1. INTRODUCTION

Excellent high temperature corrosion resistance, high strength to weight ratio and low cost of the constituent elements make Fe$_3$Al based intermetallics a good choice for high temperature applications. However, their poor room temperature ductility and loss in mechanical strength, especially creep strength, above 550°C have impeded them from use in structural applications. Several approaches have been made to address the above two problems. Exclusion of moisture, which is the source for hydrogen causing embrittlement, alloying intermetallics with Cr, production of fine grain alloy through powder metallurgy route of synthesis and even low aluminium Fe-alloys as another alternative are considered effective methods to counter the embrittlement and hence the low ductility of these intermetallics. To promote high temperature creep strength, alloying additions such as B, heat treatment and dispersing the intermetallic with oxides are considered as effective methods. The reduction in Al content, though, improves the low temperature ductility, does not impart high temperature creep resistance.

To address the twin problems of room temperature ductility and high temperature creep strength, Baligidad et al. developed dual-phase iron aluminides by alloying Fe-16 wt% Al alloy with C. A variety of dual-phase iron aluminides were produced by air induction melting, followed by electroslag refining techniques. In comparison with single-phase Fe$_3$Al-based intermetallics, these intermetallics have been shown to exhibit better mechanical properties. This has been attributed to carbon exerting interstitial solid solution strengthening at low concentrations and precipitation hardening at high concentrations by the formation of carbides. These alloys further exhibit a reduced susceptibility to environmental embrittlement. Parvathavarthini et al. have shown that the carbides of these iron aluminides act as trap sites for hydrogen.
However, this improvement in ductility was not good enough to process the material for structural applications. A substantial improvement in the ductility of these alloys was further achieved by reducing the Al content for various C levels. As any change in compositions of an alloy would have its impact on its high temperature oxidation resistance, oxidation studies of these alloys have become important. Hence, a systematic study was undertaken in our laboratory to examine a series of Fe-Al-C alloys produced through electroslag refining route, where Al existed in stoichiometric 11,12 as well as less than stoichiometric 13,14 proportions corresponding to Fe₃Al intermetallic compounds. This paper briefly reviews the existing work on the oxidation behaviour of Fe-Al-C⁺ alloys, as no such review exists in the published literature. A working model is provided to explain the observed oxidation behaviour of these alloys.

2. HIGH TEMPERATURE OXIDATION BEHAVIOUR OF TWO-PHASE CARBON CONTAINING Fe₃Al INTERMETALLICS

Recently, Shankar Rao et al 11 published a detailed study on the long-term oxidation behaviour of Fe-16Al intermetallics with C ranging form 0.05 to 1%. The study was conducted in still air, in the temperature range from 700 to 1000°C, and a comparison was made with Type 310 stainless steel (SS). Their study shows that the aluminides exhibited a marginally higher oxidation rate than that of 310 SS at 700°C (Fig.1). During early stages of oxidation, the alloys seem to exhibit a marginal rise in oxidation tendency with C content, though this effect becomes less discernible during long-term oxidation. Interestingly, the alloy with the 0.14% C, namely, Fe-16Al-0.14 C exhibits the lowest oxidation rate among all the alloys examined, even though Fe-16Al-0.05C contains the lowest C among all these alloys. The cracks present in the as-produced condition of this alloy make the latter more susceptible to oxidation than the other alloys (Fig.2). The difference in long-term oxidation behaviour among the aluminides and between the aluminides as a whole and 310 SS becomes noticeable at higher temperatures. At 1000°C, where spallation occurs, after long duration of exposure, the detrimental effect of C becomes visible and at this temperature 310 SS becomes very unstable 11.

Typical weight-gain curves of all these alloys oxidised at 900°C are brought out in Fig.3 for illustration. The presence of C in these alloys results in the formation of Fe₃AlC₀.₆₅ carbide phase, as revealed by x-ray diffraction (Fig.4) and microscopy studies (Fig.5). Discounting a marginal gain in the oxidation
resistance of Fe-16Al-0.14C alloy in comparison to Fe-16Al-0.05C alloy, which occurred due to the disappearance of cracks in this system and thereafter, the increase in C content of the alloy has been found to increase the carbide content of the alloy (compare Figs. 2 and 5). Hence the second phase of Fe-16Al-xC intermetallic, namely, the carbide can influence the oxidation of Fe-16Al intermetallics different from the single-phase Fe-16Al intermetallics in one or more ways. Work by Gesmundo and Gleeson 15 shows that oxidation behaviour of binary and ternary two-phase alloys is very complex for a rigorous theoretical treatment.

