Temperature influence on the development of interdiffusion phenomena in MCrAlY-coated nickel-based superalloys

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In this work, the interdiffusion between a MCrAlY-bond coat and two different nickel-based superalloys is evaluated at three temperatures, 950 °C, 980 °C, and 1,050 °C. Of primary interest is the evolution of Kirkendall-porosity, the β-depleted zone in the bond coat and the β'-depleted zone in the superalloy. The three phenomena arise near the interface between bond coat and superalloy as a result of interdiffusion between both materials and are detrimental to the coating-substrate system. The evolution of the interdiffusion phenomena is highly dependent on the alloy composition of the materials as well as the annealing temperature. It has been found that not only the temperature dependency of the diffusion coefficient but also the temperature dependent element activities are an important factor when evaluating the interdiffusion phenomena. It has further been shown that at lower temperatures the amount of Kirkendall-porosity per volume fraction is higher than at higher temperatures, even though the overall amount of porosity is lower. Different equilibrium concentrations of main alloying elements and a correspondingly lower over-saturation of vacancies have been identified as the main explanation.

Keywords: Kirkendall-porosity / interdiffusion / thermal barrier coating / René 80 / CM 247

1 Introduction and fundamental equations

MCrAlY-type coatings are commonly used as protection against oxidation and hot corrosion of the underlying nickel-based substrate in gas turbines [1–3]. “M” stands for the base element/s of the alloy, nickel and/or cobalt. To promote resistance against oxidation and high temperature corrosion, aluminum and chromium are added to the alloy [4, 5]. Yttrium is used as a reactive element to achieve higher bonding strength of the oxide layer to the bond coat [6–8]. Apart from yttrium, other reactive elements are often added to the MCrAlY-alloy, either instead of or additionally to yttrium, e.g. Si, Ce, Hf, Ta, Zr. As the effect of these elements is

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usually not fully understood, their addition is mostly based on phenomenological observations [9, 10].

In service, at least two different mechanisms are responsible for the degradation of the coating system: The oxidation of the coating and the interdiffusion between the coating and the nickel-based superalloy. The oxidation of the coating results in the formation of the so-called thermally grown oxide (TGO), which preferably consists of stable alumina (α-Al₂O₃). The α-Al₂O₃ scale grows very dense and exhibits low diffusion coefficients for both, aluminum- and oxygen-ions. In conclusion, the growth rate of the alumina scale is also very low. Furthermore, alumina shows good adherence to the MCrAlY-alloy as well as to the overlaying ceramic thermal barrier coating. The formation of secondary oxides, due to low remaining aluminum content in the MCrAlY-coating, is often regarded as the main failure mechanism in these types of coating systems [11–15]. Secondary oxides are for example Cr₂O₃, (Ni,Al)(Co,Cr)-spinels and Ni-Oxides. These secondary oxides show significantly higher growth rates, which accordingly results in higher stresses in the thermally grown oxide and the ceramic top coat and lead to failure of the coating system shortly after [4, 12, 15]. Consequently, the microstructure of MCrAlY-alloys consist of an aluminum-rich precipitation phase, usually β-NiAl, which is embedded in the γ-(Ni,Co)-Matrix. The depletion of the β-phase is often regarded as an indicator for the depletion of aluminum. The diffusion of aluminum from the coating into the substrate largely influences the coating lifetime, as it reduces the amount of aluminum available for the alumina scale formation [13, 16].

Another detrimental diffusion phenomenon is the depletion of the γ’-Ni₃Al precipitation phase in the nickel-based substrate. The γ’-phase in Ni-based superalloys is essential for the excellent mechanical high temperature properties of the material [17]. Dissolution of the γ’-phase is therefore detrimental to the mechanical high-temperature properties of the base material.

