Metal loss and corrosion attack of FeCrAl overlay welds on evaporator tube shields of a waste-fired power plant

Abstract
Three FeCrAl alloys (APMT, EF100 and EF101) from Kanthal® and the reference Ni-Cr Alloy 625 were used as weld cladding materials on tube shields in the evaporator tube bank of a waste-fired combined heat and power plant. For each alloy type, the overlay welded tube shields were placed in both roof and floor positions within the evaporator for 6 months. The metal-loss rate, the microstructure and hardness of the overlay welds before and after exposure and the corrosion products were analysed. The results showed higher metal-loss rates in the welds placed in the roof position, confirming heterogeneities in the evaporator bank environment. Alloys were ranked from higher to lower erosion–corrosion resistance as follows: APMT \( \approx \) Alloy 625 > EF101 > EF100. The analysis of the corrosion attacks showed a significant variation among the alloys, from a primarily homogeneous corrosion attack on APMT to intergranular corrosion in EF100 and pit formation in EF101.

KEYWORDS
corrosion, erosion, FeCrAl alloys, overlay welding, power generation, waste incineration, waste-fired power plant

1 | INTRODUCTION
Increasing ecologic and economic sustainability is a driving ambition not only for waste incineration plants but for the whole energy production sector and the European Union.[1–4] Consequently, continuous research is being performed to increase the efficiency of high-temperature materials regarding extended service life, high strength/weight ratio and resistance to extreme corrosion environments.[5–7]

High-temperature corrosion caused by the combustion of biomass in boilers is a concern in waste incineration plants. The furnace walls and the superheater and evaporator banks consist of tubes containing pressurised water and steam, which are heated by the combustion gases generated in the furnace. The tubes are commonly made of low alloyed carbon steels due to their low cost, low risk of stress corrosion cracking, high heat transfer properties and low thermal expansion. However, low alloyed steels corrode very rapidly when combusting recycled wood, mainly due to the high concentration of heavy metals, alkali metals and chlorine that are found in biomass combustion environments.[8–9] To reduce the corrosion rate of those low alloyed carbon steels, one approach is to coat or clad them with a more corrosion-resistant material.

Recently, the corrosion resistance of several alloys under different environments in waste incineration plants was evaluated and compared.[9–10] Results showed that FeCrAl...
tubes of Kanthal® APMT had a lower corrosion rate than alloys, such as AISI 310S, EN 10028 16Mo3, and slightly lower than nickel-based alloy EN Ni6625, which is the reference alloy in these applications. Those results opened the door to considering the use of FeCrAl alloys in claddings, as possible alternative alloys to mitigate corrosion problems in furnace walls, and superheater and evaporator banks. FeCrAl alloys are known to form protective oxide scales at temperatures above 900°C providing outstanding corrosion protection. However, at lower temperatures (<900°C), the formation of the protective α alumina in FeCrAl alloys might be hindered due to slow kinetics, allowing other less protective oxides to form. In addition, alkali chlorides and other corrosive species present in the boiler environment can exacerbate the corrosion process and lead to breakaway corrosion. The challenge of improving the corrosion resistance of FeCrAl alloys under these conditions has been faced with the addition of alloying elements, such as silicon and small amounts of reactive elements (REs), although the underlying mechanisms are yet to be fully described.

The environment in the superheater chamber is not homogeneous and the tubes are exposed to a combination of factors, such as high temperature, corrosive species, solid particles and long-time exposure cycles. Therefore, using short-scale samples or in-lab corrosion tests would be far from replicating these factors to provide reliable results. The novel aspects in this investigation are the fact that the welds have been exposed on-site to the boiler environment in the form of real-size overlay welds on tube shields, and very importantly, the use of newly developed FeCrAl alloys as weld cladding materials. The performance of FeCrAl alloys as weld cladding materials can differ from the industrially manufactured tubes; therefore, investigation on weldability and corrosion resistance of overlay welds in boiler applications is needed.

2.1 | Materials

The substrate used was the low-alloyed Mo steel grade EN 10028 16Mo3 in the form of 4 mm thickness, 48.3 mm outer diameter and 2 m length tubes. This is a structural carbon steel with molybdenum content between 0.25% and 0.35% for service temperatures below 500°C.

