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Fusion boundary structures in a laser welded duplex Fe–Mn–Al–C alloy

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The microstructure and composition of the fusion boundary region in a laser-welded duplex (austenite plus ferrite) Fe–Mn–Al–C alloy were investigated, and a new weld region named as the incompletely melted zone (IMZ) was postulated. In this zone, the austenite matrix was totally melted and resolidified; however, the ferrite plate was melted partially. The mechanism for the formation of IMZ was also discussed. In addition, some martensite and/or austenite precipitations with different morphologies were located within the ferrite plates both in the IMZ and the heat-affected zone, suggesting that they were formed during quenching from high temperature.

1. Introduction

In general, a fusion weld is metallurgically divided into three major regions: the weld metal (WM); the heat-affected zone (HAZ); and the unaffected base metal (BM). The weld metal comprises a completely melted, mixed and resolidified portion. Sometimes the fusion zone (FZ) is used to represent the weld metal. The unmelted portion of base metal right up to the fusion line is called the heat-affected zone. In many alloys, there is a fourth region immediately adjacent to the fusion line subjected to partial (localized, incipient) melting. This is the partially melted zone (PMeZ) which experiences peak temperature between the liquidus and the solidus, or even down to the eutectic temperature [1–4].

In their study of heterogeneous welding of a low alloy steel, Savage *et al.* [5] defined a distinct region as the unmixed zone (UMZ) at the outer extremities of the weld metal. It was the base metal that was totally melted and resolidified without undergoing mechanical mixing with the filler metal. They classified the composite region (CR) and UMZ as portions of the weld metal, where the fusion boundary was composed of the UMZ, the weld interface, and the partially melted zone (PMeZ). Outside the PMeZ, a true HAZ existed, where thermally induced solid-state microstructural changes in the base metal occurred. Although the UMZ is often observed in heterogeneous welds [5–7], it can still take place in autogenous welds if the composition of the weld metal varies. The evaporation of volatile aluminium during laser welding of a Ti–26Al–11Nb alloy [8], and the pick-up of nitrogen during gas tungsten arc welding of austenitic stainless steels [9], are two examples.

In heterogeneous welds, another region occurs due to the composition gradients between the weld and base metals [10–12]. It was named as the partially mixed zone (PMiZ) by Karjalainen [11], who suggested that the reason for the formation of this zone was insufficient mechanical mixing and diffusion of elements in the melted weld pool. The relative location and concentration distribution of these regions are illustrated schematically in Fig. 1.

In this paper a new region is presented in a laser-welded duplex (austenite plus ferrite) Fe–Mn–Al–C alloy. It is named as the incompletely melted zone (IMZ) to distinguish it from the above zones. Some solidification microstructures are also postulated.

2. Experimental procedure

A 6.4-mm cold-rolled plate, with the chemical composition Fe–28.69Mn–8.90Al–0.56C–0.34Si (in wt %) was used for the present work. It was solution-treated at 1050° C for 2 h and subsequently oil quenched. The microstructure in Fig. 2 shows spheroidized ferrite plates elongated to the rolling direction in the austenite matrix. Annealing twins are visible within austenite grains. The volume percentage of ferrite was about 25%. Before welding, the coupon was ground and polished to 100-grit random finish, and then cleaned with acetone.

For autogenous laser welding, a PRC continuous wave CO_2 laser with an output of 2500 W was used. The beam was focused 3 mm below the specimen surface. The travel speed was $10\,mm\,s^{-1}$, and was controlled by a computerized working table. Both the central and assisted gases were He, with a flow rate of $36\,l\,min^{-1}$.

