ACCELERATED CREEP TEST FOR REPAIR WELDS IN THERMAL POWER GENERATION

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Abstract

An effective way to economy of energy generation is the extension of operation life of fossil fueled thermal power plants. It can be achieved by progressive repairs involving replacing the exploited components by new ones aiming to higher regimes of operation. The repairs involve welding and the heat-affected zones of the repair welds on the side of the exploited material become the critical regions of the whole upgraded structure. As after the repair the plant must be brought as soon as possible into operation, therefore one must very soon estimate its allowable operation parameters and life. For this the conventional creep tests are useless as being very long lasting. A recent physically simulated accelerated creep test (ACT) on Gleeble allows solving of this problem and from the collected materials data determining the long-term behaviour of such repair welds. In this test, executed on bulk 10mm diameter cross-weld samples, the creep cracks nucleate in the most vulnerable portion of the welded joint, which location cannot be predicted in the true weld by any method until the failure occurs. The simulative ACT on Gleeble “identifies” the location of potential crack appearance without violating the micro-mechanism of the ductile fracture characteristic of creep. It also tells if the partly exploited components of power-gen structure can be repaired and allows estimating how long the repaired structure will last in the allowable operation conditions. An example of the repair weld is presented in this paper and the microstructures after ACT are discussed.

Keywords: creep testing, welds, Gleeble, microstructures, electron microscopy

1. INTRODUCTION

Reliable supplies of energy at an economical price are a key factor for economic growth of society and become technical challenge to power plant manufacturers and utility operators across the World. Within Europe, one of main ways to secure energy requirements is maintaining conventional fossil, biomass and waste-fired plants within each country’s mixed power system by extending their life and if possible improving their efficiency. Ageing of these plants results from creep of the construction materials, which reduces the material’s strength during long-time operation under pressure at elevated temperature. Accurate detection of creep in its early stage is impossible by conventional NDT methods [1]. A better method of detecting creep damage is by taking replicas from carefully prepared outer surfaces of operating power plant components [2]. Microscopic examination of such replicas reveals creep voids and micro-cracks if they occur, but their presence disqualifies the component for a repair and extended operation because always more damage occurs from inner side of the component or inside its wall. An advanced technique comprises taking extraction replicas from components with no creep cavitation and identifying precipitates / carbides on them [3]. Then from well known sequences of carbide transformations [4] the remaining plant life can be estimated.
This method has also its limitations, as at first the largest and most transformed carbides can hardly be extracted and moreover the fine transient carbides not always retard the creep. And it is always the matrix of the material in which the creep voids nucleate while the nucleation process is controlled by the population of generated dislocations.

In a component made of 0.5CrMoV steel (grade P11), which operated for 22.5 years at 568°C at stress of 16.6MPa, the creep-cavitation was detected, Fig.1, while in its microstructure still numerous fine precipitates could be seen, Fig.2. Study on this and other materials during the launched in the last decade COST-538 EU’s R&D Action “High Temperature Plant Lifetime Extension” [5] involved an accelerated creep test, i.e. a low-cycle thermal-mechanical fatigue procedure developed on Gleeble™ physical simulator [6]. The main aim of this test is to reproduce on samples the creep failure to appear as similar as possible to the creep cracks occurring in true exploitation. Most typical, for the welds on ferritic-bainitic and martensitic-ferritic creep-resisting steels, are type-IV cracks in heat-affected zone, which often initiate below the surface and then propagate to the surface and along the weld [7].

Type-IV cracks result from accumulation of strain in the weakest grains of material, eventually compensated by accommodation of tensile strains in hundreds or even thousands of surrounding grains and subgrains. Therefore a reliable creep test must be executed on relatively large bulk samples by applying tensile stress in direction near-normal to the HAZ, like it appears in true welds, Fig.3a [8]. The micro-mechanism of such intergranular failure is by sliding of grain boundaries [9] or piling-up dislocations at favourably oriented grain boundaries, Figs 3b and 3c. SEM observations of creep voids and cracks, carried out on fractures of the...
0.5CrMoV steel crept for 22.5 years at 568°C at stress of 16.6MPa, revealed criss-crossing slip lines, Fig.4a [10], resulting from coalescence of voids formed by piled-up planar arrays of dislocations, as schematically given in Fig 4b. Most of fine precipitates observed in this steel, were identified as transient M₆C₆ (Fe₂MoC) and M₇C₃ carbides [11], characteristic of rather earlier stages of creep than of the “end-of-life”.

