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Structure and Properties of Thermomechanically-processed HSLA Steels for Naval Applications

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ABSTRACT

Four high-strength low-alloy (HSLA) steels with varying chemical compositions were forged in two different temperature ranges followed by cooling in various media. Microstructures and mechanical properties of the steels were evaluated. The microstructures obtained in water–quenched low-carbon HSLA steels were lath martensite packet within the pancaked grains. On air or sand cooling predominantly bainitic ferrite or granular bainite structure forms. The strength properties of these steels decreased with decrease in cooling rate and is accompanied by an increase in elongation and impact toughness values. The ductile-to-brittle transition temperature of HSLA-100 grade steel was found to be – 40 °C. The impact fracture surface of air cooled HSLA-100 steel showed ductile failure with formation of dimples at 20 °C and at – 20 °C. The fracture mode changed to brittle failure with formation of cleavage and river pattern at – 40 °C and at – 60 °C. The microstructures of the ultra-low carbon HSLA steel show lath ferrite or granular ferrite in water-quenched condition. With slower cooling rate, the volume fraction of lath ferrite decreased with an increase in formation of polygonal ferrite. The maximum strength value obtained in air-cooled condition is achieved due to precipitation of fine microalloying carbides and carbonitrides. Slower cooling rate increases the volume fraction of polygonal ferrite which increases the toughness value.

Keywords: HSLA steel, bainitic ferrite, polygonal ferrite, toughness, naval applications, microstructure

1. INTRODUCTION

Copper-bearing high-strength low-alloy (HSLA) steels have generated interest for offshore, onshore, and naval applications. In the series of *Cu*-bearing HSLA steels, ASTM A710 steel was the first developed by the International Nickel Company in late 1970[1]. The A710 alloy steel is a very low– carbon, copper precipitation–strengthened steel, which provides high strength along with good low-temperature fracture toughness and weldability for Arctic pipeline applications. In recent years, several countries have developed *Cu*-bearing age-hardening steels, based on modified ASTM A710 chemistries, because these offer a good combination of properties such as weldability and toughness making these suitable for structural use. One variety of these has been designed for replacement of HY-80 for US Navy surface ships, and is known as HSLA-80.

An extensive evaluation of HSLA-80 properties, welding, and structural performance demonstrated that the very low carbon, copper precipitation-strengthened steel met the requirements of HY-80 steel, and was readily weldable with no preheat using same weldable consumables and process as

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for HY-80 steel[2]. HSLA-80 steel with 500 MPa minimum yield strength has been developed using thermomechanically-controlled processing (TMCP) techniques followed by age hardening[2-4]. This steel has a leaner chemistry than the conventional A-710 steels. The *Cr* and *Mo* additions in A-710 steels are, replaced by a higher *Mn* content. A small *Ti* addition is used to enhance heat affected zone (HAZ) toughness. The TMCP practice employed involves high temperature recrystallisation-controlled rolling, low-temperature non-recrystallisation-controlled rolling above Ar_3, and controlled cooling to 550 °C. The latter process prevents premature ε-*Cu* precipitation, and ensures a maximum and consistent precipitation-hardening response on ageing. Following the HSLA-80 programme, an alloy development and qualification programme commenced by US Navy in mid 1980 which resulted in HSLA-100 steel as a replacement for HY-100 steel[1].

HSLA-100 is also a very low carbon, copper precipitation-strengthened steel, meeting the strength and toughness of HY-100 steel. Also, weldable without the preheat requirements of HY-100, using the same welding consumables and processes as used in welding HY-100. At present, tonnage of HSLA-80 and HSLA-100 grade steels is being produced on commercial scale. While most of these steels are being produced in conventional rolling route another alternative route i.e., forging came into lime light[5]. In 1992, the collaborative INDO-US programme was initiated to develop and characterise the naval ship-building steels. As an extension of this work, the authors conducted a naval steel development programme in collaboration with NMRL, India, and later on, with the Department of Science and Technology, Govt of India.

The present paper describes some of the studies conducted in the above projects and this was done in three phases. In the first phase, an HSLA-100 grade steel which was supplied by US Navy, was thermomechanically processed via forging route followed by various post-cooling techniques. In the next phase, two HSLA steels without *Cu* were prepared, and in the last phase, an ultra-low carbon HSLA steel (*Ni-Cu-Mo-B* bearing) steel with leaner chemistry was prepared. All these steels were subjected to the same TMCP schedule and the structure and properties of these steels were characterised.

2. EXPERIMENTAL PROCEDURE

In this investigation, four HSLA steels have been studied and the chemical compositions as analysed in Quantovac and LECO are shown in Table 1. Steel 1 was supplied by US Navy as 50 mm thick slab. The remaining three steels were melted in an air-induction furnace. Steel 2 and Steel 3 had higher manganese content wrt HSLA-100 grade steel and these steels are made without *Cu*. Whereas in Steel 4, the carbon content was lowered to an extremely low value (0.01 Wt. pct) and nickel copper content was lowered with comparison to HSLA-100 grade steel. This steel is microalloyed with *Nb*, *Ti*, and also with *B*. The steels of 50 mm x 50 mm cross section were reheated at 1200 °C for 2 h. Forging was carried out in one ton capacity down stroke power hammer in two stages. In stage-I, 50 per cent deformation was given by six repeated strokes in the temperature range of 1100-1050 °C to reduce the cross section of the slab to 35 mm x 35 mm. In stage-II another 50 per cent deformation was applied by similar six repeated strokes in the temperature range of 850-800 °C to obtain the final

