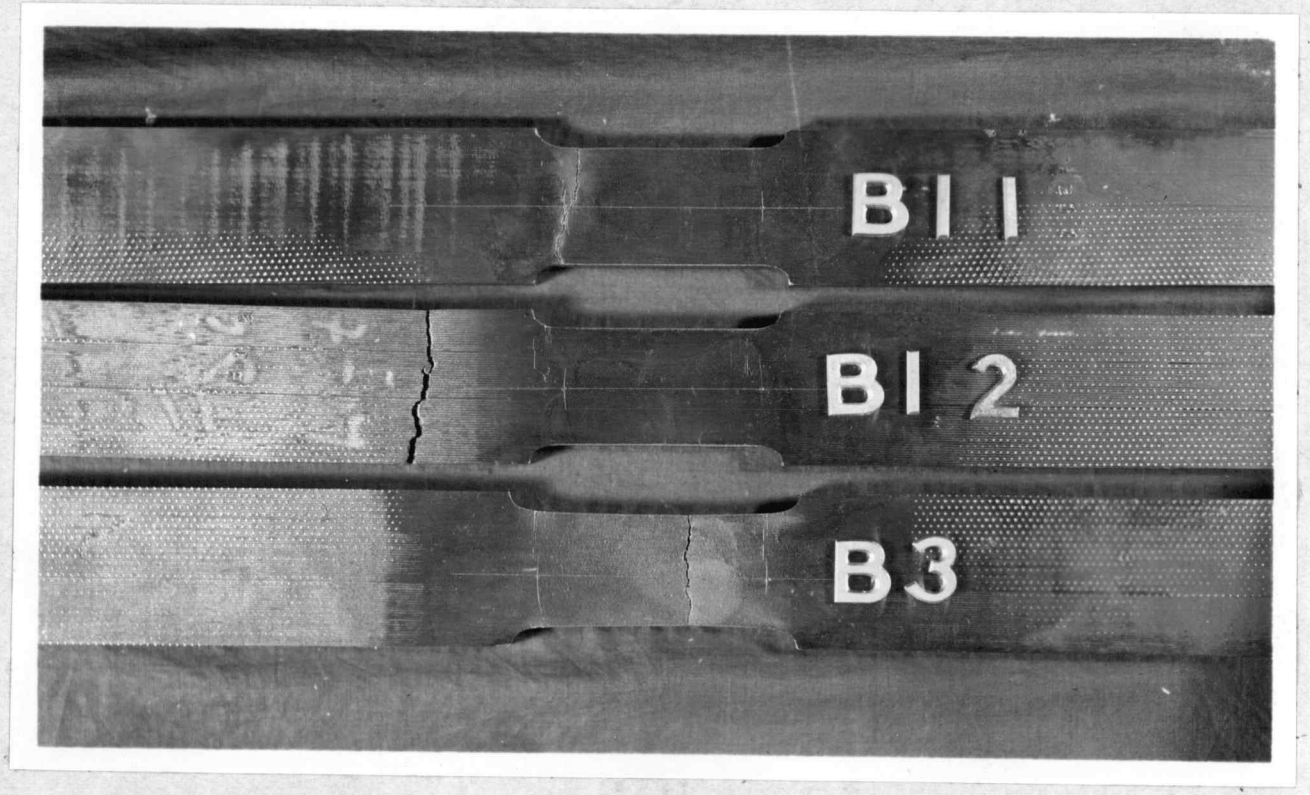

Figure 16. Tensile Tests, B-3 of "Y" Block, B-11 and B-12 of $\frac{1}{2}$ in. Plate.

Figure 17. Fractured Ends of Bars Shown in Figure 16.

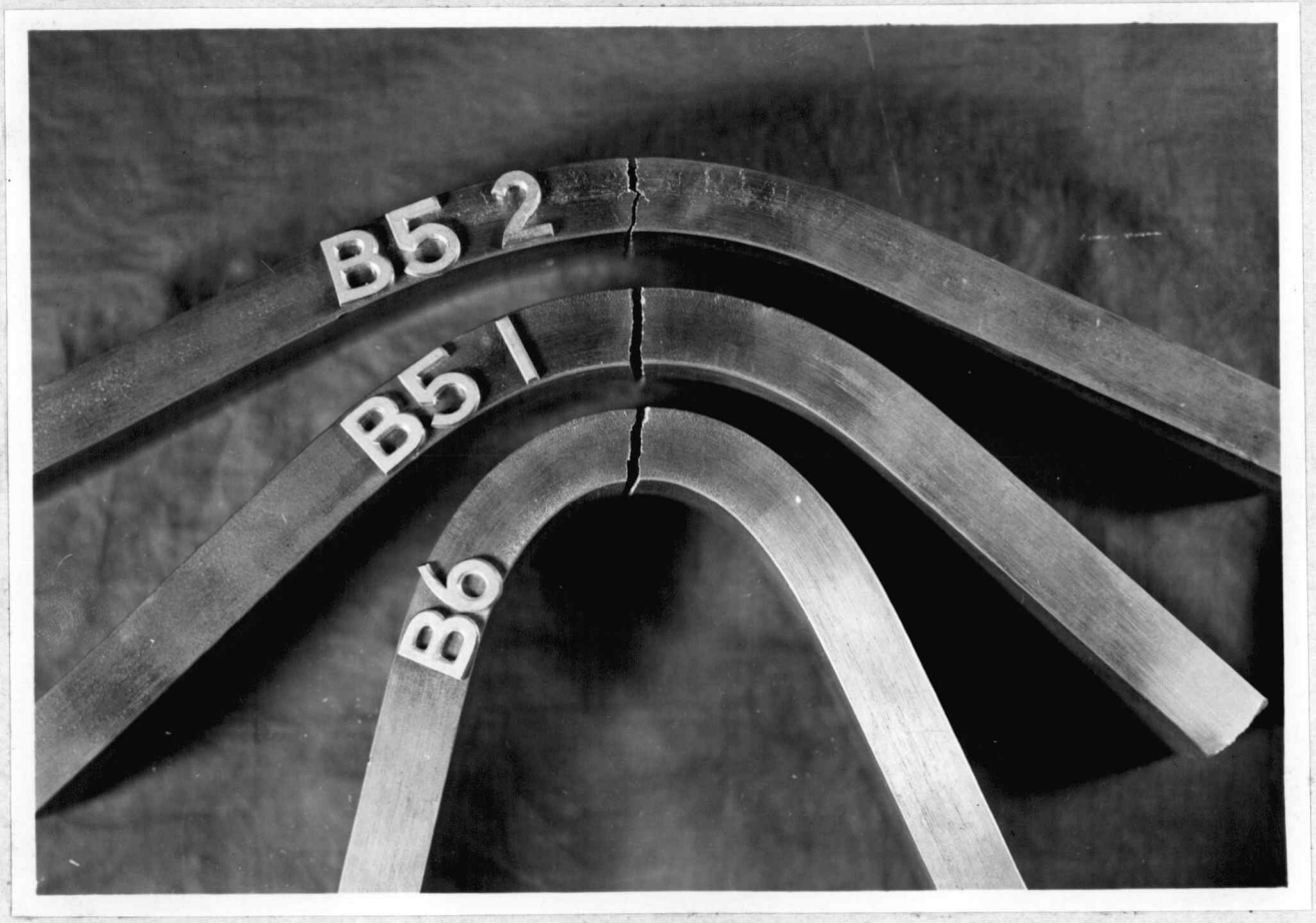

Figure 18. Results of Flexure Test. Lower "Y" Block, Upper Two Plate Material.

Figure 19. Fractured Ends of Specimens Shown in Figure 18.

in tension provided a rough surface with more light absorption while the compression side provided a less ductile type of fracture which crossed several grains at a time without change of direction, thereby providing good light reflection. The etched section from specimen B-52 clearly shows that the centerline shrinkage in this case is of macroscopic magnitude.

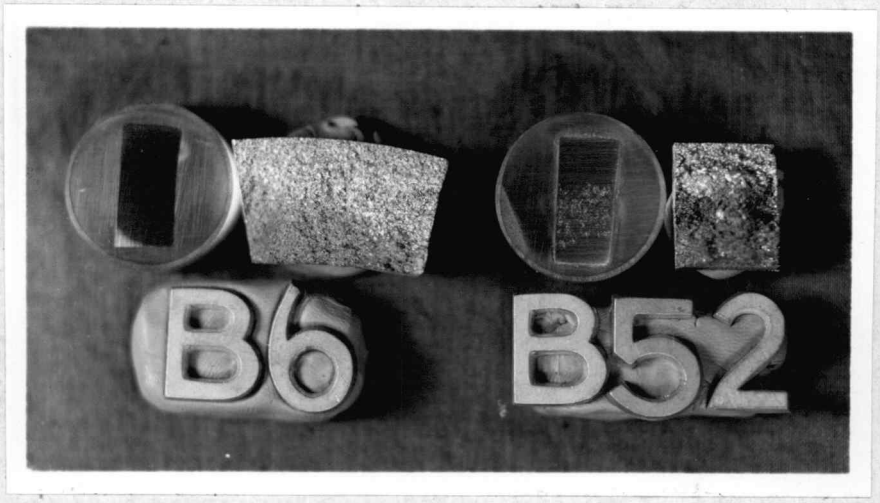

Figure 20. Metallographic Specimens Centerline Shrinkage Shown in Figure B-52.

The evidence provided by the tests just described indicated that "Y" blocks were necessary in order to produce the required quality of material. This was a disadvantage in that the plate material could be processed far more rapidly and permit a broader scope of testing program.

Several castings produced drossy or sooty areas in the fractures as a result of graphite stringers associated with oxy-sulfide and silicide slag, caused by oxidation of iron and magnesium sulfide. It was necessary to reject a large percentage of the earlier test

castings produced, due to this condition which was ascertained by X-ray examination. Defective areas, white, are shown in the print of a radiograph, Figure 21, of a rejected "Y" block casting. This condition was somewhat mitigated by use of higher pouring temperatures. Iron from the fourth tap of the day's heat and from the first to be poured from the ladle was considered to be best from this standpoint. More careful skimming of slag, and use of ladles constructed to provide bottom pour of iron also helped to produce improved quality of castings.

Four castings were produced per mold, the gating system of which is shown in Figure 22. A system involving use of a pouring basin with skim gate would be far superior to pouring directly into the sprue as shown. The use of bottom gating with slag traps would also produce castings of greater soundness. These practices did not appear to be feasible under the production set-up of the Holt Equipment Company; however, due to a continued high rate of rejected castings, slag traps were eventually used in the inlet between gate and casting, which greatly improved the general quality of the castings.

A sheave housing, Figure 23, was fractured in the as-cast condition to determine if any defects could be detected. One would judge from the appearance of the fracture that the casting was perfect. However, a subsequent radiograph, a portion of which is shown in Figure 24, discloses numerous slag inclusions and gas pockets which are white in the print.

Tee beams having a 2-inch flange and 2-inch web, each 3/8-inch thick were cast in 16-inch lengths for use in flexure tests. All of these bars had extensive centerline shrinkage at the intersection of

Figure 21. Print of Radiograph of Defective "Y" Block.

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Figure 22. Gating System for Ductile Iron "Y" Blocks.

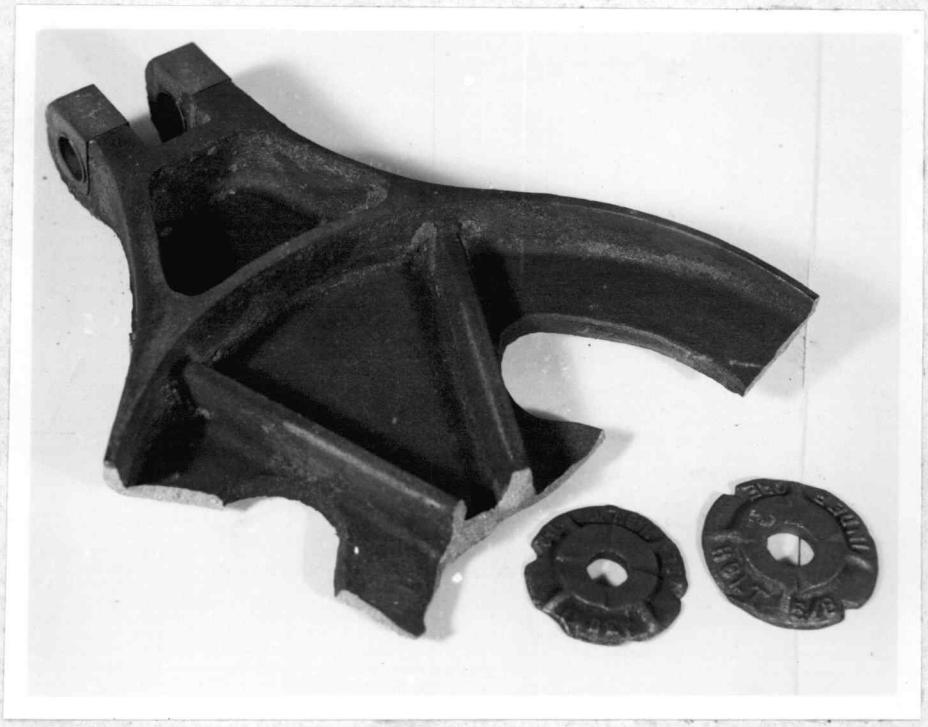

Figure 23. Section of Ductile Iron Sheave Housing and Ductile Iron Washers.

Figure 24. Radiograph of Portion of Casting Shown in Figure 23.

web and flange, with random areas of shrinkage in the flange that interfered with good correlation of test results. This is illustrated by the dark areas in Figure 25 of a broken tee beam. As a consequence, these bars were used to but a minor extent in the program.

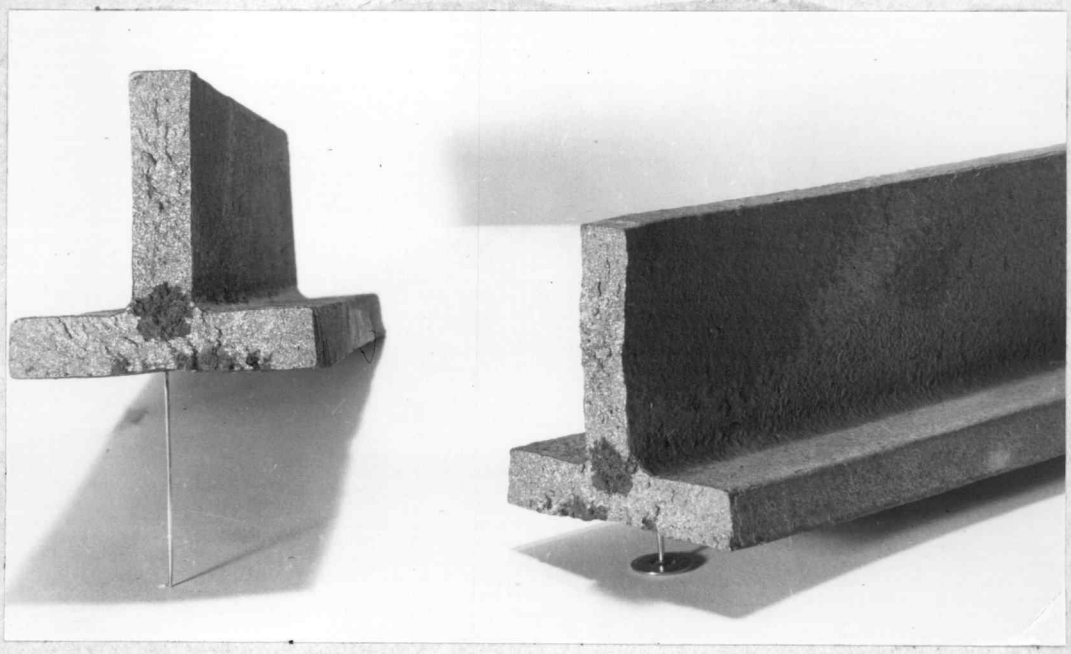

Figure 25. Fractured Tee Beam, Showing Areas of Centerline Shrinkage (dark).

THE PREPARATION OF TEST SPECIMENS

Experience has indicated that due to the high carbon content of ductile iron it is desirable to use a minimum preheat of 500 F when it is to be welded, to avoid fusion-zone cracking. While this is not mandatory for simple shapes with single-pass welds, complicated castings and multiple-pass welds frequently require preheats well in excess of 500 F. This is particularly true when iron base electrodes are used or when using electrodes of large diameters that require higher current settings.

A simple and expedient method of preheating castings was afforded by means of a preheating table with a firebrick top 32 inches square. This table, shown in Figure 26, was fabricated of welded steel pipe and angle iron and located immediately adjacent to the welding booth in order that the effect of preheat would not be lost during transfer of specimens. The table, mounted on heavy industrial casters for portability, had a convenient working heighth of 32 inches.

The preheat burner is shown in operation at the right in Figure 26. Air and gas can be regulated to provide flame of any type or intensity that may be desired. Loose firebrick can be arranged around any size or shape of casting that the table top will accommodate to provide a furnace to suit each individual job.

An alignment and positioning fixture for the specimens to be welded, shown in Figure 27, was fabricated of 6-inch channel iron. This fixture, 14 inches long and 8 inches wide, provided a fast and positive alignment of pieces to be welded and also provided a support

Figure 26. Table and Burner for Preheating Castings.

with ready accessibility during welding operations. The rigidity of the fixture was such that no warpage was experienced during any of the extensive welding operations. The test section on the fixture that has just been welded is as-cast ductile iron, welded with the carbon arc, using ductile iron filler rod.

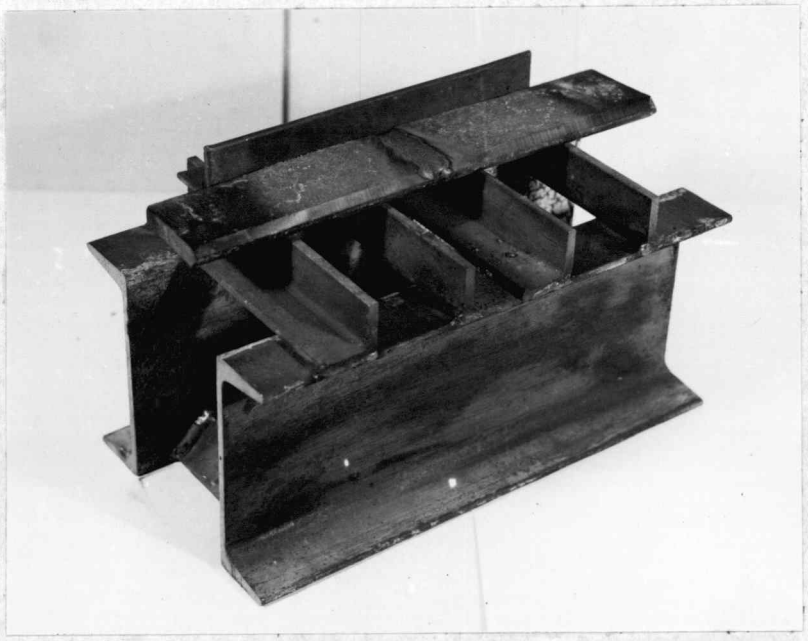

Figure 27. Positioning Fixture for Welding "Y" Blocks.

All castings used in the testing program were first radiographed in order that only those completely sound and free of defects would be utilized. A mechanically rectified 200 kilovolt Victor X-ray machine equipped with a General Electric Coolidge oil-cooled tube, type SRT-4-IND, was used for the radiographic work. This equipment was operated at the maximum setting that would permit continuous operation, which was 174 kilovolts and a tube current of 26 milliamperes. The

gamma ray, consisting of a 50-milligram capsule of radium, was initially used for one-inch thick castings. However, the X-ray equipment provided more satisfactory results, so the use of gamma ray was discontinued.

Both Eastman type K and type A film were used for radiographing the 1/2-inch thick sections. Type K film has the greatest speed of any X-ray film, and 4-minute exposures were required for 1/2-inch thick ductile iron when using lead foil screens. Lead foil screens diminish the effect of scattered radiation and, therefore, produce greater contrast and clarity of image. They also slightly reduce the exposure time for material of this thickness and density. The relative speed of type K film is four times greater than that of type A, so calcium tungstate intensifying screens were necessary when using type A film in order to reduce exposure time. It was found that 1-1/4-minute exposures with type A film and calcium tungstate screen were equivalent to 4-minute exposures with type K film and lead foil screens. X-rays cause calcium tungstate to fluoresce and the light thus emitted intensifies the exposure of the film. It was estimated that only 1/13 of the normal exposure time was required when using intensifying screens with type A film. Intensifying screens produce graininess in the developed film, which interferes somewhat with detection of minute defects; however, the improved contrast of the slower type A over type K film more than compensated for this and consequently produced superior negatives.

It was found that 15-minute exposures were required when using type K film with calcium tungstate intensifying screens for 1-inch

thick ductile iron. It was also discovered that very careful blocking is required over all portions of the cassette not covered by the casting to be radiographed, as a very small area of free radiation on one portion of the intensifying screen would fog the entire film. Although blocking is desired in all cases, it was not found essential when using lead foil screens. Although special X-ray film developer is recommended, it was found that D-19 developer provided quite gratifying results if the time of development was increased by about one-half over that recommended for ordinary film.

A frequent industrial fabrication requirement of ductile iron castings is that they can be integrated with wrought steel shapes in the production of machinery parts. A minor portion of this study, therefore, consisted of the properties of welds connecting ductile iron with steel.

The 3/8-inch thick reduced section tensile bars described in the previous chapter were sectioned at the center and each piece welded to 3/8-inch thick by 2-inch wide steel bar stock to provide rectangular reduced section tensile test bars 14 inches in length. In preparation for welding, the ends of both pieces were ground to form a double vee butt joint with a 60-degree included angle. All surface skin and scale was removed by grinding adjacent to the weld to prevent contamination of the weld. These welds were all made with Ni-Rod 55 electrodes using direct current with the electrode positive. A minimum preheat of 500 F was maintained during the welding process. Slag was thoroughly removed from each pass prior to subsequent

deposit of weld metal.

Flexure test specimens were also prepared of ductile iron welded to steel, using the tee beam sections described in the previous chapter. Steel tee beams 8 inches in length, were fabricated by welding 2-inch by 3/8-inch mild steel bar stock together with fillet welds, using 3/16-inch diameter E-6012 electrodes with direct current straight polarity, and a current setting of 165 amperes. The ductile iron tee beams were fractured in flexure to determine the property of the parent metal. The ends of the broken bars and the ends of the fabricated steel bars were then beveled by grinding to form double vee butt joints with a 60-degree included angle for welding. All scale was removed adjacent to the weld which was to be made and the bars were preheated to 600 F prior to welding. These tee-beam sections, shown in Figure 28, were all welded with Ni-Rod 55 electrodes in the same manner as was described for the tensile test bars.

In continuation of the study of the weldability of ductile iron, using various types of welding processes, filler rods and techniques of welding, it was decided to discontinue use of the shapes just described in favor of "Y" block material, for the reasons stated in the previous chapter. The International Nickel Company, which has conducted some exploratory work in this field, was consulted regarding the most desirable program and their recommendations were followed, together with other reasonable possibilities.

In the Pacific Northwest, virtually all ductile iron is used in the annealed condition at the present time. For this reason, it was decided that testing of welded as-cast material would be of minor

economic importance. Emphasis was, therefore, placed on testing of castings that had been annealed prior to welding and also on castings that had been reannealed subsequent to the welding operation, to determine if this would provide material advantages. Welding of material in the as-cast condition followed by annealing was also included in the program as this would provide economic advantages in the foundry repair of defective castings.

The scope of investigation included metallic arc welding with both ferrous and non-ferrous electrodes, inert gas-shielded arc welding and carbon arc welding. It also included oxy-acetylene fusion, welding and brazing. The complete schedule of welding and heat-treating operations, together with the specimen designations, is given in Table 8.

The original schedule of operations for preparation of castings for welding was as follows:

1. Anneal castings 5 hours at 1650 F, followed by furnace cooling through 1250 F at a rate not to exceed 35 degrees per hour.

