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EVALUATION OF Pb–17Li COMPATIBILITY OF ODS Fe-12Cr-5Al ALLOYS

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Abstract

The Dual Coolant Lead Lithium (DCLL: eutectic Pb–17Li and He) blanket concept requires improved Pb–17Li compatibility with ferritic steels in order to demonstrate acceptable performance in fusion reactors. As an initial step, static Pb-17at.%Li (Pb-17Li) capsule experiments were conducted on new oxide dispersion strengthened (ODS) FeCrAl alloys ((1) Y₂O₃ (125Y), (2) Y₂O₃+ZrO₂ (125YZ), (3) Y₂O₃+HfO₂ (125YH), and (4) Y₂O₃+TiO₂ (125YT)) produced at ORNL via mechanical alloying (MA). Tests were conducted in static Pb–17Li for 1000 h at 700°C. Alloys showed promising compatibility with Pb–17Li with small mass change after testing for 125YZ, 125YH and 125YT, while the 125Y alloy experienced the highest mass loss associated with some oxide spallation and subsequent alloy dissolution. X-ray diffraction methods identified the surface reaction product as LiAlO₂ on all four alloys. A small decrease (~1 at.%) in Al content beneath the oxide scale was observed in all 4 ODS alloys, which extended through 60 μm beneath the oxide/metal interface. This indicates improvements in alloy dissolution by decreasing the amount of Al loss from the alloy. Scales formed on 125YZ, 125YH and 125YT were examined via scanning transmission electron microscopy (S/TEM) and revealed incorporation of Zr-, Hf-, and Ti-rich precipitates within the LiAlO₂ product, respectively. This indicates an inward scale growth mechanism. Future work in flowing Pb–17Li is needed to further evaluate the effectiveness of this strategy in a test blanket module.

Key words: Pb–17Li Compatibility, Cast FeCrAl, ODS FeCrAl, Mass change, Oxide Scale, LiAlO₂

1. Introduction

The US leading blanket model for the ITER test blanket module (TBM) is the Dual Coolant Lead Lithium (DCLL: eutectic Pb–17Li liquid and He) blanket concept. In order to demonstrate acceptable performance in fusion reactors [1-3] it requires improved Pb–17at.%Li (Pb–17Li) compatibility with ferritic steels. Current reduced activation ferritic-martensitic (FM) steels and 9Cr-ODS constrain the DCLL to ~475°C due to rapid Fe and Cr dissolution in Pb–17Li) above ~500°C [4-12]. This compatibility issue is considered a key technological barrier for the DCLL

concept. There are two strategies that are currently being investigated; (1) *coating* [13-18] and (2) *alloying* [19,20]. Both strategies rely upon the presence of Al to form a protective barrier (surface scale) against dissolution in Pb-17Li. The first strategy was used to protect the base metal from dissolution in [15,16,19,21,22]; however, this concept did not provide an adequate solution above 600°C because it still did inhibit the dissolution of Al from the metal-oxide interface [17]. Moreover, above 600°C, Al loss was not only due to the dissolution in Pb-17Li but also due to Al/N interaction at the metal/coating interface, which was identified as a long-term issue that relates to microstructural evolution. For example, a thin (~50 µm) Al diffusion coating deposited via chemical vapor deposition (CVD) methods on Grade 92 steel evaluated for up to 5000 h at 600° and 700°C showed promising mass loss reduction when compared with bare Grade 92 steel tested in Pb-17Li [14,19]; however, Al loss at the coating surface, was significant [19]. For the case of a thicker coating, which would have a significantly higher amount of Al reservoir, was not a suitable solution above 600°C and also showed a significant drop in the surface Al content [19], which in the long term can cause a problem due to Al reservoir depletion and subsequent inability to supply Al to the surface scale in order to maintain the barrier between the alloy and Pb-17Li. At lower temperatures (<500°C), specimens needed to be pre-oxidized in order to form a protective oxide layer regardless of the Al-rich coating system was used due to low solute diffusivities [21].