Table 1. Chemical composition (Wt. pct.) of the steels

Type	Chemical composition											
	C	*Mn*	*P*	*S*	*Cu*	*Ni*	*Si*	*Cr*	*Mo*	*Ti*	*Al*	*Nb*	*B* (ppm)
Steel 1	0.04	0.86	0.004	0.002	1.5	3.55	0.27	.57	0.6	--	0.032	.03	-
Steel 2	0.04	1.7	0.008	0.009	--	1.71	0.23	.04	0.7	.03	0.03	.05	-
Steel 3	0.04	1.8	.01	0.01	--	3.52	0.28	.03	0.66	.02	0.03	.05	-
Steel 4	0.01	1.28	0.009	0.009	1.25	1.98	0.21	0.01	0.62	0.11	0.02	0.06	14

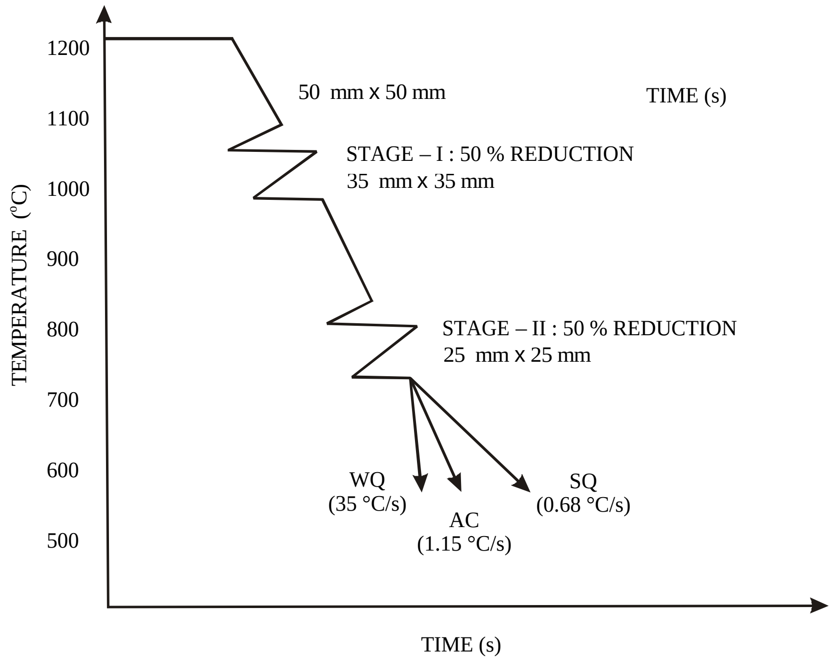


Figure 1. Schematic diagram of controlled forging operation.

cross section of 25 mm x 25 mm. Subsequently the forged slabs were subjected to water, air or sand cooling. Temperatures at different stages and during cooling were measured by inserting thermocouples into the forged bars. The schematic diagram of the forging operation is shown in Fig. 1. Optical, scanning electron microscopy (SEM) and transmission electron microscopy (TEM) were used for microstructural characterisation.

Optical metallographic samples prepared by conventional grinding and polishing techniques were etched with 2 per cent nital solution and observed in a light microscope. Thin foils for transmission electron microscopy were prepared by twin jet polishing in an electrolyte of 90 per cent acetic acid and 10 per cent perchloric acid. Thin samples were observed in TEM, Philips CM 200 with EDAX at 200 kV operating voltage. Tensile testing was carried out in an Instron tensile testing machine Model No. 4204 keeping a constant cross head speed of 8.3 x 10^{-3} mm/s. Standard impact specimens (ASTM; Vol 03.01:E23-96) were prepared and Charpy impact testing was carried out at ambient and at subambient temperatures. Fracture surfaces of a few impact specimens were examined in a SEM (Model No. JEOL JSM 840A).

3. RESULTS AND DISCUSSION

3.1 Characterisation of HSLA-100 Steel supplied by US Navy (Steel 1)

3.1.1 Microstructure

The microstructures of the steel at different post-cooling conditions have been studied and shown in Figs 2 and 3. Pancaked grains are obtained for all post-cooling conditions which have been generated due to deformation at non-recrystallisation region of austenite. However, different transformed structures of austenite have been obtained due to different post-cooling rates from the same finish forging temperature. At higher post-cooling rate (water-quenched steel) lath martensite packet is obtained in the pancaked grains [Fig. 2(a)]. TEM study shows thin retained austenite film at the interlath regions, [Fig. 2(b)]. Similar observation has been reported by earlier researchers in HSLA-80 and HSLA-100 grade steels[6,7]. With decrease in cooling rate (air cooling) the lath martensite structure changed to predominantly bainitic ferrite or granular bainite structure, [Figs 3(a) and (b)]. TEM study shows lath ferrite and chunky-type second-phase particles with dark contrast at the lath boundaries [Fig. 3(c)]. These second-phase particles in low-carbon

