Fracture Mechanism and Toughness of the Welding Heat-Affected Zone in Structural Steel under Static and Dynamic Loading

H. QIU, H. MORI, M. ENOKI, and T. KISHI

Due to the influence of the welding thermal cycle, the toughness of structural steel generally degenerates. Recently, the intercritically reheated coarse-grained heat-affected zone (IC CG HAZ) was found to demonstrate the worst toughness in welded joint, which was associated with its fracture mechanism. In this article, two IC CG HAZs of a structural steel were prepared by welding thermal-cycle simulation techniques. For the two IC CG HAZs, the static and dynamic fracture toughness were evaluated; the fracture mechanism was also studied. Under both static and dynamic loading, cracks in the IC CG HAZ were found to initiate at the intersection of bainitic ferrite $\alpha_b$ packets with different orientations, followed by propagation in cleavage. In some crack propagation regions, adjacent cleavage facets are connected by shear, thus producing dimple zones. Though the brittle fracture initiation mechanism remains unchanged, the cleavage facet size, the proportion of the dimple zones between facets, and the distance from the cracking initiation site to the crack tip vary with loading speed and welding conditions. These changes were found to be related to the variations caused by strain rate and welding conditions in fracture toughness for the IC CG HAZs.

I. INTRODUCTION

A special region denoted as the heat-affected zone (HAZ) forms as a result of thermal cycle experience in the parent metal during welding. For single-pass welding, the region of lowest toughness is generally associated with the coarse-grained heat-affected zone (CG HAZ).[1,2] During multipass welding, the HAZ formed by the previous welding heat cycle is modified by the subsequent thermal cycles, forming localized and discontinuous zones. In the HAZ formed by multipass welding, the IC CG HAZ is found not only to demonstrate low toughness,[3-6] but also to have lower toughness than CG HAZ.[3,6,7]

The low toughness of the CG HAZ and the IC CG HAZ was attributed to local brittle microstructure M-A constituent (high-carbon martensite with some retained austenite).[3,5,6,8-10] Further studies discovered that for submerged arc welding, brittle fracture initiation points are associated with the intersections of $\alpha_b$ areas with different orientations rather than with the M-A constituent.[7,11,12] Moreover, other microstructures, such as a banding structure comprised of ferrite and cementite similar to pearlite, precipitated in high heat input welding, also affect toughness.[13] Therefore, for a given steel, its fracture mechanism should relate to its microstructures and welding parameters.

In this work, the brittle fracture initiation and crack propagation of the IC CG HAZ of a structural steel under static and dynamic loading were examined. Through the analysis of its fracture mechanism, the effect of strain rate on fracture toughness and the dependence of fracture toughness on fracture mechanism were revealed. The influence of thermal experience caused by welding on fracture toughness under static and dynamic loading was also investigated.

II. EXPERIMENTAL

A hot-rolled steel, SN490, with chemical compositions 0.11C/0.29Si/1.39Mn/0.01P/0.02S wt pct was selected for this work. SN490 consists of ferrite and pearlite, as shown in Figure 1.

In fact, the actual IC CG HAZ in a welded joint is so narrow that it is sometimes impossible to prepare test specimens. To overcome this problem, welding thermal-cycle simulation techniques were applied in this work. Since simulated HAZ tests give the same toughness ranking, although not the same absolute values,[3] as those on actual welding HAZs, a simulated IC CG HAZ can be used to substitute for an actual IC CG HAZ.

The typical welding thermal cycles are shown in Figure 2. The terms $T_{p1}$ and $T_{p2}$ are the peak temperatures of the first and second thermal cycles, respectively. The time cooling from 800 °C to 500 °C is denoted as $t_{8/5}$. Two thermal series, TA ($T_{p1} = 1350 °C$, $T_{p2} = 750 °C$, and $t_{8/5} = 30 s$) and TB ($T_{p1} = 1350 °C$, $T_{p2} = 750 °C$, and $t_{8/5} = 75 s$), were applied. For the welding of thick plate, $t_{8/5}$ is approximately proportional to heat input $E$

$$t_{8/5} = kE$$  \[1\]

where $k$ is in the range of 4 to 5 seconds per kJ/mm for conventional structural steels.[14] We calculated the welding heat input with Eq. [1] and obtained the heat inputs of TA and TB of about 60 and 125 kJ/cm, respectively. The simulated HAZs of the two thermal cycle series are equivalent to the actual HAZs generated by submerged arc welding.

