J Mater Sci (2006) 41:5417–5423
DOI 10.1007/s10853-006-0266-6

Effect of rolling and epitaxial ferrite on the tensile properties of low alloy steel

E. Ahmad · M. Sarwar · T. Manzoor · N. Hussain

Received: 29 March 2005 / Accepted: 11 October 2005 / Published online: 8 June 2006

Abstract A low alloy steel containing 0.09% C was thermomechanically processed at the intercritical annealing temperature of 790 °C to produce dual-phase microstructure from 50% of austenite. After applying different rolling reductions at this temperature, the specimens were quenched in boiling water to promote the growth of epitaxial ferrite. Warm rolling at 790 °C decreased the hardenability of austenite due to increased in interfacial area of austenite and ferrite. Tensile strength was improved by increasing the rolling reductions both in longitudinal and transverse directions without any significant loss in ductility attributed to the presence of epitaxial ferrite. Microvoid formation in the necked region and their percentage area fraction was measured. The correlation between the area fractions of microvoids formation with strain in the necked region ultimately defined the mode of failure.

Introduction

The microstructure developed from conventional low carbon low alloy steel consisting of martensite and ferrite phases, in which martensite phase imparts high strength and ferrite matrix supplies good elongation that can produce a desirable combination of strength and ductility for application which required good formability.

E. Ahmad (✉) · M. Sarwar · T. Manzoor · N. Hussain
Nuclear Materials Division, PINSTECH, P. O. Nilore, Islamabad, Pakistan
e-mail: dreahmad@yahoo.co.uk

For structural applications of automobile industry a better combination of strength and ductility is the primary requirement in fabrication, to improve the fuel economy. The higher strength of micro-alloyed steels, however, suffers from the disadvantages in that they show lower formability compared with conventional low carbon steel increase in strength/weight ratio [1] which can necessitate a redesign of components and forming equipment. To overcome this disadvantage dual phase steels were developed which combined the conflicting requirement of high strength and improved formability [2–4].

The annealing temperature in the $(\alpha + \gamma)$ phase field controls the volume fraction of austenite content, thus affecting the hardenability of individual austenite pools. At the critical cooling rates, the austenite present fully transforms to martensite resulting in simple mix of ferrite and martensite. On slow cooling rates, the austenite pool first decreases in size by *epitaxial ferrite* growth on retained ferrite and at still slower rates, the remaining austenite enriched in carbon transforms to martensite, bainite and/or pearlite depending upon the alloy content of austenite and cooling rate. The data of constituent phases of microstructure obtained by heat treatment in practice is routinely used in alloy design.

Strength of dual phase steels is related with the amount of plastic deformation applied during thermomechanical processing in the intercritical region due to the formation of substructure in the ferrite [5, 6]. Despite the much published work on structure–property relationship of dual phase steel, adequate attention has not been paid to study the role of epitaxial ferrite grown after themomechanical processing, along with old ferrite and martensite on tensile properties of dual phase steel to find any correlation with strength and ductility. The epitaxial ferrite formed during relatively slow cooling may be free of substructure and ultimately affects the tensile properties, which are aimed to be studied in the present communication. In tensile testing dual phase steels deformed by the formation of micrvoids due to fracture of martensite particles and by decohesion at the martensite–ferrite interface [5, 7–9]. However, the explicit role of microvoid formation, controlling the straining behavior in the necked region is still not clear and needs to be further studied particularly in thermomechanically processed steel where fibrous microstructure is formed.

Experimental work

Chemical composition (wt%) composition of the steel used is shown in Table 1.

The material was provided in the form of hot rolled slabs of 12 mm thickness. Metallographic examination of the as-received microstructure showed it consisted of ferrite and pearlite; and no rolling bands of previous treatment were observed. An initial experimental work was designed to estimate the volume fraction of martensite formed at different intercritical annealing temperatures. For this purpose samples of 10 mm square were intercritically heat treated in the range of 725–830°C in argon atmosphere for 20 min and then quenched in brine solution. Due to very fast cooling rate after intercritical annealing the austenite was almost fully transformed to martensite. A swift semi-automatic point counting machine was used to determine the volume fraction of austenite present. On the basis of the above experimental data, a temperature of 790 °C was selected, where 50% of austenite is formed on heating, although on slow cooling (quenching in boiling water) austenite will partially transform to new or epitaxial ferrite.