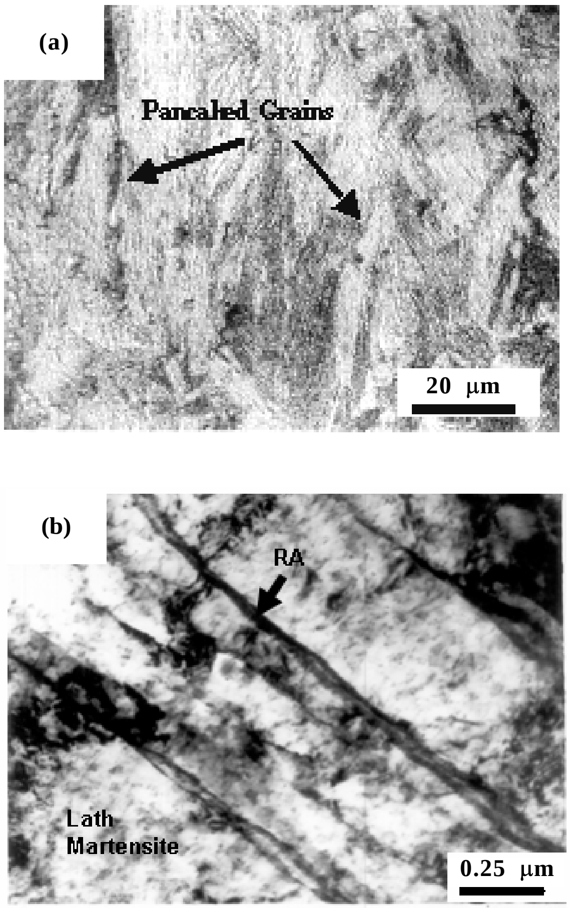


Figure 2. Microstructure of steel 1 in water-quenched condition: (a) optical and (b) TEM.

HSLA steel have been identified as MA constituents or retained austenite islands in TEM by the previous workers[6-12]. In the sand-cooled steel, microstructures were predominantly polygonal ferrite and the dispersed second-phase particles (Fig. 4).

3.1.2 Mechanical Properties

The tensile property of alloy 1 shows a decrease in strength and increase in ductility with decrease in cooling rates (Fig. 5). The strength values obtained in this steel are in the range of 996 MPa–842 MPa UTS and 942 MPa–672 MPa YS along with 16-22 per cent elongation. It is noteworthy that the YS/UTS ratio is very high (~ 0.95) in water-quenched condition and this is due to the dislocated lath martensite structure. In air-cooled or sand-cooled condition this ratio is lowered down to 0.81–0.80. This is due to the bainitic structure of the steel at slower cooling rates.

The impact toughness values obtained at different cooling rates have a similar trend with per cent elongation, (Fig. 6). However, a sharp drop in toughness is observed at –40 °C for water-quenched steel.

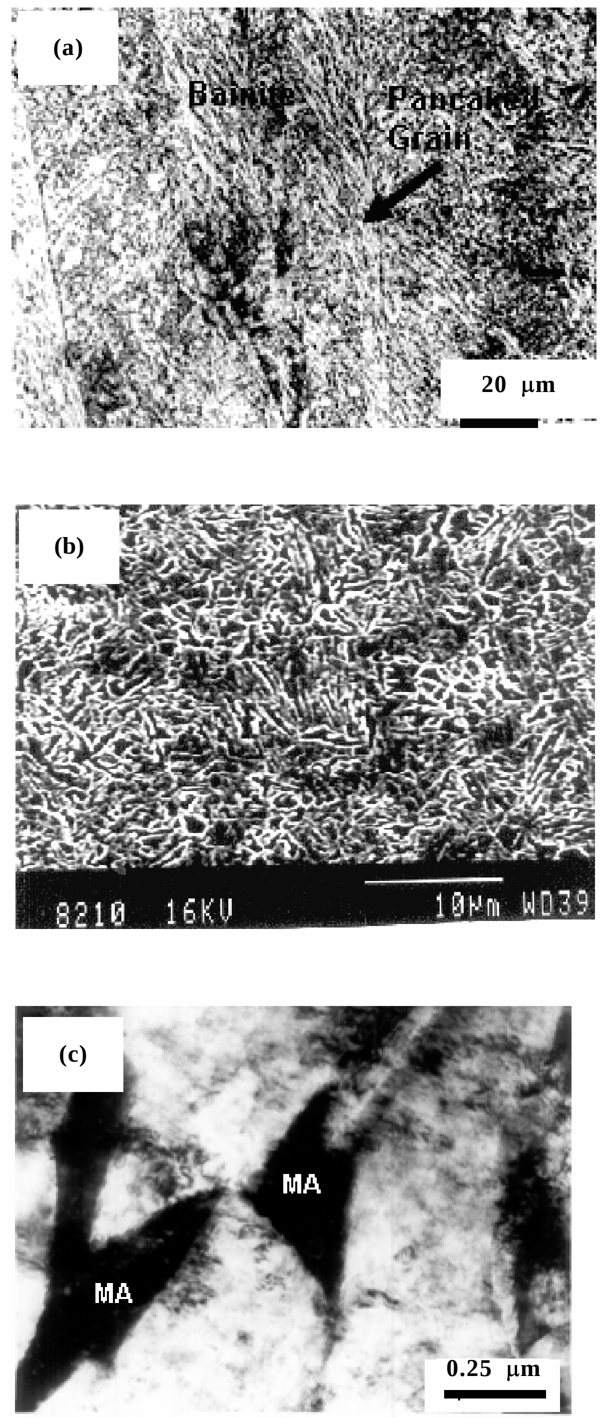


Figure 3. Microstructure of Steel 1 in air-cooled condition: (a) optical, (b) SEM, and (c)TEM.