Three-point bending fatigue precracked specimens with
lengths of 55 mm, widths of 10 mm, and thicknesses of 10 mm were applied for fracture toughness tests. A static fracture toughness $K_{IC}$ was obtained from $J_{IC}$ that was measured in accordance with the ASTM E813-81 standard\cite{15}. The dynamic fracture toughness $K_{Id}$ tests were conducted on an instrumented Charpy testing machine at an impact velocity of about 4.9 m/s. In the $K_{Id}$ tests, the strain signal was taken from the strain gage, which is very near the fatigue precrack tip, stuck on the specimen. Briefly, the $K_{Id}$ measurement procedure includes two parts: one is to get the curve of load vs strain under static loading, and the other is to obtain the impact curve of strain against time with an instrumented Charpy testing machine. The maximum strain point in the impact curve is regarded to be the cracking initiation point, and the load corresponding to this point could be determined from the calibration curve of load vs strain obtained under static loading. Substituting the load into the following equation\cite{16} yields

$$K_{Id} = \frac{PS}{BW^{3/2}} f \left( \frac{a}{W} \right)$$

where $P$ is the load, $S$ is the span, $B$ is the specimen thickness, $W$ is the specimen width, $a$ is the crack length, and $f(a/W)$ is a function of $(a/W)$ given by ASTM E399-83\cite{16}. Both the $J_{IC}$ and $K_{Id}$ tests were performed at room temperature.

The microstructures of the IC CG HAZ were examined with both an optical and a transmission electron microscope (TEM). The fracture surface of fracture toughness test specimens were observed with a scanning electron microscope (SEM). To identify the microstructural features of brittle fracture initiation, the fracture surfaces were etched with 3 pct nital.

## III. RESULTS

### A. Microstructures

The CG HAZ microstructures of TA and TB are shown in Figures 3(a) and (b), respectively. Their major microstructures are upper bainite with a large prior austenite grain size. Prior austenite grain boundaries are distinct in the CG HAZs of TA and TB. Precipitated ferrite is clearly distinguished, especially at the prior austenite boundaries of the TB CG HAZ. The second welding thermal cycle reheats the CG HAZ up to $T_{p2}$ (750°C). Since the $T_{p2}$ is above $A_{C1}$, precipitated ferrite located at prior austenite boundaries preferentially transforms into austenite. The formed austenite grows and retransforms into upper bainite on cooling. The upper bainite within prior austenite grains could be recognized as unaffected by the second thermal cycle. Because of insufficient time, phase transformation from ferrite into austenite at the prior austenite boundaries is not complete in the CG HAZ of TB. It can be seen in Figures 3(c) and (d) that distinct prior austenite boundaries with precipitated ferrite are still present in the IC CG HAZ of TB, but disappear in the IC CG HAZ of TA. The prior austenite grain sizes of TA and TB are about 100 and 150 μm, respectively. It should be noted that the prior austenite grain size of TB was obtained from the micrographs of the IC CG HAZ, but that of TA was obtained from the CG HAZ instead of from the IC CG HAZ. The M-A constituents are not found in the IC CG HAZs of both TA and TB (in the following, the IC CG HAZs of TA and TB are simply denoted as TA and TB, respectively).

In terms of the nomenclatures defined in Reference 17, the microstructures of TA and TB are mostly bainitic ferrite $\alpha_{q}$ and quasipolygonal ferrite $\alpha_{q}$, as shown in Figure 4. They distribute in $\alpha_{B}$ packets and mixed regions of $\alpha_{B}$ and $\alpha_{q}$. Within each $\alpha_{B}$ packet, $\alpha_{B}$ almost holds the same orientation. The $\alpha_{B}$ packet area fractions in TA and TB are about 7.3 pct and 10 pct, respectively. The size distribution of $\alpha_{B}$ packets in TA and TB is shown in Figure 5, which indicates that for TB, each $\alpha_{B}$ packet distributes over a wide range, and large packets are present. The average sizes of $\alpha_{B}$ packets in TA and TB are about 27 and 43 μm, respectively.

### B. Fracture Toughness

The values of $K_{IC}$ for TA and TB are 53 and 48 MPa√m, respectively, and the $K_{Id}$ values for them are 42 and 37 MPa√m, respectively. TA has a higher fracture toughness than TB at both static and dynamic loading. For the impact tests in which the impact velocity is about 4.9 mm/s, the strain rate $\dot{e}$ near the crack tip in TA and TB is about 5.2 $\times$ 10^{-3} s^{-1} and 4.5 $\times$ 10^{-3} s^{-1}, respectively, and the changing rate of dynamic fracture toughness $K_{Id}$ for TA and TB is 7.6 $\times$ 10^{-5} and 8.1 $\times$ 10^{-5} MPa√m/s, respectively.

### C. Fractographic Features of $J_{IC}$ and $K_{Id}$ Specimens

The SEM fractographs of $J_{IC}$ and $K_{Id}$ specimens of TA are shown in Figure 6. It can be seen that ductile fracture