Effect of Al content on oxidation behaviour of ternary Fe–Al–C alloys

V. Shankar Rao^{a,*}, R.G. Baligidad^b, V.S. Raja^a

^a*Corrosion Science and Engineering, Indian Institute of Technology, Bombay, Mumbai-400 076, India*

^b*Defence Metallurgical Research Laboratory, Hyderabad-500 058, India*

Abstract

Iron aluminides produced by the electroslag refining technique, having the compositions: (1) Fe–16Al–1C, (2) Fe–10Al–1C, and (3) Fe–8Al–1C were used to investigate the effect of Al on the oxidation behaviour of the Fe–1C–Al system at 700 to 1000 °C. Prior to oxidation studies, phase and microstructure of alloys were analysed. The carbide phase, Fe₃AlC_{0.69}, was found to be distributed in the Fe₃Al matrix in alloy 1 and α (Fe–Al) matrix in alloys 2 and 3. The low Al content alloys displayed inversion in the oxidation kinetics below 800 °C, while, high Al content alloy displayed inversion phenomena at 1000 °C. The mechanism involving inversion in oxidation kinetics was found to be different in the two cases. In the former, it was attributed to the preferential oxidation of Al, while in the latter, to the phase transformation within the Al₂O₃. Carbides in the alloy having low Al content showed instability during oxidation.

Keywords: A. Iron aluminides, based on Fe₃Al; A. Ternary alloy systems; B. Oxidation; D. Microstructure

1. Introduction

Development of iron aluminides based on Fe₃Al for high temperature applications is an important area of research, as these materials possess unique properties such as high strength to weight ratio, excellent high temperature oxidation and sulfidation resistance [1–4]. Low room temperature ductility has been a major impediment for mechanical processing of these materials, while a drop in strength above 600 °C has limited their possible applications [1–4]. The first review on iron aluminides by McKamey et al. [5] and a more recent review by Stoloff [6] bring out salient features of these materials. Most of the literature on iron aluminides is concerned with those aluminides having very low (<0.01 wt.%) C content. High-carbon iron aluminides were not of interest for research, as carbon is known to embrittle these alloys [7]. On the contrary, iron aluminides produced recently by the electroslag remelting process (ESR) were found to improve its strength when carbon is added as an alloying element [8–12]. This is attributed to precipitation hardening by formation of carbides. These alloys were further shown to exhibit a

reduced susceptibility to environmental embrittlement. This may be attributed to hydrogen trapping by the carbides of the alloy to minimise hydrogen content in the iron aluminide lattice [13]. However, this improvement in ductility was not good enough to process the material for structural applications. It is a well-established fact that on reducing the Al content, an improvement in ductility can be achieved in iron–aluminium alloys [1,11]. Therefore, in order to achieve substantial improvement in ductility these alloys have been developed by reducing the Al content for the same carbon content. However, in doing so, the alloy loses its ordered structure. Additionally, lowering of the aluminium content could affect the oxidation resistance. Hence it is pertinent to examine the effect of Al on the oxidation behaviour of carbon containing Fe–Al alloys. The details of the oxidation behaviour of alloy 1 have been discussed in our previous communication [14]. For comparing with the other alloys, the reader is referred there.

2. Materials

Three iron–aluminium alloys prepared by air induction melting and subsequently processed through the

electroslag refining process were taken for investigation. The nominal compositions of the alloys in wt.% are: (1) Fe–16Al–1C, (2) Fe–10Al–1C and (3) Fe–8Al–1C. This range of compositions will bring out the influence of Al on oxidation behaviour of iron aluminides. Details on processing technology and mechanical properties can be found elsewhere [15,16].

3. Experimental

3.1. Microstructural and phase analysis

Samples for microscopy were obtained by polishing them on various grades of SiC papers starting from 220 to 1000, followed by polishing with a diamond paste of 1 μm size particles and then etching. The etchant consisted of 33%CH₃COOH + 33%HNO₃ + 1%HF + 33%H₂O by volume. Microstructures of the alloys were examined using a Leica, Reichert MeF3A optical microscope. X-ray diffraction studies were carried out to identify the phases using a Philips PW-1820 diffractometer. Various phases were identified using the PCDFWIN powder diffraction software package.