Hence, Shankar Rao et al have explained the role of C on the observed oxidation behaviour of two-phase iron aluminides based on microscopy and scanning Auger electron spectroscopy analysis 12. Secondary electron image of Fe-16Al-1C alloy oxidised at 800°C for 10 min. is shown in Fig.6. Existence of two distinctly differing regions of oxides is very clear from the micrograph. Comparing Fig.5 and Fig.6 it can be said that the thicker oxide regions of Fig. 6 correspond to the dark-contrast, carbide phases of Fig. 5. They further substantiated this proposition by showing that the substrate below the thicker oxide is rich in C content corresponding to carbide phase 11. These authors carried out a detailed scanning Auger analysis of the oxides of the two regions. The depth profile analysis of the oxide on the dendritic phase,
namely, Fe$_3$Al is exemplified in Fig.7. Table 1 summarizes the data obtained on both the Fe$_3$Al and Fe$_3$AlC$_{0.69}$ phases. The following characteristics of oxides become clear. Firstly, both Fe$_3$Al phase (dendrite) and Fe$_3$AlC$_{0.69}$ phase (interdendrite) form bilayer-oxides, the top layer being Fe-oxide and the bottom layer Al-oxide. Fe$_3$AlC$_{0.69}$ forms thicker oxides than Fe$_3$Al, though Al/Fe oxide ratio and the chemistry of oxides formed over both the phases remained close to each other. The higher oxidation tendency of carbides as revealed by the thicker oxide on it than the Fe$_3$Al phase can be attributed to two factors. Firstly, it may due to the low Al (8.7 at %) content of Fe$_3$AlC$_{0.69}$ than the high Al (28 at %) content of Fe$_3$Al. Secondly, a possible detrimental role C plays in Fe-Al alloys, though it has been reported to exhibit a beneficial effect towards oxidation resistance of pure Fe $^{16}$. A close look at the surface morphology of oxides formed on the carbide and the matrix indicates that the oxide formed over carbides are more porous than that formed on Fe$_3$Al phase. Kao and Wan suggested that carbon present in Fe-5.5Al-0.65C $^{17}$ and Fe-7.5Al-0.65C $^{18}$ leads to the formation of CO and/or CO$_2$, which eventually escape from the surface leaving behind pores and cracks in the oxide. As a consequence, easy diffusion of O$_2$ through these defects becomes possible. The resultant less protective oxide formed on carbide needs to grow thick before it can offer adequate protection to the base alloy.

Kao and Wan showed that the rise in Al content form 5.5 to 7.5 % Al has significantly retarded that decarburization tendency of Fe-Alx-0.65C alloy$^{17, 18}$. Indeed, the cross-section analysis of the current Fe-16Al-xC alloys indicates that they are

![Fig. 7: Scanning AES depth profile data of various species of the oxides of the matrix phase (Fe$_3$Al) are shown in the figures 5 and 6. Stratified iron-rich outer most oxide and aluminium-rich inner most oxide scale are brought out for illustration $^{11}$](image)

<table>
<thead>
<tr>
<th>Region / Phase</th>
<th>Oxide Film Thickness at Dendritic and Interdendritic Regions</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Al-oxide (nm)</td>
</tr>
<tr>
<td>Dendritic / Fe$_3$Al</td>
<td>200</td>
</tr>
<tr>
<td>Interdendritic / Fe$_3$AlC</td>
<td>850</td>
</tr>
</tbody>
</table>
resistant to decarburization. The accumulation of carbides (Fig.8) beneath the oxide scale of the alloy clearly supports this view. Nevertheless, decarburization of surface carbides of the alloy can continue to occur, even in Fe-16Al-xC alloys, until the porous oxide grows thick enough to stop the diffusion of O₂ to the substrate/oxide interface. Perhaps, at this stage the oxidation rate of carbides might become comparable to that of Fe₃Al phase. This could be one of the reasons why the effect of C is not seen in samples oxidised for very long time. Possibly, the carbides too develop protective oxides albeit being thick. However, should the spallation occur, the process of oxide scale formation needs to restart again resulting in more loss of material over the carbide than Fe₃Al phase. In the context of spallation, carbides are found to be inferior to Fe₃Al phase. Thus, at high temperatures, the spallation tendency of Fe-16-xC alloy increases with the C carbon content, which is seen at 1000 °C.