For themomechanical processing, steel slabs with thermocouple embedded in the centerline hole to monitor the temperature were heat treated at 790 °C for 20 min. Rolling reductions of 0%, 20%, 30% and 50% were applied at this temperature, although the actual reductions were little different. All the specimens were quenched in boiling water, offering a cooling rate of about 13 °C/s, and the austenite transformed partially to epitaxial ferrite. After rolling tensile specimens in both the longitudinal and transverse directions were machined from each of the slabs. It was also required that the final thickness be 5 mm, for consistency of cooling, after each rolling reduction. Therefore, the specimens were machined initially to different lengths and thicknesses. All the specimens after rolling were quenched in the boiling water.

Table 1 Chemical composition of steel used

Fe	C	Mn	Cr	Si	Ni	Cu	Mo
Bal	0.088	1.2	0.78	0.26	0.15	0.2	0.04

Tensile testing was conducted on an Instron Machine, model 1185, having 100 KN capacity. All tests were performed at the cross-head speed of 0.5 mm/min, in normal atmosphere. An extensometer with 25 mm gauge length was used, and plots of load versus strain were obtained in each case.

For metallographic examination long-transverse sections were cut from the rolled and not-rolled specimens. After grinding and polishing, the specimens were etched according to an etching technique developed by Lawson et al. [10], to differentiate the old and new ferrite. It consisted of etching in 5% picral followed by etching in 2% natal solution and immersion in a boiling alkaline chromate solution (8 g CrO_3 + 40 g NaOH + 72 mL H_2O). A swift point counting technique was used to for quantitative phase measurements. The total number of points was 1000–1500, with a standard deviation of about 0.014 from the point counting variables. Image analyzer was used for the measurement of area fraction of microvoids using un-etched specimens. Scanning Electron Microscope was used to study microvoids and fracture surfaces.

Results and discussion

Thermomechanical processing and hardenability of austenite

The microstructures developed after intercritical heat treatment at 790 °C, followed by quenching in boiling water are shown in Figs. 1 and 2 for not-rolled and rolled specimens, respectively.

This slow cooling rate allowed the epitaxial ferrite to grow on the existing ferrite and along the martensite particles. The results of quantitative metallographic results are shown in Table 2. The amount of epitaxial ferrite formed

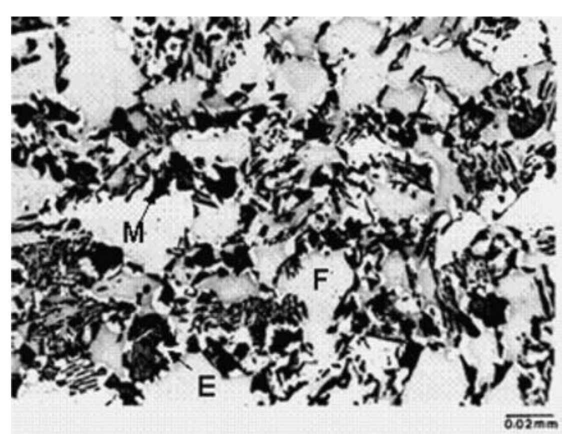

Fig. 1 Microstructure of dual-phase steel after intercritical annealing at 790 °C and the quenched into hot water shows M = Martensite (black), F = Retained ferrite (grey), E = Epitaxial ferrite (white)

Fig. 2 Microstructure of dual-phase steel after intercritical annealing at 790 °C followed by 48% rolling reduction and quenching in boiling water

in rolled specimens is more than in not rolled conditions which reflects that rolling has decreased the hardenability. The carbon content of martensite were estimated from the approximate carbon content of the ferrite (0.01%), the carbon content of the steel (0.088%) and volume fraction of the martensite, using the following equation [11]

$$C_m = C_o + \rho_\alpha/\rho_m(100/P_m - 1)(C_o - C_\alpha)$$

where C_m, C_o and C_α are the carbon contents of martensite, steel and ferrite respectively. P_m is volume percent of martensite. The value of ρ_α/ρ_m is taken as unity.

