High Temperature Oxidation Behaviors of Alloy 617 and Haynes 230 in Impurity Controlled Helium Environments

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Abstract  The high temperature oxidation behaviors of alloy 617 and Haynes 230 have been investigated for VHTR intermediate heat exchanger applications. Oxidation tests were carried out for up to 500 hours at 900 °C and 1000 °C in impure helium environments containing H₂, H₂O, CO, CO₂, and CH₄. The oxidation kinetics of the alloys followed a parabolic law of oxidation in all cases. In the impure helium environments with very low oxygen, the external oxides of alloy 617 were composed of a Cr₂O₃ layer, TiO₂ ridges on the grain boundaries, and isolated MnCr₂O₄ grains on top of the Cr₂O₃ layer. On the other hand, those of Haynes 230 consisted of a Cr₂O₃ inner layer and a protective MnCr₂O₄ outer layer, which increased the oxidation resistance. The effect of small amounts of CH₄ and H₂ on the oxidation kinetics of the alloys was insignificant. Irregular oxide morphology, such as cellular Cr₂O₃ oxides for alloy 617 and MnCr₂O₄ platelets for Haynes 230, was formed in the impure helium environment at 900 °C. For Haynes 230, along with platelets, whiskers were frequently found at the tip of the MnCr₂O₄ oxide crystals.

Keywords  Oxidation: Nickel-base superalloys; VHTR; Impure helium; Alloy 617; Haynes 230
Introduction

A very high temperature gas-cooled reactor (VHTR) is being developed with the aim of attaining higher efficiency and for potential application to massive hydrogen production. The VHTR will be operated at a high temperature of up to 1000 °C and in a corrosive helium coolant containing small amounts of H\textsubscript{2}, CO, CO\textsubscript{2}, H\textsubscript{2}O, and CH\textsubscript{4}[1]. These extreme operating conditions will likely necessitate the use of metallic components comprised of better temperature resistant alloys. Taking this into account, nickel-based superalloys—alloy 617 and Haynes 230—are being considered as constituent materials for the reactor. Upon exposure to high temperature helium environments, alloys are carburized or oxidized/decarburized depending on the gas compositions and temperature, which significantly degrades their mechanical properties [2-7]. Therefore, the formation of protective and tenacious oxide layers and their performance in impure helium environments at high temperatures are among the most important factors to be considered in reducing material damage by gaseous impurities.

Based on experimental results and the operating experience of previous high temperature helium-cooled reactors, a slight oxidizing condition was recommended as the optimum helium coolant chemistry of the VHTR in terms of material stability [8]. Slight oxidizing and decarburizing conditions can generally be achieved in the impure helium environments that contained 0.5-5 µbar H\textsubscript{2}O, below 50 µbar CO, and below 20 µbar CH\textsubscript{4}[9-12].

As the first step toward understanding the oxidation behaviors of alloy 617 and Haynes 230 in the anticipated VHTR environment, we studied the general oxidation behaviors of these alloys in air. Characterization and stability of oxide layers and the role of alloying elements at 900°C and 1100°C have previously been discussed [13, 14]. The present study focuses on the properties of the oxide layer of these alloys in slight oxidizing, impure helium environments containing H\textsubscript{2}, H\textsubscript{2}O, CO, CO\textsubscript{2}, and CH\textsubscript{4}. The effects of temperature, alloying elements, and helium gas chemistries on the oxide properties were evaluated and then compared to the corresponding properties in air.

Materials and Experimental Procedure
Two commercial nickel-based superalloys, alloy 617 and Haynes 230, containing high Cr content (about 22 wt% Cr) were used in this study. Table 1 provides the chemical compositions of the materials, which were analyzed by inductively coupled plasma mass spectroscopy (ICP-MS) at the Korea Testing and Research Institute.

Coupon type specimens with dimensions of 12 mm in diameter and 1 mm in thickness were used. The specimens were oxidized for up to 500 hours in air and impure helium environments at 900 °C and 1000 °C. A helium gas loop system was used to control the gas chemistries and to maintain the gas flow rate at 500 cc/min and a pressure of 110 kPa. The impure helium environments and test temperatures are listed in Table 2.

After the oxidation tests, the weight gain was measured. The properties of the oxide layers were investigated by X-ray diffraction (XRD), scanning electron microscopy (SEM), energy dispersive X-ray spectroscopy (EDS), and an electron probe micro-analyzer (EPMA) analysis.

