13

Ultrahigh Strength Steel: Development of Mechanical Properties Through Controlled Cooling

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1. Introduction

Structural steels with very high strength are referred as ultrahigh strength steels. The designation of ultrahigh strength is arbitrary, because there is no universally accepted strength level for this class of steels. As structural steels with greater and greater strength were developed, the strength range has been gradually modified. Commercial structural steel possessing a minimum yield strength of 1380 MPa (200 ksi) are accepted as ultrahigh strength steel (Philip, 1990). It has many applications such as in pipelines, cars, pressure vessels, ships, offshore platforms, aircraft undercarriages, defence sector and rocket motor casings. The class ultrahigh strength structural steels are quite broad and include several distinctly different families of steels such as (a) medium carbon low alloy steels, (b) medium alloy air hardening steel, (c) high alloy hardenable steels, and (d) 18Ni maraging steel. In the recent past, developmental efforts have been aimed mostly at increasing the ductility and toughness by improving the melting and the processing techniques. Steels with fewer and smaller non-metallic inclusions are produced by use of selected advanced processing techniques such as vacuum deoxidation, vacuum degassing, vacuum induction melting, vacuum arc remelting (VAR) and electroslag remelting (ESR). These techniques yield (a) less variation of properties from heat to heat, (b) greater ductility and toughness especially in the transverse direction, and (c) greater reliability in service (Philip, 1978). The strength can be further increased by thermomechanical treatment with controlled cooling.

1.1 Medium carbon low alloy steel

The medium carbon low alloy family of ultra high strength steel includes AISI/SAE 4130, the high strength 4140, and the deeper hardening and high strength 4340. In AMS 6434, vanadium has been added as a grain refiner to improve the toughness and carbon is reduced slightly to improve weldability. D-6a contains vanadium as grain refiner, slightly higher carbon, chromium, molybdenum and slightly lower nickel than 4340. Other less widely used steels that may be included in this family are 6150 and 8640. Medium-carbon low alloy ultrahigh strength steels are hot forgeable, usually at 1060 to 1230°C. Prior to machining, the usual practice is to normalise at 870 to 925°C and temper at 650 to 675°C. These treatments yield moderately hard structures consisting of medium to fine pearlite. It is observed that maximum tensile strength and yield strength result when these steels are tempered at 200°C. With higher tempering temperature, the mechanical properties drop sharply. The mechanical properties obtained in oil-quenched and tempered conditions are shown in Table 1**.**

Designation	Tempering temperature (°C)	Tensile strength (MPa)	Yield strength (MPa)	Elongation (%)	Hardness (HB)	Izod impact (J)	Fracture toughness (MPa√m)
4130	205 425	1550 1230	1340 1030	11 16.5	450 360	- -	70
4140	205 425	1965 1450	1740 1340	11 15	578 429	15 28	49
4340	205 425	1980 1500	1860 1365	11 14	520 440	20 16	46-AM 60- VAR
300M	205 425	2140 1790	1650 1480	7.0 8.5	550 450	21.7 13.6	-
D – 6a	205 425	2000 1630	1620 1570	8.9 9.6	- -	15 16	99

Table 1. Mechanical properties of medium carbon alloy steel.

1.2 Medium alloy air hardening steel

The steels H11, Modified (H11 Mod) and H13 are included in this category. These steels are often processed through remelting techniques like VAR or ESR. VAR and ESR produced H13 have better cleanness and chemical homogeneity than air melted H13. This results in superior ductility, impact strength and fatigue resistance, especially in the transverse direction, and in large section size. Besides being extensively used in dies, these steels are also widely used for structural purposes. They have excellent fracture toughness coupled with other mechanical properties. H11 Mod and H13 can be hardened in large sections by air-cooling. The chemical compositions and the mechanical properties of these steels are given in Table 2.

Designation	C (%)	Mn (%)	Si (%)	Cr (%)	Mo (%)	V(%)
H11 Mod	0.37 – 0.43	0.20 – 0.40	0.80 – 1.00	4.74 – 5.25	1.20 – 1.40	0.40 – 0.60
H13	0.32 – 0.45	0.20 – 0.50	0.80 – 1.20	4.75 – 5.50	1.10 – 1.75	0.80 – 1.20

Designation	Tempering temperature (°C)	Tensile strength (MPa)	Yield strength (MPa)	Elongation (%)	Hardness (HRc)	Izod impact (J)
H11 Mod	565	1850	1565	11	52	26.4
H13	575	1730	1470	13.5	48	27

Table 2. Chemical compositions and mechanical properties of medium alloy air hardening ultra high strength steel.

1.3 High alloy hardenable steel

These steels were introduced by Republic Steel Corporation in the 1960's and have four weldable steel grades with high fracture toughness and yield strength in heat treated condition. These nominally contain 9% Ni and 4% Co and differ only in carbon content. The four steels designated as HP9-4-20, HP9-4-25, HP9-4-30 and HP9-4-45 nominally have 0.20, 0.25, 0.30 and 0.45%C respectively. Among these steels, HP9-4-20 and HP9-4-30 are produced in significant quantities and their chemical composition and mechanical properties are given in Table 3 (Philip, 1978). As the carbon content of these steels increases, attainable strength increases with corresponding decrease in both toughness and weldability. The high nickel content of 9% provides deep hardenability, toughness and some solid solution strengthening. If the steel contains only higher amount of nickel but no cobalt, there would be a strong tendency for retention of large amounts of austenite on quenching. This retained austenite would not decompose even by refrigeration and tempering. Cobalt increases the Ms temperature and counteracts austenite retention. Chromium and molybdenum content are kept low for improvement of toughness. Silicon and other elements are kept as low as practicable.

Designation	C (%)	Mn (%)	Si (%)	Cr (%)	Ni (%)	Mo (%)	V (%)	Others (%)
HP 9-4-20	0.16–0.23	0.20–0.40	0.20 max	0.65–0.85	8.50–9.50	0.90–1.10	0.06–0.12	4.25– 4.75 Co
HP 9-4-30	0.29–0.34	0.10–0.35	0.20 max	0.90–1.10	7.0 – 8.0	0.90–1.10	0.06–0.12	4.25– 4.75 Co

Designation	Tensile strength (MPa)	Yield strength (MPa)	Elongation (%)	Hardness (HRc)	Izod impact (J)
HP 9-4-20	1380	-	-	-	-
HP 9-4-30	1650	1350	14	49 - 53	39

Table 3. Chemical compositions and typical mechanical properties of high alloy hardenable ultra high strength steel.

1.4 18 Ni maraging steel

Steels belonging to this class of high strength steels differ from other conventional steels. These are not hardened by metallurgical reactions that involve carbon, but by the precipitation of intermetallic compounds at temperatures of about 480°C. The typical yield strengths are in the range 1030 MPa to 2420 MPa. They have very high nickel, cobalt and molybdenum and very low carbon content. The microstructure consists of highly alloyed low carbon martensites. On slow cooling from the austenite region, martensite is produced even in heavy sections, so there is no lack of hardenabilty. Cobalt increases the Ms transformation temperature so that complete martensite transformation can be achieved. The martensite is mainly body centred cubic (bcc), and has lath morphology. Maraging steel normally contains little or no austenite after heat treatment. The presence of titanium leads to precipitation of Ni_3Ti. It gives additional hardening. However, high titanium content favours formation of TiC at the austenite grain boundaries, which can severely embrittle the age-hardened steel (Philip, 1978). The nominal chemical compositions of the commercial maraging steels are shown in Table 4. Typical tensile properties are shown in Table 5.

One of the distinguishing features of the maraging steels is their superior toughness compared to conventional steels. Maraging steels are normally solution annealed (austenitised) and cooled to room temperature before aging. Cooling rate after annealing has no effect on microstructure. Aging is normally done at 480°C for 3 to 6 hours. These steels can be hot worked by conventional steel mill techniques. Working above 1260°C should however be avoided (Floreen, 1978). Maraging steels have found varieties of applications including missile casing, aircraft forgings, special springs, transmission shafts, couplings, hydraulic hoses, bolts and punches and dies.

Grade	C (%)	Ni (%)	Mo (%)	Co (%)	Ti (%)	Al (%)	Other (%)
18Ni (200)	0.03 max	18	3.3	8.5	0.2	0.1	-
18Ni (250)	0.03 max	18	5.0	8.5	0.4	0.1	-
18Ni (300)	0.03 max	18	5.0	9.0	0.7	0.1	-
18Ni (350)	0.03 max	18	4.2	12.5	1.6	0.1	-
18Ni (cast)	0.03 max	17	4.6	10.0	0.3	0.1	-
18Ni (180)	0.03 max	12	3	-	0.2	0.3	5.0% Cr

Table 4. The nominal chemical compositions of maraging steel.

Grade	Heat treatment	Tensile strength (MPa)	Yield strength (MPa)	Elongation (%)
18Ni (200)	A	1500	1400	10
18Ni (250)	A	1800	1700	8
18Ni (300)	A	2050	2000	7
18Ni (350)	B	2450	2400	6
18Ni (cast)	C	1750	1650	8

A: solution treat 1h at 820°C, aging 3h at 480°C; B: solution treat 1h at 820°C, aging 12h at 480°C;
C: anneal 1h at 1150°C, aging 1h at 595°C, solution treat 1h at 820°C, aging 3h at 480°C.

Table 5. Mechanical properties of the heat treated maraging steel.

1.5 Issues and objective

In addition to high strength-to-weight ratio, ultra high strength steels should possess good ductility, toughness, fatigue resistance and weldability. Some of the currently employed steels, like maraging steels, are highly alloyed and are expensive. Search for less expensive steels with better properties, is therefore a continuing process. High strength in these alloys is obtained by exploiting all the strengthening mechanisms, by careful control of alloying and subsequent processing. Often when strength is raised by alloying and thermomechanical treatment, ductility and toughness suffer. Additionally one can have serious problems with fatigue properties. Many defects are introduced, and inferior properties are obtained during the solidification process. It is, therefore, advantageous to exercise great control during this process. Secondary refining processes like vacuum arc remelting (VAR) and electroslag refining (ESR) are often employed to obtain superior properties in these materials for critical applications. Electroslag refining is known to give low inclusion content, low macro-and micro-segregation, and low microporosity due to near-directional solidification from a small pool with application of controlled cooling. Many alloys for critical application now use this process to ensure reliability and good properties.