Another possibility was the *alloying* approach [19,20]. Pb-Li exposure experiments on model cast FeCrAl alloys suggested that using alloys with Al contents at or above 5 wt.% minimized mass loss during exposure to Pb-17Li [19]. Compositions of 10-15 wt.% Cr and 4.5-5 wt.% Al showed promising behavior with the formation of a LiAlO₂ surface layer during exposure in static Pb-17Li at 700°C. This led to the development of oxide dispersion strengthened (ODS) Fe-12wt.%Cr-5wt.%Al alloys with similar composition to cast model alloys that would combine improved Pb-17Li compatibility with superior high-temperature mechanical properties and irradiation resistance.

Therefore, the aim of this study was to fabricate ODS FeCrAl with the compositions of 12 wt.%Cr and 4.5-5 wt.% Al and to evaluate their Pb-17Li compatibility. ODS FeCrAl alloys were exposed for 1000 h at 700°C in isothermal static Pb-17Li capsules as a preliminary evaluation prior to flowing Pb-17Li experiments and showed small mass change after exposures.

2. Experimental

In this study, four ODS FeCrAl alloys were produced with different oxide dispersions via mechanical alloying (MA): (1) Y₂O₃ (125Y) (2) Y₂O₃ + ZrO₂ (125YZ), (3) Y₂O₃ + HfO₂ (125YH), and (4) Y₂O₃ + TiO₂ (125YT) with the compositions shown in Table 1. For comparison, two cast FeCrAl alloys (Fe12Cr5Al) are also listed that were produced by arc-melting in inert gas (Ar) followed by water cooling of the Cu mold after casting. Detailed microstructural characterization of the ODS FeCrAl alloys after mechanical alloying revealed the

presence of Al_2O_3 and yttrium aluminum garnet (YAG) phases ($\text{Y}_3\text{Al}_5\text{O}_{12}$) in the 125Y alloy, while in the 125YZ alloy additional $\text{Zr}(\text{C},\text{N})$ precipitates were discovered [23]. The 125YH alloy had the most complex precipitation sequence and in addition to YAG and Al_2O_3 phases, $\text{Hf}(\text{C},\text{N})$, $\text{Y}_2\text{Hf}_2\text{O}_7$ and HfO_2 precipitates were also found. Detailed characterization of the 125YT alloy is still in progress.

For exposure in static Pb–17Li specimens (~ 1.5 mm thick and $\sim 4\text{--}5$ cm²) were polished to a 600 grit SiC finish. All specimens were cleaned ultrasonically in acetone and alcohol and their mass was recorded before and after Pb–17Li exposure, using a Mettler-Toledo model XP205 balance with an accuracy of ± 0.04 mg.

Static capsule tests were performed using Mo (inert to Pb–17Li) inner welded closed capsules and type 304 stainless steel (SS) outer capsules to protect the inner capsule from oxidation [17,19]. One specimen was exposed in each capsule and held with 1 mm diameter Mo wire. The capsules were loaded with 125 g of Pb–17Li in an Ar-filled (< 10 ppm O_2 and < 1 ppm H_2O) glove box. The Pb–17Li was purchased from GMH Stachow-Metall GmbH, Germany, with Li content of 0.68 ± 0.01 wt.% (17.0 at.%). Other elements such as Al, Cr, Fe, Mn, Mo, Ni, Si, Ti, Y, Zr, S, N were less than detection limits in 1–10 wppm range. Average O level was 1200 wppm (range 540–2400), while average C content was 240 wppm (range 130–420). In this test cast Pb–17Li at ORNL was also used with 200–400 ppmw of O. The Mo and SS capsules were welded shut within an argon glove box to prevent the uptake of impurities during the isothermal exposure carried out in a resistively heated box furnace. After exposure in Pb–17Li, residual Pb–17Li on the specimen surface was removed by soaking in a 1:1:1 mixture of acetic acid, hydrogen peroxide and ethanol for up to 72 h, followed by cleaning in methanol.

