WEB EXCLUSIVE VERSION

DAMAGING PROCESS IN CREEP-FATIGUE-OXIDATION ENVIRONMENT AND ITS CONSIDERATION IN LIFE PREDICTION

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In the actual development of energy production and utilisation facilities, there is an essential need in the use of significantly higher application temperatures, to meet the requirements of higher efficiency and frequent cycling of power plants imposed by the aimed energy source diversity. The existing approach to high-temperature creep-fatigue life prediction used by many design codes, the simple linear-damage rule, is not sufficient in this respect. The critical point of this method is the significant reduction in fatigue and/or creep lifetime capacity and reliability of results. The investigations for creep-fatigue loading conditions at very high temperature with the Alloy 800H have shown that the oxidation due to grain boundary sensitisation defines an important additional damage mechanism which has to be considered in life time prediction under these conditions. The simple modification of the conventional Strain Range Partitioning method accounting for the effects of exposure time and creep damage has not led to a necessary improvement in this respect. In this work introduces is other well-founded representation of time-temperature effects. By the method, excellent results have been achieved in the life prediction of all cyclic dwell history tests in the relevant temperature range between 750 to 850 °C, for the material used.

Key Words: Creep-fatigue, oxidation, life prediction, SRP correction, test acceleration, non-isothermal Loads

1. Introduction

Under the condition of actual development in the area of energy production and utilisation facilities, involving a great variety of components such reactors, gasifiers, pipelines, heat exchangers and many other power-generation structures, there is an essential need in the utilisation of significantly higher application temperatures, to meet the requirements of higher efficiency and in the frequent cycling of power plants imposed by the aimed energy source diversity. Within the combined load-temperature-time environment, high temperature becomes a major parameter influencing deformation and damage behaviour both concerning creep and fatigue loading conditions and their interaction. Under these circumstances using conventional methods results in significant limitations and drawbacks concerning reliable practical applications. As a consequence the probability of failure of those components, if their expanded service condition are not properly considered will be, in contrary to their high economic and safety concern, significantly increased. To avoid this, more adequate, advanced lifetime prediction procedures, which are based on the availability of a suitable material, efficient design and adequate testing methods, are necessary.

The efficiency of boiler steam turbine fossil power plants is a strong function of steam temperature and pressure. Currently, the conventional coal-fired electricity generating plants are typically 35 to 40% efficient. By developing power plants capable to work at higher temperatures, it is possible to boost efficiencies up to 50%. These efficiency gains, alone, would cut the release of CO_2 and other emissions by nearly 30 percent. To this end, a wide range of electricity-generating technologies, as supercritical and ultra-supercritical, integrated gasification combined cycle, oxy-fuel combustion and ultra clean coal are available that could significantly reduce the greenhouse gas emissions. But, to reach a target steam condition of so-called ultrasupercritical equipments at 760° C/35 MPa, for example, a significant effort on the material side is required.

Fully new solutions, their introduction in the practical application is expected in the 21st century, require, however, even higher temperatures. To get the thermonuclear energy by fusion typical temperature ranges up to 1000 °K is expected. Fuel reforming and hydrogen production requires temperatures above 800° C and by applying the service temperatures of 1000 °C in Solid oxide fuel cells (SOFC) efficiency of 50 – 65 % could be achieved. Here, using the waste heat efficiency up to the 90 % is possible.

Traditionally the possible solutions could be based on the development of new materials and/or material substitution. The common reasons for materials substitution are to improve service performance, including longer life and higher reliability, to meet new legal requirements (accounting for change operating conditions) and taking advantage of new materials or processes, to reduce cost and making the product more competitive. Both new material and material substitution, however suffer on long testing and insufficient data for design.

Generally, a simple substitution of one material for another does not produce an optimum solution. This is because it is not possible to realize the full potential of a new material unless the component is redesigned to exploit its strong points and usable characteristics. To this end selection and understanding of material behaviour and the development of new advanced methods extending the range of materials application is a vital part of the development of future advanced power facilities.

2. Current life prediction methods and their limitations

The existing approach to high-temperature creep-fatigue life prediction, which is used by many design codes, is the simple linear-damage rule and its variation [3]. It was originally developed from tests on austenitic stainless steels for which, according to the actual use, tensile creep damage (intergranular cavitation) rather than environmental effects plays a major role in determining creep-fatigue life. It is assumed that the damage is linearly accumulated and thus can be superposed no matter whether it is fatigue or creep damage. An acceptable design should satisfy the following criterion:

$$\sum_{j=1}^{p}\left(\frac{n}{N_d}\right)+\sum_{k=1}^{q}\left(\frac{\Delta t}{T_d}\right)\leq D \qquad (1)$$

where

$\sum_{j=1}^{p}\left(\frac{n}{N_d}\right)$ and $\sum_{k=1}^{q}\left(\frac{\Delta t}{T_d}\right)$ correspond respectively to the fatigue and creep damage summation under different types of loads, during a specific period of service time.

Unlike to the original linear model, for taking into account nonlinear damage accumulation effects based on experimental evidence, the critical value of fatigue-creep damage D is not specified as a constant equal to 1 but as a variable dependent on the creep and fatigue damage fractions and actual material. The corresponding bi-linear dependence is given graphically in the form of a creep-fatigue interaction envelope (Fig. 1), which for design purposes shall not be exceeded by the sum of creep and fatigue damage fractions. The different envelopes given in the current codes are limited to five materials: 304 and 316 steels, 2¼Cr1Mo and alloy 800H and a very restrictive line for 9Cr1MoV, where the intersection point is extremely low (0.02, 0.1) indicating a supposed very severe creep-fatigue interaction.

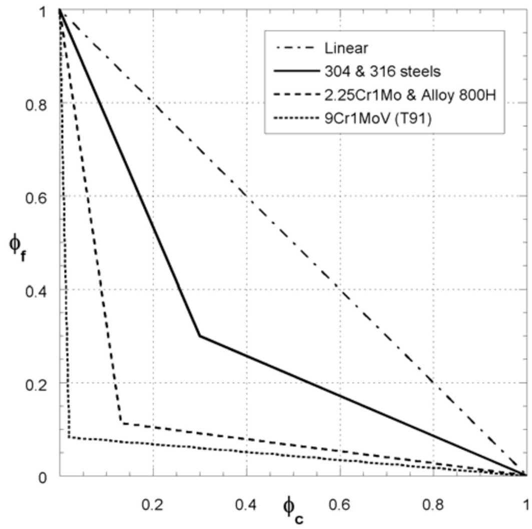


Figure 1: Bilinear damage summation curves for creep-fatigue assessment used in the current codes

The critical point of the bi-linear envelopes is the significant reduction in fatigue and/or creep capacity even though the quantity of the compliment damage portion is still very low. The origin of this phenomena, however is not very well defined and, because of this, the management of the corresponding design appears very difficult.

