Shape effect of ultrafine-grained structure on static fracture toughness in low-alloy steel

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Abstract
A 0.4C-2Si-1Cr-1Mo steel with an ultrafine elongated grain (UFEG) structure and an ultrafine equiaxed grain (UFG) structure was fabricated by multipass caliber rolling at 773 K and subsequent annealing at 973 K. A static three-point bending test was conducted at ambient temperature and at 77 K. The strength–toughness balance of the developed steels was markedly better than that of conventionally quenched and tempered steel with a martensitic structure. In particular, the static fracture toughness of the UFEG steel, having a yield strength of 1.86 GPa at ambient temperature, was improved by more than 40 times compared with conventional steel having a yield strength of 1.51 GPa. Furthermore, even at 77 K, the fracture toughness of the UFEG steel was about eight times higher than that of the conventional and UFG steels, despite the high strength of the UFEG steel (2.26 GPa). The UFG steel exhibited brittle fracture behavior at 77 K, as did the conventional steel, and no dimple structure was observed on the fracture surface. Therefore, it is difficult to improve the low-temperature toughness of the UFG steel by grain refinement only. The shape of crystal grains plays an important role in delamination toughening, as do their refinement and orientation.

Keywords: medium-carbon steels, toughening, warm caliber rolling, microstructure design, grain refinement

1. Introduction

Achieving high strength in combination with resistance to fracture is an eternal issue in the field of structural metallic materials. With the current materials science and technology, it is not difficult to achieve high strength. However, as shown in figure 1, strength and toughness in materials are strongly correlated, and toughness decreases with increasing strength [1, 2]. Few structural metallic materials are limited by their strength; rather, they are limited by their fracture toughness. Unfortunately, materials research is still mostly focused on the quest for high strength. It is reported that the refinement of crystal grains is an effective method for developing strength and toughness in metallic materials without the addition of alloying elements; hence, ultrafine-grained materials are very attractive in materials science [3–9]. However, compared with microstructure evolution through severe plastic deformation and strengthening by grain refinement, there have been few studies on the toughness improvement in ultrafine-grained materials.

Natural materials such as bamboo, wood, nacre, dentine, rock and bone achieve an optimum combination of strength and toughness through complex hierarchical designs [10]. In particular, nacre (abalone shell), which is often cited as ‘gold standard’ in biomimetic design, consists of a fine-scale layered brick-like structure comprising ∼95vol.% of sub-micrometer-thick (about 500 nm) layered aragonite (CaCO₃) plates bonded by thin (20–30 nm) layers of organic protein material [11]. The resulting composite has a fracture toughness that is an order of magnitude higher than those of either the aragonite or the protein layers. This
property results, however, not simply from multidimensional architecture, but also from precisely designed interfaces. Such biological materials can generate fracture toughness primarily by extrinsic toughening mechanisms that shield any crack during applied loads. Extrinsic toughening mechanisms affect only the crack growth and have little effect on the crack initiation [12]. The microstructures are heterogeneously designed to be fail-safe [13]; i.e. the material does not fail completely even if many microcracks occur at weak sites under applied load. Based on such design, composite materials and laminate materials have been developed [12, 14–16]. The crack-arrestor type [17–19] shown in figure 2(a) is normally referred to as a lamella fracture, and it is widely reported in the literature as a method for enhancing toughness. Kum et al. [20] reported an extremely low ductile-to-brittle transition temperature of 133 K and very high upper shelf energy of over 325 J in laminated composites containing ultrahigh-carbon steels. This excellent toughness has been related to crack branching as a result of delamination at the weak interface shown in figure 2(a). McEvily and Bush [21] observed the lamella fracture and measured a high absorbed energy of 325 J for V-notch impact testing at 473 K of an ausformed 0.2%C–3%Ni–3%Mo steel. At this time, a main crack branched along elongated austenite grain boundaries. However, because of the relatively large transverse size of the grains, no lamella fracture was observed at room temperature, and the absorbed energy was reduced to 33 J for the samples with a tensile strength of 1.6 GPa. Munch et al. [22] reported that bio-inspired ceramic composites consisting of brick-and-mortar structures are more effective for toughening than conventional lamellar composites. On the other hand, the separation/splitting of the crack-divider type shown in figure 2(b) has often been observed elsewhere, such as in rolled steel pipes and plates and Al-Li alloys. The delamination of this type leads to the improvement of toughness at low temperatures because of relaxation of the triaxial tension stresses generated by the localized plastic constraint at the crack tip. Bourell [23] showed that for low-carbon steel, the separations markedly appeared with increasing rolling strain and decreasing rolling temperature due to the development of {100}<110> texture (i.e. the aligned {100} cleavage planes acted as the weak sites); hence, the upper shelf energy decreased and the lower shelf energy increased. Rao and Ritchie [24] reported that Al–Li alloys with pancake-shaped grains grossly elongated in the rolling direction, which are used for the main fuel tank of the Space Shuttle, had high toughness at liquid helium temperatures [12]. This is attributed to crack divider delamination toughening, which is caused by the presence of weak sites, such as coarse Fe–Cu or Mg–Cu rich constituent particles (1–2 µm in diameter) and the segregation of Na, K and Li to grain boundaries.