In addition to the depletion of β-phase in the bond coat and γ’-phase in the nickel-based alloy, the formation of Kirkendall-porosity near the interface between bond coat and substrate is sometimes regarded as a lifetime reducing factor. Diffusion in metals is dominated by the vacancy mechanism [18]. As the diffusion coefficient varies between the bond coat and nickel-based alloys, differences in the overall flux of atoms in both alloys lead to the accumulation of vacancies at the interface. The accumulated vacancies can coalesce into pores, the so-called Kirkendall-pores. A certain supersaturation of vacancies is necessary for the voids to form, which is influenced by the presence of suitable nuclei and actual stresses [19, 20]. Huge amounts of Kirkendall-porosity are believed to significantly weaken the bonding strength between coating and substrate [21].

The diffusion process is usually described by Fick’s laws [22]. For a stationary state, Fick’s first law describes the flux $J_i$ of a species $i$ as dependent on the diffusion coefficient $D_i$ and the gradient of the concentration of the species $\nabla c_i$:

$$J_i = -D_i \nabla c_i$$

The diffusion coefficient $D_i$ can be written as [23]:

$$D_i = D_0 \cdot e^{\frac{Q}{RT}}$$

$D_0$ is the intrinsic diffusion coefficient, $Q$ the activation energy for the diffusion process, $R$ the universal gas constant and $T$ the temperature.

Fick’s second law describes the time-dependent change of the concentration of species $i$.

$$\frac{\partial c_i}{\partial t} = \nabla (D_i \nabla c_i)$$

The diffusion process is highly influenced by the temperature, Equation 2. Atoms in a metallic lattice can overcome the energy barrier $Q$ and move to a neighboring vacancy more easily at higher temperatures. This results in the exponential dependency of the diffusion rate of the elements on the temperature. Therefore, annealing at higher temperatures is expected to result more severe appearance of the interdiffusion phenomena Kirkendall-porosity, β-depleted zone in the bond coat and γ’-depleted zone in the nickel-based substrate.

In Fick’s laws, the driving force for the diffusion of an element are spatial differences in the concentration of the element. For most complex systems, it is more accurate to employ differences in the chemical potential $\mu$ or the activity $a$ of an element as the driving force [24, 25]. As the system
strives to minimize the Gibbs energy $G$, diffusional flow of an element occurs in the direction of the lower chemical potential/activity, Equation 4.

$$G(N_i) = U + pV - TS + \sum_{i=1}^{n} N_i \cdot \mu_i$$  \hspace{1cm} (4)

$U$ is the internal energy, $p$ the pressure, $V$ the volume, $T$ the temperature, $S$ the entropy and $N_i$ the mole fraction of $i$.

The chemical potential $\mu_i$ and the activity $a_i$ of a component $i$ are correlated by Equation 5.

$$\mu_i = \mu_i^0 + RT \cdot \ln a_i$$  \hspace{1cm} (5)

The activity $a_i$ can be interpreted as an effective concentration of the component $i$, where the chemical environment of the component is considered [24]. The activity is correlated to the concentration $c_i$ by means of the activity coefficient $\gamma_i$ of component $i$, Equation 6.

$$a_i = \gamma_i \cdot c_i$$  \hspace{1cm} (6)

2 Scientific methods

The specimens for the experimental work were prepared in the same way and the same materials were used as reported in previous publications [26, 27]. Cylindrical rods consisting of the conventionally cast nickel-based superalloys René 80 (diameter 10 mm) and CM 247 (diameter 8 mm) were coated with a NiCoCrAlYSi bond coat (BC1). Since the diameter of the superalloy rods is much bigger than the coating thickness, no effect of the different diameters is expected. The element contents of the materials used in this work are listed in Table 1.

The coatings were deposited by an industrial manufacturer using low-pressure plasma spraying (LPPS). The low-pressure plasma spraying-method is state of the art for producing high quality MCrAlY-coatings. The resulting coatings are characterized by their high density and their excellent adhesion to the substrate as well as the low amount of embedded oxide-particles [28–30]. Prior to the coating deposition process the specimens were corundum blasted to clean the surface. The MCrAlY-coating thickness was measured to be about 280 μm. The specimens were annealed in a box furnace at 950 °C and 980 °C for up to 10,000 hours and at 1,050 °C for up to 2,000 hours. Metallographic sections of the annealed specimens were prepared to be examined by optical microscopy to evaluate the formation of Kirkendall-porosity as well as the β- and γ’-depleted zones.