These alloys showed interesting oxidation kinetics with change in temperature. Fig.9 is shown to illustrate the effect of temperature on the oxidation behaviour of Fe-16Al-0.14C alloy. The alloy shows an inversion in the oxidation tendency above 900°C. In the early stages (<75h) of oxidation, when no spallation was noticed, the alloy gained more weight at 900°C than at 1000°C. Previous studies on binary Fe-Al alloys report that this inversion can occur due to phase transformation of Al₂O₃ from less protective q 19 and g 20 type phases to a more protective α-phase. Since Shankar Rao et al did not find any such transition, as they found α-Al₂O₃ to be present in samples oxidised at all the temperatures, 700-1000°C, they examined the role of possible phase transformation with in the alloys. Based on the published literature on the phase transformation in Fe-Al-C alloys 21-23, they suggested that, at 1000°C, the oxidation resistance of the alloy increases because of the following simultaneous events occurring, namely, decreasing volume fraction of carbides and the rising Al content of the remaining carbides. The first one contributes to overall resistance of the alloy and second to the resistance of the carbide phase by itself. This sounds logical because of the fact that carbides are the weakest link in oxidation resistance.

Fig.8a: XRD pattern of Fe-16Al-1C alloy after oxidation at 900 °C for 1000 h in still air. Comparing XRD pattern of Fig.4, it becomes clear that the oxidised alloy exhibits a relatively higher intensity peaks for carbide 11.

Fig.8b: Enrichment of carbides on the surface of Fe16Al-0.14C alloy after oxidation at 1000 °C for 1000 h, brought out through a cross-sectional imaging of the samples through an optical microscope. The dark contrast particles are carbides 11.

Fig. 9: Weight-gain behaviour of Fe-16Al-0.14C at various temperatures. Note a reversal in the weight-gain tendency of the alloy at 1000°C.
of Fe-16Al-xC alloys. It should, however, be pointed out that their study corresponds to 25 h of cyclic oxidation, in still air, and therefore the carbide phase would be regenerated at the end of every cycle. So the carbides formed during cooling can affect the oxidation at the beginning of every cycle.

3. HIGH TEMPERATURE OXIDATION BEHAVIOUR FE-AL-C ALLOYS CONTAINING MODERATE AL LEVELS

As mentioned in the introduction, lowering Al content enhances the ductility of Fe-Al alloy. However, it does not improve the high temperature mechanical properties of the alloy and worst, it can lead to very poor oxidation resistance, if its Al content is lowered beyond certain levels. While the two-phase structure of the alloy can improve the mechanical properties, a critical Al content in two-phase alloys needs to be balanced so as to ensure the formation of a protective oxide scale. In binary Fe-Al alloys, Al content just above 7% is known to develop external Al₂O₃ scale. The critical Al content needed to develop protective oxide would also depend on the alloying additions and the nature of phases present in the alloy. Because of this, Shankar Rao et al 13 and Shankar Rao and Raja 14 examined respectively the oxidation behaviour of Fe- (10 & 8)Al-1C and Fe-(10 & 8)Al-0.5 C alloys, in which Al has been present in moderate levels.

