COMPARISON OF CREEP FRACTURE IN AIR AND IN SIMULATED HIGH TEMPERATURE REACTOR HELIUM

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As part of an extensive investigation of the behaviour of high temperature alloys in simulated high temperature reactor service environments, the influence of surface cracks resulting from the interaction between creep strain and corrosion on the creep-rupture properties is investigated. From a study of surface crack depth as functions of stress, strain and time a diagram is proposed which should indicate whether corrosion effects need to be considered in design.

INTRODUCTION

In the materials testing programme of the German Nuclear Process Heat Project (PNP) which is concerned with the development of advanced high temperature reactors to supply process heat for coal gasification, the stress rupture properties of iron- and nickel-base alloys are being determined (Nickel, Schubert, Schuster (1)). The alloys under investigation are the candidate materials for the heat exchanger components (Table 1). They operate at temperatures up to around 950 °C. An important aspect of the stress rupture testing is the influence of the service environments. One of them is the reactor primary coolant helium containing unavoidable impurities (hydrogen, methane, carbon monoxide and water) at levels in the μbar range. The compositions of two HTR helium environments simulated in the creep and stress rupture testing are given in Table 2. The data from these tests in HTR helium are being compared with data obtained in air.

Examination of specimens exposed in impure helium and air shows significant differences in the type of corrosive attack (Figure 1). In the low oxidation potential helium environments, for example, nickel, iron and cobalt do not oxidize, whereas chromium, silicon, titanium, aluminium can still form oxides. Depending on the effective gas composition and temperature (Menken at al (2)) the presence of carbon monoxide and methane in the atmosphere can lead to carburization, whereas water can lead to decarburization if the often complex oxide films which contain metallic inclusions and carbide particles do not offer protection against transport of carbon through the scale.

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In spite of the differences in the corrosion behaviour the creep-rupture data determined to date in both environments with test durations up to 20 000 h fall within the same scatterbands. As an example, the stress for 1 % creep strain and creep rupture strength of INCONEL 617 tested in air and in various simulated HTR helium atmospheres are plotted in Figure 2.

For reliable extrapolation of the data to the required component operation times of 140 000 h, the influence of corrosion on the creep-rupture behaviour needs to be determined and the mechanisms of corrosion must be understood. Besides the investigations on the influence of carburization phenomena on mechanical properties by Mohr et al (3), methods for the study of the effect of surface cracks induced by the interaction of the alloy with the test environment on the creep-rupture properties have been introduced and first results will be outlined.

CHARACTERIZATION OF SURFACE CRACKS

For the investigation of surface cracking in creep specimens, a special specimen geometry was used. It is of the "hour-glass" shape shown in Figure 3. Using this specimen, a range of stresses causes a strain distribution in a single test piece and a statistical analysis of the distribution and size of the surface cracks as a function of stresses and strains is made possible. Small indentations, 3 mm apart, in the gauge length allow the measurement of the strain in the various parts of the specimen. Most of the creep-rupture tests are carried out in multi-specimen creep furnaces, with strain measurements at room temperature at interruptions of the tests after 500 and 1000 h intervals and after specimen rupture. Tests with cylindrical specimens in single-specimen creep furnaces are added if continuous strain measurement is necessary.

After creep testing an hour-glass specimen its gauge length is longitudinally sectioned and the depth and distribution of surface cracks are measured optically. A surface crack depth a_{90} defined as a with 90 % cumulative depth probability is used to obtain a characteristic parameter. Figure 4 a shows the crack length a_{90} as a function of the initial stress (load divided by the smallest initial cross-sectional area of the 3 mm sections of the hour-glass specimen) and Figure 4 b shows a_{90} as a function of the strain for a 800 °C creep-rupture test with INCOLOY 800 H.