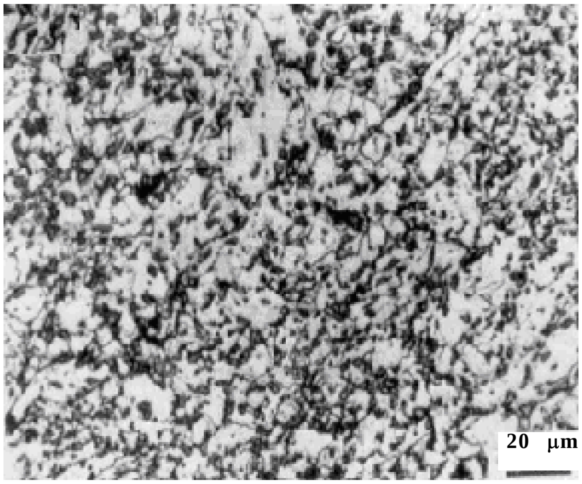


Figure 4. Optical microstructure of Steel 1 in sand-cooled condition.

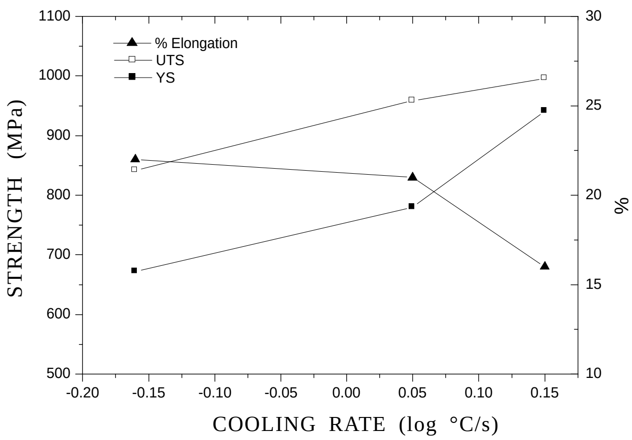


Figure 5. Tensile properties of Steel 1.

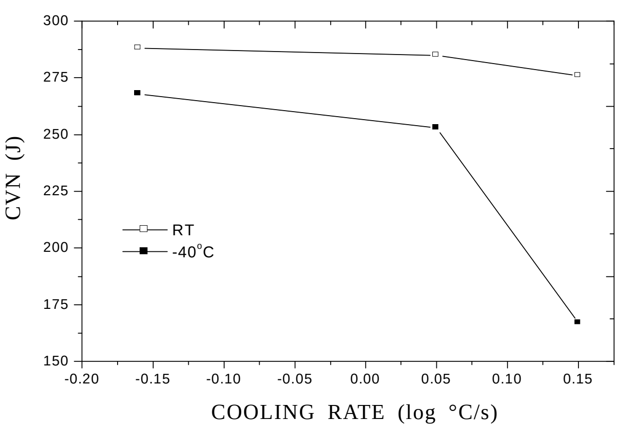


Figure 6. Impact toughness values of Steel 1.

Toughness values are slightly higher in sand- cooled condition than those of air-cooled steel at both the test temperatures.

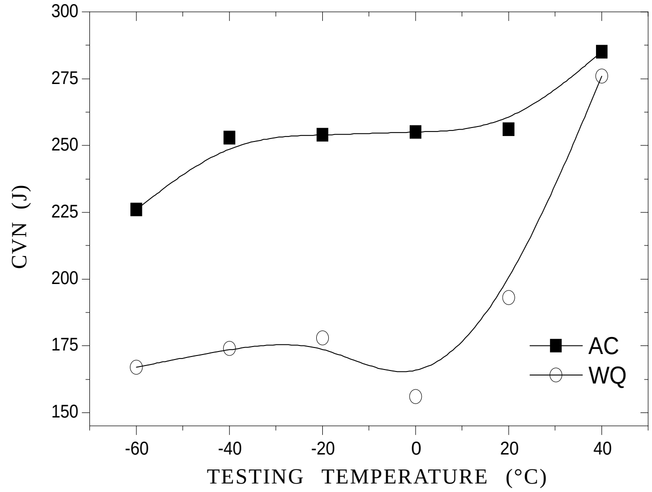


Figure 7. Ductile-to-brittle transition temperature study of Steel 1.

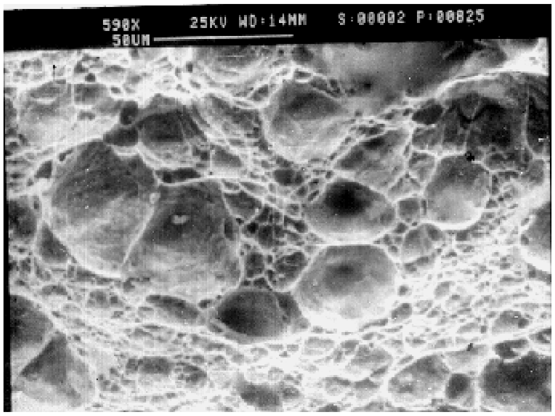
The ductile-to-brittle transition temperature (DBTT) of this steel has been studied at different post-cooling conditions from the CVN value versus testing temperature curve[13] and shown in Fig. 7. It is observed that in water-quenched or air-cooled condition, the impact toughness value drops in the temperature range 40 °C to 20 °C and this drop is comparatively slower in air-cooled steel. After that, a platue in impact toughness values is obtained in the temperature range of 20 °C to – 40 °C. However for air-cooled steel, this platue is obtained at higher toughness values (256-253 J) than in water-quenched steel (193-174 J). A fall in CVN value is observed from – 40 °C to – 60 °C.