The investigation in [1] carried out on Alloy 800 H showed that increasing the tensile relaxation hold time has a strongly detrimental effect on the cyclic life of specimens (Fig. 2). The corresponding increased creep damage should be demonstrated by adequate addition in creep deformation. The relaxation lines taken within of the same test programme (Fig. 3) however, show a clear trend to the reduction in the relaxation amount in the dependence on holt time, so that the corresponding increase in creep deformations becomes too low to give an explanation for the apparent addition in creep damage. The essential feature of creep, i.e. the significant time-dependent plastic deformation is here entirely absent.

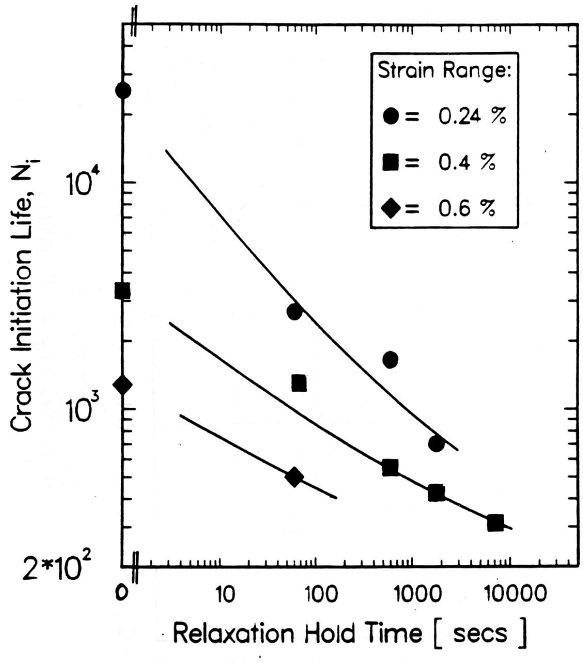


Figure 2: Reduction in crack initiation life for different dwell times for Alloy 800 H

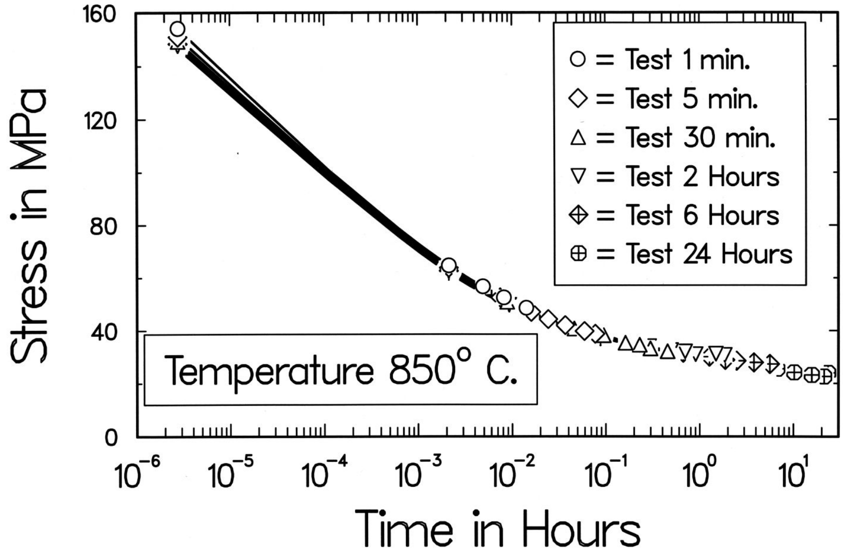


Figure 3: Relaxation lines for different tensile hold times

In addition, even though the creep loading in the creep-fatigue test in the bulk material did not exhibit either grain boundary cavitation, void formation, or grain boundary cracking typically associated with creep damage the resistance to the fatigue and the corresponding fatigue lifetime capacity were significantly reduced.

This extreme deviation related to the such characteristic signs of the creep damage indicates the existence of the additional mechanisms which enhance the overall damaging process and reduce the fatigue life and which evidently are not properly taken into account by the linear summation rule. With the increasing application temperature these additional mechanisms dominate the damaging process and if not correctly considered may unexpected limit the material application.

Due to its simplicity of use, however the original linear damage rule has been forced in the application of the codes and as the envelopes in Fig. 1 show, more or less force-fitted for different materials to allow for strong interaction features and probable, but not recognised and calculated additional effects accompanying damaging process at high temperatures.

There are other life predictive models, such as Damage Rate Equations, Strain Range Partitioning Equations, the Frequency Separation Equation, and the Ductility Exhaustion Equation, which are appreciated to be potentially more accurate. In spite of their well-founded expectation, however, the practical application shows that all of the current life prediction models are able to predict life only with limited accuracy.

Between knowing models for example, Strain Range Partitioning (SRP) is considered to be most directly adjusted to the fatigue-with-dwell situations. One of the intended advantages of this model is the insensitivity of the basic damage relationships and their material-related constants to temperature changes. The insensitivity on temperature, however, does not mean that the lifetime is not dependent on temperature. The well known strain rate dependence on temperature level for equal load conditions causes that at higher temperatures the amount of inelastic strain significantly increases creating the corresponding reduction in the lifetime.

The main improvements of the basic SRP method are limited to the consideration of cyclic conditions for the creep loading and the possible interaction if the creep strain in the cycle is compensated by plastic strain. The rest of the method is based on the same superposition principle as in the case of linear damage rules. Based on this the improvements are poor and insufficient to cover the increased complexity in the properties evaluation and calculation.

In fact, it has not been possible to identify the most satisfactory model for the problem of life prediction of industrial components subjected to creep-fatigue for materials in use in the varying high temperature environment. As a consequence the, linear-damage rule, with its relative ease of application over a wide range of conditions and with the huge availability of application data, still gives the accepted basis for life assessment in actual design codes.

For the purposes of the increased application temperature and new materials, however it is obvious that a fundamental understanding of the damaging processes under higher temperatures does not exist. This indicates a weak point in the failure criteria in many current codes that hamper the safe and economic use of materials. In addition, a more robust and perhaps material or material class specific life prediction methodology should address the fact that different mechanisms likely exist in different classes of materials, requiring different approaches to get the efficient solution(s).

3. Experience concerning additional effects and conditions

"Life of a component" is the essential quality for practical application defining the time during which the component performs its planned function safely, reliably and with acceptable economical values.

The main approach in lifetime prediction is resting on the consideration of the component life necessary for the so-called crack initiation and propagation life, expressed as cycles required to propagate the initiated crack to the failure of the component. The growth phase starts when the crack becomes observable. This crack initiation concept is rather based on a rational simplification of a more complex situation necessary to solve practical problems than on a precise determination of the relevant damage phenomena.

