Corrosion of 316 Stainless Steel in High Temperature Molten Li_2BeF_4 (FLiBe) Salt

Guiqiu Zheng*, Brian Kelleher, Guoping Cao, Mark Anderson, Todd Allen, Kumar Sridharan

Department of Engineering Physics, University of Wisconsin-Madison

1500 Engineering Drive, Engineering Research Building 909, Madison, WI, 53706

*Corresponding author, guiqiuzheng@gmail.com

Abstract

In support of structural material development for the fluoride salt-cooled high temperature reactor (FHR), corrosion tests of 316 stainless steel were performed in the potential primary coolant, molten Li_2BeF_4 (FLiBe) at 700°C for an exposure duration up to 3000 hours. Tests were performed in both 316 stainless steel and graphite capsules. Corrosion in both capsule materials occurred by the dissolution of chromium from the stainless steel into the salt which led to the depletion of chromium predominantly along the grain boundaries of the test samples. The samples tested in graphite capsules showed a factor of two greater depth of corrosion attack as measured in terms of chromium depletion, compared to those tested in 316 stainless steel capsules. The samples tested in graphite capsules showed the formation of Cr_7C_3 particulate phases throughout the depth of the corrosion layer. Samples tested in both types of capsule materials showed the formation of MoSi_2 phase due to increased activity of Mo and Si as a result of Cr depletion, and furthermore corrosion promoted the formation of a α -ferrite phase in the near-surface regions of the 316 stainless steel. Based on the corrosion tests, the corrosion attack depth in FLiBe salt was predicted as 17.1 $\mu\text{m}/\text{year}$ and 31.2 $\mu\text{m}/\text{year}$ for 316 stainless steel tested in 316 stainless steel and in graphite capsules respectively. It is in an acceptable range compared to the Hastelloy-N corrosion in the Molten Salt Reactor Experiment (MSRE) fuel salt.

1, Introduction

The Fluoride-Salt-Cooled High-Temperature Reactor (abbreviated conventionally as FHR) is emerging as a leading reactor concept for the next generation nuclear reactors [1-3]. The FHR features molten fluoride salt as the primary coolant, which provides for a number of potential benefits such as high degree of passive safety, solubility for many fission products, atmospheric pressure operation, high thermal efficiency, and lower spent fuel per unit energy [4-6]. However, corrosion of structural materials in molten fluoride salts is an important consideration which must be understood and mitigated for the successful fruition of the FHR. The protective oxide layer relied upon for corrosion resistance in most high temperature environments has a propensity to dissolve in molten salt [7-9]. The containment alloy for the FHR should not only exhibit minimum corrosion in the molten fluoride salt coolant, but it should also exhibit high air-side oxidation resistance and creep strength at the intended high

temperatures of FHR operation. These unique needs were recognized in the earlier Aircraft Nuclear Program (ANP) and the Molten Salt Reactor Experiment (MSRE) in the United States between 1950 and 1970 which led to the successful development and operation of the molten salt-cooled reactor [10, 11]. These programs used a NaF-ZrF₄-UF₄ and LiF-BeF₂-ZrF₄-UF₄ salt where the uranium fuel was dissolved in the molten salt coolant. Extensive corrosion studies in these programs demonstrated the strong propensity of chromium in the test alloys to dissolve in the molten fluoride salts. Rigorous experimentation eventually led to the development of Hastelloy-N, a Ni-based alloy low in Cr (7 wt.%) and high in Mo (16 wt.%) that exhibited outstanding long-term corrosion resistance in long-term corrosion tests in thermal flowing loops, as well as good air-side corrosion resistance. However, Hastelloy-N is not code certified in the ASME Boiler Pressure Vessel Code under sections I, III, B31.1, and B31.3, and therefore cannot presently be used in a commercial reactor without extensive testing [12].

Unlike the MSRE, the recent FHR design proposes to use a non-fuel bearing FLiBe salt as primary coolant with TRISO fuel particles submerged in this coolant [2, 3]. Although the containment material for FHR has not been decided, 316 stainless steel is being actively considered because of its existing code certification at the intended FHR operation temperatures. Experiments associated with MSRE using flow loops have shown 316 stainless steel to have acceptable corrosion resistance in molten FLiBe at 650°C, particularly when the redox potential of the salt was controlled with beryllium additions [7]. Lastly, nuclear graphite has been selected as an in-core material for the FHR because of its high temperature strength and chemical resistance to molten fluoride salts. Recent work has shown that graphite when present in conjunction with structural alloys in molten fluoride salt, FLiNaK can adversely affect their corrosion performance [13, 14].