Post-test specimen surfaces were examined using x-ray diffraction (XRD), scanning electron microscopy (SEM) imaging combined with energy dispersive x-ray spectroscopy (EDS) analyses for chemical analysis. After surface characterization, the specimens were sectioned, polished and examined by light microscopy and electron probe microanalysis (EPMA) using wavelength dispersive x-ray analysis. The analyzed results from the ODS FeCrAl alloys were compared with the cast FeCrAlY alloys containing 12Cr–5Al–0.06Y in wt.%. Detailed characterization of selected surface oxides formed during the Pb–17Li experiments was also carried out using scanning transmission electron microscopy (S/TEM) and EDS analysis using a Philips model CM200 scanning-transmission electron microscope equipped with a Schottky field emission gun (FEG) and EDAX system operated at 200 kV. The specimens for S/TEM analysis were prepared via the *in situ* lift-out method using the focused ion beam (FIB) milling technique using Hitachi NB5000 FIB-SEM.

3. Results and Discussion

The mass change data for the four ODS FeCrAl alloys (125Y, 125YZ, 125YH (repeat), 125YT) exposed in Pb–17Li for 1000 hours at 700°C are summarized in the bar plot shown in Figure 1. For reference, the mass change results for the cast FeCrAl alloys (Fe12Cr5Al = 125Y) [13], the 14YWT (ODS Fe–Cr) ferritic alloy [17] and the commercial PM2000 (ODS FeCrAl) alloy were also included (Tab. 1). Minimal mass change was observed for ODS FeCrAl alloys containing Y and either Zr, Hf or Ti elements. ODS FeCrAl alloy with only Y showed significantly higher mass loss. Two scenarios could account for the 125Y results; (1) the oxygen level within the alloy and/or (2) reactive elements (REs). All three 125YT, 125YH and 125YZ had higher O content than 125Y (Table 1), which may reflect that more of the Al was tied up as fine ~3nm oxides ($Y_3Al_5O_{12}$ (YAG phase)) in these alloys than in 125Y alloy. It can be hypothesized that small Al-rich oxides act as a reservoir for Al, since they are uniformly distributed throughout the matrix to form a surface oxide. Thus, the amount of these nano-sized oxides to support surface oxide would be reduced in 125Y. In terms of reactive elements, previous research [24,25] on high-temperature oxidation has shown that in order to improve oxidation behavior of the material, co-doping (more than one) of reactive elements is more desired than a single doping, which is evident in this study. The three alloys 125YT, 125YH, and 125YZ in addition to Y, also contained second element such as Ti, Hf or Zr, respectively. In general, REs improve high-temperature oxidation by slowing down the oxide growth, improving scale adhesion, and decreasing the possibility of scale spallation. The highest mass loss was recorded for the 125Y specimen and linked to the spallation of the surface oxide and also to the presence of deep pits (~300 μm) at the surface (Fig. 2) associated with material dissolution in Pb–17Li due to absence of protective barrier (surface oxide). This suggests that the scale was not adherent or might be damaged. In terms of pits, alloy powder with little or no Al might be incorporated in these local areas of mechanically alloyed material, which was unable to reform a protective film after spallation, and therefore, material was selectively dissolved from the specimen. EPMA results generated beneath the pit (solid line, Fig. 2) showed a large Al content drop from ~9 at.% measured in the bulk alloy to 3.5 at.% measured beneath the pit, which supports low Al level in these areas. This also shows that the Al level needs to be higher than 3.5 at.% to reform protective surface layer, which supports our previous results [19], in order to prevent dissolution of the steel. Nevertheless, the small mass change of three ODS FeCrAl (125YZ, 125YH and 125YT) alloys was consistent with the cast FeCrAl alloys of similar composition (Fig. 2/Tab. 1) confirming the beneficial effect of the presence of more than one reactive element in the alloy in addition to higher Al content. Regarding the different microstructures, the fine-grain structure in ODS alloys versus large grains in the cast alloy (cast 125Y) does not seem to have an effect on the dissolution rate, as was found by Li *et al.* [10,11] in Al-free 9Cr-ODS alloy.

