Effect of alloyed Al on the Corrosion Behaviour of Ni-base Alloys in Molten Glass under Static Condition

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Abstract

Ni-base alloys with different Al contents of 0-10 wt. % were designed and the corrosion behaviour was evaluated by static immersion in molten glass at 1050 °C. Results indicate that a small amount of Al (2-5 wt.%) is advantageous to mitigate the corrosion of the commercial Ni-base alloy, whereas further increase of Al content (8-10 wt.%) accelerates the corrosion process. In Al-added alloys, the corrosion of Cr element is suppressed by the Al dissolution, and the formation of inner alumina in high Al alloys leads to the surface exfoliation. The corrosion mechanism strongly depends on the Al content in alloys.

Keywords: A. Ni-based alloy; A. Glass; B. SEM; B. EPMA; B. XRD; C.

High-temperature corrosion

1. Introduction

In the glass industry, some metal components in contact with the molten glass suffer from severe corrosion issue when operating at high temperatures (>1050 °C). The molten glass corrosion is similar to that caused by the molten salt [1, 2], where metallic elements are easily oxidized by the dissolved oxygen or other oxidizing agents in glass [3, 4], for example, Fe^{3+} . Meanwhile, the reaction products are normally non-protective due to the high solubility in melts and a flowing of molten glass. Thus, alloys corrode rapidly in molten glass. The dissolution of metal ions is normally noticeable from the colour change of the glass [5]. In order to reduce the influence of the alloy corrosion on the production quality as much as possible, alloys used in this circumstance need to meet a basic requirement of corrosion resistance to molten glass. In some application for fiber-glass production, alloys should possess other desirable properties such as a good oxidation resistance and mechanical properties, due to the high-speed rotation during fiberizing operation [3].

Extensive investigations on the corrosion of metals or alloys in molten glass have been carried out in previous literature [3, 5-11]. Typical pure metals such as Pt [12], Ni, Fe, Cr and Co [5], Zr [11], together with alloys of Ni-base [7], Co-base [7] and

MoSi_x [13-15] were focused on their corrosion rate, interfacial reactions, electrochemical process and so on. Results suggest that the corrosion behaviour of alloys is strongly dependent on the glass temperature and composition. Carton [6] reported that chromia layer develops on pure Cr at temperatures lower than 1025 °C. As a consequence, the corrosion rate decreases significantly. Unfortunately, such a thick, protective layer is not stable when temperature is above 1050 °C [5, 6]. In addition, the composition of glass also influences the electrode reaction at the melt/alloy interface. Since the standard potential of redox for each species differs from others [8], oxidation-reduction reaction varies with the change in glass composition [9]. For instance, the reduction redox in SiO₂-PbO, SiO₂-Na₂O, and SiO₂-B₂O₃ glass is Pb²⁺/Pb, Na⁺/Na, and H⁺/H₂, respectively [9]. Moreover, the corrosion process is affected by the composition of alloys to some extent. Lizarazu [8] found that the corrosion rate changes with the Cr content in Ni-Cr alloys: Alloys with low and high Cr-content exhibit a corrosion rate more than a decade higher than that with intermediate Cr content, which is related to the different reactions referred to the alloys with various Cr contents.

Currently, most alloys used in glass industry are chromia-forming Ni-base superalloys because they meet the requirements of good oxidation resistance and mechanical properties at high temperatures. However, on basis of above investigations, a protective Cr₂O₃ layer hardly forms on the high-Cr alloys at practical operation temperatures (at least 1050 °C). Besides chromia, alumina is another protective oxide layer commonly used in high-temperature applications. However, different perspectives have been proposed on the availability and protectiveness of such an alumina layer. Previous studies on Fe-Al alloys indicate that the Al-containing alloy is capable to form Al₂O₃ layer in molten glass and thus slows down the corrosion rate [16, 17]. However, Abdullah [18] found that Ni-8Al-28Cr alloy cannot be used in molten glass because of the extremely high solubility of Al₂O₃. Moreover, earlier studies mainly focused on model alloys with single Al content. Until now, the effect of Al content on the corrosion mechanism has not been systematically investigated. Therefore, it is important to clarify the effect of alloyed Al on the

corrosion mechanism of alloys.