A transverse macroscopic view of the weld is shown in Fig. 3. The fully penetrated weld shows a typical wine-cup appearance, with a wide-and-swallow crown and a narrow-and-deep root. Porosities are also visible in the root part. The microstructural characterization of the laser weld was made using both optical and scanning electron microscopes, while pertinent

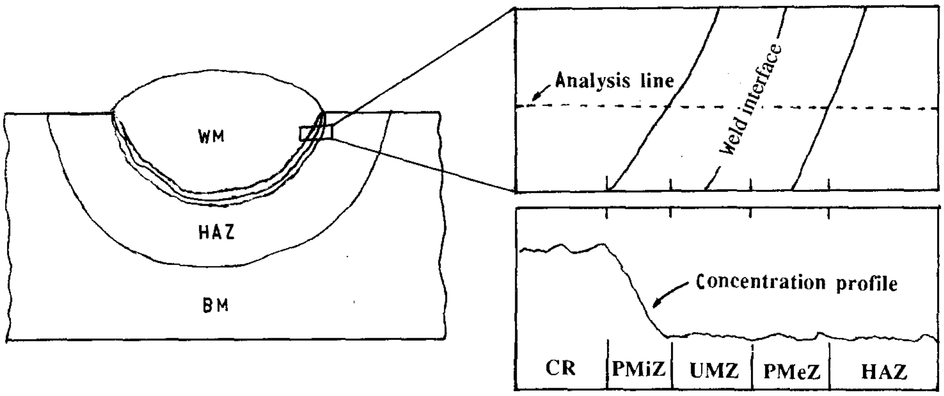

Figure 1 Schematic illustration showing various regions of a weld and the corresponding concentration profile of the alloying element.

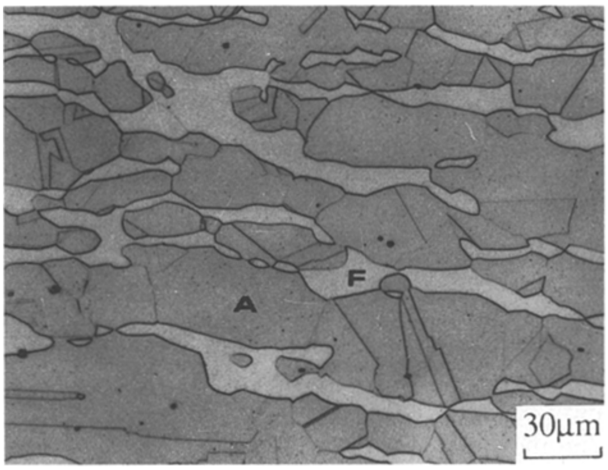

Figure 2 Optical micrograph showing the microstructure of the solution-treated base metal. A: austenite; F: ferrite.

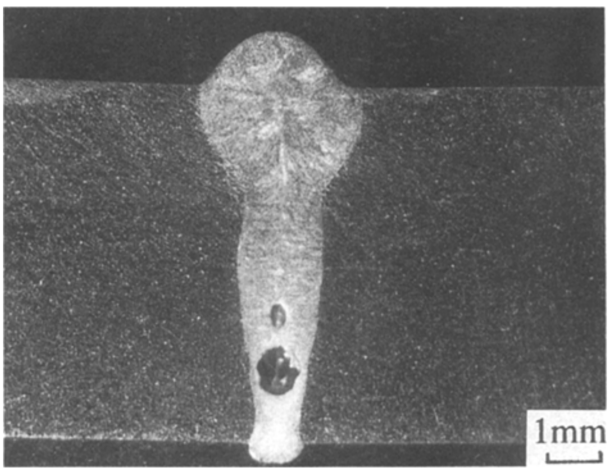

Figure 3 Optical micrograph showing transverse cross-sectional view of the laser-welded Fe–Mn–Al–C alloy.