In all alloys containing Al, the XRD analysis indicated the presence of the LiAlO_2 layer and with the grazing incidence setup the x-ray penetration depth of LiAlO_2 (scale thickness) was estimated to be around 3 μm . This corrosion product (LiAlO_2 layer) was formed [17] from O impurities in the Pb–17Li (200–400 ppmw of O in as-cast ORNL Pb–17Li; average O level of 1200 ppmw in Pb–17Li purchased from Germany). Even though Al_2O_3 was not formed/detected, LiAlO_2 layer still acted as an advantageous barrier for decreasing/preventing alloy dissolution into the liquid metal [26]. The benefit of the Al presence in ODS FeCrAl alloys can be clearly

seen when mass changes of ODS FeCrAl alloys are compared with the ODS Fe-Cr alloy (14YWT) (Tab. 1) for which the highest mass loss (2.9 mg/cm^2) was recorded (Fig. 1) [14,17]. Both Fe and Cr have a high solubility in Pb-17Li, and without presence of Al to form a stable oxide, dissolution occurs rapidly. The dissolution rate decreases, when the liquid reaches saturation [14]. The commercial ODS alloy, PM2000, had a smaller mass loss ($\sim 0.2 \text{ mg/cm}^2$) than 14YWT due to the presence of Al in the alloy, resulting in the formation of a surface scale, which protects the alloy from the external environment. However, PM2000 showed poorer performance than the other three ODS FeCrAl alloys (Fig. 1). One reason could be alloy chemistry and the difference in precipitation type and their distribution. PM2000 contains mainly Y as a dopant with minor addition of Ti. As previously mentioned, in order to improve oxidation behavior co-doping is more favorable over single doping with the careful selection of the reactive elements [25,27]. Combination of Y and Ti is less effective [27] than combination of Y and Hf or Y and Zr. On the contrary, the cast 125Y FeCrAl alloy with an amount of Cr (12 wt.%) and Al (5 wt.%) similar to ODS FeCrAl alloys had a comparable mass change to the ODS FeCrAl alloys even though it contained only one reactive element (Y). This shows that there must be other factors in addition to REs such as either uniform distribution of YAG phase in ODS FeCrAl or oxygen level affecting Pb-17Li compatibility that requires further investigation. The higher oxygen level in PM2000, which was much higher (8056 ppma) than in ODS FeCrAl ($< 7500 \text{ ppma}$), could affect the oxide precipitation and Al distribution.

It should be mentioned that the 125YH specimen was tested twice in Pb-17Li. Surprisingly, the first results showed similar high mass loss to the 125Y specimen [20]. There was no indication of significant alloy dissolution but spallation near the specimen edges was observed. This could be either due to the specimen preparation for the test, the oxygen level in Pb-17Li (first test was preformed with ORNL Pb-17Li (200-400 ppmw of O) while second Pb-17Li exposure/test with Pb-17Li purchased from Germany (1200 ppmw of O)), or non-uniform distribution of elements throughout the material near the surface in the alloy due to mechanical alloying. Nevertheless, the second exposure in Pb-17Li was run with additional care during specimen preparation (grinding with SiC) and as shown in Figure 1 the mass change was small and comparable with mass losses to alloy 125YZ and 125YT.

