Thermo-Mechanical Processing of Armor Steel Plates

Taher El-Bitar, Maha El-Meligy, Eman El-Shenawy, Almosilhy Almosilhy, Nader Dawood

Abstract—The steel contains 0.3% C and 0.004% B, beside Mn, Cr, Mo, and Ni. The alloy was processed by using 20-ton capacity electric arc furnace (EAF), and then refined by ladle furnace (LF). Liquid steel was cast as rectangular ingots. Dilatation test showed the critical transformation temperatures Ac_1, Ac_3, M_s and M_f as 716, 835, 356, and 218 °C. The ingots were austenitized and soaked and then rough rolled to thin slabs with 80 mm thickness. The thin slabs were then reheated and soaked for finish rolling to 6.0 mm thickness plates. During the rough rolling, the roll force increases as a result of rolling at temperatures less than recrystallization temperature. However, during finish rolling, the steel reflects initially continuous static recrystallization after which it shows strain hardening due to fall of temperature. It was concluded that, the steel plates were successfully heat treated by quenching-tempering at 250 °C for 20 min.

Keywords—Armor steel, austenitizing, critical transformation temperatures, dilatation curve, martensite, quenching, rough and finish rolling processes, soaking, tempering, thermo-mechanical processing.

I. INTRODUCTION

TROOP carriers are safe armored vehicles for transportation of troops and can withstand against ballistic shooting from the terrorist side. Ballistic resistance of steel plates is obtained as a result of combination between many mechanical properties. None of the properties alone are self-sufficient to predict correctly the ballistic behavior. An optimum combination of hardness, strength, and toughness is essential for good ballistic performance, which could be achieved by proper heat treatment cycle [1].

Much effort is being devoted on the development of armor materials. Ultra high strength (UHS) is the most extensively used metallic armor today. They possess a unique combination of high strength, high hardness with good toughness, and weldability. Moreover, the ease of processing and heat treatment makes them attractive for ballistic resistance applications.

Martensite phase provides the highest level of strength in such steels. However, its creation is associated with internal stresses and consequently it is rarely used in an untempered

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condition. Tempering increases both the ductility and toughness, which are essential for enhancing impact energy absorption. Tempered martensite also provides best dynamic strength in steel [2], [3].

Jena et al. [1] determined the critical transformation temperature (Ac_1 and Ac_3) for high strength armor steel, by differential thermal analysis (DTA) at 1400 °C under vacuum at a heating rate of 10 °C/min. It was focused on correlating microstructure, strength, hardness, ductility, and stress state of the material with the ballistic resistance of armor steel [4], [5]. Ballistic performance of steel generally increases with the increase in hardness up to certain value, beyond which it decreases due to the onset of failure mechanisms such as shear plugging and cracking [6]. Adiabatic shear bands (ASB) may also form when plates are subjected to high rates of deformation such as high velocity projectile impacts. The formation of hot spots and cracks during ASB propagation can lead to failure of the material [7].

The present article is dealing with processing and thermo-mechanical treating of armor steel 6 mm plates for troop carriers.

II. MATERIALS AND EXPERIMENTAL WORK

The steel alloy was processed by melting in 20-ton EAF, and followed by LF treatment for secondary refining. Finally, the liquid steel was treated in the vacuum unit to get rid of the majority of the dissolved gases. The liquid steel was then cast in metallic permanent molds with rectangular cross section (360x1000 mm). Each cast ingot was weighting 5.5 to 6.0 ton. Table I represents the final chemical composition of the processed steel.

The ingots were then subjected to stress relief at a temperature less than Ac_1 (650 °C). The ingots were then furnace cooled to 300 °C followed by air cooling. Fig. 1 shows a schematic presentation of the stress relief cycle.

Polymorphic and allotropic transformations temperatures of the steel were determined by using the thermo-mechanical simulator (Gleeble 3500) using heating up to 1100 °C at a rate 0.45 °C/s and free cooling with a rate – 1.0 °C/s. These polymorphic temperatures are phase change temperatures (Ac_1 & Ac_3) and allotropic transformations temperatures (M_s, M_f, B_s, B_f) and recrystallization temperature (T_r).

