JOURNAL OF MATERIALS SCIENCE **34** (1999) 3957–3966

Influence of high-temperature protective coatings on the mechanical properties of nickel-based superalloys

Y. ITOH, M. SAITOH, Y. ISHIWATA
Power and Industrial Systems Research and Development Center, Toshiba Corporation, Tsurumi-ku, Yokohama 230-0045, Japan
E-mail: yoshiyasu.ito@toshiba.co.jp

The low-pressure plasma spray coating process has been established in the field of gas turbines and is used for hot parts, such as turbine blades and duct segments, which are exposed to corrosive gases at high temperatures. Overlay coatings based on the MCrAlY alloy system (M is Ni, CO or both) are commonly employed as oxidation- and corrosion-resistant coatings. Mechanical properties, such as creep and fatigue lives, of various MCrAlY coating systems were investigated at high temperature as compared with the uncoated substrates, such as eqiaxis IN738LC, directional solidified CM247LC and single-crystal CMSX-2. It was clear that the MCrAlY coatings had no significant influence on the creep lives of substrates for the sake of superior ductility of MCrAlY coatings at high temperature. The low-cycle fatigue lives of MCrAlY coated superalloys at high temperature showed only a little superior performance in comparison with the uncoated results. However, the high-cycle fatigue lives of MCrAlY coated superalloys at high temperature showed inferior performance in comparison with the uncoated results. It was because that the low-cycle fatigue cracks initiated at casting cavities inside the substrate in both the coated and the noncoated cases. However, the high-cycle fatigue cracks initiated at interface defects, such as small pores and grit residues, between the MCrAlY coating and the substrate and grew into the MCrAlY coating, and then into the substrate.

1. Introduction

For gas turbines, the working environment has been extremely severe. The hot section components must endure various stresses and severe environment at operating temperature. Recentry, coating technologies become indispensable to endure against the severe combustion environment at high temperature. Current status of the coating technologies for gas turbines, such as a high temperature protective coating, has been investigated by paying attention to the coating processes and materials [1, 2].

In recent years, a low-pressure plasma spray coating process has been established in the field of gas turbines and is used for hot parts, turbine blades and duct segments etc., which are exposed to oxidation and corrosive atmosphere at high temperature. Overlay coatings based on MCrAlY alloy systems (M is Ni, Co or both elements) are commonly used as oxidation- and corrosion-resistant coatings. Hitherto, the performance evaluation of coatings, such as oxidation- and corrosion-resistance, has been chiefly conducted using a burner rig from the viewpoint of relative comparison. The difference of mechanical properties between coatings and substrates may cause the interactions which influence the mechanical properties and lives of a coated component. Namely, the presence of MCrAlY coating may affect the mechanical properties and lives of superalloy substrates.

A number of studies on the mechanical properties of coated superalloys at high temperature have been performed [3–6]. However, the strength behavior of coated superalloys is much complicated, because of the discontinuity of mechanical properties and of metallurgical inhomogeneities. Namely, the mechanical properties of coating materials are still poorly described and the lack of information become difficult to make up the adequate modelling of the behavior of coating-substrate composite systems. The hot section components of gas turbines are subjected to static loadings (the materials undergoes creep) and to cyclic loadings (the materials undergoes fatigue). In both cases the coating has some influence either through changes in the substrate properties due to a heat treatment for the coating or through changes in the surface layer of a component due to the presence of the coating [4]. Especially, it is very important that a matching design of coating-substrate composite systems, which is based on the analysis and data base of coating materials, is inevitable to maintain the coating

TABLE I Chemical composition of materials used

Substrates	Chemical composition (mass %)									
	Ni	Cr	W	Mo	Co	Al	Ti	Nb	Ta	Hf
IN738LC[1]	Bal.	15.8	2.46	1.78	8.11	3.45	3.60	0.92	1.92	—
CM247LC[2]	Bal.	8.04	9.38	0.51	9.31	5.63	0.72	0.01	3.27	1.56
CMSX-2[3]	Bal.	7.73	7.89	0.59	4.56	5.66	1.03	<0.01	6.07	<75 ppm

Heat treatment: [1]1393 K $\times$ 2 h Ar gas cooled, 1116 K $\times$ 24 h Ar gas cooled.
[2]1503 K $\times$ 2 h Ar gas cooled, 1143 K $\times$ 20 h Ar gas cooled.
[3]1588 K $\times$ 3 h Ar gas cooled, 1143 K $\times$ 20 h Ar gas cooled.

Coating powder	Chemical composition (mass %)						Powder size
	Co	Ni	Cr	Al	Y	Ta	
CoCrAlY	Bal.	<0.10	28.9	6.15	0.28	—	−400 mesh
CoNiCrAlY	Bal.	32.08	21.22	8.33	0.39	—	−400 mesh
CoNiCrAlY + Ta	Bal.	9.7	24.6	7.0	0.48	5.5	−400 mesh
NiCrAlY	—	Bal.	31.45	11.55	0.825	—	−400 mesh
NiCoCrAlY	23.23	Bal.	16.94	12.47	0.49	—	−325 mesh

reliability [7, 8]. However, because of much combination, such as coating processes, coating materials and substrate materials, it is not always sufficient to evaluate the coatings best able to protect the substrate materials against oxidation and corrosion.

In this paper, it was shown the results of investigations on the influence of high temperature protective coatings on creep lives, low-cycle fatigue lives and high-cycle fatigue lives of nickel-based superalloys. Definitely, we chose the three kinds of typical superalloys, such as equiaxial IN738LC, directionally solidified CM247LC and single crystal CMSX-2, and the five kinds of MCrAlY overlay coatings, such as CoCrAlY, CoNiCrAlY, CoNiCrAlY + Ta, NiCrAlY and NiCoCrAlY, which were sprayed by the low-pressure plasma spray coating process for our experimental objects.

2. Materials and experimental details

This investigation used the three kinds of as casted commercial superalloys as substrates, such as equiaxis IN738LC, directional solidified CM247LC and single crystal CMSX-2 which were typical nickel-based superalloys for gas turbine blades. And the five kinds of commercial spraying powders, such as CoCrAlY, CoNiCrAlY, CoNiCrAlY + Ta, NiCrAlY and NiCoCrAlY were selected for a low-pressure plasma spraying. The chemical composition of these materials used and powder size are given in Table I. The normal heat treatment conditions, that is to say a diffusion heat-treatment for each nickel-based superalloys are shown in the middle of Table I.