Metallographic cross-sections were prepared from the specimens after exposure to Pb-17Li to investigate surface scales that were supposed to be a barrier to prevent dissolution of the material in Pb-17Li and to compare their morphology with scales formed on the cast 125Y, commercial PM2000 and 14YWT alloys (Tab. 1). The metallographic cross-sections are shown in Figure 3. A thin ($1.5\text{-}3 \mu\text{m}$) scale can be observed in all Al containing alloys, while in 14YWT (ODS Fe-Cr alloy) a thick $19.8 \pm 3.7 \mu\text{m}$ “interaction layer” was found. However, this layer was not an oxide but it was mixture of Pb-17Li and alloy, which was like a “sponge” filled with Pb (detected by EPMA). This suggests penetration of the liquid Pb-17Li into the solid structure without suitable protective barrier [7,9,28,29] and alloy dissolution (mass loss of 14YWT Fig. 1) of the base metal showing the importance of its presence.

Due to the metallographic mounting almost all of the scales were detached from the substrates. This introduced challenges in analyzing elemental segregation and thickness of the scale cross-

sections via either light microscopy or SEM. However, evaluation of the elemental depletion beneath the exposed oxide/metal interface via EPMA is still valid, and it was carried out. The EPMA results for normalized Al and Cr elements for the 125Y, 125YZ, 125YH, 125YT specimens are plotted in Figure 4. For comparison, EPMA results from the cast 125Y and PM2000 alloys are also included. For the 125YT specimen, there was no variation in Al content (~9 at.%) and only a slight change in Cr level (from 12 to 11.5 at.%) at the oxide/metal interface with a small Cr depletion reaching about 4 μm depth beneath the interface. For the other three ODS FeCrAl specimens (125Y, 125YH and 125YZ), a small drop in Al content at the oxide/metal interface was observed extending through about 60 μm depth beneath the interface. In all three alloys measured Al level in the bulk was 8.5 at.% and decreased to 8.0 at.% at the oxide/metal interface for 125Y, $\sim 7.75 \pm 0.4$ at.% for 125YH and 7.4 ± 0.1 at.% for 125YZ. The Cr depletion depth of ~ 4 μm was similar among these three ODS FeCrAl alloys and was comparable with the Cr depletion depth found in 125YT. In alloy 125YH and 125YZ, there was a drop in Cr content of 0.5 at.% between location measured in the bulk (12 at.%) and at the oxide/metal interface (11.5 at.%), while 1 at.% drop in Cr was measured for 125Y alloy at the oxide/metal interface. Comparing results with the cast 125Y and PM2000 alloy, it was found that the Al depletion depth was similar (~ 60 μm) to ODS FeCrAl alloys with Al content drop of 1 at.%. A larger (~ 10 μm) Cr depletion depth with a drop of 1 at.% was found for cast 125Y alloy, while no variation in Cr concentration was found in PM2000 beneath the oxide/metal interface. In Al-free ODS Fe-Cr alloy (14YWT), the drop in Cr concentration at the surface was about 3 at.% (Fig. 5a) with the depth of 60 μm , which is similar to the depth for aluminum in Al-containing alloys. Additionally in 14YWT, C depletion was found (Fig. 5b), which accompanied Cr depletion with its drop of 2 at.% at the surface.

Detailed analysis of the cross section of the surface reaction products was not feasible from the bulk specimens exposed to Pb-17Li for 1000 h at 700°C, due to oxide layer separation/detachment from the substrate during metallographic mounting (Fig. 3). Consequently, TEM specimens were prepared for 125YT, 125YZ and 125YH specimens using FIB-milling methods from the top surface. An example of the bright-field (BF) and high-angle annular dark-field (HAADF) STEM images for alloy 125YZ are shown in Figures 6 and 7 (higher magnification). The scale thickness is approximately 1.36 ± 0.2 μm and it is continuous with some small separation at the oxide/metal interface. Similar separation at the interface was found on these alloys after exposure to steam at 1200°C for 4h [20], which suggests it is alloy specific and not due to exposure to Pb-17Li. It could be speculated that interface separation is most likely due to void (introduced during mechanical alloying) coalescence during high-temperature exposure. The oxide scale grains appeared to be columnar indicative of inner scale growth with a number of small voids throughout the scale incorporated during its growth. Additionally, precipitates were incorporated from the substrate throughout the scale, which additionally supports the inward scale growth mechanism. The scale itself was extremely beam sensitive and electron diffraction analysis was not possible due to the scale damage. Similar observation of LiAlO_2 scale sensitivity under the beam was observed by Pint and More [26], which further supports the presence of LiAlO_2 , and is consistent with our XRD results.