Results

Oxidation kinetics

Fig. 1 shows weight gain vs. time ($t^{1/2}$) plots of alloy 617 and Haynes 230 after isothermal oxidation at 900 °C and 1000 °C up to 500 hours in air and impure helium environments. It is clear from the figure that the weight gain of alloy 617 was considerably higher than that of Haynes 230 in all environments. The plots displayed linear relationships in air, following a parabolic rate law,

$$\Delta W / S = \sqrt{k_p t}$$  \hspace{1cm} (1)

where $\Delta W$ is the change in weight, $S$ is the surface area of the specimen, $k_p$ is the parabolic rate constant, and $t$ is the oxidation time in hours.

For alloy 617, the oxidation kinetics in air and impure helium environments (He-K1 and He-K2) approximately followed a parabolic growth law at both temperatures. As temperature was increased from 900 °C to 1000 °C, the oxidation rate was also increased, but the effect of the test environment was negligible. On the other hand, the oxidation kinetics of Haynes 230 was different at the two temperatures. At 900 °C, a change of oxidation kinetics was observed after exposure of about 200
hours to air, but this change in oxidation kinetics was not observed in the impure helium environments. At 1000 °C, the transitional behavior was not observed in any of the test environments. In contrast with the case of alloy 617, the oxidation kinetics of Haynes 230 was lower in the impure helium environments compared to that in air.

**XRD analysis**

Fig. 2 shows the XRD results of alloy 617 and Haynes 230 after oxidization for 500 hours at 900 °C and 1000 °C in He-K1 and He-K2. Diffraction peaks for the underlying metal substrate were found in all cases, even though the angle of the incident X-ray beam of thin-film XRD was fixed at 2°. This indicates that X-rays completely penetrated the oxide layer, minimizing the possibility of undetected oxide phases. According to a previous study [13], alloy 617 had NiO, NiCr2O4, and Cr2O3 while Haynes 230 had an external oxide layer comprising MnCr2O4 and Cr2O3 after oxidization at 900 °C in air. In this study, the XRD results for alloys oxidized at 900°C and 1000°C in He-K1 and He-K2 revealed that alloy 617 had TiO2, MnCr2O4, and Cr2O3 oxides on the surface, and Haynes 230 had MnCr2O4 and Cr2O3. The ratio of the peak intensity of MnCr2O4 to that of Cr2O3 was relatively low in alloy 617 compared to Haynes 230 alloy.

**Discussion**

**Effects of alloying elements on oxide structure**

It is known that Cr2O3 is formed as an external oxide layer on the surface of Ni-Cr alloys, rather than as an internal oxide, at 1000 °C when Cr content is above 10 wt% [15]. A previous work revealed that the external oxide layer of alloy 617 and Haynes 230 with 22 wt% Cr is mainly composed of Cr2O3 after oxidation of the alloys in air. Al in both alloys caused internal oxidation of Al2O3 below the external oxide layer, because an oxide layer could not be formed at 900 °C thermodynamically with Al content below 1.5 wt% [13]. In the impure helium environments, the Cr2O3 peaks of alloy 617 were much greater than those of MnCr2O4 and TiO2, as shown in Fig. 2. The Al2O3 internal oxide was not detected, despite considerable formation for alloy 617, because the X-ray did not penetrate the internal...
oxidation region. On the other hand, the 0.46 wt% Mn contained in Haynes 230 contributed to the formation of MnCr$_2$O$_4$. MnCr$_2$O$_4$ peak intensities were much greater than those of Cr$_2$O$_3$ for Haynes 230.

Fig. 3 shows the typical morphology of top surface oxides formed on alloy 617 and Haynes 230 oxidized for 500 hours. The same macroscopic surface morphology was also observed for all specimens oxidized under different temperatures and environments. Oxidized alloy 617 is mainly covered with Cr$_2$O$_3$, and the evolution of ridges is observable on the oxide surface. The formation of ridges generally occurs along the grain boundary of the metal substrate below the oxide layer, the so-called grain boundary ridge [16]. On the other hand, Haynes 230 is completely covered with MnCr$_2$O$_4$ as an outermost layer, thus indicating that Cr$_2$O$_3$ underlies the MnCr$_2$O$_4$ layer. No grain boundary ridge was observed for Haynes 230.