The details of the cylindrical specimens for creep tests and fatigue tests are shown in Fig. 1a and b. The machined specimens from as casted body were sprayed by a low-pressure plasma spraying system (Plasma Technik A-2000V VPS System) under the condition that preheating temperature 843 K during transfered-arc treatment, voltage 64 V, electric current 685 A and spraying distance 270 mm in argon gas atmosphere 6 kPa. An oblique line region of specimens described in Fig. 1 was sprayed MCrAlY coatings with coating thickness of 0.30 mm after the pretreatment of blasting and transfered-arc cleaning. After spraying, each diffusion heat-treatment coincided with the substrate materials as described in Table I was performed. The final coating thickness was 0.25 mm and the surface roughness at smooth region of specimens was about $Ra = 1.0\ \mu$m like gas turbine blades using a shot-peening treatment. In case of directional solidified and single crystal, the longitudinal direction of cylindrical specimens were parallel to the [0 0 1] direction, which shows the excellent high temperature strength corresponding to the blade's axial direction. For the comparison in creep tests and fatigue tests, the cylindrical specimens without the MCrAlY overlay coating were machined as shown in Fig. 1. The surface of specimens without the MCrAlY overlay coating was polished about $Ra = 0.6\ \mu$m using #600 emery paper.

Creep rupture tests were carried out in air atmosphere at 1123 and 1173 K using a vertical lever-loading type testing machine with a electric furnace. An axial displacement was measured by a linear voltage differential transformer attached to the specimen surface. And using a closed loop, servo-controlled testing machine with a electric furnace, pulsating constant load controlled tests (high-cycle fatigue tests) were performed on the specimens within the frequency range of 5 to 10 Hz in completely reversed load control (sin waves) at 1123 and 1173 K. Total axial strain controlled fatigue tests (low-cycle fatigue tests) were conducted at a frequency of 0.1 Hz in completely reversed axial strain control (triangle waves) at 1123 and 1173 K. The axial strain was controlled by a strain detector attached to the specimen surface, which contained a linear voltage differential transformer. The stress was determined by the measured results at $N/N_f = 0.5$ (N: number of cycles, N_f: number of cycles to failure). In all test results, the applied stress was calculated under the assumption of a non-load-bearing coating [2, 4, 8].

Fractured surface of specimens after the creep tests and the fatigue tests was examined with a scanning electron microscope (SEM) and an optical microscope

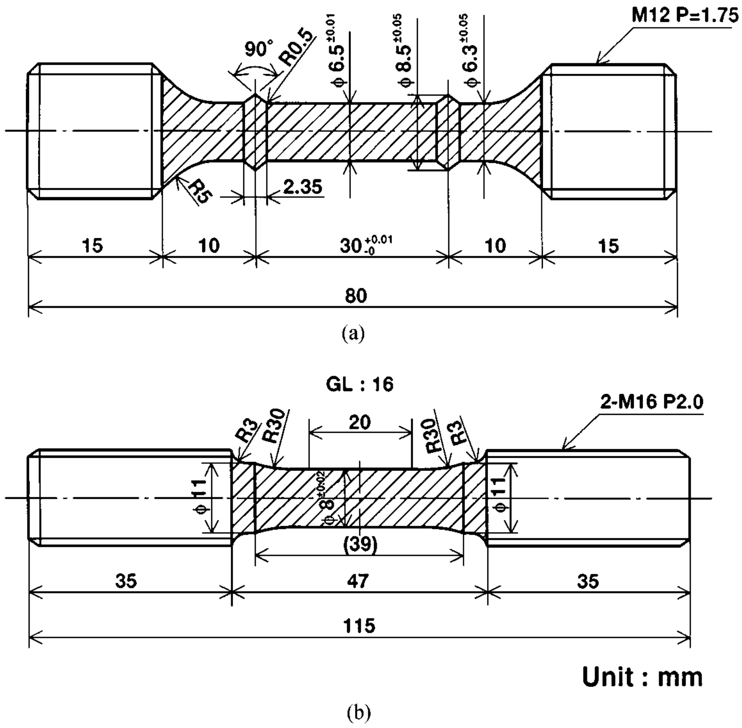


Figure 1 Geometry of specimens: (a) creep test specimen and (b) fatigue test specimen.

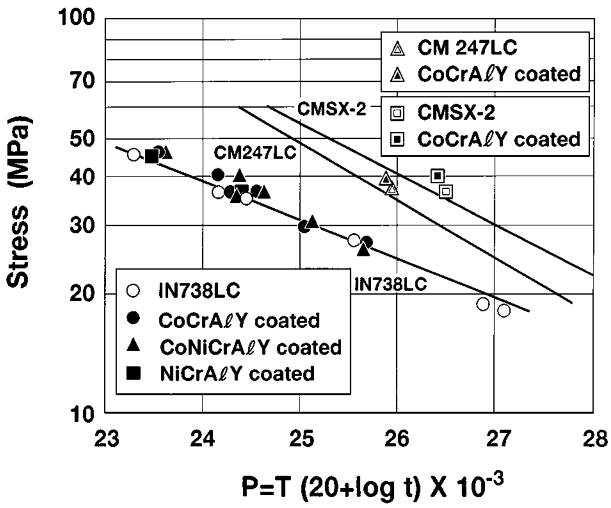


Figure 2 Effect of MCrAlY coatings on master rupture curves.

to observe the behaviors of crack initiation and growth. Also, metallographic sections of selected specimens, from both fractured and/or interrupted tests, were prepared for the observation of fracture mode.

3. Creep life properties at high temperature

It was found that no remarkable difference could be observed in creep rupture curves between nickel-based superalloys and their MCrAlY coated specimens. Therefore, master rupture curves [9] for equiaxis IN738LC, directional solidified CM247LC and single crystal CMSX-2 were shown in Fig. 2 in comparison with their MCrAlY coated specimens. Namely, Fig. 2 shows the creep rupture curves obtained by coated and uncoated specimens using a "Larson-Miller" type approach. The uncoated IN738LC and the MCrAlY coated IN738LC specimens were tested at 1123 and 1173 K in air atmosphere. And the uncoated CM247LC and the uncoated CMSX-2 and their CoCrAlY coated specimens were tested at 1173 K in air atmosphere [4, 8]. As a matter of course, the creep strength of substrate specimens falls off very rapidly with increasing temperature and stress. It is found that the creep strength can be set in order like IN738LC < CM247LC < CMSX-2. This improvement in creep properties is directly attributed to minimizing transverse grain boundaries and axial alignment of the crystallographic structure within each grain. It is very important in this study that the remarkable difference between uncoated substrate specimens and the MCrAlY coated specimens can not be observed. That is to say, the load-bearing ability of MCrAlY coatings can disregard in these experiments. And it is found that no remarkable difference can not be observed in creep lives among the CoCrAlY, CoNiCrAlY and NiCrAlY coated specimens.