2. Clean castings by tumbling to remove scale, sand or other foreign material adhering to the surface.

3. Radiograph to determine quality of casting.

4. Cut test section from feeder "Y" section by means of a power band saw.

5. Bevel casting as required for welding.

The operation of sawing was very time-consuming, and the expenditure of band saw blades was quite expensive. The schedule of operations was, therefore, revised in due time to the extent of

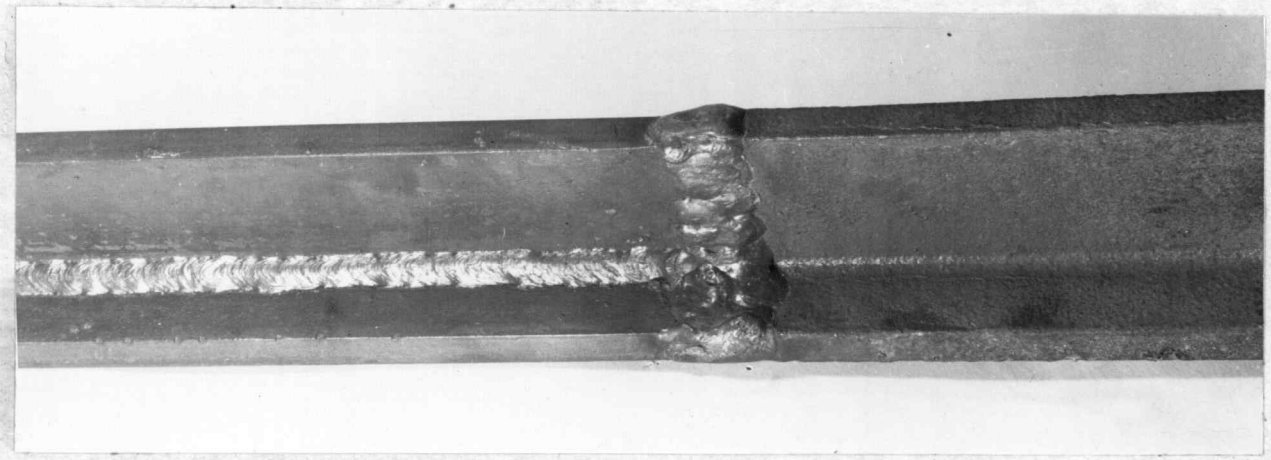

Figure 28. Two-Inch Tee-Beam Flexure Test Specimen, Consisting of Ductile Cast Iron Welded to Steel.

TABLE 8

SCHEDULE OF WELDING AND HEAT TREATMENT OF DUCTILE IRON

Specimen Designation	Welding Process and Electrode	Heat Treatment
A-1		As cast
B-1		Normalized
B-2		Normalized
C-1	Arc, 18-8 Cr-Ni stainless steel	Annealed, not preheated
C-2	Arc, 18-8 Cr-Ni stainless steel	As cast, preheated, welded, then annealed
D-1	Arc, E6015, low hydrogen mild steel	Annealed, not preheated
D-2	Arc, E6015, low hydrogen mild steel	Annealed, preheated
D-3	Arc, E6015, low hydrogen mild steel	As cast, preheated, welded, then annealed
D-4	Arc, E8015, low hydrogen	Annealed, welded, then reannealed
E-1	Arc, 25-20 Cr-Ni stainless steel	Annealed, not preheated
E-2	Arc, 25-20 Cr-Ni stainless steel	Annealed, preheated
E-3	Arc, 25-20 Cr-Ni stainless steel	As cast, preheated, welded, then annealed
F-1	Oxy-acetylene welded, parent metal filler rod	Annealed, preheated
F-2	Oxy-acetylene welded, parent metal filler rod	As cast, preheated, then annealed
G-1	Heliarc welded, parent metal filler rod	Annealed, preheated, welded, then annealed
G-2	Heliarc welded, parent metal filler rod	As cast, preheated, welded, then annealed

TABLE 8 (CONTINUED)

Specimen Designation	Welding Process and Electrode	Heat Treatment
H-1	Arc, Ni-rod 55	Annealed, not preheated
H-2	Arc, Ni-rod 55	Annealed, preheated
H-3	Arc, Ni-rod 55	As cast, preheated, welded, then annealed
I-1	Arc, E6013, mild steel	Annealed, not preheated
I-2	Arc, E6013, mild steel	Annealed, preheated
I-3	Arc, E6013, mild steel	As cast, preheated, welded, then annealed
J-1	Oxy-acetylene brazed manganese bronze filler rod	Annealed, brazed, no preheat
K-1	Carbon arc, parent metal filler rod	Annealed, preheated, welded, then re-annealed

notching the casting in the as-cast condition to a depth of 1/8 inch on each side, on a shaper, with a sharp pointed, high speed cutting tool. The test section was then broken off by means of a 30-ton hydraulic press. If the ratio of rejected defective castings was high, the castings were radiographed prior to being notched. When the rejection ratio was running low, the castings were radiographed after severence of the test section. This provided for availability of castings to be welded in the as-cast condition when desired, since the carbidic condition rendered the castings too hard to be cut by means of a band saw until after they were annealed.

Parent metal tensile test specimens were prepared from numerous heats in order to obtain correlation with welded specimens of similar composition. Test bars were machined on the shaper from the 1/2-inch thick "Y" blocks to cross-sections of 1-1/2-inch in width and 3/8-inch or greater in thickness, with the machining marks parallel to the length. A reduced section 1-1/4 inch wide was then milled with a 2-1/2-inch diameter milling cutter, which provided sufficient radius to prevent stress concentration, of sufficient length to permit a 2-inch gage length. These specimens were prepared both in the as-cast and annealed condition. Rectangular tensile test bars having approximately the same area, and also standard 0.505-inch diameter test bars with 2-inch gage lengths were also prepared from representative samples of one-inch thick "Y" blocks in both the as-cast and annealed conditions.

All specimens to be fusion welded were beveled at the center by means of a shaper, to form double vee butt joints with included angles of 60 degrees. Although a greater included angle, usually 90 degrees, is frequently used for oxy-acetylene welding, the thickness was not sufficient in this case to make an appreciable difference, so the 60-degree angle was likewise used for this process. An exception in joint preparation was made in the case of the bars to be brazed, where a shear vee joint was prepared to provide additional surface. Brazing consists of a surface bond between the parent metal and deposited metal, and strength can be obtained only by means of special joint preparation. All surface scale was removed by grinding for a distance of about 1/2 inch from the prepared joint to prevent contamination of the weld deposit.

Lincoln arc welding machines with dual control of current and voltage were used for all metallic and carbon arc welding operations. The first ferrous welding electrodes used were type AWS 308-15 stainless steel, containing 18 per cent chromium and 8 per cent nickel. The bars to be welded were preheated to 800 F and the heat of welding was sufficient thereafter to maintain this temperature in the joint area of the base metal. Beads were successively deposited on alternate sides of the joint, with sufficient weaving to permit flow of weld metal to each side of the joint in each pass. Electrodes 1/8 inch in diameter were used for the root beads with a current setting of 90 amperes, and 5/32-inch diameter electrodes were used for the remaining passes with a current setting of 140 amperes. Complete slag

removal was effected on each pass prior to deposit of subsequent weld metal. The flow of weld metal was smooth, providing even deposit of metal and ease of manipulation of electrode. Specimens were prepared of both annealed ductile iron as welded and with a reanneal following the welding operation.

Flexure test specimens were prepared for welding from one-inch "Y" block material in the same manner as described for tensile tests of 1/2-inch thick material. These bars were also welded with type AWS 308-15 stainless electrodes, using the same procedure as for the tensile test bars. The results of these tests, which will be discussed later, were of such nature that it was decided to discontinue further flexure tests until the entire tensile testing program was completed.

Low hydrogen mild steel electrodes of the E-6015 type were then used in preparation of test specimens. Some of these specimens were welded without benefit of preheat, while others were preheated as previously described. Electrodes 1/8 inch in diameter were used for the root beads and 5/32-inch diameter electrodes were used for the remaining passes. Each size of electrode required about 10 amperes more current than was required for the type 308 stainless electrodes of corresponding size. This created a problem of greater admixture of weld and base metal, so care was taken to permit the weld metal to roll against the sides of the vee during the weaving manipulation rather than to permit direct contact between the arc and parent metal. The current setting for this type of electrode was quite critical for although a minimum current is definitely preferred, there was a

greater tendency for slag entrapment at lower amperage values. Welds with this type of electrode were made in the as-cast condition followed by annealing; in the annealed condition as welded; and in the annealed condition followed by a reanneal. Welds were also made on material containing less than 3 per cent silicon and over 4 per cent silicon in order to determine the effect of this element on fusion-zone cracking.

Stainless steel electrodes of the AWS type 310-15, containing 25 per cent chromium and 20 per cent nickel, were used in preparation of test specimens in order to determine the advantage, if any, of increased alloy content of weld metal. The welding procedure and current settings were similar to those for type 308-15 electrodes. These electrodes provided the greatest ease of manipulation and smoothness of transfer of weld metal of any type used either earlier or later in the program. Specimens were prepared for test purposes both in the annealed as welded and also in the annealed, welded and then reannealed conditions.

The final type of ferrous electrodes used in welding ductile iron for test purposes were the AWS class E-6013. These are straight polarity mild steel electrodes with shallow penetration into base metal due to a soft, non-digging type of arc. This characteristic is provided by a high ratio of potassium silicate to sodium silicate in the coating. Among the mild steel electrodes with coatings that contain organic materials, this type was selected as it would provide a minimum of carbon pickup from the base metal. In addition, the high

titania-potassium silicate content of the coating, with a corresponding reduction of cellulose, provides less hydrogen to react with the base metal than AWS type E-6010 or E-6011 electrodes. The characteristics of weld metal transfer for this type of electrode are of such nature that lower current settings can be used than for the previously discussed types. A current setting of 80 amperes was used for 1/8-inch diameter electrodes and 110 amperes for 5/32-inch diameter electrodes. A noticeable boiling reaction took place as the weld metal was deposited. This was attributed to the probable formation of acetylene gas due to reaction between moisture which this type of coating contains and carbides in the weld metal due to the sudden carbon pick-up from the base metal. When the slag was removed from the weld metal, the deposit was porous and rough. Even before testing, it was readily apparent that the weldability of ductile iron with this type of electrode was very poor.

A number of test bars were welded with Ni-Rod 55, an electrode consisting of core wire containing 60 per cent nickel and 40 per cent iron, with an extruded carboniferous limespar coating. This electrode is particularly recommended by the manufacturer for welding of ductile iron castings. Root passes were deposited with 1/8-inch diameter electrodes at 85 amperes and intermediate open line voltage and the remaining passes were made with 5/32-inch diameter electrodes at 105 amperes. All passes, with the exception of the root beads, were made by weaving to the full extent of the width of the joint rather than by deposit of stringer beads. This electrode has excellent operating

characteristics, and the base metal responded well to its application. The low current setting provided a minimum of alloying between the parent metal and the weld. The material was welded in the annealed condition without benefit of preheat and also in the preheated as-cast condition followed by an anneal.

For heliarc welding, a 600 ampere alternating current welding machine was used with a super-imposed high frequency current of 10,000 cycles for stabilization of arc, and ease of striking and maintaining an arc without touching the electrode to the parent metal. A 1/8-inch diameter thoriated tungsten electrode, contained within a ceramic cup size No. 8, provided the arc for the heat for welding. Argon, an inert gas, passed through the cup at a rate of 22 cubic feet per hour to surround the weld deposit and protect it from atmospheric contamination. The electrode holder was protected by water cooling. The gas was regulated by means of a flow meter. Filler rod from parent metal of composition similar to that to be welded was cut into approximately 1/4-inch square bars by means of a band saw. The parent metal was preheated to 1000 F and welded on the preheating table in order to maintain a high temperature in the base metal during the welding operation. A rather high current of 200 amperes was used in order to effect a rapid deposit of weld metal. All specimens were reheated to 1000 F immediately following the welding for five minutes to eliminate a condition of stress. Excellent weldability was experienced by this process and the welds were completed at a more rapid rate than by any other means. Flux was not required as oxides were readily floated to the surface due to the fluid condition provided by the high welding

temperature. Specimens were welded both in the as-cast and annealed condition and all welded specimens were given a full post-weld anneal.

The microstructure of the heliarc welded specimens disclosed a partial reversion to a quasi-flake graphite, that reduced the mechanical properties. This was attributed to the high temperature involved in the welding process, so it was accordingly determined to use a minimum of current during the operation of carbon arc welding. Cored carbon electrodes, 1/4 inch in diameter were used with electrode negative, direct current, at 120 amperes and high open line voltage. Flux was not required, as the fluidity of the welding puddle was such that oxides were readily floated to the surface. Parent metal filler rod was used, and the welding technique was similar to that of oxy-acetylene welding, with the exception that the speed of carbon arc welding was much greater. Even when high preheats were maintained, it was found that if the layer of deposited metal was too thick, there was a tendency for hot cracks to form in the center of the weld. This was easily avoided, however, by depositing less metal per pass. A special carbon electrode holder was not available, and it was, therefore, found necessary to periodically immerse the conventional type of electrode holder in water to prevent overheating. The material welded was annealed ductile iron, which was reannealed after welding.

Parent metal filler rod was also used with the oxy-acetylene welding process. An Airco style 800 welding blowpipe was used, with No. 10 tip. Both Oxweld Ferro-Flux and Smith cast iron welding flux were used; however, neither was very effective in removal of oxides

formed during the welding process. The best technique under the circumstances was to use sufficient heat to enable the rod to be retained in the molten pool, maintaining a puddling action, without disturbance of the surface film of oxide. The high carbon content of the weld produced a relatively low melting condition, near the eutectic, and the oxides formed had a higher melting point. The failure of the flux to adequately lower the melting point provided a viscous surface condition that prevented complete gas escapement. As a result, gas pockets existed in the surface layers of the weld, and to some extent, deeper within the weld. It is recommended that further experimentation be conducted regarding the correct flux to use if extensive oxy-acetylene welding is to be done on ductile iron. All oxy-acetylene welding was done on annealed ductile iron, and some of the specimens were given a reanneal prior to testing.

The ductile iron test bars that were prepared with a shear vee joint, were brazed or bronze welded with Oxweld No. 25M manganese bronze filler rod. An Airco style 800 torch was used with a No. 8 tip. The oxy-acetylene flame was adjusted until it was slightly oxidizing. Molten bronze readily absorbs hydrogen and carbon monoxide gas, which forms porosity upon solidification due to decreased solubility. The slightly oxidizing flame assured that these gases were not present. No preheat was used or considered necessary, as this function automatically took place by means of the torch while bringing the pieces up to the tinning temperature. The joint was tinned with 1/8-inch diameter rod, using Smith's Amalgam high temperature

brazing flux. The weld was then completed with 3/16-inch diameter rod and Oxweld Brazo flux. All specimens were annealed prior to brazing and no subsequent heat treatment was given.

All welded tensile test bars were machined to rectangular cross sections of about 1-5/8-inch wide by 3/8-inch thick. A reduced test section was then milled or shaped to a 1-3/8-inch width with a 1-1/4-inch radius to provide a 2-inch gage length on each bar. Those specimens that were readily machinable had the reduced section milled, and those that required carbide tools for machining were shaped to provide the reduced section. The flexure test bars, with exception of the tee beams, were machined to a rectangular cross section of 1-1/2 inches wide and 3/4-inch thick. The composite welded tee beams were not machined.

Both the fusion zone and weld zone of the specimens welded with mild steel and stainless steel electrodes were exceedingly hard. Carbide tools were required to satisfactorily machine these bars, and even then, periodic sharpening of the tools was necessary. Carbide tools were also required for oxy-acetylene welded material that was not subsequently annealed.

All ductile iron that was welded by the carbon arc, inert gas shielded arc or the oxy-acetylene flame, using parent metal filler rod, was readily machinable with high-speed cutting tools, provided a post-weld anneal was applied. All material that was brazed was likewise readily machinable with high-speed steel bits. The castings welded with Ni-Rod 55 could be machined with difficulty with high-speed tools; however, carbide tools were much preferred.

After the test specimens were machined, the test sections were sanded on a belt sander and draw filed longitudinally to remove any tool marks remaining from the machining process. The corners were slightly rounded and 2-inch gage marks were applied to the test sections. The specimens were then ready for mechanical testing.

THE METALLOGRAPHY OF WELDED DUCTILE IRON

Numerous advantages are offered by a metallographic study of welds. An examination of the structure enables one to understand the extent to which integration has taken place between the weld and base metal. The effect of welding heat upon the base metal, the alteration products that exist in the fusion zone, the nature of defects and inclusions, and many other problems are solved by microscopic examination. It assists in interpretation of test results and enables one to predict many physical properties with reasonable accuracy prior to testing.

Specimens were prepared for metallographic study from welds made with each process, type of filler rod and condition of heat treatment. Many welds were too hard to be cut with a hacksaw and these were sectioned by means of an abrasive cut-off machine. The specimens were mounted in bakelite by means of a mounting press to facilitate polishing and handling.

The procedure for dry polishing consisted of first surfacing the specimen on a belt sander equipped with a 240 grit belt. The specimen was then rotated 90 degrees and hand polished on emery paper with No. 1 grit until all scratches were removed from the preceding operation. It was then polished on No. 1/0, 2/0 and 3/0 emery paper in succession, with rotation between each. Graphite was applied to the 3/0 paper to help prevent removal of the graphite nodules in the specimen.

Very little trouble was experienced in retaining the carbon nodules prior to the wet polishing operations. The first attempt at polishing was with a canvas-covered wheel with 600X alundum abrasive which was followed by a broadcloth wheel with levigated alumina. Carbon nodules were oompletely removed by this method, and a wax-covered lap was prepared as a replacement for the canvas-covered wheel. The situation was improved, but the nap on the broadcloth wheel also removed particles so it was then replaced with a silk-covered wheel. It was eventually found that the best procedure was to go from the 3/0 emery paper directly to the silk-covered wheel and etching and repolishing several times until all scratches were removed. The nodules were retained in this manner, but they were slightly depressed at the center and their structure was obscured.

In view of the fact that the results of the wet polishing operations were not entirely satisfactory, a very fast-cutting abrasive might prove more suitable. For this purpose, the Buehler Diamet-Hyprez diamond dust compound offers good possibilities. The procedure would be to follow the 3/0 emery polishing paper with use of 600X alundum on waxed billiard cloth. The final polishing operation would consist of the use of 1/4 micron diamond dust on microcloth.

A Buehler Model No. 1700 Electro Polisher was used to prepare the stainless steel weld deposits for microscopic examination. This eliminated the time-consuming fine polishing operations and provided a surface free of disturbed metal, for stainless steel is particularly susceptible to cold working. A non-explosive electrolyte was used,

consisting of 750 cc of 95 per cent ethyl alcohol, 140 cc of distilled water and 50 cc of perchloric acid. The same solution was used for the electrolytic etch, without removing the specimen from the polisher, by merely reducing the current density.

Attempts to use an electrolytic polish on the base metal were not successful, as the electrolyte produced a strong galvanic action between the nodules and the surrounding matrix that removed the graphite and enlarged the holes where it had been. The dissimilar metals of the fusion zone also produced a corrosive galvanic action, and it was concluded that the electrolytic polish would be of advantage only for weld deposits. The specimens that were prepared by means of the polishing wheels were etched with 2 per cent nital.

An examination was made of the parent metal, Figure 29, in the as-cast condition at a magnification of 100X. As cooling progressed from the liquid melt, gamma iron saturated with carbon was the first to solidify. As cooling continued, silicon also dissolved in the austenite caused a reduced solubility for carbon, which, nucleated by minute particles of magnesium oxide, formed graphite nodules in the gamma iron areas. The last portion to solidify formed a network of eutectic, consisting of cementite and gamma iron, around the primary austenite.

Twin nodules of graphite in as-cast ductile iron are shown in Figure 30 at a higher magnification of 500X. The gamma iron with which they were surrounded formed coarse pearlite upon cooling to below the transformation temperature. Examination of a broad area disclosed

that as the material cooled from the solidus and the solubility of carbon in austenite decreased, stringers of iron carbide precipitated out in the austenite areas to form a Widmanstätten pattern. The section of casting that was examined was relatively thin and the cooling rate was sufficiently rapid to prevent the excess carbon from migrating to the nodules. This represents a very brittle condition which can be eliminated by annealing.

The standard procedure for annealing consisted of heating for five hours at 1650 F, followed by very slow cooling through the transformation temperature. During the 1650 F treatment, all of the excess carbon which was shown as carbides in Figure 30, had time to migrate to the nodules, which were surrounded by gamma iron saturated with carbon. As cooling took place, and the solubility of carbon in gamma iron decreased, it migrated to the nodules when given time by slow cooling, and with very slow cooling during transformation, all of the remaining carbon in the gamma iron moved to the nodules. In this manner, a completely ferritic matrix was provided that was quite soft and ductile, as shown in Figure 31 at 100X.

A piece of ductile iron was removed from the furnace and air cooled, after it had been held at 1650 F for five hours. The normalized structure thus obtained is shown in Figure 32 at 100X. Very little time was afforded for migration of carbon to the nodules from any distance during cooling. Some carbon-free ferrite existed around almost all of the nodules. The matrix was fine pearlite, some of which was well resolved at 500X, as shown in Figure 33. Carbon appeared to migrate quite rapidly along grain boundaries and to a

lesser extent along certain preferred planes within the grain itself. It may be noted that the small amount of free ferrite present in Figure 33 had a tendency to exist in grain boundaries, from whence the carbon has moved to nearby nodules.

The use of mild steel electrodes in ductile iron will now be used to illustrate the nature of some of the more serious problems that can exist. At the right of Figure 34, the parent metal was not heated above the critical temperature by the welding process. At the left in the same figure, the base metal was heated to just above the critical temperature and carbon started to diffuse from the nodule into the surrounding austenite. Transformation products resulted when heat was rapidly conducted away by the base metal.

As the weld is approached, Figure 35, the temperature has been increasingly higher and more transformation products surround the nodules in areas where carbon has diffused. At the left sufficient carbon diffused to saturate the austenite adjacent to the nodules, now represented by extremely fine pearlite. In the upper left portion, nodules are surrounded by a white area. The temperature here was just above the solidus, and the carbon saturated area surrounding the nodule melted as it was of eutectic composition. Upon cooling, iron carbide surrounded by austenite transformation products including martensite existed.

At the extreme upper left of Figure 35 and at the right of Figure 36, the liquid surrounding the nodules met and they had completely dissolved. A hypereutectic condition existed in the area

of the nodule, which was surrounded by metal of hypo-eutectic composition. Carbon migrated rapidly from this area into the low carbon steel weld deposit at the left, increasing it to a eutectoid composition. At the upper left, farther into the weld, the carbon content is slightly less and a small amount of ferrite exists around the high carbon pearlitic areas.

An extensive crack exists in the parent metal just beneath the weld deposit, Figure 36. The coating of the E-6013 electrodes, used in making the weld, contained cellulose, which produced hydrogen gas during the welding process. Much of the hydrogen, in atomic form due to the heat of the arc, dissolved in the molten iron. When the weld solidified, the solubility for hydrogen decreased. The hydrogen removed from solution was quite free to move around in the iron in atomic form. When it reached a void, it combined with other atoms of hydrogen to form molecular hydrogen. As this process continued, a tremendous pressure was built up in the void, which, in many cases, was sufficient to cause extensive cracking which extended from the void. The material welded in Figure 36 also contained 4.22 per cent silicon, which was high enough to make it prone to fusion-zone cracking. These cracks were extended by the hydrogen gas to the magnitude indicated in the photomicrograph.

An area farther into the weld-zone is represented by Figure 37. With progression to the left, increasing amounts of ferrite exist around the grain boundaries. At the upper left, much ferrite is precipitated out within the prior austenite grains, forming a

Widmanstätten pattern. A considerable carbon pick-up is noted, even in the center of the weld.

The effect of annealing a weld made on parent metal in the as-cast condition with the same electrode will now be discussed. Figure 38 extends from the parent metal where no liquid existed around the nodules to the area where a completely liquid area existed. The area that formerly consisted of free carbides, martensite and very fine pearlite, now consists of a ferrite matrix with nodules of graphite. These nodules, formed well below instead of above the solidus, are very small and placed quite close together. Groups of these small nodules form a pattern providing a straight line continuity that lowers the strength and ductility to some extent.

The area immediately to the left of the renodulated area, Figure 39, is parent metal that was in contact with the weld. Carbon transferred from this region to the weld to such an extent that an insufficient amount remains to provide renodulization. A hypereutectoid condition then exists with some free carbides present.

The next region to the left appears to be very fine-grained in Figure 39. This same region at higher magnification, 500X, Figure 40, shows that it had sufficient carbon so that it was at least of eutectoid composition. The rapid cooling effect of the base metal that adjoined this region to the right produced a martensitic or at least a very fine pearlitic condition. The subsequent anneal produced a partial spheroidizing effect on this very fine-grained area, and massive carbides embedded in a ferrite matrix are noted in coexistence with clearly resolved pearlite.

The area to the left of the partially spheroidized area, Figure 41, is farther into the weld. When this area is magnified at 500X, Figure 42, it is found to be 100 per cent coarse pearlite as a result of the anneal. The fact that this region is of eutectoid composition leads to the conclusion that the region closer to the parent metal is of hypereutectoid composition.

From the fully pearlitic area, progression to the left, Figure 43, shows first a ferrite network surrounding pearlite. Farther into the weld to the left, increasing amounts of ferrite appear. Magnification of this region to 500X, Figure 44, shows areas of ferrite (white) and pearlite.

The low cost of mild steel electrodes make them quite attractive to the consumer. From the foregoing discussion, it is apparent that severe problems exist regarding both the complex structures produced and the content of hydrogen gas, that renders their use undesirable. This situation might be somewhat mitigated by the use of low hydrogen electrodes that do not contain either moisture or organic material in the coating. The gas-forming ingredient of the coating that functions for exclusion of air from the molten metal, consists of inorganic material. The electrodes are baked at 1100 F to remove moisture and then sealed in airtight containers, in which condition they are available to the consumer. They should not be exposed to normal atmospheric conditions for more than a few hours prior to use. The nature of the coating is such that higher currents are required during welding and the integration between weld metal and parent metal is accordingly

increased.

