Effect of Strain Rate on the Low Cycle Fatigue behavior of 316L(N) Stainless Steel Weld Joints

Sayan Kalyan Chandra\textsuperscript{a}, Vani Shankar\textsuperscript{b*}, K. Mariappan\textsuperscript{b}, R. Sandhya\textsuperscript{b}, P. C. Chakraborty\textsuperscript{b}

\textsuperscript{a}Materials and Metallurgical Engineering, Jadavpur University, Kolkata-700032, India
\textsuperscript{b}Mechanical Metallurgy Division, Indira Gandhi Centre for Atomic Research, Kalpakkam-603102, Tamil Nadu, India

Abstract

The aim of the present paper is to study the effect of strain rate on the low cycle fatigue (LCF) behavior of indigenously developed 316LN stainless steel weld joint (0.14 N wt. % base metal welded with 0.1 wt. % N electrodes). Fully reversed total axial strain-controlled LCF tests were conducted at 823 K using a servohydraulic machine, equipped with a resistance heating furnace. Tests were conducted in air at ±0.6% strain amplitude employing strain rates $3 \times 10^{-3}$ s\(^{-1}\), $3 \times 10^{-4}$ s\(^{-1}\) and $3 \times 10^{-5}$ s\(^{-1}\) to investigate the effect of strain rate in the dynamic strain aging regime (DSA). Initial brief hardening, negative strain rate stress response and lowering of fatigue life with decreasing strain rate have been correlated to DSA operating at 823 K. The LCF behaviour of weld joints are correlated to the microstructural changes during fatigue cycling in order to explain the underlying deformation mechanism and fracture behaviour.

Keywords: Low cycle fatigue; 316L(N) austenitic stainless steel; DSA; microstructure

1. Introduction

Low cycle fatigue (LCF) is an important consideration in the design of high-temperature systems subjected to thermal transients. LCF resulting from thermal transients occurs essentially under strain controlled conditions, since the surface region is constrained by the bulk of the component. At high temperatures the fatigue deformation and life are influenced by several time-dependent mechanisms such as dynamic strain ageing (DSA), oxidation, creep and phase transformations. For accurate prediction of LCF at elevated temperatures, determination of rate controlling time dependent damage process that influences the cyclic deformation and fracture behaviour of alloys is essential. At intermediate temperatures, where the effects of time dependent processes such as creep and oxidation are found to be minimal, the drastic reduction in LCF life observed with increasing temperature and decreasing strain rate in alloys such as type 304 stainless steel [1] and Nimonic PE 16 superalloy [2] and other 300 series austenitic stainless steels [3] has been ascribed primarily to
the deleterious effects of DSA. These alloys displayed anomalous cyclic hardening and an increase of maximum cyclic stress with increasing temperature or decreasing strain rate due to DSA. 316L(N) austenitic stainless steel is the material chosen for the primary components in liquid metal-cooled fast reactor due to its excellent high temperature mechanical properties and compatibility with the heat transfer medium i.e., liquid sodium. Weld joints are the weak links in any structural component and hence in order to evaluate the structural integrity of a component, it is important to assess the performance of a weld joint. Hence the present investigation is focused on assessing the effects of DSA on LCF properties of 316 L(N) SS weld joint.

2. Experimental

The chemical composition in wt. % of the 316L(N) base metal is provided in Table-I. The base metal was solution treated at 1373 K for 1 h and then water quenched in order to have both carbon and nitrogen in solid solution. Cylindrical samples of 10 mm gauge diameter and 25 mm gauge length were used for LCF testing. Sections of 450×250×25 mm, cut from the mill-annealed plate were joined along the length direction by shielded metal arc welding (SMAW) process. The pads were made using type 316(N) welding electrodes. The electrodes were soaked for 1 h at 473 K before the commencement of welding. During welding, the voltage and current were maintained at approximately 25V and 150 A, respectively. Weld joint specimens were machined from the weld pads fabricated with a double-V configuration, with an included angle of 70°, a root face of 2 mm, and a root gap of 3.15 mm. An interpass temperature of 423 K was maintained during welding. The weld pads were examined by X–radiography for their soundness. Specimens were machined from the central sections of defect-free double-V weld pads. Fully reversed total axial strain-controlled LCF tests were conducted at ambient and 823 K employing constant strain amplitude of ±0.6%, at various strain rates 3×10⁻³s⁻¹, 3×10⁻⁴s⁻¹ and 3×10⁻⁵s⁻¹, using a servohydraulic machine and equipped with a resistance heating furnace. The general microstructure of the weld joints was revealed by etching electrolytically using a solution containing 70% HNO₃. The initial microstructure of the base and weld metal contained in the weld joint are shown in Figs. 1(a) and (b) respectively. The weld metal showed typical morphologies of vermicular/lacy δ-ferrite (Fig. 1(b)). The coarse grained region adjoining the weld metal consisted of austenite which could act as a metallurgical notch and lead to failure during fatigue testing. The LCF tested samples were sectioned parallel to the stress loading direction, polished, etched and examined under an optical microscope.

Table 1. Chemical composition of base metal of 316 L(N) stainless steel in wt %.