3.1.3 Fractography

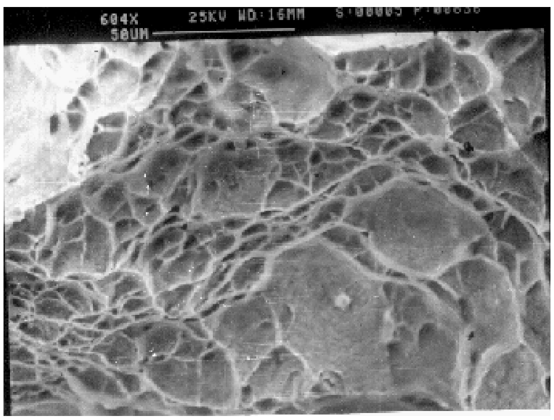
The impact fracture surface of air-cooled steel at different testing temperatures have been studied and shown in Fig. 8. The impact fracture surfaces show ductile failure with formation of dimples at 20 °C [Fig. 8(a)] and at –20 °C [Fig. 8(b)]. The fracture mode changed completely to brittle failure with the formation of cleavage and river pattern at –60 °C [Fig. 8(d)].

3.2 HSLA Steels Without Copper (Steel 2 and Steel 3)

Figure 9 shows the optical micrographs of Steel 2 and Steel 3 at different post-cooling conditions. At higher cooling rate lath martensite packet is



(a)



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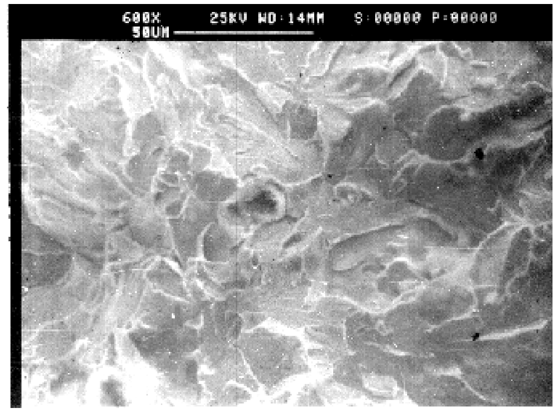
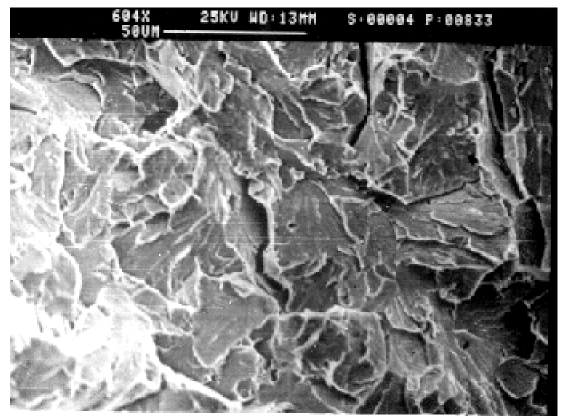


(c)



(d)

Figure 8. Fractography of the impact specimens (Steel 1) tested at various testing temperatures: (a) 20 °C, (b) –20 °C, (c) – 40 °C, and (d) – 60 °C.

obtained [Figs 9(a) and 9(d)]. With decrease in cooling rate, the lath martensite changed to predominantly bainitic ferrite or granular bainite structure [Figs 9(b) and 9(e)]. When the cooling rate is further slowed, polygonal ferrite or quasi-polygonal ferrite formed in microstructure with random distribution of second-phase particles, [Fig. 9(c) and 9(f)]. TEM of the water-quenched specimen of Steel 2 shows lath martensite packet (Fig. 10).

The UTS values of the alloys have been plotted against cooling rate [Fig. 11(a)]. For the present compositions, a high strength value (~1200 MPa UTS) has been obtained at a faster cooling rate. The main contribution of such a high strength in water-quenched steel is the solution strengthening effect of mainly *Mn*. A higher *Mn* content of these steels than HSLA-100 grade steel has enhanced the hardenability of austenite, which might have lowered the transformation temperature of austenite. Faster cooling (water-quenching) from 800 °C finish forging temperature, has produced predominantly fine lath martensite structure. This low temperature transformation product of austenite contains high dislocation density in the structure as observed from TEM micrograph (Fig. 10) and the formation of such low-temperature product, i.e., martensite, is favoured by shear mechanism. Hence, the invariant plane strain (IPS) of transformation generates high dislocation density in the structure.

A sharp fall in UTS value has been observed from water-quenched to air-cooled steel and then remained mostly unchanged up to sand cooling. In slower cooling, the strength value sharply drops (20–22 %) due to a predominant charge in

