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Improvement of Mechanical Properties in Structural Steels by Development of Acicular Ferrite Microstructures

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The weld heat affected zone (HAZ) can develop coarse austenite grain sizes when high heat inputs are applied, in the absence of second phase particles inhibiting grain growth. As a consequence of this, the final microstructure at room temperature will be coarse, exhibiting poor toughness. In this work, the improvement of toughness in the HAZ by the production of acicular ferrite is considered. The material studied is a low carbon V-microalloyed steel (0·10%C, 0·96%Mn, 0·19%Si, 0·03%V, 0·0104%N) produced by ARISTRAIN (Olaberria, Spain). After austenitization at 1250°C ($D_\gamma = 182\ \mu m$), the steel is quenched to a temperature in the range 550–350°C in which the isothermal transformation of the austenite takes place.

After holding at 550°C, the microstructure is composed of acicular ferrite with some mixture of allotriomorphic and widmänstatten ferrite, and bainite. In the microstructure corresponding to the treatment at 450°C, allotriomorphic ferrite is not found but widmänstatten ferrite and bainite are still present. The fraction of acicular ferrite is now significantly higher than in the preceding case. On transforming at 350°C, a very similar microstructure to the one obtained at 450°C is reached but the size of each acicular ferrite unit is smaller. The size of the active particles nucleating acicular ferrite is between 0·3 and 2 μm and those have been identified as oxides (MnO, SiO_2 and Al_2O_3) and Mn sulfides, both of them covered by a S–Cu rich shell.

Tensile test results show that the acicular ferrite has a yield strength in the range from 480 to 515 MPa and a UTS from 630 to 660 MPa with an elongation of 35%. At 0°C, the absorbed Charpy energy of the 550°C isothermally transformed sample is very low (brittle fracture). The acicular ferrite structures obtained at 450 and 350°C exhibit Charpy energies higher than 100 J at 0°C and, at lower testing temperatures, there is a smooth ductile-brittle transition, with an absorbed energy at −40°C higher than 50 J.

In the brittle regime, the fracture surface of the 550°C treatment is mainly composed of coarse cleavage facets; on the other hand, the samples corre-

sponding to 450 and 350°C treatment show very fine facets (some isolated coarse cleavage facets can be identified also). In the macroscopic ductile behaviour range, the fracture surfaces of the 350 and 450°C transformed samples exhibit some cleavage islands totally surrounded by ductile regions. In some cases, inside the islands the brittle particle responsible for cleavage nucleation has been identified (complex calcium aluminates and silicates).

The brittle cleavage fracture in steels has been attributed to the nucleation of a microcrack in some microstructural feature (step 1), followed by its propagation across the particle-matrix boundary (step 2) and across the matrix-matrix barriers (step 3). In the present steel, in some cases the cleavage origins have been related to the cracking of non-metallic inclusions; in other cases, it has not been possible to relate the initiation to a microstructural feature (carbides). When the microstructure of the steel is acicular ferrite, steps 2 and 3 of cleavage propagation seem to exhibit some special characteristics. The propagation of the microcrack across particle-matrix interface would be easier with a good parallelism between the cleavage planes of the particle and those of the matrix. In the case of acicular ferrite, the probability of having this type of configuration is low and the cracks will stop at the interface and blunt, acting as nuclei for void formation.

If the microcrack surmounts the particle-matrix interface barrier, step 3 of the cleavage process must be considered. A coarse microstructure contributes to an easy propagation of the microcrack across matrix-matrix boundaries. When the microcrack has trespassed the particle-matrix boundary, the stress required to continue the catastrophic propagation depends on the microcrack size, being lower for coarser microstructures. The coarse cleavage facets, in the case of a 550°C isothermally formed microstructure, show that the size of the transformation products is not able to stop the cracks at step 3, giving place to poor Charpy energies in a wide testing temperature range.

In the case of 350 and 450°C isothermal treatments, owing to the low volume fraction or even the lack of other transformation products, the toughness is mainly controlled by the interlocking acicular ferrite microstructure. Taking into account the small size of each acicular ferrite unit (smaller for lower transformation temperatures), microcracks are stopped at high angle boundaries (higher toughness values for a wide working temperature range).

The results obtained show that it is possible to achieve good strength-toughness combinations with conventional constructional steels with acicular ferrite microstructures, formed from coarse austenite grain sizes, similar to those formed in the HAZ. From the point of view of industrial application, it would be necessary to complete the research considering the production of acicular ferrite microstructures by continuous cooling transformations.

KEYWORDS

Acicular ferrite, vanadium microalloyed steels, toughness, cleavage fracture, intragranular transformation, fracture micromechanisms.