Low cycle thermomechanical fatigue of VVER-440 reactor structural materials

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Budapest - Baja
2016
1. INTRODUCTION TO THE RESEARCH TOPIC AND OBJECTIVES

Various damage processes can occur in the materials of mechanical structures due to different loading situations during operation, and this can lead to an ongoing reduction in their safety margin. Estimating the remaining lifetime of affected components is only possible if the actual condition of their structural materials is known.

This is particularly the case for power generating structures subjected to a cyclic variation of intensive thermal and mechanical loads. Many countries worldwide are intending to extend the lifetime of their operating nuclear reactors. In Hungary, four Russian-designed VVER-440/V-213 pressurized water reactor units are in operation in Paks Nuclear Power Plant, and the owner of the plant is already implementing plans for operating these reactors beyond their design life. This requires estimation of the remaining lifetime of all the key components, taking into account the damage processes that affect them.

One of the possible damage mechanisms in pressurized water reactors is the low cycle thermomechanical fatigue (TMF) caused by simultaneous thermal and mechanical loading during transient operating processes (such as startup and shutdown) or accident conditions. During the thermal fatigue of the pressurized water reactor structural materials, the cyclic heat load (or simultaneous heat and mechanical loads) causes alternating plastic strain in the materials of the pressurized components as follows:

a) in the layers near the surface of the component due to cyclic temperature and pressure loading,

b) in the inside layers of the component due to the inhomogeneous temperature field,

c) in the cross section of the pipes or vessels due to inhomogeneous temperature of the flowing media (e.g. thermal stratification).

When considering the safety of nuclear power plant operation and a possible lifetime extension, the key component is the reactor pressure vessel. It is essential that the integrity of the reactor pressure vessel is ensured during normal and off-normal operating conditions. The low-cycle fatigue transients listed above can often be responsible for reducing or even eliminating the safety margin of a reactor pressure vessel in terms of its usage factor; they may therefore be life-limiting factors in long-term operation. Consequently, knowledge of low cycle fatigue (LCF) degradation phenomena has become increasingly important to the operation of nuclear power plants beyond their design life. However, there have been no TMF studies concerning the base metal of the pressure vessel of this type of reactor, and only limited data are available on the fatigue behavior of this component’s cladding material.

My motivation in carrying out this study was to develop a sophisticated test facility for the GLEEBLE-3800 thermomechanical physical simulator, which is able to perform
TMF tests under the in-service temperature conditions of VVER-440 nuclear reactors. The study focused on the reactor pressure vessel's base metal 15Ch2MFA and anticorrosive cladding alloy 08Ch18N10T. My goal was to develop a novel low cycle fatigue model, based on the advanced material test methods carried out for these steels, which can be used for the remaining lifetime evaluation and design of power plant components. To study the microstructural evolution of the materials during the fatigue process, transmission electron microscopy (TEM) and X-ray diffraction (XRD) investigations were performed using samples at different stages of fatigue life; this can provide knowledge on the kinetics of the fatigue process. My intention was to relate the investigated cyclic mechanical behavior to the dislocation substructure resulting from the TEM observation together with X-ray diffraction analysis of samples at different stages of fatigue life.