In this paper, the role of Al content in the corrosion process of alloys in molten glass was investigated by designing Ni-based alloys with various Al contents of 0, 2, 5, 8, and 10 in mass percentage. The alloys were statically immersed in molten glass at 1050 °C for durations up to 100 h. The corrosion behaviour of alloys in the early and late stages were evaluated through the microstructure characterization. The corrosion depth of the designed alloys was compared with that for the chromia-forming commercial alloy. The corrosion mechanism for alloys with different Al contents was also established. The knowledge of alloyed-Al effect will provide useful information for alloy design in future.

2. Materials and experimental procedure

The nominal composition of the investigated alloys in this study is listed in Table 1. Among which the A30 is a commercial Ni-based alloy with high Cr content, and it has been widely used in the glass industry as the nozzle plate alloy. The other four alloys were designed based on the composition of A30 with various Al contents in the range of 2-10 wt. % by decreasing the other elements in proportion. For example, when adding 2 wt. % Al into the A30 alloy, the contents of the other elements except Al correspondingly decreased by 2 % to form a new alloy (2Al). As shown in Fig. 1, the designed alloys are termed as 2Al, 5Al, 8Al, and 10Al, respectively.

The cast ingots of these newly-developed Ni-based alloy were prepared via vacuum induction melting. Samples for corrosion test were cut from alloy ingots into a dimension of 8×10×2 mm, and then ground with 600 # SiC sandpaper, followed by ultrasonic cleaning in ethanol. Each sample was placed in an alumina crucible full filled with soda-lime glass for glass-fiber production (66.0%SiO₂-17.0%Na₂O-0.2%K₂O-2.5%MgO-8.0%CaO-2.0%Al₂O₃-4.3%B₂O₃ in weight percentage). The Al concentration in the starting glass before the corrosion test was determined to be 1.0 wt.% by EPMA, which would be utilized to modify the Al concentration after corrosion test by subtracting this original value. The immersion test was performed at 1050 °C for durations of 5 h and 100 h, respectively. When the

furnace temperature reached 1050 °C, the crucible with sample and glass was put into the furnace. After the immersion test, the crucibles were taken out of the furnace and cooled in air.

The samples after corrosion were sealed in polymer for microstructure observation by TEOL JSM-6060 type field emission scanning electron microscopy (FE-SEM). The average of corrosion depth for each sample was measured based on the SEM images taken from six different regions. The phase structure of alloys was characterized by Rigaku RINT-TTR III X-ray diffraction (XRD). In addition, electron probe micro-analyzer (EPMA) was utilized to analyze the element distribution of alloys.

3. Results and discussion

3.1 Microstructures of cast alloys

The microstructures of cast alloys under back-scattered electron mode are presented in Fig. 1. Correspondingly, the phase structures of raw materials are shown in Fig. 2. Combining the results in Fig. 1 and 2, it can be observed that A30 and 2Al with low Al content are composed of fcc γ -(Ni, Cr), a solid solution of Cr and Fe in Ni lattice, and Cr-rich carbide precipitated along grain boundaries (as indicated in Fig. 1a and b). The latter has been identified to be Cr_{23}C_6 as shown in Fig. 2. From results in Fig. 1, there is a notable contribution to the microstructure evolution from alloyed Al that the amount of Al-rich phase increases with the increase in Al content: when Al content reaches 5 wt. %, minor Al-rich particles with a dark contrast are present at grain boundaries in addition to the Cr_{23}C_6 precipitates (Fig. 1c); a large amount of Al-rich phase starts to appear when Al content exceeds 8 wt.% (Fig. 1d and e), which has been proved to be NiAl phase in Fig. 2. The newly-emerged NiAl phase along with Ni-Cr phase become the dominant phase for 8Al and 10Al alloys, and this is consistent with the results in Fig. 2, where peaks for NiAl phase are detected only in high Al alloys. Therefore, the microstructure in Fig. 1 is in good agreement with the results of XRD patterns in Fig. 2.