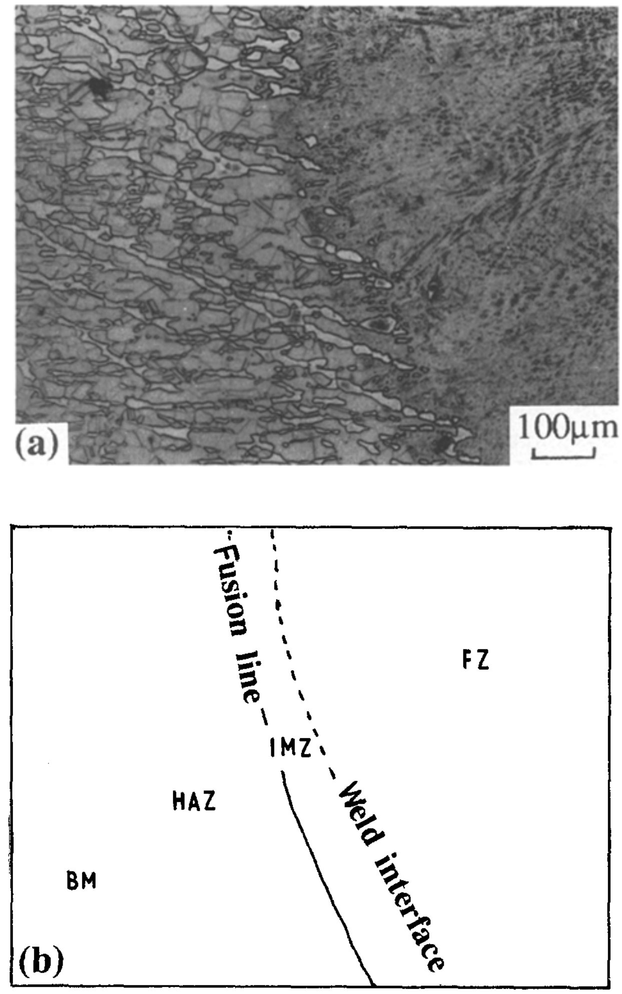

Figure 4 Optical micrograph (a) and schematic representation (b) showing discrete regions of the weld.

regions of the weld microstructure were revealed by a 10% nital etching solution. Energy dispersive spectroscopic analysis (EDS) was used to obtain composition profiles of the austenite matrix. The spot analyses at 20 μm intervals were performed through the fusion boundary with a total travel distance of 220 μm.

3. Results

As illustrated in Fig. 4, four regions of the weld are classified: the fusion zone (FZ); incompletely melted zone (IMZ); heat-affected zone (HAZ); and base metal (BM). The fusion zone was identified by the dendritic networks, which was completely melted and resolidified. In the incompletely melted zone, the austenite matrix was also melted completely and resolidified, having a similar morphology to the fusion zone, and left the fusion line as a lower border. On the other hand, ferrite plates in the IMZ were extended from BM and HAZ, although the continuity was broken down into pieces. It is believed that these ferrite plates were not completely melted. Since the weld interface was used to define whether a region experienced complete melting or not [5], it is referred to as the upper border of the IMZ.

Figs 5–8 show fusion-boundary structures in different areas of the weld: i.e. the top of the crown; middle of the crown; middle of the root; and bottom of the root, respectively. Three common features can be categorized. In the incompletely melted zone, the appearance of the very serrated ferrite grain boundary suggests that it is partially melted [13].

In addition, in the right side of the fusion line, the presence of many ferrites with mixed globular and rod-like shapes in the resolidified austenite matrix suggests that they are solidified through a eutectic reaction, just the same as those in conventional stainless steel weldings [14, 15]. However, in the left side only twins within austenite grains can be seen, suggesting that this region had been annealed.