Three FeCrAl alloys, Kanthal® EF100 (formerly called L198), Kanthal® EF101 (formerly called L197) and Kanthal® APMT, in the form of ø 1.2 mm solid wires were used as cladding materials. These alloys contain in their formulation small amounts of REs, which are undisclosed due to intellectual properties rights, and they are intentionally added to improve the oxidation resistance. As reference cladding material, Sanicro® 60 (AWS/SFA 5.14 ERNiCrMo-3, EN ISO 18274 Grade Ni6625), commonly known as Alloy 625, was used in the form of ø 1.2 mm solid wire.

Table 1 shows the nominal chemical composition of the cladding materials used.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Cr</th>
<th>Al</th>
<th>Si</th>
<th>Mn</th>
<th>C</th>
<th>Mo</th>
<th>Fe</th>
<th>Ni</th>
<th>Nb</th>
</tr>
</thead>
<tbody>
<tr>
<td>Kanthal® EF100</td>
<td>10.0</td>
<td>4.0</td>
<td>0.3</td>
<td>&lt;0.7</td>
<td>&lt;0.08</td>
<td></td>
<td>Bal.</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Kanthal® EF101</td>
<td>12.5</td>
<td>3.7</td>
<td>1.3</td>
<td>&lt;0.7</td>
<td>&lt;0.08</td>
<td></td>
<td>Bal.</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Kanthal® APMT</td>
<td>21.0</td>
<td>5.0</td>
<td>&lt;0.7</td>
<td>&lt;0.4</td>
<td>&lt;0.08</td>
<td>3.0</td>
<td>Bal.</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Alloy 625</td>
<td>22.11</td>
<td>0.09</td>
<td>0.09</td>
<td>0.07</td>
<td>0.01</td>
<td>9.09</td>
<td>0.42</td>
<td>64.22</td>
<td>3.51</td>
</tr>
</tbody>
</table>

Note: Bal. means that iron (Fe) weight composition is the remaining element until reaching the 100% in wt.

Abbreviation: Bal., Balance.

2.2 | Welding

Mechanised gas metal arc welding was used to prepare overlay welds on the 16Mo3 tube shields. The power source was a TPS 5000 CMT coupled with a robotic arm and the shielding gas was pure argon (99.99%).

Four tubes were longitudinally overlayed, one alloy per tube. For the FeCrAl alloys, one-layer overlay welds were prepared, whilst for Alloy 625, a two-layer overlay weld was prepared to avoid exceeding the maximum limit of 5%–10% iron content due to the dilution with the 16Mo3 substrate. Higher iron content in this nickel-based alloy would have been detrimental to its corrosion resistance.
The welding settings are shown in Table 2. Once the tubes were overlay welded, they were longitudinally cut into two halves, resulting in two tube shields for each alloy to be exposed in the boiler. To keep the interpass temperature below 150°C, there was forced air cooling inside the tubes during the welding process.

### 2.3 On-site exposures

The exposure of the overlay welded tube shield took place in the waste-fired combined heat and power plant designated as P14 at Händelö, E.ON utility, Norrköping, Sweden. The boiler is a 75 MW circulating fluidised bed (CFB) boiler generating steam with 65 bar pressure and a final temperature of 450°C. It is designed for fuel flexibility, using a fuel mix of 30%–50% combined household waste, 50%–70% classified industrial waste and up to 20% of sewage sludge. The superheater channel, following the empty pass, comprises both superheater banks and evaporator banks, see Figure 1 (view from the right side of the superheater channel). The exposure took place in the evaporator bank just below the superheaters. There are 100 tubes in the evaporator, 50 of them located on the floor and 50 located on the roof of the chamber. Two steam soot-blowing systems are placed in the chamber, one on each side to cover the entire volume of the chamber. The type of steam soot blowers were long retractable and rotating, and they were operational for 6 min every 24 h. There was no dwell time between the enter and return sequence of the blowing operation of the soot-blowing system.

As previously mentioned, two tube shields for each alloy were prepared and the exposure was designed so that each alloy was exposed in both positions: floor and roof, as detailed in Table 3. The tube numbers are counted from one wall to the other, with a total of 50 tube positions. The chemical composition of the ash deposit that got attached to the tube shields during the exposure was characterised using energy-dispersive spectroscopy (EDS) (see Table 4) and composed primarily of alkali chlorides.

During the 6-month exposure period, the sensors in the boiler continuously monitored several operational

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### Table 2 Overlay welding settings.