For the evaluation of the Kirkendall-porosity, two different methods were used in the scope of the present work. Method I produces a single value of Kirkendall-porosity per specimen by measuring the overall pore area near the interface and standardizing the measurement to the length of the examined interface; Equation 7. The standardization allows the determined value for the Kirkendall-porosity to be independent of the size of the measurement area. The described analysis has been performed on at least 20 micrographs per specimen. The measurement area for method I has been marked exemplary in a model micrograph, Figure 1.

Kirkendall – porosity

$$= \frac{\text{Pore area near the interface [μm}^2\text{]}}{\text{Length of examined interface [μm]}}$$  \hspace{1cm} (7)

To further analyze the formation of Kirkendall-porosity, a series of porosity profiles perpendicular to the bond coat/substrate-interface was determined using a script (method II). The script converts a picture into a black and white image (via a specific threshold value) and counts the resulting percentage

| Material   | Ni | Co | Cr | Al | Ti | Mo | W  | Zr  | C   | B   | Ta | Hf | Y  | Si |
|------------|----|----|----|----|----|----|-----|-----|-----|-----|----|----|----|----|
| René 80    | Bal.| 9.5| 14 | 3  | 4.8| 4  | 4   | 0.03| 0.17| 0.007| –  | –  | –  | –  |
| CM 247     | Bal.| 9.2| 8.1| 5.6| 0.7| 0.5| 9.5 | 0.01| 0.07| 0.01| 3.2| 1.4| –  | –  |
| BC1        | Bal.| 28 | 8  | –  | –  | –  | –   | –   | –   | –   | 3.2| 1.4| –  | 0.6|

Table 1. Nominal composition of the materials used in this work (wt.%).

Tabelle 1. Nominelle Zusammensetzung der untersuchten Werkstoffe (Gew.-%).
of black pixels (pores) for each row of pixels. To achieve statistically meaningful values, the script was run on 20 pictures per specimen and the profiles were arranged by the position of the bond coat/substrate-interface.

The measurement of the $\beta$-depleted zone at the bond coat/substrate-interface (inner BDZ) and the $\gamma'$-depleted zone has been performed at five different positions on eight micrographs per specimen, each taken at a different location of the metallographic section. Figure 1 shows an example of how both depletion zones have been evaluated.

Energy dispersive x-ray spectroscopy (EDX) element concentration-profiles perpendicular to the interface were obtained by performing area scans and averaging the horizontal values. Figure 1 exemplarily displays an energy dispersive x-ray spectroscopy-scanning area. The analysis was done with respect to all elements in Table 1, with the exception of Zr, C and B. The analysis also accounted for oxygen. The measured element compositions have been plotted versus the distance to the interface.

To further analyze the interdiffusion process, an evaluation procedure has been developed, where a characteristic value for the diffusion of an element either into the substrate or out of the substrate, depending on the direction of the diffusional flow of the element, is determined. The diffusion width describes the distance from the interface to the location in the substrate where the concentration profile of an element reaches its original value (composition in the as-sprayed condition).

3 Results and discussion

To evaluate the interdiffusion behavior of the alloying elements at the different temperatures the concentration profiles of Al, Cr, Ni, Co and Ti in BC1-coated René 80 and CM 247, respectively, after heat treatment at 950 °C, 980 °C and 1,050 °C for 2,000 hours is displayed, Figures 2, 3.