CRACK DEPTH DIAGRAM

From data shown in Figure 4, surface crack lengths can be incorporated in a double logarithmic stress-time diagram used for plotting creep-rupture strength. From this diagram the following information can be obtained:

- an area in which surface cracks are not observed ("incubation region"),
- an area in which a technically acceptable surface crack depth a_t is not exceeded,
- the effect of different test environments on the crack depth distribution.

The data available at the present time are insufficient to allow a crack depth diagram to be quantitatively determined. Figure 5 shows first results obtained with INCOLOY 800 H at 800 °C in HTR helium.

NET STRESS CONCEPT

At temperatures above 850 ^{o}C and strain rates below $10^{-7} sec^{-1}$ rupture of the specimens tested was never initiated by the surface cracks alone. Creep pores were formed eventually which finally linked to the cracks grown into the specimen from the surface. It can be assumed that relaxation reduces stress concentration at crack tips and the net stress L/A (L:Load, A: residual cross section of the specimen taking in account pores and cracks) determines the deformation.

A measure for damage under net stress condition was introduced by Kachanov (4) which may be interpreted as being the fraction of load-carrying cross section lost under creep loading:

$$D = \frac{A^* - A}{A^*}, \quad A^* \text{ : cross section calculated assuming constant volume} \qquad (1)$$

We relate D in the secondary creep range to the reduction of A^* by a hypothetical circumferential crack which reduces the specimen radius R to (R - a_{90}). For the crack depth a_{90} we used the relationships

$$a_{90} = 0 \qquad \text{and} \qquad (2)$$

$$a_{90} = k_{He} \cdot \varepsilon^m, \quad \varepsilon \text{ :true strain} \qquad (3)$$

in order to check the corrosion induced crack growth influence on the creep rate, k_{He} and m were fitted to crack depth data measured in HTR helium.

Norton's law

$$\dot{\varepsilon} = k\, \sigma_{net}^{\ n} \qquad \text{with} \qquad (4)$$

$$\sigma_{net} = \frac{\sigma_o}{1 - D} \exp \varepsilon, \quad \sigma_o\text{: initial stress} \qquad (5)$$

was used with k and n fitted to minimum creep rates formed in a series of creep tests with different σ_o. Figure 6 shows the results in comparison to the creep rate found on an 10 mm diameter specimen. In this diagram the lines obtained with equations (2) and (3) show that a corrosion induced crack growth rate following equ. (3) should influence the measured creep rate. If there is no difference found in creep data obtained in air and helium (Fig. 2) it implies that

- the dependences of crack depth on time and strain in air and HTR helium are very similar or

- structural changes in the bulk material suppress the corrosion crack influence.

ACKNOWLEDGEMENT

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2. Menken, G., Graham, L.W., Nieder, R., Schuster H. and Thiele, W., Sept. 1982, "Review of the gas metal interactions in HTR helium up to 950 °C", see (1)

3. Mohr, K., Ennis, P.J., Nickel, H. and Schuster, H., 1982, "The effect of carbide morphology on the fracture of nickel-base alloys", this conference

4. Kachanov, L.M., 1967 "The theory of creep", Nat. Lending Library for Science and Technology, Yorkshire, England

	C	Fe	Ni	Mo	Co	Cr	Al	Ti
INCOLOY 800 H	0.1	basis	32	-	-	20	0.4	0.4
INCONEL 617	0.05 to 0.09	-	basis	9.0	12.5	22	1.0	0.3
HASTELLOY X		18		9.0	-	22	-	-
NIMONIC 86		-		10	-	25	-	-

Tab. 1: Chemical composition of candidate alloys for the PNP-project

	He	H_2	H_2O	Co	CO_2	CH_4	N_2
HTR-helium			μ bar				
PNP	basis	500	15	15	-	20	5
HHT	basis	50	5	50	5	5	5

Tab. 2: Gas composition of HTR helium; standardised compositions for environmental testing