<table>
<thead>
<tr>
<th></th>
<th>FeCrAl alloys</th>
<th></th>
<th>Alloy 625</th>
</tr>
</thead>
<tbody>
<tr>
<td>Current (A)</td>
<td>160–200</td>
<td>Voltage (V)</td>
<td>185–200</td>
</tr>
<tr>
<td>Voltage (V)</td>
<td>22</td>
<td>Polarity</td>
<td>19–21</td>
</tr>
<tr>
<td>Polarity</td>
<td>DCEP</td>
<td>Wire feed speed (m/min)</td>
<td>DCEP</td>
</tr>
<tr>
<td>Wire feed speed (m/min)</td>
<td>6.3–7.2</td>
<td>Welding speed (mm/min)</td>
<td>7.2–7.5</td>
</tr>
<tr>
<td>Welding speed (mm/min)</td>
<td>480</td>
<td>Stick out (mm)</td>
<td>455–466</td>
</tr>
<tr>
<td>Stick out (mm)</td>
<td>12–14</td>
<td>Shielding gas flow (L/min)</td>
<td>10–12</td>
</tr>
</tbody>
</table>
| Shielding gas flow (L/min) | 15           | Abbreviation: DCEP, direct current electrode positive.
parameters such as the steam temperature, and the concentration of HCl, SO2, humidity and oxygen in the flue gas. A complete set of data was extracted every 30 min to analyse the exposure conditions in this work. The average values for the operational data during the 6-month exposure are shown in Table 5. These values only refer to the steady-state regime. The operational data measurements during the days of ramp up and ramp down of the boiler, when the flue gas temperature was lower than 450°C were not considered. The flue gas temperature was measured to be on average 620°C at the location of the exposed tube shields. It varied within the interval 510–690°C and the standard deviation was 30°C. The temperature of the tubes is estimated to be 30–50°C above the saturation temperature of the water at the operational pressure of the boiler (281°C), and the temperature of the tube shields is estimated to be somewhat above that temperature, with some variations caused by possible differences in contact between the tube shields and the tubes.

### 2.4 Metal-loss rate

The clad “thickness” is the commonly measured parameter to evaluate and compare the performance of claddings. When defining the total overlay thickness, the ISO-15614-7 standard related to overlay welding[20] does not consider the uneven surface of the overlay welds (bead overlapping causing a wavy surface), nor the case of curved substrates (as it happens in tubes). Due to the impossibility of finding relevant standards to measure the clad thickness of overlay welds applied on tubes, an alternative methodology was used. It was developed by the authors and described in detail in a previous publication.[16] In short, the metal-loss rate and therefore the erosion–corrosion rate are evaluated from the cross-sectional area and the angle of the circular segment. They are measured before and after the exposure to calculate the area loss and are normalised to 180° to be comparable among the different sections. By using the area loss, the calculated arc length, and the exposure time, it is possible to estimate an average thickness loss per time unit, which results in the erosion–corrosion rate or in general, the metal-loss rate. A linear rate is assumed along the timeline. The localised heterogeneity of the environment in the evaporator tube banks is well-known[16] and an estimation of the overall metal-loss rate is needed to compare the performance of the different alloys in the whole environment.

After the exposure, three specimens were cut from each tube shield to investigate their cross-sections. Figure 2 shows the position where the three specimens were cut from the cladded section of the tube shield: one in the middle (Centre, C) and the other two at about 25 mm from the end of the cladded section (Outer Rear and Outer Front toward the rear side and front side of the boiler system, respectively).

To preserve the scale during the cutting and subsequent examinations, clear epoxy resin was used to cast the areas where the specimens were cut. Afterwards, the sections were ground and polished according to conventional metallographic preparation procedures. To evaluate the metal-loss rate, digital images of the cross-sections were inspected using image analysis (ImageJ[21]).