3.2 Morphologies observation of alloys after corrosion

The cross-sectional morphologies of the alloys after corrosion in molten glass at 1050 °C for 5 h and 100 h are shown in Fig. 3. At the early stage of immersion (5 h), the corrosion mainly occurs along grain boundaries for A30 alloy (Fig. 3a). This is related to the diffusion and dissolution of Cr along grain boundaries, which is in accordance with the phenomenon of Ni-Cr alloy reported in previous investigations [3, 7]. In contrast, Al-containing alloys exhibit a different corrosion behaviour when Al content varies. As to the alloys with relatively low Al contents of 2Al and 5Al, molten glass penetrates into alloys uniformly. The surface corrosion layer for 2Al and 5Al seems thinner than that of A30. When the Al content in alloys is more than 8 wt. %, a discrete Al_2O_3 layer with a thickness of $<1\text{ }\mu\text{m}$ forms in the corrosion front, together with the voids caused by alloy dissolution as indicated in Fig. 3g and i. Moreover, the surface Al-depletion layer is evident by SEM observation in Fig. 3g, indicating a substantial consumption of Al.

After the long-term immersion test for 100 h, no chromium oxide forms on A30 surface, probably resulting from its instability at higher temperatures [5]. The glass still mainly attacks the grain boundaries (Fig. 3b). Similar to A30 alloy, the thickness of the corrosion layer increases after 100 h-immersion for 2Al and 5Al alloys (Fig. 3d and f). However, A30 evidently suffers from more severe attack than 2Al alloy according to the thickness of corrosion layer. When the Al content in the alloy is more than 8 wt. %, the inner Al_2O_3 layer becomes continuous. For 10Al alloy, a non-continuous alloy scale, less than $10\text{ }\mu\text{m}$ -thick, has peeled off from the base alloy (Fig. 3j). The glass penetrates further deep into the alloy below Al_2O_3 layer. Therefore, such an alumina layer is non-protective in molten glass. In addition, the Al-rich phase in the surface layer within a thickness of $\sim 83\text{ }\mu\text{m}$ has almost disappeared.

The corrosion depth for alloys after immersion for 100 h was measured based on SEM images taken from 6 different areas for each sample. The average values as a function of Al content are displayed in Fig. 4, where the error bar stands for the standard deviation of data extracted from different areas. The value for Al-free A30 is $\sim 75.3\text{ }\mu\text{m}$, while adding 2 wt.% Al decreases the value to $\sim 52.5\text{ }\mu\text{m}$. For other

Al-containing alloys, the corrosion depth increases with the further increase in Al content from ~70.8 μm for 5Al alloy, to ~80.8 μm for 8Al, and then to ~88.3 μm for 10Al alloy. Therefore, minor Al (< 2 wt. %) is favourable to reduce the corrosion rate in molten glass, whereas an excessive Al content (>5 wt. %) has an opposite effect.

The macro photos of alloys after corrosion for 100 h are shown in Fig. 5. The morphologies of alloys with different compositions obviously fall into two typical categories that, with and without surface spallation. Alloys of A30, 2Al and 5Al exhibit no spalling or products except for local residual glass, and the alloy surface exhibits metal lustre (Fig. 5a, b and c). However, visible delamination occurs to 8Al and 10Al as seen in Fig. 5d and e. The detached metal layer has been found on the glass side immediately in contact with the alloy. The colour of the glass close to the alloy surface changes into dark blue-purple in all cases due to the dissolution of metal ions, and it is hard to make a clear distinction from the colour change based on the macro observation. The above results are consistent with the micro morphologies in Fig. 3. Therefore, the alloy failure mode alters with the increase in Al content.

3.3 Characterization of phase structure and element distribution

Figure 6 shows the XRD patterns of the alloys after exposure in molten glass for 100 h. A30, 2Al, and 5Al possess the same phase structure of single fcc γ -(Ni, Cr) after corrosion, indicating that the corrosion products are all soluble in melts, which is in agreement with the microstructure results in Fig. 3. The NiAl phase initially present in cast 8Al and 10Al alloys disappears after corrosion, and a small amount of NaAlSiO₄ is detected. This NaAlSiO₄ (nepheline) is a reaction product between Al₂O₃ and molten silicates [19, 20], which is preserved in residual glass on the sample surface. The disappearance of NiAl diffraction peaks further confirms the Al impoverishment on surface as shown in Fig. 3h and j.

The elemental distribution of alloys after corrosion for 100 h determined by EPMA is presented in Figures 7 to 11. In Fig. 7, a Cr-depletion layer with a thickness of ~80 μm exists just below the surface layer of A30 (Fig. 7b). The glass is found to fill grain boundaries within the corrosion layer (Fig. 7e and f). W element acted as a

strengthening constituent is not significantly depleted as seen in Fig. 7g. Therefore, tungsten is expected to be more stable than Cr in melts. In addition, Si enriched on surface (Fig. 7e) demonstrates that the Si^{4+} in the glass is reduced to Si^0 and diffuses into the alloy, which has been reported in previous literature [8, 18]. Taking account that other alloying elements except Cr are not depleted in the surface layer, selective corrosion of Cr is the main cause of alloy failure in the case of A30.