The main advantage of this simplified approach is the possibility to consider residual crack growth life, i.e., the portion of life after crack initiation, using a fracture mechanics approach without methodological restrictions or uncertainties, for example to allow for short crack behaviour, multiple cracks or cracks in the residual stress field at notches. For these purposes, the crack after "initiation" has to be geometrically well defined and, to be "selected" from the other micro cracks, material defects and so on, as representative for the onset of crack growth, has to propagate continuously with the crack growth rate in dependence on the actual crack size.

For engineering design, crack initiation lifetime is commonly defined as crack nucleation and growth of small crack up to the NDT detectable crack level. Short cracks usually start by growing faster than long cracks with similar ΔK's and then slow down as they get longer, occasionally retarding completely (Figure 4). In fact, this for many researchers "anomalous behaviour" is "normal behaviour" of material which is able to overcome, adapt and forget the structure anomalies and the corresponding abilities should not be ignored. Small-crack behaviour is not an important issue for applications in which initial defects are large, i.e. for many traditional mechanical and aeronautical designs and analyses based on damage tolerance concepts, because the initial flaw size and/or conventional NDE inspection limits are usually beyond the small-crack regime.

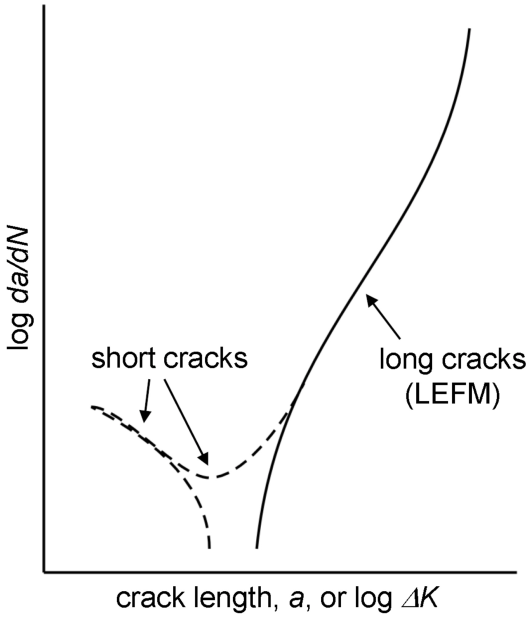


Figure 4: Different behaviour of short and long cracks

The design approach based on crack initiation also assumes the existence of a "perfect" material, free of flaws, where, under the impact of operational load conditions cracks initiate and grow and the crack initiation life is defined as the lifetime required to produce, by this means, a crack of, even though controversial, "engineering size". Opposite to the "perfect materials" is the approach of "bed materials", where the initial crack always exists (mostly the case with welding). In fact, it is always on safe side to neglect this period and calculate a non-zero incubation time assuming that crack growth occurs on the first loading. The assessment will be then conservative. However, it may be too conservative for use as a design criteria. The cracked component is usually further used, but it continually losses some of the important quality features concerning the properties as the residual strength and reliability and its use can be additionally limited, for example, if the end of life is leakage (especially in case of dangerous mediums). Its further use is mostly dominated by economic aspects.

For creep-fatigue conditions at high temperatures the above classification of the stages of damage development, which is based on fatigue experience, is further complicated by the introduction of the creep mechanisms and their interaction with the fatigue processes. Creep deformation involves grain boundary sliding with void nucleation and growth in the bulk of the material. Therefore, the combination of local crack growth and creep damage development throughout the material volume (by void nucleation and growth) makes the separation and corresponding treatment of crack initiation and crack growth for the purposes of the creep-fatigue life assessment extremely difficult. Additionally, crack growth under creep-fatigue conditions is accelerated by creep and environmentally induced internal damage in the region ahead of a crack tip. Because of this, in the high temperature range, the "local event" approach in life analysis, as in the case of fatigue crack initiation, has to be extended to a more general approach based on damage incubation [1] and considering both interacting processes of a local crack development and early growth, as well as bulk damage.

In spite of the significant difficulties in defining the conditions for crack initiation the life prediction within of the design of new structures according to the different codes (R5, 2-criterion method, etc.) is based on crack initiation, where residual life evaluation is founded on growth of cracks or similar damage features discovered by NDT during operation. For the life prediction based on crack initiation crack growth works as additional safety margin covering at the same time some scatter. When the crack growth is based on discovering crack of actual size the analyses require additional own safety margins.

The applicability of crack initiation as a basic design concept is controversial, but it is, based on practical experience, a reasonable engineering assumption that has allowed the economic life assessment for structural components over the years. Models for the accurate evaluation of the propagation of microscopic sized shorts cracks, especially under high-temperature creep-fatigue-oxidation interaction conditions are still under development.

4. Material behaviour under oxidation in creep-fatigue regime

As already shown (Fig. 2) the investigations for creep-fatigue loading conditions at very high temperature (850 °C) with the Alloy 800H demonstrated that the addition of a tensile hold time at maximum load to a fatigue cycle tends to significantly reduce the lifetime. Nickel-base alloys such as alloys 600, 690 and X-750, but also iron and cobalt alloys are highly susceptible to grain boundary deterioration. The related time-dependent intergranular cracking of superalloys at high temperatures in air is dominated by environmental interactions at the crack tip. Grain boundary chemistry, therefore is shown to be a significant factor that considerably affects strengthening, weakening, and damage behaviour of corresponding materials [4].

Based on the results [1] it becomes evident that the oxidation due to grain boundary sensitisation by second phase development defines an important mechanism of damage which has to be considered in life time prediction procedures under these conditions. The following microstructure features of thermally activated processes accompanying aging under load are of primary importance:

- Although austenitic steels like the alloy used (800 H) exhibit excellent bulk oxidation resistance grain boundary carbide precipitation takes place at high temperature. Precipitation is favoured at grain boundaries which are short diffusion pathways for oxygen. In all the loading conditions, the precipitates typically consisted of chromium-rich $M_{23}C_6$. During the process, carbon atoms migrate to the grain boundary from all parts of the crystal, whereas chromium is depleted from more localized regions, near the grain boundary, thus creating in the surrounding area of grain boundary an envelope of chromium depleted material (fig 5). Further along the grain boundary, a practically carbide free portion is seen, after which the unoxydised carbides being, as is normal for this alloy.
- The precipitation phase at grain boundary is a central site for oxidation because the composition differs from that of the matrix, local disorder makes it chemically more active than the matrix and its function as a rapid diffusion path so that oxygen diffusion along the grain boundary is much faster. Due to the grain boundary oxidation, therefore, which penetrates deeper than surface oxidation, accelerated fatigue crack nucleation (or "damage incubation") has appeared (fig. 6) and significant shortening in the lifetime of the cyclic dwell tests is found. The availability of oxygen in-depth of the crack is confirmed based on ED analysis [1] its results are shown in Fig. 7 (Interesting here is also the difference in Cr amount depending on the crack depth.) The developed oxides are brittle solids that have very low fracture strength and fracture toughness. Therefore, oxidation also accelerates fatigue damage both in terms of fatigue crack initiation and fatigue crack

propagation. Once initiated, the cracks will open, exposing new metal surfaces to the gas phase and oxidation of the crack. It is assumed that the materials' resistance to crack growth is reduced as oxygen in front of the crack tip diffuses into the grain boundary and then crack can pass through the damage zone.