With the above background, the present study was aimed at investigating the corrosion behavior of 316 stainless steel in molten FLiBe salt at the designed FHR outlet temperature of 700°C under static test conditions. The tests were performed in 316 stainless steel and graphite capsules to determine the effect of the presence of graphite on corrosion.

2, Experimental

The FLiBe salt used for this work originated from Oak Ridge National Laboratory (ORNL). This ⁷Li enriched salt was produced for the MSRE secondary loop and was stored in nickel-based alloy canisters after the program's termination. Since corrosion in molten fluoride salts is impurity driven, the salts were subjected to re-purification by the hydrofluorination process prior to this research. This insured a high quality salt after fifty years of storage. Five hundred grams of the as-received FLiBe were transferred in its molten state into a small nickel vessel where it was bubbled with hydrogen-fluoride and hydrogen gas with a ratio of 1:10 HF/H₂, flowing at 1.5 L/min, at approximately 550°C, for two hours. The hydrofluorination process time though short was deemed acceptable as the salt was being regenerated rather than being manufactured from raw materials. The hydrofluorination process was then followed with a 12 hour sparge with hydrogen. A 38.1 mm diameter cylinder of nuclear graphite was cut

to a 152.4 mm in height to serve as the corrosion crucible. The crucible was partitioned with six, 10.3 mm diameter, cylindrical capsules. After machining, the crucible was cleaned and baked for 15 hours at 850°C in 10% hydrogen environment to remove any trace moisture or graphite powder, as was done during the MSRE program [15]. A 316 stainless steel welded liner was inserted into three of the capsules to prevent FLiBe salt contact with graphite during the corrosion test, thus effectively creating three stainless steel capsules for the tests, shown in Figure 1. The thickness of the liner and the liner-graphite gap is 0.88 mm and 0.38 mm respectively. The purified salt was transferred into the salt dripper's reservoir through an open ended transfer line with a 40 μ m stainless steel filter on the end. More purification procedures and details about the salt filling are described in our earlier paper [16].

Figure 1: Schematic of (a) the dripping system for precisely filling molten FLiBe salt and (b) a corrosion crucible including six corrosion capsules with and without 316 stainless steel liner.

Commercial purity, 316L stainless steel for corrosion testing was procured in sheet form from a commercial vendor in the U.S. and Table 1 lists the nominal chemical composition of this material. 7 mm $\times$ 13 mm $\times$ 1 mm test samples were cut from the sheet stock. The sample surfaces were mechanically polished to a 1200-grit final finish with silicon-carbide paper, and cleaned with deionized water and acetone. Two test samples were suspended in each capsule using 316 stainless steel wires. Following this, FLiBe was drip loaded into each corrosion capsule in an atmosphere-controlled glove box and allowed to solidify (as shown in Figure 1). This salt was then re-melted from top to bottom with a specially designed, linear power gradient, cartridge heater until the salt temperature reached 550°C [16]. Subsequently, for the corrosion tests, the crucible was heated in a ceramic furnace to 700°C in

the glove box where it was maintained for exposure durations of 1000, 2000, and 3000 hours. These corrosion test times are longer than those reported in literature and were used to improve accuracy in corrosion rate measurements and conduct experiments well past the initial impurity-driven corrosion stage. The corrosion temperature of 700°C was selected to meet the intended FHR outlet temperature.

Table 1: Nominal composition of 316 stainless steel (wt.%) used in this study.

C	Cr	Cu	Mn	Mo	N	Ni	P	S	Si	Fe*
0.0225	16.825	0.379	1.5305	2.0115	0.051	10.025	0.031	0.0016	0.309	68.8134

* as balance

Upon the termination of the corrosion tests, the samples were cleaned in deionized water for 24 hours and air dried at room temperature. Their mass and dimensions were measured and compared to the pre-corrosion measurements to calculate weight change (mg/cm^2) as an initial metric of the extent of corrosion. The microstructure of the corrosion layer was characterized using x-ray diffraction (XRD), and scanning electron microscopy (SEM) equipped with energy dispersive spectroscopy (EDS). The corrosion surface of one 3000 hour test sample was ground in small increments to analyze the phase transitions in the near surface layers by XRD. Additionally, all samples were mounted and polished to observe their cross-sectional microstructure from the corrosion surface to inner substrate using SEM-EDS. The affected depth in the alloy was evaluated by measuring the maximum Cr depletion depth using EDS analysis.

