A Study of Heat-Affected Zone Structures in Ductile Cast Iron

Adequate weld preheat and postheat are necessary to prevent martensite formation in the heat affected zone and thereby provide improved toughness and ductility

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ABSTRACT. Poor weldability for ductile cast iron is due primarily to the formation of high carbon content martensite and massive iron carbide in the heat-affected zone and partial fusion zone, respectively. Even after postweld annealing, a fine distribution of secondary graphite particles in the heat-affected zone can prevent the weldment from attaining base metal toughness and ductility.

The formation and morphology of carbide in the partial fusion zone are affected primarily by welding heat input and are directly related to temper carbon morphologies in annealed welds. Martensite formed in the heat-affected zone is due to less than desirable preheat procedures. Postweld heat treatments lower the hardness of the martensite. However, they do not restore full ductility and toughness to the heat-affected zone due to the formation of a fine dispersion of secondary graphite that accompanies martensite decomposition resulting in a lower tensile elongation and reduced upper shelf toughness.

Preheat control to avoid martensite is essential to avoid problems associated with secondary graphitization as well. As long as preheat temperature, interpass temperature, and postweld temperatures are maintained above the martensite start temperature of ductile iron, the formation of martensite will be avoided. This temperature may have to be maintained for a considerable time after welding.

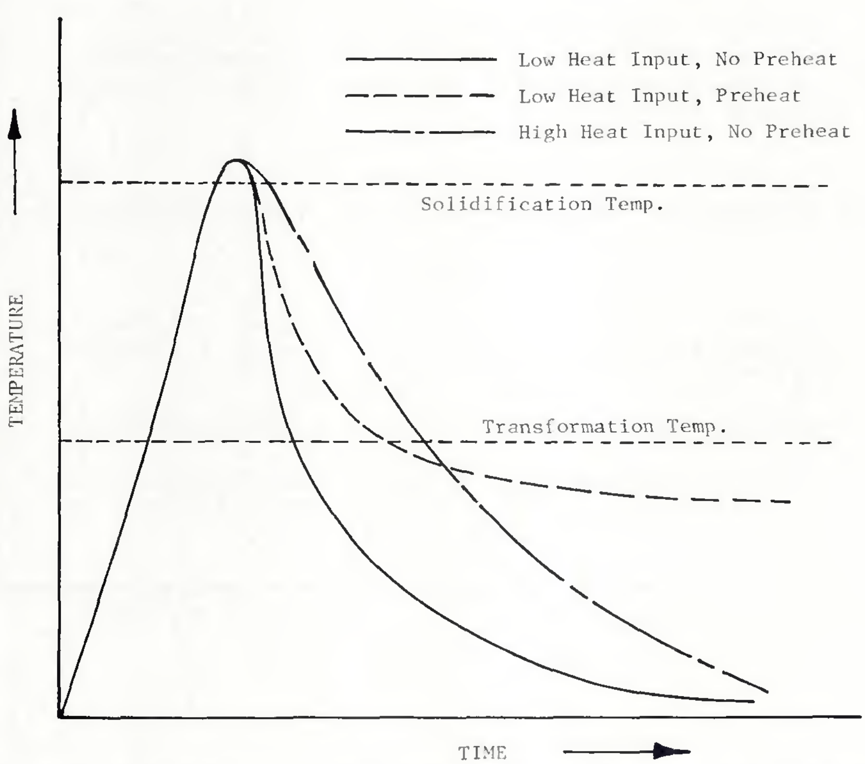

Fig. 1—Schematic representation of the effect of heat input and preheat temperature on solidification cooling rate and transformation cooling rate

Introduction

Ductile cast iron is a material which presents unique weldability problems because of its strongly heterogeneous microstructure consisting of spheroidal graphite in a matrix of alloyed ferrite and/or pearlite. Furthermore, these difficulties are more severe than those encountered with gray cast iron (flake graphite) because of the desire to obtain full base metal toughness and ductility in the heat-affected zone (HAZ). The severe carbon concentration profiles in ductile iron, combined with the rapid heating and cooling cycles associated with most welding processes, result in a myriad of microstructures in the HAZ.

What is generally accepted as the poor weldability of ductile cast iron can be attributed to two factors: the formation of martensite in the HAZ, and the development of hard, brittle iron carbide in the zone of partial fusion (Ref. 1-5). It is noteworthy that the weld metal is not considered a primary factor.

Many investigators have studied the fusion zone of ductile iron weldments and have successfully achieved acceptable weld metal properties through filler metal composition control (Ref. 6-9) or by using special welding techniques (Ref. 10). Composition control entails utilizing filter metal analyses that will not form the

Paper presented at the 63rd AWS Annual Meeting held in Kansas City, Missouri, during April 26-30, 1982.

