

Article

Protection of 316L Steel Against LBE Corrosion by a CaO-MgO-Al₂O₃-SiO₂ Glass–Ceramic Coating

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Abstract: A CaO-MgO-Al₂O₃-SiO₂ glass–ceramic coating was prepared by the slurry method and subsequent sintering to improve the corrosion resistance of 316L steel in liquid lead–bismuth eutectic alloy at high temperatures. The glass–ceramic coating, sintered at 884 °C, was dense and demonstrated strong adhesion to the substrate. It was composed of the crystalline phases diopside (CaMgSi₂O₆) and anorthite (CaAl₂Si₂O₈) and had an average Vickers hardness of 595 HV, which was over three times that of 316L steel. After corrosion in an oxygen-saturated, static lead–bismuth eutectic alloy at 500 °C for 1000 h, the uncoated 316L experienced significant mass gain (0.04 g) due to severe oxidative corrosion, resulting in the formation of Fe₃O₄ and Pb₂O on its surface. In contrast, the glass–ceramic-coated specimens showed a very small mass gain (0.0012 g) after corrosion. The coating maintained good thermal stability; its crystalline phase composition remained largely unchanged after the corrosion test. The glass–ceramic coating still exhibited dense microstructure and tightly adhered to the substrate after corrosion. There was no evident penetration of lead–bismuth into the coating, and no dissolution of the coating’s elements into the lead–bismuth alloy was detected. These observations confirm that the glass–ceramic coating possessed superior corrosion resistance in liquid lead–bismuth eutectic environments.

Keywords: glass–ceramic coating; LBE corrosion; protection; 316L steel



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1. Introduction

Lead-cooled fast reactors (LFRs) have become one of the most promising reactors for Generation IV nuclear systems, renowned for their favorable neutronics, thermal-hydraulics, and inherent safety features [1]. However, despite these benefits, material challenges continue to pose a significant barrier to the progress of LFR technology [2]. Unlike the structural materials used in the commercially established Generation II and III nuclear systems, the materials used in LFRs face a more rigorous service environment [3]. Notably, corrosion caused by liquid lead, or the lead–bismuth eutectic (LBE) alloy, is a critical factor in the deterioration of structural materials within LFRs over extended periods of service [4–9].

So far, there are three types of reference reactors in the world: SSTAR (Small, Secure Transportable Autonomous Reactor, 20 MWe) in the United States [10], ELFR (European Lead-cooled Fast Reactor, 300 MWe) in the European Union [11], and BREST-OD-300 (Bystriy Reaktor Estestrennoy Bezopasnosti, 600 MWe) in Russia [1]. Furthermore, other LFRs have been conceptualized, including the MYRRHA (Multi-purpose hybrid research reactor for high-tech applications) in the European Union [1] and CLEAR (China LEAD-based Reactor) in China [12]. Candidate structural materials for these reactors encompass a range of options such as ferritic/martensitic steels (e.g., T91), austenitic stainless steels (e.g., 15-15Ti, 316L), ODS steels, FeCrAl alloys, and refractory metals [13–17]. Notably, 316L steel or its equivalents are considered for cladding in MYRRHA and as a potential material

for pump impellers in ELFR [11]. However, studies have indicated that 316L steel in LBE was subject to more severe oxidative corrosion at high oxygen concentrations or dissolution corrosion at low oxygen concentrations between 480 and 550 °C, which are the typical design temperatures of LFRs [18,19]. These corrosion issues could significantly compromise the service life of 316L as a structural material in LFR applications. Consequently, there is a pressing need to develop corrosion-resistant coatings for 316L to protect it against lead or LBE corrosion.

Previous studies have developed a variety of coating materials for steels against lead or LBE corrosion, such as ceramic coatings (e.g., Al_2O_3 , SiC, $\text{Al}_2\text{O}_3\text{-TiO}_2$) [5,20], alloy coatings containing aluminum (e.g., FeCrAl, FeCrAlY) [14], high-entropy alloy coatings [21], and ternary MAX cermet coatings [22]. In addition to the above-mentioned types of coatings, glass and glass–ceramic materials, known for their excellent thermal stability, wear resistance, corrosion resistance, and oxidation resistance, are emerging as promising coating materials for steels against high-temperature oxidation and LBE corrosion [23,24]. For instance, Liu et al. developed a glass and ceramic composite coating on 304 stainless steel, which consisted of a $\text{SiO}_2\text{-Al}_2\text{O}_3$ glass matrix reinforced with Cr_2O_3 particles, demonstrating good corrosion resistance in LBE [23]. However, compared to composite coatings made by simply mixing a glass matrix with ceramic particles, glass–ceramic coatings that contain a significant amount of microcrystalline and glassy phases produced by controlled crystallization of a specific base glass during heating or sintering exhibit superior high-temperature stability, mechanical properties, and corrosion resistance [25–28]. Despite these advantages, there have been relatively few studies on the preparation of glass–ceramic coatings for structural materials considered for use in LFRs.