Deposited weld metal from E-6015 low hydrogen electrodes, in an area close to the parent ductile iron which was not preheated, is shown in Figure 45. The prior austenite grains are surrounded by a cementite network. They are much larger than the corresponding grains in a deposit from E-6013 electrodes, due to the higher temperature of application. The carbide network caused fractures of a brittle nature.

A view of the heat-affected zone of the parent metal, which is very wide, is shown in Figure 46. The heat of welding produced a eutectic condition around the nodules in areas that almost joined. The chilling effect of the base metal, which was not preheated, produced martensite (white) with embedded stringers of carbide (also white) surrounded with an area of very fine, hard pearlite.

When the annealed base metal was preheated to temperatures of above 800 F prior to welding, the heat-affected zone is much more narrow, Figure 47. Free carbide stringers still exist, although to a lesser extent, and the exceedingly hard martensitic matrix is not in evidence. This is indicative of the beneficial effect of preheating. Some underbead cracking was found in as-cast ductile iron welded with low hydrogen electrodes, Figure 48. One of the functions of annealing is to degasify the metal. Hydrocarbon gas existing in the as-cast base metal is believed to have been decomposed by the welding arc, producing some atomic hydrogen. This condition can, therefore, be expected when welding the casting prior to annealing. In this case, annealing subsequent to welding, has eliminated both martensite and carbide stringers.

From the preceding discussion, it was anticipated that a solution to the problem might lie with welding of annealed ductile iron followed by a reanneal. Higher strength, low-hydrogen electrodes of the E-8015 type were used, and the fusion zone is shown in Figure 49. The change of electrodes was probably a disadvantage. These electrodes had apparently been in storage too long and the moisture content of the coating caused some underbead cracking.

A portion of the weld zone is shown in Figure 50 at 500X, in order to resolve the pearlite and disclose a slightly hyper-eutectoid condition. This condition was found to a greater extent closer to the parent metal, and the carbide network provided by the anneal caused brittle fractures.

The International Nickel Company encouraged experimentation with stainless steel electrodes. The first used were type AWS 308-15, austenitic 18-8 stainless steel electrodes. A very brittle martensitic heat-affected zone resulted with excess carbide flash, as shown in Figure 51. The weld metal with its high carbon pick-up solidified into austenite dendrites as shown in Figure 52. The center of the dendrites, lower in carbon were first to solidify from the liquid melt. The edges of dendrites, richer in carbon solidified last. The electrolytic etch attacked the high carbon areas, permitting the centers of the dendrites to stand in relief.

When the same electrodes were used on as-cast ductile iron, with annealing subsequent to welding, the martensite in the fusion zone was converted to very fine pearlite with some embedded carbides,

as shown in Figure 53. This structure was stabilized by chromium carbide. Chromium was evidently imparted to this region of the base metal by the electrode. Carbon that remained in solution due to the fast rate of cool of the as-welded specimens was precipitated in the dendrite boundaries as chromium carbide upon annealing. These carbides, Figure 54, formed a very brittle network that resulted in weld failures at low physical values.

Welds were made with electrodes of the AWS-310-15 type, containing 25 per cent chromium and 20 per cent nickel, to determine if any advantages would be derived from the increased alloy content. Figure 55 is of the fusion zone without use of preheat, and Figure 56 is of the fusion zone with use of preheat. The heat-affected zone is more narrow when the base metal is not preheated. It would appear from this that preheating might not be of advantage in some instances, such as for single-pass welds. The austenite dendrites of the weld metal deposit, Figure 57, are somewhat shorter and wider than those of type 308 welds.

Ductile iron welded in the as-cast condition with type 310 electrodes and subsequently annealed, was studied. The fusion zone, Figure 58, shows renodulation in the portion of the base metal that was altered by the heat of welding. With progression toward the weld area, appreciable amounts of chromium diffused into the parent metal. The chromium carbides formed were so stable that the anneal had very little effect on them. Fine pearlite was stabilized with many carbide stringers throughout, particularly in the boundaries of prior austenite grains. Carbides also precipitated out in the weld to form

a continuous network around the austenite dendrites.

Oxy-acetylene fusion welds were made on annealed ductile iron, with filler rod similar to the parent metal. In the as-welded condition, the weld metal upon solidification consisted of primary austenite dendrites with a substantial network of ledeburite. In Figure 59, the austenite with embedded nodules was converted into fine pearlite surrounded by a heavy network of cementite, upon cooling to below the critical temperature. The sustained high heat of the welding process converted the adjacent parent metal into fine pearlite with some ferrite present, Figure 60. Although the weld possessed relatively high strength, both it and the adjacent pearlitic parent metal were hard and difficult to machine.

When oxy-acetylene welds were annealed, their structure was quite similar to that of annealed parent metal. Figure 61 shows this condition with the nodules slightly smaller than those in the base metal. This method of welding is metallurgically more promising than any other.

Welds made with the inert gas-shielded tungsten arc process had an appearance in the annealed condition, very much like the oxy-acetylene welds, as shown in Figure 62. The high temperature involved converted some of the graphite to a quasi-flake form as shown in Figure 63. These semi-flake graphitic areas, that tend to exist in the outer layers of the weld, produced lower strength than that of the oxy-acetylene welds.

Lower currents were used for welding with the carbon arc process than with the heliarc, in order to prevent conversion to the

semi-flake graphite form. The carbon arc welds had the highest physical values tested, but the nodules of graphite were smaller and spaced closer together than those in the oxy-acetylene or heliarc welded specimens. This was due to the lower welding current and maintenance of a smaller puddle, which resulted in rapid solidification.

The fusion zone of the specimens welded with Ni-Rod 55 nickel-iron electrodes was fairly hard and difficult to machine with tool steel bits. The fusion zone at the root of the weld was partially martensitic with carbide stringers, Figure 64. When the weld was annealed after welding, all trace of martensite and carbide flash was removed as shown in Figure 65. Primary cored dendrites of weld metal, Figure 66, were somewhat homogenized in the as-welded condition by the anneal, as shown in Figure 67.

The micro-hardness of the weld zones of the mild steel weld deposits was greater than that of any other as-welded specimens, with the exception of the as-welded oxy-acetylene weld. Knoop hardness of 452 for the low hydrogen weld zone was greater than that of 352 for the E-6013 weld. The oxy-acetylene weld was intermediate between the two (432). When annealed, the Knoop hardness was reduced to 332 for the low hydrogen weld and to 108 for the E-6013 weld deposit.

The hardness of the stainless steel weld deposit, Knoop 142, was lower than that of Knoop 192 for the Ni-Rod 55 weld in the as-welded condition. The hardness of the annealed heliarc weld, Knoop 187, was a little more than that of the parent metal. The softest weld

deposit was the manganese bronze, Figure 68, with a Knoop hardness of 108. The deposit of bronze had no effect on the structure of the parent metal to which it was bonded, Figure 69.

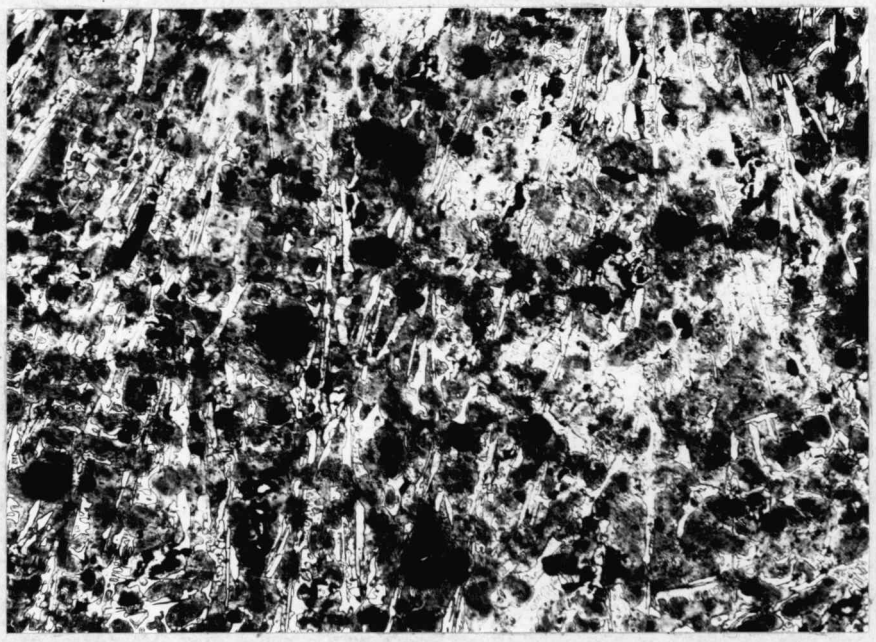

Figure 29. Ductile Iron As Cast. 100X.

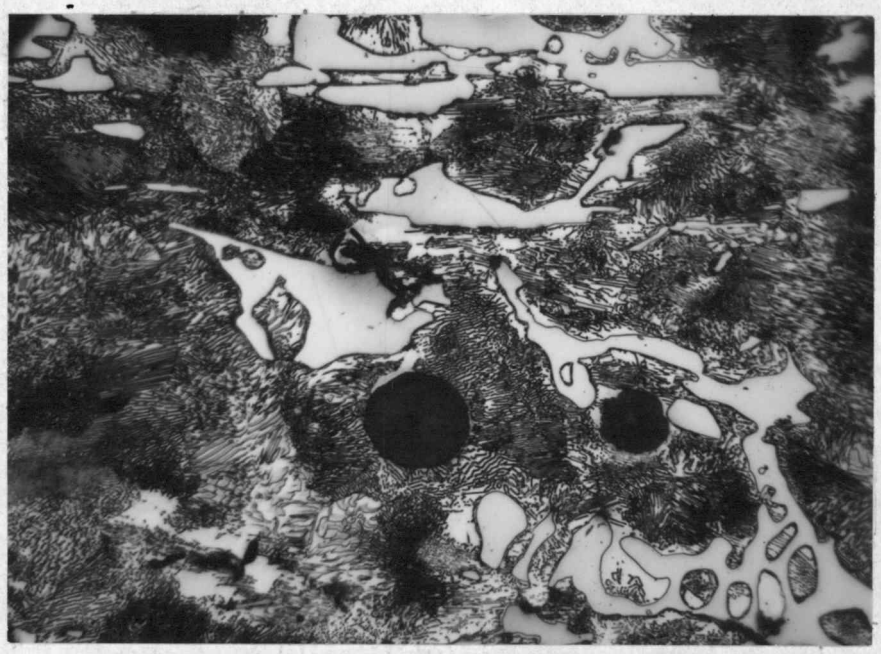

Figure 30. Ductile Iron As Cast. 500X.

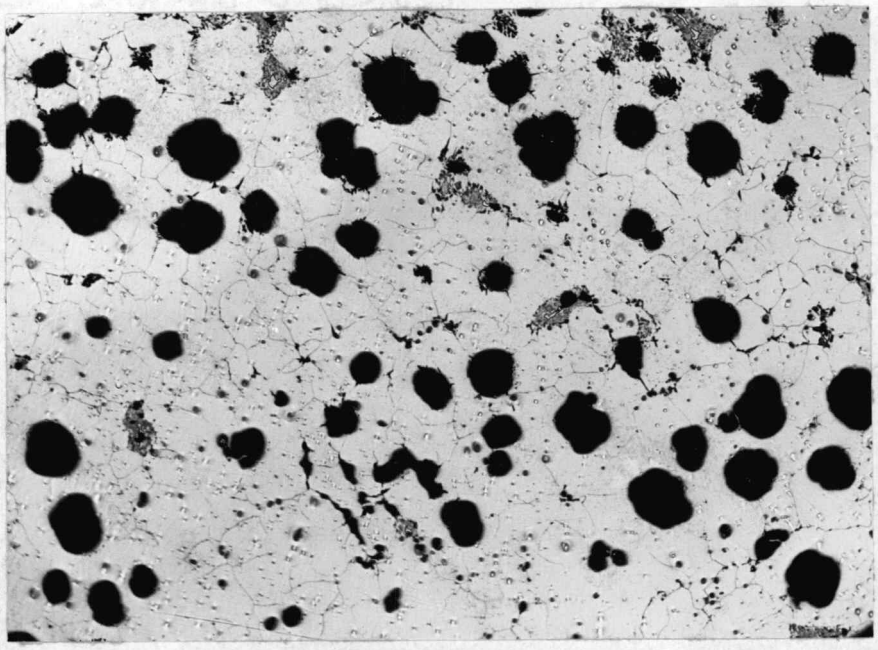

Figure 31. Annealed Ductile Iron. 100X.

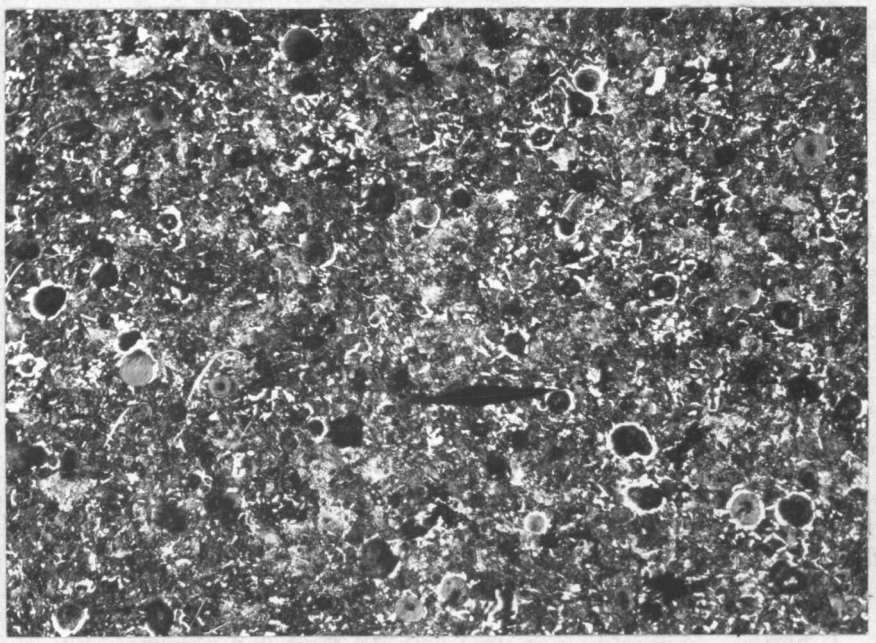

Figure 32. Normalized Ductile Iron. 100X.

Figure 33. Normalized Ductile Iron. 500X.

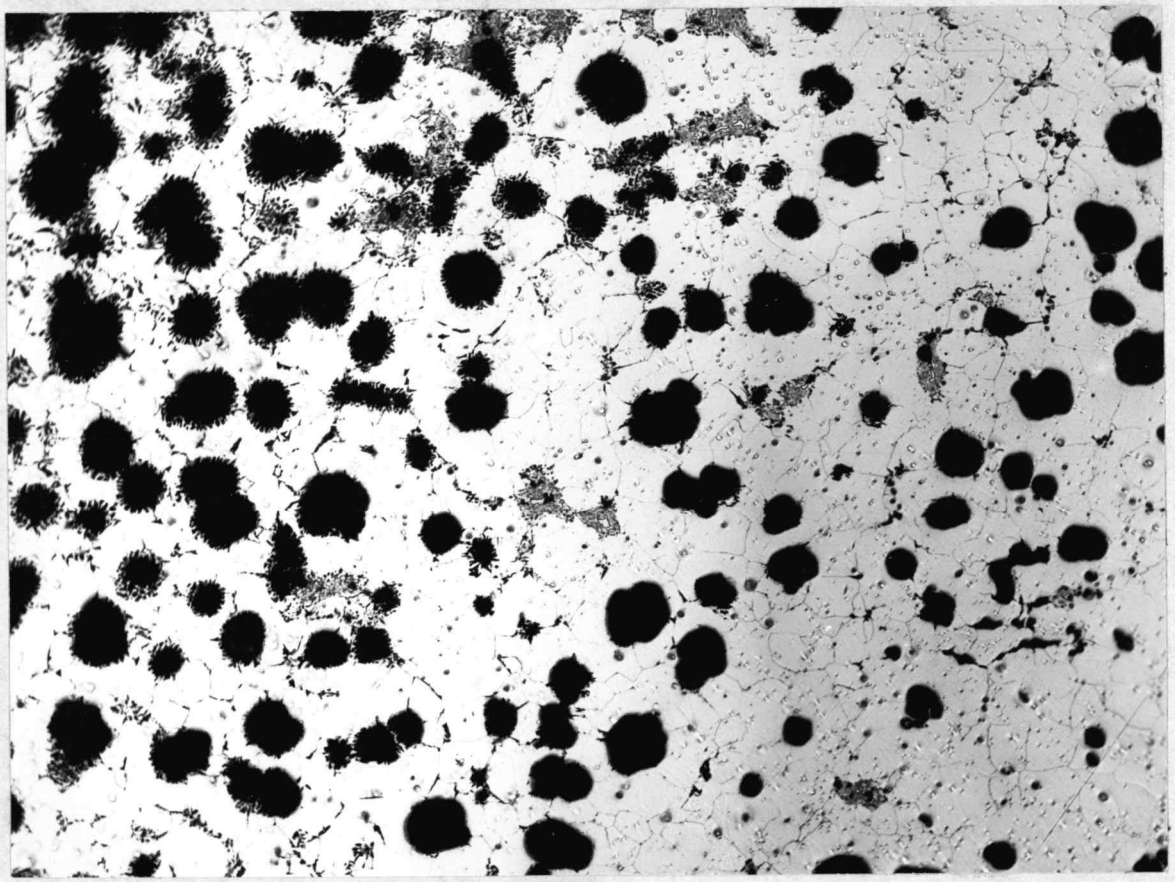

Figure 34. Edge of Heat Affected Zone (left). Weld with Mild Steel Electrodes. 100X.

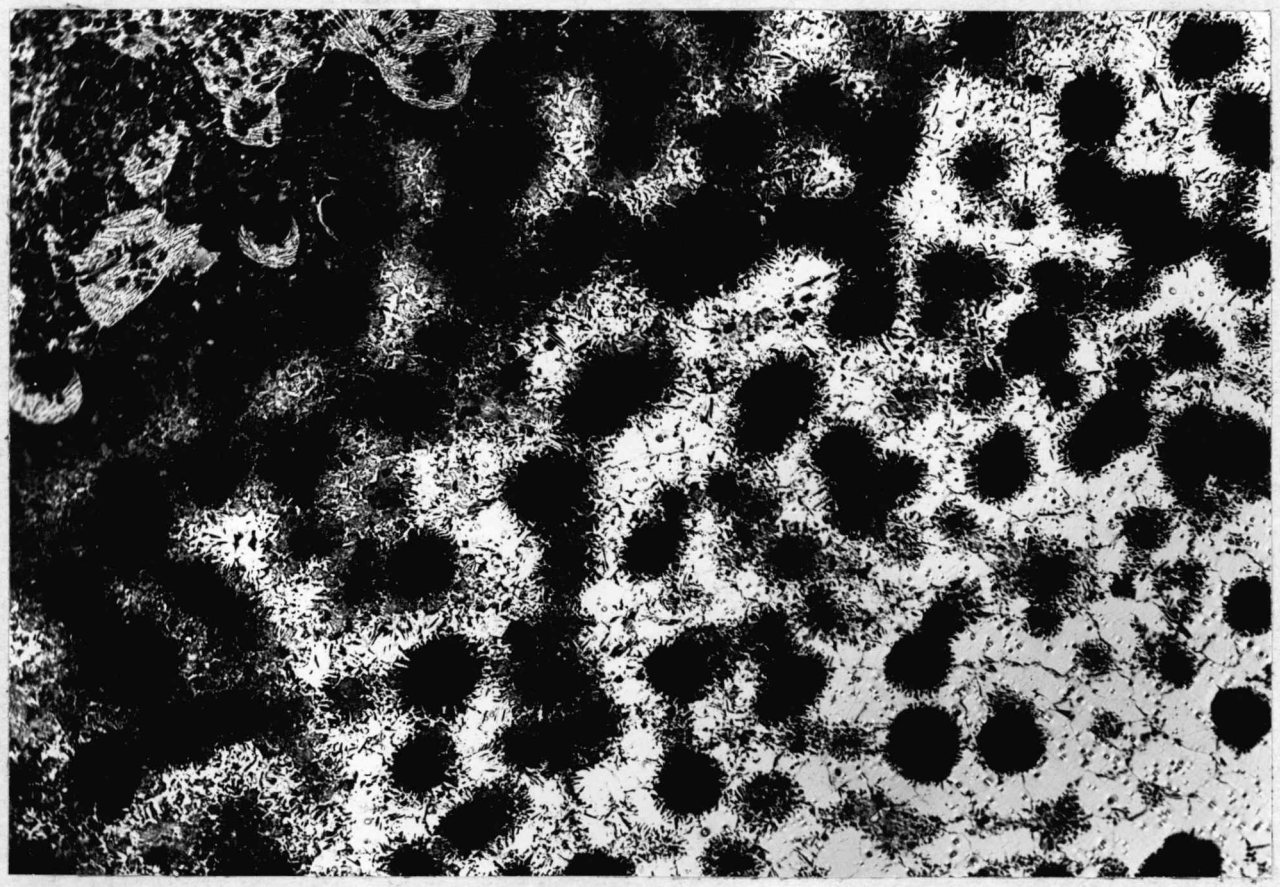

Figure 35. Heat Affected Zone (right) and Portion of Fusion Zone (left). 100X.

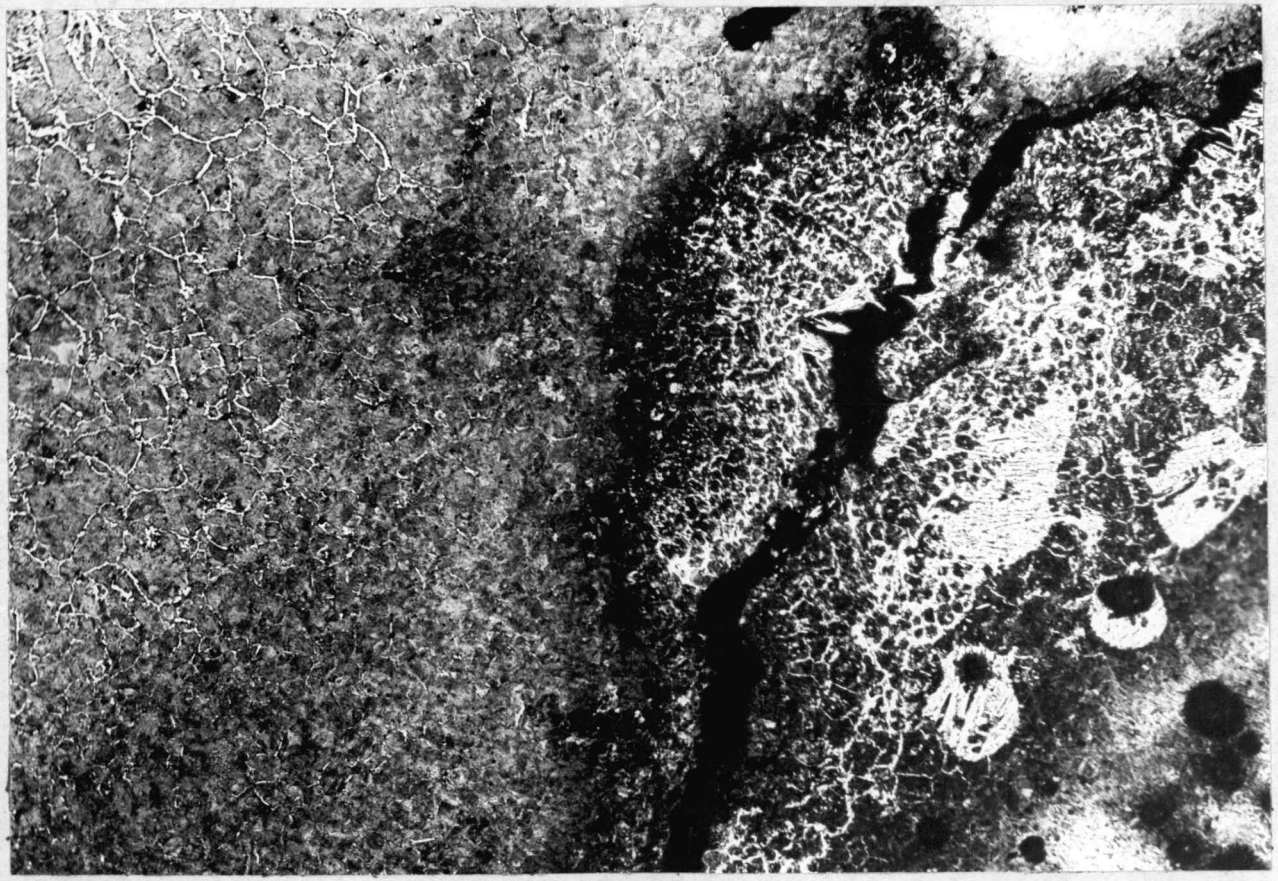

Figure 36. Fusion Zone (right) and Weld (left). 100X.

