INFLUENCE OF SMALL SURFACE DISCONTINUITIES AT THE WELD TOE ON BENDING FATIGUE STRENGTH

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Results of bending fatigue strength on specimens with synthetic CGHAZ microstructure which were smooth and flawed at the surface were discussed in respect of flaw size and mechanical properties. Specimens were grooved to cause similar stress concentration as that existing at the weld toe. Artificial flaws were single and series of Vickers indentations of the size comparable to the characteristic dimensions of CGHAZ microstructure. Threshold stress intensity factor is a linear function of strength and a parabolic function of flaw size. Bending fatigue strength increases with the strength in the case of significant flaw size and/or harder CGHAZ microstructure. But, it is also increased by local plastic deformation which has happened at the weld toe in the case of insignificant flaw size and/or softer CGHAZ microstructure.

INTRODUCTION

The main fatigue strength controlling parameters of macroscopic homogeneous full penetrated butt weld joints are level of stress concentration and presence of small material discontinuities at the weld toe. Actual fatigue properties of microstructure at the weld toe, where coarse grained heat affected zone (CGHAZ) is usually found, are seldom experimentally studied. Microstructure and properties of CGHAZ depend mainly upon chemical composition of base material (BM), temperature gradient during cooling sequence of weld thermal cycle and of eventual subsequent heat treatment.

The results of bending fatigue strength measurement on different synthetic CGHAZ microstructure are shown and discussed in this article. Stress concentration caused by the transition of weld reinforcement to BM was modelled by a groove. The surface was either smooth or flawed by Vickers indentations.

MATERIAL, SPECIMENS AND EXPERIMENTAL PROCEDURE

The material used was 690 MPa yield stress (CrNiMo) N9ncral 70 (grade HT 80) high strength structural steel. Actual chemical composition and mechanical properties of the steel were: 0.09 C, 0.27 Si, 0.30 Mn, 0.015 P, 0.010 S, 1.05

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Cr, 0.27 Mo, 2.63 Ni, 0.07 V, 0.045 Al, 0.026 Ti, 713 MPa yield stress, 764 MPa tensile strength and 170 J impact toughness at -50°C. The steel was made by quenching and tempering. Its microstructure is tempered martensite with some bainite (grain size ~ 22 µm).

Microstructure of different CGHAZs were made by a computer controlled thermal cycle simulator. The samples were cut off 30 mm thick plate in the rolling direction in size 15x15x70 mm (mechanical properties), 11x11x55 mm (impact toughness) and 15x9x70 mm (fatigue strength). They were heated rapidly (up to 200°C/s) to the peak temperature T, over 1350°C and then cooled down. The consequence of such a weld thermal cycle simulation on as-delivered BM is synthetic microstructure very similar to CGHAZ microstructure of the real weld joint adjacent to the bond. Different cooling times ∆T, were chosen in the range 4-400 s. They correspond to low, standard and high heat input welding.

Dilatation was measured during thermal cycle applications. Temperatures of austenite decomposition start (finish) during cooling and corresponding moments were determined by dilatation curves analysis. Microstructure was metallographic analyzed and hardness was measured. These data were sufficient for CCT diagram designing which is validating under welding conditions (1).

Mechanical properties at room temperature (RT) were tested by hourglass shaped tensile specimens with diameter D₀=4.5 mm and stress concentration factor K₁=1.05 (2). Instantaneous cross section during the test was measured by the circumference change x at the narrowest part of specimen by firm, flexible treat. True stress σ and strain ε at this position are function of load F, D₀ and x. Both were measured and calculated during the test up to the strain of -0.20.

\[ σ = K₁ \frac{4F}{π(D₀-x/π)^2} \quad ; \quad ε = -2 \ln \left(1 - \frac{x}{πD₀}\right) \quad . \quad . \quad . \quad . \quad (1) \]

Their relation describes Ludwick's equation.

\[ σ = A (ε)^n \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad (2) \]

Strain hardening exponent n and strength coefficient A were estimated by mean slope of log σ-log ε dependence in the range of 0.01-0.20 and by A=σ₀., True fracture stress σₜ and fracture deformation εₜ were measured by fracture load, final cross-section and undeformed cross-section after each tensile test. Yield stress σₚ and ultimate tensile strength σₜₚ were determined using engineering σ-ε graphs.