In this study, a CaO-MgO- $\text{Al}_2\text{O}_3\text{-SiO}_2$ (CMAS) glass–ceramic coating was prepared on 316L steel by the slurry method and subsequent sintering. Additionally, the corrosion behavior of the 316L steel and the glass–ceramic coatings in LBE at 500 °C for 1000 h was also studied. The findings of this study not only broaden the potential applications of glass–ceramic coatings but also provide significant guidance for the development of coating materials suitable for advanced nuclear energy systems.

2. Experimental

2.1. Specimen Preparation

A 316L stainless steel sheet was machined into 10 mm × 10 mm × 3 mm (length × width × height) specimens by the wire electrical discharge machining method. A 1 mm diameter circular hole was drilled in the upper right corner of each specimen's surface so that the specimens could be suspended on the specimen holder for the high-temperature LBE corrosion test. Before the glass–ceramic coating was prepared, the surfaces of the 316L specimens were mechanically ground using a sequence of #600 to #1500 SiC abrasive papers. Subsequently, the specimens were cleaned by anhydrous ethanol and dried in a drying oven. The chemical compositions of 316L steel are shown in Table 1.

Table 1. Chemical compositions of 316L stainless steel before corrosion.

Element	C	Cr	Ni	Mo	Mn	Si	Fe
Content (wt%)	0.02	16.20	10.03	2.06	1.48	0.48	Bal.

2.2. Glass Powder Preparation

CMAS glass was prepared using the conventional quenching method. All the starting raw materials CaCO_3 , $\text{Mg}(\text{HCO}_3)_2$, Al_2O_3 , SiO_2 , and K_2O were of reagent grade and were purchased from Sinopharm Chemical Reagent Co., Ltd. in Shanghai, China. After thoroughly mixing, the batches were placed in an alundum crucible and then heated to 1600 °C in the electrical furnace. It was held at this temperature for 2 h, and then the molten glass was poured into water to produce glass particles. Then, the water-quenched glass particles, the grinding balls, and the ethanol were put into the tank for ball milling. In the

process of ball milling, the mass ratio of glass particle to ball was 1:3, and the mass of glass particle was the same as that of anhydrous ethanol. After milling for 36 h, the glass powder could be obtained by drying and sieving (200 mesh). The chemical compositions of the CMAS glass are listed in Table 2, which shows experimental starting compositions.

Table 2. Chemical compositions of CMAS glass.

Oxide	CaO	MgO	Al ₂ O ₃	SiO ₂	K ₂ O
Content (wt%)	27.3	9.2	13.3	49.7	0.5

2.3. Coating Preparation

The CMAS glass–ceramic coating was prepared using the slurry method followed by a sintering process. Initially, CMAS glass powder, a binder (methyl cellulose solution), and distilled water were mixed in a mass ratio of 0.75:0.75:1 to form a slurry. This slurry was then air-sprayed onto the surfaces of spare 316L specimens and subsequently dried in an oven at 80 °C for 24 h. The dried specimens were then subjected to a three-stage high-temperature heating process in a muffle furnace. This process included heating the samples to 500 °C and maintaining them for 1 h, followed by 1 h at 735 °C, and finally 1 h at 884 °C for sintering. This meticulous sintering regimen ensured the coating’s optimal density and adherence to the substrate.

2.4. LBE Corrosion Tests

The study of material corrosion behavior in liquids mainly includes two methods. The first method is the immersion method, which involves directly immersing the samples in the corrosive medium. By studying the mass changes in samples before and after corrosion, as well as the morphology and phase of the corrosion products, the material corrosion behavior and mechanism can be determined. This paper adopts this method, which is also a commonly used method for studying the corrosion behavior of materials in liquid metal. The second method for material corrosion research is the electrochemical method. However, this method requires the presence of an electrolyte in the corrosive medium (such as water, water containing mixed liquids, molten salts, etc.). However, there are no electrolytes in liquid metal, and materials do not undergo electrochemical corrosion in liquid metals.

The liquid LBE alloy served as the corrosive medium in our corrosion tests, with a composition of 44.5 wt% lead (Pb) and 55.5 wt% bismuth (Bi).

To evaluate the LBE corrosion resistance of the CMAS glass–ceramic coating, the high-temperature static LBE corrosion test was carried out using the LBE corrosion test device, as shown in Figure 1. The test device, as shown in Figure 1a, comprised a stainless-steel tank, a thermocouple for temperature monitoring, and a heating and insulation system. For the corrosion tests, LBE and the specimens were introduced into the steel tank, with the specimens suspended from a stainless-steel specimen holder, as illustrated in Figure 1b, using molybdenum wire. Throughout the corrosion process, the dissolved oxygen content in the liquid LBE was not actively controlled, and it was assumed to be in a state of oxygen saturation.