The notion of mimicking natural structures has generated enormous interest [10, 11, 25–27] but has not been sufficiently developed for structural metallic materials because it is very difficult to design fine-scale complex hierarchical structures with conventional fabrication techniques. Heterogeneous microstructural designs such as those of natural and biological materials should be applied to ultrahigh-strength materials, and we should search for an optimum relation between toughness and microstructure factors, such as grain size, grain shape, texture, impurities and second-phase particles, in the simple production process without using much energy. We have studied an ultrafine elongated grain structure steel with a strong fiber deformation texture in the <|110>||rolling direction, processed by multipass warm caliber rolling (WCR), and reported an improvement in the upper-shelf energy of the Charpy V-notch impact properties upon strengthening [8, 28, 29]. This enhanced impact toughness is attributed to extrinsic toughening mechanisms of the crack-arrestor type shown in figure 2(a). Fracture mode of the crack branching planes occurred primarily by a quasi-cleavage. The property of toughness is sensitive to not...
only the size of crystal grains and their texture but also to their shape [8, 19, 28–31]. It is of interest to systematically study the effect of these parameters on the static fracture toughness in ultrafine-grained steel.

In this study, we focus on the microstructural design of grain refinement for improved toughness. Low-alloy steels with an ultrafine elongated grain (UFEG) structure and an ultrafine equiaxed grain (UFG) structure were fabricated by multipass caliber rolling at warm working temperatures and subsequent annealing. A static three-point bending test was conducted at ambient temperature and at 77 K, and the shape effect of refined crystal grains on delamination toughening was studied, including crack propagation dependent on the microstructural features.

2. Experimental details

2.1. Specimen preparation

A low-alloy steel with a chemical composition of 0.39 C, 2.01 Si, 1.02 Cr, 1.0 Mo, 0.21 Mn, <0.001P, <0.001 S, 0.004 Al, 0.0022 N, 0.001 O and the balance Fe (all in wt%) was used in this study to reduce the effect of chemical impurities on fracture behavior. A 100 kg ingot was prepared by vacuum induction melting and casting, homogenized at 1473 K, and then hot-rolled to a 40-mm-thick plate. A 40 × 40 × 120 mm³ block was cut out of the plate, solution-treated at 1473 K for 1 h to reduce undissolved carbide particles, and then hot-rolled into a rectangular bar of about 31 × 31 mm² cross section, followed by water quenching to obtain the martensitic structure. The quenched bar was soaked at 773 K for 1 h and then subjected to a caliber rolling simulator of the square/square type without any lubricant [32], as shown in figure 3(a). The total reduction in area through the WCR was about 78%; thus, a 14.3 × 14.3 × 930 mm³ rolled bar has been produced. The sample was held for 300 s in a furnace after every three passes during the rolling process to maintain the rolling temperature of 773 K. For a final groove of 14.3 mm square, the sample was passed through twice to control the cross-sectional shape of the bar, and then air cooled. Hereafter, this sample is designated as the TF sample. To obtain quenched and tempered samples with the same tensile strength for comparison, a normalized bar has been produced. The sample was held for 300 s in a furnace after every three passes during the rolling process to maintain the rolling temperature of 773 K. For a final groove of 14.3 mm square, the sample was passed through twice to control the cross-sectional shape of the bar, and then air cooled. Hereafter, this sample is designated as the TF sample. To obtain quenched and tempered samples with the same tensile strength for comparison, a normalized bar was solution-treated at 1223 K for 0.5 h, followed by oil quenching. Next, the bar was tempered at 773 K for 1 h and then water-cooled. Hereafter, this sample is designated as the QT sample. To clarify the shape effect of crystal grains on toughness, the TF sample was annealed at 973 K for 1 h [8]. Hereafter, this sample is designated as the TFA sample. The principal axes of the rolled bar in this study are defined as shown in figure 3(b). Note that the loading direction (LD) in the three-point bending tests has an angle of ~45° to the transverse direction (TD) and normal direction (ND) of the rolled bars.