The steel ingots were hot rolled in two stages. The first one was the rough rolling from 360 to 80 mm thickness and 1500 mm width thin slabs. The second stage is the finish rolling from 80 to 6.0 mm thickness plates and sizing to 6 meters long.

TABLE I
FINAL CHEMICAL COMPOSITION OF THE PROCESSED STEEL

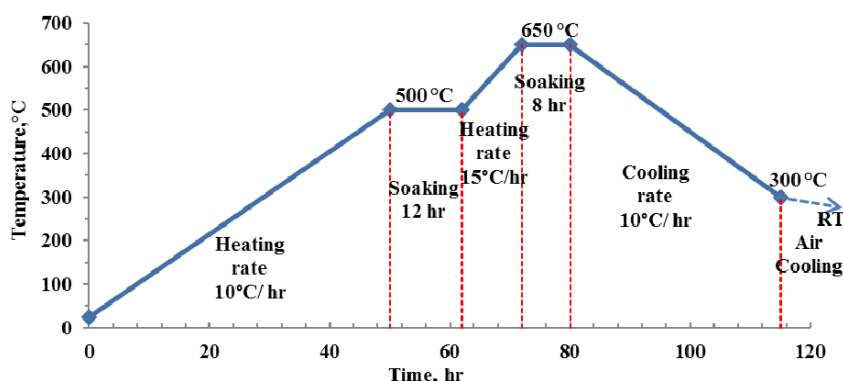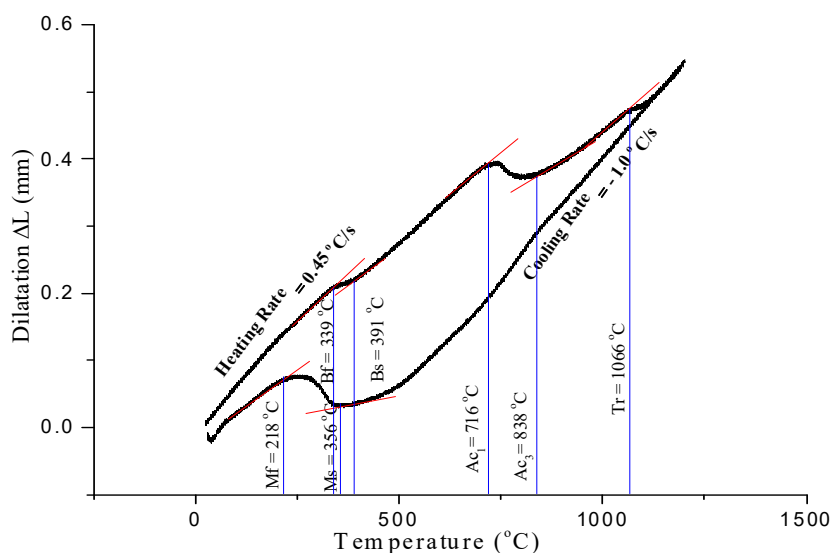| Element	C	Si	Mn	P	S	Cr	Mo	Ni	Al	Cu	B
[wt.%]	0.3	0.24	1.11	0.010	0.009	0.93	0.62	1.76	0.034	0.07	0.004

Fig. 1 Schematic presentation of the stress relief cycle

Fig. 2 Dilation accompanying heating and free cooling cycle

The hot rolled plates were then subjected to subsequent heat treatment cycles. One of the cycles was consisting of austenitizing at 910 ºC, and followed by a rapid water quenching. A tempering step was done on the quenched plates at 250 ºC as well as on hot rolled plates at 300 °C for 20 and 30 min.

TABLE II
CRITICAL TEMPERATURES ABSTRACTED FROM THE DILATION CURVES

Critical Temperature	M_f	M_s	B_f	B_s	Ac_1	Ac_3	T_r
Value, [°C]	218	356	339	391	716	835	1066

III. RESULTS AND DISCUSSIONS

Dilatations accompanying heating and free cooling were graphically presented in Fig. 2 to determine the polymorphic, and allotropic phase transformations temperatures.

The critical temperatures were abstracted at the inflection points located on the heating as well as cooling curves. Table II [8] shows the abstracted critical temperatures.