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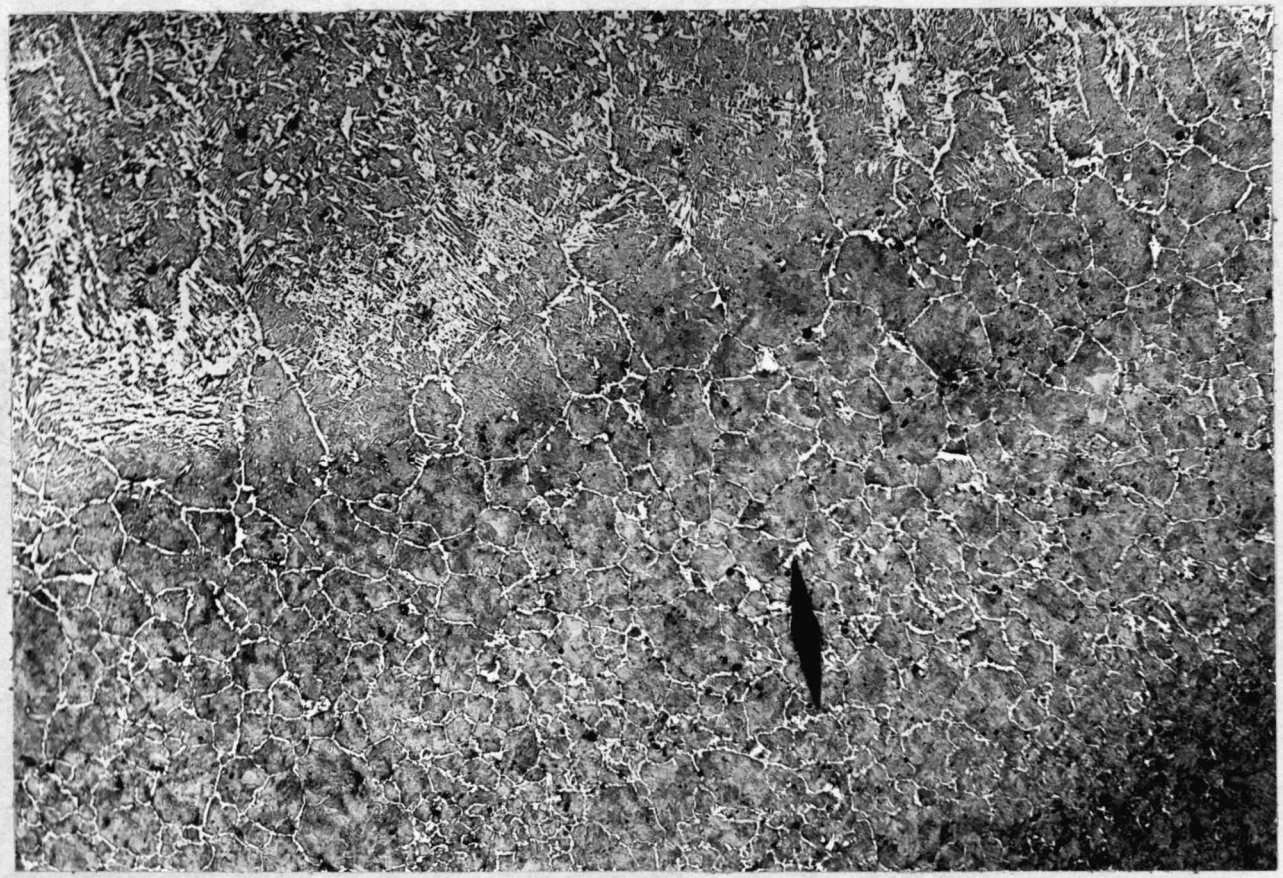

Figure 37. Farther into Weld Zone than Figure 36. 100X.

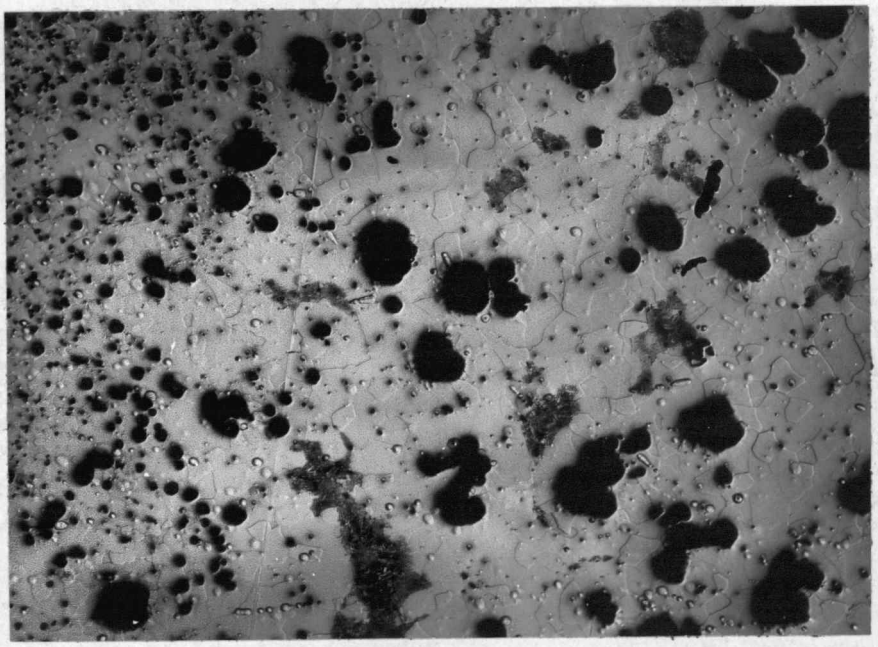

Figure 38. Parent Metal (right) and Renodulated Area (left) of Annealed Weld Made with Mild Steel Electrodes. 100X.

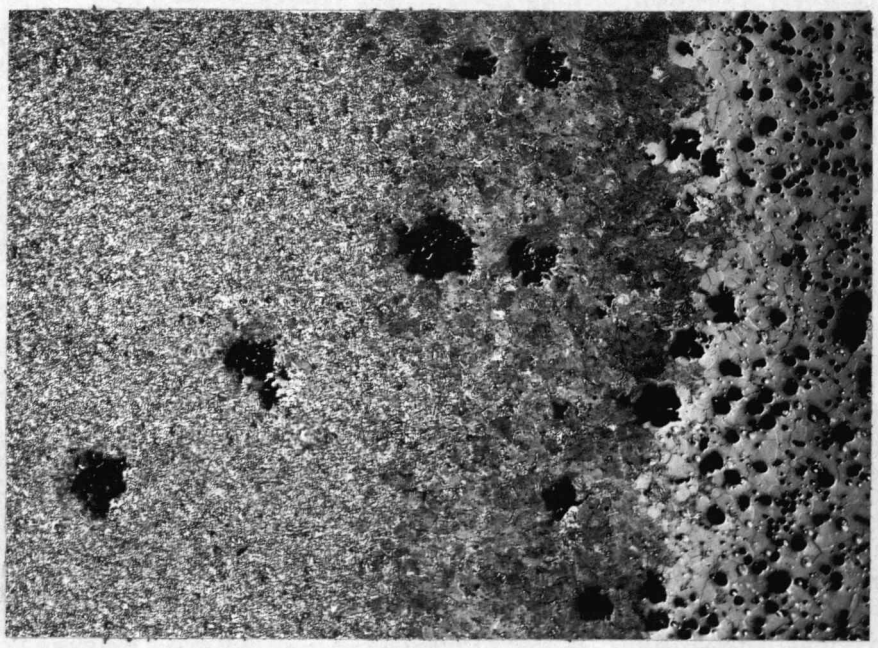

Figure 39. Fusion Zone of Specimen in Figure 38. 100X.

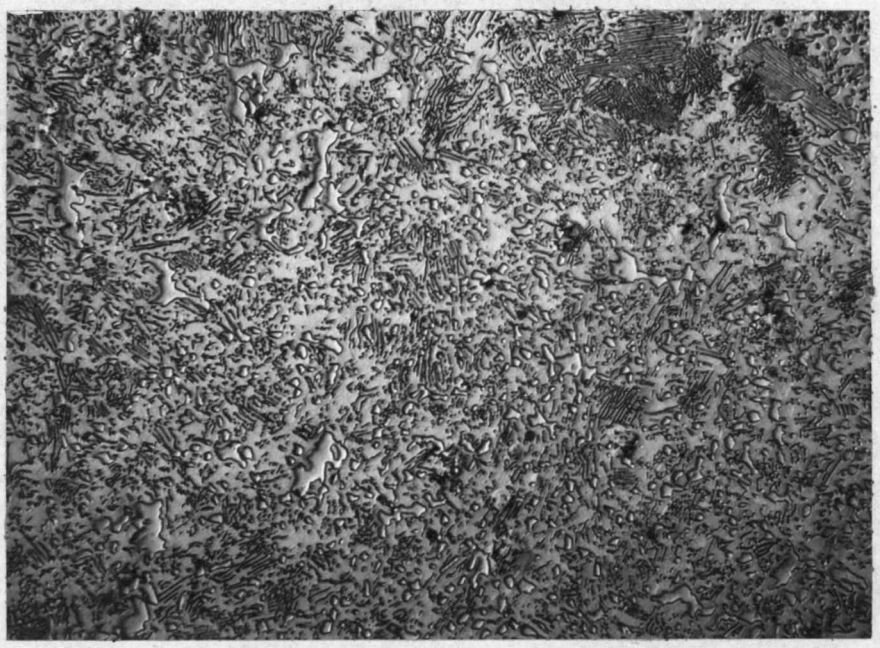

Figure 40. Fine Grained Region of Figure 39. 500X.

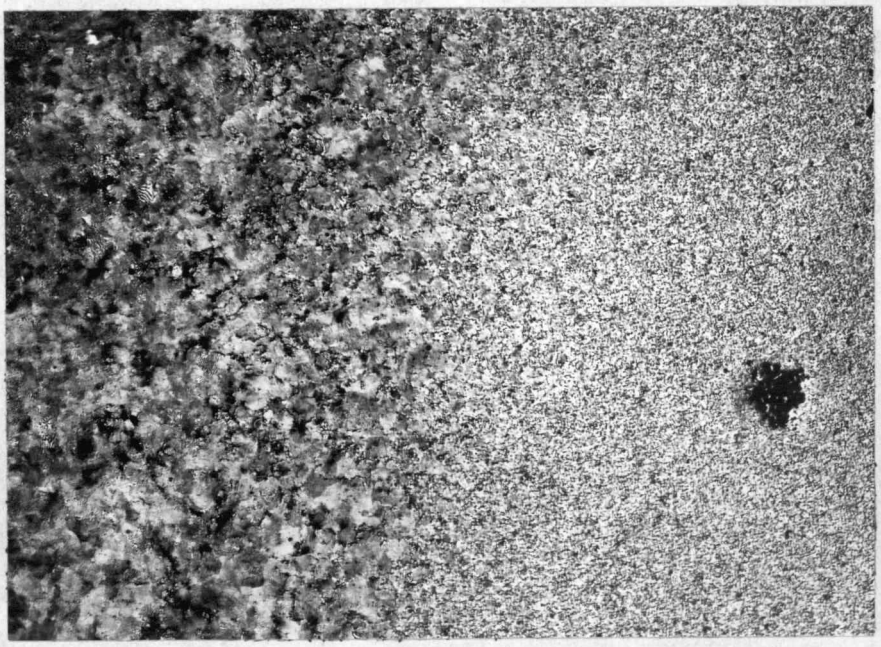

Figure 41. Farther into the Weld. 100X.

Figure 42. Coarse Grained Region of Figure 41. 500X.

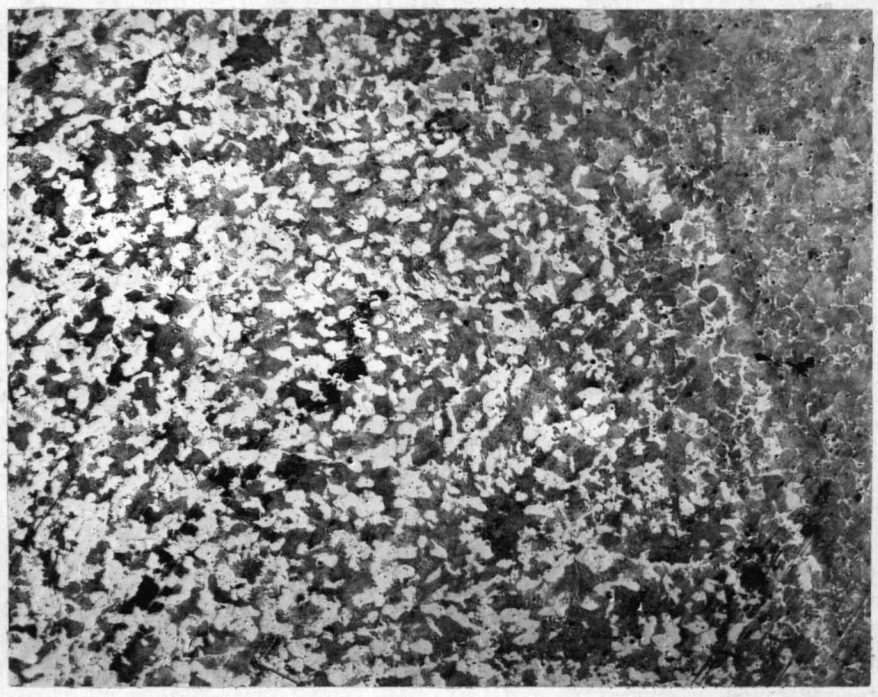

Figure 43. Farther into the Same Weld. Ferrite (white) and pearlite (dark). 100X.

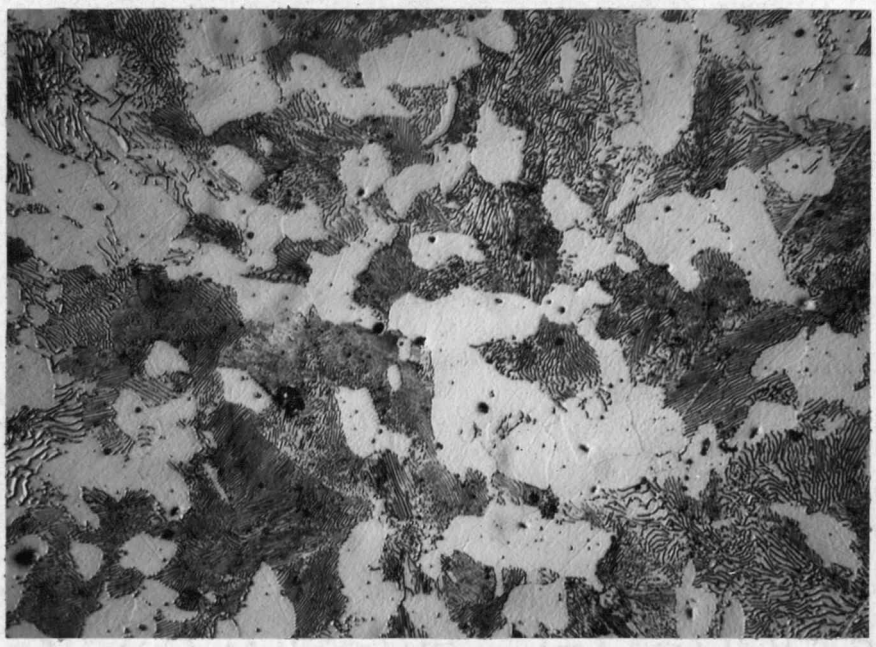

Figure 44. Same as Figure 43. 500X.

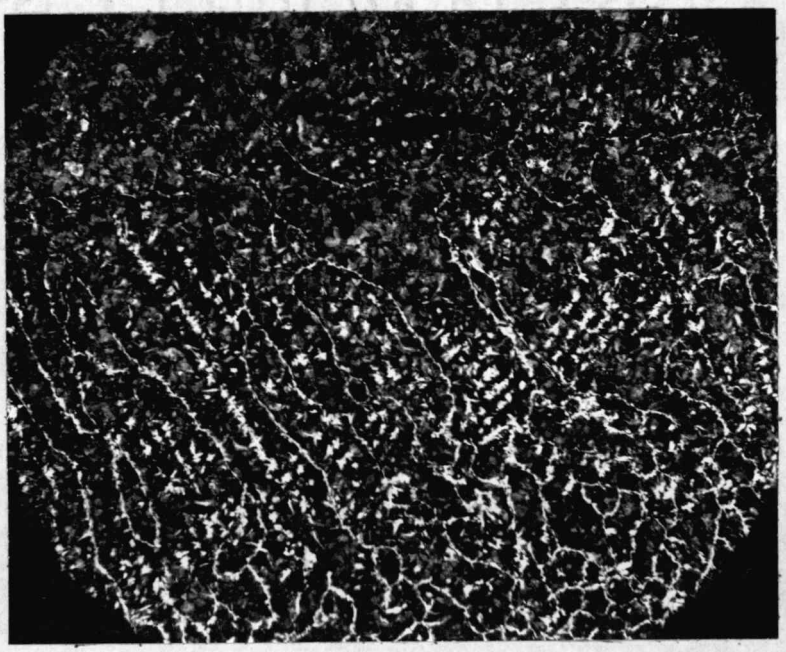

Figure 45. Weld Deposit Made with E-6015 Electrodes. 100X.

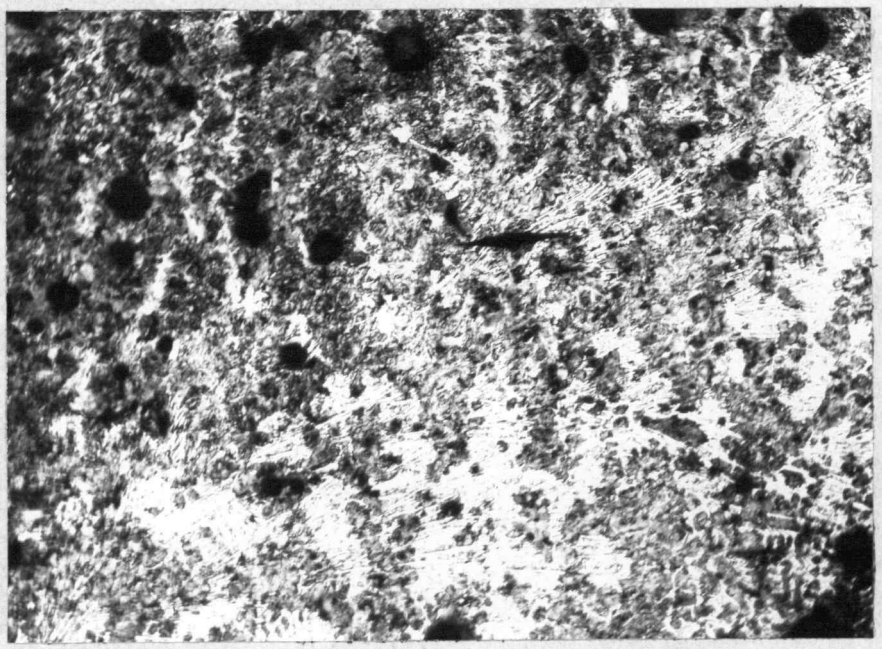

Figure 46. Parent Metal Heat Affected Zone of Weld in Figure 45. 100X.

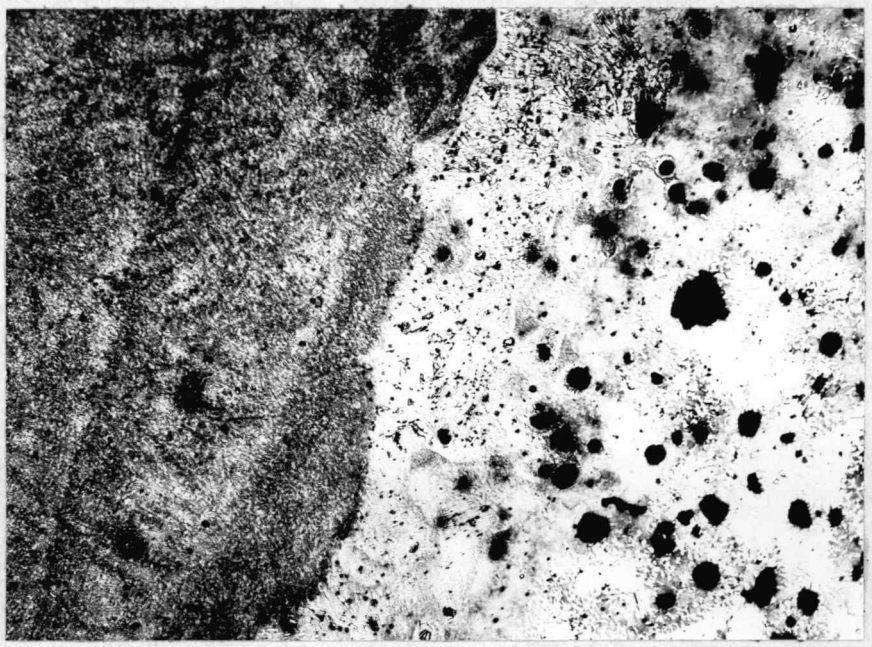

Figure 47. Heat Affected Zone When Base Metal is Preheated. 100X.

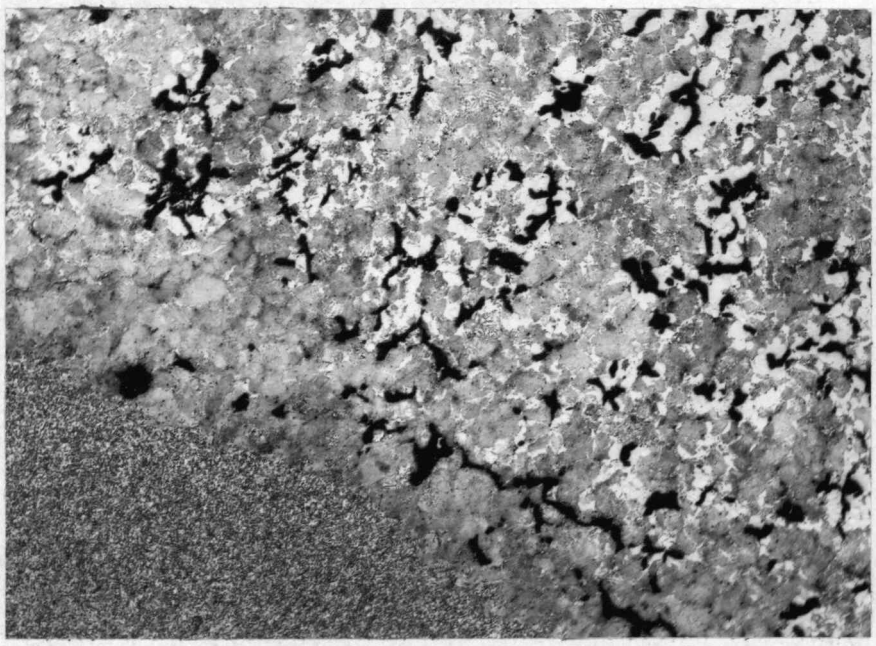

Figure 48. Weld on As-Cast Ductile Iron with E-6015 Electrodes. Fusion Zone. 100X.

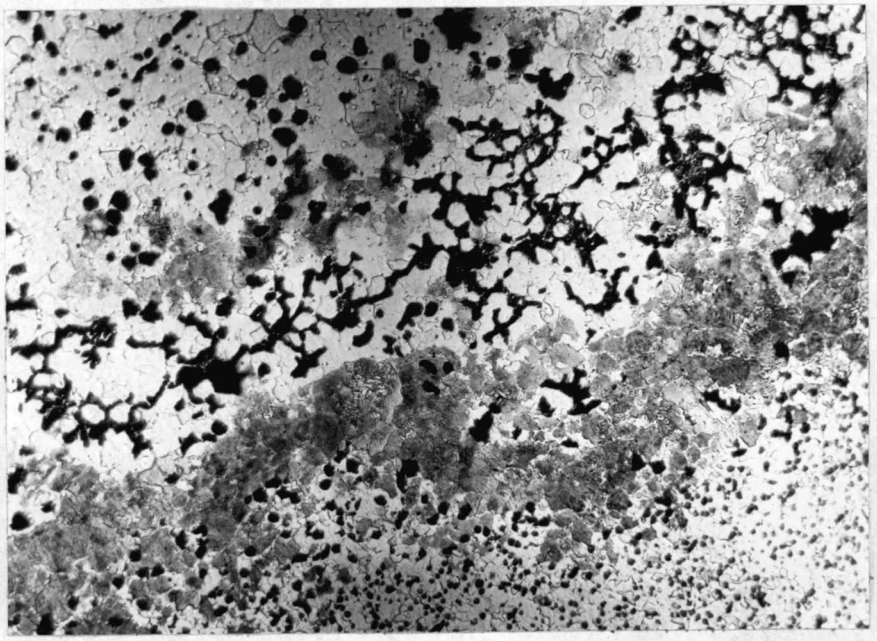

Figure 49. Fusion Zone, using E-8015 Electrodes. 100X.