In both material combinations, the element concentration profiles at each temperature show similar characteristics:

- The elements chromium, aluminum and cobalt are diffusing out of the bond coat, into the nickel-based substrate; nickel and titanium are diffusing in the opposite direction
- Aluminum diffuses faster than the other elements, as is evident by how far aluminum has diffused into the substrate compared to how far the other elements have diffused into or out of the substrate under equal annealing conditions
- For most elements, a steep transition from high to low element concentration can be observed. The transition is less steep at higher temperatures
- The transition from high element concentration to low element concentration on the substrate-side of the interface in a certain specimen is located at the same distance from the interface for the elements nickel, chromium, cobalt and titanium

After 2,000 hours, the interdiffusion process has undoubtedly progressed faster when annealing at 1,050 °C compared to annealing at 950 °C and 980 °C. This can be seen by how far the elements diffusing from bond coat into the substrate (Cr, Co, Al) have penetrated the substrate and by how far the elements diffusing in the opposite direction (Ni, Ti) have diffused out of the substrate, Figures 2, 3. To additionally visualize the difference in the interdiffusion rate at different temperatures, the diffusion width of nickel for both materials combinations is displayed in dependency on the annealing
time, Figure 4. From the depiction of the diffusion width of nickel, it is obvious that after identical annealing times, at higher annealing temperatures more nickel has diffused out of the substrate than at lower temperatures.

It can further be concluded that the diffusion rate of nickel in BC1-coated CM 247 is higher than in BC1-coated René 80, Figure 4. At all temperatures, the respective diffusion width is higher in BC1-coated CM 247. The faster diffusion rate of the elements in BC1-coated CM 247 can also be seen by comparing the element concentration profiles, Figures 2, 3. The difference in diffusion rate between the two base materials can be explained by the formation of secondary β-Phase in the inter-diffusion zone of the substrate of BC1-coated CM 247, Figure 3. The diffusion coefficient in the NiAl-β-Phase has been reported to be about one or- der of magnitude higher than in the γ-Phase [31]. The faster diffusion through the β-particles and along the grain boundaries result in the higher overall diffusion rate of the alloying elements in BC1-coated CM 247 compared to BC1-coated René 80 [32, 33].

There is a strong influence of the annealing temperature on the evolution of the β- and γ'-depleted zones, Figure 5a, b, respectively. The combination of BC1 and CM 247 showed no distinct β-depleted zone, regardless of the annealing temperature and is therefore not depicted in Figure 5a.

The width of the β-depleted zone in BC1-coated René 80 is clearly influenced by the annealing temperature. A higher annealing temperature results in a wider β-depleted zone after a certain annealing time, Figure 5a. When annealing at 950 °C, no β-depleted zone can be measured, independent of the annealing time. At 980 °C, a β-depleted zone can first be measured after 4,000 hours of annealing. In contrast to that, at 1,050 °C a β-depleted zone can be observed even after very short annealing times. The faster β-depletion of the bond coat at higher temperatures can be directly correlated to the faster diffusion rates, Figure 4. While BC1-coated CM 247 shows faster aluminum diffusion into the substrate than BC1-coated René 80, no β-depleted

**Figure 2.** Concentration profiles of a) Al and Cr and b) Ni, Co and Ti in BC1-coated René 80 after annealing for 2,000 h at 950 °C, 980 °C and 1,050 °C.

**Bild 2.** Konzentrationsprofile von a) Al und Cr und b) Ni, Co und Ti in BC1-beschichtetem René 80 nach Auslagerung für 2.000 h bei 950 °C, 980 °C und 1.050 °C.
The reason can be found in the higher aluminum content of CM 247 compared to René 80, Table 1. The high aluminum content in CM 247 causes the Al-concentration in the bond coat to not fall below the amount required for $\beta$-phase precipitation. Similar to the $\beta$-depleted zone, the width of the $\gamma'$-depleted zone is largely influenced by the annealing temperature. In both material combinations, the $\gamma'$-depletion takes place faster at higher temperatures than at lower temperatures. Furthermore, a significant difference between the rate, at which the two materials combinations deplete of $\gamma'$-phase, can be observed. BC1-coated CM 247 depletes much faster of $\gamma'$-phase than BC1-coated René 80. The depletion of $\gamma'$-phase is usually caused by the depletion of $\gamma'$-forming elements, such as aluminum and titanium. This holds true for BC1-coated René 80, where in the $\gamma'$-depleted zone very low amounts of titanium have been measured, while aluminum content in the $\gamma'$-depleted zone is similar to the initial amount, Figure 2. In BC1-coated CM 247 however, the aluminum content in the $\gamma'$-depleted zone...
is increased, which leads to the transformation of the $\gamma'$-phase into the $\beta$-phase. Diffusion of titanium only plays a minor role in this material combination as the titanium content in CM 247 is very low to begin with, Table 1.