2. BACKGROUND AND RESEARCH METHODS

Traditionally, isothermal (ISO) low cycle fatigue tests have been used to assess the performance of materials subjected to thermomechanical fatigue. However, mechanical properties of crystalline materials are a function of temperature, and so basic crystalline deformation mechanics need not necessarily be identical during ISO and TMF testing. Under TMF conditions, the material is subjected to a more complex stress path compared to that with ISO testing. TMF tests will activate the anisothermal damage mechanism, simulating service-like conditions more closely than the ISO experiments. Reliable modeling of the fatigue durability of structures remains a major challenge in both mechanical engineering design and the evaluation of remaining lifetime. A lack of conservatism concerning the lifetime prediction for structures with the importance of nuclear reactor components can lead to catastrophe. An important issue with respect to the LCF of reactor components is the engineering description of the fatigue life. In general, fatigue life assessments may be based on stress, strain or energy parameters. The strain-based models are predominantly used to characterize LCF behavior of structural materials operating at elevated temperatures and high levels of stress and strain, as are observed in nuclear power plants during operation. However, during the LCF of reactor components, the strain changes in magnitude and cannot be summed. The energy parameters have several advantages: (a) both strain and stress parameters are included in these models; (b) these variables are independent of the direction as they are scalar value; (c) cumulative damage is easy to estimate; and (d) the model results are more suitable for transferring to the actual structure. The weakness of the existing LCF engineering models is that the damage parameter includes the dissipated heat generated during the tests, resulting in an inappropriately higher value for fatigue
toughness. This form of energy does not contribute to the damage process during fatigue, and so it should be eliminated from the engineering models. Only limited information is available in the literature regarding microstructural evolution during the fatigue process; furthermore, the investigations have been based either on qualitative (TEM) or on quantitative (XRD) evaluations, and this has made understanding the phenomena more difficult. Combining qualitative investigations with quantitative methods could result in a deeper understanding of the microstructural background of the LCF process.

A sophisticated test facility has been developed for the GLEEBOOK-3800 thermomechanical physical simulator, which is able to perform TMF tests under the in-service temperature conditions of VVER-440 nuclear reactors (Fig. 1). The test details are presented in Table 1. My aim was to observe the effect of thermal cycles on lifetime, and to compare precisely the isothermal behavior with thermomechanical tests under the relevant temperature conditions. The tests were run under a specified temperature profile with total strain-controlled mechanical cycles at different strain amplitudes.

**Table 1** – LCF test parameters

<table>
<thead>
<tr>
<th>Controlled parameter</th>
<th>15Ch2MFA</th>
<th>08Ch18N10T</th>
</tr>
</thead>
<tbody>
<tr>
<td>Total strain, $\varepsilon_{mech}$</td>
<td>0.6-2.0%</td>
<td>0.9-2.5%</td>
</tr>
<tr>
<td>Triangular, $R_{mech} = -1$</td>
<td>0.3-1.5%</td>
<td>0.9-2.0%</td>
</tr>
<tr>
<td>Frequency</td>
<td>0.083 Hz</td>
<td></td>
</tr>
<tr>
<td>Total mechanical strain amplitude</td>
<td>2.5x10^{-3}s^{-1}</td>
<td>3-6.7x10^{-3}s^{-1}</td>
</tr>
<tr>
<td>Total mechanical strain rate</td>
<td>2.5x10^{-3}s^{-1}</td>
<td>3-6.7x10^{-3}s^{-1}</td>
</tr>
<tr>
<td>Temperature</td>
<td>260 °C</td>
<td>260 °C</td>
</tr>
<tr>
<td>Temperature change rate</td>
<td>-</td>
<td>20 °Cs^{-1}</td>
</tr>
</tbody>
</table>

I developed a new low cycle fatigue criterion based on the stored energy that accumulates in the material’s microstructure during fatigue loading. The new damage
parameters are based on the assumption that only the stored part of the introduced mechanical energy causes changes in the microstructure, with the remainder not contributing to the fatigue process. Furthermore, the dissipated part may depend on the fatigue conditions, especially the strain rate, and this could affect the lifetime prediction accuracy of the applied models. The fraction of plastic work dissipated to heat was investigated to provide information on the thermo-mechanical behavior of the tested materials, which can be used as an input parameter of the new model. This parameter is a macroscopic quantity and is often referred to as the Taylor–Quinney coefficient ($\beta$). Its value is influenced by many microscopic mechanisms and depends on the strain amplitude and strain rate. The value of $\beta$ for the investigated materials was determined using an inverse method based on an experimental–computational analysis. The experiments were performed on a GLEEble-3800 simulator with uniaxial tension–compression loading with total strain control, under precision measurement of the temperature evolution of the gauge surface of the specimens (Fig. 2a). The experimental details are presented in Table 2. High strain rate deformation results in an adiabatic condition and $\beta$ can be determined directly from experimental data of the temperature, stress and plastic strain versus time.