Effect of C on the oxidation behaviour of low aluminium Fe-Al alloys was studied as early as 1971 by Boggs 24. His short-term oxidation studies indicate that Fe-5.5%Al alloy shows significant rise in their oxidation tendency with C content at 800°C in 700 Torr O₂ atmosphere. However, the oxidation tendency at 500°C is not much different from that of Fe-5Al even when the alloy contains 0.1%C, the largest C addition they made to the alloy. Oxidation studies involving the next higher level of Al and C containing Fe-Al-C alloys are due to Kao and Wan 17,18. Gravimetric data of Fe-7.5Al-0.65C is shown in Fig. 10. To get an idea about what happens if further C is added to the alloy, the weight-gain data of Fe-8Al-1C is shown in Fig. 11 13. The weight-gain data of 10 and 16% Al alloys with similar C content are also included in this figure for a comparison. Fig.10 shows a significant rise in weight-gain when the temperature is raised from 600 to 700 °C, but the subsequent raise in temperature causes a substantial reduction in weight-gain by the alloy. These alloys are much more prone to spallation than high Al containing Fe-Al-C alloys. Kinetic data of selective alloys are summarized in Table 2. Though, these studies were conducted with varying degree of O₂ partial pressures, all of them correspond to high O₂ partial pressures. Hence, with reasonable approximation, these data can highlight the effect of Al and C on the oxidation behaviour of Fe-Al alloys.
Table 2
SUMMARY OF 800°C OXIDATION DATA OF VARIOUS ALLOYS

<table>
<thead>
<tr>
<th>S.No.</th>
<th>Alloy and Environment</th>
<th>Weight gain in 25 h, µg/cm²</th>
<th>Rate constants, g²/cm⁴/s</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>kp¹</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>Fe-7Al, 200 Torr dry O₂</td>
<td>50</td>
<td>4.72×10⁻¹³</td>
<td>1.2×10⁻¹³</td>
</tr>
<tr>
<td>2</td>
<td>Fe-7.5Al-0.65C dry air/100 ml/min. 200 Torr dry O₂</td>
<td>700</td>
<td>4.9×10⁻¹²</td>
<td>6×10⁻¹³</td>
</tr>
<tr>
<td>3</td>
<td>Fe-8Al-1C⁰, Still air</td>
<td>2000</td>
<td>6.9×10⁻¹¹</td>
<td>3.6×10⁻¹²</td>
</tr>
<tr>
<td>4</td>
<td>Fe-5.5Al-0.5C dry air/100 ml/min.</td>
<td>4000</td>
<td>2.1×10⁻¹⁰</td>
<td>5.7×10⁻¹⁰</td>
</tr>
<tr>
<td>5</td>
<td>Fe-5Al, 200 Torr dry O₂</td>
<td>550</td>
<td>5.9×10⁻¹¹</td>
<td>2.4×10⁻¹²</td>
</tr>
<tr>
<td>6</td>
<td>Fe-16Al-0.5C, Still air</td>
<td>130</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

The first rate constant, kp¹, having the low value, which also occurs in the early stages of oxidation, is attributed to the formation of the outer iron-rich oxide scale. In comparison, the second one, kp², derived from the later part of the oxidation data and having the higher value than kp¹ is attributed to the formation of inner aluminium rich oxide scale (Fig.12). Generally, the first rate constant is higher than the second one, which is in agreement with the fact that growth kinetics of iron oxides are faster than that of aluminium oxides. It is instructive to compare the rate constants of these alloys from the top to bottom of the table. Addition of C to Fe-7Al alloy (compare alloys 1 and 2) leads to a raise in kp¹ and kp² by ten and five times respectively. If the C content is further enhanced (compare alloys 2 and 3) then the values of rate constants rise even further. Thus C seems to promote the growth of iron nodules. In fact, Shankar Rao and Raja has shown that Fe-8Al-1C alloy exhibit iron oxide whiskers, even though it has 8 % Al. Such whisker formation is not seen in Fe-7Al alloys. Interestingly, alloy-4 exhibits a lower value of kp¹ than kp², which implies that if Al content is further reduced, higher concentration of C affects the formation of aluminium rich oxide more than that it does for iron rich oxide, though the growth of both the oxides scale are facilitated by the C content of the alloy. Alternatively it can be argued that the alloy lost its ability to form a