### 2.5 Microstructural inspection and microanalysis

For the microstructural inspection, Zeiss Axio Imager.M2m and Olympus BX60M were the light optical

### Table 3

<table>
<thead>
<tr>
<th>Position/Position/</th>
<th>Kanthal®</th>
<th>Kanthal®</th>
<th>Kanthal®</th>
<th>Alloy</th>
</tr>
</thead>
<tbody>
<tr>
<td>Roof</td>
<td>EF100</td>
<td>EF101</td>
<td>APMT</td>
<td>625</td>
</tr>
<tr>
<td>Floor</td>
<td>#21</td>
<td>#13</td>
<td>#18</td>
<td>#12</td>
</tr>
<tr>
<td>Floor</td>
<td>#7</td>
<td>#8</td>
<td>#9</td>
<td>#6</td>
</tr>
</tbody>
</table>

### Table 4

<table>
<thead>
<tr>
<th>O</th>
<th>Cl</th>
<th>K</th>
<th>Ca</th>
<th>Na</th>
<th>S</th>
<th>Si</th>
<th>Fe</th>
<th>Al</th>
<th>Zn</th>
<th>Mg</th>
<th>Ti</th>
<th>Cu</th>
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<tr>
<td>30.5</td>
<td>23.2</td>
<td>17.1</td>
<td>11.4</td>
<td>5.9</td>
<td>5.8</td>
<td>1.6</td>
<td>1.5</td>
<td>1.4</td>
<td>0.4</td>
<td>0.4</td>
<td>0.3</td>
<td>0.3</td>
</tr>
</tbody>
</table>

Abbreviation: EDS, energy-dispersive spectroscopy.

### Table 5

<table>
<thead>
<tr>
<th>HCl in flue gas (mg/Nm³)</th>
<th>SO₂ in flue gas (mg/Nm³)</th>
<th>H₂O in flue gas (vol%)</th>
<th>O₂ in flue gas left side (vol%)</th>
<th>O₂ in flue gas right side (vol%)</th>
<th>Steam temperature secondary superheater (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>507 ± 145</td>
<td>16 ± 41</td>
<td>15 ± 3</td>
<td>6 ± 1</td>
<td>6 ± 1</td>
<td>419 ± 6</td>
</tr>
</tbody>
</table>
microscopes used. For scanning electron microscopy (SEM) and EDS microanalysis: Zeiss EVO 50 with EDS probe Oxford INCA x-stream-2, FE-SEM Gemini 450 and Zeiss Sigma VP300 equipped with EDS X-maxN were used. The SEM software was SmartSEM (v5.4) and the EDS software was Aztec (v6.0).

2.6 | Hardness test

The hardness of the overlay welds was measured in the as-deposited condition and after having been exposed for 6 months as tube shields in the evaporator bank.

The hardness tester was a Struers Duramin 40 AC1 previously calibrated (calibration block 304 ± 6.5 HV0.3). Vickers hardness was chosen with an applied load of 300 g and 10 s dwell time. Ten individual measurements were conducted along a horizontal linear pattern along the centre line of the overlay weld. As the clads were deposited on a tube, that pattern gave the possibility to measure hardness in positions closer to the surface and the fusion boundary, with potential local differences in the chemical composition (due to dilution or diffusion of certain elements during the exposure). Therefore, that testing pattern was expected to give a representative overall hardness value for the overlay. For Alloy 625 (two-layer overlay), the reported hardness measurements correspond to the second layer, as that was the one exposed to the environment.

In the APMT overlay welds, there is a relative increase of 22.5% in hardness (from 269 to 330 HV0.3) when the alloy is exposed in the evaporator bank. The trend observed is the opposite in the overlay welds of EF100, as the mean value shows a relative decrease of 17.3% (from 234 to 193 HV0.3) due to the exposure conditions.

For EF101 and Alloy 625 overlay welds, there seems to be a minor reduction in the mean value (5.4% and 2.6% respectively) when the alloys are exposed. However, the variability (spread) within each series is in the same range of magnitude as the increase; therefore, it is not possible to confirm that the exposure had a clear influence on the resulting hardness.

3 | RESULTS

3.1 | Hardness of the overlay welds before and after the exposure

Figure 3 shows the range (min, max) and the average hardness for each overlay welded alloy in the as-deposited condition and the exposed condition after being part of the tube shields for 6 months. Based on these results, the overlay welds can be ordered from higher to lower hardness as follows: APMT > Alloy 625 > EF101 > EF100.

3.2 | Microstructural characterisation of the FeCrAl overlay welds

The welding settings established for the FeCrAl alloys (Table 2) resulted in overlay welds with an average layer thickness of 3 mm (Figure 4). The three FeCrAl alloys showed long columnar ferritic grains following the direction of heat dissipation (Figure 5). Different types of micro-nano precipitates were found in the grain boundaries, but also intragranularly in these alloys. EF101 presented REs in the grain boundaries (Figure 6). Similarly, the grain boundaries of deposited EF100 also showed a concentration of several REs (Figure 7). In APMT, the exposure resulted in a noticeable Cr enrichment in the grain boundaries (Figure 8). After the exposure, it was possible to observe that the grain boundaries were preferential locations for the progression of the corrosion in the three FeCrAl alloys (Figure 9).