The material developed earlier at Indian Institute of Technology (IIT) Bombay and Vikram Saravai Space Center (VSSC), Trivandrum, India with a yield strength of 1450 MPa, is qualified as aerospace application (Suresh et al., 2003; Chatterjee et al., 1990). This was a medium-carbon low alloy steel used mostly in tempered condition. The chemical composition of the alloy is: 0.3% C, 1.0% Mn, 1.0% Mo, 1.5% Cr, 0.3% V and named as 0.3C-CrMoV (ESR) steel (Suresh et al, 2003). The microstructure of heat treated alloy primarily consists of tempered lath martensite. The primary objective of the present work is to develop an alloy with yield strength in excess of 1700 MPa with adequate ductility and impact toughness. It has been achieved through:

a. ESR processing of the alloys
b. Thermomechanical treatment with controlled cooling

1.6 Plan of investigation

UHSS is mostly developed by interplay of all strengthening mechanisms. Grain refinement is achieved either by fine precipitates which pin the austenite grain boundaries by micro alloys (Tanaka, 1981; Umemoto et al., 1987). Precipitation of carbides and carbonitrides both at high temperatures or during cooling and tempering helps to improve the mechanical properties for specific needs (Bleck et al., 1988). Ductility and toughness suffer in most methods of strengthening when one tries to increase strength. The approach in the present work, therefore, is to adjust the chemistry and optimise the production process to obtain clean steel with finer microstructures by special melting process. Therefore, it is advantageous to process these materials through a secondary refining process like electroslag refining (ESR), which ensures the cleanliness and chemical homogeneity (Shash, 1988; Choudhary & Szekely, 1981). Further improvement of mechanical properties is to be obtained by a control thermomechanical treatment (TMT). Melting and casting of alloys and subsequent processing like TMT are the two main aspects in this study.

In the first part of the study, the alloys were prepared with variation of chemical composition starting with a basic composition of 0.3%C, 4.2%Cr, 1%Mn, 1%Mo and 0.35%V. In the previous study, the effects addition of titanium and niobium, and increase of chromium and vanadium contents on the mechanical and microstructural properties were investigated (Maity et al., 2008a, 2008b). Most of these alloys in as cast tempered condition displayed minimum yield strengths of 1450 MPa with elongation of about 9-12% and impact toughness in many cases was in excess of 300 $kJ.m^{-2}$. For further improvement of mechanical properties especially to increase the toughness values, the basic steel is alloyed with 1-3% of nickel in this study. Nickel is generally added in many low alloy steels to improve low temperature toughness and hardenability (Maity et al., 2009). It also strengthens the steel by solid solution hardening, and is particularly effective when it is used in combination with chromium and molybdenum (Umemoto et al., 1987). Nickel is known to increase the resistance to cleavage fracture in steel and decreases ductile-to-brittle transition temperature. The medium-carbon low-alloy martensitic steel attains the best combination of properties in tempered condition owing to the formation of transition carbides

(Malakondaiah et al., 1997). It decreases the ductile-to-brittle transition temperature by promotion of a cross-slip of dislocations in ferritic as well as martensitic steels (Arsenault, 1967; Jolley, 1968; Norstr¨om & Vingsbo, 1979). This effect promotes deformation rather than cleavage fracture and therefore increases toughness. Also, nickel is known as the alloying element, which slightly increases the hardness of martensite and has a weak effect on hindering in decrease of hardness with the tempering temperature with tempered martensite and retained austenite (Grange, 1977).

In the second part of the investigation, it was attempted to further increase the strength and toughness by optimised schedule of thermomechanical treatment. It normally increases the tensile properties and toughness (Dhua, 2003) without reducing ductility or brittle fracture resistance (Akhlaghi, 2001; Jahazi & Egbali, 2000). With controlled rolling it is possible to refine the ferrite structures directly after finish rolling or by using additional accelerated cooling. The processes can be divided into the following stages (Kern et al., 1992): i) forming in the region in which the austenite matrix recrystallises, and/or ii) forming in a heterogeneous austenite-ferrite region after partial decomposition of austenite to ferrite followed by iii) a process of accelerated cooling after the controlled rolling (Umemoto et al., 1987; Kern et al., 1992). The essential hot rolling parameters of the thermomechanical process are: i) slab reheating temperature for dissolution of the precipitated carbonitrides, ii) roughing phase for producing a fine, polygonal austenite grain by means of recrystallisation, ii) final rolling temperature, and iv) degree of final deformation in the temperature range. For controlled cooling the additional parameters are: a) cooling rate, and b) cooling temperature. Controlled rolling and accelerated cooling play important role in the modification of final microstructures. The main result in the first phase of rolling is to increase of yield strength and toughness. This is attributed to the resultant fine grain microstructures. In the second phase (accelerated cooling), the contribution of the increase of the resultant mechanical properties is caused not only by refining of ferrite grains, but also by the change of the morphology of the various phases in the ferrite matrix (bainite or martensite). It is reported that if rolling is completed at a relatively high temperature (in the high temperature austenite range) and the sample is cooled in air, one gets a mixed microstructure of upper bainite and martensite (Tanaka, 1981). Accelerated cooling results in the formation of mostly martensite phase.

The combine influence of alloying elements and thermomechanical treatment allows to exploit of different mechanisms of strengthening, such as precipitation hardening, grain refinement, and transformation hardening by means of bainite and martensite transformations (Bleck et al., 1988). Although, application of thermomechanical treatment especially to high strength low alloy steel (HSLA) is known, little systematic work has been carried out with application of this technique to ultra high strength steels (UHSS) (Jahazi & Egbali, 2000). Present study, therefore is also aimed to produce ultrahigh strength steel through optimised schedule of the processes parameters of thermomechanical treatment so that such high strength materials can be rolled in the existing rolling mill and a minimum yield strength of 1700 MPa along with good impact toughness is achieved.

2. Experiment

2.1 Preparation of as-cast alloys

The alloys were produced by induction melting followed by electroslag refining (ESR). The electrodes, which were produced by induction melting, were remelted using the pre-fused flux of 70 CaF_2: 30 Al_2O_3. About 800g slag was used for each experiment and it was preheated in a muffle furnace at 800°C for 5-6 hours to eliminate free and combined moisture, before charging into the ESR furnace. Remelting was done in a water cooled steel mould of 80 mm diameter, with electrode connected to positive end of DC power. At equilibrium the current and voltage during this process were about 730 amps, and 25 ± 2V respectively, with mould water flow rate of 30 litre/minute and base-plate water flow rate of 20 litre/minute. After the ESR process, cooled ingots were taken out from the mould. ESR ingots were approximately 150 mm long and 75 mm in diameter. The ingots were annealed in a muffle furnace at 975°C for 8-9 h. After annealing, 20 mm and 10 mm lengths were discarded from the bottom and top of the ingot respectively. Samples for chemical analysis and mechanical tests specimens were taken from the ingot. The mechanical test specimens underwent for heat treatment. Heat treatment was organised in a tubular furnace of approximately 80 mm constant temperature (±2.5°C) zone. Argon atmosphere was provided to prevent any oxidation. The base alloy (ESR1) were austenitised at 975 °C, quenched in oil and tempered at 475°C (Chatterjee et al., 1990). For nickel containing steels, the specimens are austenitised at 930°C and tempered at 475°C (Maity et al., 2009). After heat treatment, the samples were prepared for mechanical tests by machining and grinding to final size.

2.2 Thermomechanical Treatment (TMT)

The size of the as-cast ESR ingots selected for rolling was about diameter of 75 mm and length of 60-65 mm. These ingots were soaked at 1200°C for about one hour pre rolled to □ 23.7 x 23.7 mm bars. During TMT, the bars were reheated again to 1200°C, soaked for 90 minutes, transported to the rolling mill and held there till it reached to 950°C, and then rolled to a size of approximately □ 16.5 x 26.5 mm. The second pass was applied in equal deformation as soon as the temperature reached to 850°C and finally it was rolled to □ 11 x 29 mm plates. Immediately thereafter, the samples were quenched in the different cooling mediums. Total reduction was approximately 30% in area and 50% in thickness. About 10 mm and 2 mm lengths were discarded from both the ends and the sides respectively and the samples were prepared for mechanical properties and microstructural studies correspond to the rolling direction. The sampling plan is shown in Figure 1**.**

2.3 Characterisation

Most of the chemical analysis was carried out by atomic absorption spectroscopy (AAS). Carbon was analysed in a Strohlien apparatus. Nitrogen and aluminium were analysed by a spectrometer. Sulphur and phosphorous were analysed in a SPECTROLAB analytical instrument. The heat treated specimens were analysed for various mechanical properties. For tensile test, round specimens of 4 mm diameter and 24 mm gauge length were prepared, as per DIN 50125-A 12 x 60, 1991 specification and tested at room temperature using the Servo Hydraulic UTM. Charpy U-notch impact toughness specimens were prepared as per DIN 50115-DVM, 1975 specification and also tested at room temperature. Hardness of quenched as well hardened and tempered materials was measured at on Rockwell C hardness tester with application of 15 kg load. Optical, SEM and TEM specimens were prepared by standard method. The TEM-carbon replica technique was employed to extract the precipitates from the specimens. The carbon replicas were examined using field emission electron microscope equipped with energy dispersive X-ray spectrometer (EDS).