Finally, many precipitates and associated precipitation-free zones within the ferrite plates, both in the

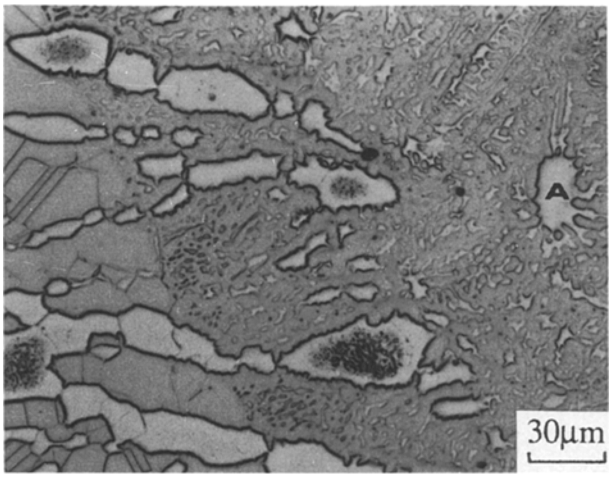

Figure 5 SEM of the fusion boundary at the top of the crown part. A: fragmented ferrite island along the weld interface.

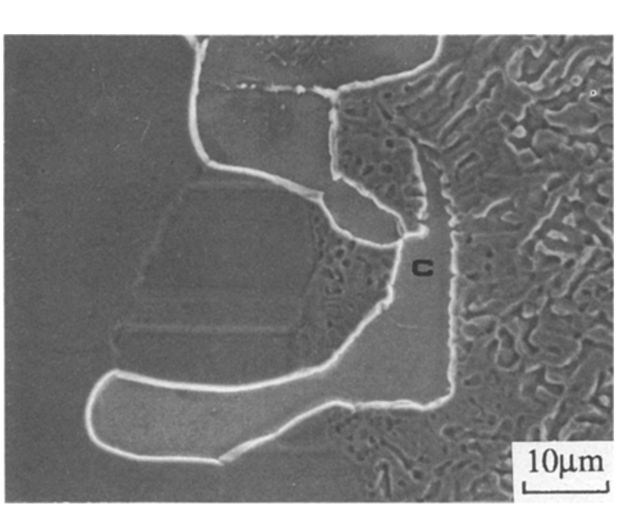

Figure 6 Optical micrograph showing the fusion boundary in the middle of the crown part. A: fragmented ferrite island in the fusion zone.

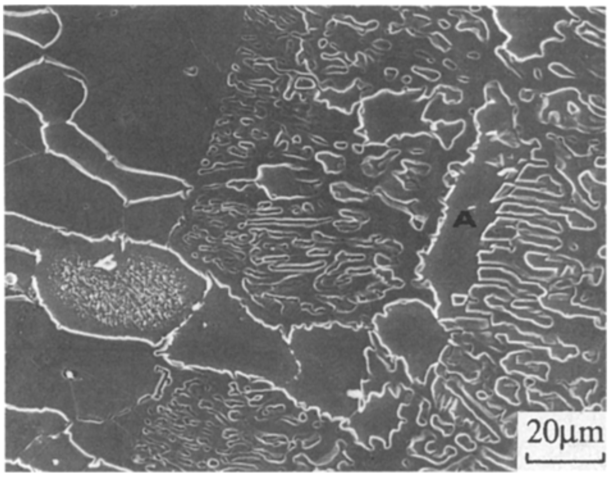

Figure 7 Optical micrograph of the fusion boundary in the middle of the root part. B, C: partially and sharply bent ferrite plates, respectively.

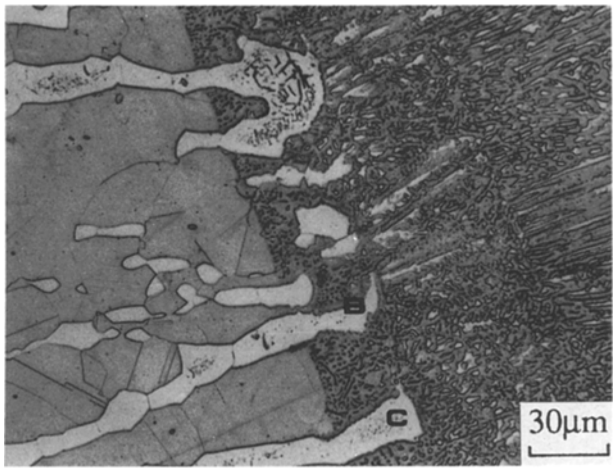

Figure 8 SEM showing the fusion boundary in the bottom of the root part. C: ferrite plate with sharply bent edge.