When steels are quenched from the fully austenite phase field, two most important parameters which influence the hardenability of the austenite are the grain size and composition of the austenite. Hardenability increases with increasing the grain size, due to the decrease in grain boundary area and the consequent reduction in the density of sites and in the rate of nucleation of ferrite or pearlite. Most alloying elements play an important role in slowing down the ferrite and pearlite reaction. In case of dual-phase steel, the nucleation of new ferrite is not essential, since ferrite existing at the intercritical annealing temperature (ICAT) can grow epitaxially into austenite. Also, the austenite volume fraction and its composition depend upon the intercritical temperature. It seems sensible and basic, therefore, to define martensitic hardenability of dual-phase steel in term of the fraction of the austenite present at the intercritical temperature which transforms to martensite. Apparent anomalies may then be understood, such as when a smaller amount of martensite is obtained by quenching from a low ICAT than quenching at the same rate from a higher temperature, although the austenite present at the lower temperature contains more carbon and would be expected to be more hardenable.

Table 2 Thermomechanical processing data of various rolling reductions and quenched in boiling water (annealing temperature = 790 °C)

Rolling reduction (%)	Martensite (%)	Carbon content of martensite (%)	Epitaxial ferrite (%)
0	32	0.25	18.9
21	25.9	0.31	27.2
29	23.4	0.34	24.1
48	30.5	0.27	24.4

Priestner and Ajmal [12] determined the effects of various rolling reductions on the austenite's hardenability, for the steel having composition of 0.11% C, 1.48% Mn and 0.34%Si. They observed that rolling progressively increased the hardenability of austenite. They explained this result by suggesting that grain rotation occurred during rolling reduction, so changing the orientation relationship between the two phases (ferrite + austenite). This, it was proposed, would inhibit the epitaxial growth of ferrite back into austenite.

The present hardenability results can be more simply explained by a model of Priestner [13] which described that the epitaxial ferrite is formed by re-growth of existing (old) ferrite and volume that forms is the product of interfacial area, average growth rate and time taken to cool to the M_s temperature. Warm rolling of ferrite and austenite mixture in the two phase field would elongate ferrite and austenite grains, thus increasing the interfacial area. This would increase the amount of ferrite formed and decrease the hardenabity of the austenite. It is suggested, therefore, the results obtained in the present study are the expected effect of warm rolling on the hardenabilty of austenite, and that the results of Priestner and Ajmal are anomalous.

Tensile properties

Tensile data for material intercritically annealed at 790 °C for the rolling reductions of 0%, 21%, 29% and 48% is presented in Table 3.

The specimens are coded according to planned rolling reduction and direction of tensile testing. For example '20L' shows 20% rolling reduction tested in the longitudinal direction.

Plastic deformation during warm rolling in the two-phase region results in the development of substructure in the ferrite [5, 6]. Tanaka [14] suggested that even heavy deformation in the fully austenite region produced only grain refinement of the resultant ferritic structure after cooling, whereas deformation of a mixture of ferrite and austenite produced a substructure in the ferrite which contributed to the strength of the material. The epitaxial ferrite is grown after rolling reduction during quenching in the boiling water; it might be free from deformation effects of dislocation substructure development, as shown in Fig. 3.

Table 3 Tensile data of various rolling reductions and quenched in boiling water (annealed temperature = 790 °C)

Specimen code	Martensite %	Rolling reduction obtained %	Max. true stress (MPa)	True uniform strain (%)	Total elongation (%)	UTS (MPa)	Uniform elongation (%)	0.2% proof stress (Mpa)
0L	32	0	762.3	11.56	21.6	679.1	12.3	333
0T		0	702.1	10.64	–	631.29	11.2	363
20L	25.9	21	773.2	11.43	19.6	689.6	12.1	407
20T		21	761.4	11.65	18.8	671.7	12.4	305
30L	23.4	29	817.0	12.45	22.8	721.38	13.3	438
30T		29	775.1	12.17	17.4	686.3	13.0	357
50L	30.5	48	835.1	11.3	18.8	745.8	12.0	414
50T		48	766.2	8.4	17.2	704.65	8.8	377
0.06 C steel [11]	21.7					595	10.0	
	24.8					624	10.5	
	28.1					630	14.0	
0.17 C steel [5]	29.4	–				872	5.3	
0.2 C steel [9]	25	–				798	2.1	