2.2. Microstructure and mechanical properties

The appearance of specimens after the three-point bending test was observed through a digital camera, crack paths in the central parts of the cross sections in the rolled bars were examined under a digital microscope (OM), KEYENCE VHX-900, and the fracture surfaces were imaged with a scanning electron microscope (SEM, KEYENCE VE-7800), operated at 15 kV. The microstructures were observed using the electron back-scattered diffraction (EBSD) method in an SEM equipped with a field emission gun.

Samples for mechanical tests were taken from the middle of the rolling direction (RD) plane, as shown in figure 3(b). Tensile tests were conducted at ambient temperature with a cross-head speed of 0.85 mm min⁻¹, and at 77 K with a crosshead speed of 0.5 mm min⁻¹, using cylindrical specimens with a diameter of 6 mm and a gage length of 30 mm. To prepare single-edge bend specimens for three-point bending tests, rectangular bars of 10 × 10 × 55 mm³ were first machined along the RD, and then a notch with a depth of 1a0 = 5 mm and a root radius of ρ = 0.13 mm was introduced using electrodischarge machining with fine wire 0.1 mm in diameter. A three-point bending test was conducted at ambient temperature and 77 K, using a support distance of 40 mm at a cross-head speed of 0.5 mm min⁻¹. They were terminated when the specimen fractured completely, the load dropped below 100 N, or the displacement reached 10 mm, which equals the specimen width. The load-displacement response was recorded with a time step of 0.05 s. Nonlinear fracture mechanics methods based on ASTM Standard E1820-01 [33] were used to evaluate the fracture toughness, J [34].
3. Results

3.1. Microstructure evolution

Figure 4 shows the image-quality maps of the TF and TFA samples obtained by the EBSD analysis on the cross-section of the plane normal to the RD and of the plane along the RD, as well as the corresponding inverse pole figures (IPFs). Compared with the QT sample, exhibiting a randomly oriented martensitic structure, the TF and TFA samples were dominated by an \( \alpha \)-fiber texture parallel to the RD (RD\|\langle 110 \rangle) regardless of the annealing treatment. Furthermore, the IPFs show that the LD\|\langle 001 \rangle texture and LD\|\langle 110 \rangle texture, which cause delamination by crack branching as shown in figure 2(a) [28, 29, 34], are produced in both bars, and the intensities of the LD\|\langle 001 \rangle texture are larger than those of the LD\|\langle 110 \rangle one. Although the \( \alpha \)-fiber texture is somewhat stronger in the TF than in the TFA sample and the LD\|\langle 001 \rangle and LD\|\langle 110 \rangle textures are somewhat weaker in the TF than in the TFA sample, there are no significant differences in the texture between the TF and TFA samples. In the TF sample, the average values of the transverse linear interceptions for the elongated grains with misorientation angles of more than 10° and 15° in the TF sample were 310 and 330 nm, respectively (figure 2(b)). Moreover, spheroidal nanometer-sized (<50 nm) carbide particles were dispersed in the elongated grain matrix, as shown in [8]. A granular structure has developed in the TFA sample (figures 4(d) and (e), and the average transverse sizes of the ferrite grains with misorientation angles of more than 10° and 15° increased to 660 and 720 nm, respectively, as shown in figure 4(e), in response to the growth of the carbide particles. The sizes of not only ferrite grains but also carbide particles increased as shown in ref. [8]. Relatively large carbide particles (200–300 nm) appeared at the grain boundaries, while finer carbide particles were dispersed in the grains. We have also analyzed the distributions of kernel average misorientation (KAM), which has a strong correlation with the dislocation densities within the cell interior of the deformed structures, using the EBSD maps of figures 4(b) and (e). The average KAM values for the TF and TFA samples were 0.65° and 0.38°, respectively, indicating a higher dislocation density for the TF sample.