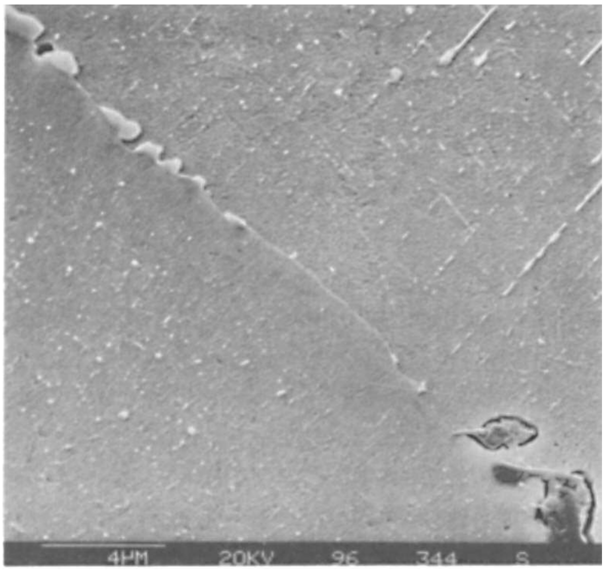


Fig. 5: Details of the crack tip region [2]

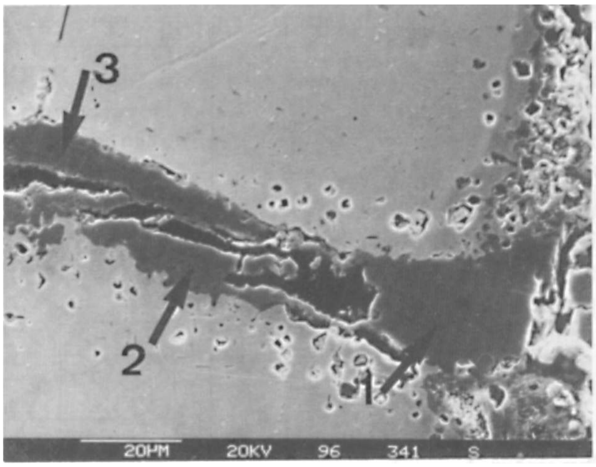


Fig. 6: Surface crack: detail positioning for ED analysis [2]
(850°C $\Delta\varepsilon = \pm 0.2$ $t_R = 2$ Hours)

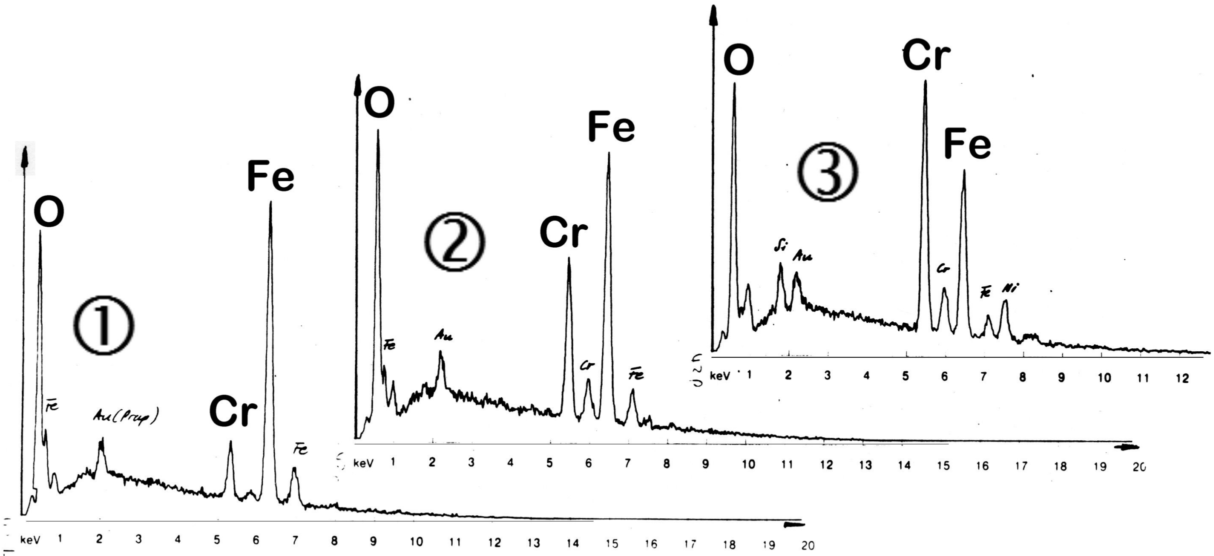


Fig. 7: Results of ED Analysis for points given in Fig. 6

The next picture (Fig. 8) completes the total image showing how deeply the reduction through the oxidation at long the grain borders penetrates. In the first phase the grain boundaries were not oxidised since they are covered with the massive Cr-precipitates and consequently are oxidation more resistant. Through the precipitation phase, however comes at the both sides of the grain boundary to a Cr- impoverishment, what favours the oxidation.

Irrespective of the loading condition, therefore the largest effect on life was exerted by oxidation processes. The oxidation also accelerates the growth rate of microcracks by enhancing the growth rate of individual cracks as well as through multiple crack coalescence.