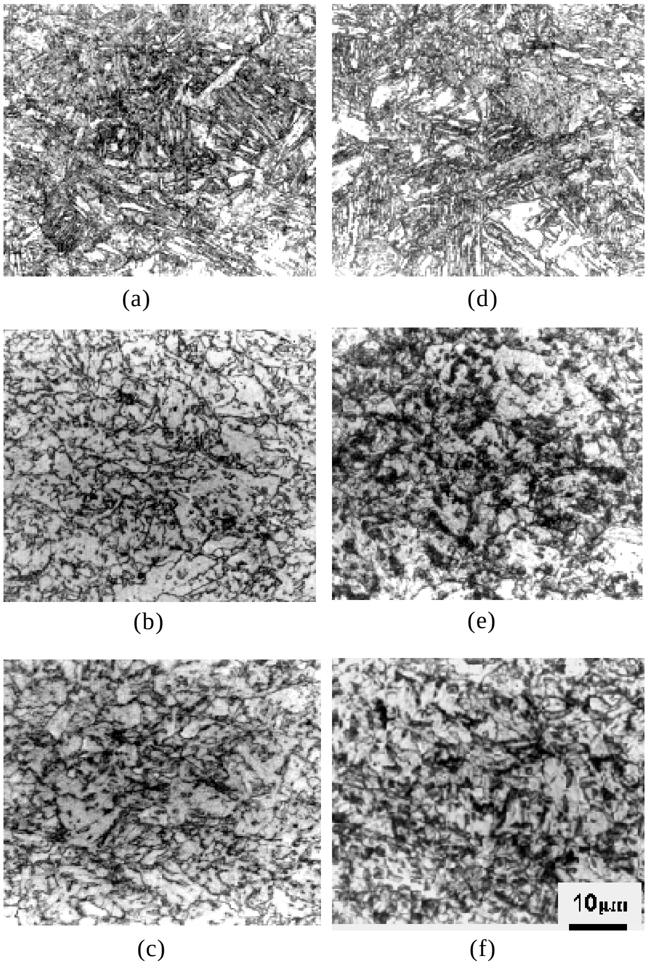


Figure 9. Optical microstructures of Steel 2: (a), (b), (c) and Steel 3: (d), (e), (f) at different post-cooling conditions. (a) and (d): WQ (35 °C/s); (b) and (e): AC (1.15 °C/s); and (c) and (f) SC: (0.68 °C/s).

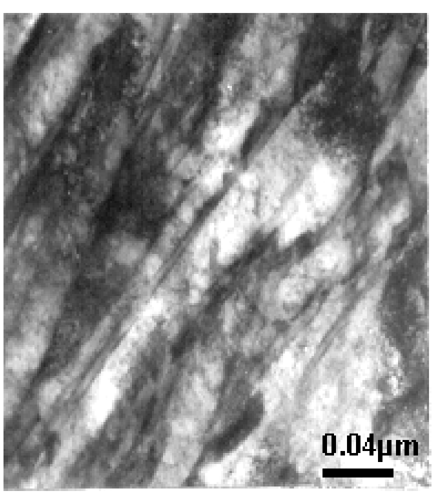


Figure 10. TEM shows lath martensite in WQ (35 °C/s), specimen of Steel 2.

microstructure. The continuous cooling diagram of HSLA-80 and HSLA-100 grade steels show that under a wide range of cooling rates, granular ferrite (bainite) is obtained with highly dislocated microstructure[6]. In the present study, the microstructure observed for air-cooled and sand-cooled steels is in good agreement with the previous observations[6,10]. As bainite forms by a mixed mode of transformation mechanism, i.e., a combination of shear and diffusion, the strength due to the solid solution strengthening of interstitial carbon atom is lost in air-cooled and sand-cooled steel. It is noteworthy that the change in strength value in air-cooled and sand-cooled low-carbon HSLA steels is very negligible. The microstructural study is also in good agreement with the above observation, though in sand-cooled steel, the volume fraction of polygonal ferrite is

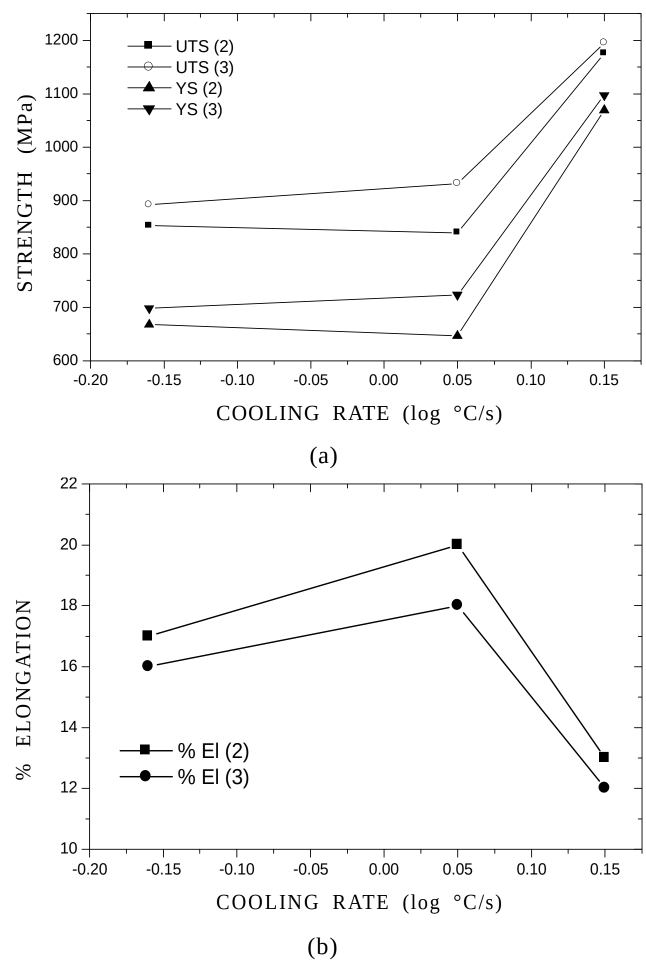


Figure 11. Tensile properties of Steel 2 and Steel 3: (a) UTS and YS, (b) percentage elongation.

slightly higher than that of air-cooled steel. It is obvious because slowest cooling (in the present study) would allow more time for high temperature transformation as well as faster diffusion rate of interstitial carbon at elevated temperature, and thereby, favour polygonal ferrite formation.

The per cent elongation has lowered in water-quenched steels, in comparison to air-and sand-quenched steels [Fig. 11 (b)].