3.2. Static mechanical properties

The static mechanical properties, including fracture toughness, \( J \), at ambient temperature, are summarized in table 1.

Figure 5 shows the tensile test results at ambient temperature. Although strengthening is usually accompanied by a drop in ductility, the UFEG steel exhibited superior ductility, despite the high yield strength. Generally, it has been reported that the plastic instability or necking during tensile tests occurred immediately after the tensile stress reached the yield point by grain refinement [6, 35]. However, the TF and TFA samples exhibited superior uniform elongation, as well as post-uniform elongation related to the area reduction. Such superior tensile properties have been reported for other caliber-rolled steels with different compositions [29, 36, 37]. Furthermore, steel wires with UFEG structures characterized by an \( \alpha \)-fiber texture as in the present steel bars, exhibit superior tensile ductility despite their ultrahigh strength [38]. In a study related to grain refinement, the UFG structure formed by the WCR exhibited superior reduction in area despite showing an increase in strength and a decrease
Table 1. Static mechanical properties at ambient temperature.

| Sample | $\sigma_{ys}$ (GPa) | $\sigma_B$ (GPa) | Uniform elongation $\varepsilon_u$ (%) | Total elongation $\varepsilon_t$ (%) | Reduction in area $\delta$ (%) | Fracture energy $J$ (kJ m$^{-2}$) | Strength–toughness balance $\sigma_{ys} \times J$ (GPa kJ m$^{-2}$) | $\sigma_B \times J$ (GPa kJ m$^{-2}$) |
|--------|---------------------|------------------|--------------------------------------|-------------------------------|-------------------------------|----------------------------------|-------------------------------------------------|----------------------------------|
| QT     | 1.51                | 1.82             | 4.60                                 | 9.20                          | 28.3                          | 129                              | 195                                             | 235                              |
| TF     | 1.86                | 1.86             | 7.00                                 | 14.8                          | 40.2                          | 5184                             | 9642                                            | 9642                             |
| TFA    | 0.99                | 1.06             | 9.80                                 | 22.0                          | 51.3                          | 2193                             | 2171                                            | 2325                             |

Figure 5. Stress–strain curves at ambient temperature.

Figure 6. (a) Bending load versus displacement curves at ambient temperature. Appearance of (b) QT, (c) TF and (d) TFA samples after bending test.

in uniform elongation [29, 39]. On the other hand, the presence of finer carbide particles, homogeneously dispersed in the ferrite matrix, improves the uniform elongation characteristics of UFG steels [40, 41]. These results explain the superior strength–ductility balance of the TF and TFA samples.

Figure 6 shows the bending load versus displacement curves, recorded at ambient temperature, and the appearance of the samples after the bending test. In the QT sample, the cracks propagated directly across the center part of the test bar, as observed normally after the bending test. As shown in figure 6(a), the QT sample fractured at a peak loading $P_{\text{max}} = 12.3$ kN and displacement $u = 0.56$ mm (see also figure 8(a)), exhibiting catastrophic fracture, i.e. typical brittle fracture behavior. The fracture surface showed a quasi-cleavage of a martensite structure (figure 7(a)). In the TF sample, the crack propagated vertically to the LD, i.e., the crack branched parallel to the longitudinal direction of the test bar. As a result, the sample did not break completely, as shown in figure 6(c). The fracture surface for the crack branching planes parallel to the RD was characterized by a quasi-cleavage (figure 7(b)), and that for the planes roughly normal to the RD was characterized by a very fine dimple structure (figure 7(c)). In other words, the fracture surface (delaminating surface) consisted of the delamination structure ($\perp$LD) and the dimple structure ($\parallel$LD). On the other hand, the TFA sample exhibited fully ductile fracture (figure 6(a)), and the test was terminated at $u = 10$ mm; lip shear was observed (figure 6(d)), and the fracture surface consisted of a fine dimple structure (figure 7(d)).

