Effect of Simulated HTR Helium Atmosphere on Creep Rupture Properties of Nickel-base Wrought Alloys*

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Synopsis

A study has been made of the effect of a simulated HTR (High Temperature Gas-cooled Reactor) helium atmosphere on the creep rupture properties of nickel-base wrought alloys.

Carburization during creep in helium accelerated the growth of surface cracking, and led to losses in rupture life and ductility of Inconel 617. This was characterized by the early onset of tertiary creep, and was explained in terms of the reduction of net cross-sectional area by surface cracking. Similar premature failure in helium was observed also in SSS113MA, but not in NSC-1 and SZ at 950 °C. The latter alloys had low rupture ductility, and are believed to have less chances of receiving the influence on the growth of surface cracks from the atmosphere.

A change of the rupture mode on the low stress side, the strengthening by carburization, and the coarsening of carbide were examined to investigate the carburization effect on the long-term strength, and it was predicted for Inconel 617 that the rupture time in helium at 900 °C on the low stress side was satisfactorily given by multiplying the rupture time in air by 1/2.

I. Introduction

Since the research and development¹⁾ of the direct steel making technology utilizing high-temperature reducing gas were decided as a large-scale project of the Agency of Industrial Science and Technology of Japan in 1973, research on the behaviors of heat-resisting alloys in helium containing a small amount of impurity gases has also been actively promoted in Japan. Intermediate heat exchanger materials were mainly investigated, and the test gases can be classified into two kinds. One is the primary system helium circulating through a high temperature gas-cooled reactor (HTR) and intermediate heat exchanger, whose simulated composition is predicted, referring to the gas composition in experimental reactors in Europe and the USA. The other is the secondary system helium flowing from intermediate heat exchanger to various heat utilization systems. Its gas composition will mainly depend on the purification capacity and leak rate in the secondary system.

In the research and development of a 1.5 MWt high-temperature heat exchanger²⁾ in the above-mentioned large-scale project and its preceding small test loop³⁾ of the Iron and Steel Institute of Japan, the behaviors of heat exchange materials in the secondary system helium were studied. Various important information has been obtained from these studies about the decarburization characteristics and creep properties in this atmosphere. The creep strength reduction due to decarburization was recognized not only in Inconel 617 but also in Hastelloy X and SSS113MA, although the degree of strength reduction differs more

or less.⁴⁾ Furthermore, this environmental effect was tried to be reflected in high-temperature design allowable stress of Inconel 617.⁵⁾

On the other hand, the studies using the primary system (simulated HTR) helium have been carried out by far more people than those studying the secondary system helium. Simulated gas compositions and test temperatures vary from laboratory to laboratory. These factors delicately change the atmospheric effects on creep and rupture strength in some steels and alloys. According to the studies disclosed until now, it has been reported on Hastelloy X^{6,7)} that the onset of tertiary creep was likely to be earlier in helium than in air. However, its causes remain still unknown. On the other hand, various similar experiments are being continued in West Germany and other European countries. However, the atmospheric effect is reported to be insignificant.⁸⁾

In the present paper, the reduction of the creep rupture strength in carburizing HTR helium and its cause are elucidated using Inconel 617. Similar experiments are conducted on several alloys developed in Japan to examine if the same phenomenon appears for these alloys. Finally, the empirical relationship between the reduction of creep rupture strength and the amount of carburization is obtained to discuss the effect of simulated HTR helium atmosphere on the long-term creep rupture strength.

II. Experimental Procedure

1. Materials

An Inconel 617 bar (13 mm in diameter) was mainly employed as the test material. NSC-1, SSS113MA and SZ alloys supplied by each manufacturer in Japan were also used in some experiments. Table 1 shows their chemical compositions and grain sizes. From these materials, creep specimens having a gauge length of 30 mm and diameter of 3 or 6 mm were sampled.