Cast ingots were mechanically surface conditioned before the rough rolling stage. The ingots were preheated (austenitized) then holed for soaking. Soaking is beneficial to ensure temperature homogenization inside the ingot. Furthermore, soaking is dealing with dissolution of the boron carbide (B_4C) precipitates to facilitate their homogenous distribution during subsequent hot rolling. This would lead to fine precipitates which would be reflected positively on the mechanical properties [9].

For a purpose of estimating the austenitizing temperature, the melting temperature T_M of the steel is essentially calculated. T_M can be calculated as a function of the steel chemical composition, using (1) [10];

$$T_M = 1537 - 88 \,(\% \text{ C}) - 25 \,(\% \text{ S}) - 5 \,(\% \text{ Cu}) - 8 \,(\% \text{ Si}) - 5 \,(\% \text{ Mn}) - 3.6 \,(\%\text{Al}). \tag{1}$$
$$- 2 \,(\%\text{Mo}) - 4 \,(\% \text{ Ni}) - 1.5 \,(\% \text{ Cr}) - 18 \,(\% \text{ Ti}) - 2 \,(\% \text{ V}) - 30 \,(\% \text{ P})$$
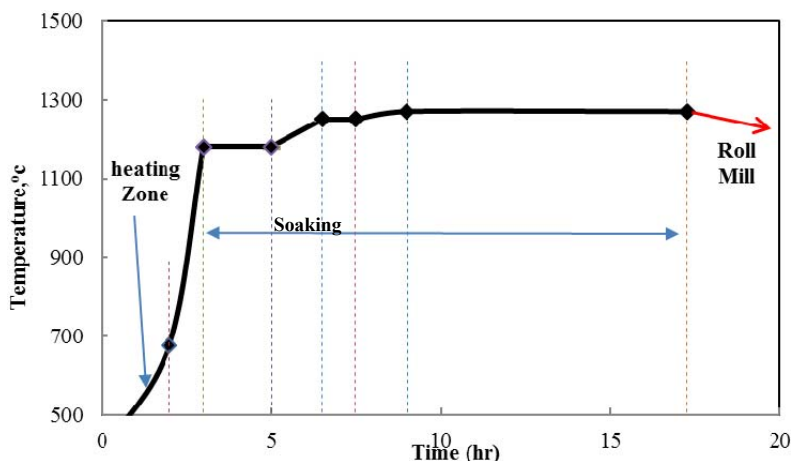

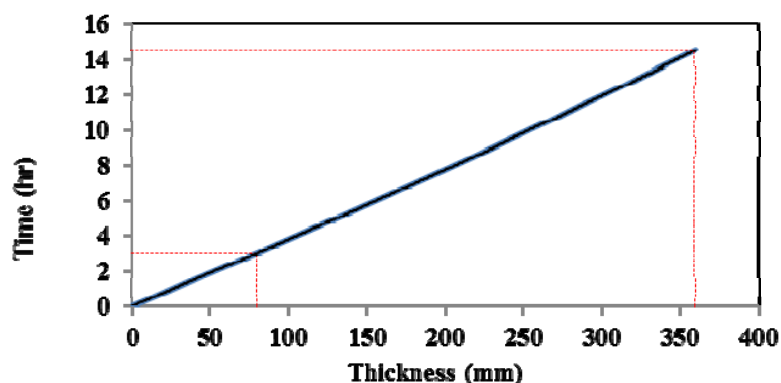Fig. 3 Heating (austenitizing) and soaking cycles for 360 mm thickness ingot

Fig. 4 Soaking time – ingot/slab thickness relationship

On the other hand, soaking time is also important, where longer soaking time may lead to burning (thick surface scale formation) and consequently material losses and subsequent rolling surface defects. Furthermore, long soaking times may lead also to austenite grain coarsening which deteriorates the final mechanical properties. A beneficial relationship was estimated during the present investigation between the soaking time (tsoak) and ingot or slab thickness. The relationship is presented as a 2nd order polynomial (2) and is graphically presented in Fig. 4;

$$t_{soak}(\text{hr}) = 0.00001(\text{thickness, mm})^2 + 0.036(\text{thickness, mm}) \tag{2}$$

In the rough hot rolling stage, it is recommended to exert with the highest possible amount of thickness reduction/pass to get rid of the as-cast dendritic structure, which would result in a homogenous chemical composition (eliminating micro-segregation) [11]. In the present roughing stage, the total thickness reduction reaches to 78%, while the minimum recommended thickness reduction in roughing stage is ranging 50-60% [12].