4. Discussion

4.1. Strength–toughness balance

As shown in table 1, although $\sigma_{ys}$ was higher in the TF (1.86 GPa) than in the QT sample (1.51 GPa), the $J$ value of the TF sample also improved markedly to 5184 kJ m$^{-2}$, about 40 times higher than that of the QT sample. In contrast, the $\sigma_{ys}$ of the TFA sample decreased to 0.99 GPa, and, similarly, its $J$ decreased to 2193 kJ m$^{-2}$ by annealing, still being 17 times higher than that of the QT sample. The decrease in strength from the TF to the TFA sample is mainly attributed to the increase in transverse sizes of ferrite grains and carbide particles and to the decrease in dislocation density. The TF and TFA samples had superior strength–toughness balance than the QT sample, and the TF sample with UFEG structures exhibited the best strength–toughness balance among the studied three samples despite its highest yield strength.

Although in the TF sample, the first drop in the load occurred at $P_1 = 10.1$ kN, which was smaller than the $P_{\text{max}} = 12.3$ kN of the QT sample, the steel exhibited a non-catastrophic fracture behavior and yielded a maximum load of 14 kN (figure 6(a)). Finally, the test was terminated
at $u = 10 \text{ mm}$. To clarify the mechanism responsible for the zigzag crack propagations, an interrupted bend test was carried out. The test was stopped at $u = 0.74$ and 1.58 mm. Figure 8(a) presents the corresponding $P$–$u$ curves and compares them with the curves from figure 6(a). It reveals the same features for the three curves of the TF sample. Figures 8(b)–(f) show optical microscopy (OM) images near the initial notch at mid-thickness for the tests interrupted at $u = 0.74$, 1.58, and 10 mm. It can be seen from figure 8(b) that a main crack started to propagate vertical to the LD, from near the initial notch root. Subsequently, the cracks propagated in a zigzag pattern along the longitudinal direction normal to the notch orientation of the test bars (figure 8(c)). At $u = 10 \text{ mm}$, many zigzag cracks branching from the zigzag crack, starting from the notch root were observable in the test bars (figure 8(d)). Furthermore, some microcracks ($\perp \text{LD}$...
Figure 9. Typical load-displacement curve of a laminate composite in three-point bend test.

and $\pm 45^\circ$ LD) were seen ahead of and near the zigzag cracks (figures 8(e) and (f)) [34].