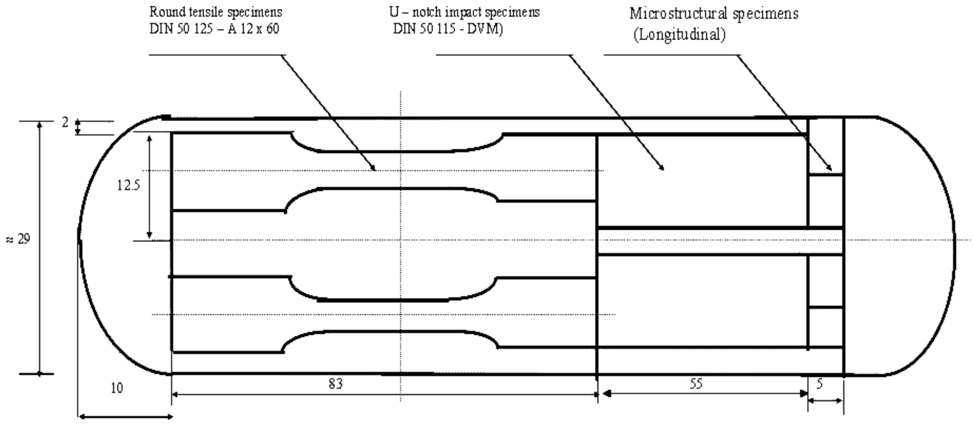


Fig. 1. Sample plan of the rolled plate in rolling direction.

3. Result and discussion

3.1 Properties of as-cast alloy

All ingots prepared by ESR process had smooth and bright surface with few blemishes. The loss of alloying elements was about 5% during ESR processes. Four alloys were prepared starting from a base alloy composition of 0.28%C, 1.0% Mn, 1.0% Mo, 0.35% V, 4.2% Cr. In other three alloys about 1% to 3% Ni was added with the basic composition. The chemical composition of the ESR ingots is illustrated in Table 6. The amount of nickel in ESR2 is about to 1%, 2% in ESR3 and 3.3% in ESR4, respectively. It can be also noticed that the amount of sulphur in all ESR ingots is substantially low. In our earlier work, it is reported that the inclusions in the electrodes were mainly of oxide and sulphide type, which were substantially removed during ESR process (Maity et al., 2006). Chemical homogeneity of the ESR steel is confirmed from glow discharge optical emission spectroscopy (GDOES) analysis and reported in our study (Maity et al., 2006). This study showed that the micro-segregation of chromium, carbon, silicon, manganese, and vanadium were minimal in ESR alloys.

Sample	Chemical composition of ESR ingot (in wt.%)										
	C	Mn	Cr	V	Mo	Si	Ni	Al	N	P	S
ESR 1	0.28	1.00	4.20	0.34	1.01	0.24	00	0.067	0.0161	0.031	0.011
ESR 2	0.28	0.91	4.50	0.35	0.97	0.19	1.07	0.057	0.0158	0.038	0.011
ESR 3	0.28	0.86	4.10	0.47	1.37	0.27	1.97	0.037	0.0128	0.034	0.010
ESR 4	0.30	1.20	4.60	0.47	0.94	0.19	3.29	0.110	0.0108	0.040	0.008

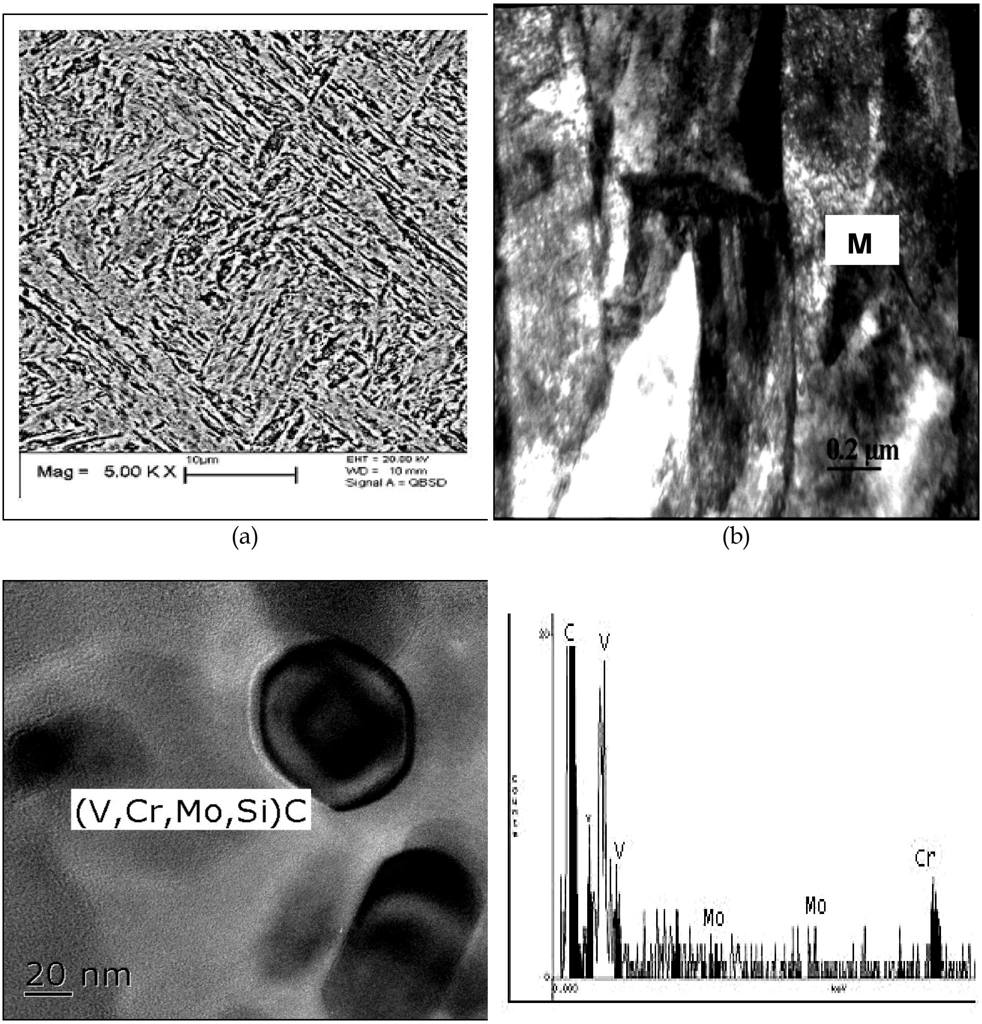
Table 6. Chemical composition of ESR ingot.

In this study, ESR1 is the basic steel. The mechanical properties of the as-cast and as-tempered alloy are shown in Table 7. The yield strength of this alloy is 1450MPa with good ductility and charpy impact toughness. The optical, SEM and TEM micrographs of base alloy (ESR1) are shown in Figure 2. The optical, SEM studies reveal that the microstructures of the tempered specimens mostly consist of lath martensites. The bright field TEM micrograph confirms that the inter lath martensite spacing is of the order of 550-700 nm. The carbon replica micrographs of this steel and the associated EDS analysis (also shown in Figure 2) show the precipitation of complex carbides. The precipitates are spherical in shape with rounded edges evenly distributed in the metal matrix. The EDS analysis of the precipitates shows that these are complex carbides/carbonitrides comprising vanadium, molybdenum and chromium. It can be noted that at the austenitising temperature of 975°C, the precipitates in ESR1 alloy are expected to contain very little amount of chromium and molybdenum because at this temperature most of these precipitates go into solution (Maity et al., 2006). Only vanadium carbonitrides remains partly undissolved in this temperature and chromium and molybdenum would have been precipitated into these pre-existing vanadium carbonitride precipitates during cooling and subsequent tempering.

Sample	Chemistry highlights	Room temperature mechanical properties					Grain Size (μm)
	Ni (wt%)	UTS (MPa)	YS (MPa)	Elongation (%)	Impact strength (kJ.m^{-2})	Hardness (HRc)	
ESR1	-	1660	1450	11.2	300	44	65
ESR2	1.07	1670	1500	9.5	400	45.5	51
ESR3	1.97	1712	1506	12.6	328	46.7	55
ESR4	3.29	1758	1542	9.6	274	46.5	56

Table 7. Mechanical properties of as-cast alloy in quenched-and-tempered condition.

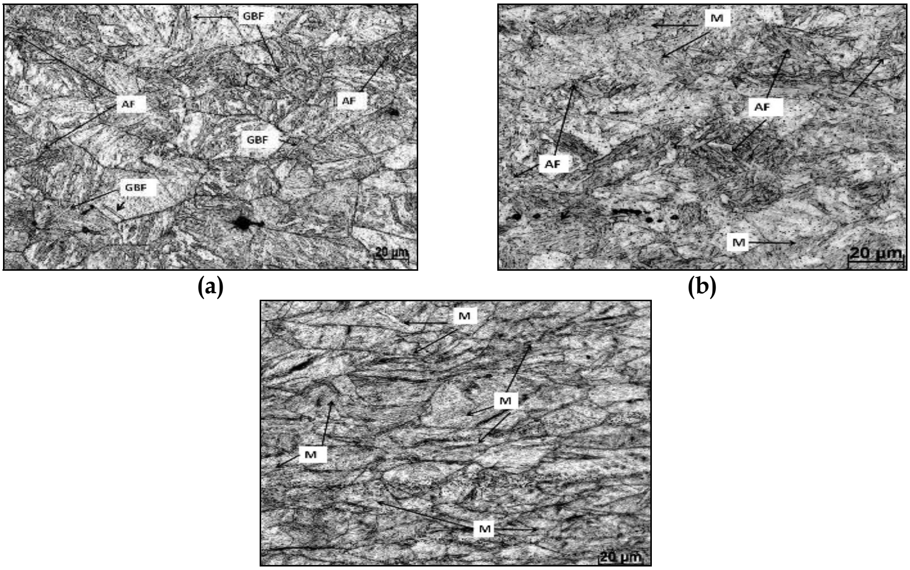
It can be noticed from Table 7 that on addition of 1% nickel in ESR2 alloy, the grain size is marginally reduced to 51 μm. The yield strength has been increased, and there is a marginal increase in hardness and tensile strength. It may be observed that the impact toughness significantly increases with the increase of nickel content up to 1%. On further increasing nickel to 2% (ESR3) and 3.2% (ESR4), the tensile strength and yield strength progressively increases and the later reaches a value of 1542 MPa in 3.2 % nickel steel (ESR4). Impact toughness drops sharply. At the same time grain size remains unchanged. The trend of increase in impact toughness from base alloy to 1% nickel steel and the decrease of its values at higher nickel containing alloys are interesting. The optical micrographs of the lightly etched specimens of nickel steels are shown in Figure 3. It reveals that the microstructures of nickel containing steel differ significantly from 1% nickel steel (ESR2) to 3% nickel steel (ESR4). In ESR2 alloy, the microstructures consist of some amount of grain boundary ferrite (GBF) and acicular ferrite (AF) in the martensite matrix. When nickel content is further increased in ESR3 to ESR4 steel, the GBF and AF phases significantly decreases. It is interesting to note that ESR4 steel consists of predominantly lath martensite microstructures. The effect of the GBF phases and the AF phases on the toughness of the steel is discussed later. The SEM micrographs of these steel are shown in Figure 4. It can be noticed from this figure that all the specimens consist of tempered lath martensite. The lath is uniform and seems to be finer in higher nickel alloys. The thermodynamic stability of precipitates of these steel is estimated by CHEMSAGE software as shown in Figure 5. It can be noticed from the figure that at the austinitising temperature of 930°C all the precipitates except the vanadium carbides are dissolved. The fraction of undissolved vanadium is about 60% which has locked equivalent atomic percentage of carbon (considering the precipitate as VC) and forms corresponding vanadium carbides/carbonitrides. The calculated dissolved carbon contents at the austenitising temperature are: 0.15% in the alloy containing 2% nickel steel (ESR3) and similarly 0.19 % in the alloy containing 3.3% nickel steel (ESR4). The slight increase (0.04 wt.%) of dissolved carbon in 3.2% nickel steels might have some effect in strengthening the martensite. It might be one of the reasons for improvement of strength values apart from possible solid solution hardening effect of nickel.