3.3 | Metal-loss rate

The visual inspection of the exposed tube shield samples revealed two preferential zones to the erosion–corrosion phenomena. Considering the cross-section of the tube shield as a 180° circular sector, one preferential location for corrosion was found at 0° and 180° (symmetric and
considered as only one location), at the interface between the overlay clad and the substrate. The second sensitive location was found at the area around 90°, where the exposed surface was more irregular and where lower remaining thickness was observed. Both characteristics are presented in Figure 10.

Figure 11 shows examples of cross-sections of the four exposed alloys. A small portion of tube shield #7 (EF100) exhibits the loss of overlay weld down to the substrate, approximately at the 90° position according to Figure 10. The encircled areas correspond to the remaining overlay weld after exposure. Figure 12 shows the metal-loss rate for the overlay welded tube shields determined based on the encircled areas shown in Figure 11. Specimens exhibiting local losses of overlay weld down to the substrate are marked with an asterisk in Figure 12. In general, the tube shields located on the roof exhibited a higher metal-loss rate than the ones located on the floor. Alloy 625 exposed on the floor location exhibited very low metal-loss rates throughout the whole tube, while alloys EF101 and EF100 exhibited enhanced metal-loss rates in the middle-length samples. APMT alloy showed a relatively consistent metal-loss rate in both roof and floor positions, as well as overall across the edge and middle sections of the tube shield.
Considering the maximum metal-loss rate values obtained the alloys have been ranked from more to less erosion–corrosion resistant as follows: for roof position APMT > Alloy 625 > EF101 > EF100, and for floor position Alloy 625 > APMT > EF101 > EF100.

3.4 Characterisation of the corrosion attack

SEM-EDS analyses were performed on edge and middle specimens for all alloys to characterise the corrosion attack regarding its composition and morphology. Analysed deposits generally contained the same species in all examined specimens (Na, Cl, Ca, K, S and O). However, the corrosion morphology varied significantly between alloys, exhibiting intergranular corrosion and the formation of pits to different degrees.

The EF100 specimens exhibited a distinctive corrosion attack, which can be interpreted as intergranular corrosion. The corrosion products are distributed in a layered structure, exhibiting a slight enrichment of Cr and Al closer to the metal, covered by an iron-based oxide. The EDS analysis (Figure 13) identifies Cl in the same interface, suggesting the formation of metallic chlorides. Even though EF100 exhibited a higher metal-loss rate in the roof position (tube #21), the corrosion attack followed the same pattern in all exposed specimens.

The corrosion attack on EF101 in the floor location was characterised by pit formation. The EDS analysis from Figure 14 suggests the formation of both Cr- and Al-rich oxide layers enveloping the pits, with a Fe-rich oxide covering it. Additionally, a thin Si-based oxide layer was observed close to the metal-oxide interface. However, in localised areas the Cl signal follows this corrosion layer

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**FIGURE 6** Scanning electron microscopy image of Kanthal® EF101 as deposited, showing reactive element enrichment in the grain boundaries (bright in the image).

**FIGURE 7** Scanning electron microscopy image of Kanthal® EF100 exposed overlay, showing reactive element located inter- and intragranularly (bright in the image).

**FIGURE 8** Energy-dispersive spectroscopy showing chromium enrichment in the grain boundaries of Kanthal® APMT after the exposure. Dark and bright precipitates in the scanning electron microscopy image correspond to reactive elements. [Color figure can be viewed at wileyonlinelibrary.com]
boundary, indicating the potential formation of metal chlorides. EF101 in the roof position (#13) experienced a much more severe attack. It is worth noting that all specimens corresponding to tube shield #13 suffered a major metal loss.

APMT generally exhibited a more homogeneous form of corrosion attack. However, in some regions, there was evidence of grain boundary attack. The SEM image in Figure 15 shows an incipient corrosion attack at the grain boundary. The bright particles depicted in the image correspond to RE elements. The accompanying EDS analysis identified both the Cr- and Al-rich oxide scale and Fe to some extent. The Cl signal was relatively weak, but it extended deep into the corrosion attack, overlapping with metal signals, thus suggesting the formation of metal chlorides.