Table 2 - Experimental details

<table>
<thead>
<tr>
<th>Controlled parameter</th>
<th>Total mechanical strain amplitude</th>
<th>Total mechanical strain rate</th>
<th>Temperature measurement</th>
<th>Sampling frequency</th>
</tr>
</thead>
<tbody>
<tr>
<td>Total strain, $\varepsilon$</td>
<td>6.0%</td>
<td>0.1-6.0$\times$10$^{-2}$ s$^{-1}$</td>
<td>K-type thermocouple</td>
<td>100 Hz</td>
</tr>
</tbody>
</table>

Figure 2 – Energy partition experiments. (a) Layout of the experiments. (b) Experimental and modelling results.

As the current plasticity experiments were carried out at low strain rates, in accordance with the achieved strain rates in our fatigue tests, heat transfer had to be taken into account. I have implemented a one-dimensional coupled thermo-elasto-plastic model using MATLAB+SIMULINK® software to take into account the conductive,
convective and radiative heat losses during the test. Applying this model, $\beta$ was derived using an inverse method that solved an optimization problem: given the experimental temperature history, the optimal value of $\beta$ was determined by searching for the best fit between the temperature calculated using the governing equation system and the measured temperature history. Example results of this optimization are shown in Fig. 2(b). Evaluation of the dislocation substructure of samples at different stages of fatigue life – namely at usage factor UF = 0%, 5%, 25%, 50% and 70% – was carried out using TEM and XRD measurements to observe the kinetics of the fatigue process and investigate the microstructural background of the cyclic mechanical response.

3. SUMMARY OF THE RESEARCH WORK

In the first part of the research, the thermomechanical fatigue tests were performed with the GLEEBLE simulator for both structural materials. The low deviation of the data on cycles to failure and the correspondence of the Coffin–Manson curves to the experimental data indicate that the experimental setup and the test program were properly designed. Plastic strain energy per cycle and strain energy to failure were determined and found to be useful characteristics of fatigue life for both isothermal and thermomechanical low cycle fatigue. It was observed that the fatigue toughness increased with decreasing total strain amplitude, in agreement with previous studies. The input parameter ($\beta$) for the suggested low cycle fatigue model was determined by the inverse method described in the previous section. This parameter has not previously been published for these structural materials. A thermodynamically consistent thermo-elasto-plastic model was developed to take into account the heat transfer during the tests. The thermoelastic heat source was also taken into consideration in the heat equation; this showed that thermoelastic heating is non-negligible in determining the Taylor–Quinney coefficient. The results for 15Ch2MFA (bainitic structure with fine grains) showed a rather strong dependence of the Taylor–Quinney coefficient $\beta$ on strain rate, whereas with 08Ch18N10T (austenitic structure with coarse grains) it was practically independent (Fig. 3).
Based on this parameter, I developed and validated a new low cycle fatigue prediction model. The table embedded within each diagram in Figure 4 presents the best fit results for material constants for the model along with the coefficients of determination ($R^2$). It was demonstrated that in most cases the accuracy of the presented stored energy-based model was better than that of the widely accepted Coffin–Manson and classical strain energy-based models, as can be seen in Figure 5.
The evolution of the fatigue structure was observed through samples that interrupted the nominal fatigue life at different stages. Representative TEM micrographs for both investigated steels in their as-received and 50% fatigued states are shown in Figures 6 and 7, together with the evolution of the dislocation density investigated with XRD and TEM micrographs.