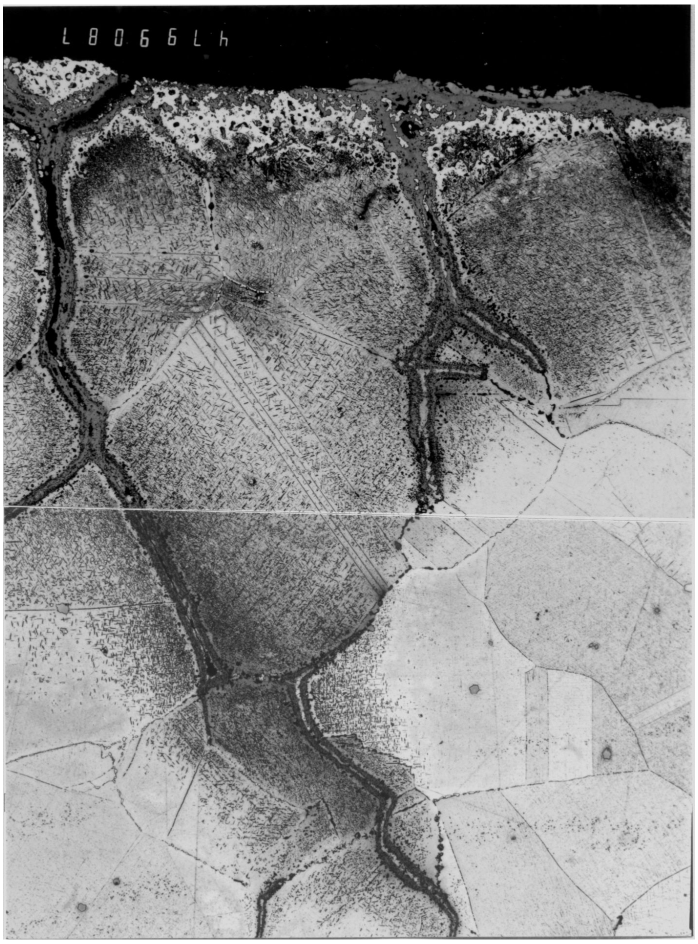


Figure 8: Oxidation penetration at grain boundaries

5. Oxidation consideration in life prediction

The earliest life prediction model to take into account environmental effects is the frequency-modified strain-life equation [3] which included, in addition to the inelastic strain range, a frequency factor which accounted for the effect of oxidation, but the primary mechanisms and parameters associated with the development of creep inelastic deformation and the corresponding damage has been not directly considered.

The improvement in this respect is given by the Strain Range Partitioning Method. Originally, it was developed to handle the life prediction of a strain cycle, containing both plastic and creep strain components. Unfortunately essential mechanisms and parameters associated with the oxidation were not directly included in the method formulation.

Recently, the conventional Strain Range Partitioning life relationships were modified to account for the effects of exposure time. This is accomplished by modifying the conventional Strain Range Partitioning life relationships using either Steady- State Creep Rate (SSCR) or Exposure Time (ET) [5]. The resulting SSCR- and ET-Modified Strain Range Partitioning life relationships correlate the creep-fatigue experimental data of 316 stainless steel at 816 °C, significantly better than the conventional Strain Range Partitioning life relationships. On the other hand, the both parameters are parameter standing in relationships and therefore the doubt still prevails whether hold-time effects have to be attributed to oxidation or creep damage in the material.

1.1. PROPOSED SOLUTION

Determination of the mean creep rate during the stress hold period of each cycle in our CP tests [1] leads to an examination of the change in this rate over the duration of the test (Fig. 9). In the first phase the creep rate is not constant with the slight linear increase, indicating that the number of the cavities increases linearly with time. Near the end the acceleration in creep rate is caused by cavity growth, coalescence and cracking. The linear portion of the curve in Fig. 9 represent the period of "damage incubation" in the bulk of the material and the subsequent transition to an unstable increase in creep rate can be used as a more accurate criterion of "crack initiation" for creep-fatigue conditions.

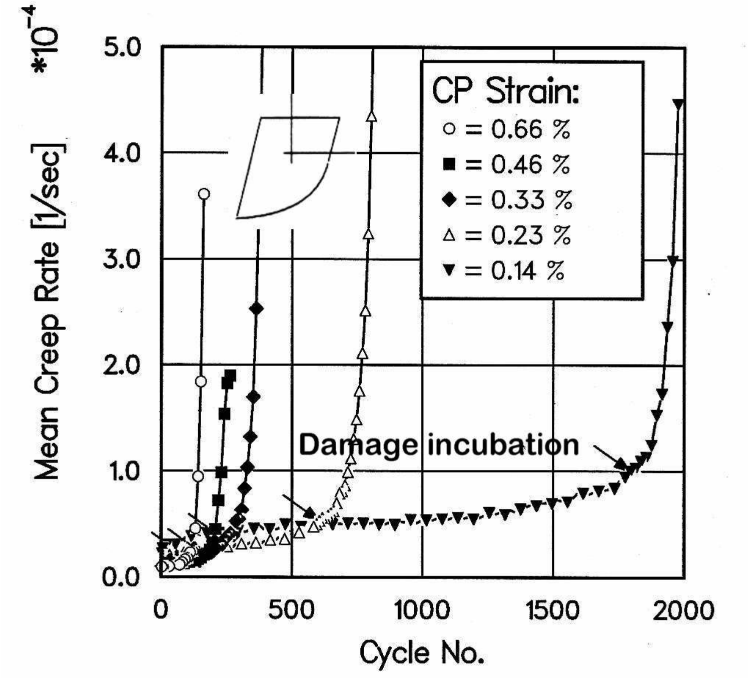


Figure 9: The mean creep rate during the stress hold period for CP Tests on Alloy 800H

Oxidation at the crack tip contributes to the creep crack growth rates in the low temperature range (650°C), where the crack growth rates are small. In the high temperature range (760°C), the acceleration of creep crack growth due to oxidation is negligible. The kinetics of cavity growth are mainly controlled by a power law creep deformation.

In contrast, concerning the fatigue damage development, a grain boundary oxide crack develops into a fatigue crack nucleus or a pre-crack which shortens the crack nucleation period and the total fatigue life. As shown in our experiments, this reduction in nucleation life is substantial, if the temperature is very high and the oxidation exposure time is very long.

Generally, the mechanisms concerning the oxidation at high temperature and especially the grain boundary oxidation have been very intensively treated and there is many literature sources concerning the subjects, but the quantitative measures for the application in life assessment are still scarce.

Therefore, extending the formulation of the very popular SRP life prediction model to incorporate the effect of surface cracking due to environmental effects need to be developed for the purposes of accurate lifetime prediction. Based on the our investigation we decide to introduce another more logical correction in the SRP method.

The phenomenological basis of the conventional SRP method can be represented as follows: During holding periods and time-dependent stress change in addition to plastic deformation (P) relevant creep (C) strains are occurring. Simultaneous creeping at tension and pressure side results in CC-strain partition. The within one cycle not balanced creep strains yield either CP or PC partitions, why within a same load loop just either tension or pressure partitions can occur as a result of creep and plastic strain blend.

$$\frac{\Delta\varepsilon_{in}}{N_f}=\frac{\Delta\varepsilon_{pp}}{N_{pp}}+\left(\frac{\Delta\varepsilon_{cp}}{N_{cp}}\ or\ \frac{\Delta\varepsilon_{pc}}{N_{pc}}\right)+\frac{\Delta\varepsilon_{cc}}{N_{cc}} \tag{2}$$

To take into account the effects of the enbrittled surface layers, aroused through the oxidation that penetrates alongside the grain boundaries, on the crack initiation, we introduce a correction [1, 2] into the known Strain Range Partitioning method, which is then in the case of the tensile relaxation dwell determined by the relationship:

$$\frac{1}{N_i}=\left(\frac{f_{PP}}{N_{PP}}\right)\cdot\alpha+\frac{f_{CP}}{N_{CP}} \tag{3}$$

With $\mathbf{f}_{PP}=\Delta\varepsilon_{PP}/\Delta\varepsilon_{in}$ and $\mathbf{f}_{CP}=\Delta\varepsilon_{CP}/\Delta\varepsilon_{in}$.