<table>
<thead>
<tr>
<th>Alloying element</th>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>N</th>
<th>Mn</th>
<th>S</th>
<th>P</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt. (%)</td>
<td>0.025</td>
<td>17.5</td>
<td>12.1</td>
<td>2.53</td>
<td>0.14</td>
<td>1.74</td>
<td>.0041</td>
<td>.017</td>
<td>Balance</td>
</tr>
</tbody>
</table>

Fig. 1. Initial microstructure of (a) base metal and (b) weld metal.
3. Results and discussion

3.1. Effect of temperature and strain rate on LCF behavior

Effect of temperature and strain rate on the cyclic stress response curves are shown in Fig. 2 (a) and (b) respectively. The weld joint generally exhibited a very rapid strain hardening to a maximum stress followed by a nearly stable peak stress. However, it was observed that whereas at room temperature (Fig. 2(a)), there was a gradual increase in the peak tensile stress, that at temperatures such as 823 K and 873 K, large and rapid increment in the peak tensile stress values was attained. The overall stress response values increased as the strain rate was decreased from $3 \times 10^{-3}\text{s}^{-1}$ to $3 \times 10^{-5}\text{s}^{-1}$ (Fig. 2(b)). Also, fatigue life decreased with an increase in the test temperature (Fig. 2(a)) and decreased with a decrease in strain rate (Fig. 3 (a)). Stress range showed an increase with decreasing strain rate (Fig. 3 (b)) but no trend was observed for the plastic strain range with decreasing strain rate (Fig. 3(c)). The stress range and plastic strain range were obtained from stable hysteresis loops. Negative strain rate stress response, the initial brief hardening, and lowering of fatigue life, decrease in plastic strain range and increase in stress range at half-life with decreasing strain rate have been correlated to dynamic strain ageing operating at 823 K and is caused by the interaction between dislocations and solute atoms in this alloy [4]. A continuous decrease in fatigue life with a decrease in strain rate and increase in temperature in the base metal has been reported to occur due to DSA under conditions where the effects of oxidation and creep were non-existent [5]. An increase in dislocation density with reduced strain rate was reported earlier in this material at 773 and 823K [5]. It was pointed out that DSA would enhance the degree of inhomogeneity of deformation during LCF by the solute locking of slow moving dislocations between slip bands. Presumably, the dislocation velocities inside the slip bands were too high for dynamic ageing of mobile dislocations to take place and consequently DSA enhanced the partitioning of strains into separate regions characterized by high and low amplitudes of dislocation movement. During DSA, slow moving dislocations become aged by the solute atmospheres and additional dislocations were generated to maintain the imposed deformation rate. This process caused an increase in the total dislocation density. The negative strain rate dependence of cyclic stress response over the temperature and strain rate range where DSA operated resulted from an increase in total dislocation density during deformation. The matrix was hardened during DSA, causing an increase in flow stress needed to impose the same total strain during successive cycles. Other manifestations of DSA such as serrations on the stress strain hysteresis loops were not observed in this alloy at 823 K.

![Fig. 2. Effect of temperature (a) and strain rate (b) on the cyclic stress response curves of 316 LN weld joint.](image-url)
3.2. Failure location and failure mode

In order to evaluate the failure location under various testing conditions, the longitudinal section of the failed samples were analysed to study the microstructure near the fractured surface. It was found that at both room temperature and 823 K and at the three strain rates, the samples failed in the weld metal region (Fig. 4 (a) through (c)). It was seen that at higher temperatures such as 823 K, microcracks formed at the austenite-δ-ferrite interface. Joining of these microcracks led to large cracks that resulted in final failure (Fig. 5). Even though the fine duplex austenite-ferrite microstructure of the weld metal, with its many phase boundaries, offered a good resistance to the extension of fatigue cracks by causing deflection of the crack path (Fig. 5) than the coarser base metal, the final failure invariably occurred in the weld metal. This could be attributed to the reduced tensile ductility of weld metal. It is known that fatigue life is mainly governed by the tensile ductility of material. It is also observed that even though failure occurred in the weld metal, many secondary cracks were also present in the base metal, especially at lower strain rates (depicted in Fig. 6). The degree of intergranularity also increased with the decreasing strain rates. Total length of intergranular and transgranular crack length was measured on the specimen surface in the DSA regime such as 823 K. It was noticed that at 823 K i.e. in the DSA regime, number of both transgranular and intergranular cracks increased with decrease in strain rate (Fig. 6) and very long cracks were seen (a consequence of crack coalescence) at low strain rates. Whereas at $3 \times 10^{-5}$ s$^{-1}$ strain rate, the cracking mode was purely transgranular, that at $3 \times 10^{-4}$ s$^{-1}$ strain rate the fraction of intergranularity was 0.19 and which increased to 0.36 at $3 \times 10^{-5}$ s$^{-1}$ strain rate as a consequence of crack coalescence. As mentioned earlier, an increase in dislocation density with reduced strain rate was reported earlier in this material at 773 and 823K [5]. With the reduced strain rates, higher back-up stresses due to pile up of dislocations get generated near the grain boundaries leading to larger amount of microcracking.
Fig. 5. Crack propagation along austenite-delta ferrite interface in the weld metal region (test conducted at 823 K, +0.6% and 3×10^{-3}s^{-1} strain rate.

Fig. 6. Crack propagation mode occurring in the weld metal (WM) and base metal (BM) during low strain rate (3×10^{-4}s^{-1}) experiment at 823 K.

4. Conclusions

Signatures of dynamic strain ageing such as initial brief hardening, negative strain rate stress response, lowering of fatigue life, lowering of plastic strain range and an increase in stress range at half-life with decreasing strain rate have been evidenced at 823 K. Mode of cracking in the base metal also changed from purely transgranular to mixed mode and the degree/fraction of intergranularity increased with a decrease in strain rate at the same temperature. Failure occurred in the weld metal due to reduced ductility of the weld metal.

Acknowledgements

The authors would like to thank Mr. S.C. Chetia, Director IGCAR, Dr. T. Jayakumar Director, MMG, Dr. A.K. Bhaduri, Associate Director, MDTG, and Prof. S.K. Ray, Jadavpur University, Kolkata for their encouragement and constant support.

References