The rough rolling stage was carried out through 15 passes. Fig. 5 shows presentation of the roll force and rolling pass temperature at different passes during the roughing stage. It is obvious that the roll force increases initially at the early passes, where temperature became lower than Tr. The roll force then continuously decreases, as a result of the adiabatic heating due to continuous deformation from pass to the other [13].

Fig. 6 shows the microstructure the hot-rolled 80 mm thickness slab. The microstructure contains numerous packets of plate martensite.

For more precise characterization, a high magnification (60000 X) microstructure of the 80-mm thickness slab is showed in Fig. 7. The micrograph shows a colony of plate and little lath martensite. The martensite packets are surrounded by unstable retained austenite. Furthermore, the microstructure contains intra-lath (embedded) carbides. The minimization of brittle twinned plate martensite along with the prevention of instability of the retained austenite envelops is of prime interest to avoid tempered martensite embrittlement and low toughness [14]. Complete dissolution of coarse carbides is also essential during the subsequent finish rolling stage [15].
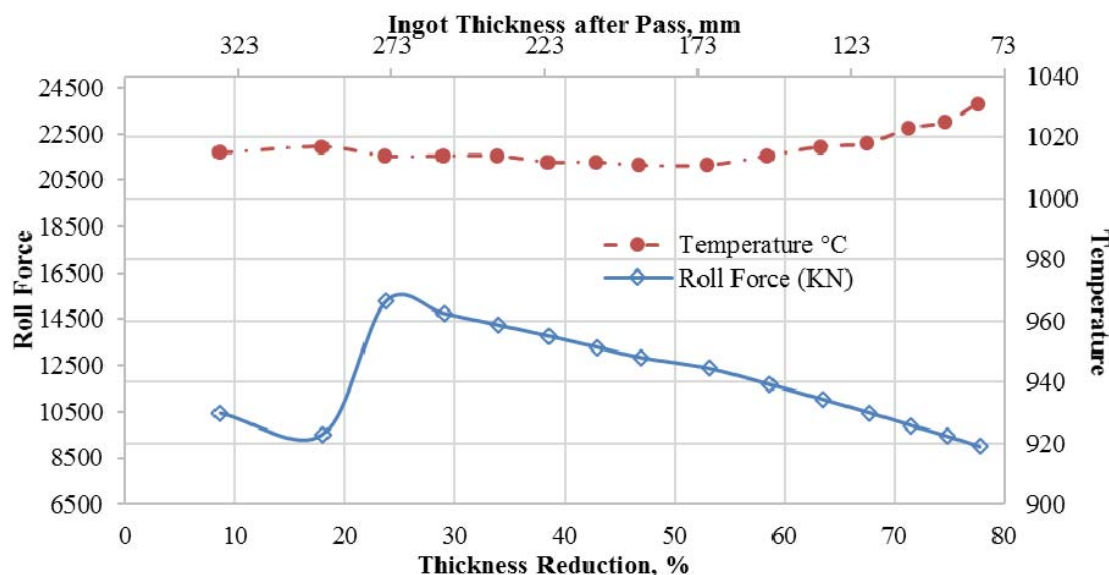
Fig. 5 Roll force and rolling pass temperature at different passes during the roughing stage
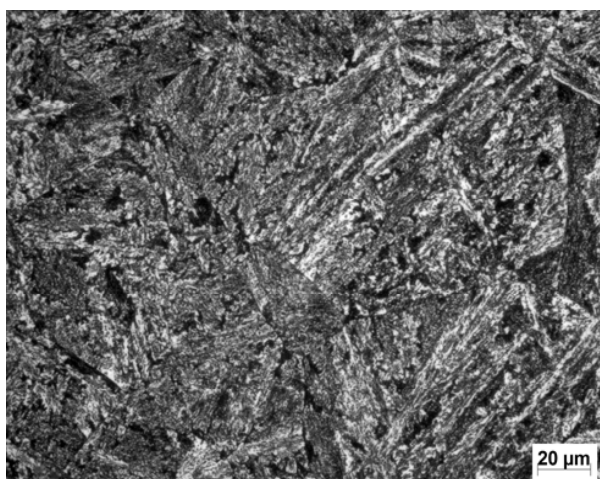