Figure 50. Weld in Figure 49. 500X.

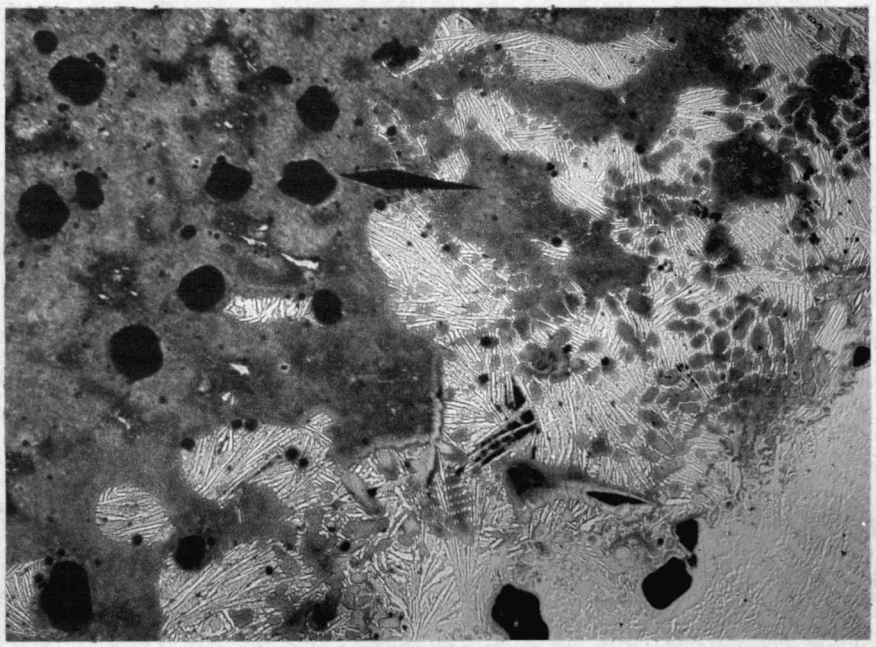

Figure 51. Heat Affected Zone. AWS 308-15 Electrodes. 100X.

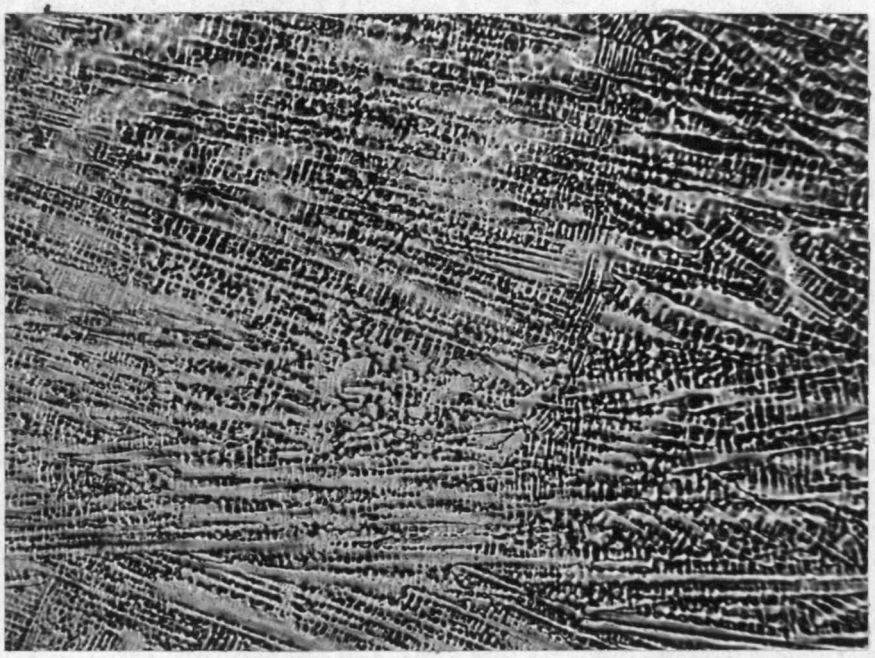

Figure 52. Weld Metal of Specimen in Figure 51. 100X.

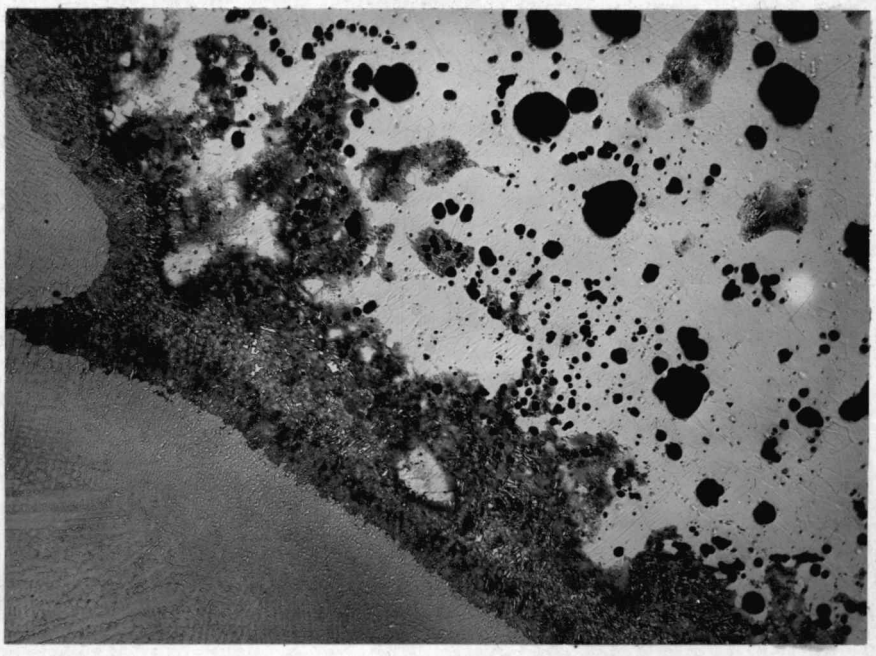

Figure 53. Fusion Zone of Ductile Iron as Cast, Welded, then Annealed. 308-15 Electrodes. 100X.

Figure 54. Carbide Network in Annealed Weld with 308-15 Electrodes. 100X.

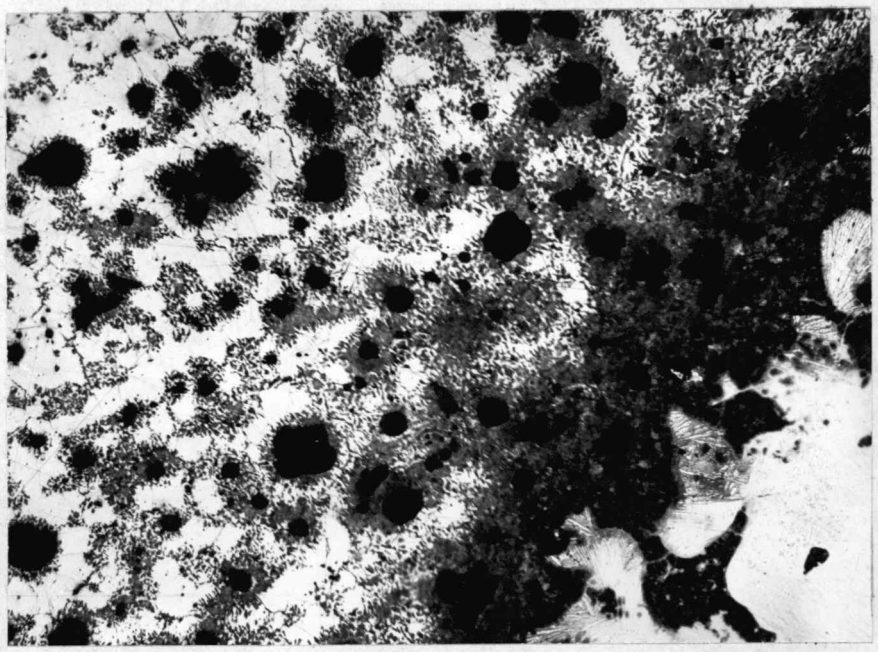

Figure 55. Fusion Zone of Weld with 310-15 Electrodes, without Preheating. 100X.

Figure 56. Preheated Fusion Zone. 100X.

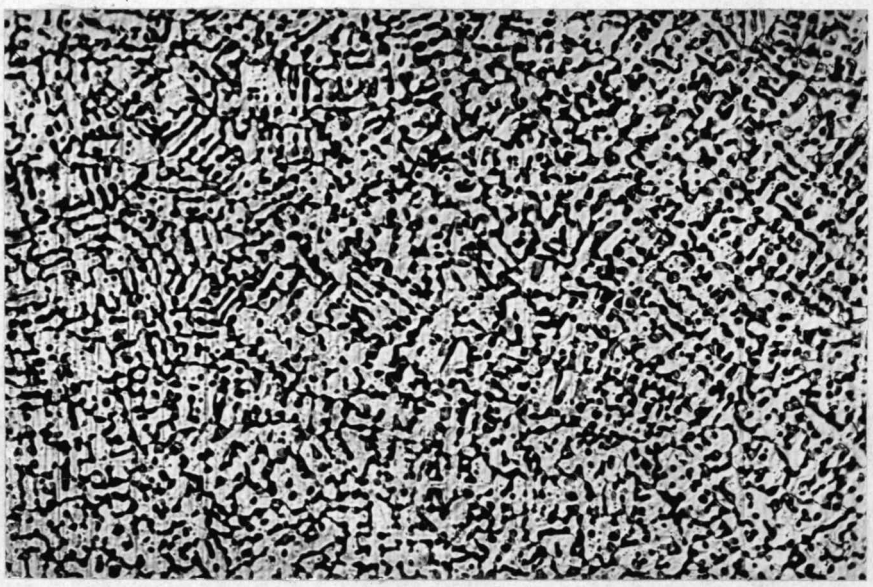

Figure 57. Weld Deposit of Type 310-15 Electrode.

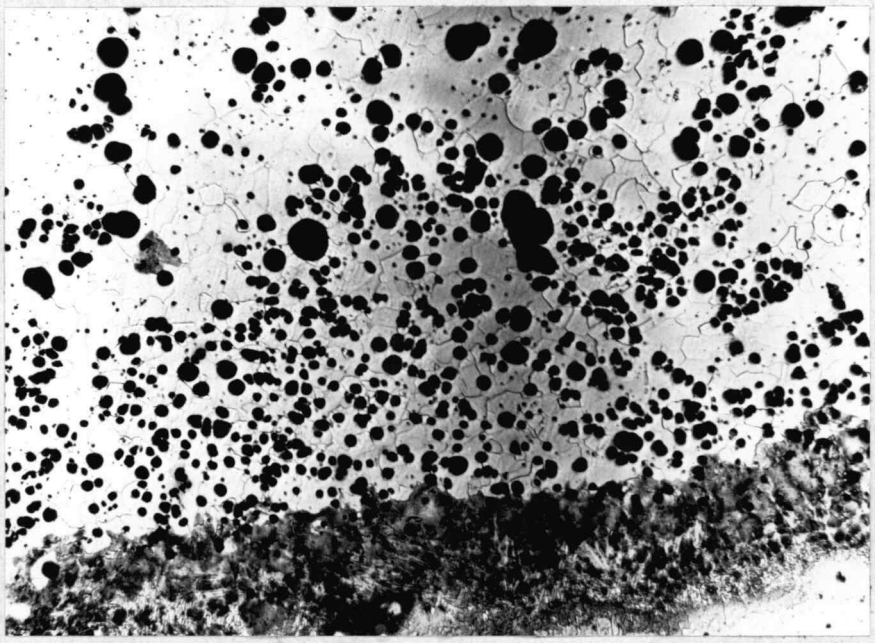

Figure 58. Annealed Fusion Zone of Weld made with 310-15 Electrodes. 100X.

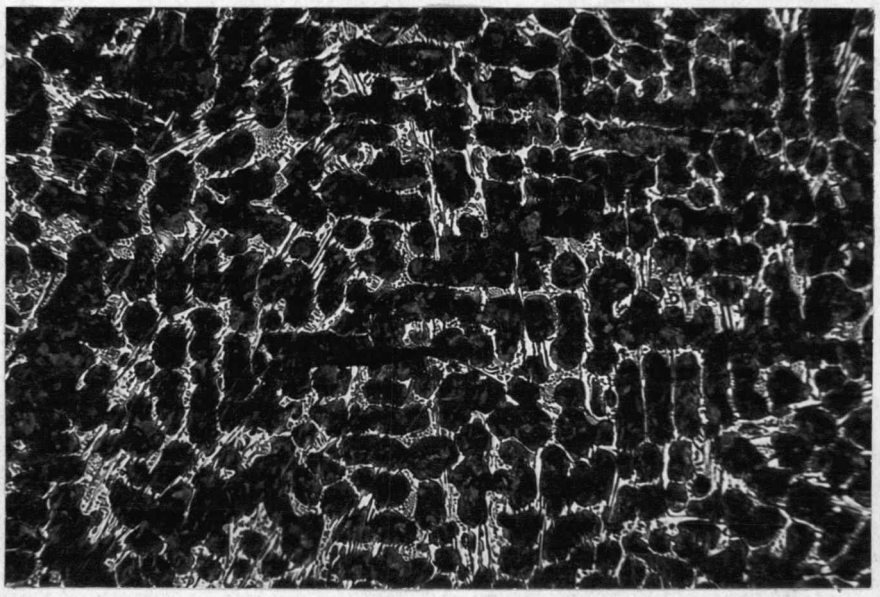

Figure 59. Oxy-Acetylene Weld Metal as Welded. 100X.

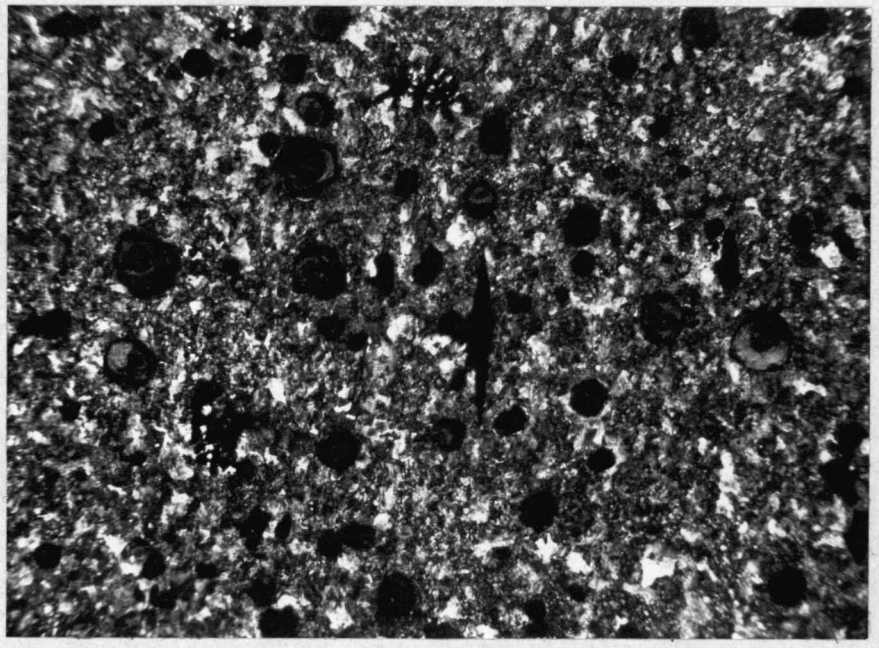

Figure 60. Parent Metal Heat Affected Zone of Weld in Figure 59. 100X.

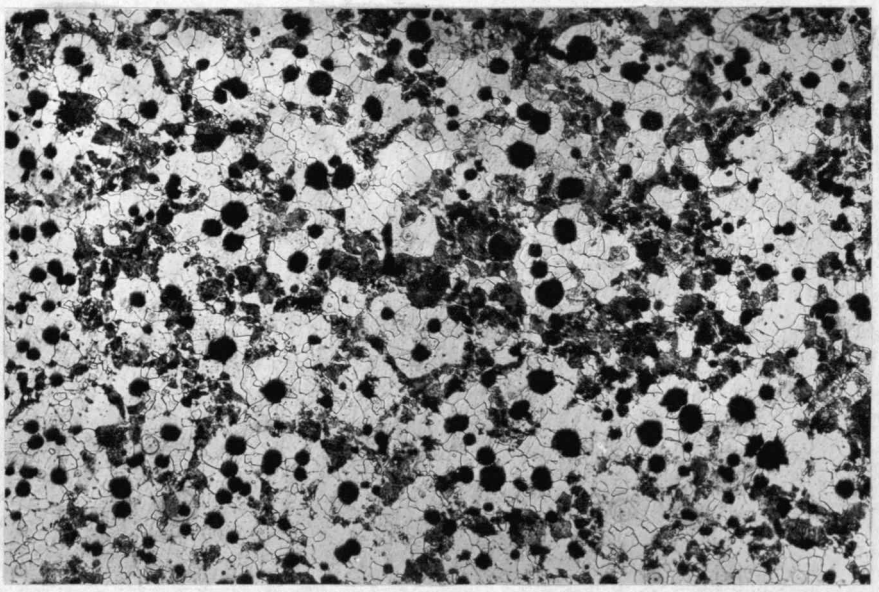

Figure 61. Annealed Oxy-Acetylene Weld. 100X.

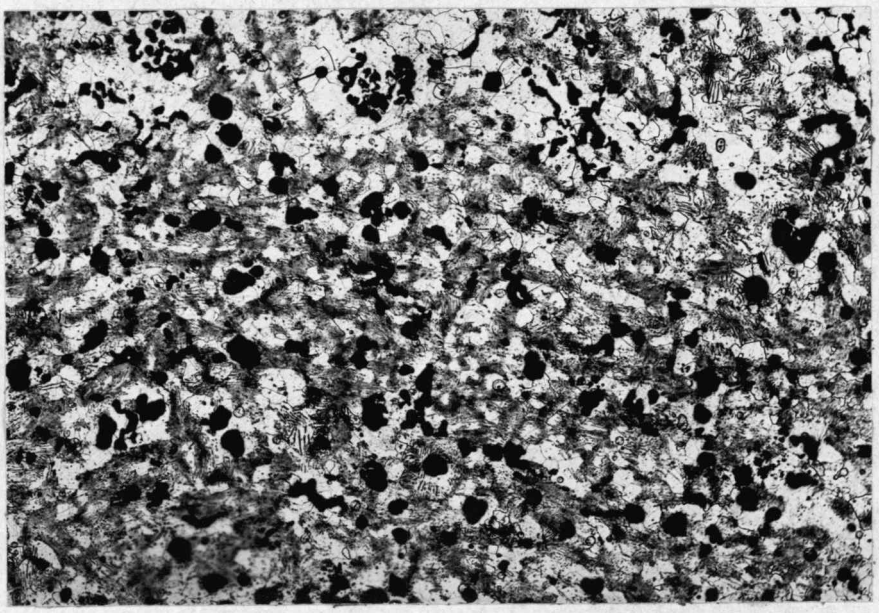

Figure 62. Annealed Heliarc Weld Deposit. 100X.

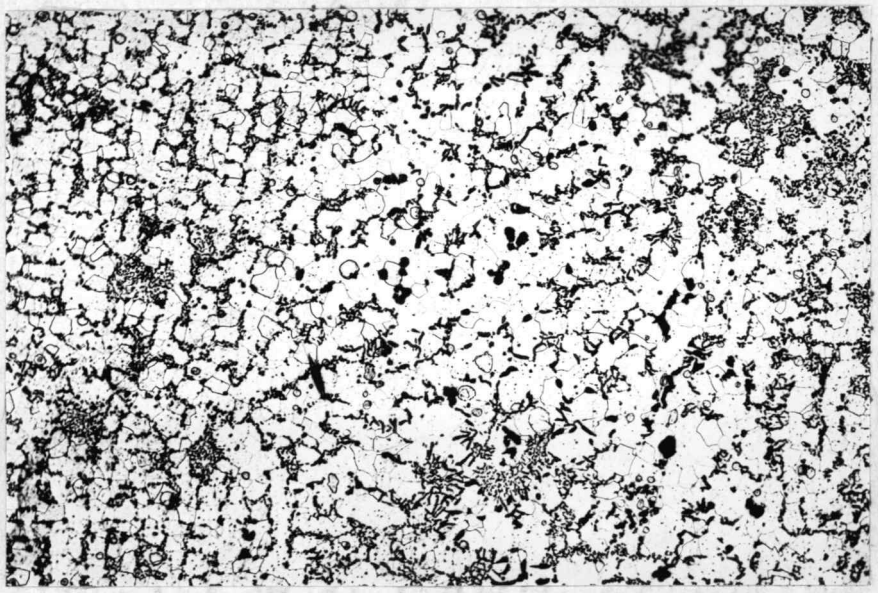

Figure 63. Same as Figure 62, in Area of Partial Graphitization. 100X.

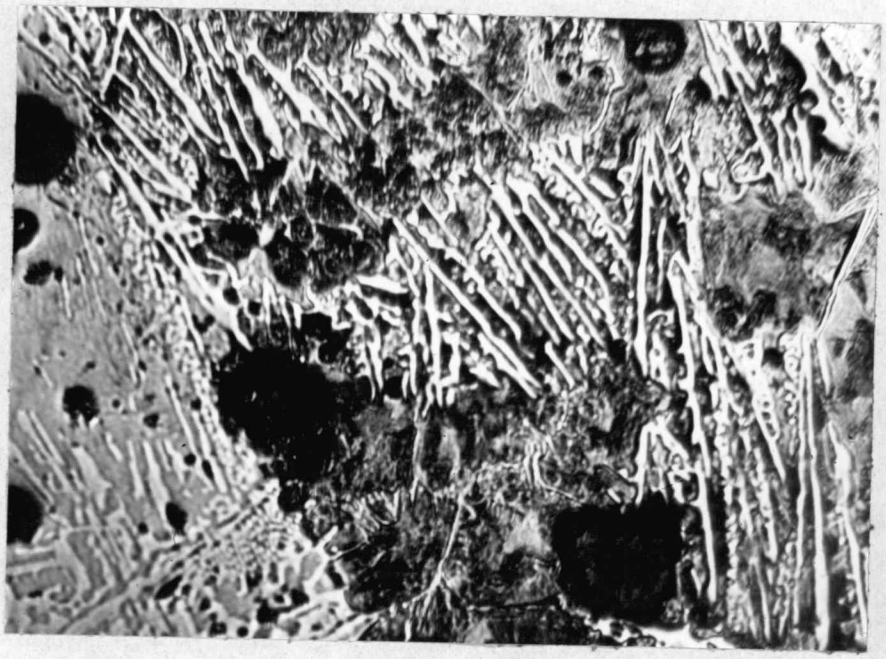

Figure 64. Fusion Zone. Ni-Rod 55 Electrodes. 500X.

Figure 65. Same as Figure 64 after Annealing. 100X.

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Figure 66. Weld Deposit. Ni-Rod 55 as Welded. 500X.

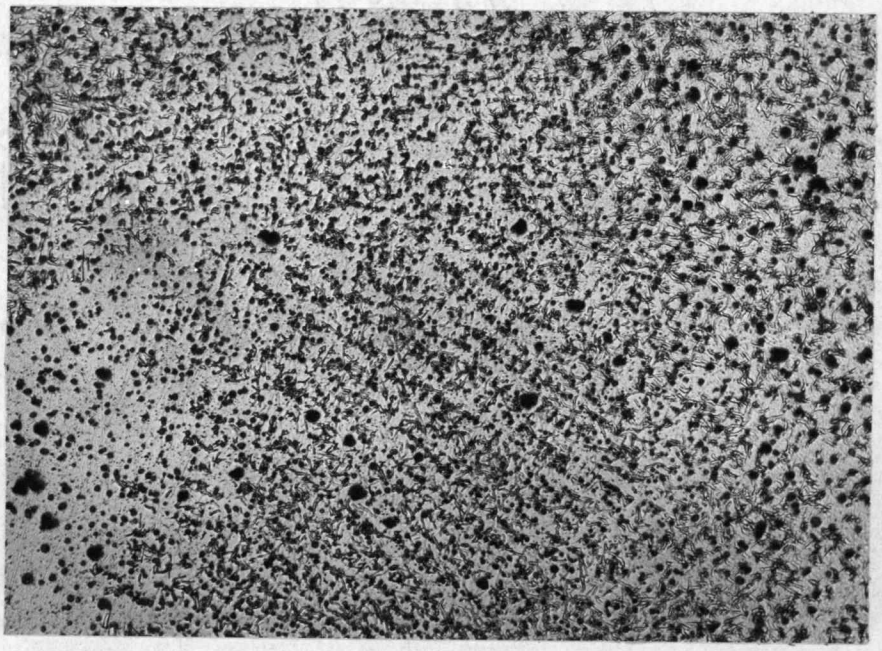

Figure 67. Weld Deposit. Ni-Rod 55 Annealed. 100X.