Figure 8 presents the EPMA results for 2Al alloy. Al is observed merely in the glass within corrosion voids (Fig. 8b). Similar to A30, the Cr-depletion layer appears again, however, the colour change in Cr-depleted layer becomes much smaller compared to the results in Fig. 7b, indicating that the corrosion of Cr is inhibited to some extent. Moreover, Si is found to be enriched in the surface layer. When adding 2 wt. % Al into A30, the corrosion of Al and Cr proceeds at the same time, and Si is generated as the reduction product.

The element distribution for 5Al (Fig. 9) exhibits the same feature as 2Al. Both Al and Cr react with molten glass and yield the depletion layer on the surface. Since the Al content increases to 5 wt. %, the boundary between the Al-rich substrate and the Al-depletion layer can be clearly distinguished. Enrichment of Fe and Ni near the surface layer is observed and this is probably related to the relatively high stability of Fe and Ni in molten glass [5, 10, 21-23]. The corrosion rates of pure metals of Fe and Ni have been reported to be much lower than that of Al and Cr [5]. From this point of view, increasing the Fe content in an appropriate range would be helpful to improve the corrosion resistance.

Mapping results of 8Al alloy in Fig. 10 shows that an inner layer of Al_2O_3 develops beneath surface alloy, which is not present for 2Al and 5Al alloys. The Al-rich phase with dark grey contrast in the substrate has been proved to be concentrated with Ni and Al. At the same time, the dissolution of Cr is significantly retarded since the Cr depletion becomes less obvious on surface for this alloy (Fig. 10c). Therefore, the selective dissolution of Al is the predominant oxidation reaction in 8Al alloy.

Figure 11 shows the EPMA results of 10Al alloy, the surface layer of alloy has peeled off from the substrate in some region, which is ascribed to the extremely high

solubility of alumina in molten glass [18, 24-26]. According to the results by Manfreido [24], the alumina solubility is more than 20 times higher than that for chromia. Thus, Al_2O_3 dissolves fast when in contact with melts and results in the exfoliation of the top layer along alumina interface. Meanwhile, some Al_2O_3 particles forms beneath the cracking interface which are indicated in Fig. 11b. The formation of alumina cannot prevent the inward penetration of melts, and corrosion voids are found to extend into the alloy. Moreover, the concentrated Al present in the glass near alloy surface (Fig. 11b) further proves the migration of Al from alloy to glass. By contrast, there is no evidence of Cr-migration in the same area (Fig. 11c), instead, the enrichment of Cr on surface reveals a suppressed corrosion of Cr. Thus, the dissolution of Al dominates the corrosion process of 10Al alloy.

It is notable that the surface layer of 8Al and 10Al seems to be depleted in Ni (Fig. 10e and 11e). Actually, the Ni concentration in the glass near the alloy surface measured by EPMA is at extremely low level of 0.15-0.20 wt. %, suggesting the dissolution of Ni could be negligible. According to the EDS results of 10Al alloy, the Ni content on surface layer is ~52 at. %, close to the original composition of cast alloy, however, much lower than 65 at. % in NiAl phase. Therefore, the appearance of “Ni-depletion” on high-Al alloys is caused by the contrast of Ni relative-concentration between the surface layer and the NiAl phase in far field. This explanation could also been proved by the intensity counts in EPMA results, because the values on surface layer read from the temperature scale in Ni profiles are at the same levels around 3100 to 3200 for all Al-containing alloys.

In order to quantitatively evaluate the dissolution of Al and Cr in molten glass, the concentration of active elements across the alloy/glass interface was measured by EPMA, and the concentration profiles are shown in Fig. 12, where Al concentrations in glass obtained after corrosion (Fig. 12b, c, d and e) have been modified by subtracting the original Al concentration in glass of 1.0 wt.%. The Cr content decreases significantly in surface alloy with a thickness of ~100 μm for A30 (Fig. 12a), and this is caused by the fast dissolution of Cr element. Approximately 3.25 wt. % Cr is detected in the glass near the interface. The composition with the Cr content of

~70 wt. % in the profile corresponds to the **Cr-rich** precipitates, which is a common phenomenon in **high Cr** alloys [27, 28]. The concentration profile of Si in alloy proves the reduction of SiO₂ in the glass network as shown in Eq. 1 [18].