Impact toughness on standard V-notched specimens (10x10x55 mm, 2 mm deep notch) was measured at -40°C.

Bending fatigue strength on grooved specimens (14.5x8x70 mm, 3.2 mm deep rounded groove with r=3 mm) was measured at RT. Grooves were made by machining and subsequent fine polishing, both in the stress direction. Stress concentration factor K due to the groove was 1.74 (2). Artificial flaws were single and series of Vickers indentations shown in Figure 1. The flaw dimensions were ~110, ~165, ~220 µm (d) and ~365, ~680 µm (l). Characterizing dimensions of these flaws in the fracture mechanics sense were 35, 50, 70 and 100 µm (3). The specimens were fatigued at ~115 Hz in pulsating bending (R=0). Load range was progressively increasing up to the fatigue crack appearance.
RESULTS

CCT diagram for CGHAZ (thermal cycle to $T_r > 1350^\circ$C) with cooling curves and hardness valid under welding conditions is shown in Figure 2. Ultimate tensile stress $\sigma_t$, yield stress $\sigma_y$, and nominal bending fatigue strengths $\sigma_{lb}$, $\sigma_{lb}^{50}$, $\sigma_{lb}^{100}$ and $\sigma_{lb}^{150}$ on cooling time $\Delta t_{k3}$ at RT are shown in Figure 3. Fracture strength and deformation $\sigma_f$, $\epsilon_f$, strain hardening exponent $n$ at RT and impact toughness $\psi$ at $-40^\circ$C on $\Delta t_{k3}$ are shown in Figure 4. Data valid for as-delivered BM obtained under the same conditions are also shown in Figure 3 and 4.

DISCUSSION

Solid lines in Figure 2 represent mean values of all measured points, dash lines are logical assumptions based on metallographic analysis. Pure martensitic CGHAZ (M) is formed only at the shortest $\Delta t_{k3}<7$ s (first and second cooling curve in Figure 2, $\Delta t_{k3}=5$ and $-6.5$ s). As soon as $\Delta t_{k3}$ attains $7-9$ s, some bainite (B) is already present in CGHAZ (third curve, $\Delta t_{k3}=9$ s). The portion of bainite in CGHAZ increases to $\sim 50\%$ at $\Delta t_{k3}=25$ s (fourth curve). The result of austenite decomposition at $\Delta t_{k3}=100$ s is solely bainite (fifth curve). At $\Delta t_{k3}=400$ s traces of ferrite (F) are found in more or less bainitic CGHAZ (sixth curve). Hardness shown in Figure 2 was measured at the load 100 N (HV10). The hardest is martensitic CGHAZ microstructure ($\sim 380$ HV). As the proportion of bainite grows hardness decreases with $\Delta t_{k3}$ ($7-200$ s). Pure bainitic CGHAZ microstructure hardness at $\Delta t_{k3}=100$ s is $\sim 270$ HV. Due to ferrite formation at the start of austenite decomposition during the slowest cooling ($\Delta t_{k3}=400$ s) the rest of austenite is enriched on carbon. Bainite which is formed later is harder than that at $\Delta t_{k3}=100$ s. The reason is higher amount of coagulated cementite in it. This mixture of bainite and a bit of ferrite has hardness $\sim 250$ HV. Hardness of unaffected BM is $\sim 244$ HV.

Mechanical properties in Figure 3 decrease with cooling time $\Delta t_{k3}$ as hardness in Figure 2. Fracture stress $\sigma_t$ and impact toughness $\psi$ attain maximums at $\sim 9$ s. Maximum fracture deformation $\epsilon_f$ is recorded at higher $\Delta t_{k3}$ ($\sim 25$ s). Hardening exponent $n$ increases with $\Delta t_{k3}$, but its two distinct levels are obvious. Lower level is referred to CGHAZ microstructure where martensite prevails ($\sim 0.07$), while higher to that consisting mostly of bainite ($\sim 0.11$). Impact toughness $\psi$ of CGHAZ are much coarser at $133 \mu$m at $\Delta t_{k3}=5$ s and $250 \mu$m at $400$ s than grains of BM ($22 \mu$m) (1).