(a) (b)

(c) (d)

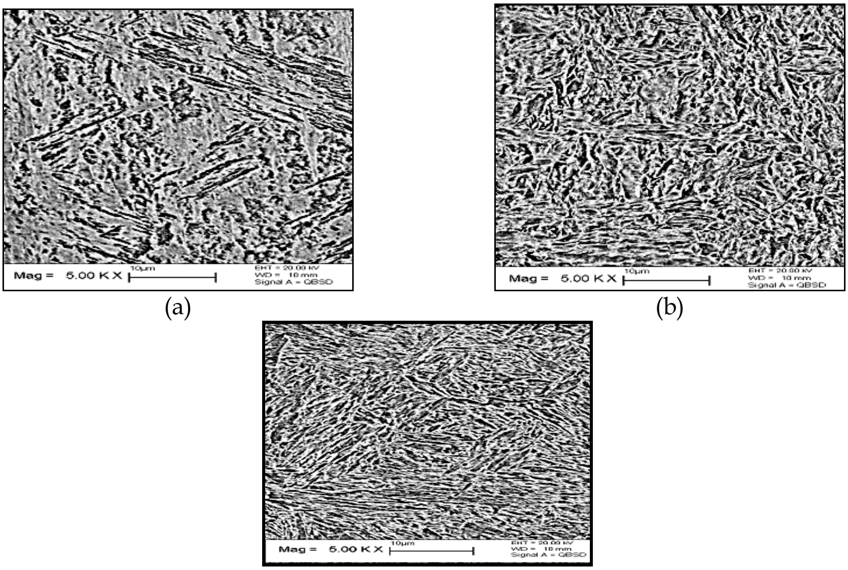
Fig. 2. (a) SEM image, (b) TEM bright field image, (c) TEM-carbon replica micrograph and (d) EDS analysis of the precipitates of as-cast base alloy (ESR1) sample quenched at 975°C in oil and tempered at 475°C.



(a) **(b)**

(c)

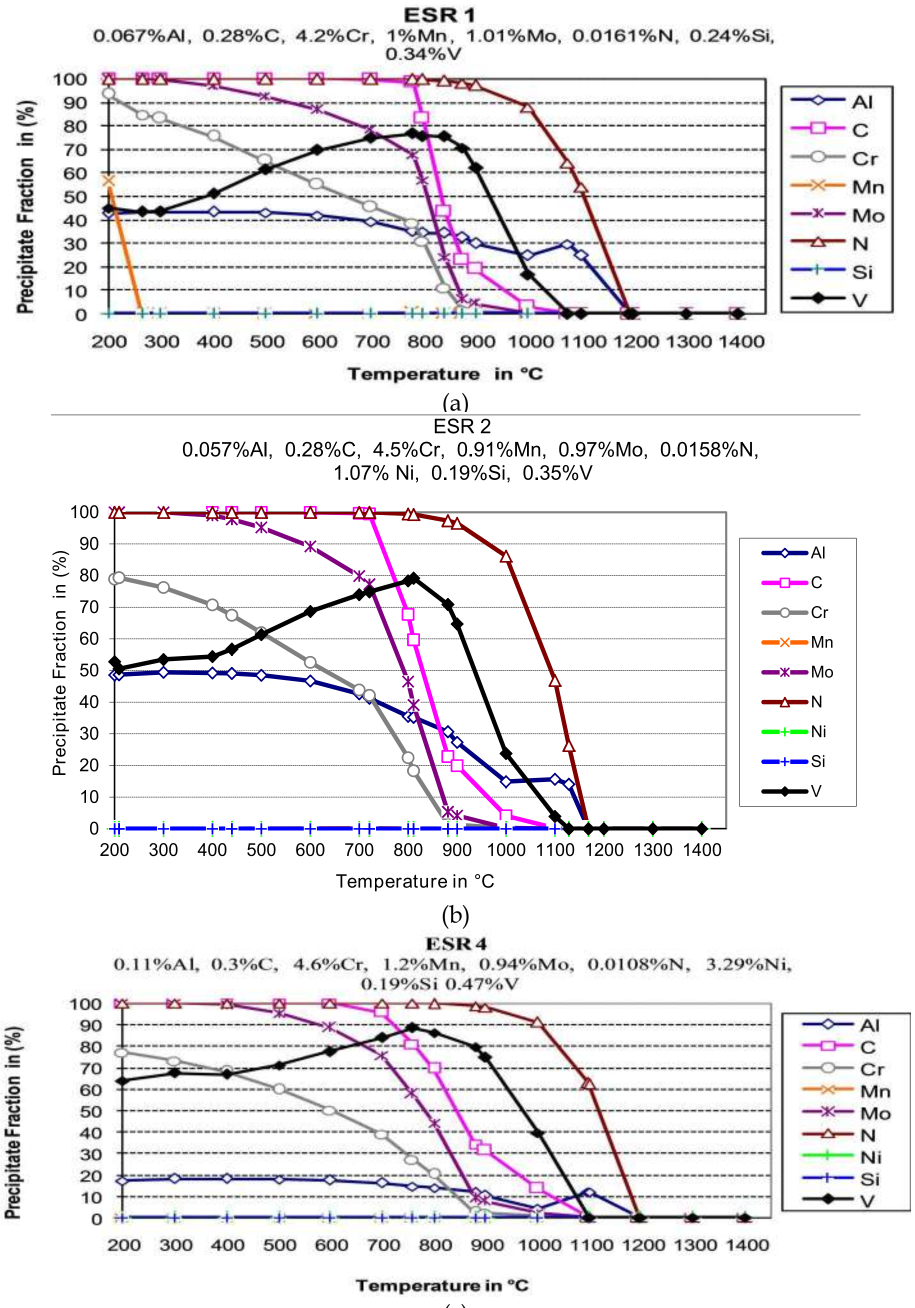
Fig. 3. Optical micrographs of nickel steels, showing the decreasing tendency of formation of acicular ferrite (AF) and grain boundary ferrite (GBF) in as-cast, quenched and tempered specimens of (a) ESR2, (b) ESR3, and (c) ESR4 alloy.



(a) (b)

(c)

Fig. 4. SEM micrographs of steels, showing the effect of nickel on the fineness of martensite laths in (a) ESR2, (b) ESR3, and (c) ESR4 alloy in as-cast tempered condition**.**



(a)

(b)

(c)

Fig. 5. Calculation of precipitate stability using CHEMSAGE software for ESR1, ESR2 and ESR4 alloy showing the volume fraction of precipitates.

3.2 Optimisation of processes parameters of TMT

It is possible to obtain optimum combination of strength and toughness by a control process parameters of thermomechanical treatment such as slab reheating temperature, deformation temperature, deformation per pass, cooling rate, etc (Kim et al., 1987). In the present study, it was attempted to optimise some of the process parameters like slab reheating temperature, deformation temperatures and the cooling rate of the cooling medium, etc which are discussed in the following section.

3.2.1 Soaking temperature

The initial stage of any hot rolling process usually consists of a selection of proper soaking temperature. At this temperature, attempt is normally made to dissolve all the carbides or carbonitrides present in the steel, so that these can be re-precipitated at smaller sizes in the later stage of the process. At the same time, too high soaking temperature leads to increase in austenite grain size, which controls the final microstructure. Therefore, it is necessary to select the appropriate soaking temperature at which the optimum results may be achieved. The microalloys form different carbides and carbonitrides, which go into solution at different temperatures, and therefore one needs to know these temperatures. Equilibrium stability of the carbides and carbonitrides in the alloys were calculated using CHEMSAGE software and the result are shown in Figure 5. The calculation is based on the chemical composition of the steel. Calculations were done for temperatures in the range of 200°C to 1400°C and in the intervals of 100°C. It may be noticed from these figures that the precipitates of carbides in ESR1, ESR2, ESR3, ESR4 are almost completely dissolved at around 900-1000°C and nitrides at 1200°C. The soaking temperature of these steels was therefore fixed at 1200°C.