Fig.12: SEM cross-sectional image of Fe-8Al-1C oxidised at 700 °C for 1000 h. Line scan profiles of Al and Fe are shown in the top and bottom images shown above.
protective aluminium oxide, instead it forms spinel phase(s) such as FeAl$_2$O$_4$. Fig.12 indeed shows that the inner layer consists of both Al and Fe cations even in Fe-8Al-1C. This phase forms in Fe-Al binary alloys, when their aluminium content lays around 2\% 26. Thus, the detrimental effect of C seems to be high when its content in alloy is more and/or when Al content of the alloy is reduced. The above fact is further substantiated by comparing rate constants of alloys 6 and 7, where addition of even 0.5 \% C to Fe-5Al alloy enhances the growth both the inner and outer oxides significantly, the effect being pronounced more for the inner oxide than the outer oxide. Attention is drawn to the fact, that even in Fe-16Al-1C alloy, the carbide phase formed a thicker oxide scale than Fe$_3$Al phase (Fig.6). The oxide was found to be porous, though once a thick oxide is formed; its oxidation kinetics slowed down. This has become obvious in the case of low Al containing alloys. The increase in C content might facilitate the formation of large quantity of CO/CO$_2$ and thereby raising the oxide porosity. However,

Two phases oxidise independently

![Schematic model explaining the oxidation behaviour of C containing iron aluminides and Fe-Al alloys](image)

Differences in low temperature (independent oxidation of carbide and the matrix) and high temperature (co-operative oxidation of the matrix and the carbide through inter-diffusion of C and Al and the simultaneous inward diffusion of C) behaviours, formation of carbides at the interfaces on cooling the high-temperature oxidized sample, the effect of lowering Al and enhancing C informing a less protective FeAl$_2$O$_4$ oxide instead of highly Al$_2$O$_3$ oxide are explained.

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Fig.13: Schematic model explaining the oxidation behaviour of C containing iron aluminides and Fe-Al alloys is shown. Differences in low temperature (independent oxidation of carbide and the matrix) and high temperature (co-operative oxidation of the matrix and the carbide through inter-diffusion of C and Al and the simultaneous inward diffusion of C) behaviours, formation of carbides at the interfaces on cooling the high-temperature oxidized sample, the effect of lowering Al and enhancing C informing a less protective FeAl$_2$O$_4$ oxide instead of highly Al$_2$O$_3$ oxide are explained.
Once decarburization is stopped/or C disappears from the surface, the oxidation kinetics can slow down. The role of C on oxidation also becomes clear, when the oxidation tendency of Fe-6Al alloy is compared with that of Fe-8Al-1C alloys. Oxidation of Fe-6Al alloy \(^{24}\) at 800°C for 5 h resulted in a mere weight gain of 0.2 mg/cm\(^2\), whereas for the similar condition, Fe-8Al-1C alloy \(^{13}\) exhibited 2 mg/cm\(^2\) weight gain, about an order of magnitude higher than the former. This happens in spite of the fact that the latter alloy has more Al than the former. This is further amplified in Fe-5Al alloys (compare alloy 4 and 6) Notably the effect of C is not very significant when the Al content is enhanced to 10% and above (Fig.11).

The following reason seems to explain why the effect of C on the susceptibility of alloy to oxidation is marginal for high Al containing alloys. Unlike chromium oxide, aluminium oxide has low growth rate. Hence it has less ability to cover those (microstructures) areas, which form less protective oxide. This effect will be further compounded, when these microstructures contain less Al content and/or detrimental elements such as C. Moreover, in high oxygen partial pressure conditions preferential oxidation of reactive element such as Al is not possible and this makes it even more difficult for the vulnerable areas to form a protective oxide scale. When the alloy possesses high Al content, then the formation of Al\(_2\)O\(_3\) is facilitated and hence the vulnerable microstructures can be quickly protected.