The EDS results of alloy 617 are shown in Fig. 4. It is clear that Ti was concentrated only on the grain boundary ridges, although Cr was also detected on the grain boundary ridges of alloy 617, as shown in Fig. 4(a). Based on the XRD analysis results, Ti existed on the ridges as an oxide form, i.e., as TiO$_2$. In order to determine the cause of the formation of grain boundary ridges, the Ti concentration was also measured in a cross-section of the oxidized specimen, as shown in Fig. 4(b). Ti was not observed within the grains of the metal, but it was shown to be concentrated on the Al$_2$O$_3$ internal oxide formed along the grain boundaries and the Cr$_2$O$_3$ oxide layer. This indicates that Ti diffused from the grain lattices to the grain boundaries, and the diffused Ti was subsequently transported to the specimen surface through the grain boundaries.

The images in Fig. 5 sequentially show the continuous growth of ridges at the grain boundary on the oxide surface as oxidation progresses. The average width of the grain boundary ridges is approximately 4 µm after 24 hours, 8 µm after 100 hours, and 20 µm after 500 hours. The rate of increase of the average width of the grain boundary ridges also follows a parabolic function up to 500 hours. Until Ti depletion, which results from limited Ti content at the grain boundaries below the oxide layer, the growth of grain boundary ridges is considered to be in progress.
Spallation of the grain boundary ridges is infrequently observed, as shown in Fig. 6. In the spallation region (Fig. 6(a)), a dense Cr$_2$O$_3$ layer with no pores is observed below the irregular oxide surface near the ridges. A metal substrate is found below the grain boundary ridges (Fig. 6(b)), and it is notable that a gap is formed between the grains of the metal substrate. Such spallation and the formation of the gap might occur during the cooling process due to different thermal expansion of the metal and the oxide layer. It is thought that inter-granular oxides formed along the grain boundaries are spalled out together with the grain boundary ridges.

Mn in Haynes 230 forms MnCr$_2$O$_4$ as the outer layer of a double layer, because Mn ions generally diffuse two orders of magnitude faster than Cr ions in a Cr$_2$O$_3$ layer [17, 18]. As oxidation progresses, Mn is gradually consumed from the surface with the formation of MnCr$_2$O$_4$. Since the Haynes 230 alloy used in this study contains only 0.46 wt% Mn, Mn is eventually depleted below the external oxide. As shown in Fig. 7(a), a very small amount of Mn is detected just below the interface between the oxide and metal. As the Mn depletion zone develops slowly, the supply of Mn through the Cr$_2$O$_3$ inner layer is reduced, and eventually the growth of MnCr$_2$O$_4$ is limited by the availability of Mn. The MnCr$_2$O$_4$ outer layer thereafter acts as a barrier to oxygen diffusion into the metal surface, resulting in a change in the oxidation kinetics and good oxidation resistance of Haynes 230. On the other hand, as shown in Fig. 7(b), when Haynes 230 is oxidized in impure helium environments at 1000 °C, the Mn depletion zone is continuously developed and the Mn content below the surface remains high at 500 hours. Therefore, changes in the oxidation kinetics were not observed during exposure for 500 hours to He-K1 and He-K2.

**Effects of gaseous impurities on oxidation**

The main oxidizing species in air is oxygen, whereas water vapor and CO$_2$ are the main species in the present impure helium environments. In air, it is reported that the oxide scale is composed of Cr$_2$O$_3$, NiO, and NiCr$_2$O$_4$ for alloy 617 at 900 °C, and Cr$_2$O$_3$ and MnCr$_2$O$_4$ for Haynes 230 [13]. On the other hand, with low oxygen activities of He-K1 and He-K2, both alloy 617 and Haynes 230 have an oxide layer of Cr$_2$O$_3$ and MnCr$_2$O$_4$, as shown in Fig. 2.
The CO/CO\(_2\) ratio and the partial pressure of water vapor are fixed in He-K1 and He-K2, and oxygen partial pressure can be determined by the following reactions:

\[
\begin{align*}
\text{CO}_2 &= \text{CO} + \frac{1}{2}\text{O}_2 \\
\text{H}_2\text{O} &= \text{H}_2 + \frac{1}{2}\text{O}_2
\end{align*}
\]  

(2)  

(3)

\[
P^\text{eq}_{\text{O}_2} = \left( \frac{P_{\text{CO}_2}}{P_{\text{CO}}} \right)^2 \exp \left( \frac{-2\Delta G^\text{a}_{\text{CO}_2}}{RT} \right) = \left( \frac{P_{\text{H}_2\text{O}}}{P_{\text{H}_2}} \right)^2 \exp \left( \frac{-2\Delta G^\text{a}_{\text{H}_2\text{O}}}{RT} \right).
\]