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metastable carbide eutectic during solidification, nor form martensite during solid state transformation, when diluted with the base ductile iron. Most commonly this is accomplished through the use of nickel-bearing filler metals. Special welding techniques that restrict base metal dilution but enable adequate fusion have also been proposed to minimize weld metal problems.

This investigation is concerned solely with the problems associated with the HAZ microstructures and properties. The effects of welding heat input and preheat temperatures are shown schematically in Fig. 1. Of particular interest are the cooling rates experienced:

1. Over the solidification temperature range where high cooling rates promote carbide formation in the partial fusion zone.
2. Over the solid state transformation temperature range where martensite can form.

The solidification cooling rate, and thus the amount and morphology of the carbides, is controlled primarily by welding heat input (Ref. 1, 2, 5, 11, 12); unless excessive preheat above about 800°F (430°C) is used (Ref. 2, 11, 12). Both heat input and preheat temperature influence transformation cooling rates (Ref. 1, 2, 5, 11, 12).

Ductile cast iron is formed by the controlled solidification of a near eutectic iron-carbon-silicon alloy resulting in spheroidal graphite dispersed through a matrix of austenite which subsequently transforms to pearlite, ferrite, or ferrite plus pearlite. Solidification takes place with the graphite spheroid encased in austenite so that the last regions to solidify are the interspheroid or intercellular areas.

The melting of ductile iron is even more complex, with the lowest melting temperature regions of the microstructure at the graphite-matrix interface and in the intercellular areas. The high carbon, silicon and manganese contents of unalloyed ductile irons also give them significant hardenability. Both melting and solidification and matrix transformation effects must be considered for successful welding of ductile iron.

The difficulty in obtaining tough, ductile machinable welds in ductile iron has certainly impeded some growth in the use of this material. Typical hardness, toughness, and tensile elongation values for several classes of ductile iron and the properties achieved in weldments are shown in Table 1 (Ref. 13). Similar results have been obtained for oxyacetylene welds in ductile iron (Ref. 3). In practice, it is not a problem to achieve full base metal tensile and yield strengths. However, even upon full annealing, weldments do not achieve full base metal toughness and ductility due to the graphitization of carbides to form temper carbon in the partial fusion zone (Ref. 2, 5, 10, 11) and due to the possible formation of secondary graphite (Ref. 12) in all parts of the HAZ.

In this study the HAZ structures of ductile iron welds have been examined and analyzed in the as-welded, subcritically annealed, and fully annealed conditions; this has been done in light of recent data concerning the continuous cooling transformation behavior of ductile iron (Ref. 15) and the formation and properties of secondary graphite in heat treated ductile iron (Ref. 14). The importance of adequate preheat and maintaining preheat temperature after welding to avoid martensite in the HAZ is discussed, along with the formation and effect of secondary graphite in the HAZ upon annealing or multi-pass welding.

Procedure

A series of single pass, manual shielded metal arc welds were prepared on 0.375 in. (10 mm) thick sections of commercial ductile cast iron (3.8% C, 2.5% Si) having a typical pearlite-ferrite "bullseye" microstructure. These welds were produced using a 0.125 in. (3 mm) diameter ENi-CI electrode and a preheat temperature of 600°F (316°C).

To examine the microstructures influenced by this welding procedure and subsequent transformation, one set of weldments was quenched in water immediately after solidification of the weld, while the other was returned to the 600°F (316°C) preheat furnace after welding, held for 1.5 hours (h), and then quenched to room temperature. Welds were thus produced having essentially identical solidification cooling rates but different solid state transformation cooling rates. Using these procedures, the importance of maintaining preheat for a sufficient time to achieve complete solid state transformation was evaluated.

Each weld HAZ was examined metallographically in the as-welded, subcritically annealed, and fully annealed conditions. Subcritical annealing was performed at 1250°F (677°C) for 2 h. Full annealing was carried out by holding at 1650°F (899°C) for 2 h, furnace cooling to 1250°F (677°C), holding at 1250°F (677°C) for 4 h, followed by air cooling.

In addition, a wide variety of heat treatments were performed on quenched and tempered ductile iron to determine the multitude of factors affecting secondary graphite formation in multipass welds or in annealed ductile iron weldments. Base iron composition, alloy contents, quenching cooling rate, and heat treatment temperature were examined to determine their effect on secondary graphite formation and morphology (Ref. 14). The Charpy impact toughness of ferritic ductile iron, with and without the presence of secondary graphite, was measured and correlated to the microstructure and the fracture surfaces observed using the scanning electron microscope (SEM).