The impact toughness value at room temperature for different cooling rates has been plotted for the two alloys and shown in Fig. 12. The toughness value increases for slower cooling rates.

3.3 ULC HSLA Steel Containing Boron (Steel 4)

Figures 13(a) and 13(b) show the optical microstructures of the ultra-low carbon steel at

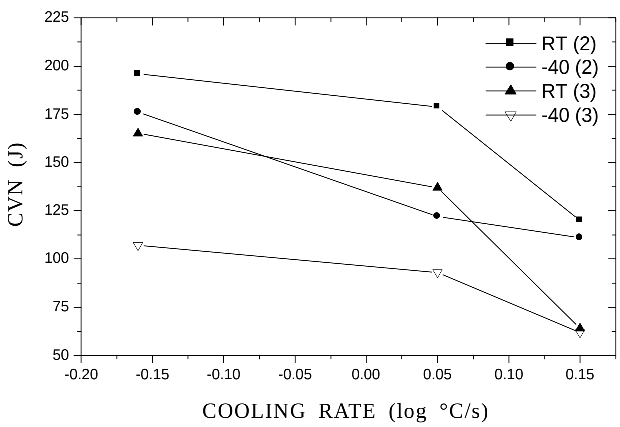
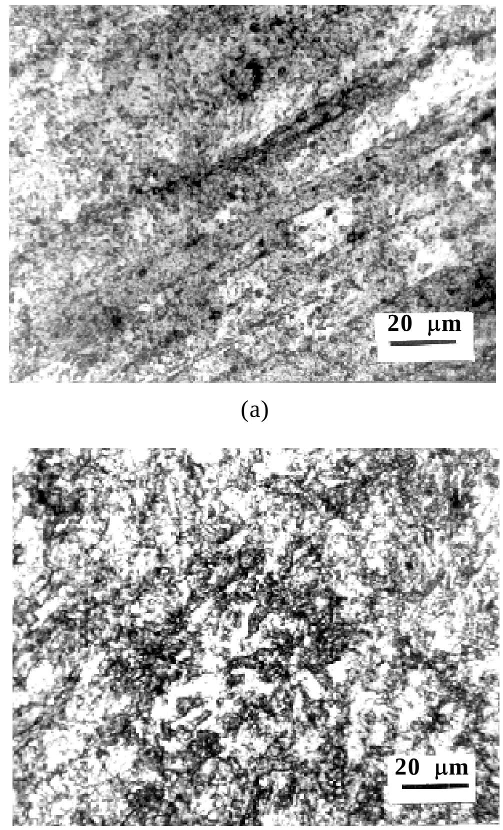


Figure 12. Impact toughness values of Steels 2 and 3.

different post-cooling conditions. Lath ferrite microstructure is obtained at faster cooling rate [Fig. 13(a)]. With slower cooling rate, the volume fraction of lath ferrite is reduced by the formation of polygonal ferrite [Fig 13(b)]. The TEM study shows different types of lath ferrite structure in the water-quenched specimen [Fig 14(a) and 14(b)]. High dislocation density has been observed in lath ferrite. The TEM of air-cooled specimen shows that very fine precipitates are interacting with dislocations [Fig. 14(c)].

The tensile property of Steel 4 shows the maximum strength values (927 MPa UTS, 860 MPa YS) in

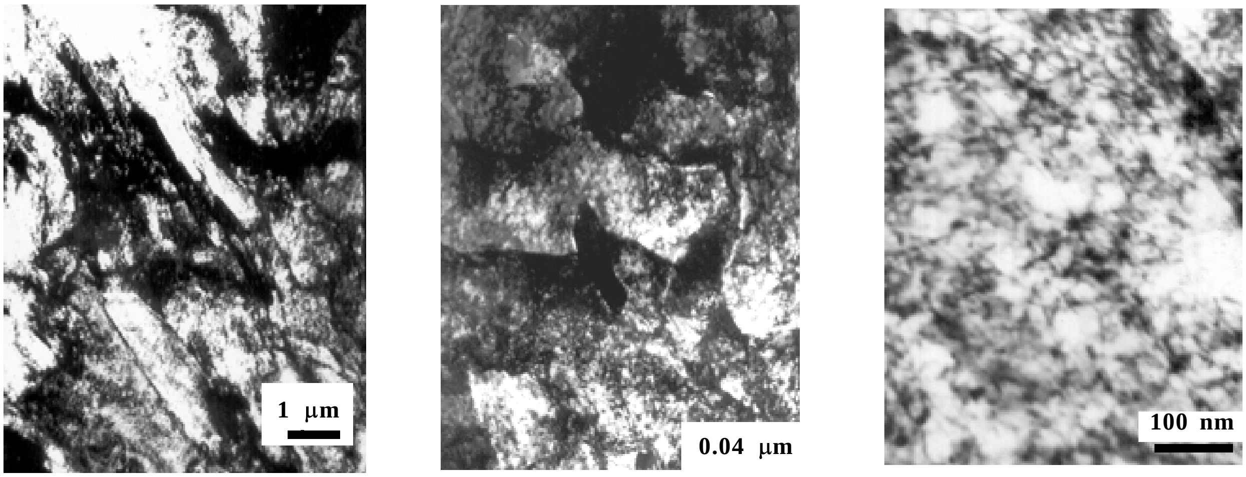


(a)

(b)

Figure 13. Optical micrographs of the ULC-HSLA steel containing boron (Steel 4) at different cooling conditions: (a) WQ (35 °C/s); and (b) SC (0.68 °C/s).

air-cooled condition, [Fig. 15]. The strength values obtained in water-quenched and sand-cooled conditions are very close to each other (~ 860 MPa UTS). Precipitation of fine microalloyed carbides and carbonitrides is responsible for the increase in strength value in air-cooled condition. These precipitates are carbides and carbonitrides of *Ti* and *Nb* and also *Cu* particles. These fine precipitates pin mobile dislocations generated during second-stage deformation of TMCP or from the IPS of formation and increase the strength value, which is accompanied by a decrease in the per cent elongation. For ultra-low carbon HSLA steel, the optimum combination of strength and toughness has been obtained in air-cooled condition. Slower cooling rate increases the volume fraction of polygonal ferrite in ultra- low carbon steels which increases the toughness value [Fig. (16)] by reducing the strength.