3.2.2 Deformation and deformation temperature

Hot compression tests were performed to get an idea about the required load during hot rolling for a given amount of deformation. The specimen size was identical for all alloys. It was cylindrical in shape with 8 mm diameter and 14.4 mm height. The samples were reheated in a controlled atmosphere in a cast iron mould. The compression tests were performed at 1200°C with a strain rate of 1.0 s^{-1} with 50% total reduction. Result of hot compression test is represented by stress vs. degree of deformation (flow stress curve). The entire test was performed within 10 seconds. Visual observation showed that no major defect occurred in the compressed samples. Figure 6 shows the flow stress curves of ESR1 (base alloy), ESR2 (1% Ni), ESR3 (2 %Ni), and ESR4 (3.2% Ni). Except ESR3 alloy, the curves are similar for all the steels. The gradual increase of stress in all the alloys reflects the work hardening of the austenite. It can be inferred from Figure 6 that the required stresses for 50% hot deformation of the steels for all alloys are in the range 60 and 70 MPa, except in ESR3 (2% Ni) requiring the highest stress (80 MPa). TTT diagram of the base alloy (ESR1) has been predicted and reported using a model based on the chemistry of the metal (Maity et al., 2006). The calculated diagram for ESR1 steel is shown in Figure 7. This figure predicts that AC_1 temperature of this steel is about 825°C and martensite start transformation (Ms) temperature is above 300°C. Fast cooling below Ms temperature, could lead to transformation of martensite. Relatively slower cooling may result in a mixture of bainite and martensite. It was not possible to model the TTT diagram for the nickel containing alloys, as the γ-loop shifted extremely to the right. The diagram provides probable information regarding the beginning and end of transformation into stable and metastable phases. It was planed to roll the material in the two-phase α-γ region between AC_3 and AC_1 temperatures. As the α- phase in the two phase region being softer than the γ-phase in the stable γ-region (Yu et al., 2006), the high strength steels could then be rolled with the existing equipment. Additionally, if the first phase of rolling is done at a relatively high temperature in the two-phase region (above the recrystallisation temperature), one can get dynamic recrystallisation and finer austenite grains. The final pass can be made just above the AC_1 temperature so that recrystallisation can be limited and work hardening effect can be achieved (Kawalla & Lehnert, 2002). These arguments are based on equilibrium temperature. In reality, austenite to ferrite reaction may be sluggish enough throughout the rolling range. Small amount of ferrite may of course forms during rolling due to deformation induced transformation.

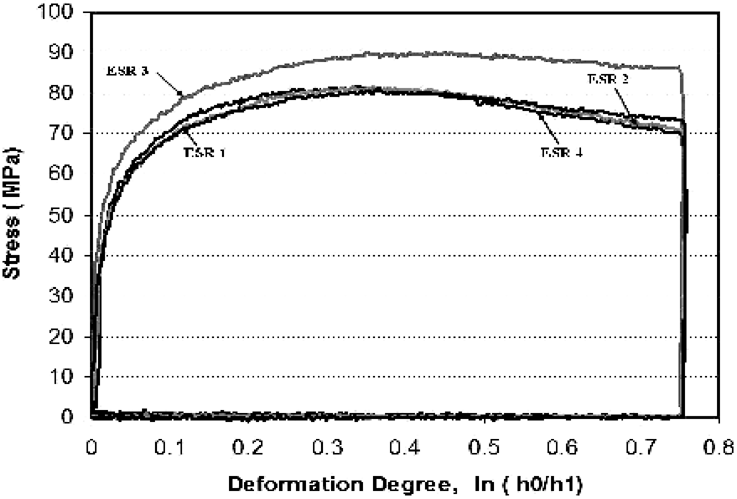


Fig. 6. Result of hot compression tests (50% reduction) on as-cast samples of ESR1, ESR2, ESR3 and ESR4 alloy.

3.2.3 Cooling rate of the medium

The cooling rate of the as-cast alloys was determined experimentally. The as-cast specimens were heated to 1200°C and after soaking at this temperature, the samples were held outside the furnace till it cooled to 850°C, and were then allowed to cool in different coolants. The selected coolants were air, oil, polymer-water mixture (1:1), polymer-water mixture (1:1.5) and the polymer-water mixture (1:2). The progress of cooling of the specimens in these coolants is shown in Figure 8. The figure shows that the rate of cooling is slowest in air, and polymer-water (1: 2) mixture results in the severest cooling. Cooling in oil is faster than the other two polymer-water mixtures down to a temperature of 250°C. The polymer-water (1:2) mixture was not selected for the final experiments, as it was considered too severe and therefore may lead to cracks. Use of the polymer-water (1:1) and (1:1.5) mixtures results in similar cooling profiles in the 300-700°C range. The polymer –water (1:1.5) mixture was used along with air and oil cooling in the final experiments. The average cooling rate for these coolants was estimated and it was 1.3°C.s^{-1} for air, 16°C.s^{-1} for polymer-water (1:1.5) mixture and 28°C.s^{-1} for oil, in the temperature range of 700°C-300°C. At temperatures below 300°C, oil cools slower than the polymer water solution.

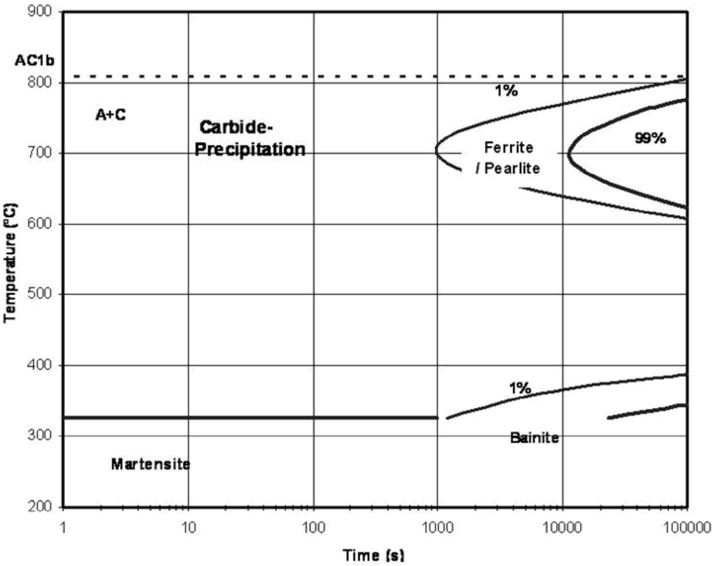


Fig. 7. Modelled TTT diagram of ESR1 (base alloy) showing AC_3 and M_S temperature.

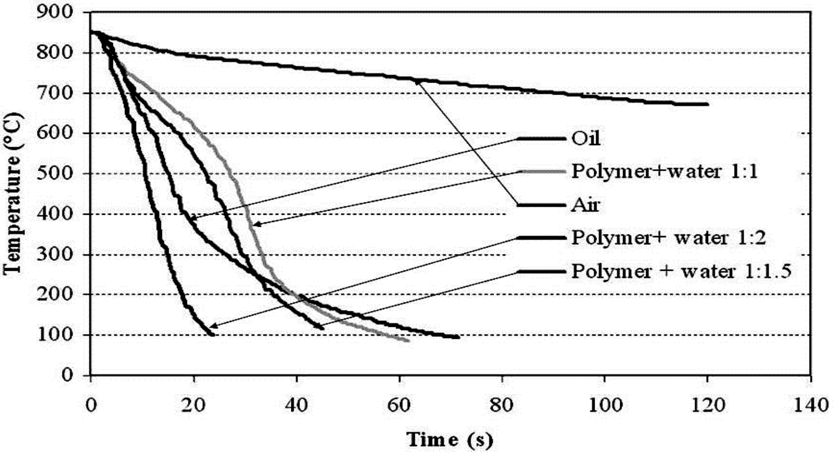


Fig. 8. Estimated average cooling rate of the ESR1 (base alloy) in different coolants.

3.2.4 Modelling of Continuous Cooling Transformation (CCT) diagram

Estimation of different phases was modelled to obtain a relationship of the phases to be appeared in different cooling conditions. The data predicts the transformation of various phases on application of continuous cooling conditions. The model used for this purpose was neural network based and claimed an error band of ± 14K for Ms temperature and ±10% for phase percentages (Ion, 1984; Doktorowski, 2002). Starting temperature for the model calculation has been considered as 900°C. The CCT diagram obtained by this model is shown in Figure 9. It predicts that at the slower cooling rate (less than 2-5K/s) the microstructures consist of a mixture of bainite, martensite and some amount of ferrite. Fast cooling (>10 K/s) on the other hand results in complete transformation to martensite. The results of these models are useful in analysing the results obtained after TMT.

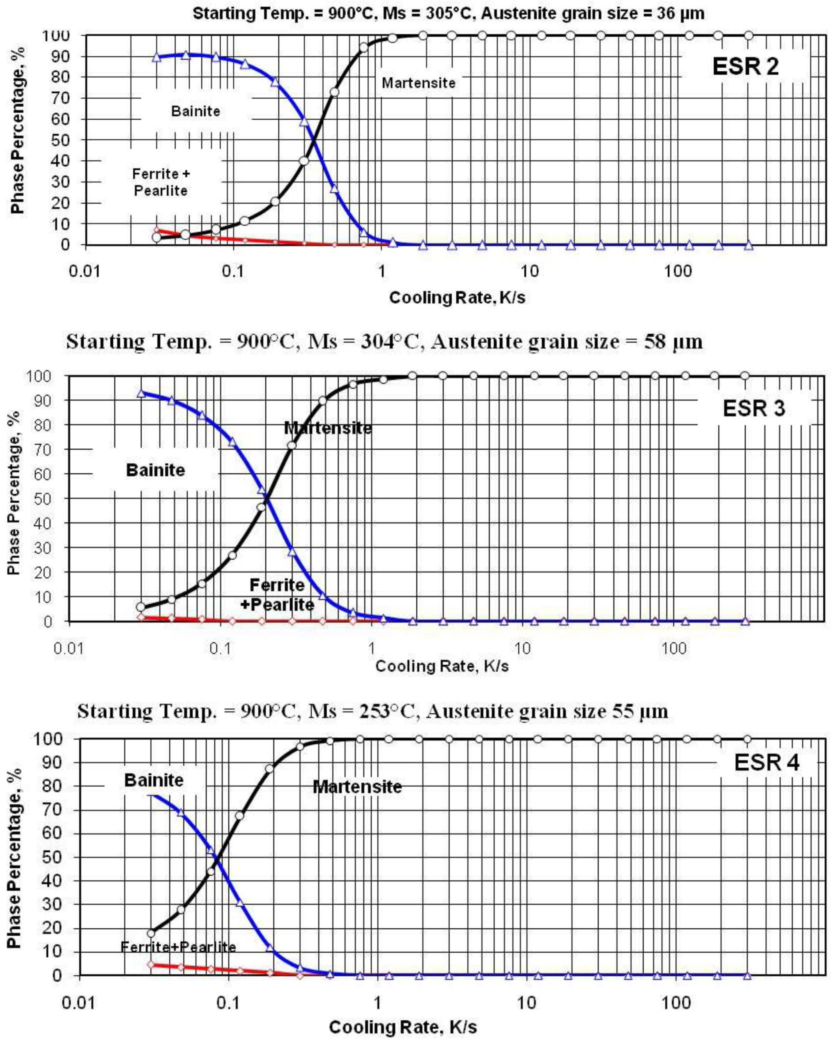


Fig. 9. Modelled CCT diagram predicts the microstructure constituents and Ms temperature for ESR 2, ESR3, and ESR4 alloys.

