International Journal of Fatigue 33 (2011) 1505–1513

Contents lists available at ScienceDirect

International Journal of Fatigue

journal homepage: www.elsevier.com/locate/ijfatigue

Behaviour of small cracks during their propagation from Vickers indentations in coarse-grain steel: An experimental investigation

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ARTICLE INFO

Article history:
Received 20 January 2011
Received in revised form 21 May 2011
Accepted 5 June 2011
Available online 23 June 2011

Keywords:
Coarse grain
Vickers indentation
Residual stresses
Crack initiation
Crack propagation

ABSTRACT

In this investigation we look at the influence of the local residual stresses caused by Vickers-pyramid indenting on the initiation and early propagation of small cracks from indentations in coarse-grain martensitic steel. The size of these indentations is comparable to the grain size. Specimens with and without a local residual stress field were tested on a rotary bending machine. A focused ion beam and a scanning electron microscope were used to reveal the influence of those stresses on the location of the cracks' initiation and the mechanism of the small-crack propagation. The existing local residual stresses assist in the initiation of two cracks at a level lower than the fatigue limit. The early small-crack propagation is gradually obstructed by the residual stress-field configuration until the cracks become non-propagating cracks. At levels higher than the fatigue limit, both cracks succeed in breaking through the compressive stressed domain and link together. From that moment the crack begins to behave as a long crack, penetrating outside the indentation into the tensile-stressed domains.

1. Introduction

The phenomenon of metal fatigue is a complex process involving several steps. However, the most clearly recognised are crack initiation and crack propagation. Crack initiation starts with the slip within the grains, with dislocations moving along crystallographic planes parallel to the maximum shear stress. This slip gradually localises, persistent slip bands form, and then extrusions and intrusions appear. In accordance with the Tanaka–Mura model [1], the stress necessary for slip is lower in coarse grains. However, since metals are composed of randomly oriented grains, the strain–stress response depends on the grain orientation.

A model based on the surface-strain redistribution introduced by Abdel-Raouf and co-workers was used to predict the fatigue limit and early small-crack propagation in an aluminium alloy [2,3]. Namely, the cyclic loading of smooth specimens leads to a local strain redistribution along the surface. The strain is the highest within the most favourably oriented grains at the surface. However, due to an increasing constraint, the maximum strain below the surface decreases and attains a nominal value not earlier than at a certain distance. The strain gradient is inversely proportional to the grain size. This model was successfully used by Plumtree to explain the decrease in the fatigue limit with an increase in the grain size [4].

Because of the easier slip and the higher strain amplitude at the surface, the first cracks initiate in the coarsest grains at the surface. Here, cracks that are smaller than grains can propagate faster than long cracks [5], and grain boundaries are the barriers for small-crack propagation. The driving force for the cracks' propagation is the effective range of the stress-intensity factor, ΔK_{eff}. However, due to crack closure ΔK_{eff} is often lower than the whole range of the stress-intensity factor, ΔK. Then, as cracks approach the grain boundaries, the propagation rate is gradually retarded because ΔK_{eff} decreases. Finally, at a lower ΔK_{eff} the cracks stop propagating and become non-propagating cracks [5,6]. In contrast, if ΔK_{eff} exceeds the threshold value for crack propagation, the crack will propagate further as a long crack.

A series of investigations was performed by Murakami, Endo, and co-workers in order to explain how small cracks initiate and propagate, in different metals, from artificial and real small defects [7–9]. They found both the hardness and the parameter $\sqrt{area}$ to be crucial. The magnitude of the parameter $\sqrt{area}$ is square root of the defect projection onto the plane of stresses. No attention was given to the effects of the local residual stresses (LRS) that arise due to the indenting. In fact, drilled small holes were mostly used in the past as the artificial small defects. Data based on the effects of other artificial defects are very rare.

A Vickers indentation is a very practical sort of artificial small defect. For example, only the hardness of the material needs to be known for it to be prepared with a specific size of indentation. However, the result of pyramid indenting is an extensive localised material deformation, which is partly irreversible. After the force is

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doi:10.1016/j.ijfatigue.2011.06.008

removed the LRS field springs out into the close surroundings. The level and the configuration of LRS have an influence on the crack initiation and its early propagation. Both are reflected in the fatigue limit.

Our previous papers have dealt with the use of Vickers indentations as artificial small defects [10–12]. Individual and series of indentations of different sizes were found to be appropriate for our investigations [11]. The first findings relating to the effects of the LRS caused by the preparation of artificial small defects have already been published [12,13].