Fig.4a Pattern of “slip-lines” on surface of creep void in 0.5CrMoV steel [11]

Fig.4b Schematic of creep void formation (F) and coalescence (C) resulting in criss-crossing slip-lines pattern like in Fig 4a

2. EXPERIMENTAL

2.1. Initial material and its testing

On the same grade of 0.5CrMoV steel (P11) which operated for ~20 years at 568°C at stress of 3.8 MPa and showed no creep-cavitation, a test weld coupon was made using P24 grade coated stick electrodes. Multi-bead welds were laid down using quasi temper-bead method, adjusted to reduce hardness differences between the old exploited base material and the fresh weld metal without PWHT. As the weakest portion of such welded joint is expected to be the heat-affected zone on the side of exploited old material, the main aim of this study was to characterize this HAZ and to describe if the ACT causes microstructure transformation, simulating that of further operation of the repaired component at creep condition.

The used plate of P11 steel had microstructure of well recrystallised ferrite grains with fine precipitates inside them and large carbides located mainly along the grain boundaries, Fig.5. The microstructure of P24 weld metal was martensitic, while the HAZ in the parent steel plate consisted of two distinguished layers, Fig.6:

- the “white layer” having dual-phase microstructure of ferrite with fine, compact islands of high-carbon martensite; in which the initial large carbides were entirely dissolved, Fig.7;
- the transient “layer” consisting of ferrite with globular islands of martensite formed in sites of entirely or partly dissolved large carbides, Fig.8.
Micro-hardness measurements, \( \mu \mathrm{HV}_{200G} \), showed hardness of the “white” layer similar to the hardness of the P24 weld metal, see Fig.6, while the transient layer was of the highest hardness, especially in locations where the partly dissolved large carbides were surrounded by the high-carbon martensite. In non-affected by the welding thermal cycle parts of the P11 parent plate an average micro-hardness was \( \sim 187 \, \mathrm{HV} \).

From the welded coupon cross-weld bulk samples were machined for the ACT in such way that the weld’s HAZ appeared in the middle of the sample’s gauge portion, and this portion was subjected to axial loading and thermal cycling in Gleeble, according to the procedure described elsewhere [12].

### 2.2. Results of ACT

The accelerated creep tests were carried out at two different base temperatures: \( 600^\circ \mathrm{C} \) and \( 625^\circ \mathrm{C} \).

From the duration of ACT the following parameter has been calculated:

\[
P_{\text{ACT}} = (7 + \log t) \cdot \frac{T}{100}
\]

where: \( t = \) time of ACT in [ksec] till onset of fracture, and \( T = \) base temperature in [K].

Then, the creep strength factor in the ACT has been given as:

\[
F_{\text{ACT}} = P_{\text{ACT}} \cdot R_S / 100
\]
where $R_S$ is the average from all cycles tensile stress measured after relaxation.

<table>
<thead>
<tr>
<th>Material</th>
<th>Sample</th>
<th>Temp [°C]</th>
<th>Time [ksec]</th>
<th>$R_S$ [MPa]</th>
<th>$P_{ACT}$</th>
<th>$F_{ACT}$ [MPa]</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weld P11-P24</td>
<td>P11-HAZ1</td>
<td>600</td>
<td>18.4</td>
<td>258</td>
<td>72.2</td>
<td>186</td>
</tr>
<tr>
<td>Weld P11-P24</td>
<td>P11-HAZ2</td>
<td>625</td>
<td>16.5</td>
<td>239</td>
<td>73.8</td>
<td>176</td>
</tr>
</tbody>
</table>

Analysis of data collected by Gleeble during the accelerated creep test on this HAZ gave the result of 11245 hours for 100MPa at 600°C. For the 20 years exploited P11 pipe material, which was additionally tested, similar calculation of the ACT data resulted in 3823 hours for 100MPa at 600°C.

Micro-hardness measurements on the cross-section of the ACT sample, Fig.9, revealed decrease of the hardness in P24 weld metal, as compared with that before the ACT, while increase of hardness on the side of the parent P11 steel in which deformation bands appeared. The highest micro-hardness after the ACT was measured in the former transient layer of the HAZ. Also the hardness of “white” layer slightly increased, as compared with the initial as-welded state. After the ACT the both characteristic layers of P11 steel’s HAZ appeared well preserved, however tempering of the dual-phase microstructure by thermal cycles of the ACT caused the fine martensitic islands very hard to distinct from the ferrite matrix.

Observation of the sample during the ACT revealed that two cracks formed one after another in its gauge portion. The first crack initiated in the former “white” layer of the HAZ, and then deviated towards the transient layer of the HAZ, where was retarded. The second crack initiated at the end of gauge portion i.e. in the exploited P11 steel. The test on this sample was interrupted before full breaking of the sample.

2.3. Electron–microscopy study

To explain these results, metallographic observations followed by TEM examination of thin foil specimens were carried out. Slices for the preparation of thin foil specimens were taken from the gauge portion of ACT sample by cutting parallel to the fusion plane of the weld. Microstructures from the P24 weld metal through the HAZ up to the parent P11 material were examined and compared with these of the initial state.