Results and Discussion

As-Welded HAZ Structure

The width of the HAZ in ductile cast iron is affected by the original matrix structure of the casting, as well as by the welding heat input and the preheat temperature. The higher carbon content of a pearlitic matrix, or of the pearlitic fraction of the matrix, results in more rapid transformation of that portion of the matrix to austenite. On the other hand, the low carbon content of ferrite can be expected to significantly retard the formation of austenite. Under the same welding conditions, then, the HAZ of a pearlitic ductile iron will be wider than that of a ferritic ductile iron.

It is also evident that the zone of partial fusion is a major factor in the HAZ. This is due not only to the broad solidification temperature range of ductile iron, but to the significant elemental segregation resulting during solidification. This strongly heterogenous cast structure renders the metallurgical problems associated

Table 1—Selected Room-Temperature Mechanical Properties of Welded Butt Joints Made in Pearlitic and Ferritic Ductile Irons[a]

		Elongation, %	Unnotched Charpy impact, ft-lb (J)	Maximum HAZ hardness, HV
Ferritic	Base metal	12-20	100-106 (135-144)	—
	As-welded	1-9	14-20 (19-27)	650-700
	Subcriticallly annealed	5-14	28-36 (38-49)	400-430
	Fully annealed	6-12	30-36 (41-49)	160-180
Pearlitic	Base metal	3-5	10-12 (14-16)	—
	As-welded	1-3	12-15 (16-20)	650-700

[a]Reference 13

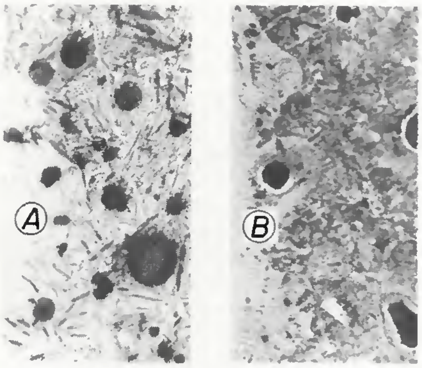

Fig. 2 — Partial fusion zone structure; as-welded condition. A — quenched after welding; B — slow cooled after welding

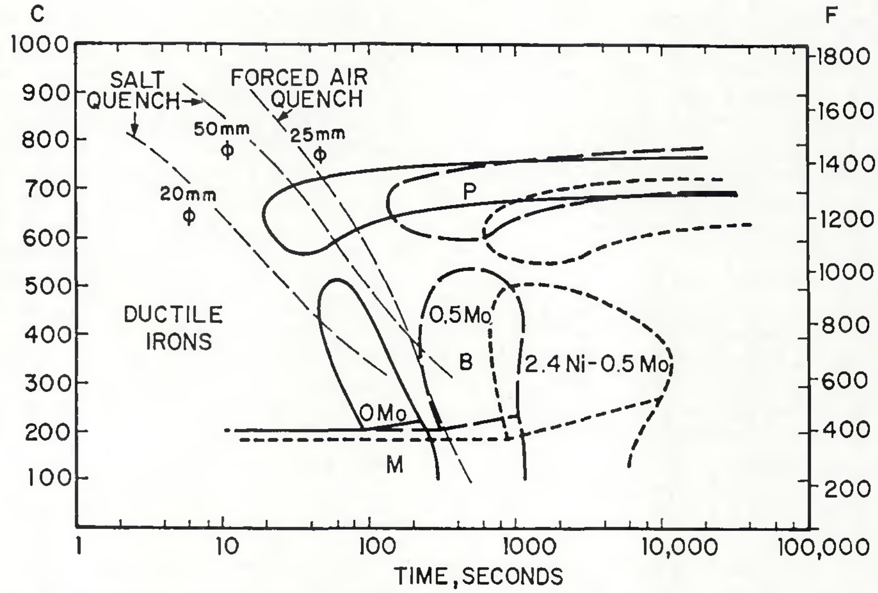

Fig. 3 — Composite continuous cooling transformation diagram for 3.35% C, 2.5% Si, 0.3% Mn ductile irons (Ref. 15)

with the formation and morphology of the partial fusion zone the most difficult to solve to achieve acceptable weldability. Generally, carbides are observed to form in the zone of partial fusion in two regions:

1. The periphery of the graphite spheroids where carbon diffuses into the matrix lowering the solidus and liquidus temperatures of the austenite.
2. The intercellular regions of the structure between the graphite spheroids where the bulk of alloy segregation occurs, and where solidification occurs last.

The subsequent rapid cooling of these liquid pools results in the formation of the metastable carbide eutectic structure. However, extreme hardness in the HAZ (due to carbides and/or martensite), resulting in poor machinability, is not the only mechanical property difficulty. Joint toughness and ductility are severely lowered when the specific welding conditions result in the formation of a continuous carbide network in the partial fusion zone. A discontinuous carbide structure can be achieved at either relatively high or relatively low heat inputs (Ref. 4, 6, 7). At very high heat inputs a large HAZ forms and the zone of partial fusion is generally broad, but the irregularly shaped fusion line causes the carbide structure in the zone of partial fusion to be discontinuous.