(4)

Assuming gas reactions are thermodynamically in equilibrium, oxygen partial pressure in He-K2 is obtained as follows: \(P^\text{eq}_{\text{O}_2} = 3.37 \times 10^{-18}\) at 900\(^\circ\)C and \(P^\text{eq}_{\text{O}_2} = 3.10 \times 10^{-16}\) at 1000\(^\circ\)C. In He-K1, CH\(_4\) supplies additional carbon to the metal surface by the following reaction:

\[
\text{CH}_4 = \text{C} + 2\text{H}_2.
\]

(5)

Some water vapor reacts with the carbon in the metal surface by the following reaction:

\[
\text{C} + \text{H}_2\text{O} = \text{CO} + \text{H}_2.
\]

(6)

Consequently, oxygen partial pressure can be written as follows:

\[
\left( P_{\text{O}_2} \right)^{1/2} = \left( P^\text{eq}_{\text{O}_2} \right)^{1/2} \left( 1 - \frac{k_5 P_{\text{CH}_4}}{k_6 P_{\text{H}_2\text{O}}} \right)
\]

(7)

where \(k_5\) and \(k_6\) are the reaction rate constants for eq.(5) and eq.(6), respectively. Oxygen partial pressure can be lowered by the second term in the bracket. When \(k_5/k_6\) is small as the decomposition kinetics of CH\(_4\) is much slower than that of H\(_2\)O, then P\(_{\text{O}_2}\) in He-K1 containing CH\(_4\) is of the same order as He-K2 without CH\(_4\). Quadakkers and Schuster [19] revealed that CH\(_4\) does not have a significant effect on oxygen partial pressure for alloy 617, as long as the CH\(_4\)/H\(_2\)O ratio is not overly high. Moreover, the effect is much smaller under flowing gas. They reported that there was no difference in oxidation behaviors at 950\(^\circ\)C when the alloy was oxidized in impure helium with or without 20 \(\mu\)bar CH\(_4\). Therefore, it is expected that CH\(_4\) within a range of 15-20 \(\mu\)bar, as contained in He-K1, has little impact on oxygen partial pressures. Accordingly, the difference in oxidation kinetics was negligible between He-K1 and He-K2, as shown in Fig. 1.
Notably, low oxygen partial pressure retarded the formation of Ni-rich oxides of alloy 617 that were formed in air. According to the Ellingham diagram, the oxygen partial pressure for the formation of NiO by the reaction between Ni and O is between $10^{-10}$ and $10^{-12}$ in a temperature range of 900°C to 1000°C in equilibrium. NiO that formed in air for the case of alloy 617 is no longer stable at the very low oxygen partial pressures of He-K1 and He-K2, and subsequently NiCr$_2$O$_4$ is not formed due to the absence of NiO. On the other hand, the Gibbs free energy of the formation of MnCr$_2$O$_4$ is of the same order as Cr$_2$O$_3$ in a temperature range of 900 °C and 1000 °C [20], and is low enough that MnCr$_2$O$_4$ exists as an outer layer under the oxygen partial pressures of He-K1 and He-K2. Therefore, no change of oxide type was observed for Haynes 230.

One of the important changes was decarburization depth depending on gaseous impurities. In spite of the negligible difference in oxidation kinetics, decarburization was much more extensive below the oxide layer of the alloys in the He-K1 environment, which contained CH$_4$ and H$_2$, than in He-K2. As a result, loss of ductility changed significantly for both alloys, as has been noted in other papers [7, 14].

Surface morphology was also influenced by helium chemistries. Specifically, cellular oxides formed on alloy 617 and platelets were observed in Haynes 230 in He-K1 and He-K2 at 900 °C, as shown in Figs. 8(a) – 8(d). The formation of an irregular oxide morphology has previously been reported when water vapor was present in the environments [21-23] and was also observed in cases where gas mixtures contained CO$_2$ [24]. The formation of these oxides with irregular morphology is attributed to more rapid dissociation of water vapor and CO$_2$ molecules compared to oxygen molecules [25, 26]. Similarly, in hot steam environments containing H$_2$, extensive formation of platelet oxides was observed in Haynes 230 as well as in alloy 617 at an early oxidation stage due to accelerated diffusion of cations under the oxide layer [27]. At high oxygen activities, molecular oxygen adsorbs preferentially on the entire surface, which reduces local variation of surface morphology caused by catalysis of H$_2$O and CO$_2$ dissociation [22]. As a result, grains on the oxide surface have a nodular structure after oxidation in air, as shown in Figs. 8(e) and 8(f).