Alloy 625 exposed on the floor position, exhibited a limited corrosion attack that was characterised by a layered and relatively dense structure, as seen in Figure 16. The EDS analysis shows a chromium-rich oxide layer across the metallic interface, in connection with weak signals of chlorine and calcium, followed by a nickel-rich oxide. It was observed that nickel migrated from the overlay weld and became mixed with the deposit.

4 | DISCUSSION

4.1 | Influence of the location in the evaporator chamber on the metal-loss rate

The results in Section 3.3 show considerable differences in metal-loss rate by the same material in different positions or orientations (Figure 12). This confirms that the environment in the evaporator chamber is not homogeneous, as it is also supported by previously published results from the authors.[16]

There was a clear trend when comparing a specific alloy. The overlay welds exposed in the roof position experienced substantially higher erosion–corrosion rates than the ones exposed on the floor. The only exception found was in the overlay weld of EF100 located in tube #21 in the roof, showing a similar metal-loss rate to the same alloy located on the floor in tube #7. The explanation could be the lower efficiency of the soot blowers in the tubes that are far from the chamber sides from where the soot blowers are traversing towards the boiler's centre. Each soot blower should clean the deposits of around 25 tubes, so tube #21 is quite close to the central area of the chamber. An inefficient soot

**FIGURE 9** Overlay welds after exposure. The grain boundaries seem to be preferential locations for the progression of corrosion.
blowing could cause an accumulation of deposits in the tube that would have partially acted as a self-protector for the overlay weld. Additionally, the overall visual inspection of the tube shields revealed that the environmental conditions on the 90° exposed surface were more aggressive than in other locations of the exposed surface, as described in Section 3.3 and Figure 10.

In this type of boiler, the flue gas stream passes through the superheater banks vertically from the bottom to the top just after a 180° turn (Figure 1). Before the superheater...
FIGURE 12  Metal-loss rate (erosion–corrosion rate) for overlayed tube shields (roof and floor positions). The specimens were cut from the tube shields at positions R, F and C. The edge specimens R and F correspond to the Rear and Front sides of the boiler system, respectively, whilst C corresponds to the mid-length of the tube shield. Specimens exhibiting local losses of overlay weld down to the substrate are marked with an asterisk. [Color figure can be viewed at wileyonlinelibrary.com]

FIGURE 13  Scanning electron microscopy-energy-dispersive spectroscopy analysis of edge section of tube #7. Intergranular corrosion is observed in this specimen. [Color figure can be viewed at wileyonlinelibrary.com]
channel, the flue gas speed is approximately 8 m/s, and because of the presence of tubes in the channel, the free space for the gas circulation is reduced and therefore, the speed is expected to increase. The tube shields exposed in the roof of the chamber are the ones experiencing the direct impact of any solid particles carried by the flue gas. Dixon et al.\cite{22} used CFD to support the experimental erosion observed in the convection tube banks of the superheater in a Bagasse-fired boiler. The gas flow dynamics showed that in some orientations, the ash particle streams were concentrated and impacted on tubes at specific locations, whilst in other areas the density of particles and the impact angle were lower. The same principle is likely to apply to this CFB boiler and might support the observations in this work. The impact of ash particles and small amounts of entrained quartz sand bed particles would preferentially influence the tube shields placed on the roof position. Another element that could have influenced the heterogeneity of the

**FIGURE 14** Scanning electron microscopy-energy-dispersive spectroscopy analysis of centre section of tube #8. The corrosion attack is characterised by the formation of pits filled with a layered oxide structure, including Si-containing oxides. [Color figure can be viewed at wileyonlinelibrary.com]

**FIGURE 15** Scanning electron microscopy-energy-dispersive spectroscopy analysis of edge section of tube #9. A type of grain boundary-related attack is observed in this specimen. The bright spots correspond to reactive elements. [Color figure can be viewed at wileyonlinelibrary.com]
environment, as mentioned earlier, is the soot blowing. Probably it helped in removing the corrosion product preferentially in the mid-length of the tube shields, and consequently leaving a fresh metal surface exposed, which, in turn, would decrease the self-protection of the FeCrAl alloys against corrosion locally. However, this influence would be expected on both roof and floor tube shields. Moreover, flue gas temperature variations could affect the adherence of deposits by the difference in thermal expansion coefficients, in comparison with the substrate. Nonetheless, this effect is believed minor since the variation in temperature of the tube shields is considerably lower than the variation of the flue gas temperature. This is due to the temperature of the tube shields being primarily controlled by the constant temperature of the evaporator tubes.