Similar to Fe-16Al-xC alloys, the low Al containing alloys also exhibit inversion in oxidation kinetics with temperature. For example, Fe-8Al-1C alloy at 800°C gains only one tenth of the weight as it gains at 700°C \(^{13}\). Further raise in temperature to 900°C, however, leads to higher weight gain. At 600°C, however, the alloy shows weight gain that lies between those weights gained between 700°C and 800°C. Such behaviour is also well exemplified in Fig.10 from the work of Kao and Wan \(^{18}\) on Al-7.5Al-0.65C. Shankar Rao et al \(^{13}\) carried out a few planned experiments to examine the reason for this inversion. A sample was initially oxidised at 800°C for 25 h and then at 700°C for 100 h and another sample was oxidised at 700°C for 125 h in a single step. In the first case the sample exhibited a mere 1.48 mg/cm\(^2\) rise in weight, while in the latter case it showed a weight-gain of 17 mg/cm\(^2\). What makes the first sample different from the second is that the initial high temperature (800°C) exposure for 25h in the first case enables the alloy to form a protective aluminium oxide scale, because of faster diffusion kinetics of Al at 800°C in comparison with that at 700°C. Higher diffusion kinetics of Al not only enables the formation of Al\(_2\)O\(_3\) protective scale just above the surface, but also helps in its outward growth to cover less Al containing phases such as Fe\(_3\)AlC\(_{0.69}\). Thus different explanations were advanced for the observed inversion kinetics of Fe-16Al-xC \(^{11}\) alloys and low Al \(^{13}\) containing alloys.

4. DEVELOPMENT OF A PHYSICAL MODEL FOR THE OXIDATION OF Fe-16Al-xC ALLOYS

Various stages of scale formation during oxidation of single phase Fe-xAl alloys at 800°C in dry O\(_2\) of pressure 200 Torr have been schematically illustrated by Tomaszewicz and Wallwork \(^{24}\). According to them, alloys with less than 7% Al form stratified oxide layers, where the outermost layer consists of iron oxide and the inner most Al\(_2\)O\(_3\) layer. If the Al content of the alloy becomes very less (1-2%), instead of Al\(_2\)O\(_3\), the spinel FeAl\(_2\)O\(_4\) phase can form. Above 7% Al, the alloys form protective Al\(_2\)O\(_3\) scale. In addition, alloys containing Al content less than 7% shows two different parabolic rate constants; the first one occurring at the earlier stages of oxidation, being lower than the second one, occurring at later stages of oxidation. In contrast to this, Fe-8Al-0.5C \(^{14}\) and Fe-8Al-1C \(^{13}\) exhibit thick oxide scale, which contains Fe-rich outermost layer and Al rich inner most layer, despite these alloys contain 8% Al that is sufficient to form a protective Al\(_2\)O\(_3\) scale in binary alloys. In fact, even Fe-10Al-1C alloy is prone to nodular iron oxide growth at 900°C in spite of its high Al content. The presence of carbide phase in these alloys, as opposed to its absence in single-phase binary Fe-Al alloys seems to be responsible for this deterioration in the oxidation tendency of Fe-8Al alloy. When the Al content is raised to 16%, the formation of iron oxide nodules is stopped and the scale becomes a thin protective aluminium oxide. Thus the modification in the oxide scale chemistry, due the presence of C, is responsible for the higher oxidation kinetics of Fe-Al-C alloys.
As seen from the Table-2, the addition of C is detrimental to the oxidation resistance of Fe-Al alloys. The effect becomes very pronounced at low Al concentrations. Fe-Al-C alloys exhibit distinctly two-phase structures, as opposed to Fe-Al binary alloys. So the mechanism of oxidation of these alloys must take into account the oxidation behaviour the two phases. Gesmundo and Glesson have treated two-phase oxidation of binary alloys quantitatively and ternary alloy qualitatively. As has been pointed out by them, an element commonly existing in both the phases needs to exhibit the same thermodynamic activity, irrespective of any concentration difference this element may exhibit between both the phases. As a result, an element can exhibit the same thermodynamic oxidation tendency and also exhibit the same kind of equilibrium with the oxide phases of the alloy irrespective of in which phase it exists. However, the continued availability of an element to maintain equilibrium with the oxide phase depends on the composition (availability) and diffusivity of the element in a given phase. At least in the context of concentration of alloying elements, Fe<sub>3</sub>AlC<sub>0.69</sub> significantly differs from Fe<sub>3</sub>Al phase. The lower Al content of the former than the latter will make the former more susceptible to oxidation than the latter. In addition, C present in Fe<sub>3</sub>AlC<sub>0.69</sub> phase creates cracks and pores during oxidation, which is responsible for easy access of O<sub>2</sub> to the oxide/ alloy interface. When composition of two phases differs significantly, the O<sub>2</sub> partial pressure can have a profound influence on the nature of protective scale formed on the phases. At low partial pressures, the more reactive element gets selectively oxidised and hence a phase can form a scale predominantly of reactive element, even if it contains low activity of reactive element. On the other hand, when the O<sub>2</sub> partial pressure is high, in the present cases, even the less reactive element can be subjected to oxidation and hence the composition of the oxide scale of the phase depends truly on alloy composition and to some extent on the oxide growth rate. Such cases can lead to mixed oxides formation until some kinetic factors over-take the thermodynamic oxidation tendency of the alloy.