Fig. 6 Microstructure the hot-rolled 80 mm thickness thin slab

Hot rolling from 80 mm thickness to 6 mm is a complementary stage after the roughing stage. Fine grained microstructure can be obtained by increasing the thickness reduction and/or rolling speed at temperatures lower than the recrystallization temperature to activate dynamic recrystallization [13]. Fig. 8 shows presentation of the roll force and rolling temperature at different thickness and amounts of reduction during finish rolling stage from 80 mm slab to 6.0 mm thickness plates on 10 passes. The temperatures of the last 4-passes are lower than the recrystallization temperature (<1066 °C). The roll force behaves as a plateau up to pass No. 6 at 88% total thickness reduction, after which the rolling force begins to increase followed by a sudden continuous decrease. The initial plateau behavior of the roll force reflects continuous static recrystallization of the steel due to deformation at a higher temperature than Tr. Fall of passes temperature (< 1066 °C) at

pass no. 7 causes the sudden increase of the roll force reflecting strain hardening phenomena and consequently leading to strain accumulation (elongated grains). Further strain accumulation in combination with further fall of rolling temperature favors dynamic softening resulting in fine austenite grains. [16], [17]. On cooling and/or quenching of steel plates, fine grained austenite would create thin lath martensite packets with better toughness than that for plate martensite [14], [18].

The final hot rolled plats (6.0 mm thickness) are naturally cooled due to the limited thickness with respect to the plate surface area. Quick cooling of plates is accompanied by hardening effect and results in deficiency of elongation with high ultimate tensile strength. Table III presents the mechanical properties of as-rolled 6.0 mm thickness plates.

TABLE III
MECHANICAL PROPERTIES OF AS-ROLLED 6.0 MM THICKNESS PLATES

Brinell hardness	Ultimate strength	Proof strength	Elongation
[HB]	[MPa]	[MPa]	[%]
474	1838	858	3.75