Kotecki, et al. (Ref. 4) suggest that heat input is not the only factor controlling carbide size and continuity. Excessive preheat (above 800°F, i.e., 427°C), torch travel speed, and base metal composition also affect carbide morphology. The specific factors governing the continuity, or lack of continuity, of partial fusion zone carbide structures are a complex interaction between welding parameters and base metal solidification characteristics.

The mechanism of carbide formation in this region appears to be straight forward. However, a number of observers have asked how carbides can develop in the partial fusion zone at cooling rates which appear to be sufficiently slow to induce graphitization. In answer, it is recognized that preheat temperatures have only a modest affect on the cooling rate at solidification temperatures and, even though higher heat inputs widen the HAZ, the cooling rate is affected only slightly. Further, one should consider the fact that a liquid pool, such as that formed in the zone of partial fusion, will solidify locally by three dimensional transfer of heat to the surrounding solid thereby enabling more rapid cooling rates to be experienced than from the overall two dimensional heat flow.

Martensite formation in the HAZ is associated not only with inadequate preheat temperatures (Ref. 1, 2, 5, 11, 12) but by neglecting to maintain the preheat temperature for a sufficient time after welding to ensure transformation to non-martensitic structures. Underbead cracking in ductile iron weldments, for example, can be prevented by using adequate preheating (Ref. 10, 11).

Various minimum preheat temperatures have been suggested to avoid martensite formation. However, it is not commonly realized that the necessary preheat temperature depends upon section thickness as well as welding heat input. In the case of thick sections, the welding preheat to avoid martensite formation must be above the martensite start (Ms) temperature, i.e., above 380–400°F (193–204°C). Lower preheat temperatures are acceptable for high heat input welds, or when welding thin sections, provided the actual HAZ temperature does not fall below the Ms during welding or before transformation to bainite or pearlite is completed. It has been proposed that high preheat temperatures, e.g., over 500°C (932°F), must be avoided because they promote continuous carbide networks in the zone of partial fusion (Ref. 12).

Maintaining the preheat temperature is particularly important when welding alloyed ductile iron where the bainite transformation times at the recommended preheat temperatures are particularly long. This is illustrated in Fig. 2

Table 2—Continuous Cooling Transformation Times for Various Ductile Cast Irons Transformed at 570°F (300°C), Minutes[a]

Composition	Transformation Start	50% Transformation	90% Transformation	99% Transformation
Unalloyed	1.0	1.5	2.2	2.5
0.25% Mo	2	5	8	10
0.5% Mo	4	7	11	18
0.5% Mo 1.2% Ni	5	12	30	120

[a]Reference 15.

where the microstructure of the partial fusion zone for two ductile iron welds made under identical welding conditions (same heat input, preheat, etc.) are presented.

The structure shown in Fig. 2A was cooled rapidly after solidification, while that in Fig. 2B was cooled slowly. While the solidification cooling rate, and hence the carbide morphology in the fusion zone, was the same for both welds, the more rapid transformation cooling rate caused martensite to form in the structure of Fig. 2A even though adequate preheat was maintained during welding.

Continuous cooling transformation curves for alloyed and nonalloyed ductile irons illustrate this behavior — Fig. 3 (Ref. 15). Due to the considerable hardenability of ductile iron, especially when alloyed, the bainitic transformation is delayed, thereby indicating that the preheat temperature must be maintained for an appreciable time after welding to avoid martensite formation.*

Transformation times from the CCT data of Fig. 3 are presented in Table 2. Note that for thick sections of alloyed ductile iron, where the transformation cooling rate is rapid, the preheat temperature must be maintained for as long as 2 h after welding to avoid the potential for cracking and excessive HAZ hardness associated with martensite formation. Avoiding martensite formation is also essential to eliminate secondary graphite in the heat affected zone of annealed weldments, or multipass welds.

Subcritically Annealed HAZ Structures

Subcritical annealing can be successfully used to lower the maximum hardness in the HAZ of a ductile iron weld from an as-welded value of 650-700 HV (58-60 HRC, equivalent) to a heat-treated value of 400-500 HV (41-49 HRC, equivalent). Microstructurally, subcritical annealing ferritizes the martensite, pearlite or bainite in the HAZ and partially graphitizes the carbides in the partial fusion zone to form temper carbon, particularly along the fusion foundry. There is still a considerable amount of primary carbide present in the zone of partial fusion that contributes to high heat-affected zone hardness and low weldment toughness and elongation — Table 1.