I found that the dislocation density for the 15Ch2MFA steel increased during the early fatigue life even though this material showed a cyclic softening behavior (Fig. 6). This contradiction can be resolved by mean free path analysis of the dislocations in the system.
Figure 7 – Investigation of the kinetics of the fatigue process for 08Ch18N10T. (a) TEM micrograph in the initial state. (b) TEM micrograph in the 50% fatigued state. (c) Evolution of the dislocation density with fatigue.

In the cyclic hardened material 08Ch18N10T, the dislocation density increased in the early fatigue life and well-developed cell-type dislocation substructures formed (Fig. 7).

4. THESES

Thesis 1 On the basis of the evaluation with Coffin–Manson model and the low deviation of the data on cycles to failure, the fatigue results for materials 15Ch2MFA and 08Ch18N10T – obtained from low cycle fatigue tests conducted between 150 °C and 270 °C with a thermomechanical physical simulator that applied Joule heating and cooling by an air nozzle – can be used to validate the energy-based low cycle fatigue model which is modified using the Taylor-Quinney parameter.

Related publications: [S1]; [S7]; [S8]; [S10]; [S14]

Thesis 2 The inverse method based on the experiment using a physical simulator with strain-controlled tension–compression plastic deformation at 6% total strain amplitude with simultaneous high-resolution temperature measurement, together with its analytical thermomechanical model, is able to predict the Taylor–Quinney coefficient described by Eq. (2-1) for the strain rate regime 0,1-6,0x10⁻² 1/s by calculating the heat transferred between the specimen and the ambient during the experiment. The Taylor–Quinney coefficient can be calculated as:

\[ \beta = \frac{\dot{Q}_p}{W_p}, \]  

(2-1)

where:

\[ \dot{Q}_p \] is the power related to the plastic heat source,
\( \dot{W}_p \) is the power related to the mechanical work.

**Related publications:** [S2]; [S5]; [S9]; [S13]

**Thesis 3**  The power of the thermoelastic heat source term defined by the formula \( \dot{Q}_e = -T \alpha_{CTE} E \dot{e}_e \) (which describes the Joule–Thompson effect) is generally neglected in the literature [Hodowany et al., 2000; Rusinek and Klepaczko., 2009]. Based on the results of the thermomechanical modeling, its value is in the same order of magnitude as the other terms of the thermo-elasto-plastic energy balance equation and so it has to be included in the modeling to investigate the partition of the mechanical work converted to heat during plastic deformation. The energy balance equation can be written as follows:

\[
\rho C_v \dot{T} = \beta \sigma : \dot{e}_p - T \alpha_{CTE} E \dot{e}_e - k \cdot \Delta T,
\]

where:

- \( \rho \) is the density,
- \( C_v \) is the specific heat at constant volume,
- \( T \) is the temperature,
- \( \beta \) is the Taylor–Quinney coefficient,
- \( \sigma \) is the Cauchy stress tensor,
- \( \dot{e}_e \) and \( \dot{e}_p \) the elastic and the plastic strains, respectively,
- \( \alpha_{CTE} \) is the coefficient of thermal expansion,
- \( E \) is the elasticity modulus,
- \( k \) is the coefficient of heat conduction.

**Related publications:** [S2]; [S5]; [S9]; [S13]

**Thesis 4**  Based on the results obtained by the inverse experimental–computational method, the Taylor–Quinney coefficient for the bainitic b.c.c. steel 15Ch2MFA is a monotonically increasing function of the strain rate in the range 0.001 to 0.06 1/s. It has a value between 0.63 and 0.91 for uniaxial tension and between 0.75 and 0.94 for uniaxial compression. In the same range of strain rate, the mean value of the Taylor–Quinney coefficient for the austenitic f.c.c. steel 08Ch18N10T is 0.87 (standard deviation (SD), 0.016) for uniaxial tension, and 0.92 (SD, 0.028) for uniaxial compression.