As two of the FeCrAl alloys in this study were recently developed, not much data has been published on their performance in waste-fired boilers. However, a previous study\[23\] was conducted in the superheater of the P14 boiler in Händelö using tube shields of 253 MA, Hastelloy G-30, Haynes HR-160, 316L, 304L, and commercial FeCrAl alloys (Kanthal® APMT, Kanthal® APM, and Kanthal® A-1) in as-delivered condition, for 321 and 492 days. When comparing the average performance of some alloys in that former study (e.g., 0.6 mm/year for APMT, 1 mm/year for 253 MA, 1.7 mm/year for 316 L or 3.3 mm/year for Haynes HR-160) with the metal-loss rates of APMT, EF101 and EF100 obtained during the present 6-month exposure (Figure 12), the performance of these alloys is promising. However, those tube shields were not overlay welded, and they were not exposed at the same tube banks; therefore, any comparison needs to be done with caution.

4.2 Influence of alloy composition on the corrosion attack

Alloy EF100 showed extensive grain boundary attack, while the corrosion attack of EF101 having a higher Si content was characterised by pit formation. A similar behaviour was reported in a long-term laboratory exposure, showing a grain boundary attack in an EF100 welded specimen compared to a more homogeneous form of internal oxide growth exhibited by the welded specimen EF101.\[24\] However, in the same study, a nonwelded EF100 specimen exhibited a layered inward/outward corrosion structure and no grain boundary attack. Moreover, a recent field study with nonwelded EF100 specimens reported only a layered corrosion attack after 12 months of exposure and no intergranular corrosion.\[25\] These observations suggest that the corrosion response of the FeCrAl alloys studied here can be influenced by welding and that it can differ from the response of the same alloy in nonwelded conditions.

The silicon content difference between alloys EF100 and EF101 (0.3 and 1.3 wt%, respectively) was expected to influence the metal loss to some extent. The SEM-EDS investigation revealed different types of corrosion attacks in both alloys, which is suggested to be related to the silicon content in the alloys. The increased Si content in EF101, compared to EF100, promoted the formation of an Al-Si-containing oxide layer, which might have mitigated the progression of an intergranular corrosion attack, such as the one exhibited by EF100. The beneficial effect of Si in model FeCrAl alloys has been previously studied,\[26–28\] indicating that an increase in Si
content can enhance the corrosion resistance of the alloy by different mechanisms. One of the mechanisms is the enhancement of the aluminium-containing scale that has been associated with the synergy between these two alloying elements. Another possible mechanism is the formation of a dense oxide layer that prevents chromium depletion and additionally hinders chlorine diffusion into the alloy. The metal-loss difference between alloys, however, was relatively small even in the more homogeneous floor position. Thus, the silicon content difference might be responsible for the type of corrosion attack, but the overall effect in the metal loss was not as evident. The complex and highly corrosive-erosive environment inside the boiler might also be responsible for the overall limited effect of silicon in EF101. Indeed, other degradation mechanisms could take prevalence, namely the erosive effect of the entrained particles and the high concentrations of alkali chlorides present in the flue gas.

The erosion–corrosion rate of APMT was the most consistent across positions within the tube (back, centre and front). The corrosion attack had a sparse number of pits in comparison to the other tested FeCrAl alloys, which might be attributed to the comparatively higher chromium content and the presence of molybdenum in APMT.

The overlay welds of Alloy 625 showed considerable variability in the metal loss across the two positions roof and floor, which could be an indication of inhomogeneities inside the boiler. However, the metal-loss rate at the floor position was on average the lowest of all assessed specimens (about 0.2 mm/year). The corrosion products in this alloy were mainly composed of nickel oxide mixed with deposits. These results agree with a field exposure at 600°C by Eklund et al.\[^{26}\] However, they reported the presence of metal chlorides underneath the scale, which was not clearly observed in the present study.

### 4.3 Influence of the Cr content, microstructure and hardness on the metal-loss rate

The hardness results of the nonexposed and exposed EF100, EF101, APMT and Alloy 625 overlay welds (Figure 3) confirm that EF101 and Alloy 625 are not clearly affected by the temperature and exposure conditions in P14, whilst EF100 experiences softening and APMT experiences hardening due to the exposure.