If continuous carbide networks are present in the as-welded partial fusion zone structure, decomposition of a portion of the carbide upon subcritical annealing may result in the formation of chains of temper carbon along the fusion

The cooling cycle experienced by the HAZ of a weld does not correspond exactly to the cooling conditions represented on either a CCT or an IT curve. However, CCT data can be used as a rough estimate of the actual transformation time in the HAZ of a weld.

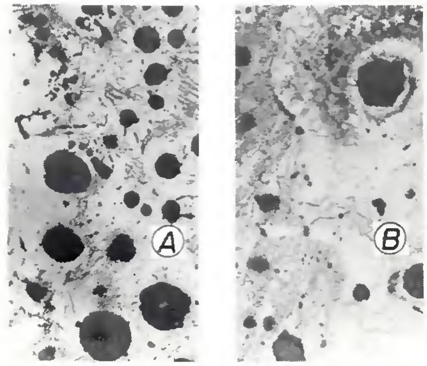

Fig. 4 — Partial fusion zone structures; subcritically annealed. A — quenched after welding; B — slow cooled after welding

boundary. This is particularly detrimental to HAZ properties. Control of the carbide morphology during welding is, therefore, also paramount even if the weldments are to be subcritically annealed, or for that matter, fully annealed.

Similarly, martensite in the as-welded HAZ also contributes to inferior properties in the weldment after subcritical annealing; this is because the martensite decomposes to ferrite plus a fine distribution of secondary graphite (Ref. 12). Decomposition of bainite or pearlite in the HAZ results in secondary graphite-free ferrite with the carbon from the bainite or pearlite matrix growing on the existing primary graphite in the ductile iron. It has been shown that this fine distribution of secondary graphite significantly lowers the tensile elongation (Ref. 16) and the toughness of the ferrite contributing to poor weldment properties.

The formation of secondary graphite from martensite in the HAZ is shown in Fig. 4. Here the subcritically annealed structure in the zone of partial fusion is shown for two welds made under the same welding and solidification conditions but with different transformation cooling rates after welding. The as-welded partial fusion zone structures for these welds was shown previously in Fig. 2.

Chains of temper carbon along the fusion boundary, from the decomposition of a portion of the carbides, can be seen in both views of Fig. 4; however, secondary graphite is only observed upon decomposition of martensite — Fig. 4A. The detrimental effects of secondary graphite on weldment toughness and ductility are particularly important in weldments that are fully annealed where carbide-free HAZ's are achieved.

Fully Annealed HAZ Structures

Full annealing is normally used to soften the ductile iron HAZ; this is done to ensure machinability and to restore maximum toughness and ductility to the weldment. While base metal hardness and tensile strength may be achieved after annealing, base metal toughness and ductility are not — Table 1. The carbides in the as-welded zone of partial fusion form temper carbon and the martensite, bainite or pearlite in the HAZ is ferritized. Full annealing usually results in complete graphitization of the partial fusion zone, while a significant amount of carbide is still present after subcritical annealing.

As before, the initial as-welded partial fusion zone carbide morphology and the original HAZ matrix structure determine the morphology of the temper carbon formed from carbide decomposition and the formation of secondary graphite, respectively. For fully annealed weldments in particular, the inability of the weldment to achieve full base metal toughness and ductility can be attributed to:

1. The formation of chain-like temper carbon at the fusion boundary caused by a less than desirable original carbide morphology in the partial fusion zone.
2. The formation of a fine distribution of secondary graphite in the HAZ from the decomposition of martensite caused by inadequate preheat temperatures, or the failure to maintain preheat temperature for a sufficient time after welding.

The formation of temper carbon cannot be avoided. However, its detrimental effect on weldment toughness and ductility can be minimized by controlling the original carbide morphology through control of the solidification cooling rate. Secondary graphite can be eliminated and fully annealed properties improved by avoiding martensite formation in the HAZ.

Fully annealed partial fusion zone microstructures for the welding procedures described previously are shown in Fig. 5. The decomposition of martensite, Fig. 5A, results in a fine distribution of undesirable graphite in the heat-affected zone. Pearlite decomposition, Fig. 5B, (or bainite) results in the growth of secondary graphite around the primary graphite and secondary graphite-free ferrite. This behavior is completely analogous to the

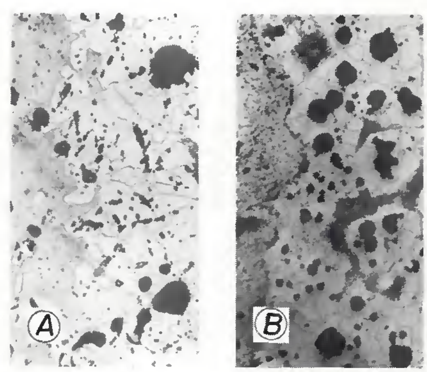

Fig. 5 — Partial fusion zone structures; fully annealed. A — quenched after welding; B — slow cooled after welding