**Related publications:** [S2]; [S5]; [S9]; [S13]

**Thesis 5**  From the physical point of view, the classical energy-based low cycle fatigue models were more effective with the incorporation of the Taylor–Quinney coefficient into the engineering models through Equations (5-1) and (5-2), as their
damage parameters excluded the heat generated, which does not contribute to the fatigue damage of the material.

\[
W_{STF} = (1-\beta)W_{SCP} = D \cdot N_{SCP}^\delta, \quad (5-1)
\]

\[
W_{STF}^{stab} = (1-\beta)W_{SCP}^{stab} = Z \cdot N_{SCP}^{\eta}, \quad (5-2)
\]

where:

- \( W_{STF} \) [J/mm\(^3\)] and \( W_{STF}^{stab} \) [J/mm\(^3\)] are the proposed damage parameters (STF = STOrered, Fatigue),
- \( W_{SCP} \) [J/mm\(^3\)] and \( W_{SCP}^{stab} \) [J/mm\(^3\)] are the introduced strain energy until failure and the introduced strain energy in one cycle, respectively.
- \( \beta = \beta(\dot{\varepsilon}_p) \) [-] is the Taylor–Quinney coefficient,
- \( N_{SCP} \) [-] is the number of cycles to failure required to start crack propagation,
- \( D \) [J/mm\(^3\)], \( Z \) [J/mm\(^3\)], \( \delta \) [-], \( \eta \) [-] are parameters of the models.

When the models were applied to the thermomechanical low cycle fatigue results for the 15Ch2MFA bainitic and 08Ch18N10T austenitic steels, the accuracy levels of the stored-energy based models presented were higher than those of the widely accepted Coffin–Manson and classical strain energy based models.

**Related publications:** [S3]; [S15]

### Thesis 6

The relationship between the increase in the dislocation density in the microstructure of the bainitic steel 15Ch2MFA during the early fatigue life, as investigated by TEM and XRD, and its cyclic softening mechanical behavior can be modelled through Equation (6-1), derived for the correlation between the stress ratio and the density of the mobile dislocations ratio concerning two different states of the fatigue process:

\[
\left( \frac{\tau_{i,k}}{\tau_{i,n}} \right)^m = \frac{\rho_{d,i,k}}{\rho_{d,i,n}}, \quad (6-1)
\]

where:

- \( \tau_i \) [Pa] is the maximal value of the shear stress for stage \( i \) of fatigue life,
- \( m \) [-] is the dislocation–velocity stress coefficient,
- \( \rho_{d,i,k} \) [1/m\(^2\)] is the density of the mobile dislocations for stage \( i \) of fatigue life.

The TEM investigations showed that the microstructure starts to change significantly after reaching half of the fatigue life, while the plastic deformation takes place through two different deformation mechanisms, namely the motion of dislocations and the motion of cell boundaries, resulting in microcracks, especially in the cell boundaries occupied by the carbides those became coarser during fatigue.

**Related publications:** [S1]; [S4]; [S5]; [S6]; [S11]; [S12]
Thesis 7  Based on the microstructural investigations with TEM and XRD analyses, the cyclic hardening behavior in the early period of the fatigue life of austenitic steel with a mean grain size of 47 µm is caused by the increased dislocation density, and the effect of the cell boundaries, which act as strong obstacles to dislocation motion. The TEM investigations performed on samples at different stages of fatigue life showed that the dislocations begin to concentrate into dislocation veins in the early stage of the fatigue life; and then with further loading the dislocations concentrate in the walls of cells and persistent slip bands, the structures that are the origin of the microcrack formation.

Related publications: [S1]; [S4]; [S5]; [S6]; [S11]; [S12]

5. PUBLICATIONS RELATED TO THE THESES


[S5] Peter Bereczki, Balazs Fekete, Viktor Szombathelyi, Fanni Mísjak: Different applications of the Gleeble thermal-mechanical simulator in material testing, technology optimization and process modeling. ASTM Materials Performance and Characterization, DOI: 10.1520/MPC20150006 (published online)