To further display the correlation between the $\gamma'$-depleted zone and the interdiffusion width in the substrate, Figure 6 shows a 45-degree diagram of the $\gamma'$-depleted zone and the nickel diffusion width of both material combinations for various annealing times at 950 °C, 980 °C and 1,050 °C. The nickel diffusion width is used in this graph representationally for the diffusion width of the other main alloying elements, except for aluminum. It has been shown previously that the elements nickel, cobalt, chromium and titanium diffuse analogous, Figures 2, 3. Figure 6 shows that all data points, especially the ones at longer annealing times and therefore larger $\gamma'$-depleted zones and nickel diffusion widths, lay close to the 45-degree line. Therefore the $\gamma'$-depleted zone can be used to describe how far the main alloying elements have diffused out of or into the substrate. This is of advantage, since for the $\gamma'$-depleted zone reliable average values can more easily be determined than for the diffusion width.

Figure 7a, b show the evolution of Kirkendall-porosity with ongoing annealing time at 950 °C, 980 °C and 1,050 °C for BC1-coated René 80 and CM 247, respectively. The measurements were performed by means of method I, described in chapter 2. The data points have been fitted by a parabolic law to guide the eye. For both material combina-
Kirkendall-porosity develops more rapidly at higher temperatures. In BC1-coated René 80, higher amounts of Kirkendall-porosity were measured at all temperatures, compared to BC1-coated CM 247.

The Kirkendall-porosity profiles for BC1-coated René 80 and CM 247, annealed at 950 °C, 980 °C and 1,050 °C for 2,000 hours are shown, Figure 8. The porosity profiles were obtained by means of method II, detailed in chapter 2. At all investigated

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**Figure 7.** Kirkendall-porosity in a) BC1-coated René 80 and b) BC1-coated CM 247 in dependency on the annealing time at 950 °C, 980 °C and 1,050 °C. Measured with method I.

**Bild 7.** Kirkendall-Porosität in a) in BC1-beschichtem René 80 und b) BC1-beschichtem CM 247 in Abhängigkeit der Auslagerungszeit bei 950 °C, 980 °C und 1.050 °C. Ermittelt mit Methode I.

**Figure 8.** Porosity profiles in a) BC1-coated René 80 and b) BC1-coated CM 247 after 2,000 hours at 950 °C, 980 °C and 1,050 °C. Measured with method II.

**Bild 8.** Porositätsprofile in a) in BC1-beschichtem René 80 und b) BC1-beschichtem CM 247 nach 2.000 h bei 950 °C, 980 °C und 1.050 °C. Ermittelt mit Methode II.
temperatures, the Kirkendall-porosity in both material combinations is forming primarily on the substrate-side of the interface. The area in which the formation of porosity can be observed is evidently smaller at lower annealing temperatures. Furthermore, at lower annealing temperatures, the maximum of the observed porosity is located closer to the interface than at higher temperatures. The value of the maximum porosity on the other side is decreasing with increasing annealing temperatures. From these observations, it is apparent that the higher amounts of porosity at higher temperature, measured by means of method I, are mainly a consequence of the larger volume in which the porosity is forming.