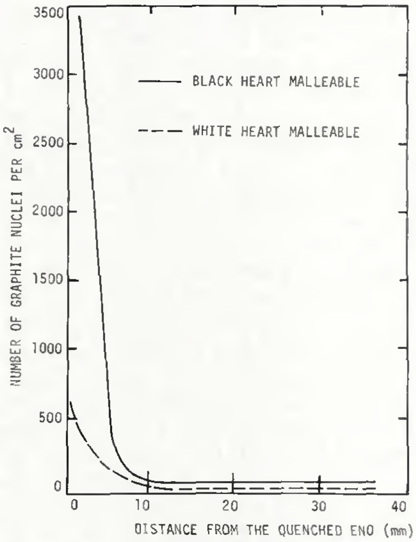

Fig. 6 — Effect of quenching a white cast iron Jominy bar on the number of graphite nuclei formed upon malleabilization (Ref. 17)

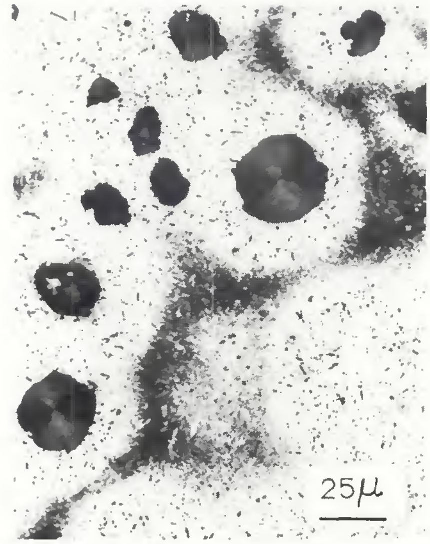

Fig. 7 — Characteristic ferritic ductile cast iron structure with secondary graphite

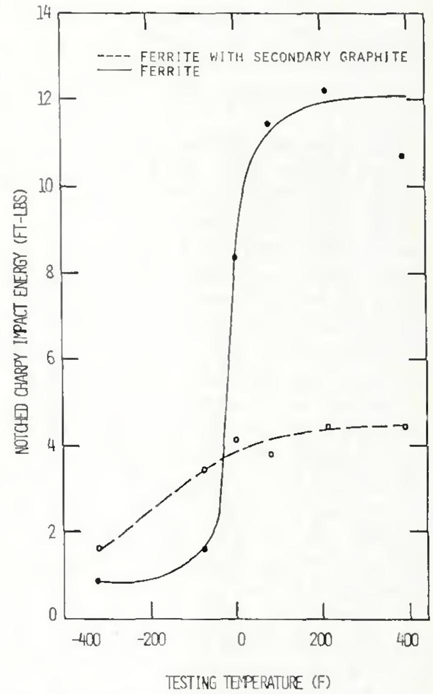

Fig. 9 — Charpy impact toughness of ductile iron with and without secondary graphite

decomposition of iron carbide from martensitic or pearlitic white cast iron to form malleable iron upon long time annealing.

As shown in Fig. 6 from the work of Schurmann (Ref. 17), malleable iron (ferrite plus temper carbon graphite structure) formed from white iron that was originally cooled rapidly (structure of carbide and martensite) resulted in a very fine distribution of graphite compared to that of slower cooled white iron (structure of carbide and pearlite). This is due to a dramatic increase in the number of temper carbon growth sites caused by the early nucleation of secondary graphite upon martensite decompensation.

Secondary Graphite

An analysis of the mechanism of formation of secondary graphite, and of the effect of secondary graphite on the mechanical properties of ductile iron, leads to an understanding of the desirability of avoiding HAZ martensite structures in ductile iron weldments. Results of the study of the heat treatment of quenched and tempered ductile irons are summarized as they relate to HAZ structures observed in ductile iron weldments.

Figure 7 illustrates secondary graphite formed from martensite upon heat treatment of quenched ductile iron at 1250°F (680°C) for 2 h. In all cases, decomposition of martensite in ductile iron (regardless of its carbon content, alloy content, or specific heat treatment parameters) resulted in the formation of secondary graphite throughout the matrix upon high temperature tempering. This fine distribution of secondary graphite occurred only upon the decomposition of martensite — not from pearlite or bainite. This is shown in Fig. 8 where a mixed ductile iron matrix structure of pearlite and martensite was subcritically annealed resulting in the formation of secondary graphite throughout the matrix only where the martensite had originally been. This behavior can be attributed to the easy nucleation of secondary graphite in the early stages of tempering and subsequent growth of secondary graphite upon the nucleated secondary graphite and the primary graphite spheroids during subsequent tempering (Ref. 14).