IMZ and HAZ, are evident. This is similar to the martensitic transformation in Fe–Mn–Al–C alloys quenched from high temperature [16–18]. A similar morphology was found in both the martensitic transformation in rapidly solidified duplex stainless steels [19] and the austenitic transformation in duplex stainless steel weldings [20, 21]. In fact, these precipitates have a variety of morphologies, such as particulate, needle-like, and plate-like, as can be seen at low magnification in Fig. 9a and at high magnification in Fig. 9b.

Figs 5–8 also show that the widths of the incompletely melted zones are 80, 100, 30–45 and 15 μm, respectively. In the crown part, ferrite islands (marked as A), having different orientations from the prior ferrite plate, were located either along the weld interface (Fig. 5) or within the fusion zone (Fig. 6). In the root part, ferrite plates with partially bent edges (mark B in Fig. 7) or sharply bent edges (marked C in Figs 7 and 8) were seen at the weld interface.

Concentration profiles of iron, manganese and aluminium in the austenite matrix along the path of the

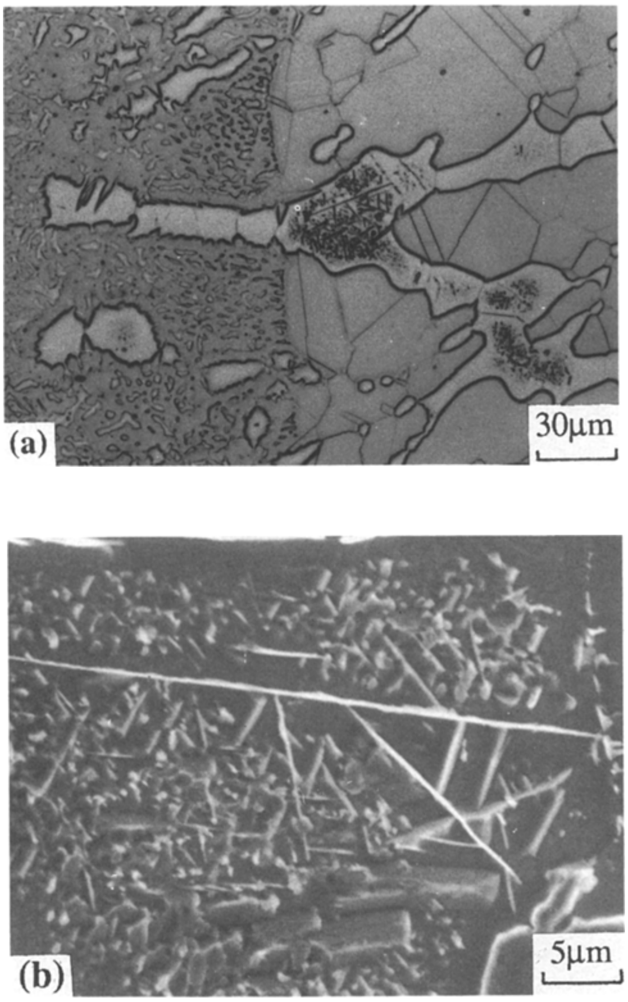

Figure 9 (a) Optical and (b) SEM micrographs of precipitates at the fusion boundary.