The corrosion tests were conducted at a temperature of 500 °C with a total duration of 1000 h. This duration could simulate the long-term corrosion behavior of materials in practical applications (such as in lead-cooled fast reactors), as well as evaluate the durability of materials in the LBE environment. Furthermore, previous studies show that a 1000 h corrosion test duration is long enough to observe the corrosion characteristics and trends in materials in the LBE environment, thereby providing a scientific basis for material selection and application. During corrosion, specimens were extracted from the liquid LBE at intervals of 250 h, 500 h, 750 h, and 1000 h for analysis. The temperature of corrosion tests was set at 500 °C because the current design temperature for the cladding of CLEAR (China LEAd-based Reactor) is 500 °C. The glass–ceramic coating designed in this paper is also intended for potential application in CLEAR in the future.

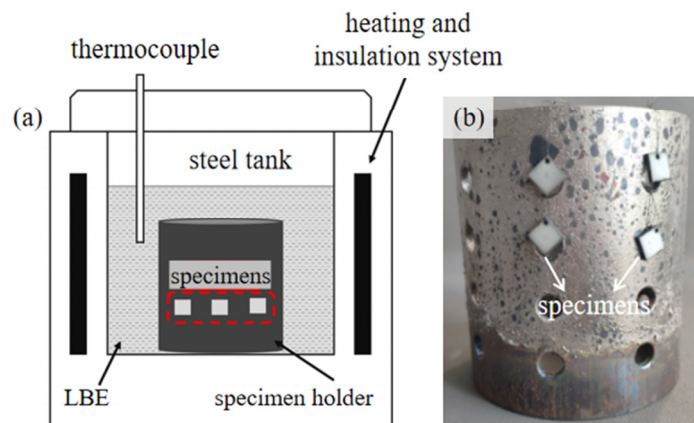


Figure 1. Schematic diagram of LBE corrosion test device (a) and physical diagram of specimen holder (b).

Post-corrosion, half of the specimens were subjected to surface and mass gain analysis. These were immersed in glycerol at 200 °C to remove any adhering LBE. The other half, designated for cross-sectional analysis, were embedded in resin and then ground and polished to prepare them for further examination.

2.5. Analytical Characterization

The thermal properties of CMAS glasses, such as the glass transition (T_g) and crystallized peak (T_c) temperatures, were determined by differential scanning calorimetry (DSC, NETZSCH in Selb, Germany) in an inert gas atmosphere (Ar) with a heating rate of 10 °C/min. The surface and cross-sectional morphology and element distribution of specimens were analyzed with a scanning electron microscope (SEM, ZEISS in Oberkochen, Germany) with an EDS spectrometer (Oxford Instruments in Oxford, United Kingdom). The phase composition of specimens was observed by an X-ray diffraction instrument (XRD, PANalytical in Malvern, United Kingdom).

3. Results

3.1. Determination of Coating Sintering Temperature

The DSC curve of CMAS glass powder revealed three distinct exothermic peaks: the glass transition temperature (T_g) at 558.09 °C and two crystallization temperatures (T_c) at 789.14 °C and 951.52 °C. To achieve the desired crystalline phases, the sintering temperature range of CMAS glass–ceramic coating was chosen to be between 789.14 °C and 951.52 °C. We referred to the sintering temperatures of glass ceramics with similar compositions. During sintering, we set the sintering temperatures at 844 °C, 864 °C, 884 °C, 904 °C, and 924 °C, respectively. Eventually, we found that only when the sintering temperature was 884 °C could we achieve a complete coating. Meanwhile, coatings obtained at other sintering temperatures peeled off. However, the reasons for the peeling are not yet clear and require further research. So, after careful consideration of the sintering process, the optimal sintering temperature was set at 884 °C to ensure the coating’s adherence without peeling.

3.2. Microstructure of the As-Prepared CMAS Glass–Ceramic Coating

Figure 2 shows the surface and cross-sectional morphology of the CMAS glass–ceramic coating prepared at 884 °C. The surface morphology (Figure 2a) shows that the coating was dense and exhibited a high level of smoothness. However, minor protrusions were observed on the surface, potentially due to the uncontrolled crystal growth during the sintering process. The cross-sectional morphology reveals strong adhesion between the coating and the substrate with no signs of cracking and peeling (Figure 2b). Notably, there was no distinct interface layer visible at the boundary between the coating and the substrate. Some pores within the coating were also observed.