It is clearly noticed that the yield strength, as well as elongation, is too low, and consequently, there is a need for subsequent heat treatment cycles to compensate the deficiency of the mechanical properties to enhance ballistic resistance of the steel plates. The most convenient treatment for such steel alloy is the quenching – tempering technique [1]. Water was used as a quenching medium. More details about the successful heat treatment cycles were published in reference [8]. Tempering at 250 °C for 20 min. after quenching reveals a lath martensitic structure with a small fraction of rosette like boron carbides (B_4C) between the martensite packets as shown in both Figs. 9 and 10. The process enhances elongation from 3% to 11.35%, without deterioration of either strength or impact toughness even at – 40 °C [8].
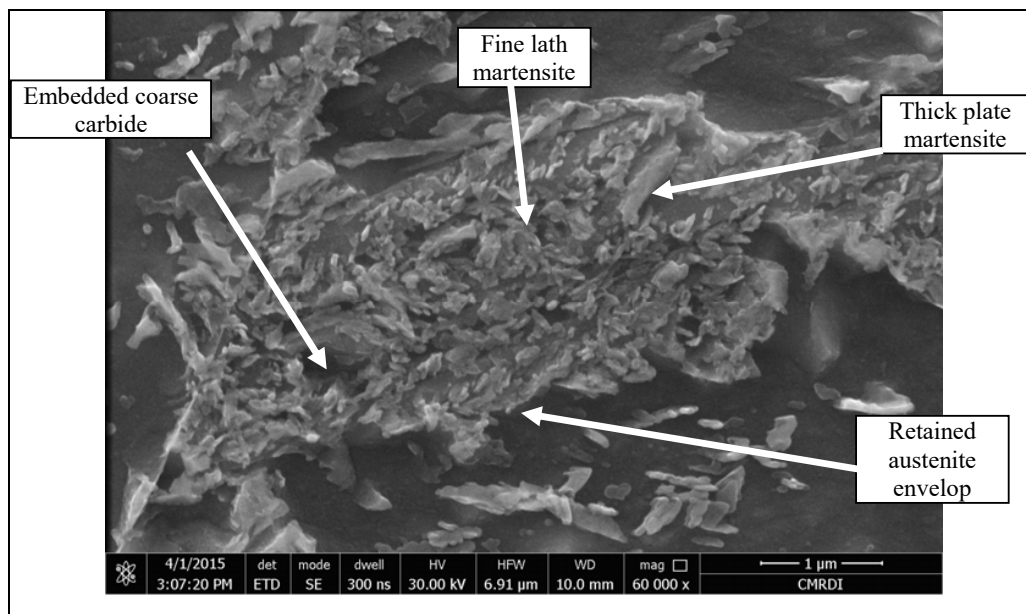
Fig. 7 A martensite colony at the 80-mm rough rolled thin slabs (60000 X)
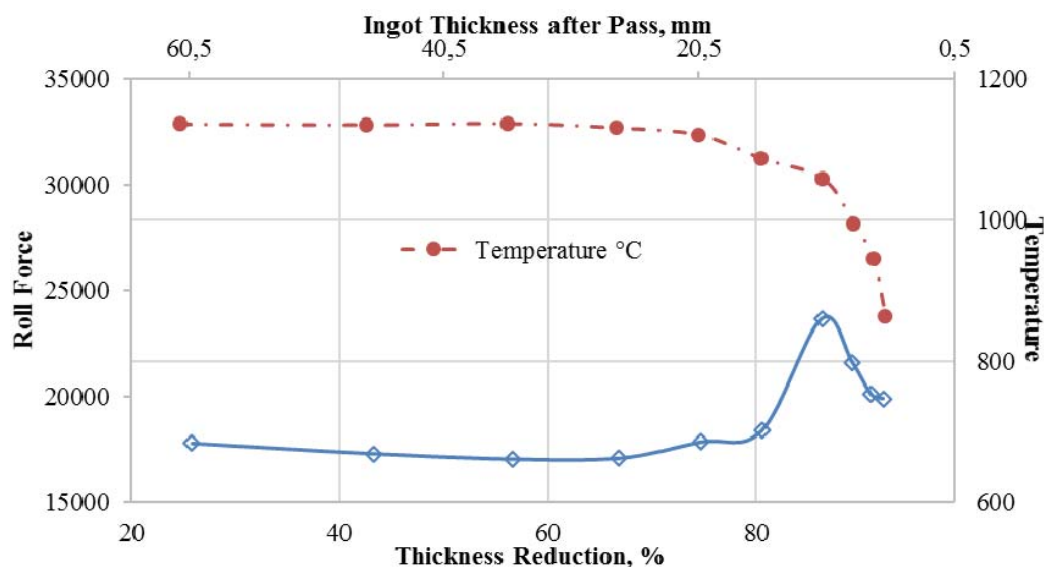

Fig. 8 Roll force and rolling temperature at different thickness and amounts of reduction during finish rolling stage

Alternately, tempering of the as-rolled plates at 300 °C for 20-30 minutes shows good hardness and tensile properties as well as excellent impact toughness. The microstructure of the as-rolled tempered plates contains a thin ferrite layer on the boundaries surrounding the martensite packets (Fig. 11), and few fractions of cotton wool shape boron carbides (B_4C), as shown in Fig. 12, between the martensite laths [8].

Finally, both of the previous heat treatment cycles, show successful ballistic resistance plates against shooting by three bullets from each side [8].

IV. CONCLUSIONS

1. A steel alloy containing 0.3% C and micro-alloyed with 0.004 boron is successfully processed by melting in EAF and followed by LF treatment and finally treated by evacuation.
2. Cast ingots were subjected to stress relief cycle at 650 °C.
3. Dilation investigation reveals the critical and allotropic martensite transformation temperatures as 716, 835, 218, and 356 °C. The recrystallization temperature Tr was observed as 1066 °C.
4. The austenitizing temperature was calculated as 1200-1270 °C, while soaking time (t_{soak}) was calculated as;

t_{soak}(hr) = 0.00001(thickness, mm)^2 + 0.036(thickness, mm).