Eckel (Ref. 16) has reported the effects of secondary graphite on the tensile elongation of heat treated ferritic ductile iron. Decreases in tensile elongation of ferritic structures from 22% to as low as 10% were observed when secondary graphite was present. Elongated secondary graphite morphologies, often seen in the HAZ of ductile iron, appeared to be the most detrimental to tensile elongation.

Similarly, the results of Charpy V-notch impact tests on ferritic ductile iron with and without secondary graphite are shown in Fig. 9. Ferritic structures were obtained by slow cooling from an austenitizing temperature of 1700°F (927°C) and holding at 1250°F (677°C) for 6 h. Ferritic structures with secondary graphite were quenched from an austenitizing temperature of 1700°F (927°C) and then tempered (subcritically annealed) at 1250°F (677°C) for 6 h.

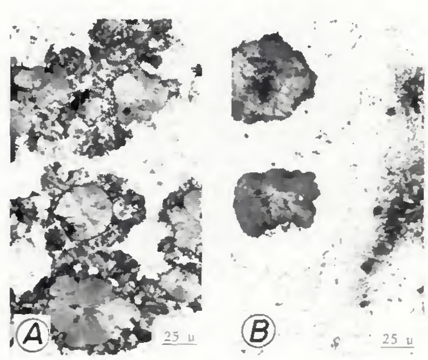

Fig. 8 — Formation of secondary graphite from mixed pearlite/martensite structures. A — pearlite/martensite structure; B — pearlite/martensite structure after annealing

The impact transition curve observed in Fig. 9 for ferritic ductile iron is typical. The formation of secondary graphite in the ferrite lowers the upper shelf* toughness considerably from more than 10 ft-lb (13 J) to approximately 4 ft-lb (5 J). However, the impact transition temperature is lowered and the lower shelf toughness is increased when secondary graphite is present.

This unique behavior is clearly explained by observing fracture surface characteristics as shown in Fig. 10 for ferritic ductile iron and in Fig. 11 for ferritic ductile iron with secondary graphite. In Fig. 10, typical ferritic ductile iron fracture surfaces are observed with considerable matrix plasticity at high impact testing temperatures (Fig. 10B) and brittle cleavage fracture of the matrix at low temperatures (Fig. 10C).

As shown in Fig. 11, secondary graphite lowers the upper shelf energy due to

**Upper shelf — the upper limits of Charpy impact transition curves where a leveling off is observed when plotting impact values vs. test temperatures.*

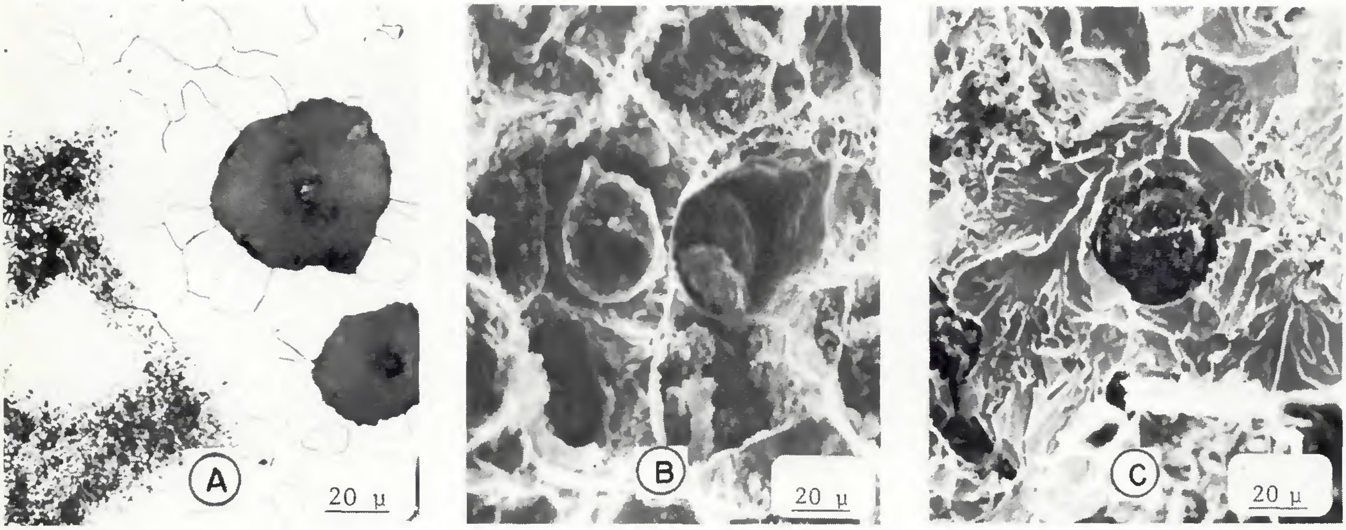

Fig. 10 — Microstructure and fracture surface characteristics of ferritic ductile iron: A — microstructure; B — fractured at 200°F (93°C)-SEM; C — fractured at −70°F (−57°C)-SEM

microtearing at the ferrite/secondary graphite interface with much less and more localized plasticity than observed for ferritic specimens. The impact transition temperature is lowered, and the lower shelf energy is increased; these phenomena occur because secondary graphite effectively causes localized tearing ahead of the cleavage crack front, giving rise to some very localized plasticity even at very low impact testing temperatures.