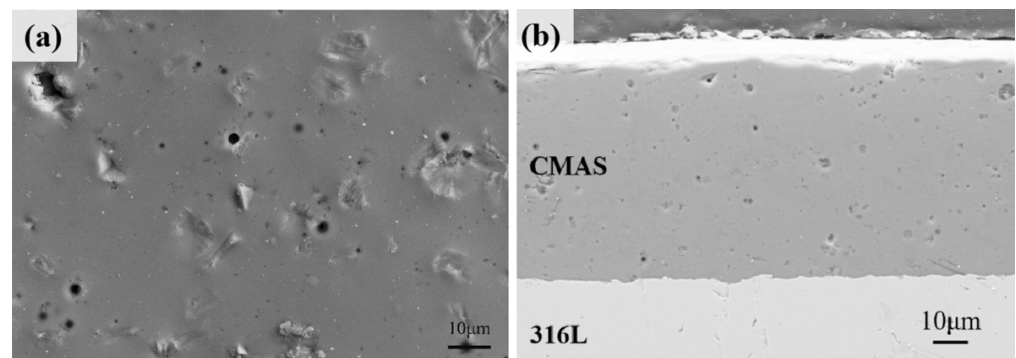


Figure 2. The surface and cross-sectional morphology of the as-prepared CMAS glass–ceramic coating: (a) surface morphology; (b) cross-sectional morphology.

The XRD pattern of the as-prepared CMAS glass–ceramic coating (Figure 3) confirmed that the main crystalline phases of the coating were diopside ($\text{CaMgSi}_2\text{O}_6$) and anorthite ($\text{CaAl}_2\text{Si}_2\text{O}_8$), which was consistent with the design objective of this study.

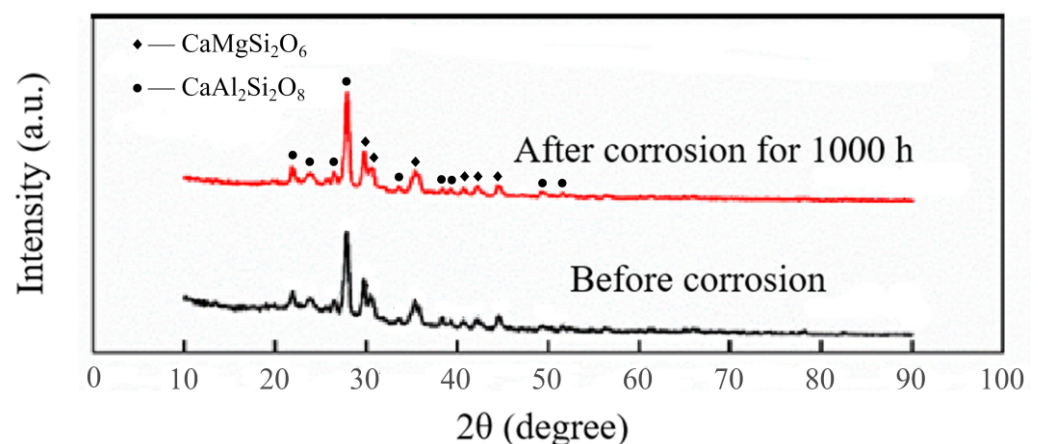


Figure 3. XRD pattern of the CMAS glass–ceramic coating before and after corrosion for 1000 h.

3.3. LBE Corrosion

3.3.1. Mass Gain of Corrosion Specimens

Figure 4 illustrates the mass gain of the 316L and CMAS glass–ceramic-coated specimens after different corrosion times. In our corrosion experiments, the corrosion mass gain at each corrosion time point was derived from the average mass gain of three parallel samples. During the corrosion experiments, we have taken strict control measures to ensure that the initial state of each sample and the corrosion environment remain consistent. Error bars were not added in the diagram of mass gain because we found that the degree of dispersion in the corrosion mass gain data at each time point was very small, within an acceptable range, and will not affect the scientific nature and reliability of this study. Figure 4 shows that the mass of 316L substrate increased with prolonged corrosion time, although the increase rate gradually slowed down. Specifically, after 1000 h of corrosion, the uncoated 316L experienced a mass gain of 0.04 g. In contrast, the CMAS glass–ceramic coating demonstrated minimal mass change throughout the corrosion periods. After 1000 h of corrosion, the coated specimens registered a very small mass gain (0.0012 g). These findings suggest that the CMAS glass–ceramic coating provided effective protection to the substrate.

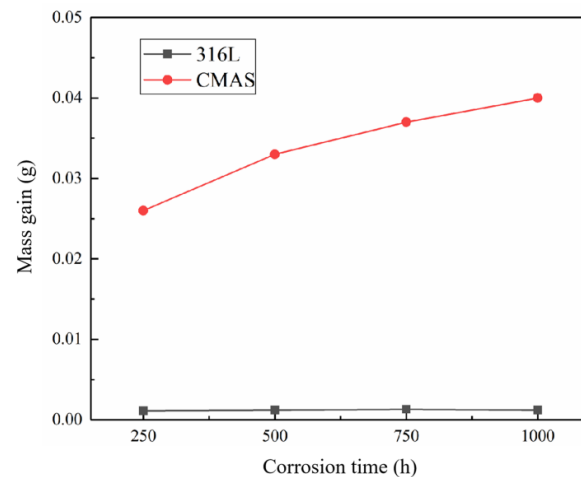


Figure 4. Mass gain of 316L and glass–ceramic coatings after different corrosion times.