The concentration profiles of Cr and Si in 2Al alloy exhibit the same tendency as A30 (Fig. 12b). Besides the enrichment of Si and the Cr depletion, the Al-depletion zone from the surface to the depth of ~ 60 µm is clearly observed. Approximately **2.24** wt. % Al and ~1.53 wt. % Cr were detected in the glass adjacent to the interface for 2Al, revealing Al and Cr simultaneously **corrode** even though the Al content is much smaller than Cr in the alloy.

Comparing the results in Fig. 12c, d, and e, it is clear the Cr concentration in the glass close to the interface **decreases** from ~ 1.26 wt. % in 5Al alloy to ~ 0.87 wt.% in 8Al alloy and further to ~0.71 wt. % in 10Al alloy. The Al content in glass, by contrast, **increases** from ~ 2.55 wt. % in 5Al alloy to ~ 2.82 wt. % in 8Al alloy and **reaches** ~3.09 wt. % in 10Al alloy. The results for alloys with high Al content are similar to the investigations on Ni-8Al-28Cr by Abdullah et al. [18]. **The contrast between the concentration change** of Al and Cr **implies** that the corrosion of Al and Cr **proceeds** simultaneously and competitively. When the Al content in alloy exceeds a critical level (~8 wt. %), the **dissolution** of Al prevails.

3.4 Corrosion mechanism of alloys

In the light of the above results, the corrosion **mechanism** for alloys with different Al contents can be schematically illustrated in Fig. 13. For the alloys without **alloyed Al**, the corrosion **proceeds** via the **dissolution** of Cr on the surface (as seen in Fig. 13a). In this case, the oxidation of Cr is the main corrosion reaction. According to the studies on Cr-containing alloys [6, 8], the reaction equation can be written as Eq. 2:

The Cr ions **dissolve in melts with no protective oxide layer generated on the surface**. At the same time, silicate in the glass **is** reduced to a Si element (Eq. 1) and then **diffuses** into the alloy. Since the diffusion of Cr from the base alloy to the

interface mainly along grain boundaries, the corrosion preferentially takes place along these short routes of diffusion.

For the alloys with low Al contents (2Al and 5Al), both Al and Cr participate in the corrosion procedure as seen in Fig. 13b, where Al is oxidized as shown in Eq.3 [9]:

The Gibbs free energy of Al_2O_3 formation is much more negative than that of Cr_2O_3 , and the higher solubility of alumina in glass has also been confirmed in the previous discussion. Therefore, Al at the glass/alloy interface is more prone to corrode than Cr. However, the Al content in 2Al is at a low level and easily consumed, which makes the migration of Al from the bulk alloy to the surface become the rate control step, thus inhibiting the corrosion rate to some extent. This is probably the reason why the 2Al alloy has the lowest value of corrosion depth (Fig. 4). In this case, minor Al has a positive effect of mitigating the uniform corrosion of alloys.

For the alloys with high Al contents of 8Al and 10Al, the dissolution of Al is the domain oxidation reaction. At the early stage of corrosion, selective dissolution occurs to the Al-rich phase, leading to the formation of corrosion voids on the surface. Meanwhile, non-protective Al_2O_3 layer forms below the surface (Fig. 13c). After the immersion for 100 h, the NiAl phase disappears as a result of the fast dissolution of Al. Since the Al_2O_3 is not stable in molten glass, the surface layer of 8Al and 10Al alloy easily peels off along the Al_2O_3 interface and the glass penetrates into the alloy. In addition, another inner Al_2O_3 layer forms again. It is predictable that this subscale will peel off with a longer immersion duration since the alumina layer developed beneath will also dissolve in molten glass.

4. Conclusions

The corrosion behaviour of Ni-base alloys with various Al contents ranging from 0 to 10 wt. % in molten silicates at 1050 °C were investigated under static mode. Based on the above results and discussion, it can be concluded that:

(1). The dissolution of the Cr element is responsible for the corrosion of alloys without alloyed Al. A30 remains the single phase of $\gamma\text{-(Ni,Cr)}$ after corrosion and no

protective oxide layer forms.