By the way, it is known that tensile properties of the MCrAlY coated materials at room temperature are affected by the high strength and low ductility of MCrAlY coatings. However, the tensile properties at high temperature are out of problems in practice for an increase in ductility of MCrAlY coatings [3, 4]. Creep elongation of above 100% has been reported at the higher temperature, deformation being superplastic [4]. It is thought that various MCrAlY coatings shows extremely high ductility and low strength above 1123 K in our experiments. Therefore there is no difference of creep lives between the uncoated substrates and the MCrAlY coated specimens. Also, Fig. 3 shows the all datum for minimum creep rate, $\varepsilon_{\min}$ obtained in our experiments. The dotted line in this figure shows the relationship of Equation 1.

$$\varepsilon_{\min} \cdot t_s = const. \tag{1}$$

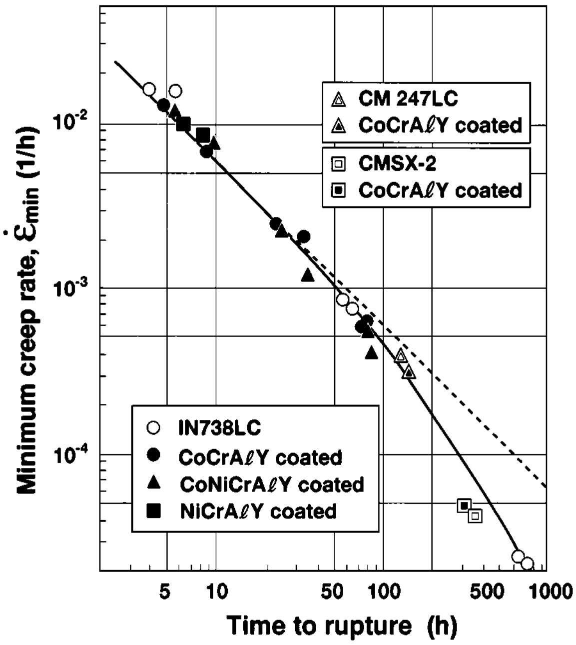


Figure 3 Effect of MCrAlY coatings on minimum creep rate curve.

It is well known that Equation 1 can be recognized between the minimum creep rate and time to rupture, t_S. In our experiments, the same relationship can be observed in the range of high creep rate and short time to rupture, and the marked difference behavior between the uncoated substrates and the MCrAlY coated specimens can not be recognized.

Fig. 4 shows the typical examples of cross-section of creep ruptured specimen in case of the single crystal CMSX-2 and the CoCrAlY coated CMSX-2 specimens at test temperature 1123 K and stress 353 MPa. It is clear that the creep cracks initiate and grow along the dendritic crystalline inside the CMSX-2 substrate and the creep crack initiation inside the CoCrAlY coatings can not be observed. Almost the same results could be obtained in any other experiments for the equiaxial IN738LC and directionally solidified CM247LC. In case of the single crystal CMSX-2, multi-cracks initiate and grow by the coalescence of cracks. Basically, it seems to be possible to select the MCrAlY coatings without marked effects on the creep lives of substrate at high temperature.

4. Low-cycle fatigue properties at high temperature

Fig. 5 shows the relationships between total strain range and number of cycles to failure, which are obtained by various MCrAlY coated specimens as compared with uncoated substrate specimens at 1123 K. The solid line in Fig. 5 shows master curve obtained for the uncoated IN738LC substrate specimens. It is clear that low-cycle fatigue lives of the CoCrAlY, CoNiCrAlY and NiCrAlY coated specimens at high temperature show only a little superior performance in compaison with the uncoated IN738LC results. The same tendency can be observed in experimental results for elastic and plastic

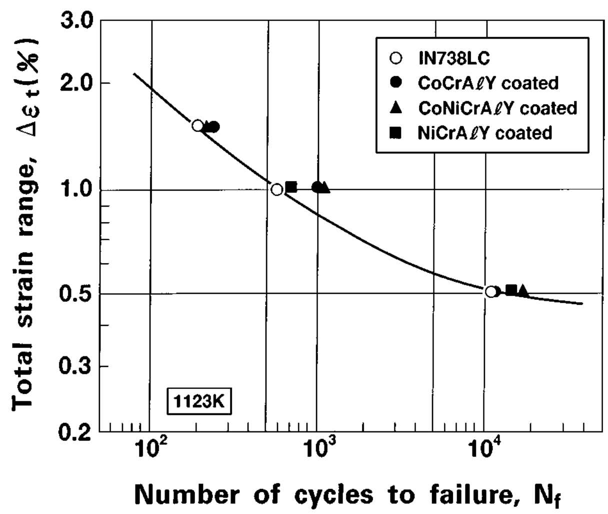


Figure 5 Effect of MCrAlY coatings on IN738LC low-cycle fatigue lives.

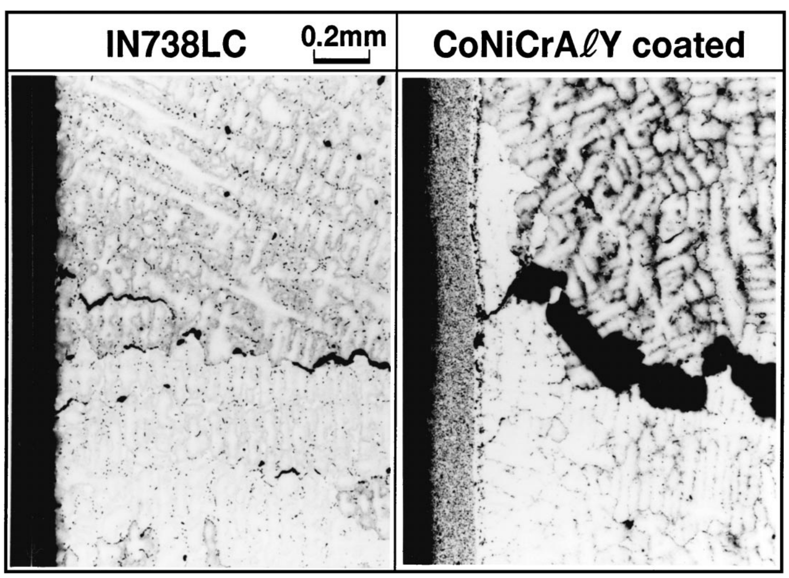


Figure 4 Comparison of creep crack growth behavior between IN738LC and CoNiCrAlY coated IN738LC.

strain range. And this tendency is remarkable for the experimental results in the range of low starin. However, the distinct difference can not be observed among the low-cycle fatigue lives of CoCrAlY, CoNiCrAlY and NiCrAlY coated specimens at high temperature. As a result, it seems to be possible to safely estimate the low-cycle fatigue lives of MCrAlY coated specimens at high temperature without the load-bearing effect of MCrAlY coatings.