3.3.2. Microstructure of Oxide Scales on Bare 316L Steel

Figure 5 presents the surface morphology of 316L steel subjected to varying durations of corrosion (ranging from 250 to 1000 h). Initially, after 250 h of corrosion, numerous corrosion pits were observed on the steel's surface, accompanied by the presence of many protruding substances. Over time, the quantity of these protrusions increased, and their distribution evolved from being scattered to forming continuous patches after 1000 h of corrosion, suggesting a progressive escalation in the severity of corrosion. The EDS scanning analysis depicted in Figure 6 reveals that the predominant elements on the corroded surface after 1000 h were Fe, Cr, Ni, Pb, and O. Notably, the protruding substances exhibited a significant enrichment in O and a marked depletion in Cr and Ni, indicating a change in the surface composition due to the corrosion process. To delve deeper into the composition of the protruding substances, point scanning analysis was conducted on specimens that had been subjected to 1000 h of corrosion. The analysis reveals that the primary constituents of these surface protrusions were Fe at 45.4 wt%, Pb at 19.6 wt%, and O at 23.7wt%, with trace amounts of Ni at 2.3 wt% and Cr at 7.9 wt%. Corroborating these findings with the XRD analysis presented in Figure 7, it was concluded that the surface protrusions on the 316L steel were predominantly composed of Fe_3O_4 and Pb_2O .

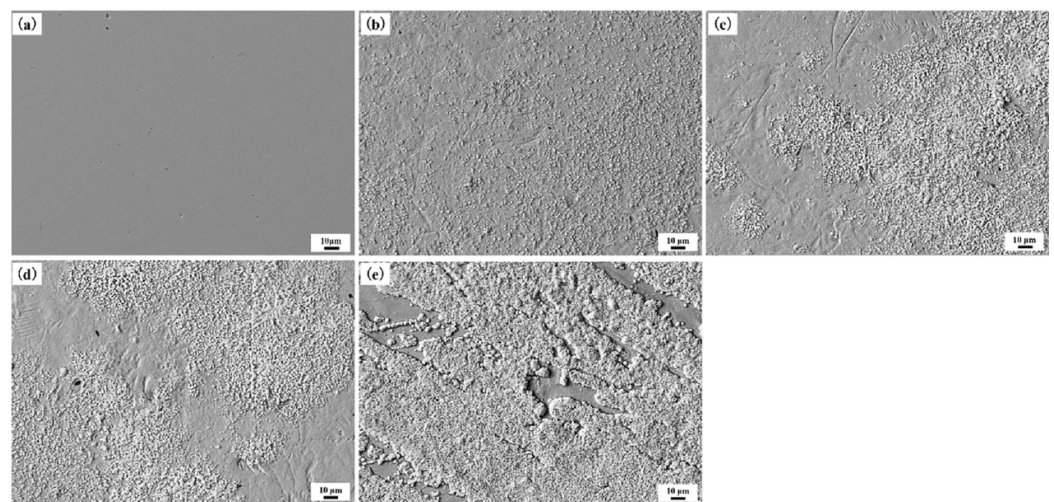


Figure 5. The surface morphology of 316L steel after different corrosion times: (a) before corrosion; (b) 250 h; (c) 500 h; (d) 750 h; (e) 1000 h.

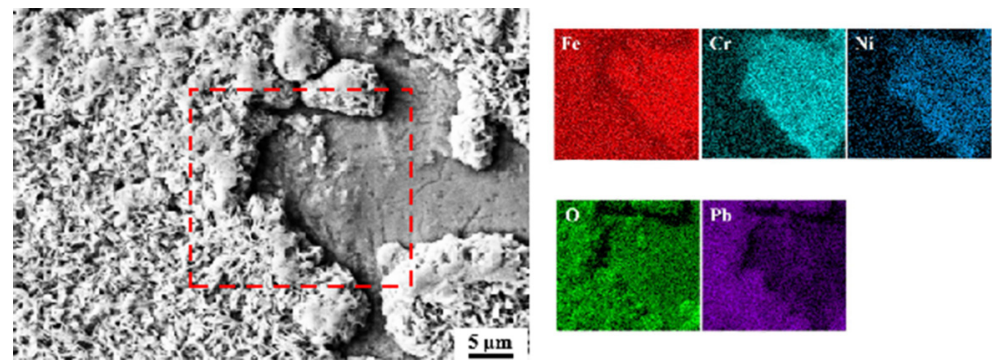


Figure 6. Localized enlarged image of surface morphology of 316L specimens after corrosion for 1000 h and corresponding EDS area scanning results.