3.3 Properties of TMT plates

The summary of the observations during the hot rolling experiments is given in Table 8. The rolling stresses for each steel were calculated by the standard method (Zouhar, 1970). The calculated rolling stresses for the different alloys are illustrated in Figure 10. It can be noted that ESR1, base alloy, required the minimum stresses (113 MPa for 1st pass and 254 MPa for final pass). The three nickel containing steels, viz., ESR2, ESR3 and ESR4 required higher

steel	Initial		First pass				Final Pass				Cooling medium
	H_o (mm)	B_o (mm)	H_1 (mm)	B_1 (mm)	Fw_1 [kN]	Av σ_1 (MPa)	H_2 (mm)	B_2 (mm)	Fw_2 [kN]	Av σ_2 (MPa)	
ESR1	21.2	23.1	16.5	26.5	122	113	11.1	30.0	326	254	Air
	21.2	23.1	16.5	26.5	129		11.1	30.0	341		Oil
	21.2	23.1	16.5	26.5	120		11.1	30.1	340		Polymer
ESR2	21.0	22.6	16.5	26.5	131	126	11.1	29.9	327	254	Air
	21.0	22.6	16.5	26.5	135		11.2	30.0	328		Oil
	21.0	22.6	16.5	26.5	140		11.2	30.1	344		Polymer
ESR3	21.4	22.5	16.5	26.5	154	136	11.2	29.0	341	263	Air
	21.4	22.5	16.5	26.5	155		11.2	29.3	340		Oil
	21.4	22.5	16.5	26.5	148		11.2	29.3	324		Polymer
ESR4	21.0	22.3	16.5	26.5	143	141	11.2	29.8	345	267	Air
	21.0	22.3	16.5	26.5	156		11.2	29.9	337		Oil
	21.0	22.3	16.5	26.5	162		11.2	29.8	360		Polymer

Table 8. Experimental data of thermomechanical treatment.
Initial dimension of steel: 22.7 x 22.7 mm, final dimension of steels: plate 11 x 29 mm, temperature: 1st pass: 950°C, final pass: 850°C, ingot soaking temperature 1200C, soaking time: 90 minutes. Fw_1 is load, σ_1 is stress, H_o is initial height and B_o initial width.

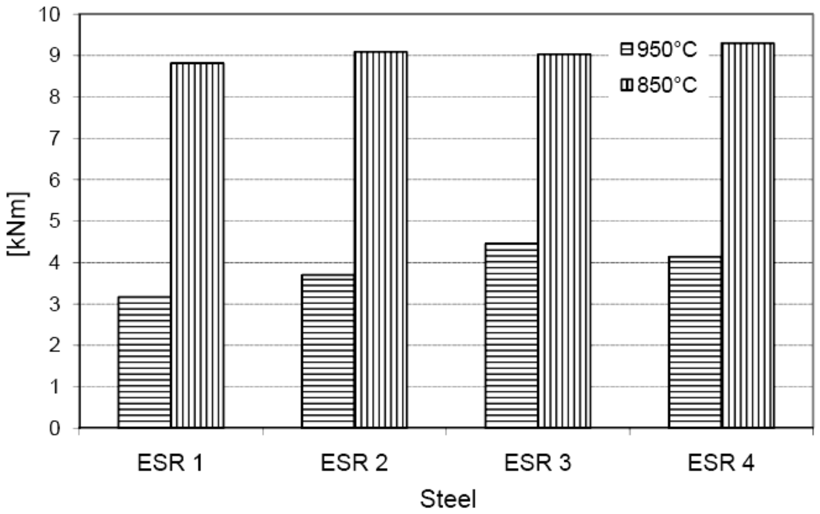


Fig. 10. Rolling stresses for first and final pass during hot rolling experiments.

stresses than that of ESR1. The result also shows that the stress for the final pass is much higher than that for the first pass in all samples. The rolling torque is also shown in Figure 11. The four selected grades of steels underwent hot rolling as mentioned in the experimental section, and were cooled in air, polymer-water mixture and oil after the final rolling. It produced total of 12 plate samples of 11 x 29 mm cross section. Preliminary investigation on the plates showed that no major surface defects like scaling, cracks, bends etc were present on the plates.

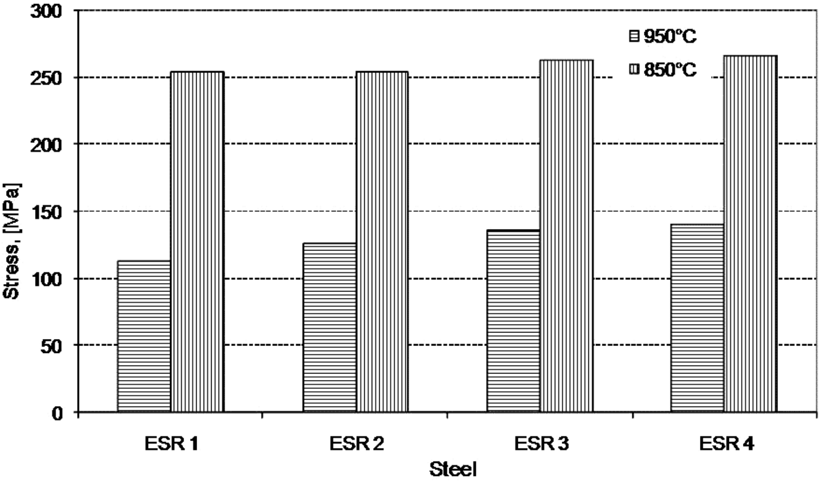


Fig. 11. Rolling torques for first and final pass during hot rolling experiment.

3.3.1 Effect of cooling rate

The tensile strengths, yield strengths and elongations of the hot rolled plates in the three cooling conditions are illustrated in Table 9. At the outset one can notice that in most of the cases the tensile strength and yield strength increase as the severity of cooling increases, best values being obtained with oil-cooled samples. It can also be seen that ductility is marginally improved in the oil-cooled samples. The hardness and impact toughness of the as rolled specimens in the three cooling conditions is shown in Table 10. It can be observed that for all steels, hardness increased as cooling became faster. Air-cooling resulted in the lowest hardness, and the highest hardness was observed in the oil cooled specimens. Among the samples, lowest and highest hardness were measured in ESR1 (base alloy) and ESR3 samples, respectively. Annealing of these samples resulted the decrease in hardness values compared to as rolled condition. It is also seen from table 10 that except of one or two cases, the impact toughness values also increase with increase of cooling rate. Highest impact toughness is observed in oil cooled specimens.

Sample	Air cooled			Polymer-water cooled			Oil cooled		
	UTS (MPa)	Y. S (MPa)	el (%)	UTS (MPa)	Y. S (MPa)	el (%)	UTS (MPa)	Y. S (MPa)	el (%)
ESR 1	1818	1525	8.8	1883	1550	8.1	2030	1615	10.7
ESR 2	1925	1600	9.8	1920	1703	9.5	2062	1721	10.4
ESR 3	1990	1667	8.9	2054	1705	9.6	2214	1750	9.9
ESR 4	1941	1635	9.8	2002	1684	9.3	2181	1715	10.1

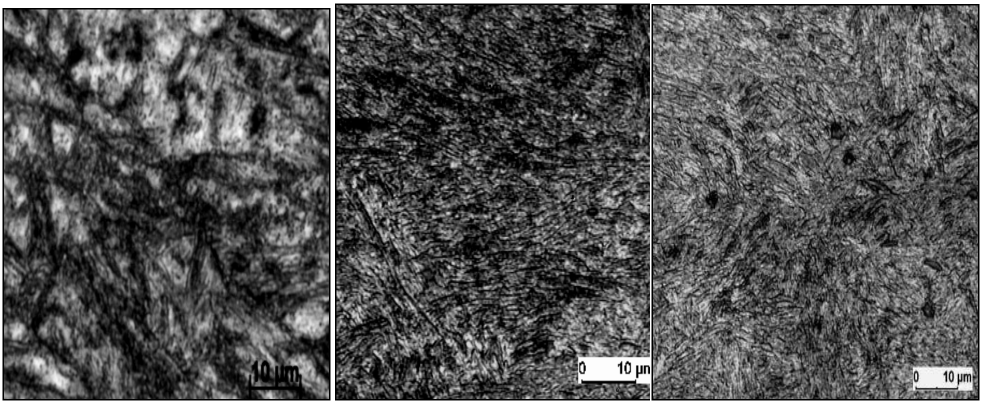
UTS: ultimate tensile strength, Y.S: Yield strength, el: Elongation

Table 9. Tensile properties of TMT plates.