Figure 8 shows the oxide cross-section formed on 125YT after 1000 h testing in Pb–17Li at 700°C. The scale thickness was $\sim 0.77 \pm 0.3 \mu\text{m}$ and it also contained precipitates within the reaction product, which supports the inward scale growth mechanism. EDS point analysis indicated that these precipitates were mainly Ti-rich and no strong oxygen or Y concentration was detected along with Ti signal generated from the precipitates. Advanced electron diffraction analysis of these precipitates was not possible due to electron beam damage of the oxide. Atom probe tomography (APT) attempts are in progress to help identify these precipitates, to investigate if there is a higher Li concentration associated with the precipitates found in the scale, and to determine if they are still oxides.

STEM was used to examine the 125YH specimen exposed to static Pb–17Li for 1000 h at 700°C, and an example of the oxide cross-section is shown in Figure 9. The scale thickness was $\sim 0.65 \pm 0.1 \mu\text{m}$ and some of the Pb–17Li remnant was still present on the scale even after 72 hours of cleaning/soaking in a 1:1:1 mixture of acetic acid, hydrogen peroxide and ethanol. Within the scale, precipitates were also observed and EDS analysis showed mainly Hf enrichment without any strong oxygen or yttrium signal concentration suggesting Hf(C,N). The reaction product was mainly single layer with columnar grains throughout the entire scale thickness. Both the columnar grains and incorporated Hf-rich precipitates support an inward scale growth mechanism. However, the scale appearance is rather different than it is generally observed in the high-temperature testing in various gas media suggesting of some differences in the mechanism of scale growth between liquid and gas media at high temperature. In the scales formed in gas media, for instance in 100% H₂O (steam) [20], usually the oxide precipitates from the alloy are not incorporated, which was found in scales formed during Pb–17Li testing. If the oxide precipitates are found in the scale, they are mainly located along oxide grain boundaries and at the triple points due to the outward diffusion of reactive elements along scale grain boundaries. On the contrary, similar precipitate distribution and incorporation throughout the scale was found after testing of Mg alloys in salt water [30]. This suggests and supports the notion that even though the scale growth is inward in the liquid media, it is still different than in the gas media and requires further investigation.

Based on the results, it can be seen that the previous prediction of the composition established on the cast specimens was a helpful strategy that allowed narrowing down the composition of the ODS FeCrAl alloys and reduced the cost of the evaluation process. Three out of the four of the ODS FeCrAl alloys produced at ORNL [24] showed promising compatibility with Pb–17Li with small mass change after testing for 1000 h at 700°C in static Pb–17Li. This alloying strategy only addresses compatibility in static testing and the new ODS FeCrAl alloys will be evaluated next in flowing Pb–17Li experiments. It should be mentioned that this alloying strategy only addresses Pb–17Li compatibility and does not address tritium control nor provide a solution to the magneto-hydrodynamic pressure drop [31].