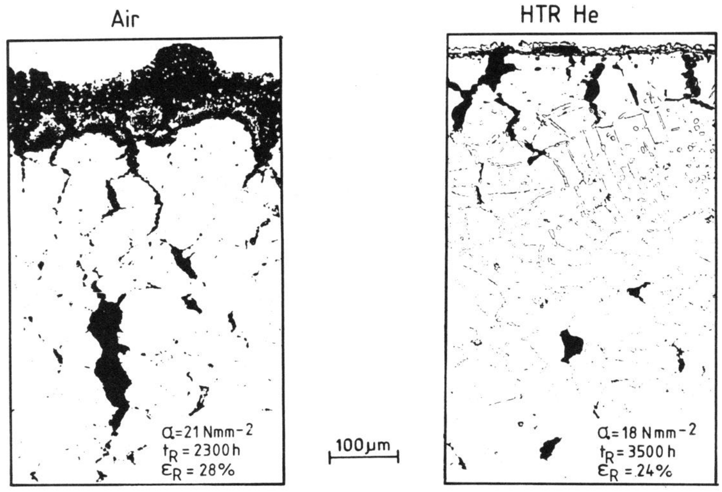

Fig. 1: Comparison of corrosion effects on creep specimens tested in air and in HTR helium
σ_0 : initial stress
t_R , ε_R : time to rupture and rupture elongation

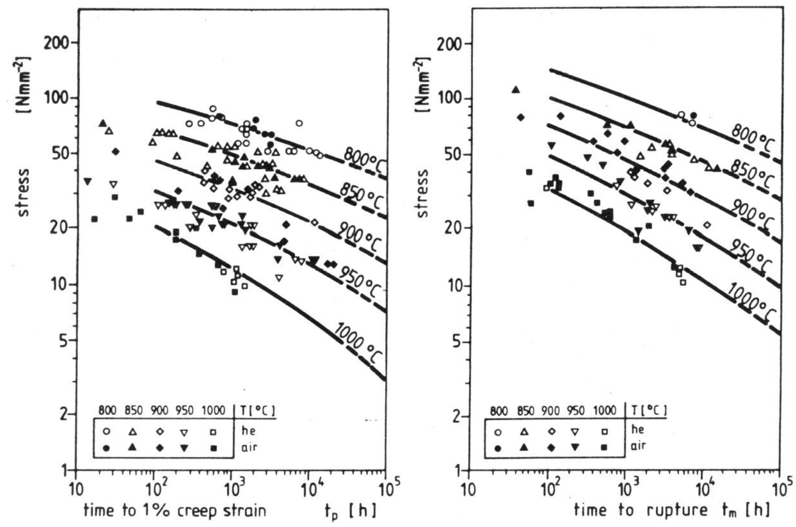

Fig. 2: Stress for 1 % creep strain and rupture strength of INCONEL 617 creep tested in air and HTR helium

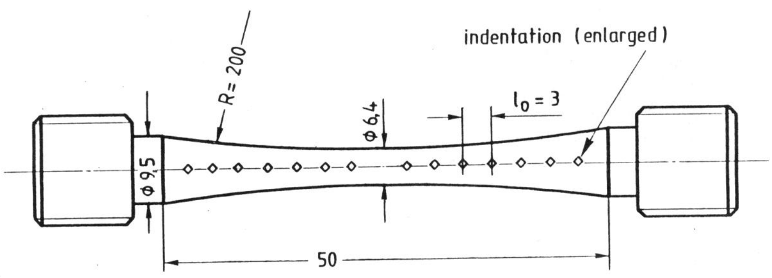

Fig. 3: Hour glass specimen used for surface crack growth studies (unit: mm)