In the case of laminate composites [42–44] having a weak interface normal to the LD (figure 2(a)), the number of delaminations is dependent on the quality of the interfaces between layers, which in turn depends on the thermomechanical processing employed during fabrication. The ideal situation for high toughness is that all the interfaces will delaminate during the applied load. As a general result depicted in figure 9, the bending load drops sharply after it attains the maximum value. Subsequently, a plateau of constant load appears, and the load decreases again. The load drops are associated with crack propagation through the block of layers until the crack is arrested at the interfaces, and the plateau region corresponds to delamination at the interfaces. The extension of delamination is given by the ductility of the next layer, where a new crack is renucleated. This pattern of crack propagation, delamination and crack renucleation, is repeated until the sample is fully fractured. Therefore, for delamination toughening, the materials must have not only a weak interface but also a layer with superior plastic deformation properties. As indicated in figures 6(a) and 8(a), for the studied steel, the load drops during bending are very small, and many such drops are seen until $u = 5$ mm. This behavior is significantly different from that of laminate composites. Originally, the microstructures in the structural metallic materials are composed of complex three-dimensional structures. Lath martensite, the initial microstructure in the present steel, shows a hierarchical microstructure consisting of prior austenite grains, packets, blocks, sub-blocks and laths [45, 46]. On the other hand, the composites bonded by two or more materials have a relatively straight and long weak interface. Furthermore, in the case of composite materials, such as carbon fiber-reinforced plastics, the ductility of high-strength fiber is much lower and its diameter is larger than the transverse grain size of the steel studied in this work. In this steel, although it is difficult to microscopically pinpoint the location of the crack path because of the ultrafine-grained microstructure, as shown in figures 8(g) and (h), the weak site that causes delamination is predicted to be located in an elongated [100] cleavage plane and grain boundaries [8, 19, 28, 29, 31, 47]. Furthermore, a UFEG structure with RD|$<110>$ texture has superior plastic deformation properties. An extended delaminating crack plane was produced with a fine dimple structure, as shown in figure 7(c). In other words, not only a cleavage plane (crack angle $45^\circ$ relative the LD) related to the $\{110\}<110>$ texture but also a fracture plane associated with a plastic deformation was induced, together with a crack normal to the LD during the bending test. Although the incidence of microcracks during the applied load is high due to the presence of many weak planes, the propagation of microcracks is arrested by grain boundaries because the ribbon-like grains elongated by the WCR are three-dimensionally intertwined, like textile. As a result, the microstructural damage is not localized but is distributed over large areas, as in nacre [10, 11] or bio-inspired ceramic composites [22]. Figure 8(a) reveals many load drops in the $P$–$u$ curve, and the load did not decrease with increasing displacement because of two effects: the stress shielding associated with the interference of multiple cracks and the improved plastic deformation associated with grain refinement and texture. As a result, high strength with excellent toughness was achieved in the TF sample.

4.2. Fracture toughness at 77 K

The bcc iron becomes brittle with a decrease in temperature. It has been reported that, in a Charpy impact test of a steel with UFEG structures of 1 $\mu$m size or less, the fracture surface was characterized by a dimple structure even at 77 K and the DBTT was lower than 77 K [3, 35, 47]. However, in the case of UFEG steel produced through heavy deformation, the major toughness mechanism is a change in the triaxial stress state near the crack tip resulting from the presence of separations.
of the crack-divider type [17–19], as shown in figure 2(b). When such a separation occurs during the impact test, the effective thickness of the test sample is reduced, and, the specimen acts as a cluster of thin samples rather than one thick sample. Because of a transition from the plane strain to plane stress in the stress state near the crack tip, the DBTT becomes lower, and dimple fractures easily occur. This fact is clearly confirmed in the vE–temperature curves for non-standard Charpy specimens of various thicknesses [17, 48] and for full-size multiple laminated Charpy specimens containing up to six layers of mild steel [17]. Namely, there are few reports on the low-temperature toughness of UFG steel in the absence of the above-mentioned separation.