2. Creep Testing Method in Various Atmospheres

Creep tests were conducted at 900 °C for Inconel 617 and 950 °C for other alloys using a direct loading creep machine in air and a lever type creep machine having an atmosphere chamber in vacuum and helium. The load transmission rod to specimen was sealed with the atmosphere chamber by an O-ring, and a friction force of about 0.6 kg was produced be-

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Table 1. Chemical compositions* and grain sizes of tested materials. (wt%)

Alloy	C	Si	Mn	Cr	W	Mo	Al	Ti	B	Zr	Y	Grain size (μm)
Inconel 617	0.067	0.15	0.01	22.33	—	8.99	1.05	0.36	—	—	—	No. 3 (130)
NSC-1	0.060	<0.01	<0.01	22.6	13.5	0.5	0.050	0.31	0.003	0.014	0.009	No. 1 (250)
SSS113MA	0.059	0.05	0.01	23.60	18.40	—	—	0.54	—	0.030	—	No.-3 (1 020)
SZ	0.041	0.02	0.35	27.61	5.00	5.24	—	—	0.0045	0.012	—	No. 0 (360)

* Manufacturer's analyses besides carbon content.

tween the pull rod and the O-ring. This value was additionally loaded in the helium test. The atmospheric pressure of 4.6 kg applied to the cross-sectional area of the pull rod was also added for the creep test in vacuum of 2×10^{-5} Torr.

The creep test in helium was conducted by feeding the mixed helium at the atmospheric pressure having the composition in Table 2 into the chamber without collecting the gas after passing through the chamber. The present test gases contain less amounts of H_2O and CO_2 than other simulated gases. In order to reduce a change of the gas composition due to the reaction of gas-gas and gas-metal in the chamber to a negligibly small extent, the flow rate in the chamber must be more than 2 l/min (1 cm/s in flow velocity conversion).⁴⁾ In this experiment, the gas injection pipe was introduced into the vicinity of specimens, and the flow rate was set to 200 to 300 ml/min instead of a high flow rate.

Figure 1 shows the relationship between the exposing duration at 900 °C and the increase in bulk carbon content of Inconel 617 of 6 mm in diameter. The carbon content after exposure was analyzed over the whole cross-sectional area of specimens. The triangle marks indicate that the specimens were heated in the creep testing chamber without stress, while the open circles indicate that they were heated in a quartz tube. When heated in a quartz tube, the outlet gas composition remained unchanged even at the small flow rate of 200 ml/min. As shown in Fig. 1, the carburization rate in the creep testing chamber was almost equal to that in a quartz tube. Therefore, the present test method with a small flow rate may be disregarded. The carburization rate in creep ruptured specimens (marked with solid circles) was larger than that of specimens exposed without stress. This may be considered to have been caused by an increase of net surface area due to the surface crack.

III. Experimental Results and Discussion

1. Creep Rupture Properties of Inconel 617 in Various Atmospheres

Creep tests were conducted by using specimens of 3 mm in diameter at 900 °C in simulated HTR helium, vacuum and air, and the effect of the test atmospheres was examined. Figure 2 shows the creep curves under 9 kg/mm². No significant difference due to the atmosphere in the steady-state creep was detectable. However, the onset of the tertiary creep was earlier in air and helium than in vacuum. As

Table 2. Impurity contents in simulated HTR helium for the present test.

Type		High methane	Low methane
Impurity (μatm)	H ₂	270	200
	CO	90	100
	CH ₄	40	8
	H ₂ O	*	*
	CO ₂	*	*
	N ₂	†	†
Test materials		Inconel 617	Others

* Not intentionally added (less than 1 ppm).

† Not measured.

Fig. 1. Carburization rate of Inconel 617 at 900 °C in He-270 μatm H₂-90 μatm CO-40 μatm CH₄.

Fig. 2. Creep curves of Inconel 617 in simulated HTR helium, air and vacuum at 900 °C, 9 kg/mm².

a result, the rupture time was shortened in these atmospheres.

Figure 3 shows the variation of the ratio of the creep

Fig. 3. Variation of relative rupture time and reduction in area with rupture time for Inconel 617 specimens 3 mm in diameter.

Fig. 4. Stress rupture curves of Inconel 617 specimens 6 mm in diameter in various atmospheres at 900 °C.

Fig. 5. Creep curves of Inconel 617 specimens 3 mm in diameter tested at 900 °C, 6.5 kg/mm² for examination of surface cracking behavior.

Table 3. Atmospheric effects on biaxial creep of Inconel 617 tube at 900 °C.