3, Results and Discussion

The results of weight change measurements after corrosion tests are shown in Figure 2. In all cases the samples experienced a loss in weight due to corrosion indicating a material dissolution mechanism, and this weight loss increased with exposure time. Up to 2000 hours, the weight loss was higher for the samples tested in graphite capsules as compared to the samples tested in stainless steel capsules, while at 3000 hours, the weight loss for the two capsule material tests were approximately identical. Herein, it should be noted that the net weight losses for the samples tested in graphite capsules should be higher as chromium carbide particle formation in the alloy matrix almost certainly resulted in weight gain after corrosion tests. This will be discussed later through microstructure analyses.

As the depletion of Cr becomes pronounced in the near-surface, the chromium inside the sample diffuses outwards due to a concentration gradient. The rate of Cr depletion therefore depends on its thermal diffusion coefficient in 316 stainless steel. As Cr is depleted deeper within the sample, grain boundaries rather than grains, become the preferred path of Cr migration because of their higher diffusion coefficient [17, 18]. Assuming a constant diffusion source, the weight change (ΔM) due to Cr depletion as a function of time can be expressed as

$$\Delta M = 2AC_0\sqrt{\frac{D_{eff}t}{\pi}} \quad (1)$$

where A is the sample's surface area, C_0 is the initial concentration of Cr in alloy, D_{eff} is the effective diffusion coefficient of Cr in 316 stainless steel, and t is the corrosion time [19, 20]. According to this formula, the D_{eff} of Cr in the 316 stainless steel samples tested in 316 stainless steel capsules, without graphite, was calculated as $4.2 \times 10^{-19} \text{ m}^2/\text{s}$ at 700°C , which is in the range of the reported $D_{\text{Cr}/316\text{ss}} = 10^{-19} \sim 10^{-17} \text{ m}^2/\text{s}$ at 700°C [17, 18]. Therefore, the weight loss of 316 stainless steel in FLiBe occurred as proportionally to the square root of corrosion time, shown in **Figure 2**, indicating thermal diffusion-controlled corrosion [20].

Figure 2: Weight change measurements for all twelve 316 stainless steel samples tested at 700°C in molten FLiBe salt in 316 stainless steel and graphite corrosion capsules (two samples in each) as a function of square root of corrosion time.

The surface images of various samples after the corrosion tests are shown in Figure 3. All samples clearly showed evidence of corrosion attack particularly along the grain boundaries, although there is evidence of attack on the grains as well. The degree of attack is seen to increase with exposure time. The samples tested in the graphite capsules had wider grain boundary attack than those tested in 316 stainless steel capsules. These initial observations indicate that graphite accelerates the corrosion attack to 316 stainless steel and is to some extent

consistent with the trends in weight loss measurements shown in Figure 2. Additionally, pitting is observed on the grains of all samples and there is evidence of round particulate phases adhered to the surface.

Figure 3: SEM images of the surface of the 316 stainless steel samples after corrosion testing under different conditions. a, b, and c were in 316 stainless steel capsules and d, e, and f tested in graphite capsules performed for 1000, 2000, and 3000 hours respectively.

Figure 4 shows SEM images of the surface of the samples exposed for 3000 hours in 316 stainless steel and graphite capsule tests, along with elemental EDS x-ray mapping of the surface. A large number of Mo-rich precipitate particles were detected on the corrosion surface regardless of capsule material. The formation of these precipitates appears to be a consequence of Cr depletion at the surface which increases the activity of Mo and Si in the near-surface regions and promotes the precipitation of MoSi_2 phase. Molybdenum is known to be quite inert to

molten fluoride salts, however pure silicon is prone to attack in this environment [9]. The formation of MoSi_2 precipitates prevents the fluoridation of Si.

Figure 4: EDS maps showing Mo-rich precipitates on the corrosion surface of 316 stainless steel samples (a) in 316 stainless steel lined capsule and (b) in graphite capsule after 3000 hours exposure to molten FLiBe salt at 700°C.