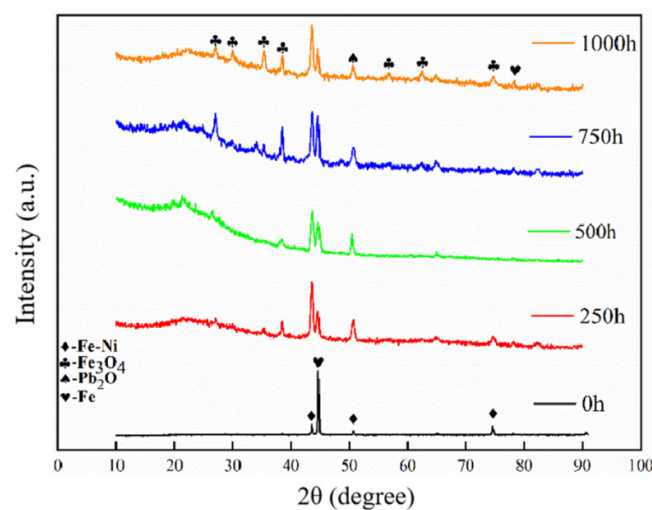


Figure 7. The XRD pattern of the 316L steel before and after corrosion for 250 h, 500 h, 750 h, and 1000 h.

The results of XRD and SEM show that the 316L substrate's surface was predominantly exposed during the early stages of corrosion, with evident oxidation corrosion occurring. As corrosion time increased, the oxide film on the substrate seemed to mitigate further material corrosion to some extent, which was reflected in the decrease in the corrosion rate corresponding to the mass gain data. This suggests that while the initial corrosion was rapid, the formation of the oxide layer helped to slow down the rate of material degradation over time.

3.3.3. Microstructure of the CMAS Glass–Ceramic Coating on 316L Steel After Corrosion

Figure 8 displays the surface and cross-sectional morphology of CMAS glass–ceramic coating after being exposed to liquid LBE for 1000 h. The surface morphology, as depicted in Figure 8a, maintained a dense microstructure post-corrosion, suggesting that the coating remained intact in the high-temperature LBE environment. A comparative examination with pre-corrosion specimens indicates an increase in both the quantity and size of the crystals within the glass–ceramic coating following the corrosion test.

The XRD patterns, as shown in Figure 3, confirm that the crystalline phases of the coating after 1000 h of corrosion were still dominated by diopside ($\text{CaMgSi}_2\text{O}_6$) and anorthite ($\text{CaAl}_2\text{Si}_2\text{O}_8$), matching the composition of the specimens prior to corrosion. This consistency demonstrates that the CMAS glass–ceramic coating possessed excellent chemical stability against LBE corrosion, retaining its structural integrity and compositional uniformity even after prolonged exposure.

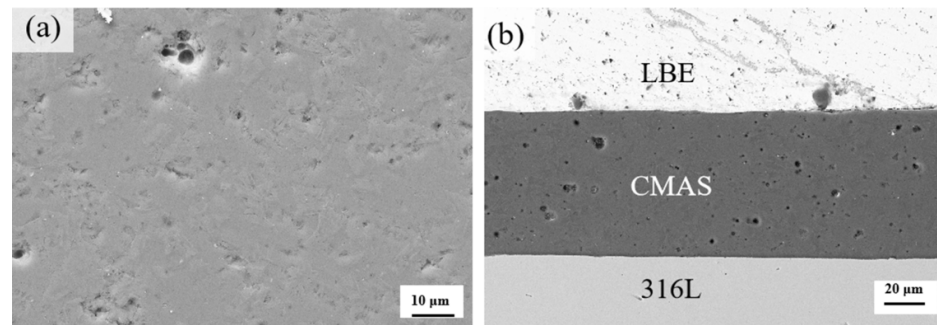


Figure 8. The surface and cross-sectional morphology of the CMAS glass–ceramic coating after corrosion for 1000 h: (a) surface morphology; (b) cross-sectional morphology.

The cross-sectional morphology of the coating presented in Figure 8b demonstrates that the coating remains firmly bonded to the substrate even after corrosion, with no signs of delamination. Additionally, there was no evident infiltration of lead and bismuth into the coating, suggesting that the coating effectively resisted penetration by these elements. The EDS line scanning analysis, as shown in Figure 9, reveals that lead (Pb) and bismuth (Bi) elements were predominantly localized on the coating's surface, with no indication of them permeating into the coating's bulk. Furthermore, there was no evidence of the coating's constituents dissolving into the lead–bismuth alloy. These observations underscore the CMAS glass–ceramic coating's robust resistance to LBE corrosion, highlighting its potential as a protective barrier in such environments.