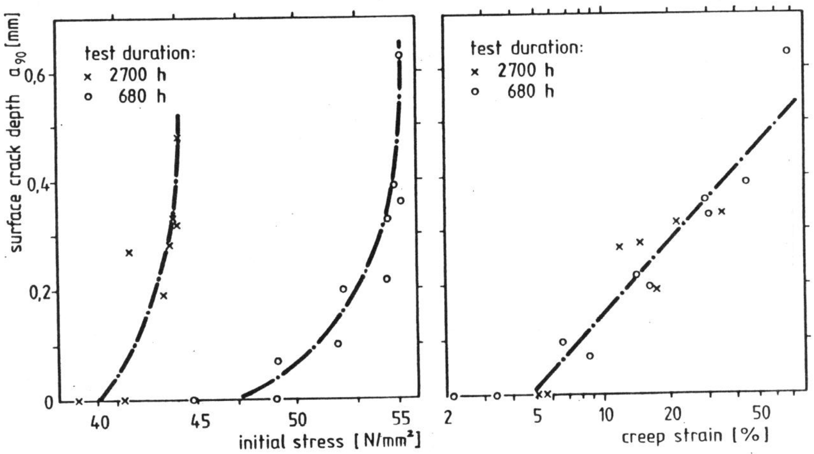

Fig. 4: Surface crack depth a_{90} versus stress (a) and versus strain (b) for creep loaded INCOLOY 800 H at 800 °C in PNP helium

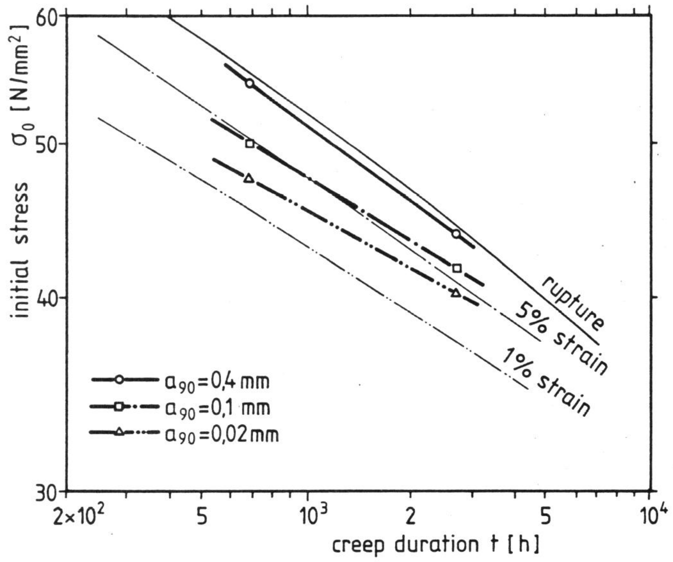

Fig. 5:
First results for the surface crack depth diagram of INCOLOY 800 H at 800 °C in PNP helium

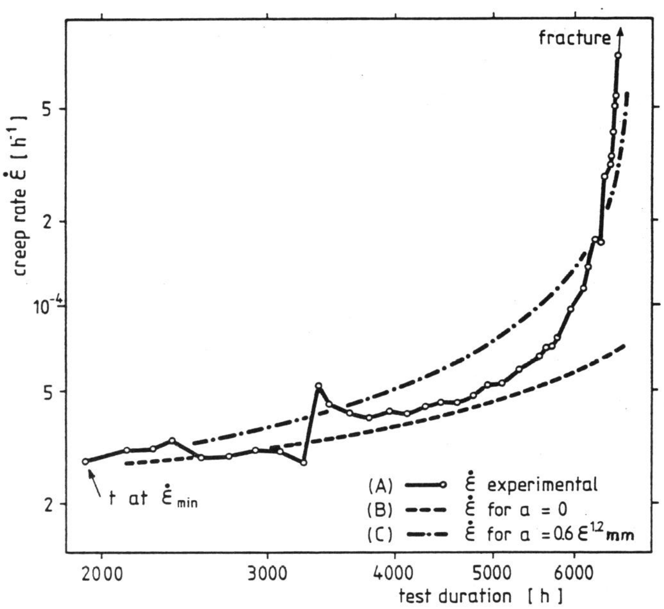

Fig. 6:
Comparison of measured creep rates (A) with rates calculated and assuming no surface cracks (B) or taking into account surface crack depth as measured for a_{90} (C, Equ. (3) of text)