High magnification SEM images of the surface of the samples tested for 3000 hours, shown in Figure 5, provide further insights into the corrosion mechanisms. For the samples tested in a graphite capsule, as shown in Figure 5b, deep, steep cleavage was observed at the grain boundary, indicative of more severe attack than those in 316 stainless steel capsule. These samples also contained hexagonal Cr_7C_3 crystals and spherical Mo-rich particles on the corrosion surface. The Cr_7C_3 crystals were embedded in grains, protruding out from the edges, while the Mo-rich particles reside on the tip of Cr_7C_3 crystals. For the samples tested in the 316 stainless steel lined capsule, shown in Figure 5a, only the spherical Mo-rich particles were observed on the corrosion surface. Compared to Figure 5b, the grain boundary was not attacked as steeply. The difference in the surface microstructure of the two types of samples demonstrates the acceleration effect of graphite.

Figure 5: High magnification (20,000×) SEM images of the 316 stainless steel sample's corrosion surface after corrosion tests (a) in a 316 stainless steel capsule and (b) in a graphite capsule after 3000 hour tests. Mo denotes Mo-rich MoSi₂ phase and GB denotes grain boundary.

The phase content of the corrosion layer was further analyzed using XRD technique. Figure 6 shows the XRD patterns of the as-received samples and samples tested for 3000 hours in the two capsule materials. After corrosion, the characteristic peak intensity of austenite phase (γ phase) decreased with the (220) peak shrinking and the (111) and (200) peaks completely disappearing. The characteristic peaks of α -ferrite phase generated as indicated by the appearance of (200), (211) and (110) peaks. Main peak (110) of both α -ferrite phase and α' -martensite phase overlapped on XRD pattern. The α' -martensite phase in the as-received sample might have been introduced during the cold-working of the alloy during sheet fabrication [21, 22]; the formation of α -ferrite phase is attributable to the corrosion environment. In order to further confirm that α -ferrite phase formation after corrosion tests was not a thermal effect, and to ascertain the depth of the ferrite transformation, the surface of the samples from the corrosion test was ground slightly and re-analyzed using XRD. As shown in Figure 7, XRD patterns of the incrementally ground 3000 hour corrosion test surfaces shows the gradual resurgence of the γ phase. After removing 45-50 μm from the surface, the XRD pattern appears identical to that of the as-received sample. The α -ferrite phase was dominant for only 5-10 μm into the sample which, as will be shown later, is on the same order as the corrosion attack depth.

Figure 6: XRD patterns collected from the sample surface of (a) as-received 316 stainless steel, (b) 316 stainless steel corrosion tested in a graphite capsule, and (c) 316 stainless steel sample corrosion tested in a 316 stainless steel capsule. (b) and (c): both tested for 3000 hours.

The XRD patterns (Figure 6) also successfully identified tetragonal MoSi₂ (PDF card 01-072-6181) on the corrosion test samples' surface from both types of capsules material, but Cr₇C₃ (PDF card 04-002-2449) only on the corrosion surface tested in graphite capsules [23]. XRD pattern presents MoSi₂ main peaks of (004) and (112) both at around $2\theta=46.2^\circ$ and (110) at $2\theta=39.7^\circ$. MoSi₂ peaks are relatively weak due to the relatively low concentration of these precipitates. Main characteristic peaks of hexagonal Cr₇C₃ crystal such as (421) at $2\theta=44.3^\circ$ and (420) at $2\theta=39.3^\circ$ overlapped with other peaks, and a weak peak of (202) at $2\theta=42.6^\circ$ in XRD pattern due to its relatively low percentages in sample [23].

Figure 7: XRD patterns scanned on (a) the original corrosion surface tested in 316 stainless steel capsule for 3000 hours, (b) the surface after grinding down 5 μm to 10 μm below the original surface of the sample in (a), and (c), the surface after grinding down 45 μm to 50 μm below original corrosion surface of the sample in (a). Photographs on right side show the XRD scanned surface.