Stress* (kg/mm ²)	Atmosphere	Rupture time (h)	$t_{\text{He}}/t_{\text{Air}}$	Max. growth in O.D. (%)
7.4	He	35	0.44	12
	Air	80		23
6.0	He	130	0.59	9
	Air	220		17
5.0	He	270	0.54	8
	Air	500		16

* Equivalent stress was calculated using the mean diameter formula.

considered to have been caused by the early initiation and growth of surface crack and resultant reduction of the effective cross-sectional area. In order to clarify this supposition, the surface cracking behaviors were investigated by interrupted creep specimens in various atmospheres. Figure 5 shows the creep curves of Inconel 617 specimens 3 mm in diameter tested at 900 °C, 6.5 kg/mm² for examination of surface cracking behavior. No significant difference in creep curves was detectable in air and vacuum under this test condition. However, the creep rate began to ac-

rupture time in helium and air to that in vacuum and the reduction in area with the test time. The effect of air atmosphere on the rupture strength disappeared when the test time was longer than 300 h. On the other hand, the minimum rupture time ratio was 0.45 at about 200 h in helium. The reduction in area decreased in each atmosphere as the test time became longer. It showed a slow changing tendency after the test time exceeded a certain value. The reduction in area in the long-term tests was in the order of helium, air and vacuum.

In the fundamental tests described above, the specimens 3 mm in diameter were employed to enhance the atmospheric effects. The stress-rupture curves using widely employed specimens of 6 mm in diameter are shown in Fig. 4. Since the test time in this figure is longer than 300 h, the rupture time in vacuum is equal to that in air. The rupture time in helium was shortened at almost the same ratio as in Fig. 3 for the rupture time of 300 to 600 h. We shall compare the reduction in area shown in Figs. 3 and 4 within the test range of 300 to 600 h where the rupture time overlaps each other. The reduction in area in vacuum was 50 to 60 %, irrespective of a difference in diameters. However, in helium and air, the reduction in area for small specimens was smaller by about 20 % than for large specimens.

The creep rupture strength reduction in helium was examined for internally pressurized tubes. A tube with a 10.7 mm outside diameter and 1.35 mm-thick wall was machined from the Inconel 617 bar materials. This 50 mm long tube was internally pressurized by argon gas and exposed to air or flowing HTR helium at 900 °C. The times to burst in helium were compared with those in air. As shown in Table 3, the rupture time ratio was 0.4 to 0.6 and the rupture ductility decreased in helium. These atmospheric effects on rupture properties seem to be only slightly severer than those in uniaxial creep tests.

2. The Causes of Early Onset of Tertiary Creep

The tertiary creep was initiated earlier in the carburizing helium than in vacuum. This may be

celerate earlier in helium when the creep strain reached 5 %.

Histograms of surface crack depth were obtained as shown in Fig. 6. Internal cracks were not observed by an optical microscope on all interrupted specimens. No surface crack was observed on the specimen interrupted in vacuum, and its surface crack density at rupture was small compared with values in other atmospheres. This means that the surface cracking was suppressed in vacuum. A significant finding is that only a small difference was present in the crack distribution between the specimen interrupted ($\epsilon=11\%$) and the specimen ruptured ($\epsilon=20\%$) in helium and that a surface crack reaching $200\ \mu\text{m}$ has already been formed at this interruption time. This indicates that main cracks which caused specimens to rupture preferentially developed after interruption, and the growth rate of other cracks was considerably smaller. By contrast, the crack distribution at rupture ($\epsilon=26\%$) largely changed from that at the interrupted stage ($\epsilon=10\%$) in air and also in vacuum. In these atmospheres the crack density of each size is supposed to increase, as the deformation progresses.

Photograph 1 shows the fracture surfaces of ruptured specimens in Fig. 5 observed by a scanning electron microscope. For the specimen ruptured in helium, flat intergranular fracture surfaces were observed, as indicated by an arrow in Photo. 1, within an area near the outer circumference at maximum depths of $500\ \mu\text{m}$. In another area, ductile fracture surfaces similar to those in air were found. In addition, fracture surfaces in vacuum were similar to those in air.

The observation of the cross sections of ruptured specimens revealed that no recrystallization occurred in the vicinity of the outer circumference where the flat fracture surfaces were seen due to the carburization in helium. In the central part where the ductile-like fracture surfaces were seen, recrystallized and fine grains were observed as shown in Photo. 2(b). Recrystallization is more clearly demonstrated in Photo. 2(d) of the specimen ruptured in air. The ductile-like fracture surfaces shown in Photo. 1(b) are caused by the creep rupture along refined grain boundaries

in the recrystallized area.