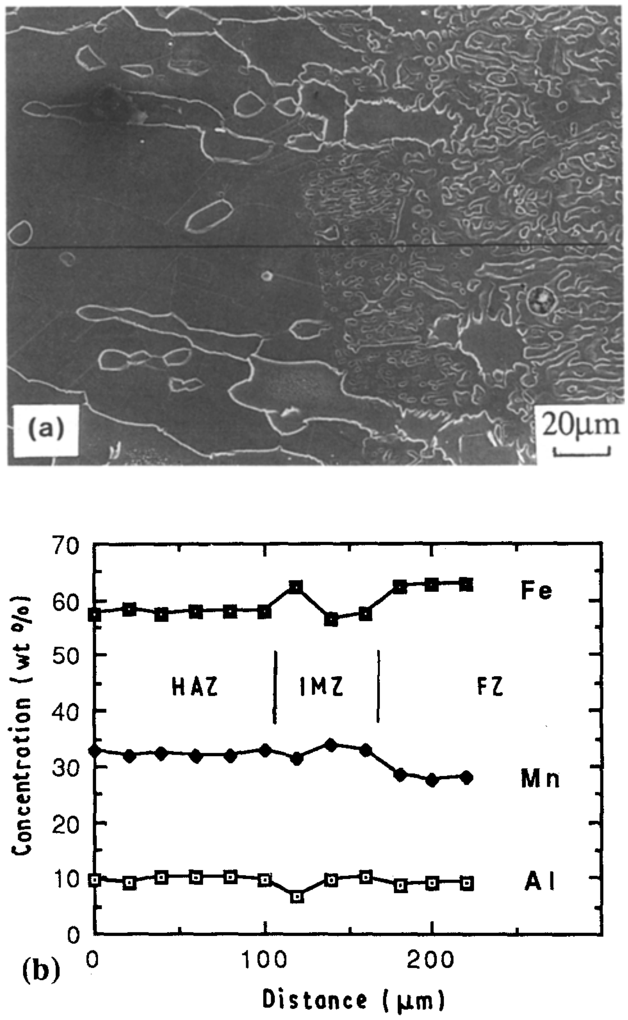

Figure 10 (a) SEM and (b) corresponding concentration profiles of iron, manganese and aluminium in the fusion zone, incompletely melted zone, and heat-affected zone.

fusion zone, incompletely melted zone, and heat-affected zone are shown in Fig. 10. The aluminium concentrations in these three zones are nearly constant. The manganese concentration in the fusion zone drops to about 28 wt %, while it remains about 32 wt % in the latter two zones. The profile of iron concentration is the reverse of that of manganese.

4. Discussion

Melting and solidification are two of the most important reactions involved in autogenous welding. They may be regarded as competing stages, especially in the welding of duplex (austenite plus ferrite) steels. Cieslak & Savage [22] have studied the arc weldability of cast stainless steels, and found special ferrite islands at the fusion boundary.They postulated that ferrite islands in the casting had a higher melting temperature than the surrounding austenite matrix, and they were not melted at the solid–liquid interface, while the adjacent austenites were melted.

The similar morphological variation was also observed in electron beam-melted stainless steel by Elmer *et al.* [15]. They believed that high-melting-point ferrite grew epitaxially at the fusion line during solidification, and produced a plane-front solidification zone. Nevertheless, this zone had been obscured by the subsequent solid state transformation of ferrite to austenite. As can be seen in Figs 5–8 in this study, it is impossible for a planar growth of ferrite to produce the oriented and bent ferrites. They are more likely to be abruptly fragmented. Therefore the morphological variation at the fusion boundary proceeded during the melting stage, not in the solidification stage. The fluid flow during melting, and the difference of melting points between ferrite and austenite, are the main reasons for the incompletely melted zone formation.