The depth of the corrosion attack and the internal microstructure of the corrosion layer were examined in sample cross-sections under the SEM. Figure 8(a-c) shows the images of samples tested in 316 stainless steel capsules and Figure 8(d-f) shows the samples tested in graphite capsules, each for 1000, 2000 and 3000 hours respectively. For the samples tested in graphite capsules, the randomly distributed particulate phases in the 316 stainless steel matrix are chromium carbide particles which were previously observed protruding from the grain boundaries in the plan view images (Figure 5b). These particles formed as inwardly diffusing carbon reacted with outwardly diffusing chromium. Increase in corrosion test time resulted in an increase in depth of the region containing these particles from approximately 5 μm at 1000 hours, to deeper than 50 μm after 3000 hours. As expected, these chromium carbide particles were not observed in the samples tested in 316 stainless steel capsules due to the absence of carbon source in the molten salt. Additionally, in all of the cross-sectional SEM images, M_{23}C_6 type carbide precipitates were observed at grain boundaries, which usually formed in 316 stainless steel during high temperature aging [24-26]. The samples exposed to salt for 3000 hours in the graphite capsule were

marked by the recession of corrosion surface, particularly at the grain boundary. As shown in Figure 8f, a crack developed several microns deep into alloy along grain boundary. It corresponds to the observation of a deep cleavage at grain boundary shown in Figure 5b. This heavy attack at the grain boundary was not observed in the samples tested in 316 stainless steel capsules.

Figure 8: Cross-sectional SEM images of 316 stainless steel samples after corrosion tests; (a), (b), and (c) were tested in 316 stainless steel capsules and (d), (e) and (f) were tested in graphite capsules for 1000, 2000, and 3000 hours respectively. Please note that for samples in (c) and (f) a Cu electroplated layer was applied for edge protection.

The corrosion behavior of the 316 stainless steel samples in the graphite capsules was different from those observed for stainless steel capsules. The graphite capsule was found to liberate carbon particles in the presence of salt even though it was cleaned and baked. Although carbon is chemically inert to salt, never forming CF_4 in fluoride salts under normal conditions, it was found to travel through the salt. The existence of carbon in the salt allowed the preferential reaction of Cr with carbon to form chromium carbides. The carburization of Cr was an extra driving force, causing Cr depletion from the alloy, in addition to chromium fluoridation. Furthermore, the

faster inward diffusion rate of carbon through 316 stainless steel ($D_{C/316ss}=10^{-15}$ m²/s at 700°C) allowed reactions with the outward diffusing chromium ($D_{Cr/316ss}=10^{-20}$ m²/s at 700°C) to form chromium carbide particles within the alloy [17, 18, 27]. The reaction which occurred in the alloy was found to be

This reaction within alloy certainly increased the samples' weight. Therefore, the net weight loss of the samples tested in the graphite capsules was the value of Cr depletion caused by fluoridation minus the weight gain due to new-formed chromium carbide particles within alloy. The real weight loss due to Cr depletion for the samples in graphite capsules should be higher than the measured value shown in Figure 2.

The depth of corrosion attack was measured in terms of the maximum Cr depletion distance using EDS elemental maps for Cr as shown in Figure 9. Based on this criteria, the affected corrosion attack depth was found to be from 6.4 µm to 11.5 µm for the samples tested in 316 stainless steel capsules, and from 13.7 µm to 22.5 µm for the samples tested in graphite capsules for corrosion time from 1000 hours to 3000 hours. As expected, increasing the corrosion test time increased the affected depth; testing in graphite capsules doubled the maximum Cr depletion depth. Based on the experimental data, the long-term corrosion attack depth can be predicted from the logarithmically fitting curves as shown in Figure 10. The fits predicted an attack depth of 17.1 µm/year and 31.2 µm/year for the samples tested in 316 stainless steel capsules and graphite capsules respectively. The EDS maps also showed Cr depletion through volume diffusion in a shallow region near the corrosion surface. This can be seen in Figure 9d where volume Cr depletion accounts for a layer approximately 3 µm thick. This observation indicates that volume diffusion is dominant mechanism for the Cr depletion during initial corrosion stage due to the large projected area ratio of grain to grain boundary on corrosion surface.

Figure 9: EDS maps of Cr distribution in near surface layer of corroded 316 stainless steel samples: (a), (b), and (c) tested in 316 stainless steel capsules and (d), (e), and f tested in graphite capsules at 700°C for 1000, 2000, and 300 hours. Maximum Cr depletion distance was labeled in each EDS Cr map.

Figure 10: The depth of corrosion attack, in terms of maximum Cr depletion depth, as a function of corrosion time. Experimental data is indicated by a solid line, while the fit is indicated by the dashed line. The fit predicted 17.1 $\mu\text{m}/\text{year}$ in 316 stainless steel capsules and 31.2 $\mu\text{m}/\text{year}$ in graphite capsules.