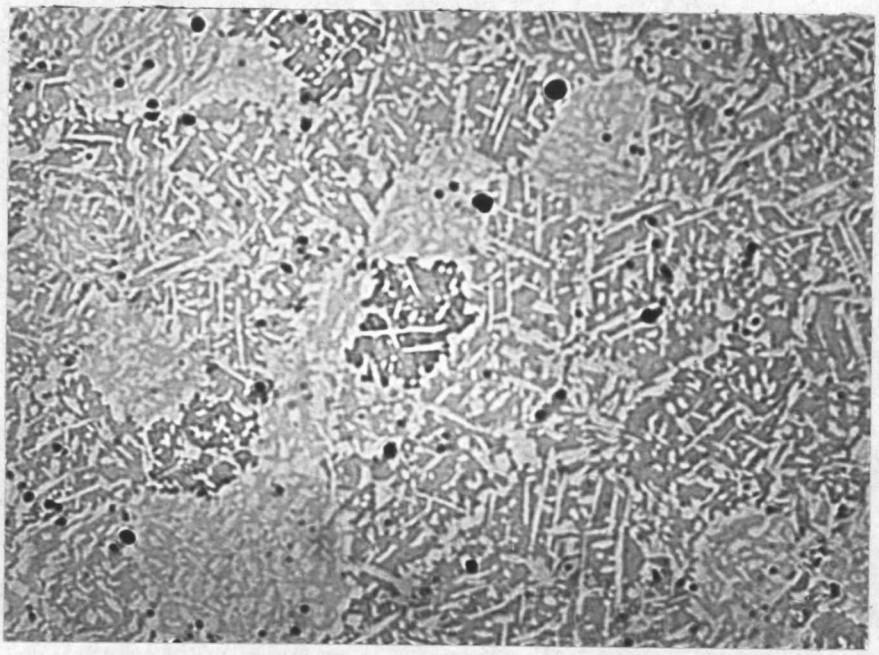

Figure 68. Manganese Bronze Weld Deposit. 100X.

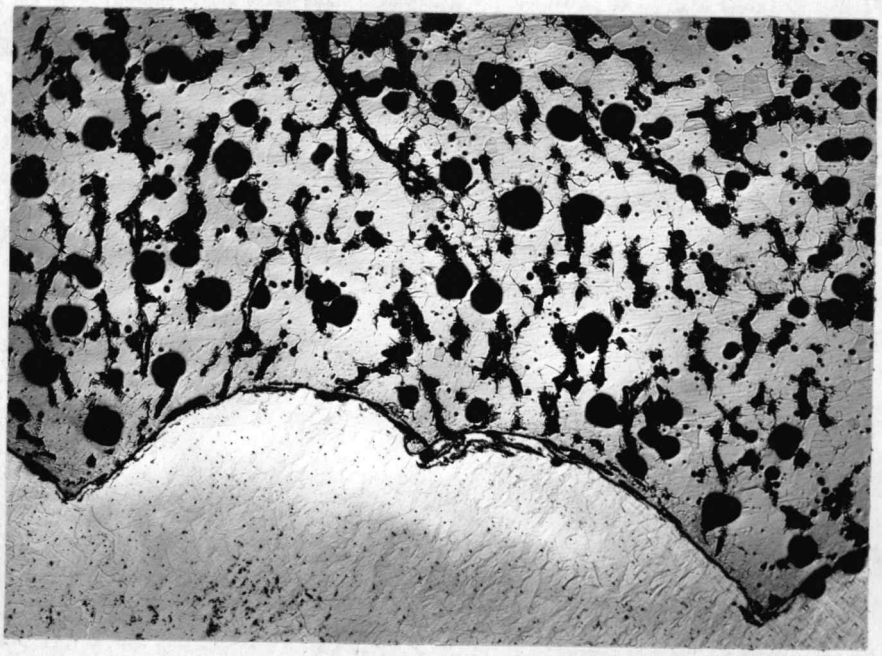

Figure 69. Ductile Iron (above) and Manganese Bronze Weld (below). 100X.

EVALUATION OF MECHANICAL TESTS

A very close correlation exists between the mechanical testing and metallography in this study. It would be virtually impossible to correctly interpret test results without the benefit of microscopic examination. Rather accurate predictions can frequently be made regarding the outcome of mechanical tests on the basis of a metallographic study.

A 60,000-pound Southwark-Emery hydraulic testing machine was used in making both tensile and flexure tests. All welded specimens were machined to rectangular cross sections for tensile tests while some parent metal specimens were of rectangular cross section and others were in the form of 0.505-inch diameter test bars. The preparation of all test bars was described in a previous section of this report.

The "tee" beam transverse flexure test bars were loaded by placing them on knife edges, flange down on 12-inch centers, and exerting a downward force on the stem at the center of the bar by means of a third knife edge. The load required to break these beams ranged from 20,000 pounds to 21,300 pounds, with an average of 20,770 pounds. The average deflection at fracture was 0.485 inch and the average outer fiber elongation was 4.0 per cent in two inches. However, due to the shape of the section being tested, this should represent a higher value in simple tension.

All of the ductile cast iron "tee" beams had a dark, brittle area of centerline shrinkage in the fracture, at the intersection of

the leg and flange. Some cope defects were also found in the bottom of the flange in the nature of drossy, sooty areas. Internal cracking in defective areas produced lower values than would have been obtained with perfect castings.

The reverse ends of the fractured parent metal "tee" beam bars were used to produce the composite ductile iron and mild steel "tee" beams. The average load supported by these beams was 18,125 pounds, which represented 88 per cent of the strength of the parent metal. Although failures were in the weld, this should be expected as the weld itself was at the point of highest applied stress. The center of the fracture was in the fusion zone of the root of the weld which was identified as a white carbidic area.

Cast reduced section tensile test bars in the annealed condition produced a maximum ultimate strength of 65,100 psi, a yield strength of 40,000 psi based on 0.1 per cent offset and 4.5 per cent elongation in two inches. The modulus of elasticity was 23×10^6 psi. Other specimens failed at lower values due to centerline shrinkage, slag inclusions and gas pockets as shown in Figure 70, with 48,500 psi ultimate strength being the lowest recorded value. Although these values compare favorably with those of mild steel, they would have been higher had it not been for defects in the castings.

The composite ductile iron and mild steel test bars produced a maximum ultimate tensile strength of 52,000 psi. Failures were primarily in the weld and carbidic fusion zones, although parent metal failures were obtained. The quality of the castings was such that the

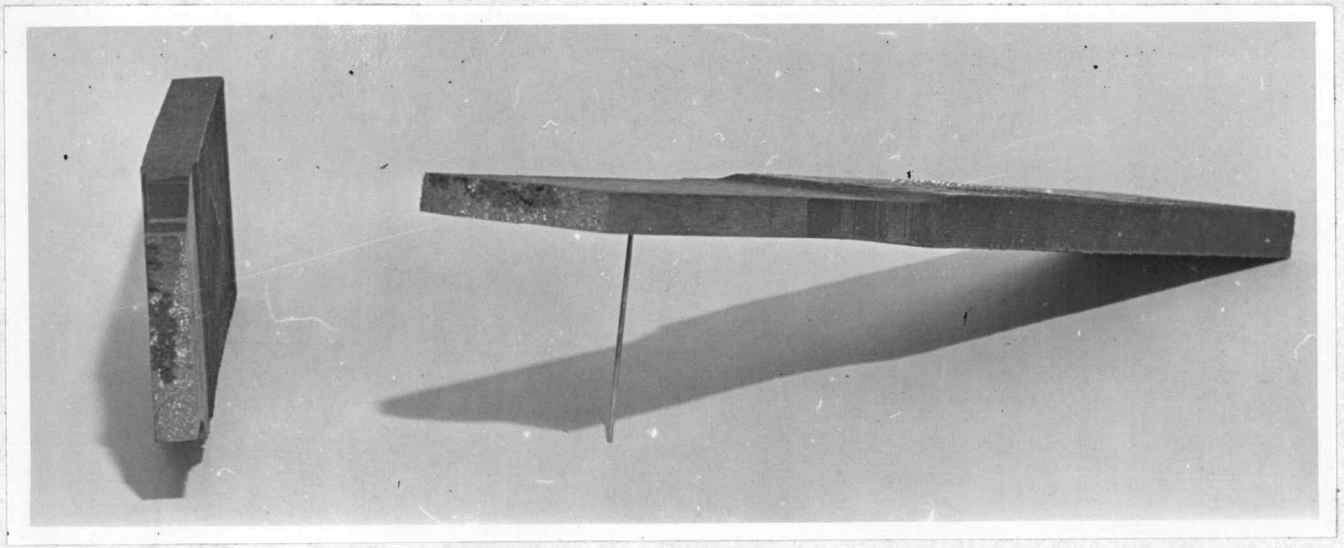

Figure 70. Defects in Broken Parent Metal Reduced Tensile Test Bar.

use of this type was discontinued.

The problem of welding ductile iron to steel is not so complex as that of welding ductile iron to itself. For this reason, subsequent testing was limited to ductile iron only, and more suitable "Y" block castings were produced for this purpose. The "Y" block castings produced ultimate tensile strengths of from 68,000 psi to 86,000 psi in the as-cast condition, with Rockwell C hardness values from 20 to 25. Typical test specimens are shown in Figures 71 and 72. Ultimate strength values from 55,000 psi to 84,000 psi were obtained for annealed castings, with Rockwell B hardness values from 80 to 90. These values depended on composition, which was discussed in detail in an earlier section of this report.

Parent metal flexure test specimens from one-inch "Y" block material and also from one-inch thick plate material, having rectangular cross sections of 1-1/2 inch x 3/4 inch were tested on 12-inch centers with third point loading applied to the width. The plate material was not homogenous and fractures occurred with little permanent set during the test. The "Y" block material had a yield stress of 23,900 psi in the outer fibers, and withstood a maximum applied load of 6,540 pounds at the maximum bending capacity of the equipment without fracture.

Flexure test specimens welded with Stainweld A-5 18-8 stainless steel electrodes failed with deflections of but 0.060 inch to 0.070 inch and with a maximum applied load of 3,300 pounds. Failures were in the carbidic fusion zone and are shown in Figure 73. The dark

Figure 71. Parent Metal As-Cast "Y" Block Tensile Test Bars.

Figure 72. Ends of Bars in Figure 71.
Few Defects are Noted.

areas represent surface fusion-line hot cracks. Rigidity afforded by the first weld passes which had contracted upon cooling prevented contraction of the hot outer weld layers and the cracks resulted. This condition, not experienced in the thinner sections, could be prevented by maintaining a higher preheat during the entire welding operation. The surface hot crack can be seen in flexure test specimen C3 prior to the test, Figure 74, and is not present in the thinner tensile test specimen B4. Due to the poor results obtained, it was decided to discontinue further flexure tests until the tensile testing program was completed, in order that more information would be available regarding the nature of the weldability.

According to the results of the tensile tests, the most promising methods of welding were fusion processes, using filler rod of composition that matched the parent metal. Welds made with the heliarc process produced an average ultimate tensile strength of 36,000 psi for those welded in the as-cast condition followed by annealing, compared with 47,000 psi for those welded after annealing and subsequently reannealed. It was assumed that the arc decomposed hydrocarbon gas in the as-cast parent metal, during the welding operation, and provided hydrogen for underbead cracking to a minor extent. These gases were removed to such an extent during the annealing operation that this difficulty was not experienced in the welding of the annealed castings. Examples of heliarc welded specimens, Figures 75 and 76, show how readily they were machined with high-speed steel milling cutters. The hardness of the weld, fusion zone and parent metal were all the same, Rockwell B-85.

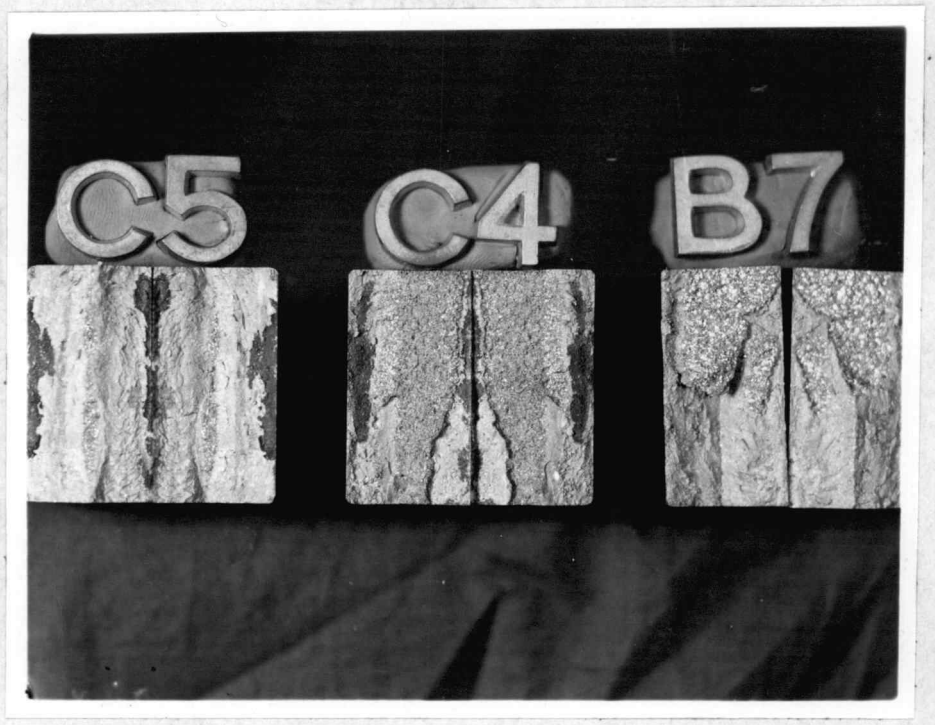

Figure 73. Flexure Test Bars After Fracture.

Figure 74. Flexure Test Bar (upper) and Tensile Test Bar (lower), before Fracture.

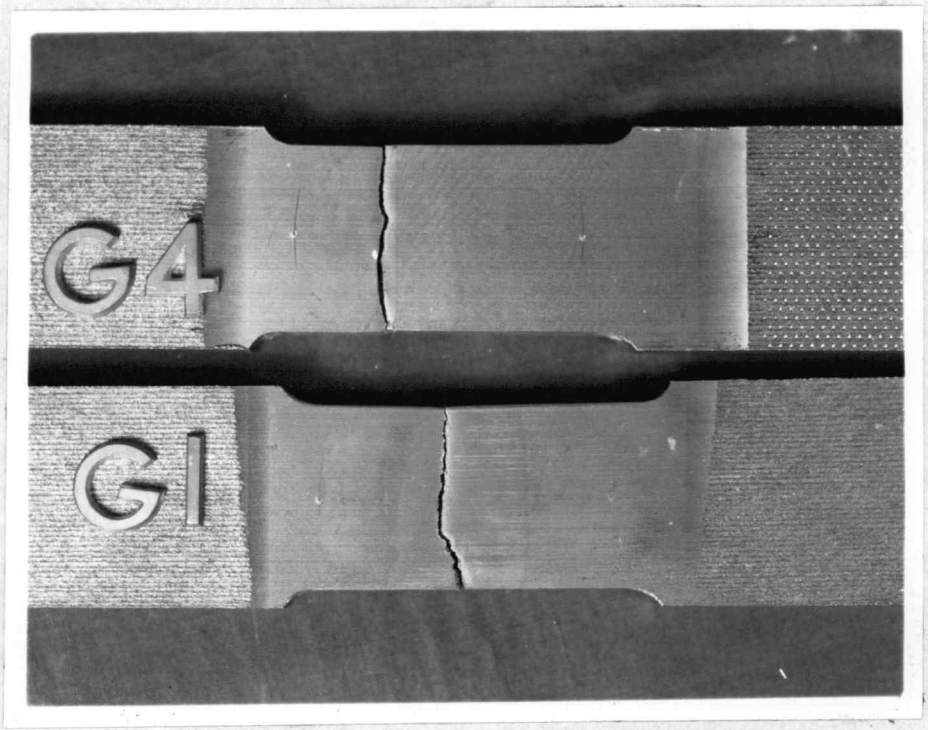

Figure 75. Fractured Heliarc Welded Test Specimen.

Figure 76. Ends of Test Bars Shown in Figure 75.

Specimens welded with the oxy-acetylene process and parent metal filler rod provided an average ultimate tensile strength of 56,000 psi in the as-welded condition and 64,000 psi after reannealing. These results are compared with those of the other processes in Figure 77. Annealed parent metal only was used in making the tests due to the experience with heliarc welding. The higher values for the reannealed specimens could be due to the higher content of silicon in both the base metal and filler rod. Blowholes present, particularly in the outer weld zone, were attributed to the flux used, which did not provide adequate fluidity of weld metal. The hardness of the weld, fusion zone and parent metal after annealing was Rockwell B-95, 96 and 90 respectively. The weld and fusion zone of the as-welded material was much harder, with respective values of C-34 and C-31.

The material that was carbon arc welded was of the same composition as the ductile iron that was oxy-acetylene welded and reannealed. An average ultimate strength of 67,500 psi, with a maximum value of 72,000 psi was the highest obtained with any process. The hardness of the weld and fusion zone was B-95. Some reversion to quasi-flake graphite during the heliarc welding process, which was not experienced with oxy-acetylene welding, led to the conclusion that this was a result of the high temperature of the welding process. As a result, lower current values were used for carbon arc welding and no graphite reversion was experienced.

The best results obtained with metallic electrodes were with Ni-Rod 55, having a nickel-iron core wire. Failure in the heat-affected zones was experienced with annealed ductile iron welded

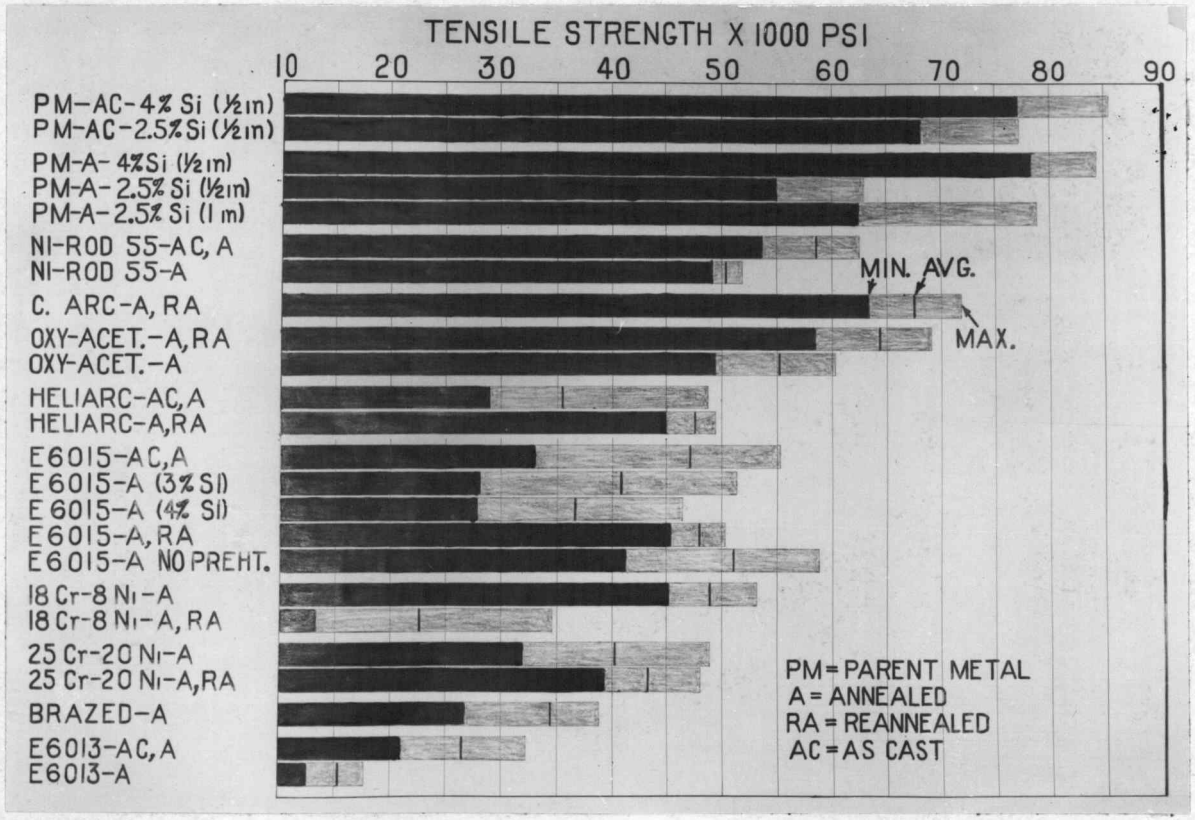

Figure 77. Results of Mechanical Tests of Welded Ductile Iron.

without preheat. The relatively low ultimate strength of slightly over 50,000 psi was attributed to the high phosphorus content of the base metal. Ductile iron of lower phosphorus content provided an average ultimate tensile strength of 58,000 psi when welded in the as-cast condition, followed by annealing. The hard narrow fusion zone (Rockwell C-40) of the as-welded material completely disappeared after annealing and had the same hardness as the base metal, Rockwell B-85. Annealing had no effect on the hardness of the deposited metal, which was Rockwell B-85 both before and after.

In general, the best results obtained with steel electrodes were with the low hydrogen mild steel electrodes of the AWS E-6015 type. Ductile iron with increased silicon content did not respond as well to the welding process as that with lower silicon content. Welding of the as-cast material followed by annealing was comparable in results with welding of annealed material without reannealing. The highest values obtained were with annealed ductile iron that was not preheated prior to welding.

The ultimate tensile strengths of all low hydrogen welds ranged between 26,000 psi and 58,000 psi with a general average of 45,000 psi. The hardness of the fusion zone, Rockwell C-45, was very high in the as-welded condition. After annealing, the fusion zone hardness dropped to Rockwell B-86. The average Rockwell C hardness of 35 to 39 for the as-welded deposited metal dropped to Rockwell C-25 after annealing. There was a strong tendency toward fusion zone failures in the as-welded condition, and weld failures after annealing. This was

attributed to the as-welded carbidic fusion zone, and to the carbide network in the hyper-eutectoid region of the annealed weld deposits.

The average ultimate strength value of 48,000 psi for welds with 18-8 stainless steel electrodes in the as-welded condition was slightly higher than the general average for low hydrogen mild steel weld deposits. This average fell to 23,000 psi when the welds were annealed, due to the chromium carbide network that precipitated in the weld zone. The weld deposit prior to annealing was relatively soft, Rockwell B-99. After annealing the weld was harder, Rockwell C-48, than the weld or fusion zone from any other process or heat treatment. The hardness of the fusion zone was Rockwell C-30 both before and after annealing. Failures were in the carbidic fusion zone in the as-welded condition and in the weld in the annealed condition.

Welds with 25-20 stainless steel electrodes provided lower ultimate strength than those made with low hydrogen electrodes or with 18-8 stainless electrodes in the annealed condition. All failures were in the fusion zone as shown in Figures 77 and 78, and annealing had little effect on the physical results. Annealed ductile iron welds in the as-welded condition had an average ultimate strength of 41,000 psi and those that were reannealed failed at 43,000 psi. Annealing had very little effect on removing the stable chromium carbides in the fusion zone, but it did reduce the Rockwell C hardness from 35 to 25. A network of chromium carbide resulted from the anneal, but the increase in hardness was very slight, from Rockwell C-23 to 26.

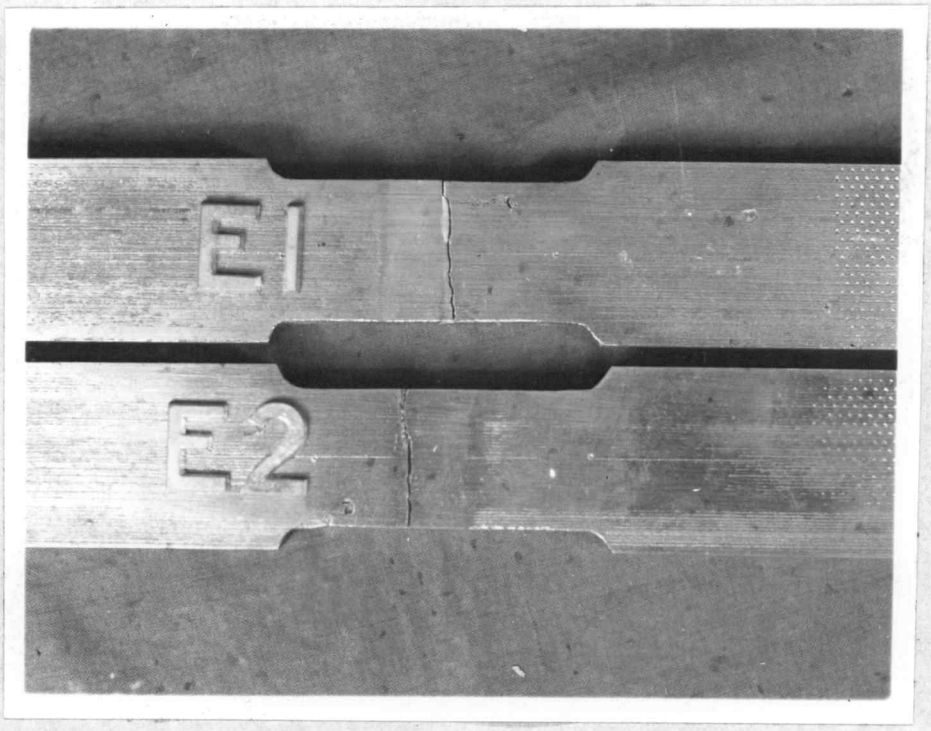

Figure 78. Fusion Zone Failure of Welds
Made With 25-20 Electrodes.