3.2. Oxidation studies

The specimens taken for oxidation test were 12×12×4 mm in size; these were polished by grinding them on successive grades of SiC paper up to 800 grade. Just before oxidation, specimens were ultrasonically cleaned using acetone, dried and finally weighed. A resistance-heated tubular furnace controlled with an accuracy of $\pm 5^\circ\text{C}$ was used for the experiments. Experiments were conducted in duplicate to check for reproducibility of data. Long-term oxidation tests were carried out at 700–1000 $^\circ\text{C}$. The weight change of the specimen was recorded by withdrawing them from the furnace at regular intervals. A semi-microbalance having an accuracy of 1×10^{-5} g was used. Crystal structures of the oxide scale formed on the alloys during oxidation were determined by X-ray diffraction as described in the previous section. The morphology of the oxide was examined by using a scanning electron microscope (SEM). Cross-sectional analysis of the oxide scale was carried out from electron probe microanalysis (EPMA).

4. Results

4.1. Phase identification and microstructure analysis

X-ray diffraction patterns of these alloys are presented in Fig. 1. Based on JCPDS card numbers 45-1203 and 03-0965, peaks of the XRD pattern of alloy 1 (Fig. 1a) can be indexed for Fe₃Al and Fe₃AlC_{0.69}

phases, respectively, whereas in alloys 2 and 3, peaks of the XRD pattern are indexed as α (Fe–Al) and Fe₃AlC_{0.69} phases (Fig. 1b and c). In alloy 1, the appearance of an extra peak at the d -value of 2.89 Å is due to the reflection of the superlattice plane (200) of Fe₃Al. This indicates that it is an ordered structure, whereas the other alloys have a disordered structure. Since the XRD patterns shown here correspond to the sheet sample used for oxidation studies which were obtained from the as-cast alloy, they do not exactly reproduce powder patterns. Thus small variation in the peak intensities and a few additional peaks appears in alloys 2 and 3. However, the peaks indexed correspond to the corresponding plane given in the JCPDS file. Microstructures of alloys 1–3 as seen in the SEM are brought out in Fig. 2. Since the carbon content is the same in all the alloys, the difference in microstructure exhibited by the alloys can be attributed to the variation in Al content. In alloy 1, the dendrites are single-phase Fe₃Al and the entire interdendritic region is a continuous network of Fe₃AlC_{0.69} carbides, as pointed out in our previous work [14,17]. In alloy 2, dendrites are more uniformly dispersed. On further reducing the Al content from 10 to 8%, there is not much change in the microstructure of alloy 3, as shown in Fig. 2c. An isothermal section of the iron-rich corner of the Fe–Al–C phase diagram at 25 $^\circ\text{C}$ is shown in Fig. 3 [18]. The location of alloys 1–3, are marked in this figure. The carbide phase, Fe₃AlC_x, is designated as K phase while Fe₃Al and Fe–Al are designated as α . All our alloys fall in the region of α + K. Carbides are formed from the liquid directly through a peritectic transformation although the temperature of transformation is uncertain. Palm et al. made an extensive investigation related to the K phase in the Fe–Al–C system between 800 $^\circ\text{C}$ and the liquidus surface [19]. According to them, the K phase forms at about 1400 $^\circ\text{C}$. The interdendritic morphology of the carbides, as revealed by the micrograph, is in agreement with the phase diagram. It should be noted that, volume fraction of α and the K phases are the same for a given carbon content, though Al content is varied. The α becomes an ordered phase once its Al content exceeds the critical value. Since the oxidation behaviour of the phase relies heavily on its chemical composition, it seems that the carbide, because of its stoichiometric composition, might exhibit similar oxidation behaviour in all the alloys studied, while the α phase would exhibit different oxidation behaviour depending on whether it exists as a super lattice structure or a random solid solution.