Warm rolling in the intercritical region progressively increased the true stress and UTS both in the longitudinal and transverse directions with little effect on the ductilities. These observations suggest that improvement of strength with rolling reduction was caused by three principal changes in the microstructure.

(a) The formation of fibrous martensite increased the area of contact with the matrix (ferrite) compared to equiaxed structure of unrolled martensite. Therefore, stress transfer from the matrix to the fiber might be more efficient.
(b) Deformation may have produced subgrains in the old ferrite which have increased the strength of the matrix.
(c) The increase in strength with rolling reductions should be at the expense of ductility but this effect is nullified here due to the presence of epitaxial ferrite except in 50T specimen and ductility remains almost constant.

The tensile data comparison of 0.06 C steel of Speich and Miller [11] shows that with the almost same carbon content of steel and volume fraction of martensite the tensile strength is lower from that of present results. Similarly with higher carbon content of steels and martensite of Sarwar and Priestner [5] and Erdogan and Tekeli [9] have reported the increase in strength with sufficient decrease in ductility. Although Sarwar and Priestner used the rolled specimens of 50% reduction and substructure was formed in the ferrite but without the presence of soft epitaxial ferrite, the uniform elongation was affected.

Effect of rolling on the distribution of strain and microvoid density

Figures 4 and 5 show the true area strain plotted against the distance from the fracture surface, for the longitudinal and transverse specimens respectively.

These curves indicate the different samples had different necking characteristics, with different stain gradients in the neck and varying true fracture strain at fracture.

The strain distribution in the region of the neck may be correlated with nucleation and growth of microvoids. Stenbrunner et al. [15] found that in dual-phase steels during localized necking the microvoids were concentrated near the fracture surface and their density decreased rapidly

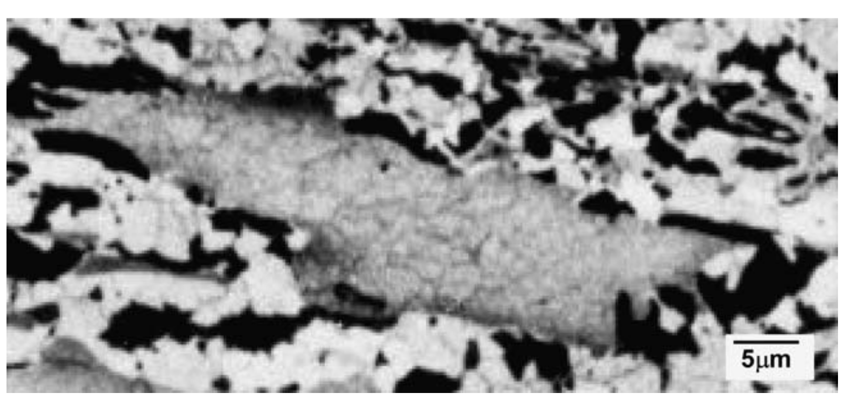

Fig. 3 The sub-structure formed in old ferrite during thermomechanical processing in the intercritical annealing