The other factor that could affect the oxidation of the two phases is their thermal stability at the oxidation temperature. Being a three-component system, the phase in FeAl-C alloy can exhibit differing composition at a given temperature and there by encourage diffusion of alloying elements between the two phases, without eliminating either of the two phases. Three things become clear from the literature: with raise in temperature a) the solubility of C in Fe<sub>3</sub>Al increases, b) Al content of Fe<sub>3</sub>AlC<sub>0.69</sub> increases at the expense of its C content and c) the volume fraction of Fe<sub>3</sub>AlC<sub>0.69</sub> decreases. Though the second and third aspects of the alloy promote its oxidation resistance, the third aspect can cause a loss in mechanical strength of the alloy only for which carbides are formed.

Before providing a physical model, a close look at the oxidation behaviour of Fe-16Al-xC alloys is important. The data shows that during early stages of oxidation at 700°C the raise in C content of the alloy increases its oxidation tendency (Fig.6). Such variations are not well observed at higher oxidation temperatures. High thermal stability of carbides below 750°C and low Al diffusivity allow the vulnerable carbide to oxidise with little influence by aluminium rich Fe<sub>3</sub>Al matrix. However, what makes these carbides different from the carbides of Fe-15Cr-0.5C alloy is that in the latter chromium carbides of the alloy, which is only a small fraction of the matrix, is rich in reactive element (Cr), while the surrounding matrix is low in this element. This is exactly a reverse situation of what is seen in Fe-Al-C alloys (where Al lean carbides are surrounded by Al rich intermetallic phase). Hence, in spite of slower growth rate of Al<sub>2</sub>O<sub>3</sub> than Cr<sub>2</sub>O<sub>3</sub>, it is able to cover the vulnerable (Al lean carbides) areas more quickly than Cr<sub>2</sub>O<sub>3</sub> could do in Fe-Cr-Al alloys. Thus, carbides in Fe-15-Cr-0.5C cast alloys cause more damage to oxidation than carbides in Fe-Al-C alloys.

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found to be detrimental to oxidation, while it can contribute to the strength and ductility of the alloys. Based on the above discussion a schematic model as shown in Fig. 13, can be proposed to explain the observed behaviour of Fe-16Al-xC alloys.

5. SUMMARY

Oxidation behaviours of C containing Fe₃Al intermetallic and Fe-Al alloys having moderate levels of Al have been discussed. Increasing C and decreasing Al contents are shown to enhance the oxidation tendency of Fe-Al alloys. While decarburisation is a problem only in low Al alloys, the carbides seem to be unstable at high temperatures even in alloys with high Al content. A schematic model has been provided to explain the role of carbides on the oxidation tendency of Fe₃Al(C) alloys. The dissolution of carbides, though enhances the oxidation resistance of these two-phase alloys, can result in losing their high temperature strength gained by their presence. Since the carbides are known to enhance the room temperature embrittlement resistance of Fe-Al alloys, their presence is essential to enable room temperature mechanical processing of these intermetallics. Hence, attempts need to be made to improve their high temperature stability and oxidation resistance if this property of the needs to be retained. Chromium carbides are known to exhibit high oxidation resistance 25 but the effect of Cr alloying with Fe₃AlC₀.₆₉ is not known. This becomes important because of the fact that Cr has been found to be detrimental to the oxidation resistance of Fe₃Al. A detailed study is required to explore the beneficial effect of Cr in these C containing aluminides, as the carbide containing iron aluminides are in any case found to be inferior even to Cr alloyed Fe₃Al aluminides 20, 11.

REFERENCES