Note that the correction relates solely to the reduction of fatigue life or respectively crack formation on the surfaces. Herein, the correction-factor (Alfa) can be represented through the relation of the Arrhenius-Type, that takes into account the dependence on the temperature and time

$$\alpha=\mathbf{A}\cdot\mathbf{t}^{n}\cdot\mathbf{exp}\left(-\frac{\mathbf{Q}}{\mathbf{R}\cdot\mathbf{T}}\right) \tag{4}$$

For simplification and to get the clear relationships the effect of N_{CC} portion of life is excluded from the investigation. The total time spend by creeping at constant load is relatively small compared to the dwell effect of relaxation at constant strain.

For the evaluation of the correction factor, a series of test (16 specimens) at a temperature in the range 750 to 850 ° C with dwell time up to 2 hours were performed (Figure 10). Based on 850 °C results only, the regression analysis gives:

$$\alpha=\mathbf{a}\cdot\mathbf{t}^{n} \tag{5}$$

where

a = 0.9751 and n = 0.35213 (with R^2 = 0.99283)

Assuming exponent **n** found is valid for all temperatures the values for different temperatures has to be scaled by the relationship

$$\alpha_{SC}=\alpha_T\cdot\left(\frac{\tau_{850}}{\tau_T}\right)^{n} \tag{6}$$

Based on data for $\Delta\varepsilon$ = const. only (Figure 11) the next regression analysis could be performed leading to the general solution

$$\alpha=\mathbf{6{,}515\cdot 10^{5}}\cdot\tau^{0,352}\cdot\mathbf{exp}\left(-\frac{\mathbf{Q}}{\mathbf{R}\cdot\mathbf{T}}\right) \tag{7}$$

where Q = 125,23 kJ/gr, and R = 8,314 J/gr

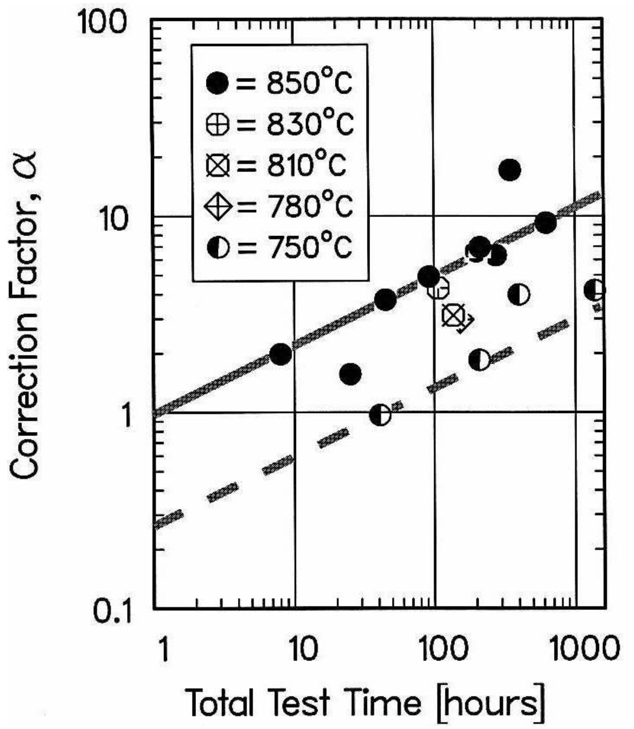


Figure 10: Evaluation of the time dependence of the oxidation correction factor α

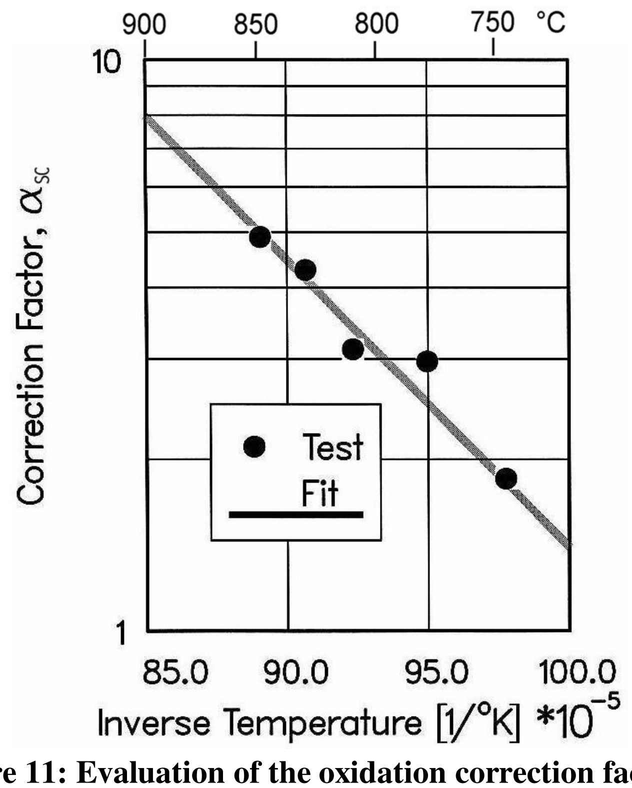


Figure 11: Evaluation of the oxidation correction factor α

6. Test Verification of the approach

The proposed method was verified with help of own experimental examinations (shown in Table 1) and the results of examination from the literature, that refers to the conditions with dwell time under tensile loading with the same material. In this way, also the numerous influences of the material-charges, heat treatment and test-conditions has been fully included.

By comparison with the prediction in Fig. 10 and 11 the fitted lines according to (7) has been shown to be valid over a range of temperature.

It is clear that the verification of any new design method requires a lengthy process, involving evaluation of different practical evidence and consideration of a number of possible random or less predictable influences, such as material charge-to-charge variability, basic data scatter and component related effects. Nevertheless an additional evidence about the accuracy of the developed method and about the influence of a number of these factors can be accounted for by investigation, prediction and comparison of available literature data for Alloy 800H. For this a broad literature search has performed to find relevant results, although these were limited by the fact that only test results with creep-fatigue interaction effects, as for example tensile strain hold tests, could be used.

Accordingly, in addition to the own verification tests carried out, the following dwell time tests were found to be applicable for this analysis:

Source:	No. of test
INTERATOM [8]	16
Jaske at all [7]	8
IABG/MAN [6]	6

It can be seen that the number of results found is only two times larger than the test results produced in the project. From the available data the longest test result is also nearly two times shorter than our own longest verification tests.

a. b.