The Kirkendall-porosity profiles show distinct differences in dependency on the material combination. While the porosity maxima in BC1-coated René 80 are much higher than in BC1-coated CM 247, the area in which Kirkendall-porosity forms is larger in BC1-coated CM 247 than it is in BC1-coated René 80. The disparity in size of the area in which Kirkendall-porosity forms is directly correlated to different interdiffusion rates of the elements in both material combinations, which has been shown to be higher in BC1-coated CM 247, Figure 6. The formation of Kirkendall-porosity on the substrate-side of the interface in BC1-coated René 80 and CM 247 has already been shown for annealing at 1,050 °C in a previous publication [27]. The formation of Kirkendall-porosity at that temperature was attributed mainly to the diffusion of nickel and partly to the diffusion of titanium. Since the Kirkendall-porosity is forming in the substrate, the accumulation of vacancies must be attributed to their interdiffusion rates, which diffuse out of the substrate and into the bond coat. Out of the main alloying elements, only nickel and titanium diffuse out of the substrate and in consequence, the formation of Kirkendall-porosity on the γ'-depleted zone has been chosen for these graphs instead of the nickel diffusion width, since the width of the γ'-depleted zone can sufficiently be approximated by a line, Figure 9. The Kirkendall-porosity on the γ'-depleted zone can therefore in different amounts of Kirkendall-porosity.

Since the formation of Kirkendall-porosity is a diffusion-controlled phenomenon, Kirkendall-porosity can only form in areas, in which sufficient diffusion takes place. For the material combinations investigated in this work, where Kirkendall-porosity is related to the diffusion of nickel and titanium, the Kirkendall-porosity on the γ'-depleted zone present adequate values to describe this area. The amount of Kirkendall-porosity, as measured with method I, is presented dependent on the width of the γ'-depleted zone for the three different temperatures, Figure 9a, b. The width of the γ'-depleted zone has been chosen for these graphs instead of the nickel diffusion width, since the width of the γ'-depleted zone can be more easily measured in a larger number of specimens.

For both material combinations and at all three temperatures, the dependency of the amount of Kirkendall-porosity on the γ'-depleted zone can sufficiently be approximated by a line, Figure 9. The slope of the line is different for each investigated temperature. For both material combinations, the dependency of the amount of Kirkendall-porosity on the γ'-depleted zone at higher temperatures is described by a line with a smaller slope than at lower temperatures. In conclusion, at lower temperatures, a larger amount of Kirkendall-porosity has formed per volume area of material, in which a significant amount of diffusion. The most reasonable explanation for this observation are the different equilibrium concentrations of the alloying elements observed inside the γ'-depleted zone, Figures 2, 3. This results in different levels of oversaturation and therefore in different amounts of Kirkendall-porosity.

Furthermore, the oversaturation of vacancies in
the crystal lattice at higher temperatures is most likely more stable than at lower temperatures, which can additionally effect the formation of Kirkendall-pores. From that conclusion, it can be presumed that the formation of Kirkendall-porosity at lower temperatures should be regarded as more critical for the reliability of a component, since the porosity forms more concentrated than at higher temperatures, even though the overall amount of porosity is lower.

A noticeable difference between the influence of the annealing temperature on the development of the Kirkendall-porosity and the $\gamma'$- and $\beta$-depletion zones can be observed, Figures 5, 7. While increasing the annealing temperature results in a significant increase in the growth rate of the $\gamma'$- and $\beta$-depletion zones, the increase in Kirkendall-porosity is comparatively small. To explore the rationale behind this behavior, calculations with the commercial program Thermo-Calc have been performed to investigate the temperature-dependence of the driving force for the interdiffusion process [34–36].

The usage of Thermo-Calc for the calculation of thermodynamic properties in nickel-based superalloys and MCrAlY-alloys has been proven valid numerous times in the common literature [37–42]. In a previous publication, the good agreement between a Thermo-Calc-Simulation and experimental data for the interdiffusion process of Bond Coat 1 and René 80 was shown [26].

Figure 10 shows the activity difference between Bond Coat 1 and René 80/CM 247 superalloy at different temperatures for the elements aluminum, titanium, cobalt, chromium and nickel, calculated with Thermo-Calc. Since the bond coat contains no titanium and a titanium-activity of the bond coat can therefore not been specified, it is set to zero. The titanium-activity difference therefore describes the titanium-activity of the superalloy. The activities of the individual elements differ in orders of magnitude. Hence, a standardization was chosen, where the activity differences are displayed in relation to the value at the lowest investigated temperature (950 °C). The standardization allows displaying the data for all main alloying elements in a single graph, while still being able to highlight large proportional discrepancies in the activity difference at the various temperatures.