The HAZ of the welded joint in the 20-years crept P11 steel, showed globular ferrite grains and subgrains dominating in the “white layer” and minor amount of the fine compact martensite islands often deposited along the ferrite grain boundaries, Fig.10. Larger islands of martensite, often embedding non-dissolved large carbides, appeared in the transient layer and around them the dislocation density was higher.
After the ACT, in the HAZ of P11 steel both characteristic layers appeared well preserved. As regards the “white” layer, it consisted of recovered / recrystalised fine grains of ferrite, Fig.11, with sub-micron size carbides often pinning the grain boundaries. The dislocation density in these grains was substantial, however visibly smaller than before the ACT. An extremely high dislocation density appeared in the location between the “white” and transient layer, at which the first ACT crack after its initiation in the HAZ, begun deviating towards the weaker non heat-affected P11 material, Fig.12. The study on the hardening micro-mechanism of this HAZ during the ACT is continued, including the examination of the transient layer of the HAZ in which were formed complex agglomerates of spheroidised and platelike carbides with martensite, Fig.13.

3. DISCUSSION

Cross-weld sample of dissimilar materials: P11 partly exploited by creep for 20 years and P24 fresh weld metal was subjected to thermal-mechanical elasto-plastic fatigue until onset of crack. The weld’s HAZ initially suspected to be the weakest portion of this welded joint did not appear to be. The first smaller crack initiated in the HAZ of weld but did not propagate too far and ended just after the transient layer of the HAZ. The second crack initiated and then propagated through the non-affected by the weld thermal cycle portion of the crept P11 material, and this crack started from the end of the gauge portion of the ACT sample, i.e. from the location being beyond the uniformly heated zone, in which the maximum temperature during the test was from 560 to 5700°C for the programmed 600°C temperature of the test. Different elevated temperature strengths of the P11 and P24 materials resulted in more accumulation of strain in P11 than in P24, thus...
affecting increase of hardness of the P11 material in the ACT gauge zone. The presence of deformation bands consisting of very fine subgrains as well as the appearance of unknown from the initial state very fine precipitates in these deformation bands can explain the origin of the hardening. In P24 weld metal the ACT caused uniform medium tempering of microstructure with recovery of ferritic matrix, resulting in the decrease of room temperature hardness. The heat-affected zone of the weld appeared stronger than the 20 years crept P11 steel. The large carbides, which during 20 years of creep were formed in P11 steel, entirely or partly dissolved in this HAZ during the welding thermal cycle thus substantially re-alloying its matrix and this way increasing its hardness. The weld’s HAZ, harder than the crept P11 steel, appeared to be a good transfer region to the stronger P24 weld metal. In certain sense the stronger in the as-welded state outer layer of the HAZ containing partly dissolved large carbides acted in the ACT as a shield protecting the inner “white” layer from accumulating strains there. The study on strengthening micro-mechanism of the inner “white” layer of the HAZ is to be continued. As regards the outer, transient layer of the HAZ, which contains compact second-phase islands of martensite embedded in soft ferrite matrix, the analogy to their role can be found in dual-phase steels [13]. Microhardness measurements along the first crack, which initiated in the HAZ of the weld, showed that the first crack’s deviation and then retardation appeared in the dual-phase microstructure due to substantial strain hardening ahead of the crack’s tip. The last has been characteristic of the dual-phase microstructures and considered as the origin of their exceptional fatigue strength [14]. After this first crack was stopped or at least its propagation rate was slowed down, the second and main crack did nucleate and propagate in the weakest material, which in this case was the crept for 20 years P11 steel. The sequence of the cracks appearance indicates substantial changes of properties of different microstructural components of the welded joint during the accelerated creep testing.

And from the stress-strain-time data collected by the acquisition system of the Gleeble thermal-mechanical simulator it was possible to calculate the creep lifetime of such P11–P24 dissimilar welded joint, intended to be applied in repair welding of a power plant component.

4. CONCLUSIONS

- The accelerated creep tests developed on Gleeble thermal-mechanical simulator allows gaining in just a few hours the materials behaviour data necessary to estimate the real creep life-time.

- In the case of testing cross-weld samples, the ACT reveals which part of the microstructure is the most prone to crack initiation and where the failure shall happen, and when the regions of the matrix substantially change their properties during the test it also allows an alternative crack to form.

- The heat-affected zone (HAZ) of weld does not need to be the weakest portion of the welded joint, for example in repair welds on some long-term crept steels, as the regeneration of crept steel’s microstructure and local improvement of properties in HAZ may be achieved by an optimum welding procedure.

- The ACT, by accumulating small tensile and compressive deformations, strain-hardens the material and stimulates strain-induced precipitation thus revealing the remaining strengthening potential of the partly exploited steel and contributing to adequate calculation of the remnant creep life-time.

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