As noted previously, secondary graphite formed from martensite from inadequate preheat results in lower inherent ductility and upper shelf toughness. Nonetheless, low temperature weldment toughness may actually be improved somewhat by the formation of this "undesirable" secondary graphite as indicated by the Charpy impact data of Fig. 9. Secondary graphite in the microstructure serves as nucleation sites for plastic deformation during fracture. Accordingly, the lower shelf energy is raised, and a mixed mode fracture is obtained, even at very low temperatures.

Summary

The results of a study of HAZ structures in ductile cast iron weldments have been presented and discussed in terms of factors to be considered in preparing acceptable welds. The composition of ductile iron and its extremely heterogeneous microstructure results in the development of a partial fusion zone containing hard, brittle carbides. This partial fusion zone cannot be effectively prevented. However, one should apply welding procedures that prevent the development of a continuous carbide network which is detrimental from a mechanical property standpoint.

The high hardenability of ductile iron also renders the matrix likely to form martensite if appropriate preheating is not employed. It was demonstrated that preheating must be sustained for a time sufficient to avoid martensite formation. This is especially important when welding alloyed ductile cast iron where the hardenability is substantially increased compared to the already high hardenability of unalloyed ductile iron.

Martensite formation must also be

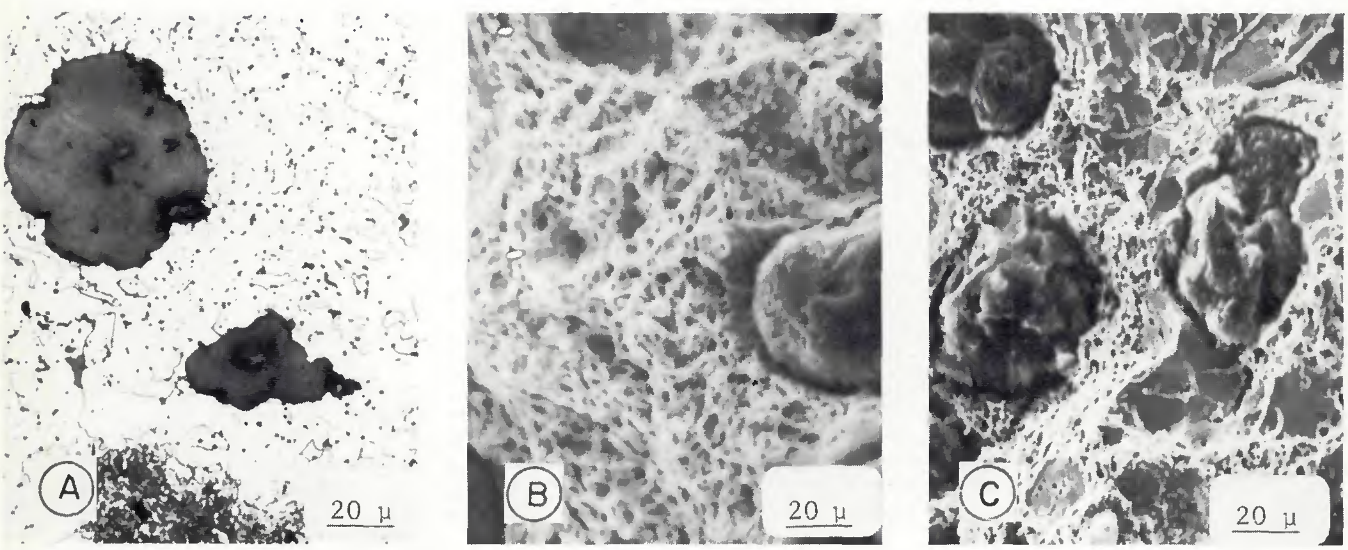

Fig. 11 — Microstructure and fracture surface characteristics of ferritic ductile iron with secondary graphite: A — microstructure; B — fractured at 200°F (93°C)-SEM; C — fractured at −70°F (−57°C)-SEM

avoided to prevent secondary graphite from developing in the matrix upon annealing or multipass welding. Maximum annealed weldment toughness will be obtained when secondary graphite formation is avoided.

Acknowledgments

The authors wish to express their gratitude to those providing support for this program: Neenah Foundry Co., Neenah, Wisconsin, for supplying the ductile iron used in this study and to R. Schorr who performed the welding experiments.

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