The formation sequence and the width of the IMZ can be discussed by roughly dividing into the crown and root parts, as represented in Fig. 11. The crown part experienced swallow temperature gradient (G) and the cooling rate was slower due to the reactions of heat absorption, multiple reflection, and molten metal flow [23–26]. Thus the wide-and-swallow shape was produced. During welding, the temperature above the bulk melting point caused both austenite and ferrite phases to melt completely and mix in the weld pool. At the place near the weld interface, a part of the prior ferrite plate was fragmented by the downward molten metal flow. It was then left around the weld interface, as fluid flow near the solid–liquid interface was less turbulent than in the central area. In the IMZ, the distance between melting of high melting-point ferrite

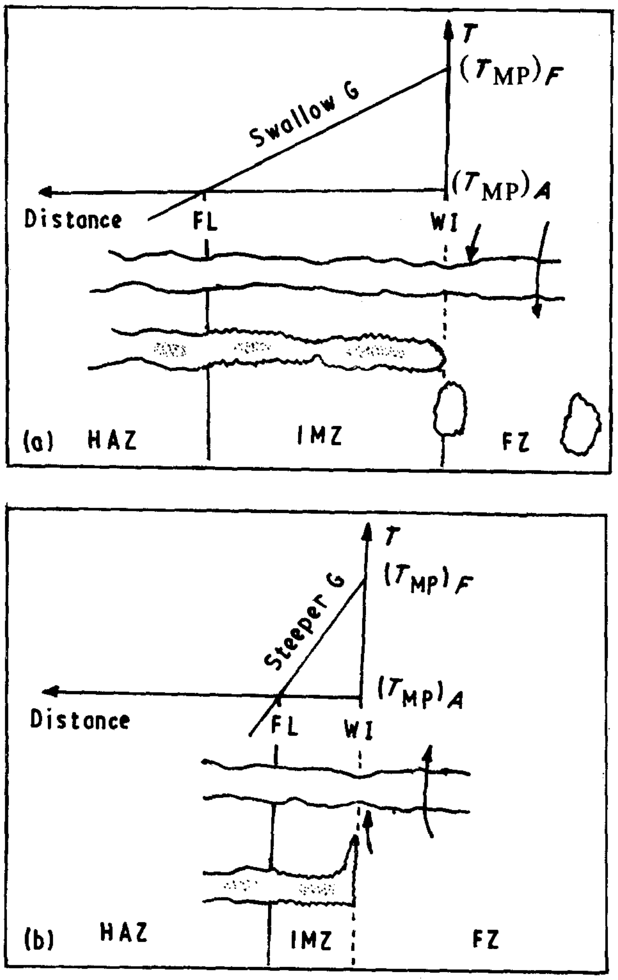

Figure 11 Proposed mechanism for incompletely melted zone formation. (a) Crown part; (b) root part (T, temperature; T_{MP}, temperature melting point; T_F, melting temperature for ferrite; T_A, melting temperature for austenite.)

(designated as weld interface, WI) and low melting-point austenite (designated as fusion line, FL) was long because of the swallow temperature gradient. Local melting might take place in the interface between the melted austenite and the unmelted ferrite by the impingement of the molten metal, and the serrated interface was formed.

In the root part of the weld, the cooling rate was faster than in the crown part. When the prior ferrite plates were fragmented by the upward molten metal flow, some ferrite plates at the weld interface were partially or sharply bent. They were retained to room temperature because of the fast cooling rate. In the meantime, the steeper temperature gradient in the solid plus liquid region shortened the width of the IMZ.

The incompletely melted zone we postulate here is different from the well-known partially melted zone in two ways. At first, the mechanism for the PMeZ formation is either wetting of grain boundary segregates, or constitutional liquation of intermetallic particles [1–4] as the melting point of these minor secondary phases are lower than the matrix phase. However, the reason for IMZ formation is the melting point difference between two main phases, where the matrix phase has a lower melting point than the second phase. Secondly, the IMZ is located above the fusion line, while the PMeZ is below that.

Although both the IMZ and UMZ have the same character in composition [5–7], metallurgical differences are also evident between them. In the latter, the base metal was completely melted during the passage of the weld pool and left a typical solidification substructure. However, in the IMZ there was incomplete melting of the base metal phases, and the final structure was a mixture of the weld metal and base metal.