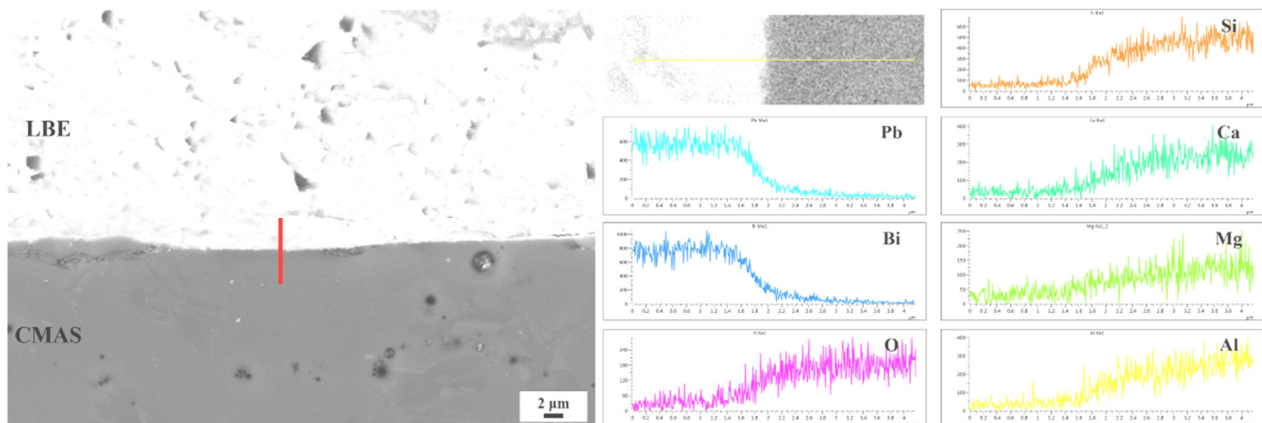


Figure 9. EDS line scan results of the CMAS glass–ceramic coating after corrosion for 1000 h.

4. Discussion

So far, the literature on the development of glass–ceramic coatings for structural materials in LFR is relatively sparse. This paper makes a commendable contribution to the field. The CMAS glass–ceramic coating, characterized by its high hardness, superior thermal stability, and outstanding resistance to LBE, stands out as an ideal candidate for applications like cladding or pump impeller coatings in LFR settings. Its unique combination of properties positions it as a promising material for enhancing the performance and durability of components in such demanding environments.

The CMAS glass–ceramic coating we developed exhibited density and a high level of surface smoothness, which are highly beneficial for enhancing the material's corrosion resistance. Nonetheless, some pores within the coating were observed. These pores are likely attributable to two main factors: first, the pores might come from bubbles within the coating slurry, and second, the bubbles produced in the process of sintering densification of the coating could not be discharged in time, which can also result in pore formation [29,30]. To optimize the density of the glass–ceramic coating and reduce porosity to a minimum,

the coating samples must be subjected to a two-stage heat treatment in a muffle furnace before the sintering temperature (884 °C). This involves holding the samples at 500 °C for one hour, followed by a further hour at 735 °C. This procedure was designed to ensure the coating achieves its maximum density and minimizes the presence of pores, thereby enhancing its overall performance and durability.

Previous research yielded a variety of ceramic coating materials for steel protection against lead or LBE corrosion, including coatings of Al_2O_3 , SiC, and $\text{Al}_2\text{O}_3\text{-TiO}_2$ [5,20]. These ceramic coatings have demonstrated effective corrosion resistance. Traditional methods for preparing these coatings include thermal spraying, physical vapor deposition, or chemical vapor deposition. In contrast to these methods, our approach to preparing glass–ceramic coatings is relatively simple. It involves air spraying or brush coating followed by high-temperature sintering processes. More importantly, the preparation process of glass–ceramics is very suitable for the preparation of surface coatings on complex components such as pump impellers, which provides a great opportunity for the widespread application of glass–ceramic coatings in nuclear reactors.

The CMAS glass–ceramic coating we developed is primarily composed of diopside ($\text{CaMgSi}_2\text{O}_6$) and anorthite ($\text{CaAl}_2\text{Si}_2\text{O}_8$) crystal phases, which are consistent with the design objective of this study. The literature reviews indicate that CMAS glass–ceramic coatings can crystallize into various phases, such as forsterite, diopside, anhydrite, and anorthite. The diopside phase is particularly noted for its excellent wear and corrosion resistance, whereas the anorthite phase is known for its superior corrosion resistance and thermal stability. Research has indicated that a well-balanced ratio of diopside to anorthite can significantly enhance the corrosion and wear resistance of CMAS glass–ceramic coatings [31]. Given that coating materials in lead-fast reactors (LFRs), especially those used for cladding and pump impellers, are subjected to the erosive effects of high-velocity LBE, it is imperative that the coatings exhibit excellent high-temperature stability, resistance to LBE corrosion, and robust wear resistance. Our CMAS glass–ceramic coating, with its specific crystalline composition, is well-suited to meet these demanding requirements.