It may be stated that the grain boundary migration and recrystallization at crack tips as shown in Photos. 2 (e) and (f) were blocked by carburization, and the growth of surface cracks accelerated. Although the mechanism of accelerated surface cracking will not be examined in more detail, we shall discuss it if it can explain the early onset of the tertiary creep. The creep rate at the interrupted time ($\epsilon=11\%$) in helium shown in Fig. 5 is $4.0\times 10^{-2}\ \%\text{h}^{-1}$, which corresponds to 1.5 times the creep rate of $2.6\times 10^{-2}\ \%\text{h}^{-1}$ in air. Assuming that this acceleration is caused by the reduction of the effective cross-sectional area due to surface cracks, the amount of reduction was calculated as about 7 % from the relation⁴⁾ of $\dot{\epsilon}_s \propto \sigma^6$. Although the interrupted specimen in helium was in the tertiary creep stage, this relation was employed, supposing that the onset of tertiary creep was only geometrical in nature. This value corresponds to the reduction of $50\ \mu\text{m}$ in the specimen radius. On the other hand, surface cracks having an average depth of $60\ \mu\text{m}$ are regarded to be produced over about 1/4 of all grain boundaries penetrating to the surface when simply averaging the histograms of surface crack depth in Fig. 6. Although it is quite difficult to determine if the factor (less than unity) 1/4 is valid, this crack distribution may be considered to be tolerably close to the expected value calculated from the increase in

Fig. 6. Histograms of surface crack in various atmospheres. (see Fig. 5 for the test conditions)

Photo. 1. Fracture surface of Inconel 617 specimens tested at $900\ ^\circ\text{C}$, $6.5\ \text{kg/mm}^2$ in (a) HTR helium and (b) air.

Photo. 2.
Microstructures of Inconel 617 specimens tested in (a-b) HTR helium and (c-f) air showing recrystallization except in the carburized zone formed in helium. (e-f) shows grain boundary migration ahead of surface crack.

Fig. 7. Summarized illustration of the cause of early onset of tertiary creep in HTR helium.

creep rate. These are summarized in Fig. 7.

In the above description, the cause of early onset of the tertiary creep in carburizing helium was discussed in detail from the purpose of this paper. Similar phenomena observed in air are supposed to be explained in the same way. The external oxide layer and intergranular oxides of Al was fractured by the grain boundary sliding during creep in air and consequently, (1) intergranular oxidation is remarkably accelerated and (2) surface cracks initiate earlier in air than in vacuum⁹⁾ as in Fig. 6.

Fig. 8. Comparative stress rupture curves of several nickel-base alloys at 950 °C in helium and air (specimen: 6 mm in diameter).

3. Atmospheric Effects in Other Ni-base Wrought Alloys

Creep rupture tests in simulated HTR helium were made of nickel-base alloys NSC-1, SSS113MA and SZ. Helium gas containing 200 μatm H₂, 100 μatm CO and 8 μatm CH₄ was used in these tests. This test gas is close to that simulated by the Japan Atomic Energy Research Institute excepting the H₂O and CO₂ contents which are reduced to very low levels in the present gas.

The stress rupture curves at 950 °C in helium and air were compared in Fig. 8 for the specimens 6 mm in diameter. The increase in bulk carbon content was examined for the specimens tested in helium and given in the same figure. Some specimens exhibited

a large difference in the creep rate between helium and air from the beginning of creep and also at the steady state. The rupture data for these specimens were not plotted in Fig. 8 since their differences in creep rate were considered to be attributable to heat variations in the creep strength. Although the carburization rate of NSC-1 was larger than SSS113MA, the reduction of life in helium was seen in SSS113MA and not in NSC-1. On the other hand, the atmospheric effect on rupture ductility was largest in SZ.

A large difference was found between the cross-sectional microstructures of SSS113MA specimens ruptured in helium and in air as shown in Photo. 3. Many internal and surface cracks were observed in the specimen ruptured in air, while a few and rather deep surface cracks were found in helium. This was also true in other alloys.

The carburization rate of NSC-1 in the present test gas is as high as that of Inconel 617 shown in Fig. 1. The reason why the atmospheric effect observed in Inconel 617 has not appeared in NSC-1 can be attributable to the low rupture ductility of NSC-1. As shown in Fig. 9, the creep rate of NSC-1 rapidly accelerated when the strain reached about 5 %, while it is around 20 % or more for SSS113MA and Inconel 617. Therefore, the chance of atmosphere to

exert an influence on crack growth will be less in NSC (and SZ) as compared with Inconel 617 and SSS113MA.