In Fig. 6, the effect of substrate materials, such as directional solidified CM247LC and single crystal CMSX-2, on the low-cycle fatigue lives is shown. Namely, the relationships between total strain range and number of cycles to failure at 1173 K, which are obtained by the CoCrAlY coated specimens as compared with the uncoated CM247LC and CMSX-2 substrate specimens are shown. The solid line in Fig. 6 shows the master curves obtained for the uncoated CM247LC and CMSX-2 substrate specimens [9]. It is clear that the fatigue lives of CMSX-2 substrate are longer than that of CM247LC substrate in the range of low strain. However, the fatigue lives of CM247LC and CMSX-2 substrate specimens show superior performance by comparison with that of IN738LC substrate specimens as shown in Fig. 5. The [0 0 1] crystal direction for CM247LC and CMSX-2 substrate specimens allows for approximately a 35% reduction in Young's modulus for this direction. This reduction in Young's modulus is directly correlated with lower thermal stress resulting in a significant increase in low-cycle fatigue lives. In our experiments, it is also clear that the low-cycle fatigue lives of CoCrAlY coated specimens at high temperature showed only a little superior performance in comparison with the uncoated CM247LC and CMSX-2 substrate results. This tendency is almost same in case of the equiaxis IN738LC specimens.

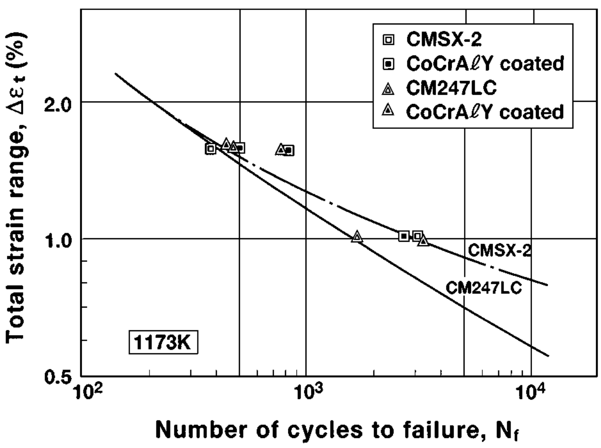


Figure 6 Effect of CoCrAlY coating on CMSX-2 and CM247LC low-cycle fatigue lives.

Typical surface observations of the low-cycle fatigue fractured specimens are given in Fig. 7 for analysing the low-cycle fatigue fracture mechanism of MCrAlY coated specimens. These photographs are the SEM

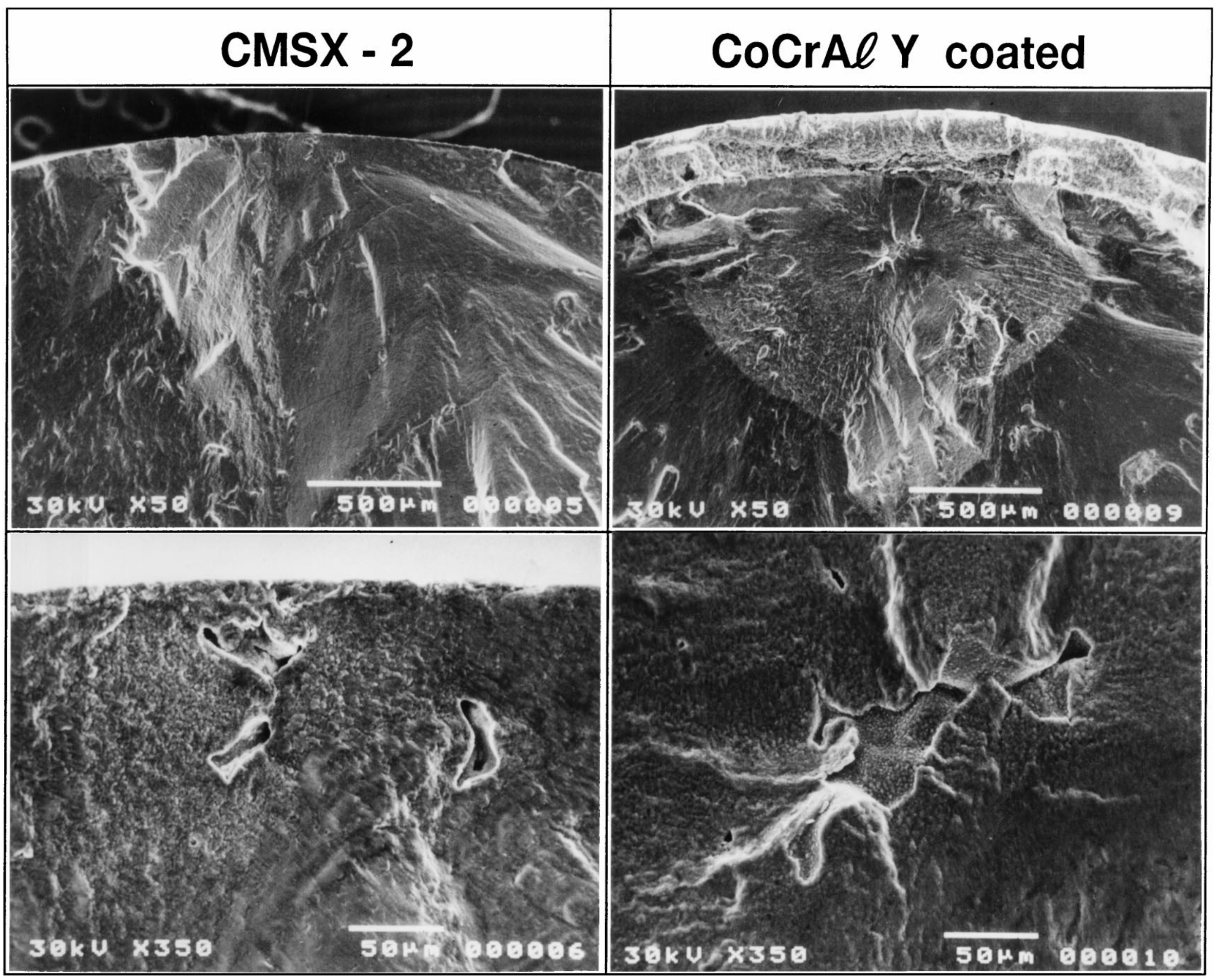


Figure 7 SEM observation of fracture appearance of CMSX-2 and CoCrAlY coated CMSX-2.

observations in case of the CoCrAlY coated CMSX-2 as compared with the single crystal CMSX-2 at total strain range, $\Delta\varepsilon_t = 1.0\%$. It is clear that the low-cycle fatigue cracks initiate at the casting cavities exposed at specimen surface in any cases. For anyother substrate specimens, such as equiaxis IN738LC and directional solidified CM247LC, the same behavior of crack initiation can be observed. On the other hand, the internal crack initiation at the casting cavities is identified as the source of failure in a majority of specimens and the low-cycle fatigue crack initiation due to the CoCrAlY coating can not be observed as shown in Fig. 7. It seems to be possible to choose the MCrAlY coatings, such as CoCrAlY, CoNiCrAlY and NiCrAlY, without inferior effects on the low-cycle fatigue lives of the nickel-based superalloy substrates at high temperature.