4.2. Oxidation kinetics

Oxidation kinetics of alloys 1–3 were studied at four different temperatures, viz. 700, 800, 900 and 1000 $^\circ\text{C}$. Fig. 4 shows the variation in the weight gain per unit area with time for all these oxidation temperatures.

Fig. 1. X-ray diffraction patterns of (a) alloy 1, (b) alloy 2 and (c) alloy 3.

(a)

(b)

(c)

Fig. 2. SEM micrographs: (a) alloy 1, (b) alloy 2, (c) alloy 3, showing the interdendritic network of carbides in alloy 1, while in alloys 2 and 3 carbides are more uniformly distributed.

Alloy 3 exhibits a much higher weight gain than alloys 1 and 2 at all the temperatures under study. At 700 °C the weight gain by alloy 2 is higher than that of alloy 1 (Fig. 4a), while at 800 °C, oxidation behaviour, seen in terms of weight gain, are almost the same for alloys 1 and 2 (Fig. 4b). At 900 °C, the alloys exhibit an

initial steep rise followed by weight drop before they reach a plateau (Fig. 4c). In the plateau region, the weight gain by the alloys follow the order alloy 3 > alloy 2 > alloy 1. Interestingly, at 1000 °C, alloys 1 and 2 follow the same kinetics up to 300 h, while alloy 3 fails within 25 h of exposure time (Fig. 4d). On further oxidation, alloy 1 displays weight loss in steps and returns almost to its initial weight, while weight gain is continued in alloy 2.

4.3. Oxide scale characterisation

No oxide peaks could be identified in the case of alloy 1 oxidised at 700 and 800 °C, possibly due to low oxide thickness. An XRD pattern of the same alloy oxidised at 900 °C for 1000 h is shown in Fig. 5. The peaks belonging to α -Al₂O₃ and Fe₃AlC_{0.69} are identified. The appearance of strong peaks of Fe₃AlC_{0.69} indicates that the oxide scale is thin. Fig. 6 shows the XRD pattern of alloy 3 oxidised at 700 and 900 °C for 1000 h. Only peaks belonging to iron oxides of Fe₂O₃ appear. The absence of the background peaks of the alloy indicates that the oxide formed is thicker than that formed on the alloy.

Morphologies of the oxide scale viewed in the SEM are shown in Figs. 7–9. Micrographs of the oxide scale formed on alloy 1 at two different temperatures, 800 and 900 °C for 1000 h, are shown in Fig. 7. The oxide layer formed at 800 °C displays merging of oxide grains. This results in a compact and protective scale (Fig. 7a). On raising the temperature to 900 °C, the morphology of the scale is changed, as shown by development of oxide nodules over the scale surface (Fig. 7b). Rupture of such oxide nodules cause spallations, as also shown by the drop in weight gain after 300 h of oxidation (Fig. 4c). Morphology of the alloy 2 oxidised at 900 °C is brought out in Fig. 8. As seen, there are two layers of the oxides. This micrograph further shows that a few sites are predominantly oxidised, which could possibly correspond to dendritic regions. A similarity between the microstructure of the alloy (Fig. 2b) and the current figure supports this view. Energy-dispersive X-ray (EDX) analyses of the oxide scale performed at two different locations, 1 and 2, as shown in the Fig. 8, are listed in Table 1. The analysis reveals that the oxide scale at location 1 is mainly due to iron oxides, the oxide scale of locations 2 consists of mixed Fe and Al oxides.

The micrograph of the alloy 3 oxidised at 700 °C for 100 h (Fig. 9a) displays typical iron oxides. The surface is not fully covered with oxide even after 100 h of oxidation. This behaviour is expected to cause a higher oxidation rate. At higher magnification, this oxide appears as a whiskers (Fig. 9b). Visual inspection showed the oxide to be reddish brown in colour, loosely adherent until 100 h of oxidation. On further oxidation, the surface slowly turned grey in colour and the oxide seemed adherent in nature. The morphology of the same

Fig. 3. Fe-Al-C phase diagram [18].