Figure 79. Ends of Fractured Bars Shown in Figure 78.

Welds with E-6013 electrodes were all very poor. A reaction took place between the base metal and electrode coating that resulted in entrapped slag in the fusion zone and also in underbead cracking. Welding was unsatisfactory on both annealed and as-cast material. Base metal with a high silicon content was much more difficult to weld than that of lower silicon content. The average ultimate strength of the low silicon base metal of 27,000 psi dropped to 16,000 psi when the silicon was increased one per cent. Annealing had no effect on reducing the hardness of the deposited weld metal, which was Rockwell C-20 both before and after heat treatment. The fusion zone, with a Rockwell C hardness of 28, was of the same hardness as the base metal, Rockwell B-85, after annealing.

The advantage of the increased weld surface area for the bronze welded specimens provided by the shear vee joint, was lost when too much metal was removed during the machining operations. The average ultimate strength of only 34,000 psi was considered quite low for this process under normal conditions, particularly when the reinforcing bead is not to be removed. The manganese bronze weld deposit was very soft with a Rockwell B hardness of 65. No heat effect was noted in the bond area between the weld and base metal, which had a hardness value of Rockwell B-88.

The stress-strain recorder was used only on annealed parent metal specimens, as premature weld failures could damage the equipment. The yield point of welded specimens, observed by a reduction in the rate of applied load, corresponded in general with the yield values obtained for the parent metal.

RECOMMENDATIONS AND CONCLUSIONS

The conclusions that are derived from the results of this investigation are threefold in nature. The scope of the study is very broad and years may be required to determine all of the answers to the problems involved. The tempo of research work in this field is increasing due to the demands of industry. The first conclusions then are in the form of recommendations for continuation of the program with the view of solving the remaining problems in a most efficient manner. The second conclusions are in the form of correlation between the work here and elsewhere. The remaining conclusions are in the form of recommended practices for welding of ductile iron.

With regard to the future program of investigation, the properties of fusion welding with ductile iron filler rod appear very good. Oxy-acetylene welding is perhaps best from a metallurgical standpoint, even if it is more expensive. One problem is to develop a flux suitable for this method of welding. The effect of adding silicon to the weld metal should be investigated. Response in general to other types of welding in the as-cast condition is poor. The oxy-acetylene process should be investigated in this regard.

The best test results were obtained with carbon arc welds. The heliarc welding process, although more expensive, should be as good as the carbon arc process. Some silicon may be lost during remelting of the ductile iron filler rod, and the effect of silicon additions to the weld should be studied. Additional tests should be made to either prove or disprove these items.

An anneal is needed to soften the fusion zone of welds made with iron base electrodes. The cementite network around grain boundaries resulting from a full anneal could perhaps be removed by a spheroidizing heat treatment. Heating for at least an hour at temperatures below the critical might suffice, or a prior normalizing heat treatment might be required. Low hydrogen electrodes produced by different manufacturers should be investigated to determine which is best adapted to ductile iron. The extent to which preheat is required should be studied.

The response of the base metal to welding with type 310-15A stainless steel electrodes was particularly good. Although higher test values were obtained when using low hydrogen mild steel and type 308-15 stainless steel electrodes, under certain conditions of heat treatment not included in this investigation the type 310-15 stainless steel electrodes might compare more favorably. It is recommended that spheroidizing heat treatments be included in a continuation of work with these electrodes.

The fact that no metallurgical changes were noted in the base metal as a result of brazing operations would warrant additional study for this technique. The use of different filler rods, particularly the aluminum bronzes, and use of different types of flux should be investigated. The type of joint and its method of preparation are other variables that could have a pronounced effect on the results obtained.

Use of non-ferrous electrodes other than the high nickel type was not included in the study. The aluminum bronzes have high

strength with good ductility and their use should be investigated. The reaction between the electrode coating and base metal is very important, therefore, the results of using aluminum bronze electrodes produced by different companies should be compared. In addition to metallic electrodes, non-ferrous filler rods applied with the double carbon arc process and alternating current should be investigated.

The microscope is one of the most valuable tools in studying the different techniques mentioned above. Much valuable time will be saved if a careful metallographic study is made prior to the conducting of time-consuming mechanical tests. A thorough investigation should be made of one process prior to proceeding to another. The variables of changing composition should be eliminated by obtaining a sufficient number of castings from one heat to conduct a large number of tests.

Mechanical properties other than tensile and hardness should be investigated. Foremost among these are impact, flexure and fatigue. Types of welded joints other than butt joints should be included in the study; for example, "tee" joints of ductile iron to ductile iron and of other metals to ductile iron. Tests of assemblies commonly used in fabrication, such as ductile iron fittings welded to steel tubing, would provide very practical information.

Some correlation has been obtained between the results of this study and other published results, although the scope of the work in general is different. The earliest published information (24, pp. 45-54) concerned ductile iron pipe welded with Ni-Rod 55.

Ultimate strengths of 47,250 psi to 58,900 psi obtained on bursting tests correspond with tensile strengths of butt joints obtained here of 49,000 psi to 62,500 psi.

In the second article published (16, pp.378-385S), the problem of Hucke and Udin was not complicated by heat transfer by the walls of grooved joints or by extensive heat-treating operations. While their observations regarding the undesirability of preheating operations could apply to their study of single-pass welds, the work at Oregon State College would indicate that for multi-pass welds preheating is not only desirable but in many instances it is essential. Their prediction that satisfactory fusion welding processes cannot be developed for ductile iron, which was quoted in section two of this report, has been proved erroneous by the results obtained here.

The investigation by Sohn, Boam and Fisk (34, pp.823-833) is of particular interest as it supplements this study. It indicates the results of welding ductile iron to steel with different types of electrodes, and of welding austenitic nodular iron. It is logical to anticipate higher physical values and more uniform results when welding relatively thin material, such as was used in their study, than when welding thick material with multi-pass joints such as were used here. The metallic electrodes that provided the most interesting results in their study, namely, type 310 stainless steel, low hydrogen mild steel and Ni-Rod 55, also produced promising results here. The post-weld heating of their welds to simulate operating conditions of jet engine parts would correspond to the spheroidizing heat treatment recommended in this study.

The only phases common between the International Nickel Company and Oregon State College programs concern the use of Ni-Rod 55 and low hydrogen mild steel electrodes. Their testing program (18, pp.947-956) was largely built around the use of Ni-Rod 55 electrodes, which is their product. Their work in this regard was more extensive than that done at Oregon State College; however, the results are comparable. Their use of low hydrogen electrodes was very limited and the values reported were higher than those obtained here. The reason for this is probably due to a difference in electrodes used or in post-weld heat treatment.

The most recent publication, May 1954 (26, p.49), was with regard to salvage of defective castings by means of the carbon arc. The North American Aviation Company substitute a carbon electrode for the tungsten electrode in a heliarc torch. The work here was performed with the carbon electrode in a metallic electrode holder and without the use of inert gas atmosphere. They report the use of cast iron welding flux; however, its use was not considered essential here. They reported hard deposits in the weld with use of heliarc welding with tungsten electrodes, while identical conditions here produced soft machinable welds. While several differences in technique exist between the two programs, and the North American Aviation Company have apparently made no mechanical tests of their welds, the conclusions of both parties are that the carbon arc is the most promising means of salvage of defective castings.

It is concluded from the results of the tests that fusion welding processes with ductile iron filler rod provide the most promising means of salvage of defective ductile iron castings, repair of broken castings or of welding simple component castings together to produce complicated built-up cast shapes. Of these processes, the highest rate of production can be obtained with the carbon arc. It is easy to manipulate and is also the least expensive. The best test results obtained were with the use of the carbon arc. It is recommended that preheats on the order of 1000 F be used with these fusion processes and that welding be done only on annealed castings. The castings should be reannealed subsequent to welding.

The use of Ni-Rod 55 electrodes is recommended for the purposes mentioned in the preceding paragraph, if it is desired to eliminate preheating and postheating operations. These are the only electrodes that can be recommended for use on ductile iron in the as-cast condition according to the findings in this investigation. Preheating is desirable for multi-pass welds and post-weld heat treatment should be used if possible when these electrodes are used.

Steel electrodes that contain cellulose or moisture in the coating should not be used for welding ductile iron under any circumstances. Ductile iron that is to be welded should contain less than 0.05 per cent phosphorus and less than 3.5 per cent silicon.

Metallic electrodes are recommended for integration of ductile iron with wrought steel. The best general choice for this purpose would be Ni-Rod 55 electrodes. If the high cost of these electrodes is objectionable, the low hydrogen mild steel electrodes are

recommended. Some sacrifice of strength and shock resistance may be encountered with the use of steel electrodes. Although spheroidizing heat treatments were not used in this investigation, they would probably be the best form of post-weld heat treatment. If ductile iron is to be welded to stainless steel, AWS type 310-15 electrodes appear to be a good choice. Ni-Rod 55 electrodes are also suitable for this purpose and they would be preferred for welding ductile iron to non-ferrous metals.

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APPENDIX

TEE BEAM FLEXURE TEST DATA

Beam 2x2x3/8 in., 16 in. long, 12 in. span, simple support center point loading.

Parent Metal

Heat 7 Load lb	Heat 7 Deflection in.	Heat 8 Load lb	Heat 8 Deflection in.	Heat 9 Load lb	Heat 9 Deflection in.
0	0.000	0	0.000	0	0.000
1000	0.027	1000	0.018	1000	0.017
2000	0.034	2000	0.025	2000	0.024
3000	0.038	3000	0.030	3000	0.030
4000	0.043	4000	0.036	4000	0.036
5000	0.047	5000	0.041	5000	0.042
6000	0.051	6000	0.047	6000	0.048
7000	0.056	7000	0.053	7000	0.054
8000	0.062	8000	0.059	8000	0.060
9000	0.070	9000	0.066	9000	0.067
10000	0.079	10000	0.074	10000	0.074
11000	0.092	11000	0.083	11000	0.082
12000	0.109	12000	0.094	12000	0.092
Fracture	21,300	Fracture	21,000	Fracture	20,800
Defl.	0.456	Defl.	0.635	Defl.	0.470

Welded Specimens

Heat 9 Load lb	Heat 9 Deflection in.	Heat 7 Load lb	Heat 7 Deflection in.	Heat 7 Load lb	Heat 7 Deflection in.
0	0.000	0	0.000	0	0.000
1000	0.020	1000	0.023	1000	0.026
2000	0.029	2000	0.031	2000	0.036
3000	0.036	3000	0.039	3000	0.044
4000	0.043	4000	0.047	4000	0.051
5000	0.050	5000	0.054	5000	0.058
6000	0.056	6000	0.062	6000	0.066
7000	0.062	7000	0.070	7000	0.074
8000	0.068	8000	0.078	8000	0.083
9000	0.075	9000	0.088	9000	0.093
10000	0.082	10000	0.102	10000	0.106
11000	0.090				
12000	0.100	Fracture	18,000	Fracture	16,400
13000	0.111				
Fracture	20,000				
Defl.	0.420				

TEE BEAM FLEXURE TEST DATA (CONTINUED)

Welded Specimens
Heat 8

Load lb	Deflection in.
0	0.000
1000	0.015
2000	0.023
3000	0.031
4000	0.038
5000	0.046
6000	0.054
7000	0.062
8000	0.072
9000	0.084
Fracture	18,250

TENSILE TEST DATA

Parent Metal

Ultimate Strength psi	Elongation % in 2 in.
63,000	4.0
48,200	1.0

Welded Specimens

Ultimate Strength psi	Elongation % in 2 in.
52,000	4.0
45,000	3.5
46,400	5.0
37,700	Broke outside gage

ROCKWELL HARDNESS DATA

Specimen: Ductile iron welded to mild steel with Ni-Rod 55

Procedure: Readings taken at intervals of 1/16 in. running completely across the weld.

Readings: Rockwell C scale

Set 1		Set 2		Set 3	
Read. No.	Rockwell C	Read. No.	Rockwell C	Read. No.	Rockwell C
1	13	1	18	1	19
2	13	2	18	2	18
3	12	3	24	3	26
4	14	4	26	4	31
5	15	5	26	5	33
6	14	6	25	6	36
7	15	7	25	7	20
8	15	8	23	8	21
9	15	9	23	9	18
10	17	10	26	10	18
11	20	11	30	11	16
12	24	12	26	12	13
13	21	13	29	13	16
14	21	14	31	14	17
15	20	15	35	15	18
16	21	16	27	16	15
17	20	17	23		
18	21	18	21		
19	24	19	23		
20	34	20	23		
21	16	21	23		
22	29	22	22		
23	25				
24	34				
25	25				
26	16				
27	18				
28	18				
29	18				
30	18				

ROCKWELL HARDNESS DATA (CONTINUED)

Set 4		Set 5	
Read. No.	Rockwell C	Read. No.	Rockwell C
1	22	1	30
2	26	2	32
3	28	3	29
4	29	4	30
5	29	5	32
6	33	6	32
7	42	7	32
8	28	8	37
9	27	9	44
10	27	10	37
11	26	11	28
12	26	12	28
13	27	13	28
14	26	14	29
15	31	15	28
16	28	16	28
17	27	17	25
18	26	18	26
19	26	19	26
20	26	20	24
21	26	21	26
22	26	22	26
23	25	23	26
24	26		

KNOOP MICRO-HARDNESS DATA

General data: Knoop Indentor, 1.000 kg load, 16 mm objective, multiplication factor 0.699

Distance mm/100	Reading Initial	Reading Final	Reading Net	Length of Indentation Microns	Length of Indentation mm	Knoop Hard. No.
0	313	692	379	265	0.265	202.6
50	303	687	384	268	0.268	198.1
100	268	709	741	309	0.309	149.0
150	267	702	435	304	0.304	154.0
200	289	681	393	275	0.275	188.2
250	282	683	401	280	0.280	181.5
300	255	706	451	315	0.315	143.4
350	203	757	552	386	0.386	124.5
400	219	696	477	333	0.333	167.0
450	197	706	509	356	0.356	146.3
500	245	693	448	314	0.314	188.0
550	236	689	453	318	0.318	183.0
600	217	714	497	348	0.348	153.0
650	217	645	428	299	0.299	207.0
700	204	746	542	372	0.372	134.0
750	204	712	508	355	0.355	147.1
800	194	724	530	370	0.370	135.5
850	248	684	436	304	0.304	201.0
900	197	724	527	368	0.368	136.9
950	214	713	496	349	0.349	152.0
1000	233	699	466	319	0.319	182.0
1050	197	714	517	354	0.354	148.0
1100	274	691	417	291	0.291	219.0
0	255	674	419	293	0.293	165.7
15	253	659	406	284	0.284	176.4
30	344	563	219	153	0.153	607.8
45	318	589	271	189	0.189	398.3
60	280	630	350	244	0.244	239.0
75	290	617	327	228	0.228	273.7
90	294	600	306	214	0.214	313.6
105	281	644	363	254	0.254	220.5
120	281	646	365	256	0.256	217.1

FLEXURE TEST DATA*

Specimen B5-1 1.490 in. x 0.730 in.

Load in lb	Total Deflection in inches	Unit Deflection in inches	Remarks
500	0.022	0.022	Specimen from left side of plate in good area, away from ingate. Did not fail, but bent to maximum capacity of equipment.
1000	0.034	0.012	
1500	0.046	0.012	
2000	0.057	0.011	
2500	0.070	0.013	
3000	0.082	0.012	
3500	0.094	0.012	
4000	0.107	0.013	
4500	0.124	0.017-yield point	
5000	0.148	0.024	
5500	0.192	0.044	
5600	0.204	0.012	
5800	0.244	0.040	
6000	0.304	0.060	
7450-maximum load.			

Specimen B5-2 1.50 in. x 0.750 in.

500	0.015	0.015	Specimen from right side of plate in area of center line shrinkage. Specimen failed during test.
1000	0.030	0.015	
1500	0.042	0.012	
2000	0.057	0.015	
2500	0.070	0.013	
3000	0.084	0.014	
3500	0.100	0.016	
4000	0.117	0.017-yield point	
4500	0.147	0.030	
5000	0.230	0.083	
5200	0.320		
5400	0.465		
5600	0.600		
6230-maximum load.			

* All tests were made between knife edges on 12 in. centers and using third point loading. Pressure was applied to the width of the bars.

FLEXURE TEST DATA (CONTINUED)

Specimen B6 1.48 in. x 0.737 in.

Load in lb	Total Deflection in inches	Unit Deflection in inches	Remarks
500	0.022	0.022	Specimen from 1 in. "Y" block. 12 in. knife edges slipped 1¼ in. apart during test. Specimen did not fail.
1000	0.035	0.013	
1500	0.048	0.013	
2000	0.060	0.012	
2500	0.072	0.012	
3000	0.085	0.013	
3500	0.097	0.012	
4000	0.109	0.012	
4500	0.122	0.013	
5000	0.136	0.014	
5500	0.154	0.018-yield point	
6540-maximum load.			

Specimen B7

500	0.014	0.014	Welded with Stainweld A-5, tested as-welded. Brittle failure in fusion zone.
1000	0.026	0.012	
1500	0.038	0.012	
2000	0.050	0.012	
2500	0.062	0.012	
3000	0.070	0.008	
3300-maximum load.			

Specimen C3 1.488 in. x 0.775 in.

500	0.012	0.012	Same as B7.
1000	0.025	0.013	
1500	0.037	0.012	
2000	0.048	0.011	
2500	0.060	0.012	
2950-maximum load.			

Specimen C4 1.485 in. x 0.768 in.

1000	0.025	0.025	Same as B7.
2000	0.048	0.023	
2500	0.059	0.011	
2600-maximum load.			

RADIOGRAPHIC LOG

Radiograph No.	Specimen No.	Type of Casting	Film	Exposure	Screen	Remarks
1	A1	1" plate-Holt	No screen	(6 hr with gamma ray)		Extensive shrinkage
2	A2	3/4" "	Blue Brand	16 min.	CaW	Shrinkage & minor porosity
3	B1-1	½" "	Type K	3-3/4 min.	lead foil	Pinholes, slag, shrinkage
4	B1-2	½" "	"	"	"	" " "
5	B2	½" "	"	"	"	" " "
6	A6	½" "Y" Blk-Holt	Type K	4 min.	lead foil	Small slag inclusions
7	B3	½" "	"	3-3/4 min.	"	Good
8	B4	½" "	"	"	"	Some slag
9	A5	½" "Y" Blk-Holt	Type K	4 min.	lead foil	Good
10	A7	½" "	"	"	"	Good
11	B5	1" plate-Holt	Type K	16 min.	CaW	Center line shrinkage
12	A4	½" "Y" Blk-Holt	Type K	4 min.	lead foil	Good
13	B5	1" plate-Holt	No screen	(6 hr with gamma ray)		(Same as No. 11)
14	B6	1" "Y" Blk-Holt	"	"		Good
15	B7	1" "	"	"		Good
16		1" "	Type F	11 sec.	CaW	Inclusions-reject
17	B12,13	1" "	"	"	"	Small inclusions
18	B20,21,22	1" "	"	"	"	Good
19	C3	1" "	"	"	"	Small inclusions
20	B31-1 to 6	1" "	"	"	"	Good
21	C4	1" "Y" Blk-Holt	Type F	11 sec.	CaW	Good
22		1" "	"	"	"	Remove section
23		1" "	"	"	"	Fair
24		1" "	"	"	"	Fair
25	B28,29,30	1" "	"	"	"	Very good

RADIOGRAPHIC LOG (CONTINUED)

Radiograph No.	Specimen No.	Type of Casting	Film	Exposure	Screen	Remarks
26	D5	½" "Y" Blk-Holt	Type K	4 sec.	lead foil	Fair - some inclusions
27	D6	½" "	"	"	"	Good
28	D7	½" "	"	"	"	Remove center section
29	D8	½" "	"	"	"	Good
30	I1	½" "	"	"	"	Good
31	I2	½" "	"	"	"	Good
32	I3	½" "	"	"	"	Fair
33	I4	½" "	"	"	"	Good
34		1" "Y" Blk-Holt	Type F	10 sec.	CaW	Fair
35		1" "	"	"	"	Slag-discard
36		1" "	"	"	"	Fair
37		1" "	"	"	"	Remove section
38		1" "	"	"	"	Poor but usable
39	H1	½" "	Type K	4 min.	lead foil	Good
40	H2	½" "	"	"	"	Good
41		½" "	"	"	"	Cut for welding rod
42		½" "	"	"	"	Slag-reject
43		½" "	"	"	"	Slag-reject
44	H3	½" "	"	"	"	Poor but usable
45	H4	½" "	"	"	"	Poor but usable
46	H5	½" "	"	"	"	Poor but usable
47		½" "	"	"	"	Slag-reject
48	E5	½" "	"	"	"	Poor but usable
49		½" "	Type A	1¼ min.	CaW	Slag-reject
50		½" "	"	1½ min.	"	Slag-reject

RADIOGRAPHIC LOG (CONTINUED)

Radiograph No.	Specimen No.	Type of Casting	Film	Exposure	Screen	Remarks
51		½" "Y" Blk-Holt	Type A	1½ min.	CaW	Slag-reject
52	J1	½" "Y" Blk-Holt	"	"	"	Fair
53	J2	½" "	"	"	"	Fair
54	J3	½" "	"	"	"	Fair
55	J4	½" "	"	"	"	Poor but usable
56		½" "	"	"	"	Slag-discard
57	J5	½" "	"	"	"	Poor but usable
58		½" "	"	"	"	Slag-discard
59		½" "	"	"	"	Slag-discard
60		½" "	"	"	"	Slag-discard
61		½" "Y" Blk-Holt	Type A	1½ min.	CaW	Slag-discard
62		½" "	"	"	"	Slag-discard
63		½" "	"	"	"	Slag-discard
64		½" "	"	"	"	Slag-discard
65		½" "	"	"	"	Good
66	E6	½" "	"	"	"	Fair
67		½" "	"	"	"	Slag-discard
68	E7	½" "	"	"	"	Poor but usable
69		½" "	"	1¼ min.	"	Slag-discard
70	E8	½" "	"	"	"	Poor but usable
71	D9	½" "Y" Blk-Holt	Type M	4 min.	CaW	Poor but usable
72	D10	½" "	"	"	"	Poor but usable
73		½" "	Type A	1¼ min.	"	Good-slag traps used
74		½" "	"	"	"	Good-slag traps used
75		½" "	"	"	"	Good-slag traps used
76		½" "	"	"	"	Good-slag traps used
77	D11	½" "	"	"	"	Good-slag traps used

RADIOGRAPHIC LOG (CONTINUED)

Radiograph No.	Specimen No.	Type of Casting	Film	Exposure	Screen	Remarks
78		½" "Y" Blk-Holt	Type M	4 min.	CaW	Slag-keep for spare
79		½" "	"	"	"	Slag-reject
80	D12	½" "	Type A	1¼ min.	"	Poor but usable
81		½" "	"	"	"	Slag-discard
82		½" "	"	"	"	Poor
83		½" "	"	"	"	Poor-metalog. spec.
84		½" "	"	"	"	Fair
85		½" "	"	"	"	Poor-metalog. spec.
86		½" "	"	"	"	Excellent
87	H7	½" "	"	"	"	Good
88		½" "	"	"	"	Remove slag section
89	H8	½" "	"	"	"	Fair
90	H9	½" "	"	"	"	Good
91		½" "Y" Blk-Holt	Type A	1¼ min.	CaW	Very good
92	H10	½" "	"	"	"	Fair
93		½" "	"	"	"	Poor-spare
94		1" "	Type F	10 sec.	"	Good
95		1" "	"	"	"	Good
96		½" "	Type A	1¼ min.	"	Fair
97		½" "	"	"	"	Remove slag section
98	H11	½" "	"	"	"	Poor
99		½" "	"	"	"	Fair
100		½" "	"	"	"	Very good
101		½" "Y" Blk-Holt	Type A	1¼ min.	CaW	Fair
102		½" "	"	"	"	Fair
103		½" "	"	"	"	Fair
104		1" "	Type F	12 sec.	"	Good

RADIOGRAPHIC LOG (CONTINUED)