APMT and Alloy 625 showed the highest hardness values and interestingly these are the alloys that presented lower metal-loss rates during the tube shielding exposure. That could be an indication of the importance of the erosion mechanism in comparison to the corrosion mechanism on the tube shields.

The evolution of hardness in EF100, EF101 and APMT during exposure is considered to be mainly the result of two competing processes: softening through annealing recovery and hardening through phase separation (of the ferritic matrix into Fe-rich $\alpha$-phase and a Cr-rich $\alpha\prime$-phase). This is commonly known as 475°C embrittlement and it is characteristic of ferritic stainless steels with high Cr content.\[^{29–31}\] The phenomenon is well-known in conventional FeCrAl alloys containing around 20% Cr, such as APMT. EF100 and EF101 contain lower Cr in their chemical composition (Table 1), and that can be the reason why hardening was not clearly observed in the overlay welds of these two alloys but observed in APMT instead.

The hardness change would reflect the combined effect of the two mechanisms mentioned: annealing and phase separation (ferrite $\rightarrow \alpha + \alpha\prime$). Therefore, the softening in EF100 can be explained by the annealing mechanism, whilst for EF101, staying at roughly the same hardness level implies that the competing mechanisms of annealing and hardening through phase separation would have been comparable in their predominance. In APMT, the higher Cr allows for a higher degree of hardening from phase separation, which would dominate over the effect of annealing and result in net hardening. For Alloy 625, the annealing over time is expected to be slower at the exposure conditions, due to the lower self-diffusivity in Ni-based alloys.

The grain boundaries enriched in REs were found to be preferential locations for corrosion to propagate internally in the microstructure. It is well known that REs with high carbon affinity are added to prevent local depletion of Cr due to an eventual formation of Cr$_2$C$_6$ and therefore to form RE carbides to avoid intergranular corrosion. However, in these alloys, the grain boundaries with RE carbides have been preferentially attacked. More investigation would be necessary to explain this phenomenon.

Combining the maximum determined metal-loss rates with the amount of local corrosion attack (intergranular or pitting), the performance of the overlay welds in the evaporator tube bank can be ranked as follows: APMT and Alloy 625 showed the best and very similar behaviour, whilst EF101 performed better than EF100.

### 5 CONCLUSIONS

Overlay welded tube shields of three FeCrAl alloys from Kanthal\textsuperscript{®} and reference Alloy 625 were placed in the evaporator bank of a waste-fired power plant and exposed for one operational cycle of 6 months. The objective of the exposure was to evaluate the metal loss in
the overlay welds and to analyse the corrosion products resulting from the exposure to the harsh environment, namely the effect of corrosion and erosion phenomena at high temperatures. The following conclusions were drawn:

- The erosion–corrosion environment in the evaporator bank is heterogeneous. The specimens placed in the roof position consistently showed a higher metal-loss rate than the specimens exposed on the floor, most likely due to the combined effect of the soot blower and bottom-to-up direction in the flue gas circulation designed for the CFB boiler. The entrained solid particles in the flue gas could have changed the erosion–corrosion regime of the roof tubes, enhancing the degradation synergy.

- The good erosion–corrosion performance of APMT can be explained by the hardening during the exposure, increasing its resistance to erosion. This alloy shows primarily a homogeneous form of corrosion attack in the presence of a Cr- and Al-rich oxide scale, and that can explain the good corrosion performance observed. However, in some regions, evidence of intergranular corrosion attack was found.

- The lower erosion–corrosion resistance of EF100 can be explained by two factors. On the one hand, the low erosion resistance can be caused by the softening occurring at the exposure conditions. On the other hand, the severe intergranular corrosion attack observed might be explained by the formation of base metal chlorides close to the metal surface.

More sustainable production of heat and power demands an effort to investigate promising alloys such as FeCrAl. With this work, we have shed some light on the challenges of the environmental heterogeneity in the waste-fired plants, and we have graded the performance of different FeCrAl alloys under that harsh environment. The use of computational fluid dynamics and the design of new FeCrAl alloys based on the knowledge obtained in this work could be promising paths to continue with the mission.

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DATA AVAILABILITY STATEMENT
The data that support the findings of this study are available from the corresponding author upon reasonable request.

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