alloy oxidised at 700 °C for 1000 h is displayed in Fig. 9c. This change in colour and morphology could possibly indicate the change in composition of the oxides with the time of exposure. A cross-sectional view of the oxide scale of alloy 3 oxidised at 700 °C is shown in Fig. 10. Two layers of the oxide scale are visible. The total thickness of the scale is 70 μm , of which inner layer occupies about 30 μm . The figure displays the outer oxide layer as porous while the inner oxide layer is seen to be compact. The EPMA line scans show that the inner layer is richer in Al than the outer layer, while iron is distributed more in the outer layer than that of inner layer. At the metal-oxide interface, the Al profile is maximum and iron is at its lowest. Notably, XRD analysis could not reveal the presence of aluminium oxide due to a thicker outer layer of iron oxide. Fig. 11 shows the cross-sectional view of alloy 2 oxidised at the same temperature for 1000 h. Unlike alloy 3, the scales formed on alloy 2 have fallen off during polishing. This could be due to more brittle and/or thin oxides formed on the latter than on the former. The thickness of the oxide layer measured under microscope with a micrometer stage is around 20 μm in this case.

5. Discussion

The results show that oxidation kinetics of Fe-Al-C alloy is less affected when the Al content is lowered from 16 to 10%. However, on further reducing the Al content to 8%, a marked deterioration in oxidation resistance occurred, as exhibited by the alloy 3. Comparison of weight gains of Fe-Al alloys with that of the Fe-Al-C alloy having the same Al content would bring out the effect of carbon on oxidation. Tomaszewicz et al. have reported a weight gain of 0.2 mg/cm^2 in the Fe-6Al alloy when oxidised at 800 °C for 25 h [20], while alloy 3 of the present study, having 8%Al, shows 2 mg/cm^2 for similar oxidation conditions, an order of magnitude higher than that of the former. This difference can be attributed to the presence of carbon in alloy 3. The present authors, using scanning Auger electron spectroscopy analysis, showed that the carbides in Fe-Al alloys undergo preferential oxidation even when the Al content of Fe-xAl-1C is as high as 16% [17]. But the effect was more significant as is shown by 8%Al alloy. Thus, it seems that the detrimental effect of carbon increases with decrease in aluminium content. While the carbides

themselves are prone to attack, the carbide/matrix interface could be another vulnerable area of attack. Diffusion of oxygen through the interface could be faster, leading to severe localised oxidation of the alloy. Interestingly, carbon seems to be behaving differently in pure Fe than in the Fe–Al alloy. Thus, Caplan et al.

have reported a decrease in oxidation rate of pure iron when carbon was added to it [21]. The indirect effect of C seems to arise for the following reason. The amount of carbide formed in the Fe–1C– x Al alloy seems to depend on the carbon content. Since the carbide stoichiometry remains the same, the same amount of Al is

Fig. 4. Oxidation kinetics of alloys 1–3 (a) at 700 °C, (b) 800 °C, (c) 900 °C, (d) 1000 °C. Insets in figures are enlarged plots of alloys 1 and 2.

Fig. 4. (continued)

utilised in forming the carbide phase. This leaves less Al available to form a solid solution with Fe. As a consequence, not only is the alloy unable to form ordered α , as it requires more Al than is available, but also makes the α prone to oxidation. Thus the critical concentration of Al needed to offer oxidation resistance for Fe-C

seems to be more than that is required for pure Fe, notwithstanding the vulnerable interfacial regions in the two-phase system.

As the alloy 3 showed significant weight gain even during the early stages of oxidation, at 700 °C, the weight gain data were utilised to determine the kinetic

Fig. 5. X-ray diffraction pattern of alloy 1 after oxidation at 900 °C.

