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Analysis of microstructural evolution during friction stir welding of ultrahigh-strength steel

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Friction stir lap-welded advanced high-strength M190 steel exhibited formation of lath martensite at weld nuggets, a ferrite layer at faying surfaces and a ferrite–pearlite microstructure at the bottom of the stir zone. The phase transformation was governed by severe plastic deformation in the austenitic region followed by cooling. The strain rate and peak temperature played key roles in controlling the prior-austenitic grain size and were correlated to the Zener–Hollomon parameter. An empirical relationship has been obtained for dynamically recrystallized austenite grains.

Keywords: Friction stir welding; Martensitic steel; Dynamic recrystallization; Zener–Hollomon parameter

M190 steel is an advanced low-carbon high-strength steel, with a strength of more than 1000 MPa, and is a promising alloy for use in the automotive sector because of its high strength-to-weight ratio and adequate formability. Joining is an essential processing step as it is required to mitigate the complexity of manufacturing large structures. Fusion welding cannot be applied as large heat input alters the microstructure, detoriates the mechanical properties, promotes solidification cracking related to segregation of alloying elements, expedites hydrogen embrittlement and creates blowholes at the weld nugget (WN). Friction stir welding (FSW) is a promising technique that has advantages such as low heat input accompanied by severe plastic deformation in the austenitic range. Friction between the tool and the workpiece plasticizes the metal below the shoulder and the plasticized material is moved by the tool's rotation and sideways movement. The weld is produced with less residual stress and distortion in comparison to fusion welding. The thermo-mechanical cycle experienced by the material in the stir zone (SZ) of steel essentially includes hot working. The microstructure resulting from the influence of plastic deformation at elevated temperature is characterized by a central WN. The nugget is thus subjected to the highest strain and strain rates, which in turn are related to the process parameters. Few reports are available on friction stir welding/friction stir spot welding of high-strength steels [1–4]. Most of the studies deal with the WN microstructure, microhardness profile and tensile shear test of the welded joints. Apart from simple grain refinement at SZ, FSW of steel involves thermal impact, thermomechanical processing and phase transformation, which are still not clearly understood. The effect of grain refinement is controlled by the strain rate and stress generation during the process at that elevated temperature and information is lacking in this respect for FSW of ultrahigh-strength steel. The present paper thus presents a clear understanding of microstructural evolution in the WN in relation to the quantified data of strain rate and stress for friction stir-welded martensitic steel.

FSW was carried out to make lap joints between 1.0 mm thick martensitic M190 steel sheets having the composition 0.19C–0.47Mn–0.17Si–0.005S–0.009P–0.0034Ti (wt.%). The welding conditions are summarized in Table 1. Optical metallography was performed for transverse sections. To reveal the prior-austenitic grain size, the samples were etched with acidified ferric chloride and the average grain size was determined by the mean linear intercept method. The microhardness distribution along the weld centre through the thickness direction was evaluated. Microhardness measurements were taken from the

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Table 1. A summary of the welding conditions for the lap joints.

Sample ID	Tool description	Tool rotation rate and tilt	Processing condition	Linear velocity (mm min^{-1})
Sample 1	Hybrid carbide tool:	1000 rpm and 2.5°	Forced air cooling from	50.8
Sample 2	shoulder ϕ 10 mm,		the top and a chilled	101.6
Sample 3	cylindrical pin ϕ 4 mm,		backing plate	203.2
Sample 4	pin length 1.1 mm			304.8

face of the weld to its root through the centre line, with a distance of 0.1 mm between successive measurements.

The nugget microstructure of the welds is shown in Figure 1. The microstructure in the SZ was predominantly martensitic. It had been reported by Arbegast [5] that the maximum temperature at the SZ was predominantly controlled by tool rotation and the rate of heating/cooling was governed by the traversing speed of the tool for constant tool dimensions and process parameters. Based on the above assumption, an empirical relationship was developed and expressed as [5]:

$$T'/T_{\mathrm{m}} = K[w^2/v \times 10^4]^{\alpha} \tag{1}$$

where T' is the temperature during welding (°C), T_{m} is the melting point of alloy (°C), w is the tool rotational speed, v is the tool traversing speed, and α (~0.005) and K (~0.7) are constants. Using this relationship, T' was calculated and found to be $\gg A_3$ for all the joints (Table 2). The starting martensitic microstructure transformed to single-phase austenite during the heating cycle. The austenite subsequently underwent martensitic transformation after the passage of the tool owing to the substantial hardenability of M190 steel [1]. Martensitic transformation was primarily governed by the tool's traversing speed: a higher tool traversing speed resulted in faster cooling of the nugget [6]. All the samples exhibited a sharp drop in microhardness (~200 VHN) at the faying surfaces (Fig. 2a, along the x′–x imaginary line, and Fig. 2b–e), where a thin ferrite layer occurred along with a small volume fraction of pearlite (Fig. 3a). The occurrence of a ferrite layer had been also reported and discussed by Hovanski et al. [4] for a 1.4 mm thick friction stir spot-welded 0.2% carbon steel sheet, but its origin is still not understood. It was postulated that in the as-received condition the processed rolled sheet contained oxide scales and that, during welding, the oxide layer reacted with the carbon because the elements responsible primarily for oxide formation, like Al and Si, dissolved. This reaction locally reduced the hardenability of M190 steel at the faying surfaces and the ferrite layer was formed. As FSW was carried out in an open atmosphere, a small quantity of decarburization during welding might be also responsible for ferrite layer formation. Below the ferrite layer, the top of the second sheet also exhibited martensite formation, and this was supported by the hardness profile (Fig. 2a, below the x′–x imaginary line, and Fig. 2b–e).

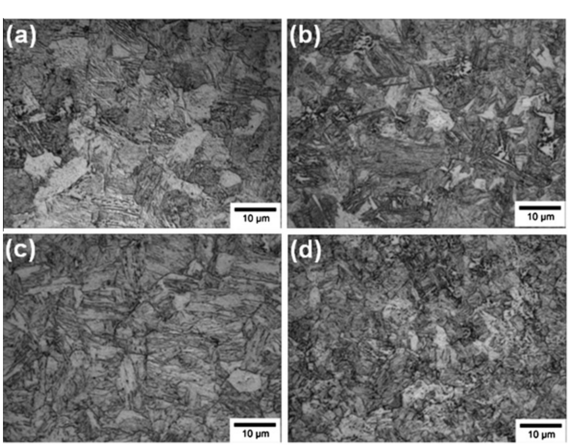

Figure 1. Optical microstructure of WN showing prior-austenitic grains: (a) sample 1, (b) sample 2, (c) sample 3, and (d) sample 4.

The bottom of the welds (Fig. 2a, below the line y′–y up to e′–e) showed the formation of ferrite–pearlite aggregate and its signature was given by downward trend in hardness plot (Figs. 2b–e and 3b). The reason for the reduction in hardness at the bottom was the slow heat dissipation through the supporting backing plate and the negligible effect of forced air cooling. In the present investigation, the martensite in the base alloy illustrated marked difference in microhardness from that in the WN. The reason for this might be the presence of soft phase constituents: granular bainite, degenerated upper bainite, and lath martensite are all present in the SZ of friction stir-welded 0.13% carbon steel [7]. The presence of the bainite was perhaps responsible for the reduction in hardness. The first part of the heat-affected zone is referred as HAZ-1, and is shown in Figure 2a as areas B′b′a′A′ and BbaA. The microstructure in HAZ-1 was polygonal ferrite–pearlite after transformation from the intercritical temperature during welding (Fig. 3c) [2]. The second part of the heat-affected zone is referred as HAZ-2 and is located between the base material and HAZ-1 (Fig. 2a, areas D′d′b′B′ and DdbB). In HAZ-2, the microstructure consisted of tempered martensite as the temperature rise in this location was below A_1 during welding (Fig. 3d) [8].

The degree of fineness of martensitic lath in WN was governed by the prior-austenitic grain size (Fig. 2a, region A′a′aA) and the cooling rate. No predominant difference in prior-austenitic grain size was observed along the thickness direction of the welds. Under severe plastic deformation during FSW, the austenite grain size depended on the strain rate and the temperature of deformation. These two variables can be expressed by the Zener–Hollomon parameter (Z) as [9]

$$Z = \acute{\varepsilon} \exp\left(\frac{Q}{RT}\right) \tag{2}$$

where $\acute{\varepsilon}$ is the strain rate, Q is the activation energy of the process, R is the universal gas constant, and T is the deformation temperature. The upper limit of the to-