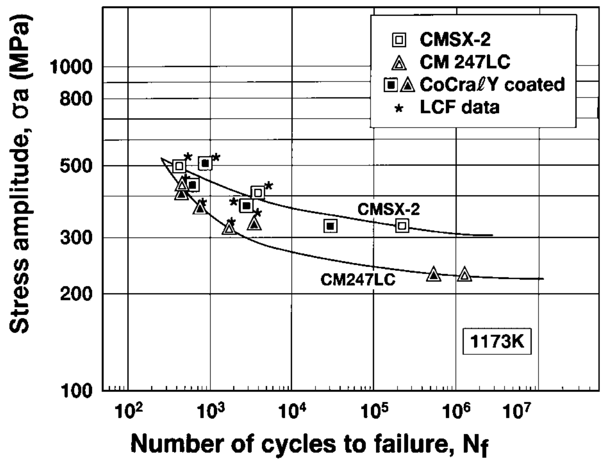


Figure 9 Effect of CoCrAlY coating on CMSX-2 and CM247LC high-cycle fatigue lives.

5. High-cycle fatigue properties at high temperature

The presence of a high temperature protective coating can significantly alter the fatigue resistance of superalloy substrates. This phenomenon has been clearly shown in the case of single crystal superalloys [4, 5]. The fatigue failure lives, N_f are plotted against the stress amplitude at $N_f/2$, are given in Fig. 8, where for comparison the fatigue lives of IN738LC substrate specimens in the load-controlled high-cycle fatigue tests are also plotted. It is interesting that the smooth extrapolation on stress amplitude vs. number of cycles to failure can be roughly seen up to the high-cycle region. This solid line in Fig. 8 shows the master curve obtained for the uncoated IN738LC substrate specimens. Also, the fatigue lives for the CoCrAlY and CoNiCrAlY coated specimens as compared with the uncoated substrate specimens at 1123 K are shown. It is clear that the high-cycle fatigue lives of the CoCrAlY and CoNiCrAlY coated specimens at 1123 K showed a little degradation in comparison with the uncoated IN738LC results in the high-cycle region. However, the difference of fatigue lives between the CoCrAlY coated specimens and the CoNiCrAlY coated specimens can not be found out in the low-cycle fatigue lives. Almost

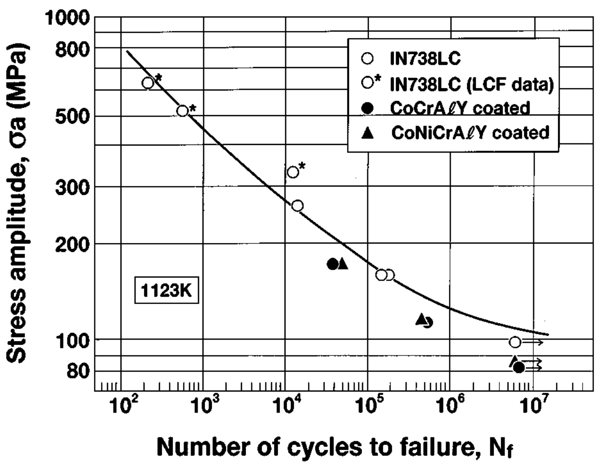


Figure 8 Effect of MCrAlY coatings on IN738LC high-cycle fatigue lives.

same experimental results can be obtained in the case of directional solidified CM247LC and single crystal CMSX-2 as shown in Fig. 9. Namely, Fig. 9 shows the relationships between the stress amplitude and the number of cycles to failure, which are obtained by the CoCrAlY coated specimens as compared with the uncoated CM247LC and CMSX-2 substrate specimens at 1173 K. The solid line shows the master curves obtained for the uncoated CM247LC and CMSX-2 substrate specimens [9]. Therefore, it seems to become the underestimation of high-cycle fatigue lives for the MCrAlY coated specimens under the simple assumption of non load-bearing coating as shown in the low-cycle fatigue lives.

By the way, the high-cycle fatigue behavior of metals is very strongly influenced by the crack initiation conditions. In casted nickel-based superalloys the crack initiation site is generally determined by internal casting cavities. Fig. 10 shows the fracture appearance of CMSX-2 substrate specimens and CoCrAlY coated specimens for investigating the behaviors of fatigue crack initiation and growth. As the fatigue tests had been conducted at 1173 K, the CMSX-2 substrate specimens without the CoCrAlY coating are excessively oxidized. It is found that the CoCrAlY coating is effective for the protection against oxidation atmosphere at high temperature. In anyother substrate specimens, such as equiaxis IN738LC and directional solidified CM247LC, the similar oxidation behavior can be observed. However, the marked reduction in thickness and surface roughening of specimens, which affects the high-cycle fatigue lives, can not be recognized by the observation of cross-section views. Secondly, it is clear that the high-cycle fatigue cracks initiate at the casting cavities, which are exposed at the surface of CMSX-2 substrate specimen, like in case of the low-cycle fatigue tests. The multi-cracks initiation from the casting cavities exposed at specimen surface and the fatigue crack growth by the coalescence of multi-cracks can be observed. On the other hand, it is observed that the high-cycle fatigue cracks initiate at the interface between the CMSX-2 substrate and the CoCrAlY coating. The cross-section views of the CoCrAlY coated CMSX-2 specimen are shown in Fig. 11. Only a main crack