In the MSRE literature, the void layer observed under an optical microscope was considered to be Cr depletion layer, and the attack depth was calculated based on weight loss by assuming uniform Cr depletion from tested alloy [28]. Our studies using SEM and EDS to identify the depth of corrosion attack indicates that MSRE literature have underestimated the corrosion attack depth values. Overall, comparing 316 stainless steel to the corrosion resistance of Hastelloy-N in liquid fluoride salts, reported approximately 25 $\mu\text{m}/\text{year}$ at the temperature up to 704°C, the corrosion attack depth of 316 stainless steel in FLiBe at 700°C of 17.1 $\mu\text{m}/\text{year}$ and 31.2 $\mu\text{m}/\text{year}$ is in an acceptable range [11, 20].

4, Conclusion

Corrosion of 316L stainless steel in ^7Li -enriched molten FLiBe salt at 700°C was investigated for exposure durations of up to 3000 hours. The salt was subjected to a brief hydrofluorination purification treatment prior to the corrosion tests which were performed in both 316 stainless steel and graphite capsules. Corrosion in all cases was dominated by the dissolution of the Cr in the alloy into the salt and corrosion attack occurred predominantly along the grain boundaries of the stainless steel. The corrosion tests performed in graphite capsules exhibited

about twice the attack depth compared to those tested in 316 stainless steel capsules indicating that the presence of graphite had an accelerating effect on the corrosion of stainless steel. In both cases Cr depletion at the surface promoted the formation of MoSi_2 precipitates at the surface. For the samples tested in graphite crucibles, Cr_7C_3 particulate phases was clearly observed through the depth of the corrosion layer, due to outward diffusion of Cr from the alloy and inward diffusion of C into the sample. The affected corrosion attack depth as measured from SEM-EDS analysis ranged from 6.4 μm to 11.5 μm for the samples tested in 316 stainless steel capsules, and from 13.7 μm to 22.5 μm for the samples tested in graphite capsules. Based on experimental results, the attack depth per year was predicted of 17.1 $\mu\text{m}/\text{year}$ and 31.2 $\mu\text{m}/\text{year}$ for the samples tested in 316 stainless steel capsules and graphite capsules respectively. XRD analysis showed that the corrosion layer primarily consisted of α -ferrite phase due to massive Cr depletion at high temperature.

Acknowledgements

This research is supported by the U.S. Department of Energy Nuclear Energy University Program Contract No. DE-AC07-05ID14517. We would like to acknowledge Fred Peretz at ORNL packaging and supplying us the enriched FLiBe salt, Paul Brooks at the University of Wisconsin-Madison for his design guidance and welding which allowed us to setup these experiments, and Dr. L. M. Toth for his continued input in the area of salt chemistry.

References

- [1] C. W. Forsberg, P. F. Peterson, and P. S. Pickard. Molten-Salt-Cooled Advanced High-Temperature Reactor for Production of Hydrogen and Electricity. In *Nuclear Technology, American Nuclear Society*, pages 1-25, La Grange Park, IL, 2003.
- [2] G. F. Flanagan, D. E. Holcomb, and S. M. Cetiner. FHR Generic Design Criteria, ORNL/TM-2012/226. Technical Report June, 2012.
- [3] FHR Group. White Paper of FHR Integrated Research Project Workshop 1. Technical Report, August, University of California, Berkeley, 2013.
- [4] D. F. Williams and K. T. Clarno. Evaluation of salt coolants for reactor applications. *Nuclear Technology*, 163(3) (2008) 330-343
- [5] V. Khokhlov, V. Ignatiev, and V. Afonichkin. Evaluating physical properties of molten salt reactor fluoride mixtures. *Journal of Fluorine Chemistry*, 130(1) (2009) 30-37
- [6] S. Delpech, C. Cabet, C. Slim, and G. S. Picard. Molten fluorides for nuclear applications. *Materials Today*, 13(12) (2010) 34-41
- [7] J. R. Keiser, J. H. DeVan, and D. L. Manning. The Corrosion Resistance of Type 316 Stainless Steel to Li_2BeF_4 , ORNL/TM-5782. Technical Report, Oak Ridge National Laboratory, 1977.