rate constants. Two parabolic rate constants, one for the data obtained up to 200 h of oxidation, called k_{p1} , and the other for the data obtained between 200 and 1000 h, called k_{p2} , are calculated from the graph (Fig. 4a). The k_{p1} turns out to be $6.9 \times 10^{-11} \text{ g}^2/\text{cm}^4\text{s}$ and k_{p2} is $3.6 \times 10^{-12} \text{ g}^2/\text{cm}^4\text{s}$. As it is logical to suggest that the outer iron-rich oxide scale formed during early stages of oxidation, when the kinetics is rapid, k_{p1} can be attributed to this less protective scale formation. The lower value for k_{p2} than k_{p1} and its occurrence at later stages of oxidation, beyond which the rate of oxidation becomes low, could indicate that it corresponds to the formation of aluminium oxide. The time taken for the transition from one mode of oxide formation to another in the present case is 200 h; other authors have reported this transition to occur within 25 h of oxidation in the case of Fe–7.5Al–0.65C alloy at the same temperature [22]. Nevertheless, it is becoming clear that the formation of Al₂O₃ is crucial in the protection of Fe–Al alloys from oxidation and its formation depends on time and temperature of oxidation. This aspect is discussed further.

The results show that among all the temperatures of investigation, the weight gains in case of alloys 2 and 3 are the highest for oxidation at 700 °C. It goes through an inversion with the lowest weight gain at 800 °C, whereas the alloy 1 having 16% Al does not exhibit this inversion. Saegusa et al. reported the oxidation behaviour of Fe–5Al (wt. %) alloy from 500 to 1000 °C under 1 atm partial pressure of oxygen [23]. They observed the highest oxidation rate at 800 °C and the lowest oxidation rate at 1000 °C in their study. This inversion of oxidation at 1000 °C was attributed to the formation of Al₂O₃. Even binary Fe–Al alloys exhibit a similar

behaviour. For example, Boggs reported a similar trend in the case of Fe–4.94Al (wt. %), when oxidised between 450 and 900 °C at 700 Torr oxygen partial pressure [24]. The oxidation rate increased with temperature up to 570 °C and then decreased to a minimum at 850 °C. He showed that below 570 °C, a Fe–4.9Al alloy forms a scale composed of Fe₂O₃, Fe₃O₄ and FeAl₂O₄, and attributed the high oxidation rate at this temperature to the poor barrier offered by these oxides to the migration of Fe from the substrate to the surface. On increasing the temperature above 570 °C, formation of γ -Al₂O₃ at the expense of Fe₂O₃ and FeAl₂O₄, gives rise to better protection. Above 800 °C the alloy forms an almost pure aluminium oxide and the oxidation rate falls to its lowest level. This trend is seen not only in Al-containing alloys, but also in 9Cr–1Mo steel [25]. It should be noted that pure iron does not show this reverse trend in its oxidation [26], indicating that the inversion phenomenon is related to the ability of the alloying element to participate in the protective film formation.

Inversion in the oxidation rate kinetics is reported even in iron aluminides with high Al content. However, the reported literature shows the formation of alumina scale over a range of temperatures, before and after the occurrence of inversion. The authors of these studies have attributed their findings to polymorphic transformation of various form of Al₂O₃ such as θ , γ to form a protective α -Al₂O₃ [27,28]. The inversion in the present case can be related more to the ability of the alloy to preferentially form Al₂O₃ over either of Fe₂O₃ and FeAl₂O₄ than to such a phase transition. It is necessary to point out the fact that the present authors reported the formation of both Fe₂O₃ and Al₂O₃ even by the alloy 1, when oxidised at 800 °C for 10 min [17]. But

Fig. 6. X-ray diffraction patterns of alloy 3 after oxidation at (a) 700 °C and (b) 900 °C for 1000 h.

what makes the alloy 1 different from the remaining two is that the former forms a thin oxide, better in protection, than the thick poorly protecting oxide formed by the other two alloys.

It is known that scale formation involves two stages, viz. (1) nucleation and (2) growth of the oxides. Thermodynamically, aluminium has larger tendency for oxidation than does iron. Accordingly, more nuclei of Al_2O_3 than Fe_2O_3 are expected to form. What makes the difference in forming a protective scale is in the ability of Al_2O_3 to cover the surface. However, despite the

higher thermodynamic tendency of aluminium to form oxide than that of iron, Al_2O_3 has poor growth rate in comparison with Fe_2O_3 . So, Fe_2O_3 outgrows Al_2O_3 . Added to this, the thermodynamic tendency of Fe to oxidise becomes high, when its activity in the alloy becomes high, on reducing its Al content. Notwithstanding the higher thermodynamic tendency of 16% Al alloy to form Al_2O_3 , it does give rise to Fe_2O_3 along with Al_2O_3 as mentioned above. However, the alloy very quickly forms a more protective Al_2O_3 oxide beneath the Fe_2O_3 oxide and gains resistance to further

Fig. 7. SEM micrographs of alloy 1 (a) oxidised at 800 °C for 1000 h, showing growth of oxide nodules, (b) oxidised at 900 °C for 1000 h, showing cracks in the oxide scale.