Sample	Hot-rolled, air-cooling		Hot-rolled, polymer cooling		Hot-rolled, oil-cooling	
	Hardness (HRc)	Impact toughness ($kJ.m^{-2}$)	Hardness (HRc)	Impact toughness ($kJ.m^{-2}$)	Hardness (HRc)	Impact toughness ($kJ.m^{-2}$)
ESR 1	44.3	391	45.7	421	48.0	516
ESR 2	48.6	629	48.1	655	49.3	742
ESR 3	48.3	496	51.2	467	52.5	564
ESR 4	48.4	439	50.9	546	51.7	516

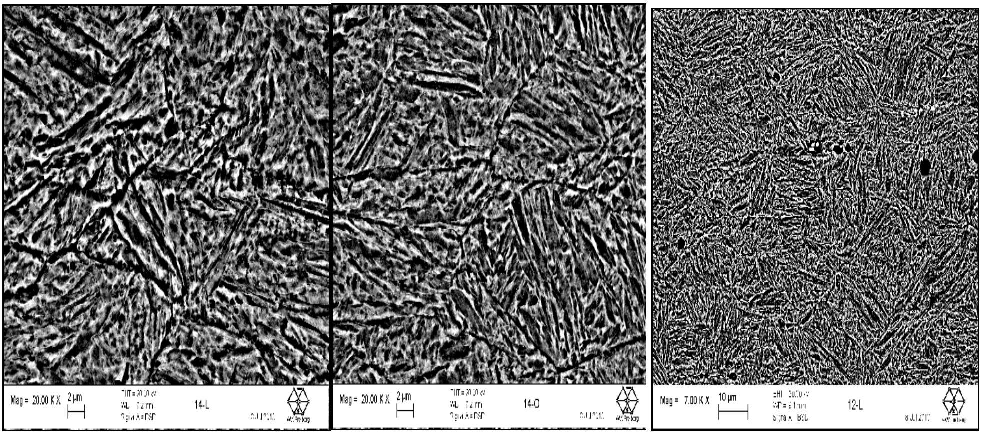
Table 10. Impact strength and hardness of TMT plates.

It can be noticed that mechanical properties of the thermomechanically treated steels are greatly influenced by the quenching medium as in evident from Table 9 and Table 10. The mechanical properties are improved substantially with increase in cooling rate. After thermomechanical treatment the as-cooled plate displays significant increase in yield strength and toughness in compare to as-cast tempered alloys. The best combination of strength and toughness has been observed in oil cooled specimens of ESR2 steel. The optical metallography of one of the ESR2 alloy in three cooling conditions is given in Figure 12. It can be seen that the structure becomes progressively finer as cooling rate become faster. Figure 12 also reveals that in the slow cooling rate the microstructure consists of many more phases. There may be some lath martensites along with austenite and bainite in the matrix. Whereas, oil cooled plates consists of predominantly finer lath martensite structures. The SEM micrographs of ESR2 alloy are also shown in Figure 13. It can be seen that the microstructures of the specimens consist of lath martensites and more uniformity and homogeneity is observed in the specimens those are cooled in faster rate. Apparently it is also seen that the microstructures in oil cooled samples predominantly consist of finer lath martenisites. The TEM micrographs of ESR2 sample in air cooled and oil cooled samples are shown in Figure 14. The TEM micrograph reveals that air cooled sample consist of lath martensite, bainite and some retained austenites. In oil cooled sample the microstructure are mainly consist of lath martensites. The martesite interlath spacing in oil cooled is observed about 200-300 nm whereas, it is 300-400 nm in the air cooled sample. It can be noticed from Figure 15 that the specimens cooled at slower cooling rates showed segregation of carbon, which indicates the presence of retained austenite and bainite (Maity et al., 2008). It is also inline with the predicted phase transformation information as shown in Figure 9. According to CCT diagrams shown in Figure 9, all investigated alloys had enough hardenability to get full martensitic microstructure in cross-section of tested samples after oil quenching (cooling rate normally greater than 15K/s) and mixed microstructures in air cooing (cooling rate less than 1.5 K/s).



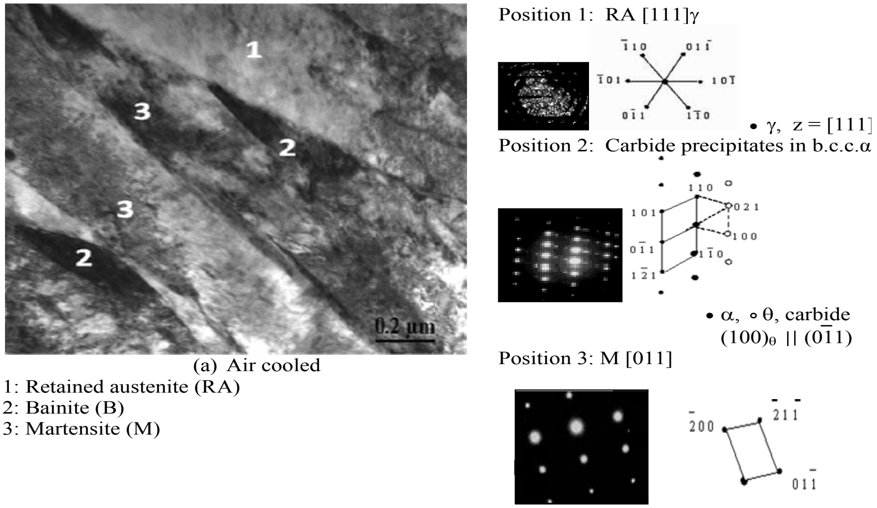
(a) Air cooling (b)Polymer+water cooling (c) Oil cooling

Fig. 12. Optical Micrographs of the TMT plates of ESR2 specimens cooled in different cooling medium.

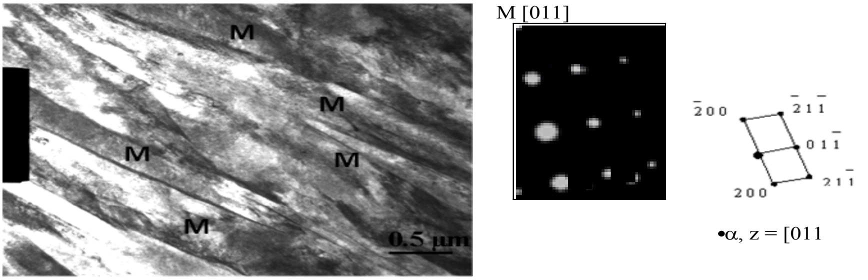


(a) Air cooled (b)Polymer - water cooled (c) Oil cooled

Fig. 13. SEM Micrographs of the TMT plates of ESR2 alloy cooled in different medium.



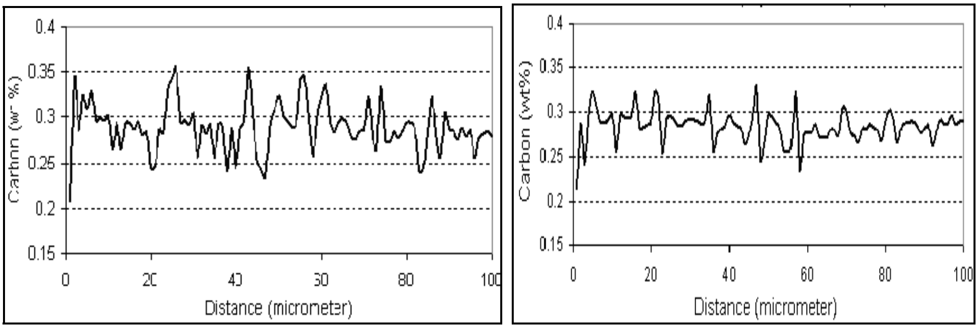
(a) Air cooled
1: Retained austenite (RA)
2: Bainite (B)
3: Martensite (M)



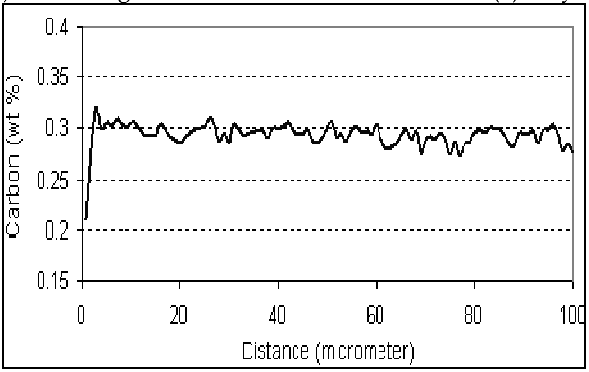
(b) Oil cooled

Fig. 14. TEM micrographs and diffraction pattern of TMT plates of ESR2 specimens cooled in air and oil showing:(a) the presence of martensite (M), retained austenite (RA) and bainite (B) in air cooled sample, and (b) predominantly martensite (M) in oil cooled specimens.

Evidences for phase identification are collected through EPMA and TEM studies. If during transformation, the temperature is high enough, carbon gets enough time to diffuse ahead of the transformation front. Higher carbon regions should be found at the boundaries of pockets of laths and retained austenite or in between upper bainite laths. Samples cooled in different quenching medium (air, oil and polymer) were subjected to EPMA analysis to reveal the segregation patterns, the results of which are presented in Figure 15 (Maity et al., 2008). One can clearly see that segregation of carbon decreases as the severity of quench increases. In the air-cooled sample, one can see peaks in carbon content nearly at regular intervals of about 15-25 µm. This may be due to retained austenite at the boundaries of packets of laths. The individual laths being less than a micron wide, inter lath segregation cannot be resolved in EPMA. In the specimen cooled at the intermediate quench rate (polymer-water 1:1.5 mixture), the extent of segregation is less indicating carbon had less time to diffuse. The interval between the peaks is also slightly less, indicating the size of packets of laths are smaller. This is in tune with the optical/SEM micrographs. The oil-cooled samples show very little long range segregation. Here the severity of quench has been high enough, and carbon could not diffuse out and austenite could not be retained. The improvement of mechanical properties in oil cooled specimens possibly due to the change of the morphology of the microstructural changes due to the change of cooling rate.



(a)Air cooling (b) Polymer cooling



(c)Oil cooling

Fig. 15. Electron probe microanalysis of the distribution of carbon in the central zone of the hot rolled steel under different cooling conditions.