5. Conclusions

1. During the laser welding of a duplex Fe–Mn–Al–C alloy, a new weld region called the incompletely melted zone (IMZ) was discovered.
2. The morphological characterizations of this zone were that the austenite matrix was melted completely, while ferrite was melted partially with a serrated interface and had precipitates within it. The concentration distributions in both the incompletely melted zone and the base metal were similar; however, manganese was lost in the fusion zone.
3. The formation mechanism of the incompletely melted zone and the variations in zone width could be explained by thermal history.

Acknowledgements

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References

1. W. F. SAVAGE and B. M. KRANTZ, *Weld. J.* **45** (1966) 13s.
2. W. A. OWCZARSKI, D. S. DUVALL and C. P. SULLIVAN, *ibid.* **45** (1966) 145s.
3. W. F. SAVAGE, E. F. NIPPES and T. W. MILLER, *ibid.* **55** (1976) 181s.
4. S. KOU, "Welding Metallurgy" (Wiley, New York, 1987) p. 239.
5. W. F. SAVAGE, E. F.NIPPES and E. S. SZEKERES, *Weld. J.* **55** (1976) 260s.
6. W. A. BAESLACK, J. C. LIPPOLD and W. F. SAVAGE, *ibid.* **58** (1979) 168s.
7. J. LUKKARI and T. MIOSIO, *Microstr. Sci.* **7** (1979) 333.
8. M. J. CIESLAK, T. J. HEADLEY and W. A. BAESLACK, *Metall. Trans.* **21A** (1990) 1273.
9. C. D. LUNDIN and C. P. D. CHOU, *Weld. Res. Coun. Bulletin* **289** (1983).
10. D. S. DUVALL and W. A. OWCZARSKI, *Weld. J.* **47** (1968) 115s.
11. L. P. KARJALAINEN, *Z. Metall.* **70** (1979) 686.
12. F. ORNATH, J. SOUDRY, B. Z. WEISS and I. MINKOFF, *Weld, J.* **60** (1981) 227s.
13. T. KILNER, G. C. WEATHERLY and R. M. PILLIAR, *Scripta Metall.* **16** (1982) 741.
14. T. WATANABE, *Trans. Nat. Res. Inst. Met.* **27** (1985) 42.
15. J. W. ELMER, S. M. ALLEN and T. W. EAGER, *Metall. Trans.* **20A** (1989) 2117.
16. S. K. CHEN, W. B. LEE, K. W. CHOUR, C. M. WAN and J. G. BYRNE, *Scripta Metall.* **23** (1989) 1919.
17. S. K. CHEN, K. H. HWANG, C. M. WAN and J. G. BYRNE, *ibid.* **24** (1990) 151.
18. K. H. HWANG, W. S. YANG, T. B. WU, C. M. WAN and J. G. BYRNE, *Metall. Trans.* **21A** (1990) 2815.
19. T. TOMIDA, Y. MAEHARA and Y. OHMORI, *Mater. Trans. Jpn Inst. Met.* **30** (1989) 326.
20. W. A. BAESLACK and J. C. LIPPOLD, *Metal Const.* **20** (1988) 26R.

21. T. OGAWA and T. KOSEKI, *Weld. J.* **68** (1988) 181s.
22. M. J. CIESLACK and W. F. SAVAGE, *ibid.* **59** (1980) 136s.
23. C. E. ALBRIGHT, "Trends in Welding Research of the United States" edited by S. A. David (ASM, Ohio, 1982) p. 653.
24. Y. ARATA, M. TOMIE, N. ABE and X. Y. YAO, *Trans. Jpn Weld. Res. Inst.* **16** (1987) 13.
25. N. POSTACIOGLU, P. KAPADIA, M. DAVIS and J. DOWDEN, *J. Phys. D: Appl. Phys.* **20** (1987) 340.
26. J. M. VITEK, A. DASGUPTA and S. A. DAVID, *Metall. Trans.* **14A** (1983) 1833.

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