Fig. 8. SEM micrograph of alloy 2, after oxidation at 900 °C for 1000 h.

Table 1
EDX analysis of the scales at locations 1 and 2, shown in Fig. 8

Location	Fe (at.%)	Al (at.%)
1	94.4	5.6
2	47.6	52.4

oxidation. When the Al content is low, the alloy suffers from not only very few nucleation of Al_2O_3 but also its inability to supply Al to grow the existing Al_2O_3 . This leads to unlimited growth of Fe_2O_3 , as observed in the present case (Fig. 9). The rise in temperature promotes the diffusion of Al from the matrix to the surface and

Fig. 9. SEM micrographs of alloy 3 oxidised at 700 °C, (a) for 100 h, displays typical iron oxides, (b) at higher magnification this oxide appears as a whiskers and (c) after 1000 h, displays porosity in the scale.

Fig. 10. Cross-sectional view of the oxide scale of alloy 3 during oxidation at 700 °C for 1000 h, showing formation of double layer of oxide, and line scan profiles of (a) Al and (b) Fe across the oxide-matrix interface.

Fig. 11. Cross-sectional view of alloy 2 after oxidation at 700 °C for 1000 h, showing spallation of oxide scales.

enhances nucleation and growth of Al_2O_3 . As a result, the alloy tends to form protective Al_2O_3 at the expense of Fe_2O_3 and FeAl_2O_4 , as observed in the present case.

The existence of Al_2O_3 as an inner layer of the scale shows that formation of Al_2O_3 has taken place from inward diffusion of O through the iron oxides rather than outward diffusion of Al, unlike as mentioned in the alloy 1. However, Fe_2O_3 forms here by outward diffusion of Fe.

In order to separate the effect of temperature on the oxidation kinetics of these two alloys from that of the oxide characteristics, the following experiments were conducted. Alloy 3 was oxidised initially at 800 °C for 25 h and subsequently it was subjected to oxidation at 700 °C for 100 h. In this case, the weight gain turns out to be 1.48 mg/cm², interestingly, which is much less compared to the weight gain corresponding to one oxidised directly at 700 °C for 125 h, that is 17 mg/cm². This is probably due to formation of some aluminium oxide at 800 °C, which inhibits the preferential oxidation of Fe on oxidation at 700 °C. Ignoring the influence of oxide, the effect of temperature on the oxidation tendency of low-Al alloy was further examined. Thus, alloy 3 was oxidised at 600 °C for 100 h. For the same intervals of time the weight gain of the alloy followed the order 700 > 600 > 800 °C. The lower weight gain by the alloy at 600 °C, in comparison with 700 °C, possibly indicates that when the alloys form less protective oxides, its oxidation tendency is lowered with lowering temperature. The drop in oxidation rate at 800 °C is due to the ability of the alloy to form a better protective oxide. Once it forms a better oxide, a further rise in temperature promotes the oxidation tendency of the alloy.

It is necessary to comment on the shape of the kinetic curves. The plots of 700 and 800 °C show a smooth gain in weight with time, while at 900 °C the plot shows some instability during the early stage of oxidation. After attaining a high weight gain initially, the weight of the alloy drops within 400 h of exposure. However, the alloys do not show much fluctuation in weight on further oxidation (Fig. 4d). The initial loss in weight could be due to spallation. This could be possibly assisted by disintegration of carbides into CO/CO₂. The alloy could stabilise from spallation either by forming a protective layer of Al_2O_3 , which could prevent further decarburization, or possibly all the surface carbides are removed during the initial stages of oxidation. This aspect needs further examination.