- [8] P. Calderoni, P. Sharpe, H. Nishimura, and T. Terai. Control of molten salt corrosion of fusion structural materials by metallic beryllium. *Journal of Nuclear Materials*, 386-388 (2009) 1102-1106
- [9] L. C. Olson, J. W. Ambrosek, K. Sridharan, M. H. Anderson, and T. R. Allen. Materials corrosion in molten LiF-NaF-KF salt. *Journal of Fluorine Chemistry*, 130(1) (2009) 67-73
- [10] D. R. Holmes. New corrosion-resistant high temperature heat exchanger materials. *Corrosion Science*, 8(8) (1968) 603-622
- [11] J. W. Koger. Evaluation of Hastelloy N alloys after nine years exposure to both a molten fluoride salt and air at temperatures from 700 to 560°C, ORNL-TM-4189. Technical report, Oak Ridge National Laboratory, 1972.
- [12] H. J. White. Haynes International Capabilities: HASTELLOY N alloy, ORNL. 2010.
- [13] R. S. Sellers, W. J. Cheng, M. H. Anderson, K. Sridharan, C. J. Wang, and T. R. Allen. Materials corrosion in molten LiF-NaF-KF eutectic salt under different reduction-oxidation conditions. In *Proceedings of ICAPP*, 2012.
- [14] W. J. Cheng, R. S. Sellers, M. H. Anderson, K. Sridharan, C. J. Wang, and T. R. Allen. Zirconium effect on the corrosion behavior of 316L stainless steel alloy and Hastelloy-N superalloy in molten fluoride salt. *Nuclear Technology*, 183 (2013) 248-259
- [15] H. G. MacPherson. Molten-salt reactor program, Quarterly progress report, ORNL-2799. Technical report, Oak Ridge National Laboratory, September 1959.
- [16] G. Zheng, B. Kelleher, G. Cao, K. Sridharan, M. Anderson, and T. Allen. Investigation of 2LiF-BeF₂ (FLiBe): Salt Transfer, Corrosion Tests and Characterization. In *Transactions of the American Nuclear Society*, volume 109 (2013) 259-260
- [17] A. F. Smith. The Diffusion of Chromium in Type 316 Stainless Steel. *Metal Science*, 9(1) (1975) 375-378
- [18] R. Braun and M. Feller-Kniepmeier. Diffusion of chromium in α -iron. *Physica Status Solidi (a)*, 90(2) (1985) 553-561
- [19] R. B. Evans, J. H. DeVan, and G. M. Watson. Self-diffusion of chromium in nickel-base alloys, ORNL-2982. Technical report, Oak Ridge National Laboratory, 1961.
- [20] J. R. Keiser. Compatibility studies of potential molten-salt breeder reactor materials in molten fluoride salts, ORNL-TM-5783. Technical report, Oak Ridge National Laboratory, 1977.
- [21] B. R. Kumar, A. K. Singh, B. Mahato, P. K. De, N. R. Bandyopadhyay, and D. K. Bhattacharya. Deformation-induced transformation textures in metastable austenitic stainless steel. *Materials Science and Engineering: A*, 429(1-2) (2006) 205-211
- [22] A. Das, P. C. Chakraborti, S. Tarafder, and H. K. D. H. Bhadeshia. Analysis of deformation induced martensitic transformation in stainless steels. *Materials Science and Technology*, 27(1) (2011) 366-370

- [23] T. Y. Xing, X. W. Cui, W. X. Chen, and R. S. Yang. Synthesis of porous chromium carbides by carburization. *Materials Chemistry and Physics*, 128(1-2) (2011) 181-186
- [24] H. U. Hong, B. S. Rho, and S. W. Nam. Correlation of the $M_{23}C_6$ precipitation morphology with grain boundary characteristics in austenitic stainless steel. *Materials Science and Engineering: A*, 318 (2001) 285-292
- [25] H. U. Hong and S. W. Nam. The occurrence of grain boundary serration and its effect on the $M_{23}C_6$ carbide characteristics in an AISI 316 stainless steel. *Materials Science and Engineering: A*, 332 (2002) 255-261
- [26] R. Jones, V. Randle, and G. Owen. Carbide precipitation and grain boundary plane selection in overaged type 316 austenitic stainless steel. *Materials Science and Engineering: A*, 496(1-2) (2008) 256-261
- [27] R. P. Agarwala, M. C. Naik, M. S. Anand, and A. R. Paul. Diffusion of carbon in stainless steel. *Journal of Nuclear Materials*, 36 (1970) 41-47
- [28] L. S. Richardson, D. E. Vreeland, and W. D. Manly. Corrosion by molten fluorides, ORNL-1491. Technical Report, Oak Ridge National Laboratory, 1953.