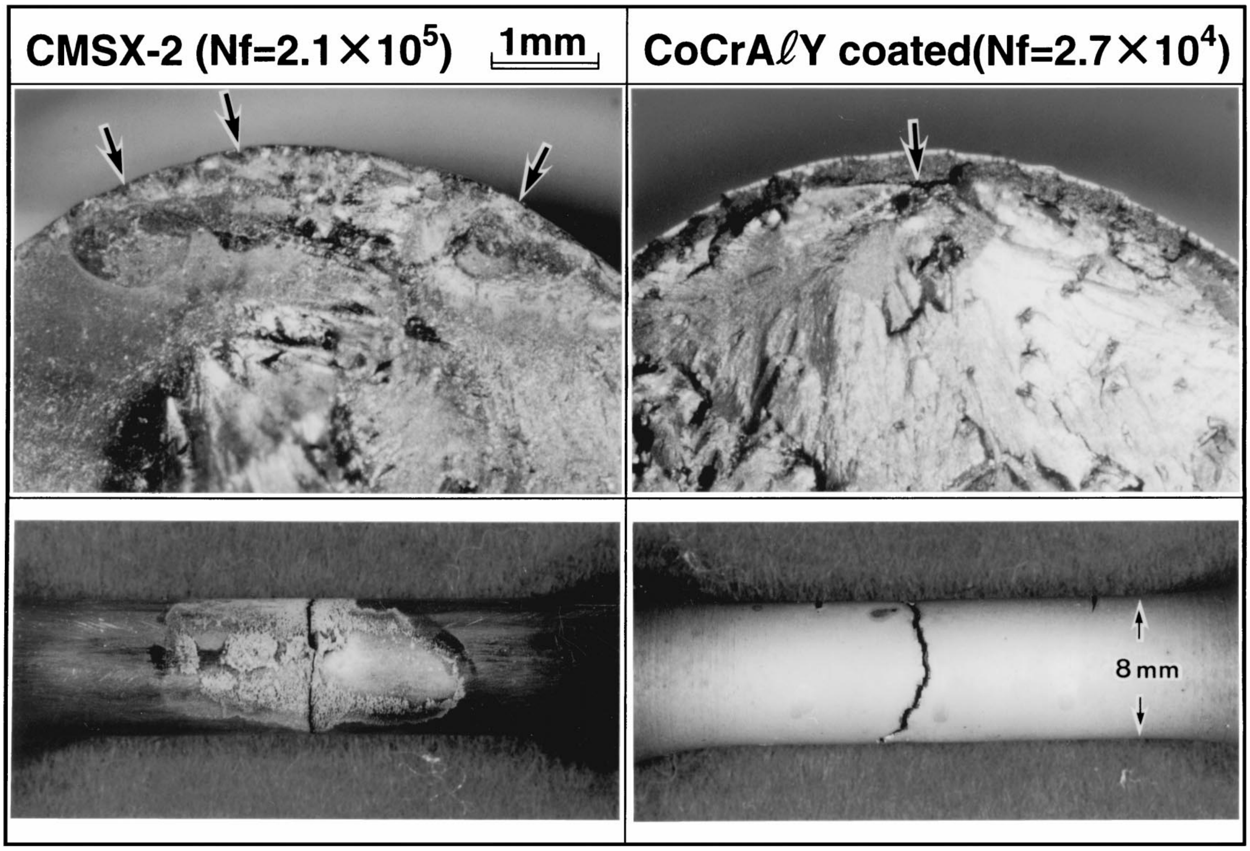


Figure 10 Comparison of fracture appearance between CMSX-2 and CoCrAlY coated CMSX-2.

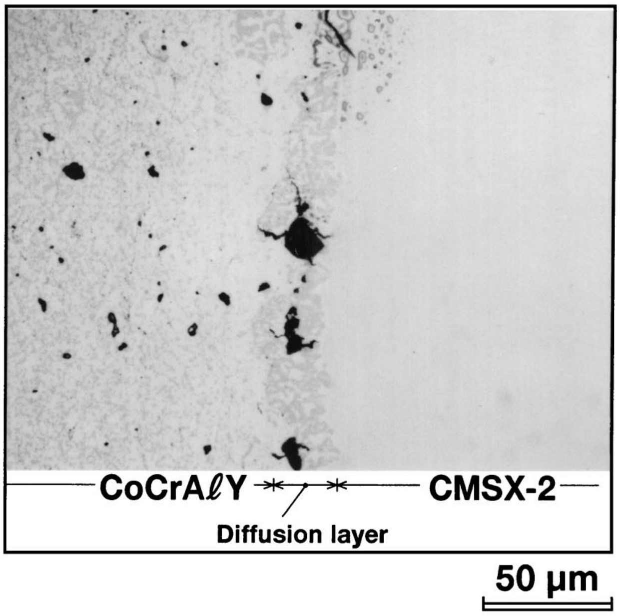
is found from the external appearance as shown in Fig. 10. However, it is clear that multi-cracks can be seen from the cross-section views as shown in Fig. 11. Also, it is clear that the fatigue cracks initiate from pores and/or residuum, which exist at the interface between the CMSX-2 substrate and the CoCrAlY coating. It is thought that the residuum are alumina grits used for the pretreatment of low pressure plasma spraying process. Both type cracks, which are parallel to the loading direction and are perpendicular to the loading direction, initiate from the pores and/or residuum. The parallel cracks to the loading direction grow along the diffusion layer at the interface between the CMSX-2 substrate and the CoCrAlY coating. And they seem not to directly influence on the high-cycle fatigue lives. The perpendicular cracks to the loading direction is thought to influence on the high-cycle fatigue lives. Firstly, the cracks initiated from the pores and/or residuum pass through the CoCrAlY coating layer. In the next, the cracks passed through the CoCrAlY coating layer grow into the CMSX-2 substrate. The same behaviors for crack initiation and growth can be seen in anyother case of the MCrAlY coated IN738LC and the MCrAlY coated CM247LC specimens.

As a result, the fatigue crack initiation for the MCrAlY coated specimens switches from the internal casting cavities in the low-cycle fatigue tests towards the interface pores and/or residuum in the high-cycle fatigue tests. The reduction of the MCrAlY coated specimens in the high-cycle fatigue lives is related to early cracking in the interface pores and/or residuum with the subsequent growth into the substrate. As a result, it seems to be impossible to safely estimate the high-cycle fatigue lives of MCrAlY coated specimens at high temperature without the load-bearing effect of MCrAlY coatings like as the creep lives and the low-cycle fatigue lives.

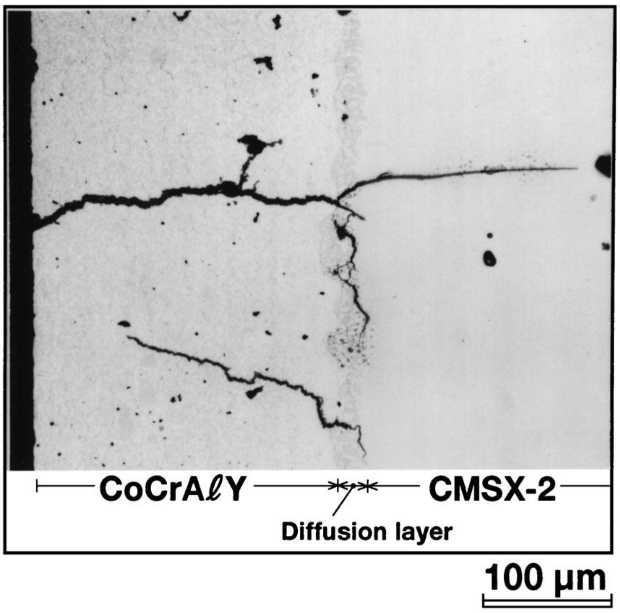
6. Fatigue fracture consideration for MCrAlY coated superalloys

It has shown that the presence of MCrAlY coatings has not remarkably affected the creep lives of nickel-based superalloys as well as the low-cycle fatigue lives of nickel-based superalloys at high temperature. On the other hand, the high-cycle fatigue lives of nickel-based superalloys have been reduced by the presence of MCrAlY coatings. This phenomenon for the reduction of high-cycle fatigue lives is very important problem to safely design the high-temperature protective coatings for the practical use of gas tubine components. The purpose of this section is to mainly investigate the reduction mechanism of high-cycle fatigue lives in case of the MCrAlY coated superalloys.