For the investigated temperature range, the activity differences between bond coat and superalloy of the elements nickel, chromium and cobalt decreases slightly with increasing temperature for both material combinations, Figure 10. In contrast, the activity difference of aluminum and titanium shows an almost threefold increase from 950 °C to 1050 °C. A decreasing activity difference represents a reduced driving force for the interdiffusion
of an element, while an increasing activity difference indicates a raised driving force for interdiffusion, Equation 4. To summarize, an increase in temperature does not only affect the diffusion coefficient in a way that results in a higher interdiffusion rate, but also affects the driving force for the interdiffusion process, Equation 2. In the case of aluminum and titanium, the increase in driving force results in an additional raise of the interdiffusion rate with higher temperatures. In contrast, the slight decrease in driving force for nickel, chromium and cobalt, leads to a less severe impact of an increased temperature on the interdiffusion rate. However, since the diffusion process causes a constant change of the element concentrations, and therefore the element activities, near the interface, this analysis of the activity differences is strictly speaking only valid for the beginning of the diffusion process.

The observed discrepancy in the influence of the annealing temperature on the development of the Kirkendall-porosity and the $\gamma'$- and $\beta$-depletion zones can mainly be attributed to the divergent influence of a temperature increase on the development of the activity-differences of the various elements. The temperature dependence of aluminum-activity difference results in higher diffusion rates of aluminum at elevated temperatures, leading to a large rise in the growth rate of the $\beta$-depletion zone, Figure 5a. In the case of BC1-coated CM 247, the temperature dependence of the Al-activity difference also results in a large rise in the growth rate of the $\gamma'$-depleted zone with increasing temperature, Figure 5b. Similarly, the growth rate of the $\gamma'$-depletion zone in BC1-coated René 80 also shows a large rise with increasing temperature, as the titanium-activity difference displays a comparable temperature dependence to the aluminum-activity difference, Figure 5b.

On the contrary, the development of the Kirkendall-porosity in the investigated material combinations is mainly linked to the diffusion of nickel. Since the nickel-activity difference decreases slightly with increasing temperature, the reduced driving force for nickel-diffusion at higher temperatures results in a higher equilibrium concentration of nickel in the $\gamma'$-depleted zone and, in conclusion, in lower amounts of Kirkendall-porosity per volume fraction.

4 Conclusions

The interdiffusion process between a MCrAlY-bond coat and the two nickel-based superalloys René 80 and CM 247 was evaluated regarding the

Figure 10. Activity difference between bond coat 1 and a) René 80 or b) CM 247 for main alloying elements at 950 °C, 980 °C and 1,050 °C. Calculated with Thermo-Calc and standardized to the value at 950 °C.

Bild 10. Unterschiede in der Aktivität zwischen Bond Coat 1 und a) René 80 oder b) CM 247 für die Hauptlegierungselemente bei 950 °C, 980 °C und 1,050 °C. Berechnet mit Thermo-Calc und normiert zu den Werten bei 950 °C.
formation of interdiffusion phenomena at three different temperatures, 950 °C, 980 °C and 1,050 °C. The interdiffusion phenomena Kirkendall-porosity, β-depleted zone in the bond coat and γ'-depleted zone in the superalloy were investigated in detail. The formation and evolution of the interdiffusion phenomena was shown to be highly dependent on the alloy composition of the materials as well as the annealing temperature. It was shown that, besides the exponential temperature dependency of the diffusion coefficient, temperature dependent element activities are an important factor when evaluating the temperature influence on interdiffusion phenomena. Furthermore, a temperature dependency of the Kirkendall-porosity per volume fraction in the interdiffusion zone has been found. In the investigated temperature range, lower temperatures lead to a higher amount of Kirkendall-porosity per volume fraction in the interdiffusion zone than higher temperatures. Different equilibrium concentrations of main alloying elements, caused by temperature dependency of the driving force for element diffusion, and a correspondingly lower over saturation of vacancies have been identified as the main explanation.

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