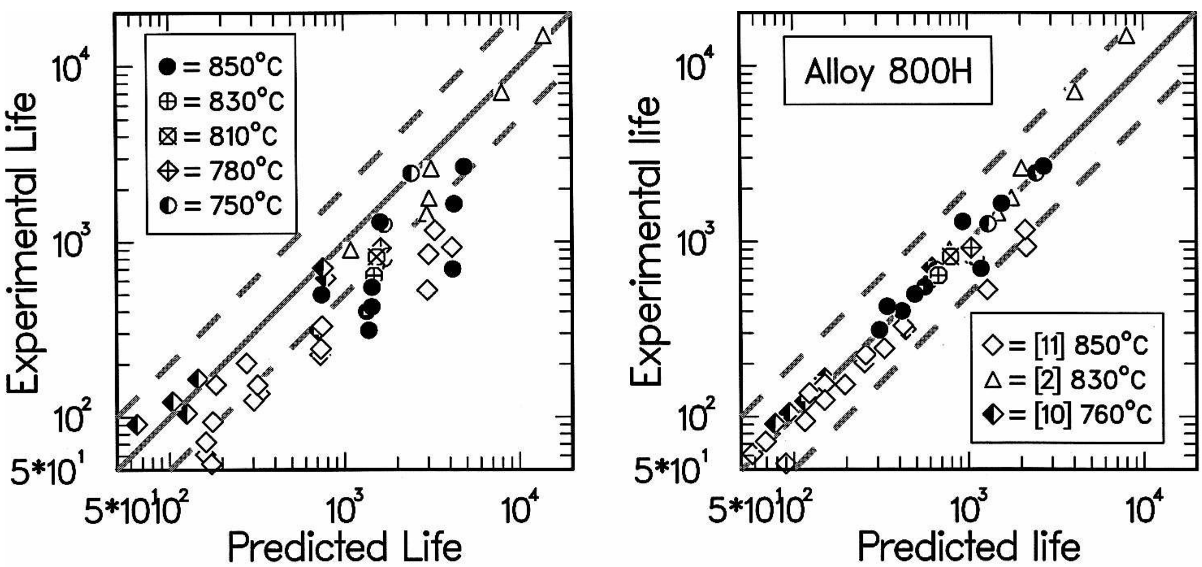
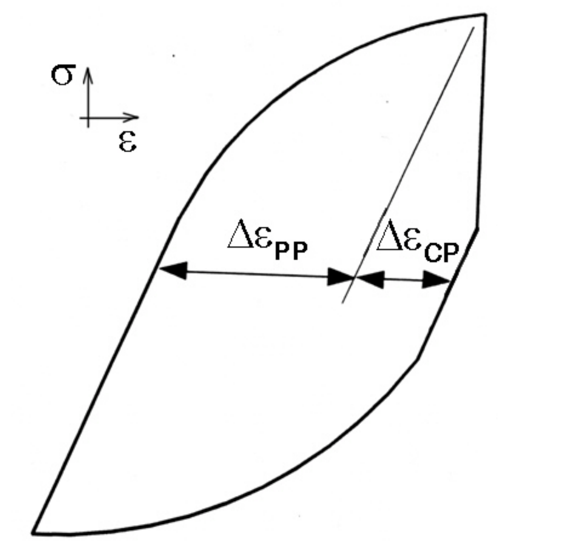


Figure 12: Comparison of the life prediction using (a) the standard SRP method and (b) using modified SRP method with the oxidation correction α

At first the life prediction was performed for all data using the conventional SRP method. The results (Figure 12, a) show large scatter of the data and a trend of non-conservative predictions which becomes even more critical as the total test time increases. On the other hand, the predictions using-the method developed by us (Figure 12, b) are seen to be very accurate, despite the fact that the data comes from different material charges of correspondingly varying specifications. All the predictions for these tests are within the 1:2 life margin, with only two results outside this margin. Since this result is not for a longer term test, its deviation can be explained as pure scatter and not indicative of any systematic effect. No trend with regard to test duration can be recognized. This demonstrates the high extrapolation capability of the method. Although, because of the time limitations in this program, the longest our verification test time of "only" 1300 hours may appear low, it must be bear in mind that the longest duration tests reported in the literature (at 760°C) have only half this value. This may give an impression as to how difficult and expensive cyclic tests of this kind are.

Table 1: Verification tests on Alloy 800H in the temperature range 750 to 850° C [1]



How the experimental data presented in this table were obtained is described in all details in [1]

$\Delta\varepsilon_{CP}$ are produced during dwell time T_R at constant strain

All tests are carried out at constant temperature given in the table.

The basic SRP data are generated by separate test described in [1]:

$$\Delta\varepsilon_{PP} = 0.48 \cdot N_{PP}^{-0.648}$$

$$\Delta\varepsilon_{CP} = 0.0839 \cdot N_{CP}^{-0.535}$$

Test	Temp °C	$\Delta\varepsilon_{PP}$ %	$\Delta\varepsilon_{cp}$ %	t_R secs.	N_i	N_{SRP}	α	N_α
IR38	850	0.395	0.071	60	500	741	1.980	490
IR37	850	0.213	0.060	68	1300	1606	1565	928
IR43	850	0.217	0.067	1795	425	1434	6.930	341
IR66	850	0.236	0.065	1795	400	1345	6.623	414
IR45	850	0.230	0.065	7192	312	1380	9.259	308
IR40	850	0.075	0.047	60	2685	4843	3.745	2698
IR41	850	0.082	0.051	600	1650	4244	6.329	1551
IR46	850	0.086	0.051	1795	700	4153	17.241	1180
IR39	850	0.223	0.064	600	550	1440	4.902	559
IR72	830	0.212	0.066	600	640	1480	4.292	671
IR73	810	0.212	0.063	600	820	1533	3.115	789
IR74	780	0.195	0.065	600	920	1621	2.967	1040
IR76	750	0.168	0.047	60	2460	2423	0.966	2444
IR75	750	0.196	0.062	600	1260	1690	1.848	1298
IR78	750	0.184	0.066	1800	800	1684	3.984	1105
IR84	750	0.199	0.069	7200	690	1514	4.188	651

By the α-correction of SRP-method the amounts which relate to insufficient methods of calculation (classical SRP, linear damage rules, etc.) were sorted out from the natural scatter and the resulting scatter reduced to the real random effects.

7. Discussion of results

Overall the use of the proposed method produces an excellent result when it is considered that the data included in the analysis covers a number of different material lots, treatments and testing conditions.