By the way, the low-cycle fatigue lives of a free-standing NiCoCrAlY coating at high temperature had been presented [6]. It was also known that the low-cycle fatigue lives of the NiCoCrAlY coating at 923 K were significantly less than that of the NiCoCrAlY coated single crystal, PWA 1480. However, the low-cycle fatigue lives of the NiCoCrAlY coating at 1323 K were five times greater than that of the NiCoCrAlY



(a)



(b)

Figure 11 High-cycle fatigue crack growth behavior observed at cross-section of CoCrAlY coated CMSX-2: (a) crack initiation at interface defects and (b) crack growth from coating to substrate.

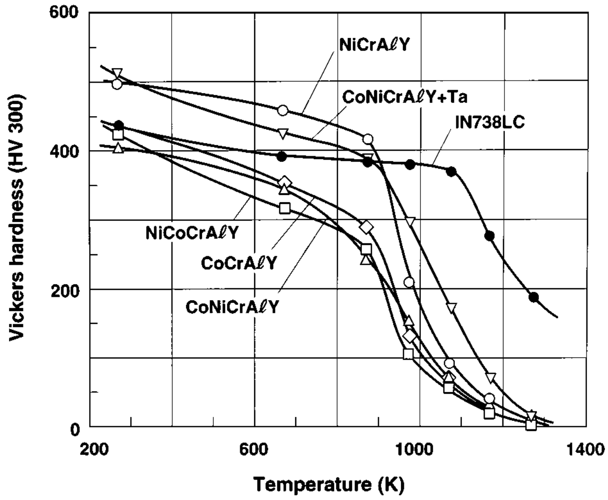


Figure 12 Temperature dependency of Vicker's hardness for various MCrAlY coatings.

coated single crystal, PWA 1480. This experimental result is due to the mechanical characteristics of NiCoCrAlY coating, which show high strength and low ductility at lower temperature (923 K) and low strength and high ductility at higher temperature (1323 K). It is thought that the temperature dependency of mechanical properties of MCrAlY coating is a key point for determining that of MCrAlY coated superalloys. Fig. 12 shows the measurement results of temperature dependency of Vickers hardness, which correspond with the strength and ductility, for various MCrAlY coatings. Also, the Vickers hardness of IN738LC substrate at high temperature are shown for comparison with the MCrAlY coatings. It is clear that the Vickers hardness rapidly decreases with increasing the temperature above 1000 K. Namely, the tendency of low strength and high ductility for the MCrAlY coatings becomes remarkable in the temperature range of above 1000 K. At 1123 and 1173 K of this study, the Vickers hardness of CoCrAlY, CoNiCrAlY and NiCoCrAlY is less than 50 and reduces to about 1/7–1/10 in comparison with the IN738LC substrate. There is a general tendency that the low-cycle fatigue lives of metals becomes longer with increasing the ductility. Consequently, it is thought for the MCrAlY coated nickel-based superalloys at high temperature that the low-cycle fatigue cracks initiate from the casting cavities inside the nickel-based superalloy substrate, previous to the MCrAlY coating layer.

On the other hand, it was showed that the negative effect due to the presence of MCrAlY coatings on the high-cycle fatigue lives of nickel-based superalloy substrates at high temperature could be found in the case that the fatigue crack initiated at the interface pores and/or residuum between the substrate and the MCrAlY coating. It is generally used practice to plot fatigue crack growth rates, $\mathrm{d}a/\mathrm{d}N$ versus cyclic stress intensity range, ΔK in a doublelogarithmic way. The fatigue crack growth rates and a fatigue crack growth threshold, ΔK_{th} for various metals had been able to be represented by the empirical Young's modulus correlation like $\Delta K/E$ (ΔK: cyclic stress intensity range, E:Young's modulus) [10]. Namely, it is obvious that the fatigue crack growth threshold becomes lower and the fatigue crack growth rate becomes faster in case of the lower Young's modulus metals. The measurement results of Young's modulus at high temperature for five kinds of MCrAlY coatings are shown in Fig. 13. There is a tendency that the Young's modulus becomes lower with increasing the test temperature like the IN738LC substrate. At test temperature, 1123 and 1173 K, it is thought that the Young's modulus of MCrAlY coatings become lower than that of nickel-based superalloys at high temperature like the Vicker's hardness as shown in Fig. 12. From this point of view, the fatigue crack growth curves of the MCrAlY coating and the nickel-based superalloy substrate are schematically represented in Fig. 14. In case of the low-cycle fatigue, it is thought that the fatigue crack initiates from the casting cavities exposed at the specimen surface, which shows the higher cyclic stress intensity range in

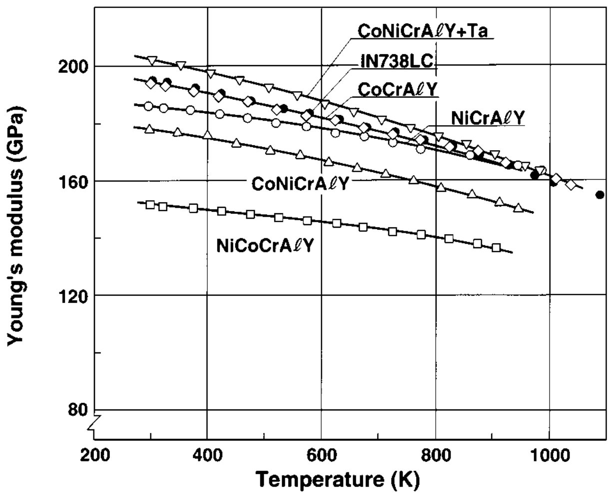


Figure 13 Temperature dependency of Young's modulus for various MCrAlY coatings.