The size of artificial small flaw was transformed to equivalent geometrical parameter (EGP) using calculation procedure in reference (3). The supposition was its influence equivalent to that of the crack which shape and size is equal to flaw contour. The suggestion of Murakami and Endo (4) for EGP equal to square root of the area occupied by projecting the flaw onto the plane normal to maximum tensile stress gives almost the same EGP size for single Vickers indentations but not for series of them.

Local stress range at the bottom of grooves $\Delta \sigma$ is higher than nominal bending fatigue strength $\sigma_{lb}$ due to stress concentration. Ratio $\Delta \sigma/\sigma$ versus EGP is shown in Figure 3. Such kind of diagram, suitable to describe short crack influence on fatigue strength, is known by Kitagawa and Takahashi (5). In some
cases $\Delta \sigma$ is greater than yield stress. Figure 6 shows fatigue strength of smooth and flawed specimens in terms of tensile strength. Stress necessary for plastic deformation of infinitesimal volume at the bottom of groove $\sigma_{pl} = \frac{\sigma}{K}$, is added.

Naturally, maximum tensile stress cannot exceed yield stress not even locally due to plastic deformation. The consequence is R-ratio shift to $R<0$ at the bottom of the groove. CGHAZ microstructure was assumed to be cyclic stable due to its ratio $\sigma_{pl}/\sigma_y$ in the range 1.2-1.4 for all treated $\Delta \varepsilon_t$ (6). So, local R-ratio was assessed to be $R = (\sigma - \Delta \sigma)/\sigma_y$. Fatigue strength at $R=0$ was calculated using well known suggestion of fatigue strength dependence upon mean stress according to Goodman. By means of fatigue strength at $R=0$ and EGP size adequate threshold stress intensity range $\Delta K_{th}$ was calculated. The results are linearly strength dependent as already shown in Reference (4). So, $\Delta K_{th}$ values were unified according to $\sigma_y$ and $\sigma_e$. Figure 7 represents the dependence of unified $\Delta K_{th}$ upon EGP size. Actual $\Delta K_{th}$ values from both sides of strength interval valid for single CGHAZ microstructure ($\Delta \varepsilon_b = 4.5$ s) and BM are also given in the figure.

CONCLUSIONS

Small surface discontinuities which size is comparable to the characteristic dimensions of CGHAZ microstructure at the weld toe could influence the fatigue strength of macroscopic homogeneous weld joints. Decreasing of nominal bending fatigue strength with the size of small discontinuities is recorded. As hardest is CGHAZ microstructure at the weld toe as highest is the fatigue strength. On the other hand, possible local plastic deformation at the weld toe in the case softer CGHAZ microstructure referring to longer cooling times increasing in nominal bending fatigue strength of weld joints.

Threshold stress intensity of CGHAZ microstructure in the case of small defects is linear function of its strength and parabolic function of defect size.

REFERENCES


Figure 1  Artificial flaw shapes

Figure 2  CCT diagram for CGHAZ

Figure 3  Mechanical properties $\sigma_{y}$, $\sigma_{y}$, $\sigma_{u}$, $\sigma_{u}$, $\sigma_{u}$, $\sigma_{u}$, $\sigma_{u}$, $\sigma_{u}$ (at RT) at RT

Figure 4  Mechanical properties $\sigma_{y}$, $\sigma_{y}$, $\sigma_{y}$ (at RT) at $-40^\circ$C

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Figure 5. $\Delta \sigma/\sigma$ in terms of EGP size for CGHAZ (••••) and BM (•)

Figure 6. Nominal fatigue strength and yield stress in terms of tensile strength

Figure 7. Unified and actual threshold stress intensity factors in terms of EGP size