(a) (b) (c)

Figure 14. TEM study of Steel 4: (a), (b) show different types of lath ferrite in the WQ (35 °C/s), and (c) shows precipitation dislocation interaction in AC (1.15 °C/s) specimen.

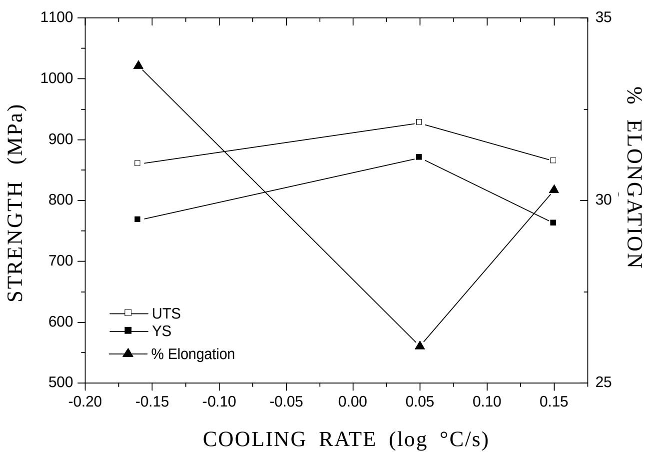


Figure 15. Tensile properties of Steel 4.

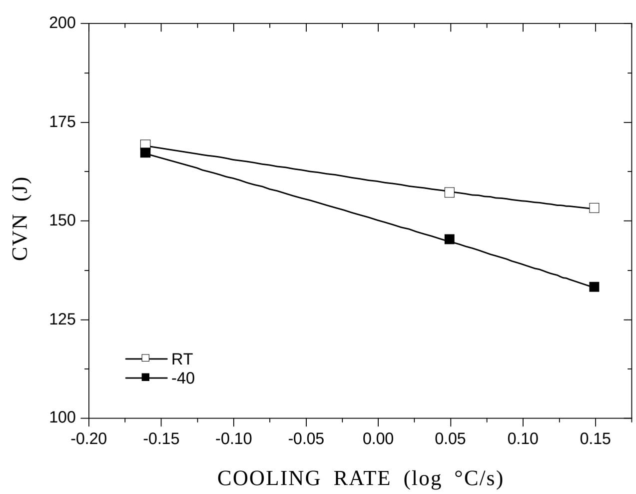


Figure 16. Impact toughness values of Steel 4.

4. SUMMARY

(1) In the HSLA-100 steel (Steel 1) microstructures obtained in water-quenched condition is lath martensite packet in the pancaked grains. On air cooling or sand cooling, the lath martensite structure changed to predominantly bainitic ferrite or granular bainite structure, with second-phase particles.

(2) Steel 1 shows a decrease in strength and an increase in elongation with decrease in cooling rate. The strength values obtained in this steel are in the range of 996 MPa–842 MPa UTS and 942 MPa–672 MPa YS along with 16-22 per cent elongation.

(3) The impact toughness values obtained at different cooling rates have a similar trend with per cent elongation. However, a sharp drop in toughness is observed at – 40 °C for water quenched steel. Toughness values slightly increase in sand cooled condition than in the air-cooled steel. The ductile-to-brittle transition temperature of this steel has been found to be – 40 °C.

(4) The impact fracture surface of air-cooled steel shows ductile failure with formation of dimples at 20 °C and at – 20 °C. The fracture surface changed to brittle failure with the formation of cleavage and river pattern at – 40 °C and at – 60 °C.

(5) In Steel 2 and Steel 3, in water-quenched condition, lath martensite packet is obtained. In air-cooled steels, the lath martensite changes to predominantly bainitic ferrite or granular bainite structure. When the cooling rate is further slowed down, i.e., in sand cooling polygonal ferrite or quasi polygonal ferrite, microstructure appears with random distribution of second-phase particles. A high strength value (~1200 MPa UTS) has been obtained in water-quenched steels due to lath martensite structure. The toughness value increases for slower cooling rates.

(6) The microstructures of the ultra-low carbon steels show lath ferrite or granular ferrite in water-quenched steel at faster cooling rate. With slower cooling rate, the volume fraction of lath ferrite decrease with an increase in polygonal ferrite. The tensile property of this steel shows the maximum strength values (927 MPa UTS, 860 MPa YS) in air-cooled condition. Precipitation of fine microalloyed carbides and carbonitrides may be responsible for the increase in the strength value in air- cooled condition. For ultra-low carbon HSLA steel, the optimum combination of strength and toughness has been obtained in air-cooled condition. Slower cooling rate increases the volume fraction of polygonal ferrite in ultra- low carbon steels which increases the toughness value by reducing the strength.

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