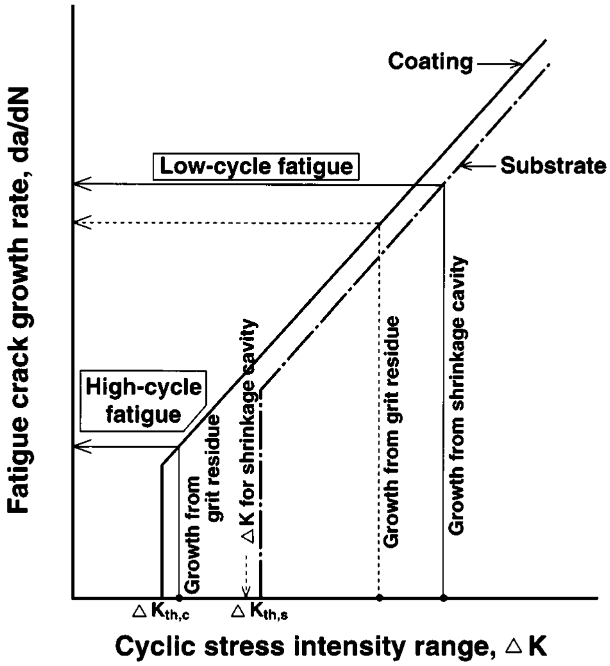


Figure 14 Fatigue crack growth mechanism of MCrAlY coated nickel-based alloy.

comparison with the interface pores and/or residuum between the substrate and the MCrAlY coating. On the other hand, it is thought that the fatigue crack initiates from the interface pores and/or residuum and grows into the nickel-based superalloy substrate in case of the high-cycle fatigue. The reason is that the cyclic stress intensity range for the internal casting cavities becomes lower than the fatigue crack growth threshold for the nickel-based superalloy substrate. This is the reason why the high-cycle fatigue crack initiates from the interface pores and/or residuum, which are smaller than the internal casting cavities.

7. Conclusions

Nickel-based superalloys employed in gas turbine blade are ofen used with a low-pressure plasma sprayed coatings for oxidation and hot corrosion resistance. These high-temperature protective coatings may affect the creep lives and the fatigue lives of nickel-based superalloys. From this point of view, mechanical properties, such as creep-rupturelives, low-cycle fatigue lives and high-cycle fatigue lives, of various MCrAlY coated nickel-based superalloys were investigated at high temperature and compared with that of the uncoated superalloys, such as equiaxis IN738LC, directional solidified CM247LC and single crystal CMSX-2 as a substrate.

As a result, it was found that the creep lives of nickel-based superalloys could be set in order like IN738LC < CM247LC < CMSX-2. This improvement in creep properties of nickel-based superalloys is directly attributed to minimizing transverse grain boundaries and axial alignment of the crystallographic structure within each grain. Also, it was clarified that the presence of MCrAlY coatings had no significant influence on the tensile strength and the creep lives of MCrAlY coated Nickel-based superalloys at high temperature. Namely, it is all because the MCrAlY coatings show extremely low strength and high ductility at high temperature. Therefore, the tensile strength at room temperature of the MCrAlY coated nickel-based superalloys was reduced for the low ductility of MCrAlY coatings on the contrary [3, 5]. And it was found that no remarkable difference could be observed in creep lives among the MCrAlY coated nickel-based superalloys. Consequently, the load-bearing ability of MCrAlY coatings can disregard in our experiments. It seems to be possible to select the MCrAlY coatings without marked effects on the creep lives of nickel-based superalloys at high temperature.

On the other hand, it was clear that the low-cycle fatigue lives of CMSX-2 substrate are longer than that of CM247LC in the range of low strain. Also, that of CM247LC and CMSX-2 substrate specimens showed superior performance by comparison with that of IN738LC substrate specimens. The [0 0 1] crystal direction shows about 35% reduction in Young's modulus for this direction, and this reduction is directly correlated with lower stress resulting in a significant increase in low-cycle fatigue lives. It was also clear that the low-cycle fatigue cracks initiated from the casting cavities exposed at the specimen surface. The low-cycle fatigue properties of MCrAlY coated nickel-based superalloys at high temperature showed only a little superior performance in comparison with the uncoated superalloys. It was because that the low-cycle fatigue cracks of the MCrAlY coated nickel-based superalloys initiated at the casting cavities inside the substrate and the low-cycle fatigue crack initiation due to the presence of MCrAlY coating could not be observed. It seems to be possible to safely estimate the low-cycle fatigue lives of MCrAlY coated superalloys at high temperature without the load-bearing effect of MCrAlY coatings. Namely, it is possible to select the MCrAlY coatings, without marked effects on the low-cycle fatigue lives of the nickel-based superalloys at high temperature.

In our experiments of high-cycle fatigue, the nickel-based superalloys without the MCrAlY coatings seemed to be fairly oxidized by the external appearance.

However, the remarkable reduction in thickness and surface roughening, which affected the high-cycle fatigue lives, could not be recognized by the detailed observation. Also, little interdiffusion occured during the high-cycle fatigue test period. A favorable influence can occur in the long run because of the suppression of environmental degradation. High-cycle fatigue properties of the MCrAlY coated superalloys at high temperature showed inferior performance in comparison with the uncoated results. It was because that the high-cycle fatigue cracks initiated at interface defects, such as pores and grit residues, between the MCrAlY coating and the nickel-based substrate. It is thought that the residuum are alumina grits used for the pretreatment of low-pressure plasma spraying. The fatigue crack initiation switches from the internal cavities in case of low-cycle fatigue towards the interface pores and/or residuum in case of high-cycle fatigue. The reduction in high-cycle fatigue lives is related to early cracking in the interface pores and/or residuum with subsequent growth inside the nickel-based superalloy. It seems to become the underestimation of high-cycle fatigue lives for the MCrAlY coated substrates under the simple assumption of non load-bearing coating.

At last, it was investigated the reduction reason of high-cycle fatigue lives in case of the MCrAlY coated substrate by using the fatigue crack growth curves of the MCrAlY coatings and the nickel-based superalloys.

References

1. D. F. BETTRIDGE and R. G. UBANK, *Mater. Sci. Technol.* **2** (1986) 232.
2. R. MEVREL, *Mater. Sci. Eng.* **A120** (1989) 13.
3. M. I. WOOD, *ibid.* **A121** (1989) 633.
4. K. SCHNEIDER and H. W. GRUNLING. *Thin Solid Films* **107** (1983) 395.
5. J. M. VEYS and R. MEVREL, *Mater. Sci. Eng.* **88** (1987) 253.
6. J. GAYDE, T. P. GABB and V. MINER, *International Journal of Fatigue* **8** (1986) 217.
7. M. OKAZAKI, T. TABATA and S. NOHMI, *Metall. Trans.* **21A** (1990) 1990.
8. Y. ISHIWATA, M. SAITOH and Y. ITOH, in "Proceedings of the 1995 Yokohama International Gas Turbine Congress," IGTC-III, 1995, pp. 99.
9. H. TAMAMOTO, M. YAMAMOTO, K. IMAI, M. MIYAZAKI, M. SATOH, Y. KOBAYASHI and H. MATSUZAKI, ASME paper, 95-GT-449, 1995.
10. M. O. SPEIDEL, "High-temperature Materials in Gas Turbines" (Elsevier Scientific Publishing Company, 1974) p. 207.

Received 12 August 1998
and accepted 13 February 1999