(2). The dissolution of Al and Cr were both detected in alloys with low Al contents (Al< 5wt. %).

(3). An inner alumina layer forms in high Al alloys (Al>8 wt. %), leading to the exfoliation of surface alloy along the Al₂O₃ interface.

Therefore, a small amount of alloyed Al is beneficial to suppress the corrosion of Cr. When the Al content exceeds 8 wt. %, the corrosion is accelerated by surface spallation.

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Figure captions and Tables:

Fig. 1 Microstructures of cast alloys: (a) A30; (b) 2Al; (c) 5Al; (d) 8Al; (e) 10Al

Fig. 2 XRD patterns of cast alloys

Fig. 3 Cross-sectional morphologies of alloys with different Al contents after corrosion at 1050 °C for 5 h and 100 h: (a, b) A30; (c, d) 2Al; (e, f) 5Al; (g, h) 8Al; (i, j) 10Al

Fig. 4 Corrosion depth of alloys after immersion in molten glass at 1050 °C for 100 h

Fig. 5 Macro photos of alloys after corrosion in molten glass at 1050 °C for 100 h

Fig. 6 XRD patterns of alloys after corrosion in molten glass at 1050 °C for 100 h

Fig. 7 EPMA results of A30 after corrosion in molten glass at 1050 °C for 100 h

Fig. 8 EPMA results of 2Al after corrosion in molten glass at 1050 °C for 100 h

Fig. 9 EPMA results of 5Al after corrosion in molten glass at 1050 °C for 100 h

Fig. 10 EPMA results of 8Al after corrosion in molten glass at 1050 °C for 100 h

Fig. 11 EPMA results of 10Al after corrosion in molten glass at 1050 °C for 100 h

Fig. 12 EPMA concentration profiles of Al, Cr and Si in alloy and glass after corrosion in molten glass at 1050 °C for 100 h: (a) A30; (b) 2Al; (c) 5Al; (d) 8Al; (e) 10Al

Fig. 13 The schematic of corrosion mechanism of alloys with different Al contents: (a) A30 without Al; (b) low-Al alloy; (c) high-Al alloy

Table 1 The composition of alloys in mass percentage

	Cr	Fe	Al	W	C	Ni
A30	28.0	6.0	-	7.40	0.70	Bal.
2Al	27.40	5.88	2.0	7.25	0.69	Bal.
5Al	26.60	5.70	5.0	7.03	0.66	Bal.
8Al	25.76	5.52	8.0	6.81	0.64	Bal.
10Al	25.20	5.40	10.0	6.66	0.63	Bal.

Fig. 1 Microstructures of cast alloys: (a) A30; (b) 2Al; (c) 5Al; (d) 8Al; (e) 10Al

Fig. 2 XRD patterns of cast alloys

Fig. 3 Cross-sectional morphologies of alloys with different Al contents after corrosion at 1050 °C for 5 h and 100 h: (a, b) A30; (c, d) 2Al; (e, f) 5Al; (g, h) 8Al; (i, j) 10Al

Fig. 4 Corrosion depth of alloys after immersion in molten glass at 1050 °C for 100 h
(Error bar represents the standard deviation of average values)

Fig. 5 Macro photos of alloys after corrosion in molten glass at 1050 °C for 100 h

Fig. 6 XRD patterns of alloys after corrosion in molten glass at 1050 °C for 100 h

Fig. 7 EPMA results of A30 after corrosion in molten glass at 1050 °C for 100 h

Fig. 8 EPMA results of 2Al after corrosion in molten glass at 1050 °C for 100 h

Fig. 9 EPMA results of 5Al after corrosion in molten glass at 1050 °C for 100 h

Fig. 10 EPMA results of 8Al after corrosion in molten glass at 1050 °C for 100 h

Fig. 11 EPMA results of 10Al after corrosion in molten glass at 1050 °C for 100 h

Fig. 12 EPMA concentration profiles of Al, Cr and Si in alloy and glass after corrosion in molten glass at 1050 °C for 100 h: (a) A30; (b) 2Al; (c) 5Al; (d) 8Al; (e) 10Al

Fig. 13 The schematic of corrosion mechanism of alloys with different Al contents: (a) A30 without Al; (b) low-Al alloy; (c) high-Al alloy