Radiograph No.	Specimen No.	Type of Casting	Film	Exposure	Screen	Remarks
105		½" "Y" Blk-Eagle	Type A	1¼ min.	CaW	Very good-Mar. 3 heat
106	A8	½" "	"	"	"	Small slag inclusions
107	A9	½" "	"	"	"	Small slag inclusions
108		½" "	"	"	"	Very good
109		½" "	"	"	"	Very good
110		½" "	"	"	"	Very good
111	A10	½" "Y" Blk-Eagle	Type A	1¼ min.	CaW	Small slag inclusions
112	A11	½" "	"	"	"	Pinholes and some slag
113	B17	½" "	"	"	"	Some pinholes
114		½" "	"	"	"	Very good
115	B14	½" "	"	"	"	Some slag and pinholes
116	B15	½" "	"	"	"	Small slag inclusions
117	B16	½" "	"	"	"	Good
118	F9	½" "	"	"	"	Fair-pinholes and slag
119	D13	½" "	"	"	"	Fair-pinholes and slag
120	D14	½" "	"	"	"	Fair
121		½" "	"	"	"	Very good
122	D15	½" "	"	"	"	Fair-small inclusions
123	D16	½" "	"	"	"	Fair-small inclusions
124	F5	½" "	"	"	"	Fair-small slag incls.
125	F6	½" "	"	"	"	Fair-small slag incls.
126		½" "Y" Blk-Eagle	Type A	1¼ min.	CaW	Excellent
127	F7	½" "	"	"	"	Good
128	F8	½" "	"	"	"	Remove slag at center
129		½" "	"	"	"	Fair-pinholes
130		½" "	"	"	"	Fair-pinholes

RADIOGRAPHIC LOG (CONTINUED)

Radiograph No.	Specimen No.	Type of Casting	Film	Exposure	Screen	Remarks
131		1/2" "Y" Blk-Eagle	Type A	1 1/4 min.	CaW	Excellent
132		1/2" "	"	"	"	Excellent
133		1/2" "	"	"	"	Fair-cut for welding arc
134		1/2" "	"	"	"	Excellent
135		1/2" "Y" Blk-Holt	"	"	"	High P content
136		1/2" "	"	"	"	High P content
137		1/2" "	"	"	"	High P content
138		1/2" "	"	"	"	High P content
139		1/2" "	"	"	"	High P content
140		1/2" "	"	"	"	High P content
141		1/2" "Y" Blk-Holt	Type A	1 1/4 min.	CaW	High P content
142		1/2" "	"	"	"	High P content
143		1" "	Type F	12 sec.	"	Good. High P
144		1" "	"	"	"	Good. High P
145	I5	1/2" "Y" Blk-Eagle	Type A	1 min.	"	Excellent-Mar. 4, 1954 heat
146	I6	1/2" "	"	"	"	Excellent-Mar. 4, 1954 "
147	I7	1/2" "	"	"	"	Excellent-Mar. 4, 1954 "
148	I8	1/2" "	"	"	"	Excellent-Mar. 4, 1954 "
149	D17	1/2" "	"	"	"	Excellent-Mar. 4, 1954 "
150	D18	1/2" "	"	"	"	Good-Mar. 4, 1954 heat
151	D19	1/2" "Y" Blk-Eagle	Type A	1 min.	CaW	Excellent-Mar. 4, 1954 "
152	D20	1/2" "	"	"	"	Excellent-Mar. 4, 1954 "
153	C5	1/2" "	"	"	"	Excellent-Mar. 4, 1954 "
154	C6	1/2" "	"	"	"	Fair-Mar. 4, 1954 heat
155	C7	1/2" "	"	"	"	Fair-Mar. 4, 1954 heat
156	C8	1/2" "	"	"	"	Excellent-Mar. 4, 1954 "
157	E9	1/2" "	"	"	"	Excellent-Mar. 4, 1954 "

RADIOGRAPHIC LOG (CONTINUED)

Radiograph No.	Specimen No.	Type of Casting	Film	Exposure	Screen	Remarks
158	E10	½" "Y" Blk-Eagle	Type A	1 min.	CaW	Excellent-Mar. 4, 1954 heat
159	E11	½" "	"	"	"	Excellent-Mar. 4, 1954 "
160	E12	½" "	"	"	"	Excellent-Mar. 4, 1954 "
161	K1	½" "	"	"	"	Excellent-Mar. 4, 1954 "
162	K2	½" "	"	"	"	Excellent
163	K3	½" "	"	"	"	Excellent
164	K4	½" "	"	"	"	Good-some slag
165		½" "	"	"	"	Excellent
166	L1	½" "	"	"	"	Fair-some slag
167	L2	½" "	"	"	"	Fair-some slag
168	L3	½" "	"	"	"	Good
169	L4	½" "	"	"	"	Fair-rough surface
170	L5	½" "	"	"	"	Excellent
171		½" "Y" Blk-Eagle	Type A	1 min.	CaW	Remove slag section
172	E13	½" "	"	"	"	Excellent
173		½" "	"	"	"	Poor-cut up for rod
174	E14	½" "	"	"	"	Excellent
175	E15	½" "	"	"	"	Excellent
176	E16	½" "	"	"	"	Excellent
177	D21	½" "	"	"	"	Excellent
178	D22	½" "	"	"	"	Excellent
179	D23	½" "	"	"	"	Excellent
180	D24	½" "	"	"	"	Good
181		½" "	"	"	"	Fair-cut up for rod (Eagle
182		½" "	"	"	"	Fair-rough surface Mar.4)
183	J6	½" "	"	"	"	Excellent "
184	J7	½" "	"	"	"	Good "
185	J8	½" "	"	"	"	Good "

RADIOGRAPHIC LOG (CONTINUED)

Radiograph No.	Specimen No.	Type of Casting	Film	Exposure	Screen	Remarks
186	J9	½" "Y" Blk-Eagle	Type A	1 min.	CaW	Fair (Eagle Mar. 4)
187	L6	½" "	"	"	"	Fair "
188	L7	½" "	"	"	"	Excellent "
189	L8	½" "	"	"	"	Good "
190	L9	½" "	"	"	"	Good "
191		½" "	"	"	"	Good "
192		½" "	"	"	"	Good "
193		½" "	"	"	"	Excellent "
194		½" "	"	"	"	Excellent "
195		½" "	"	"	"	Rough-remove section (Eagle
196		½" "	"	"	"	Fair Mar.4)
197	B18	½" "Y" Blk-Holt	"	"	"	Good - Holt Mar. 17
198	B19	½" "	"	"	"	Good - "
199		½" "	"	"	"	Fair - "
200		½" "	"	"	"	Fair - "
201	B26	½" "Y" Blk-Holt	Type A	1 min.	CaW	Good
202	B27	½" "	"	"	"	Good
203		½" "	"	"	"	Good
204		½" "	"	"	"	Good
205	B23,24,25	1" "	Type F	12 sec.	"	Good
	D1	½" "	Blue Brand	4 min.	lead foil	Good
	D2	½" "	"	"	"	Good
	D3	½" "	"	"	"	Good
	D4	½" "	"	"	"	Good
	B8	½" "	Type K	4½ min.	"	Good
	B9	½" "	"	"	"	Good
	B10	½" "	"	"	"	Good
	B11	½" "	"	"	"	Good

RADIOGRAPHIC LOG (CONTINUED)

Radiograph No.	Specimen No.	Type of Casting	Film	Exposure	Screen	Remarks
	E1	½" "Y" Blk-Holt	Type K	4 min.	lead foil	Good
	E2	½" "	"	"	"	Good
	E3	½" "	"	"	"	Good
	E4	½" "	"	"	"	Good
	F1	½" "	"	"	"	Good
	F2	½" "	"	"	"	Good
	F3	½" "	"	"	"	Good
	F4	½" "	"	"	"	Good
	G1	½" "	"	4½ min.	"	Good
	G2	½" "	"	"	"	Good
	G3	½" "	"	"	"	Good
	G4	½" "	"	"	"	Good
	G5	½" "	"	"	"	Good
	G6	½" "	"	"	"	Good
	G7	½" "	"	"	"	Good
	G8	½" "	"	"	"	Good
	G9	½" "	"	"	"	Good

SCHEDULE OF OPERATIONS

Specimen No.	Welding Process & Electrode	Heat Treatment
A3 to A7	None	Parent metal, as cast - Holt
A8 to A11	None	Parent metal, as cast - Eagle
A12 to A14	None	OSC Ht. No. 17, As cast
A15 to A17	None	OSC Ht. No. 20, As cast
B1-1 and B1-2	None	Parent metal, annealed - Holt
B2 and B3	None	Parent metal, annealed - Holt
B4	Stainweld A5	Annealed, preheated to 550 F, as welded
B5-1, B5-2, B6	None	Parent metal, annealed - Holt
B7	Stainweld A5	Annealed, preheated to 550 F, as welded
B8 to B13	None	Parent metal, annealed - Holt
B14 to B17	None	Parent metal, annealed - Eagle
B18 and B19	None	Parent metal, annealed - Holt
B20 to B22	None	Parent metal, annealed - Holt - Mar. 17 heat
B23 to B27	None	Parent metal, annealed - OSC - Mar. 17 heat
B31 #1 to #6	None	Parent metal, annealed - Holt
C1 to C4	Stainweld A5	Annealed, preheat to 550 F, as welded
C5 to C8	Stainweld A5	Annealed, preheat, weld, reannealed
D1 to D4	E-6015	Annealed, preheat to 600 F, as welded
D5 to D8	E-6015	As cast, preheat to 600 F, weld, anneal at 1650 F
D9 to D12	E-8015	Annealed, preheat to 800 F, weld, reanneal
D13 to D16	E-6015	Annealed, no preheat, as welded
D17 to D20	E-6015	Annealed, preheat to 800 F, as welded
D21 to D24	E-6015	Annealed, preheat to 800 F, reheat 2 hr at 1200 F
E1 to E4	Stainweld D, type 410	Annealed, preheat to 800 F, reheat 5 min. 1000 F
E5 to E12	Stainweld D, type 410	Annealed, preheat to 800 F, reannealed
E13 to E16	Stainweld D, type 410	Annealed, preheat to 800 F, reheat 2 hr at 1200 F

SCHEDULE OF OPERATIONS (CONTINUED)

Specimen No.	Welding Process & Electrode	Heat Treatment
F1 to F4	Oxy-acetylene, parent metal rod	Preheat to 1000 F, reheat 5 min. to 1000 F
F5 to F9	Oxy-acetylene, parent metal rod	Annealed, preheated to 1000 F, reannealed
G1 to G5	Heliarc, parent metal rod	Annealed, preheat to 1000 F, weld, reannealed
G6 to G9	Heliarc, parent metal rod	As cast, preheat to 1000 F, weld, anneal
H1 to H5	Ni-Rod 55	Annealed, not preheated, as welded
H7 to H11	Ni-Rod 55	As cast, preheated, weld, anneal
I1 to I4	E-6013, mild steel	As cast, preheated, weld, anneal
I5 to I8	E-6013, mild steel	Annealed, preheated, weld, as welded
I9 to I12	E-6013, mild steel	Annealed, preheated, reheat 2 hr at 1200 F
J1 to J5	Braze with Airco 22	Annealed, not preheated, as welded
J6 to J9	Braze with Oxweld 25M	Annealed, not preheated, as welded
K1 to K4	Carbon arc, parent metal rod	Annealed, preheated, reannealed
K5 to K9	Carbon arc, parent metal rod	As cast, preheated, annealed
K10 to K12	Carbon arc, parent metal rod	Annealed, preheated, reannealed
L1 to L5	Aluminum bronze electrode	Annealed, preheated, as welded
L6 to L9	Aluminum bronze electrode	Annealed, preheated, reheat 2 hr at 1200 F

RESULTS OF TENSILE TESTS

Specimen No.	Area Sq.in.	Yield Strength in lb	Ultimate Strength in lb	Yield Strength in psi 0.2% offset	Ultimate Strength in psi	Elongation in per cent	Remarks
A3	0.447		30,250		68,200	0.5	Small slag pocket
A4	0.504		39,300		78,000	0.5	No defects
A5	0.474		35,450		75,000	1.0	No defects
A6	0.574		44,400		77,400	2.5	3/32" slag pocket
A7	0.513		39,850		77,700	1.5	No defects
A8	0.625	47,500	53,300	76,000	85,400	4.0	No defects
A9	0.534	41,500	44,850	77,800	84,200	3.0	No defects
A10	0.576	43,900	44,800	76,200	77,800	2.0	No defects
A11	0.559	43,900	45,250	78,600	81,000	2.5	No defects
A12	0.708		29,600		41,800	0.5	Slag pockets
A13	0.750		36,700		48,900	1.5	Surface slag pockets
A14	0.764	43,000	43,850	56,300	57,500	1.5	No defects
A15	0.772		22,050		28,600	0.0	Medium grey fracture
A16	0.739		19,100		25,850	0.0	Medium grey fracture
A17	0.715		21,500		30,100	0.0	Medium grey fracture
B1-1	0.402		16,900		42,000	1.5	Dark fracture
B1-2	0.404		21,420		----	-	Failed in grips
B3	0.403		27,400		68,000	17.0	No defects
B4	0.492	24,200	28,000	49,200	56,800	2.0	Fract. in fusion zone
B8	0.585	29,900	35,750	51,200	61,200	6.5	No defects
B9	0.635	30,700	37,400	48,400	58,900	5.5	No defects
B10	0.341	16,900	20,600	49,600	60,500	7.0	No defects
B11	0.586	27,700	32,200	47,300	55,000	7.5	No defects
B12	0.796	35,500	49,500	44,600	62,200	11.0	No defects
B13	0.811	39,500	51,500	48,700	62,800	9.0	No defects
B14	0.600		47,300	68,400	78,800	5.0	No defects

RESULTS OF TENSILE TESTS (CONTINUED)

Specimen No.	Area Sq.in.	Yield Strength in lb	Ultimate Strength in lb	Yield Strength in psi 0.2% offset	Ultimate Strength in psi	Elongation in per cent	Remarks
B15	0.545		43,500	70,200	79,800	5.0	No defects
B16	0.585		49,400	71,000	84,400	8.0	No defects
B17	0.538		44,800	71,400	83,400	7.5	No defects
B18	0.595	24,400	36,900	41,000	62,000	13.5	No defects
B19	0.632	26,750	39,550	42,400	62,250	13.0	No defects
B20		9,400	15,500	47,000	77,500	10.0	No defects
B21		10,150	15,750	50,700	78,600	7.5	No defects
B22		9,000	14,475	45,000	72,400	12.5	No defects
B23		9,000	13,830	45,000	69,200	12.5	No defects
B24		8,550	13,620	42,750	68,200	13.0	No defects
B25		8,700	13,250	43,500	66,200	13.0	No defects
B26	0.459	18,000	28,200	39,200	61,500	11.5	No defects
B27	0.606	24,250	36,700	40,000	60,600	11.0	No defects
B31-1			13,400		67,000	22.0	No defects
B31-2		9,500	13,050	47,500	66,500	22.0	No defects
B31-3		9,600	13,150	48,000	65,750	21.0	No defects
B31-4		9,900	13,350	49,500	66,500	23.5	No defects
B31-5		10,100	13,175	50,500	66,000	20.0	No defects
B31-6		9,950	13,300	49,750	66,500	22.0	No defects
B28		8,900	13,250	44,500	66,250	21.0	No defects
B29		9,000	13,500	45,000	67,500	21.5	No defects
B30		9,200	13,900	46,000	69,500	19.0	No defects
C1	0.477		25,300		53,200	1.5	Fusion zone failure
C2	0.471		21,500		45,600	1.0	Fusion zone failure
C5	0.650		15,200		23,400	0.5	Slight underbead cracking

RESULTS OF TENSILE TESTS (CONTINUED)

Specimen No.	Area Sq.in.	Yield Strength in lb	Ultimate Strength in lb	Yield Strength in psi 0.2% offset	Ultimate Strength in psi	Elongation in per cent	Remarks
C6	0.640		13,250		20,700	1.5	Fail center of weld
C7	0.461		15,900		34,500	1.5	Fail center of weld
C8	0.572		7,800		13,650	0.5	Fail center of weld
D1	0.406	20,300	21,000	50,000	51,700		Fract. in weld, brittle
D2	0.367		10,400		28,400		Fract. fus. zone, gas pocket
D3	0.441	21,800	22,800	49,400	51,750		Fract. in weld, gas pocket
D4	0.469		15,000		32,000		Fract. weld, no defects
D5	0.690		32,800		47,500	2.5	Fract. in weld and fus. zone
D6	0.650		36,050		55,500	3.5	Fract. in weld
D7	0.571		30,150		52,800	3.0	Fract. in fus. zone, dark root
D8	0.635		21,100		33,300	1.5	Fract. in center of weld area
D9	0.708		33,700		47,600	1.5	Fract. in weld, dark area
D10	0.790		39,500		50,000	1.5	Fract. in weld, no defects
D11	0.774		38,950		50,400	2.0	Fract. in weld, dark areas
D12	0.698		32,050		46,000	2.0	Fract. in weld, dark areas
D13	0.640		26,600		41,500	1.5	Fract. heat aff. zone, surf. crks.
D14	0.579		34,200		59,000	1.5	Fract. heat aff. zone
D15	0.595		30,800		51,800	2.5	Fract. heat aff. zone, surf. crks.
D16	0.663		34,400		51,900	1.5	Fract. heat aff. zone, surf. crks.
D17	0.606		28,100		46,400	0.5	Fract. fus. zone, no defects
D18	0.657		26,000		39,600	1.0	Fract. fus. zone, surf. hot crks.
D19	0.635		17,900		28,200	1.5	Fract. center of weld, sm. gas pockets
D20	0.680		23,450		34,500	0.5	Fract. fus. zone, some porosity
D21	0.739		28,350		38,400	1.5	Brittle fus. zone fract.
D22	0.734		13,950		19,000	1.0	Brittle fus. zone fract. Some porosity

RESULTS OF TENSILE TESTS (CONTINUED)

Specimen No.	Area Sq.in.	Yield Strength in lb	Ultimate Strength in lb	Yield Strength in psi 0.2% offset	Ultimate Strength in psi	Elongation in per cent	Remarks
D23	0.723		24,500		33,900	1.5	Brittle fus. zone fract.
D24	0.712		26,050		36,700	1.5	Brittle fus. zone fract.
E1	0.464		22,750		49,000		Fract. in fus. zone
E2	0.517		16,600		32,200		Fract. in fus. zone
E3	0.501		22,300		44,500		Fract. in fus. zone
E4					35,700		Fract. in fus. zone
E5	0.734		35,200		48,000	1.5	Fus. zone fract., surf. hot crks.
E6	0.797		31,150		39,500	1.5	Fus. zone fract., surf. hot crks.
E7	0.769		33,000		42,900	1.5	Fus. zone fract., surf. hot crks.
E8	0.741		33,400		45,000	1.0	Fus. zone fract., surf. hot crks.
E9	0.566		19,200		33,900	1.0	Fus. zone fract., surf. hot crks.
E10	0.584		23,750		40,700	2.0	Fus. zone fract., surf. hot crks.
E11	0.548		24,500		44,600	1.0	Fus. zone fract., surf. hot crks.
E12	0.603		18,750		31,200	1.5	Fus. zone fract.
E13	0.825		22,600		27,400	2.0	Fract. in fus. & heat aff. zones
E14	0.820		25,000		30,500	1.5	Fract. in fus. & heat aff. zones
E15	0.735		10,200		13,850	0.5	Fract. in fus. & heat aff. zones
E16	0.800		13,650		17,100	1.5	Fract. in fus. & heat aff. zones
F1	0.389		21,000		54,000		Fract. fus. zone and weld
F2	0.332	19,400	20,100		60,500		Fract. weld, gas pockets
F3	0.478	25,100	27,700	52,500	57,900		Fract. in parent metal
F4	0.508		25,350		49,800		Fract. in weld, some blowholes
F5	0.666		40,600		64,000	3.0	Fract. in weld, small porosity
F6	0.774		53,600		69,400	4.0	Fract. in weld, slight porosity at root
F7	0.767		45,350		59,200	3.0	Fract. in weld, slight porosity at root

RESULTS OF TENSILE TESTS (CONTINUED)

Specimen No.	Area Sq.in.	Yield Strength in lb	Ultimate Strength in lb	Yield Strength in psi 0.2% offset	Ultimate Strength in psi	Elongation in per cent	Remarks
F8	0.722		49,500		68,600	3.0	Fract. in weld, bad porosity
F9	0.682		44,200		64,800	3.0	Fract. in weld, some porosity
G1	0.515		25,200		49,000	4.0	Failed in weld, no defects
G2	0.554		25,400		46,000	4.0	Failed in weld, dark area in cen.
G3	Ruined while being machined.						
G4	0.514		22,800		44,400	3.5	Failed in weld, no defects
G5	0.527		26,200		49,700	4.5	Failed in weld, small porosity
G6	0.594		17,900		30,200	3.0	Failed in weld, no defects
G7	0.605	28,300	29,300	46,800	48,500	4.0	Failed in weld, no defects
G8	0.580		17,000		29,300	2.5	Failed in weld, no defects
G9	0.604		21,400		35,400	3.0	Failed in weld, some porosity
H1	0.782	38,800	40,700	49,600	52,000	3.0	Failed in heat-affected zone
H2	0.702		35,250		50,300	3.0	Failed in heat-affected zone
H3	0.679		33,750		49,800	3.0	Failed in heat-affected zone
H4	0.745		37,200		50,000	3.0	Failed in heat-affected zone
H6	0.670		34,400		51,400	3.0	Failed in heat-affected zone
H7	0.702	35,000	42,100	49,900	60,100	7.0	Failed in center of weld
H8	0.613	30,900	32,900	50,400	53,700	6.0	Failed in fusion zone
H9	0.712	36,300	41,050	51,000	57,800	6.5	Failed in center of weld
H10	0.689	34,750	43,050	50,400	62,600	8.0	Failed in parent metal
H11	0.640	33,050	33,100	51,700	51,800	5.5	Failed in weld, 1/8" dia. slag pocket
I1	0.920		24,750		26,900	3.0	Failed in fus. zone, slag and gas pockets
I2	0.852		17,800		20,900	2.5	Failed in fus. zone, slag and gas pockets

RESULTS OF TENSILE TESTS (CONTINUED)

Specimen No.	Area Sq.in.	Yield Strength in lb	Ultimate Strength in lb	Yield Strength in psi 0.2% offset	Ultimate Strength in psi	Elongation in per cent	Remarks
I3	0.860		20,600		24,000	2.0	Failed in fus. zone, slag and gas pockets
I4	0.920		29,500		32,100	2.5	Failed in fus. zone, slag and gas pockets
I5	Ruined while being machined.						
I6	Ruined while being machined.						
I7	0.956		16,750		17,500		Failed in fus. zone, slag and gas pockets
I8	1.010		13,000		12,900		Failed in fus. zone, slag and gas pockets
I9	0.716		27,400		38,200		Failed in fus. zone, slag and gas pockets
I10	0.884		25,600		29,000		Failed in fus. zone, slag and gas pockets
I11	0.856		19,700		23,000		Failed in fus. zone, slag and gas pockets
I12	0.777		32,500		41,800		Failed in fus. zone, slag and gas pockets
J1	0.714		27,700		38,800		Fract. along bond area
J2	0.756		24,600		32,550	4.0	Fract. along bond area
J3	0.767		20,800		27,150		Fract. along bond area
J4	0.675		13,100		19,400	4.0	Fract. along bond area
J5	0.874		33,050		37,900		Fract. along bond area
J6	0.705		12,150		17,250		Fract. along bond area
J7	0.835		14,600		17,500		Fract. along bond area
J8	0.820		27,700		33,800		Fract. along bond area
J9	0.685		15,250		22,300		Fract. along bond area

RESULTS OF TENSILE TESTS (CONTINUED)

Specimen No.	Area Sq.in.	Yield Strength in lb	Ultimate Strength in lb	Yield Strength in psi 0.2% offset	Ultimate Strength in psi	Elongation in per cent	Remarks
K1	0.691	36,000	43,500	52,000	63,000	3.5	Fract. in center of weld
K2	0.637	34,500	43,350	54,200	68,400	3.0	Fract. in center of weld
K3	0.575	29,400	41,250	51,200	71,800	3.5	Fract. in center of weld
K4	0.694	41,500	46,900	59,800	67,700	3.0	Fract. in center of weld
K5	0.552		25,250		46,200	3.0	Fract. in fus. zone, dark areas
K6	0.569	27,200	29,600	44,600	52,000	3.0	Fract. in fus. zone, dark areas
K7	0.538		30,750		57,000	4.5	Fract. in weld zone, dark areas
K8	0.630		34,500		54,750	3.5	Fract. in fus. zone, dark areas
K9	0.608	28,000	37,000	46,000	60,800	5.0	Fract. in fus. zone, dark areas
K10	0.691		23,200		33,600	2.0	Fract. in weld, dark areas
K11	0.830		12,300		14,800	1.5	Fract. in weld, dark areas
K12	0.830		29,900		36,000	2.0	Fract. in weld, dark areas
L1	0.693		17,300		25,000		Fract. in carbidic fusion zone
L2	0.712		19,400		27,250		Fract. in carbidic fusion zone
L3	0.725		18,700		25,800		Fract. in carbidic fusion zone
L4	0.615		13,700		22,300		Fract. in carbidic fusion zone
L5	0.695		15,400		22,200		Fract. in carbidic fusion zone
L6	0.823		20,400		24,800		Failed in fusion zone
L7	0.911		25,000		27,450		Failed in fusion zone
L8	0.644		13,000		20,200		Failed in fusion zone
L9	0.681		16,550		24,400		Failed in fusion zone