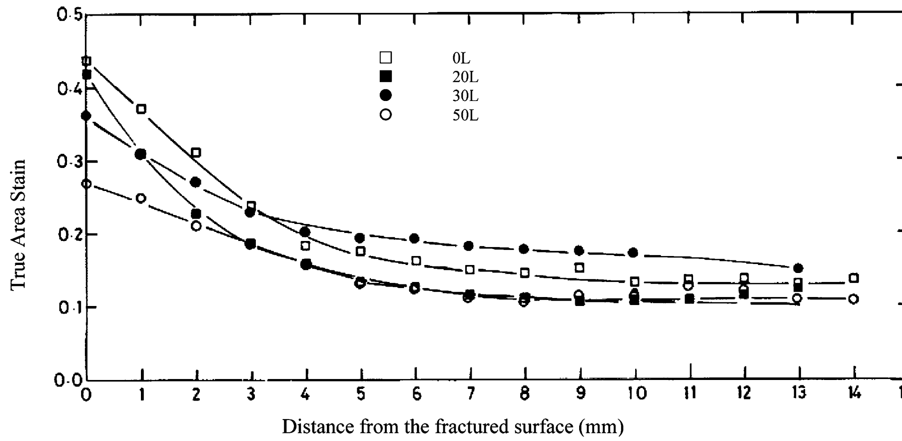

Fig. 4 True area strain at different positions for 790 °C annealed, different rolling reductions and boiling water quenched longitudinal specimens

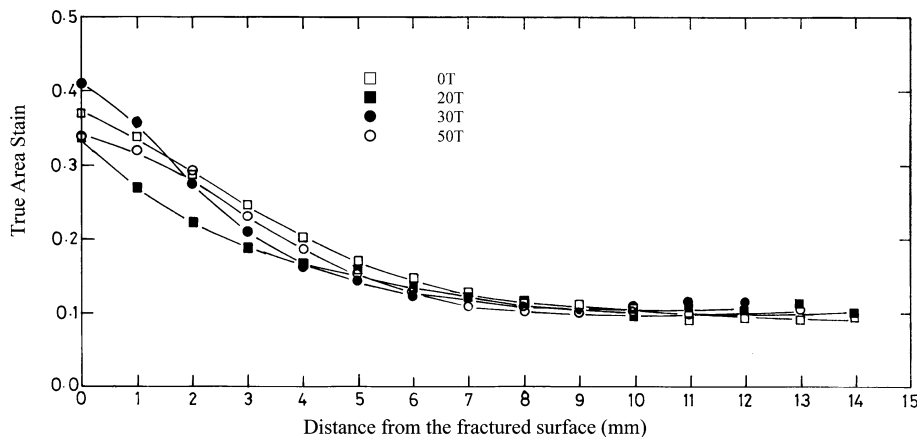

Fig. 5 True area strain at different positions for 790 °C annealed, different rolling reductions and boiling water quenched transverse specimens

with increasing distance from the fracture surface. In diffuse necking they found a more uniform distribution of microvoids in the necked region. Figure 6a and b shows microvoids near fracture surface in specimen 30L.

The voids formed by fracture of martensite as well as by decohesion at the matrix/particle interface. These phenomenons are clear in Fig. 6b. The greater incidence of decohesion at the interface may be due to the presence of softer epitaxial ferrite, although void formation by fracture of individual martensite particles in general and occasionally decohesion at the interface has been studied previously [5, 8, 9]. Similarly, Fig. 7 shows the formation of voids near the fracture surface in specimen 50L.

They also formed mostly by decohesion at the interface, with some examples of fracture of martensite.

In the present experiments the percentage area fraction of microvoids was measured as function of distance from the fracture surface along the centerline. Polished, unetched specimens were used to get a good contrast on the image manager; subsequent etching for microphotography did not appear to alter the microvoids [15].

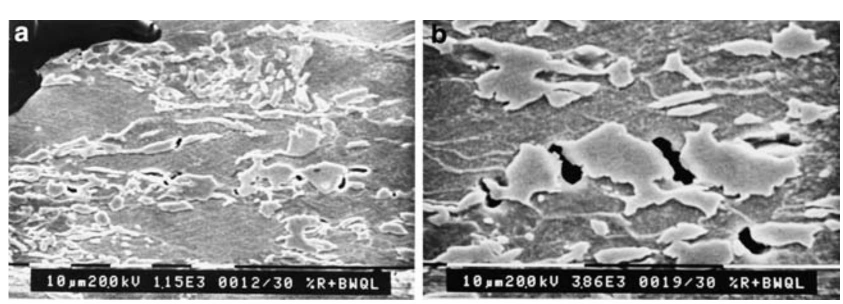

Fig. 6 The formation of Microvoids for 30L specimen (**a**) Near the fracture surface and (**b**) away from the fracture surface