We have repeated the static mechanical tests for the studied three samples at 77 K. The $\sigma_{ys}$ of the QT, TF and TFA samples at 77 K increased to 1.80, 2.26, and 1.38 GPa, respectively, compared to the $\sigma_{ys}$ at ambient temperature (see table 1). Figure 10 shows the $P$–$u$ curves and photographs of the TF and TFA samples after the bending test. The TFA sample fractured at $P_{\text{max}} = 4.95$ kN at $u = 0.19$ mm and exhibited a brittle fracture behavior, as did the QT sample that fractured at $P_{\text{max}} = 6.60$ kN and $u = 0.27$ mm, and microdelamination, indicated by arrows in figure 11, was observed on the fracture surface. The microdelamination was divided into two types, as illustrated in the inset, on the basis of the cleavage delamination mechanism [29, 31]. The [100]<110> grain has a cleavage plane normal to the LD, i.e. its plane and grain boundary cause microcracks, as indicated by solid arrows in figure 11. On the other hand, the [110]<110> grain has a cleavage plane parallel to the LD, i.e. its plane and grain boundary cause microcracks, indicated by open arrows. No dimple structure was observed on this fracture surface. That is, if a relaxation of the triaxial stress state according to the separation did not occur under an applied load, the fracture manner at 77 K would be brittle even in a UFG steel with grain size of 1 $\mu$m or less. Therefore, it is difficult to improve the low-temperature toughness of a UFG steel by grain refinement only. On the other hand, the TF sample exhibited fracture behavior with stepwise load increases beyond $P_1$ of the first load drop as well as fracture behavior at ambient temperature, although its $P_1$ value was lower at 77 K than at room temperature (4.78 versus 10.1 kN). The load drops were larger at 77 K than at ambient temperature (figure 6(a)), with an especially sharp drop after a peak at 7.0 kN and $u = 0.78$ mm. This behavior can be explained by the reduced plasticity at low temperatures. Although the steel fractured, extensive delaminating cracks were observed on the sample surface (figure 10(b)). The $J$ value was estimated as 262 kJ m$^{-2}$, which was about 8 times higher than that of the QT (31.8 kJ m$^{-2}$) and TFA samples (31.5 kJ m$^{-2}$), despite the highest strength of 2.26 GPa for the TF sample. Figure 12 shows OM images at mid-thickness of the TF and TFA samples. In the TF sample, extensive delamination starting from the notch root (solid arrow in figure 12(a) and three zigzag delaminating cracks (open arrows) can be seen. On the other hand, in the TFA sample shown in figure 12(b), the quasi-cleavage fracture along the RD was less pronounced, and no stepwise crack propagation was observed. In both samples, macroscopically, the zigzag fracture paths appeared to have an angle of ±45$^\circ$ to the LD and RD, and, finally, the samples fractured into two pieces.

The bcc iron cleaves on the [100] planes, and the coherence length on the [100] plane corresponds to the cleavage crack length [46]. In the UFEG structure, the coherence length on the [100] plane for a crack normal to the LD in the [100]<110> grains is longer than that for a crack oriented at 45$^\circ$ to the LD in the [110]<110> grains or a transverse crack. A brittle fracture occurs when the tensile stress in the process zone at the notch tip exceeds the brittle fracture stress, $\sigma_F$. Consequently, the UFEG structure produces a condition in which the main crack can run along the longitudinal direction rather than the ±45$^\circ$ direction from the LD or the transverse direction. On the other hand, in the UFG structure, when a brittle fracture appears at low temperature, the main crack runs along the ±45$^\circ$ direction from the LD because the $\sigma_F$ normal in its direction is small compared to that in the other directions. The UFEG structure that can achieve high toughness via strengthening by grain refinement can be created through a plastic deformation process such as caliber rolling, extrusion and drawing. The grain shape plays an important role in delamination toughening, as well as grain refinement and texture.
5. Conclusions

0.4C-2Si-1Cr-1Mo steel bars with an ultrafine elongated grain (UFEG) structure and an ultrafine equiaxed grain (UFG) structure were produced by multipass caliber rolling at 773 K and subsequent annealing at 973 K. They were studied for microstructure and static mechanical properties at ambient temperature and 77 K. These properties were compared with those of a quenched and tempered sample, the QT sample, having a martensitic structure. The main results are as follows:

(1) In the TF sample with a UFEG structure, transverse grain size of 310 nm, and yield strength of 1.86 GPa at ambient temperature, the fracture toughness was about 40 times higher than that of the QT sample with a yield strength of 1.51 GPa. This result is attributed to delamination toughening. The TFA sample with a UFG structure, grain size of 660 nm, and yield strength of 0.99 GPa exhibited a fully ductile fracture, and its toughness was about 17 times higher than that of the QT sample, which exhibited a brittle fracture.

(2) At 77 K, the fracture toughness of the TF sample was about 8 times higher than that of the QT and TFA samples, despite the highest strength of 2.26 GPa in the TF sample. The TFA sample, in which the ferrite grains were relatively equiaxed and carbide particles were enlarged by annealing, exhibited a fully ductile fracture, and its toughness was about 17 times higher than that of the QT sample, which exhibited a brittle fracture.

(3) The UFEG structure with α-fiber texture produces a condition in which the crack can run along the longitudinal direction rather than the transverse direction parallel to the LD, because the cleavage fracture stress in the transverse direction is very small due to the elongated grains. The shape of crystal grains plays an important role in delamination toughening, as do their refinement and orientation.

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