However, irregular morphology for nickel-based superalloys is not typical, and such morphology is not observed in similar impure helium environments with H$_2$O in a range of 1 to 6 µbar at very low
CO$_2$ at 950 °C [28]. This indicates that such morphology cannot be obtained with very low H$_2$O content. The addition of CO$_2$ to He-K1 and He-K2 might have significantly contributed to the oxide growth with irregular morphology. Irregular oxide morphology is generally related to dislocations in the oxide, as these regions are more active than other areas. Subsequently, whiskers grow rapidly, because the tip of the whisker is catalytically active and promotes dissociation of the oxidants [21, 22, 26, 29]. However, the sequential images in Fig. 9 show that whiskers are preferentially grown at the tip of the MnCr$_2$O$_4$ oxide crystals rather than at dislocations for Haynes 230; in particular, they are more active catalytically at the vertex of nodular oxide crystals. Droplets at whisker tips can be observed, and are generally thought to be evidence of a vapor-liquid-solid (VLS) mechanism for whisker formation [30, 31]. Droplets on the tips of the whiskers might be consumed as the whiskers grow into platelets. After 500 hours, as shown in Fig. 9(c), droplets are infrequently observed and the formation of platelets is insignificant.

**Effect of oxide structure on oxidation resistance**

The parabolic oxidation of the alloys is attributed to the rate-limiting process for oxide growth being diffusion-controlled as the oxide thickens. Although the growth of whiskers is controlled by a surface reaction, and thus linear kinetics will be observed, the formation of whiskers has no effect on oxidation kinetics, as shown in Fig. 1. As noted above, alloy 617 has Cr$_2$O$_3$ and MnCr$_2$O$_4$ oxides on its surface. The structure of Cr$_2$O$_3$ below the cellular oxides is found to be dense with few pores, as shown in Fig. 6, and there is no significant difference in the oxide morphology in a temperature range of 900 to 1000°C. However, MnCr$_2$O$_4$ oxide has an isolated nodular structure, and therefore the formation of MnCr$_2$O$_4$ provides little contribution to an increase of oxidation resistance for alloy 617. Consequently, the oxidation kinetics of alloy 617 was not significantly different in air and impure helium environments.

On the other hand, Haynes 230 was covered with a continuous MnCr$_2$O$_4$ layer in all cases, but the oxidation kinetics in He-K1 and He-K2 was slower than that in air. In an image showing the fractured oxide layer after bending deformation of the specimen oxidized in air (Fig. 10(a)), the MnCr$_2$O$_4$ grains
were arranged in a quasi-columnar configuration with grain sizes on the order of micrometers. This structure provides a high diffusion path for oxygen and metal ions, and a porous structure can also provide a high diffusion path for ions. On the other hand, the formation of whiskers is not observed at 1000°C in impure helium, as shown in Fig 10(b). However, it is notable that the grain size of MnCr₂O₄ with faceted grain morphology is smaller and its structure is denser. Consequently, the formation of this dense MnCr₂O₄ structure yielded better oxidation resistance in the case of Haynes 230 in the impure helium environments.

Conclusions

High temperature oxidation behaviors of nickel-based superalloys, alloy 617 and Haynes 230, were studied for VHTR intermediate heat exchanger applications. Oxidation tests were carried out for up to 500 hours in impure helium environments, He-H₂O-CO-CO₂ and He-H₂O-CO-CO₂-CH₄, at 900°C and 1000°C. Based on the results of oxidation tests and subsequent analyses, the following conclusions were drawn.

1. Parabolic oxidation kinetics was observed for all tests performed in the impure helium environments at 900 °C and 1000 °C. However, the formation of a MnCr₂O₄ outer layer accompanying Mn depletion was an important factor in increasing the oxidation resistance of Haynes 230, which resulted in slower oxidation kinetics.

2. NiO and NiCr₂O₄, which were stable in air, were not formed in the impure helium environments due to very low oxygen activities. Instead, a Cr₂O₃ layer, TiO₂ ridges on the grain boundaries, and isolated MnCr₂O₄ grains on top of the Cr₂O₃ layer were formed for alloy 617, while a Cr₂O₃ inner layer and a MnCr₂O₄ outer layer were formed for Haynes 230.