Fig. 7 The formation of Microvoids near the fracture surface for 50L

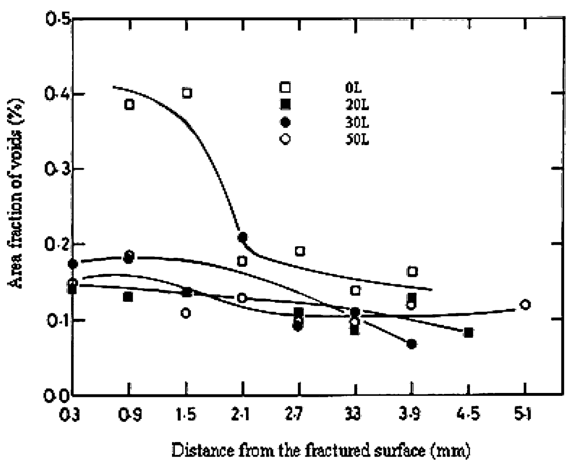

Fig. 8 Area fraction of voids along the center-line in the necked region for 790 (C annealing, different rolling reduction and boiling water quenched longitudinal specimens

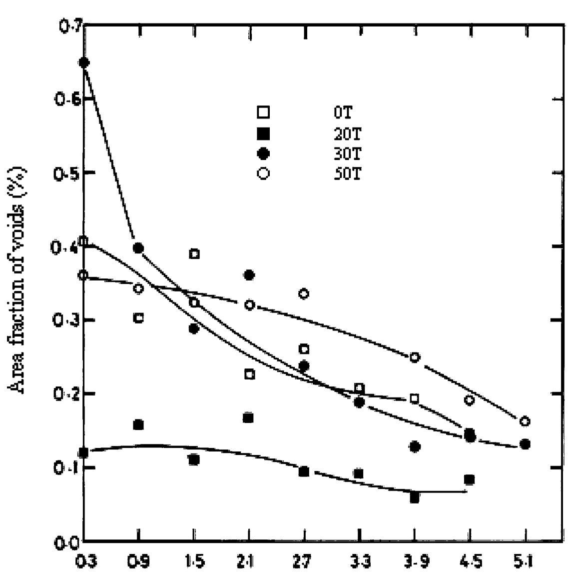

Fig. 9 Area fraction of voids along the center-line in the necked region for 790 °C annealing, different rolling reduction and boiling water quenched transverse specimens

Figures 8 and 9 show the quantitative measurement of microvoid area fraction as a function of distance from the fracture surface, for longitudinal and transverse tests.

In the longitudinal direction 0L and in the transverse direction 30T had the highest microvoid densities near the fracture surface. Very few or no voids were found in the regions of uniform deformation outside the neck. A high density of microvoids near the fracture and a rapid decrease of microvoid density with distance from the fracture surface was generally associated with high true strain. This is illustrated in Fig. 10, where the microvoid area fraction is plotted versus true strain in the neck.

It is seen that in both longitudinal and transverse tensile tests microvoids began to appear at ''threshold'' strain, and their area fraction increased, at a rising rate which define the nature of fracture: if it is assumed that a critical area fraction of microvoidage must be achieved before the fracture mechanism, can be completed catastrophically, then ductility is controlled by:

(a) The thresholds strain at which microvoids first appear.
(b) The rate of increase of voidage with respect to strain.

Figure 9 shows that up to 20% of rolling reduction the true stress for fracture was almost same both in longitudinal and transverse directions of rolling. However, at higher rolling reduction of 30% the true strain for fracture was less in the transverse than in longitudinal direction but it increased rapidly with higher area fractions of microvoids. Similar increase in area fraction of microvoids was