In addition, the application of the method provides different practical advantages. Modern life analysis involves analysing times-to-failure data obtained under normal operating conditions to quantify the life characteristics of the product, system or component. In many situations, and for many reasons, such life data (or times to-failure data) is very difficult, if even impossible, to obtain. Given this difficulty, and the need to observe failures of products to better understand their failure modes and their life characteristics, design practitioners have attempted to devise methods to force these products to fail more quickly than they would under normal use conditions (accelerated test sequence). In other words, they have attempted to accelerate their failures. For this purpose critical environmental factors such as temperature, stress, time endurance and cycles, etc. and the corresponding interaction effects have to be incorporated. Based on the time-temperature correction α this kind of life management can be successfully applied [2].

According to the original SRP method the basic lifetime relationships are not dependent on temperature, but only on stress-strain history, which has to be calculated according to the material inelastic stress-strain behaviour and component geometry. Therefore, increasing the temperature of the test leads according to (7) to

$$\tau_1^{0,352} \cdot \mathbf{exp}\left(\frac{-\beta}{\mathbf{T}_1}\right) = \tau_2^{0,352} \cdot \mathbf{exp}\left(\frac{-\beta}{\mathbf{T}_2}\right)$$

$$\left(\frac{\tau_1}{\tau_2}\right)^{0,352} = \mathbf{exp}\left(-\frac{\beta}{\mathbf{T}_2} + \frac{\beta}{\mathbf{T}_1}\right)$$

where

$\boldsymbol{\beta} = \mathbf{Q/R} = \mathbf{15062}$ (for $\mathbf{Q} = \mathbf{125{,}23\ kJ/gr}$ and $\mathbf{R} = \mathbf{8{,}314\ J/gr}$)

If the service temperature is 740 °C increasing the test temperature for 60 °C will give the relationship

$$\tau_1 = \mathbf{0{,}094} \cdot \tau_2$$

This means the accelerated exposures to oxidation which generates the microstructural damage, by increasing the test temperature for 60 °C, allows the reduction of the test duration (this means hold time in the cycle) of one order of magnitude. In this way the proposed method developed based on the consideration and understanding of materials response and degradation mechanisms can help to reduce the amount of experimental effort in the design of components for operation under creep- fatigue conditions and their optimisation and efficient redesign.

The significance of the corresponding possibilities cannot be sufficiently appreciated. The current codes should be applicable to the design life up to 30 years even though it is impractical to conduct such long-term tests in any types of testing.

The actual transient events occur over a temperature range. In contrast to this the current Code rules and the available methods are based on isothermal data. Because of this there are also differences of opinion relative to the use of isothermal data for predicting transient events.

The relationship (4) represents a correction as the function of time and temperature. Based on this the proposed model is fully suitable under non-isothermal conditions, i.e. for the purposes of thermal-mechanical creep-fatigue evaluation. The actual transients can be considered integrating the equation (4) according to the temperature change in time

$$\alpha_{ef} = a \cdot \int_{o}^{\tau} \tau^{n} \cdot \exp\left[-\frac{\beta}{T(\tau)}\right] \cdot d\tau$$

If the temperature history can be reduced to a limited number of different levels the resulting correction factor equals the sum:

$$\alpha_{ef} = \sum_{i=1}^{p} \alpha_{i}$$

To consider the variation in the end life time the end results according to (3) must be produced by iteration.

The experimental verification for the treatment of the non-isothermal conditions, however, are very complex and has been not performed.

8. Conclusions

Because of its simplicity of use, the linear damage rule has been force-fitted for different materials in current codes even though their high-temperature fatigue life is influenced more by oxidation assisted surface cracking than by bulk creep damage. Extrapolation to high temperature conditions and beyond the database using linear damage rule cannot be justified. Specialized improved, more reliable models for predicting surface cracking due to environmental effects need to be developed, validated, and integrated into the design process.

Extending the formulation of the very popular SRP life prediction model to incorporate the effect of environment is most desirable. Some modification has been made to SRP, but it has not so far led to a significant improvement in the life predictions. Based on the results of our investigation we decide to introduce other more logical analytical representations of time effects in the SRP method through the correction factor Alfa. For this purpose the correction-factor (Alfa) can be represented through the relation of the Arrhenius-Type, that takes into account the dependence on the temperature and time

By introducing the above time-dependent environmental correction in the SRP equations, excellent results have been achieved in the life prediction of all cyclic dwell history tests in the relevant temperature range between 750 to 850 °C, for the material used (Alloy 800 H). Additional evidence about the accuracy of the developed method has been accounted for by investigation, prediction and comparison of available literature cycle tensile dwell data for Alloy 800 H. The use of this method produces very accurate results for all data included in the analysis which covers a number of different material lots, treatment and testing conditions.

References:

[1] Final Technical report, BRITE Prpoject P-1209: Lifetime Prediction and extrapolation Methodologies for Computer-Aided Assessment of Component Service Behaviour Under Stress at High Temperature. Proj. Coordinator MAN Technologie AG, Munich, July 1986.

[2] Agatonovic, P. : High Temperature Lifetime Management Using Time-Temperature-Substitution, in Proceedings of the Fifth Int. Conf. on Creep and Fracture of Engineering Materials, Ed. By B. Wilshire and R.W. Evans, April 1993, The Institute of Materials, London.

[3] Shah V.N., S. Majumdar, K. Natesan: Review and Assessment of Codes and Procedures for HTGR Components Prepared by Argonne National Laboratory NUREG/CR-6816 Date Published: June 2003

[4] BHANU SANKARA RAO, H. SCHUSTER, and G.R. HALFORD: Mechanisms of High-Temperature Fatigue in Alloy 800H Metallurgical and Materials Transactions A Volume 27A, April 1996, S.851

[5] Kaluri S., S.S. Manson and G.R. Halford: Exposure Time Consideration in High Temperature Low Cycle Fatigue, in Mechanical behaviour of materials-V. Proceedings 5th ICM, Beijing China, 1987, Pergamon Press

[6] Agatonovic, P. and Heidenreich, R. In Low Cycle Fatigue, Eds. Solomon, H.D. et al. ASTM STP 942. 1987, p.500-518.

[7] Jaske. C.E., Medlin, H. and Perrin, J.S. - Influence of Hold Time and Temperature on the Low-Cycle Fatigue of Incolloy 800, ASME Paper No. 71-WA/PVP-7.

[8] Meurer, H.P. et al - Untersuchungen zum Ermüdungsverhalten von HTR-Werkstoffen bei Temperaturen bis zu 1000° C, Metallische Werkstoffe, Schriftenreihe "Energiepolitik in Noordrhein-Westfalen", Band 14, 1982, p.154-167.

[9] K. Natesan1 at all. Code Qualification of Structural Materials for AFCI Advanced Recycling Reactors, ANL-AFCI-244, Prep. U.S. Department of Energy, Nuclear Engineering Division, Argonne National Laboratory, September 2008