6. Conclusions

1. The oxidation resistance of Fe-*x*Al-1C alloy deteriorates only marginally on lowering its Al content from 16 to 10%. When the Al content is reduced further to 8%, it becomes more prone to oxidation. Comparing the literature data with the present results, it becomes clear that carbon is very

detrimental to oxidation resistance of Fe–Al when its aluminium content is low.

2. When the Al content of the Fe– x Al–C alloy is at 10% and above, it gives rise to scale, which is aluminium oxide. At 8% Al, the iron oxide coexists with aluminium oxide, the latter exists close to the substrate.

Acknowledgements

The project is sponsored by DRDO, Government of India. The authors thank Dr. D. Banerjee, Director, Defence Metallurgical Research Laboratory for the interest shown in this collaborative work.

References

- [1] Liu CT, Stiegler JO, Fores FH, Ordered intermetallics. In: Metals handbook, vol.2. Metals Park (OH) ASM Int. 1990. p. 920.
- [2] De Van JH, Tortorelli PF. Mater High Temp 1993;11:30.
- [3] Tortorelli PF, Natesan K. Mater Sci Eng 1998;A258:115.
- [4] Prakash U, Buckley RA, Jones H, Sellars CM. ISIJ Int 1991; 31:113.
- [5] McKamey CG, De Van JH, Tortorelli PE, Sikka VK. J Mater Res 1991;6:1779.
- [6] Stoloff NS. Mater Sci Eng 1998;1:A258.
- [7] Kerr WR. Metall Trans 1986;17A:2298.
- [8] Baligidad RG, Prakash U, Radhakrishna A, Ramakrishna Rao V, Rao PK, Ballal NB. Script Mater 1997;36:667.
- [9] Baligidad RG, Prakash U, Ramakrishna Rao V, Rao PK, Ballal NB. ISIJ Int 1996;36:1453.
- [10] Baligidad RG, Prakash U, Radhakrishna A. Mater Sci Eng 1999;A265:301.
- [11] Vayas S, Viswanathan S, Sikka VK. Script Metall 1992;27:185.
- [12] Baligidad RG, Prakash U, Ramakrishna Rao V, Rao PK, Ballal NB. Iron Making Steel Making 1994;21:324.
- [13] Gehrmann F, Grabke HJ, Riecke E. In: Turnbull A, editor. Proc. hydrogen transport and cracking in metals. London: Institute of Materials, 1995. p. 216.
- [14] Shankar Rao V, Baligidad RG, Raja, VS. Oxide, Met, communicated.
- [15] Baligidad RG, Prakash U, Ramakrishna Rao V, Rao PK, Ballal NB. ISIJ Int 1995;35:443.
- [16] Baligidad RG, Radhakrishna A. Mater Sci Eng 2000;A283:211.
- [17] Shankar Rao V, Norell M, Raja, VS. Corros Sci, communicated.
- [18] Raghavan V. Phase diagram of ternary iron alloys, Part-1. Metals Park (OH): ASM, 1987.
- [19] Palm M, Inden G. Intermetallics 1995;3:443.
- [20] Tomaszewicz P, Wallwork GR. Oxid Met 1983;19(5/6):165.
- [21] Caplan D, Sproule GI, Hussey RJ, Graham MJ. Oxid Met 1979; 13(3):255.
- [22] Kao CH, Wan CM. J Mater Sci 1988;23:1943.
- [23] Saegusa F, Lee L. Corrosion 1966;22:168.
- [24] Boggs E, William J. Electrochem Soc 1971;118:906.
- [25] Taylor JW, Trotsenberg. In: Holmes DR, Hill RB, Wyatt LM. editors. Corrosion of steels in CO₂. Proc. BNES, 1974. p. 180.
- [26] Paidassi J. Acta Metall 1958;6:185.
- [27] Mignone A, Frangini S, Barbera ALA, Tassa O. Corros Sci 1998; 40:1331.
- [28] Natesan K. Mater Sci Eng 1998;A258:126.