The formulation of glass–ceramic components is pivotal in creating coatings with the desired crystalline phases. To achieve this, we initially consulted the phase diagram of the $\text{CaO-Al}_2\text{O}_3\text{-SiO}_2$ system with an addition of 10% MgO to establish a preliminary composition range for the glass–ceramic coating [32]. As shown in Table 2, the CMAS glass–ceramic coating was characterized by a high SiO_2 content, which typically serves as the network-forming component in glass–ceramic coatings. The precipitation of anorthite phase ($\text{CaAl}_2\text{Si}_2\text{O}_8$) requires a certain amount of Al_2O_3 . The proportions of CaO and MgO in the CMAS glass–ceramic coatings can be fine-tuned to lower the crystallization temperature, encourage the precipitation of diopside and anorthite phases, and enhance the surface hardness and corrosion resistance of the glass–ceramic coatings. Studies have indicated that maintaining a mass ratio of CaO to MgO close to 3 in CMAS glass–ceramic coatings broadens the crystallization temperature range, thereby improving the crystallization capability and promoting the coexistence of anorthite and diopside phases [33]. Furthermore, to address the limited crystallization propensity of CMAS glass–ceramic coatings, a minor amount of nucleating agent K_2O was incorporated into the composition. This addition aims to reduce the crystallization temperature and facilitate densification post-crystallization, thereby optimizing the overall properties of the glass–ceramic coating.

The corrosion mechanism of coatings in lead–bismuth eutectic (LBE) can be attributed to three primary processes: the dissolution of the coating material by LBE, the oxidation of the coating material by dissolved oxygen in LBE, and the corrosion of the substrate steel material by dissolved oxygen in LBE that permeates through the coating. For metallic coatings like Al-based coating (FeCrAlY , AlTiN) [14,34] and Al-containing high-entropy alloy coatings (FeCrAlTiNb , AlCrMoNbZr) [21,35], when the dissolved oxygen concentration in LBE is controlled within a suitable range, it can diffuse into the coating and react with it to form a protective layer of Al_2O_3 , which mitigates the coating's corrosion. Conversely, when the dissolved oxygen concentration in LBE is minimal or too high, the coating ma-

terial is more susceptible to dissolution or excessive oxidation by LBE, leading to coating corrosion. For our CMAS glass–ceramic coatings, neither dissolution nor oxidation occurs. Our previous research has demonstrated that even at 800 °C, the diffusion rate of oxygen in glass–ceramic coatings was remarkably low [27]. Given that the experimental temperature in this study was 500 °C, the diffusion rate of oxygen through the coating was even lower, making it difficult for dissolved oxygen in LBE to penetrate the glass–ceramic coating and oxidize the 316L substrate. Therefore, the CMAS glass–ceramic coating exhibited good resistance to LBE corrosion.

Looking ahead, our research will delve deeper into the coating’s irradiation resistance. This critical assessment will provide further insights into the coating’s potential and overall value for use in the demanding conditions of LFR.

5. Conclusions

In this research, we successfully applied a CMAS glass–ceramic coating onto 316L steel to protect it against corrosion by liquid lead–bismuth eutectic (LBE). The as-prepared coating predominantly constituted of diopside ($\text{CaMgSi}_2\text{O}_6$) and anorthite ($\text{CaAl}_2\text{Si}_2\text{O}_8$) crystal phases. We investigated the corrosion resistance of both the glass–ceramic coatings and the 316L steel when exposed to static LBE at 500 °C for a duration of up to 1000 h. Post-corrosion analysis revealed that the uncoated 316L experienced significant mass gain (0.04 g) due to severe oxidative corrosion with the formation of Fe_3O_4 and Pb_2O on its surface. In contrast, the glass–ceramic coated specimens showed a very small mass gain (0.0012 g) after corrosion. Furthermore, the phase composition of the CMAS glass–ceramic coating remained largely unaffected after the corrosion test, which is a testament to the coating’s excellent high-temperature stability. There was no significant evidence of lead–bismuth penetration through the coating, and no dissolution of the coating’s elements into the LBE was detected. These findings suggest that the CMAS glass–ceramic coating provides outstanding resistance to LBE corrosion.

Author Contributions: Conceptualization, Z.J.; Investigation, L.L.; Methodology, Y.W.; Validation, J.L.; Writing—original draft, Z.X. and B.R.; Writing—review and editing, A.L. and B.Z. All authors have read and agreed to the published version of this manuscript.

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