Welds are used in the construction of structures and heavy machines. As a result, various defects can arise in these welds. Furthermore, larger defects can be found when using the appropriate NDE methods. Imperfect welds can, of course, be repaired, but smaller defects tend to remain in the welds [14]. The coarse-grain, heat-affected zone (CG HAZ) is a microstructure that is sensitive to cyclic loading, with its low resistance to crack initiation [1]. Relatively long cracks behave as small ones in the CG HAZ, because of the extensive size of the grains. The strongest welds in the as-welded condition are butt-welds. The position of CG HAZ in such a weld is shown in Fig. 1.

If small defects have an effect on the CG HAZ, crack initiation will be easier. The reasons for this are the coarse grains, the residual welding stresses, the stress concentration, the high hardness, etc. Small defects affecting the CG HAZ are an undersized lack of fusion, inclusions, precipitations, tiny cracks, and undercuts. The origins of the small cracks in the CG HAZ are cold, liquation and reheat cracking or lamellar tearing. An additional stress field can be restored around the natural small defects, too. For example, different thermal expansion coefficients of the inclusions, with respect to the steel during cooling, an enhanced residual stress field due to the stress concentration caused by pores, the tensile stress at the cold crack tip due to reversed plastic-zone formation, etc.

The present research was performed on coarse-grain martensitic steel like that found in the CG HAZ in the case of fast cooling. A groove on the specimen was used to set up a stress concentration, like in butt-welds. The designed microstructure for the experimental work was prepared in the laboratory by using a two-step heat treatment (HT), and Vickers indentations at the bottom of the groove were used as artificial small defects. They were prepared either after the first stage or the second stage of the HT. The indented specimens with and without the LRS were tested in this state. This made it possible to consider the effects of the LRS around the same size indentations on the initiation and early propagation of small cracks. The LRS caused by the indenting were calculated using the finite-element method (FEM). After the specimens were loaded into the rotary bending machine (RBM), either for the small-crack initiation and early propagation or to final fracture, some of the specimens were prepared for an experimental investigation by using a focused ion beam (FIB) and a scanning electron microscope (SEM) [15,16].

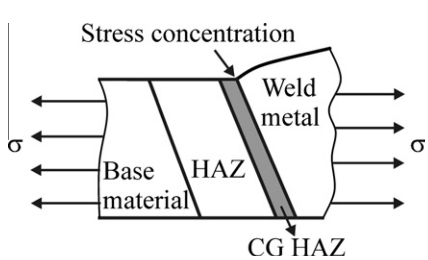

Fig. 1. Position of the CG HAZ in a butt-weld.

2. Experimental details

2.1. Material

The steel 17CrNiMo7 was used to prepare samples of coarse-grain steel, referred to here as “CG HAZ”. The chemical composition and the mechanical properties of the steel are given in Tables 1 and 2.

2.2. Experimental procedure

The procedure used for the specimens’ preparation is shown schematically in Fig. 2. The HT is the first-stage, i.e., coarse-grain annealing and water quenching, and the second stage, i.e., hardening by water quenching. During the annealing at 1100 °C for 3 h the grains grow to 200 μm, while the water quenching ensures a martensite transformation with the same grain size.

Table 1
Chemical composition of the steel (weight%).

C	Si	Mn	P	S	Cr	Ni	Cu	Mo	Al
0.18	0.22	0.43	0.012	0.028	1.56	1.48	0.15	0.28	0.023

Table 2
Mechanical properties of the steel.

	R_{p02} (MPa)	R_m (MPa)	A_5 (%)	Z (%)	Impact toughness (J)	Hardness (HV10)
“CG HAZ”	1042	1431	15	56	72 (at + 20 °C)	463

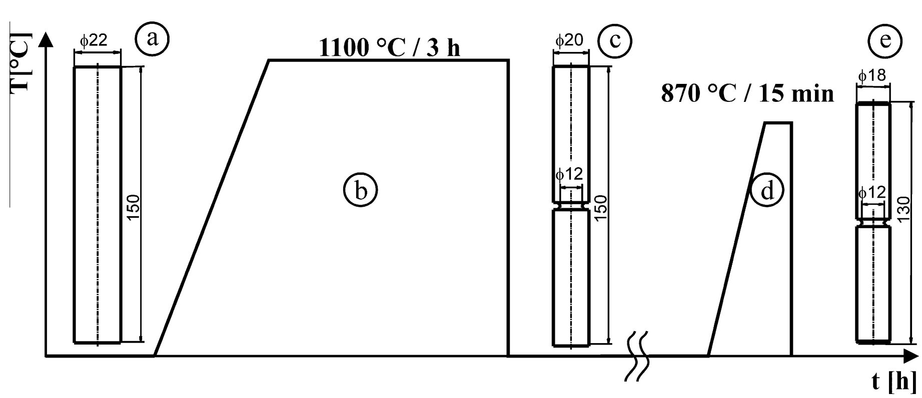

Fig. 2. Procedure for the specimens’ preparation.