IV. Prediction of Long-term Creep Rupture Strength in Simulated HTR Helium

1. Effect of Test Conditions on Atmospheric Susceptibility

1. Applied Stress

Test atmospheres will largely affect the strength under such a condition as the growth of surface cracks control the creep rupture. For instance, the fatigue cracks generally nucleate and propagate from the surface, while internal cracks dominate the creep rupture at high temperatures. According to Stegman *et al.*,¹⁰ this difference causes the effect of oxidizing atmosphere to be smaller on creep and larger on fatigue of nickel. Another example as that the low-cycle fatigue life of SUS 304 steel in air was shorter than that in vacuum under symmetric strain waveforms when the nucleation and propagation of surface cracks control the fatigue life, while the life was not affected by atmosphere under asymmetric strain waveforms when internal cracks dominate the life.¹¹

Photograph 4 shows longitudinal sections of Inconel 617 specimens ruptured in various atmospheres under the stress of 7.5 kg/mm². Surface cracks were larger than internal cracks and the growth of surface cracks are likely to affect the rupture strength in atmospheres other than vacuum. Actually, the rupture time was greatly affected by the atmospheres in these test conditions. The rupture times in vacuum, helium and air were 195, 147 and 87 h, respectively. With a decrease in stress the number of internal cracks increased, and the growth of internal cracks played an important role in creep rupture. The rupture strength reduction in air disappeared in the creep tests longer than 300 h (about 6.5 kg/mm² in stress) as shown in Fig. 3, and this can be explained by a change of this rupture mode. The strength ratio in helium and vacuum had a minimum at the rupture time of about 200 h. This may be explained by the reduced carburization in the shorter term tests and also by the increase of contribution of internal cracks in the longer term tests. The growth rate of surface cracks in helium was much larger than that in air as

Fig. 9. Comparison of creep curves in helium and air.

Photo. 3. Cross-sectional microstructures of SSS113MA specimens ruptured at 950 °C, 5.0 kg/mm² in (a) He-200 μatm H₂-100 μatm CO-8 μatm CH₄ and (b) air.

Photo. 4.
A typical example of surface and internal cracks observed in (a) helium, (b) air and (c) vacuum at 900 °C, 7.5 kg/mm² for Inconel 617.

Fig. 10. Effect of the methane content in He-H₂-CH₄ mixed gas on creep curves of Inconel 617 at 900 °C, 7.5 kg/mm².

described previously, so the rupture occurs before internal cracks sufficiently grow as shown in Photo. 2(a). As a result, the strength reduction was fully recovered at 6.5 kg/mm² in air, while not in helium.

2. Carburation Rate

As the stress is decreased and the test duration becomes longer, the amount of carburization increases and the carburization may develop in considerable depths from surface cracks. In this case, a different influence might be observed as compared with the short-term tests. In order to make this clear, creep tests were conducted using Inconel 617 in atmospheres having various carburization rates and the relationship between the amount of carburization and the creep rupture strength reduction was investigated.

Figure 10 shows the effect of the methane content on the creep curve at 900 °C, 7.5 kg/mm² in He-H₂-CH₄ mixed gas. Hydrogen was added by such a quantity as the carbon activity represented by $K \cdot P_{CH_4} / P_{H_2}^2$ was unity at 900 °C ($K=50$). The specimens were 3 mm in diameter and the carbon contents of specimens ruptured in 20 ppm and 410 ppm methane were 0.11 % and 0.16 %, respectively (the original content is 0.067 %). In 4 400 ppm methane, the carburization occurred remarkably up to the center of the specimens and the bulk carbon content may possibly exceed 0.3 %.

The rupture time was shortened as compared with that in air when the carburization was unnoticeable. These results are just the same as in simulated HTR

Fig. 11. Ratio of rupture time in carburizing helium to that in air and reduction in area plotted as a function of the amount of increase in bulk carbon content for Inconel 617.

helium. However, when the carburization exceeded a certain amount, the creep rate decreased and the rupture time was prolonged. The rupture time reduction due to the early onset of the tertiary creep amounts only to 1/2, while the decrease in the creep rate due to carburization is 1/10 or more. As a result, the rupture time was prolonged more than several times in severely carburized specimens.

Figure 11 shows the ratio of rupture time of Inconel 617 in carburizing helium to that in air and the reduction of area, plotted as a function of the amount of increase in bulk carbon content. The rupture time in helium has been discussed up to the previous paragraph by comparing it with that in vacuum. Here, the rupture time ratio in helium and air was employed from the practical view-point. The strength reduction due to carburization was maximum when the carburization quantity was about 0.05 %, and the rupture time was about 1/2 of the value in air. When the carburization exceeded 0.15 %, the rupture time was prolonged. The reason has already been discussed relating to Fig. 9. The dependence of the reduction of area on the amount of carburization is also shown in Fig. 11 for the specimens ruptured in 800 to 1 200 h. The reduction of area in air was about

45 %, but it was reduced to about 20 % due to carburization. However, the reduction of area showed a value more than 20 %, even if the amount of carburization increased to more than 0.3 %.