4. Conclusions

Static Pb–17Li capsule experiments were conducted to evaluate new ODS FeCrAl alloys ((1) Y₂O₃ (125Y), (2) Y₂O₃+ZrO₂ (125YZ), (3) Y₂O₃+HfO₂ (125YH), and (4) Y₂O₃+TiO₂ (125YT)) produced at ORNL via mechanical alloying (MA) to improve Pb–17Li compatibility of Fe-base alloys above 600°C. Tests were conducted in static Pb–17Li for 1000 h at 700°C. Alloys showed promising compatibility with Pb–17Li with small mass change after testing for 1000 h at 700°C in static Pb–17Li for 125YZ, 125YH and 125YT, while 125Y alloy experienced the highest mass loss associated with spallation of the oxide and dissolution of the substrate. XRD methods identified the surface reaction product as LiAlO₂ in all four alloys. Al depletion was observed beneath exposed surfaces in all 4 ODS alloys, with an Al drop of ~1 at.% at the surface. Scales formed on 125YZ, 125YH and 125YT were examined via STEM and showed Zr-, Hf-, and Ti-rich precipitates incorporated throughout the LiAlO₂ product, respectively, with an identified inward scale growth mechanism. The next step will be to evaluate the new ODS FeCrAl alloys in flowing Pb–17Li experiments.

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Figure Captions

Figure 1. Mass change results for several alloys after capsule exposure in Pb-Li for 1000 h at 700°C.

Figure 2. EPMA results for normalized Al content (at.%) generated from the pit/dissolved area (0µm) towards the alloy after exposure to Pb-Li at 700°C for 1000h. For reference EPMA for regular exposed surface/area is also included (dash line).

Figure 3. BSE images of as-polished cross-sections of different alloys after exposure in static Pb–17Li for 1000 h at 700°C.

Figure 4. EPMA results for Normalized a) Al and b) Cr content (at.%) generated from the exposed surface (0µm) towards the alloy for different alloys after exposure to static Pb–17Li at 700°C for 1000 h.

Figure 5. EPMA results for Normalized a) Cr and b) C content (at.%) generated from the exposed surface (0 μm) towards the alloy for 14YWT alloy after exposure to static Pb–17Li at 700°C for 1000 h.

Figure 6. a) HAADF and b) BF-STEM image of the final oxide cross-section formed on alloy 125YZ for 1000 h exposure at 700°C in static Pb–17Li.

Figure 7. Higher magnification images of BF-STEM (a,c) and HAADF (b) of the oxide cross-section formed on 125YZ after 1000 h exposure at 700°C in static Pb–17Li with EDS point spectrum (d) generated from particle within the oxide.

Figure 8. a) BF-STEM image and b) DF-STEM image of the oxide cross-section formed on 125YT during 1000 h exposure at 700°C in static Pb–17Li. c) EDS elemental maps generated from Figure 8a.

Figure 9. a) BF-STEM image and b) HAADF-STEM image of the oxide cross-section formed on 125YH during 1000 h exposure at 700°C in static Pb–17Li. c) and d) higher magnification images, respectively. Insert in Fig. 9c shows Ti EDS elemental map.

Table 1. Chemical composition (at.%) determined by Inductively Coupled Plasma Atomic Emission Spectroscopy (ICP-AES). Values for Y, O, N, C, S in ppma.

Material	Fe	Cr	Al	Y	N	C	S	O	Others
Cast 105Y	80.4	10.2	9.4	230	10	220	21	40	
Cast 125Y	79.1	12.2	8.7	160	30	220	21	60	0.02Si
125Y	78.3	11.5	9.3	1120	1710	1660	33	2760	0.04Si
125YH	77.5	11.8	9.4	1010	410	960	16	7500	0.200Hf-0.006Zr
125YZ	77.8	11.6	9.5	1060	600	1090	16	6290	0.003Hf- 0.17Zr
125YT	77.2	12.1	9.4	940	500	1520	49	7258	0.009Hf-0.02Si
14YWT	83.25	14.07	0.04	1240	5166	3001	69	5997	0.585W
PM2000	68.97	19.12	10.5	2280	319	61	13	8056	0.06Mn-0.02Ni- 0.03P-0.04Si- 0.52Ti

Figure

Figure 1.

Figure 2.

Figure 3.

Figure 4.

Figure 5.

Figure 6.

Figure 7.

Figure 8.

Figure 9.