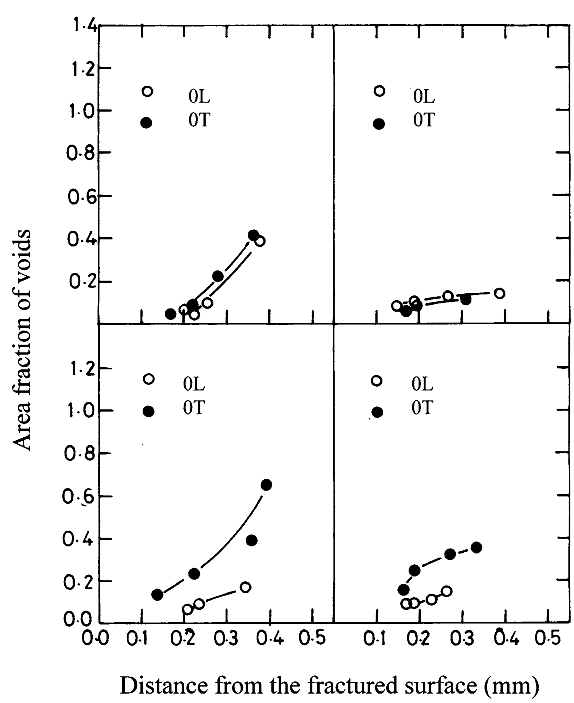

Fig. 10 True area strain as a function of area fraction of voids, for different rollingreductions at 790 °C

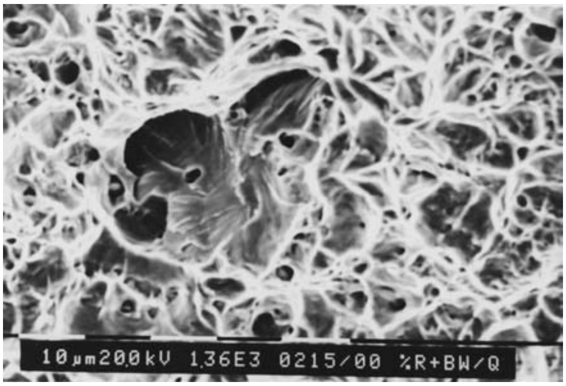

Fig. 11 Fracture surface mainly with ductile dimples of 0L

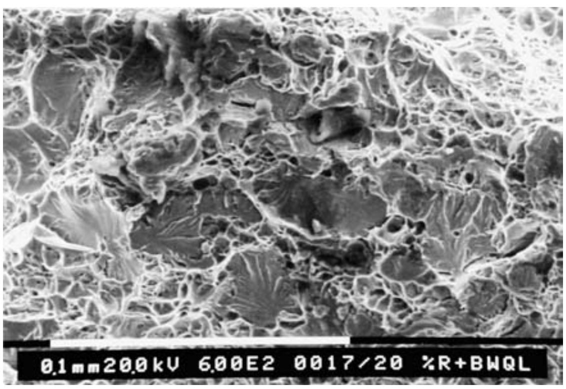

Fig. 12 Fracture surface with cleavage facets and ductile dimples of 20L

also observed in 0L and 0T, resulting in a fracture surface with mainly ductile dimples, shown in Fig. 11.

In 20L where an increase in area fracture of microvoids with strain for fracture was not observed a fracture surface appeared mainly with cleavage facets, Fig. 12

Reexamining Fig. 10, the role of epitaxial ferrite is not clear for controlling the threshold strain but in general it looks to inhibit the void formation in warm rolling.

Conclusions

Hardenability of austenite decreases by warm rolling due to the fact that increases in interfacial area of the austenite and ferrite phases.

The epitaxial ferrite which grows after the rolling reduction during slow cooling is relatively soft than the old ferrite containing substructure. The increase in strength of the material by rolling without any appreciable loss in ductility is attributed to the combined effect of two types of ferrites. The higher strengths in the longitudinal than in the transverse directions at all rolling reductions may be due to the fibrousity of the microstructure. In the longitudinal directions the martensite fibers are oriented along the tensile axis results in better stress transfer from soft ferrite matrix to the hard martensite particles.

The microvoids in the necked region of broken tensile specimens were formed by fracture of martensite particles or by decohesion at the interface. The percentage area fraction of the microvoids is related to the straining in the necked region. The threshold strain at which microvoids first appear and rate at which their area fraction increases define the ultimate failure mode. Although it was observed that epitaxial ferrite slow down the process of void formation but generally it appears that where void formation was low, cleavage becomes prevalent.

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