2. Prediction of Long-term Creep Rupture Strength

The relation between the creep rupture strength and the amount of carburization shown in Fig. 11 was obtained from data of the rupture time of 800 to 1 200 h. If it is desired to apply this relation to the long-term strength, the relation between the creep rupture behavior and the carburization structure should be clarified. As the creep test time increases, precipitation strengthening by carburization will become ineffective since carbide is aggregated and coarsened. As a harmful effect of severe carburization, carbide precipitation reduces the solid solution strengthening elements such as Mo and Cr, causing the creep resistance to deteriorate. We shall examine this in some details. Assuming that the carburization amount is 0.1 % and the carbide is $\text{Cr}_{21}\text{Mo}_2\text{C}_6$,¹²⁾ the reduction of elements in solution was calculated. It was 1.5 % for Cr and 0.3 % for Mo, being less than 1/10 of the total content. Yukitoshi *et al.*¹³⁾ examined the effect of alloying elements on the creep strength of Ni-Cr-Mo-W alloys. When a strength decrease caused by the reduction of solid solution elements is predicted referring to the above results, it is at most 1/2 of the rupture time for Cr and possibly negligible for Mo.

On the other hand, the acceleration mechanism by carburization must be elucidated, if we strictly discuss the relation between the carburized microstructure and the acceleration of the nucleation and growth of surface cracks due to carburization. However, the following is predictable when the decrease in ductility is considered as a general phenomenon of the reinforcement by carburization; the ductility loss will be restored in long-term tests since reinforcing by carburization decreases due to the coarsening of precipitates.

The above examination may be summarized as follows. In the long-term tests, the creep strengthening by carburization is weakened due to the coarsening of carbides, and also the nucleation and growth of surface cracks will not be accelerated by carburization. Consequently, Fig. 11 can be also applied to the long-term strength. The possible modification will be that the tilting of right-ascending curves in Fig. 11 may be reduced due to precipitate coarsening.

Combining Fig. 11 with the carburization amount *vs.* the time curve of Fig. 1, we can obtain the relationship between the ratio of rupture time in helium to air and the exposure time for Inconel 617. For other gas compositions and other dimensions of specimens, different carburization rate curves shall be used. Although the extrapolated curve of Fig. 4 will be used as the long-term stress rupture curve in air, other statistically analyzed curves should be used for design purposes. The predicted stress rupture curve in $\text{He}-200 \mu\text{atm H}_2-100 \mu\text{atm CO}-8 \mu\text{atm CH}_4$ is given in Fig. 12 for Inconel 617 specimens 6 mm in diameter.

Fig. 12. Estimated long-term stress rupture curves of Inconel 617 at 900 °C in carburizing or decarburizing helium.

Also shown was the strength reduction due to decarburization⁴⁾ in decarburizing helium by the same technique as described above. As clearly shown in Fig. 12, the effect of carburizing HTR helium on the creep rupture strength of Inconel 617 is very small compared with the strength reduction due to decarburization.

V. Conclusions

Creep test were conducted in simulated HTR helium to examine the effect of the carburizing helium atmosphere on the creep rupture properties of nickel-base wrought alloys. The following results were obtained:

(1) The creep rupture strength and ductility of Inconel 617 at 950 °C were higher in the order of vacuum, air and helium. The helium atmospheric effect was characterized by the early onset of the tertiary creep in the creep curves.

(2) The growth of surface cracks was accelerated due to carburization in helium, and the reduction of net cross-sectional area due to the surface cracking was enough to explain an increase of the creep rate.

(3) A similar reduction of rupture time in helium was observed at 905 °C is SSS113MA, but not in NSC-1 and SZ. The latter alloys had low rupture ductility, and were believed to have less chance of receiving the influence on the growth of surface cracks from the atmospheres.

(4) A change of the rupture mode on the low stress side, the strengthening by carburization, and the coarsening of carbide were examined to investigate the carburization effect on the long-term strength, and it was predicted for Inconel 617 that the rupture time in helium at 900 °C on the low stress side was satisfactorily given by multiplying the rupture time in air by 1/2.

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