3.3.2 Effect of nickel and other alloying elements

As discussed, in ESR2, ESR3 and ESR4 steel deliberately 1% to 3% nickel are added to the base composition of ESR1 alloy. It can be noticed from Table 9 and Table 10 that with increase of nickel content in TMT plates in three different cooling conditions, the tensile strength, and yield strength are progressively increased up to 0% to 2% with increase of nickel content. In 3% nickel steel the tensile properties are in reverse in trend. Highest tensile strength of 2214 MPa and yield strength 1750 MPa were obtained with 2% nickel in ESR3 steel. Other steels have also displayed tensile strength values of about 2000 MPa in oil cooled plates. As these steel has ductility values varies from 8-10%, so the change of elongation is not so prominent. The room temperature impact toughness of the rolled samples are shown in Table 10. It is interesting to see that the impact toughness in the most of the cases increases from 0% to 1% nickel steel and further increase of nickel content reduces the impact toughness. ESR1 (base alloy) displayed the lowest impact toughness and

ESR2 with 1% nickel gave the highest. All nickel containing steels showed higher impact toughness compared to the base alloy. This was the trend in the as-cast tempered steels too. Lower additions (up to 1% Ni) could give better toughness without sacrificing yield strength. In the alloys, all nickel containing as-cooled plate results better combination of tensile properties and toughness compare to base alloy. In the nickel alloys, one can also notice that the best combination of yield strength and toughness are obtained in the alloy containing 1% nickel (ESR2). Higher nickel contents had improved the yield strength but results comparatively lower impact toughness.

Generally nickel enlarges the γ phase region in Fe-C phase diagram, therefore it enables lower austenitizing temperature of steel, which can promote refinement of structure. Decrease in the martensite packet diameter, similar to the decrease of the grain size, improves the strength as well as the toughness of steel (Tomita & Okabayashi, 1986). Nickel can also influence increasing the stability of retained austenite (Rao & Thomas, 1980; Sarikaya et al., 1983) and the morphology of cementite precipitation at tempering (Peters, 1989). It is indeed happened in case of nickel steels. The SEM micrographs as shown Figure 4 reveal that the laths in martensite matrix are progressively finer with the increase of nickel content. Most of the cases, nickel increases toughness, but it is effective when its amount is controlled in the steel containing 1% Mn. Nickel increases the resistance to cleavage fracture of iron and decrease a ductile-to-brittle transition temperature (Bhole et al., 2006). It is also reported that increase of the nickel content, the grain boundary ferrite (GBF) and acicular ferrite (AF) decreases and as a result of the reduction of both AF and GBF, the impact toughness decreases (Bhole, 2006). It is also reported that when in C-Mn steel containing 1.4% Mn, the toughness drops if nickel content exceeds 2.25%. Kim et al. found that the combined presence of Ni and Mo decreases the volume fraction of GBF (Kim et al., 2000). This may be due to the improved wettability of the Ni as binder on the carbide phase due to the addition of Mo. Improved wettability results the decrease in micro-structural defects and an increase in the interphase bond strength and phase uniformity. The increase in nickel results in the reduction of impact toughness. It may be due to the significant reduction of the volume fraction of acicular ferrite or grain boundary ferrite. The optical micrograph (Figure 3) reveals the presence of substantial amount of acicular ferrite in ESR2 steel and trace amount in ESR3, but this phase could not be identified in ESR4 alloy. This may be one of the reason for the increase of impact toughness in ESR2 containing 1% nickel. It suggests that at the content of about 1% of nickel will have significant influence on notch toughness in these types of steels.

Nickel being an austenite stabilizer leads to retained austenite on one hand, and on the other hand it increases toughness, especially when the nickel content is low at about 1%. Nickel leads to grain refinement and improve toughness when it is used in optimum amount. As a result, all the alloys containing nickel showed high impact toughness after TMT and the one with 1% nickel shows a best combination of strength and toughness. On the other hand, hot rolling at temperatures just above AC_1, has been shown to be feasible and effective method to roll such high strength steel. It is also possible that ESR can be used effectively to reduce the major casting defects and can control the macro- and micro-segregation.

3.4 General discussion

The objective of the present work rose out of the requirement of developing an ultra high strength steel with a yield strength in excess of 1650 MPa, with a minimum elongation of 9-10%. This material is being developed primarily for application in the area of pressure vessels in aerospace vehicles. In such high strength alloys one needs to employ all modes of strengthening. There are heat treatable alloys where strength is obtained from finer martensites with additional precipitation hardening. The approach in the present work was to adjust the chemistry and the production process to obtain a optimum morphology in the microstructure in the as-cast steels. Further improvement was carried out by a optimised thermomechanical treatment with controlled cooling. These two aspects formed two parts of this work.

The primary alloying elements in this 0.3%C steel are chromium, molybdenum and vanadium, which are all carbide/carbonitride formers. At temperatures below about 500°C almost all carbon is in various precipitates at equilibrium. To obtain optimum properties one needs to balance the precipitation process between high and low temperatures. Precipitates at soaking temperatures are needed to limit austenite grain growth and modify the deformation processes. Management of precipitate size is extremely important here. Precipitation at lower temperatures, especially of carbides of chromium and molybdenum, can be coherent/semi-coherent and leads to large strength development during cooling and tempering. The alloys could only be developed because of ESR processing. Normally, most of the strengthening mechanisms lead to loss in ductility. The ability to ensure removal of all large and medium sized inclusions from near directional solidification under a high temperature gradient from a small liquid metal pool during the ESR process increases ductility, toughness and workability. Most of the defects like micro-and macro-segregations, micro porosities and looseness associated with solidification are nearly absent in ESR processed materials. Nickel containing alloys showed finer grain sizes compare to the basic steel. Addition of 1%Ni gave lower yield strength in combination with very high impact toughness. Some improvement in strength was indeed obtained at higher nickel contents. One reason for this behaviour may be the retention of austenite promoted by nickel. Softer austenite distributed in small amounts interferes the crack propagation and improves the impact toughness but decreases the strength at 1%Ni. Solid solution strengthening probably becomes important at higher percentages, more than compensating for loss due to larger proportion of retained austenite. These are the issues which need further exploration.

The thermomechanical treatment adopted, wherein the samples are rolled in the two phase region finishing the deformation just above AC_1, seems to have improved the properties enormously. This strategy permitted rolling to be done with the existing equipment, and to retain some work hardening effect to increase the strength. Controlled cooling allows one to optimise the final microstructure. It has been demonstrated that it is possible to obtain the optimum combination of strength and toughness by an appropriate control of processing parameters such as reheat temperature, deformation temperature, deformation per pass, cooling rate, etc. Cooling rate has large influence on the properties. Air-cooling generally gave lower strengths and oil cooling the highest. Interestingly oil-cooling also gave higher elongation, indicating the effect of auto-tempering. The microstructure in case of oil cooling seems to largely consist of finer lath martensite. At air cooling, there were clear evidences of retained austenite, bainite and martensite. It was also noticed that strength values increase with the increase in cooling rate and the highest yield strength were obtained in oil-cooled samples. Steels for aerospace and aircraft applications, need to possess ultrahigh strength coupled with high toughness to ensure high reliability. The ingots produced in this study are smaller size, however it should be brought to a practice of production of relevant level.

4. Conclusions

1. ESR processed ingots has low inclusion content and good microscopic homogeneity.
2. The base alloy consists of predominantly lath martensite microstructure, having lath sizes in the range of 550-700 nm. It contains complex carbonitrides precipitates of vanadium, chromium and molybdenum, of 25-70 nm size. The alloy displays a yield strength of about 1400 MPa, elongation of 11% and impact strength of 300 kJ/m^2.
3. The addition of 1 to 3 % nickel to the base alloy improves most of the mechanical properties. The yield strength of 1% nickel alloy is around 1500 MPa. The alloy containing 3% nickel results a yield strength value of 1542 MPa.
4. The process parameters for thermomechanical treatment were optimised based on model calculations and preliminary experiments. The treatment involved pre-rolling at 1200°C, followed by soaking at 1200°C and rolling in two passes starting from 950°C and 850°C respectively.
5. The thermomechanical treatment applied in the two phase region and finishing at just above AC$_1$, seems to improve the mechanical properties enormously. This strategy permits to roll this high strength steel with the existing equipment, and also helps to retain work hardening to obtain yield strength in excess of 1700 MPa in some alloys.
6. After thermomechnical treatment all the four alloys showed UTS values in the range of 1800-2200 MPa and yield strength in excess of 1700 MPa.
7. The increase of nickel content up to 1% results in increase of toughness in both as-cast tempered alloys and TMT plates. However, further increase of nickel did not beneficial in this composition of alloys. The best combination of tensile strength, yield strength, elongation and toughness are observed in 1% nickel alloy and may be the optimum composition in all alloys.
8. It can be noticed that cooling rate has large influence on the microstructure and thereby on the mechanical properties of the sample of thermomechanical treatment. It is found that the air cooled sample consists of martensite, bainite and retained austenite. The oil cooled sample consists of predominantly finer lath martensite. The air cooled sample results in low strengths compare to oil cooled plate.

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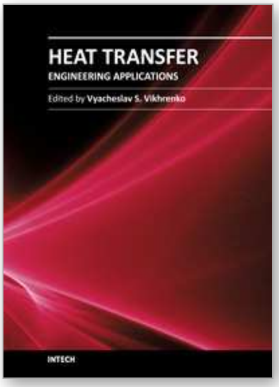

Heat Transfer - Engineering Applications

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Heat transfer is involved in numerous industrial technologies. This interdisciplinary book comprises 16 chapters dealing with combined action of heat transfer and concomitant processes. Five chapters of its first section discuss heat effects due to laser, ion and plasma-solid interaction. In eight chapters of the second section engineering applications of heat conduction equations to the curing reaction kinetics in manufacturing process, their combination with mass transport or ohmic and dielectric losses, heat conduction in metallic porous media and power cables are considered. Analysis of the safety of mine hoist under influence of heat produced by mechanical friction, heat transfer in boilers and internal combustion engine chambers, management for ultrahigh strength steel manufacturing are described in this section as well. Three chapters of the last third section are devoted to air cooling of electronic devices.

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