3. At very low oxygen activities, H₂O and CO₂ promoted the formation of cellular Cr₂O₃ on alloy 617 and whisker/platelet MnCr₂O₄ on Haynes 230. At high oxygen activities, on the other hand, grains on the oxide surface have a nodular structure, because molecular oxygen adsorbs preferentially on the entire surface, which reduces local variation of surface morphology caused by H₂O and CO₂ dissociation.
4. Formation of MnCr$_2$O$_4$ whiskers might be related to a VLS mechanism on the basis of the appearance of droplets on the tips of the whiskers, the formation of which was preferential on the vertex of MnCr$_2$O$_4$ crystals.

5. A dense MnCr$_2$O$_4$ oxide layer was grown in impure helium environments for Haynes 230, while quasi-columnar MnCr$_2$O$_4$ grains were grown in air. The formation of the dense MnCr$_2$O$_4$ layer resulted in better oxidation resistance in impure helium environments.

Acknowledgement

This work was supported by the MEST / NRF (Nuclear R&D Program, 2005-2004718 and 2009-0083392) of Korea. Part of the funding was provided by the Second Phase BK21 Program of the Ministry of Education, Science and Technology of Korea.

References

Table Captions

Table 1 Chemical composition of alloy 617 and Haynes 230 (in wt%)

Table 2 Composition of impurities in the helium gases (in µbar)
**Figure Captions**

**Fig. 1** Weight change vs. time plots of alloy 617 and Haynes 230 in air and impure helium environments: (a) at 900 °C, (b) at 1000 °C

**Fig. 2** Thin-film XRD results of alloy 617 and Haynes 230 oxidized for 500 hours in impure helium environments: (a) at 900 °C in He-K1 (b) at 900 °C in He-K2, (c) at 1000 °C in He-K1, (d) at 1000 °C in He-K2

**Fig. 3** Surface morphology after oxidation for 500 hours at 900 °C in an He-K1 environment: (a) alloy 617, (b) Haynes 230

**Fig. 4** EDS analysis results showing (a) chemical compositions in oxide surface and (b) the Ti content on grain boundary and lattice of alloy 617 aged for 500 hours at 1000 °C in an He-K2 environment

**Fig. 5** Growth of grain boundary ridge of alloy 617 during oxidation at 900 °C in He-K2: (a) 24 hours, (b) 100 hours, (c) 500 hours

**Fig. 6** Spallation of grain boundary ridge of alloy 617 oxidized for 500 hours at 900 °C in the He-K2: spallation region (a) at low magnification, (b) at high magnification

**Fig. 7** EPMA analysis results showing depletion of Mn for Haynes 230: (a) after 1000 hours exposure to air at 1000 °C, (b) after 500 hours exposure to He-K1, He-K2 at 1000 °C

**Fig. 8** Surface morphology of alloy 617 and Haynes 230 after being aged for 500 hours at 900 °C: (a) alloy 617 in He-K1, (b) Haynes 230 in He-K1, (c) alloy 617 in He-K2, (d) Haynes 230 in He-K2, (e) alloy 617 in air, and (f) Haynes 230 in air

**Fig. 9** Surface morphology of Haynes 230 aged at 900 °C in He-K2: (a) after 24 hours, (b) after 100 hours, (c) after 500 hours

**Fig. 10** Oxide morphology of Haynes 230 oxidized for 500 hours at 1000 °C: (a) in air, (b) in He-K2
Table 1 Chemical composition of alloy 617 and Haynes 230 (in wt%)

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<th></th>
<th>Ni</th>
<th>Cr</th>
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<th>Mo</th>
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<th>Fe</th>
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<th>Mn</th>
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<td>.38</td>
<td>-</td>
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Table 2 Composition of impurities in the helium gases (in µbar)

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<th>CO</th>
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<tr>
<td>He-K1 at 1000°C</td>
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<tr>
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<tr>
<td>He-K2 at 1000 °C</td>
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<td>-</td>
<td>0.7</td>
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<td>10</td>
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Fig. 9 Surface morphology of Haynes 230 aged at 900 °C in He-K2: (a) after 24 hours, (b) after 100 hours, (c) after 500 hours
Fig. 10 Cross-section view of Haynes 230 oxidized for 500 hours at 1000 °C: (a) in air, (b) in He-K2