5. At the early rolling passes, the steel resists deformation due to low temperature (<Tr). On later passes of roughing stage, adiabatic heating has a pronounced effect on lowering resistance to deformation.

6. During finish rolling stage, steel deformation reveals static recrystallization up to pass no. 6, where it begins to strain hardened due to fall of temperature (< Tr).

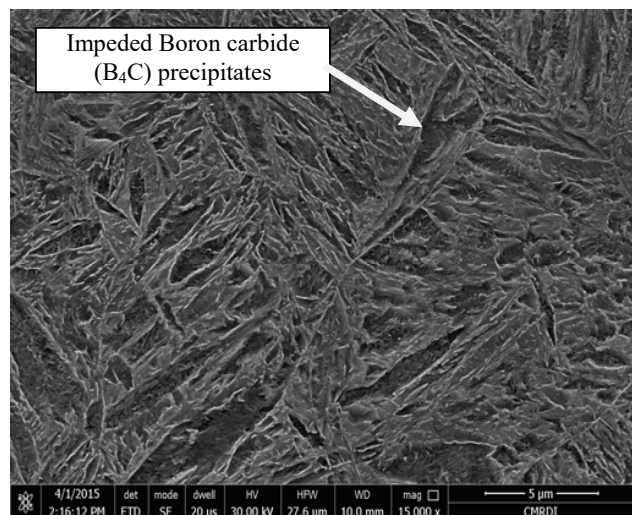7. Heating to 250 °C for 20 min. is the optimum tempering parameter.

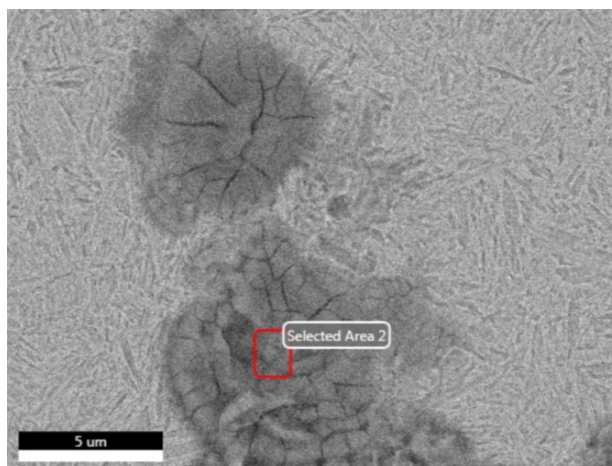Fig. 9 Microstructure of quenched-tempered plates at 250 °C for 20 min (15000X)

Fig. 10 Rosette like carbide precipitate of quenched and tempered plates at 250 for 20 min
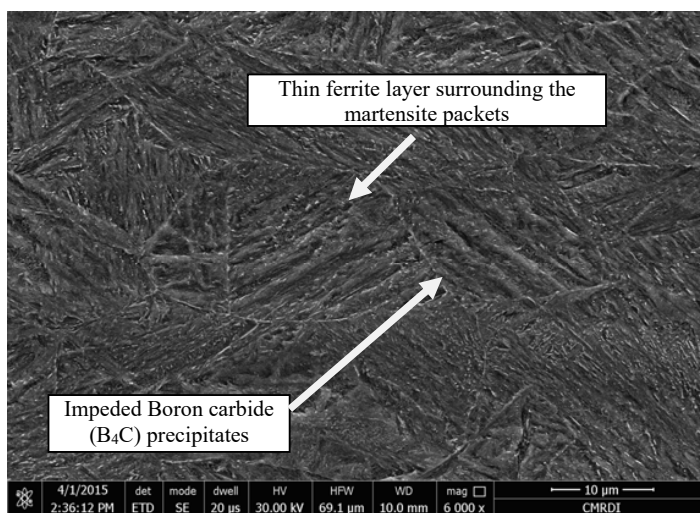

Fig. 11 Microstructure of as-rolled and tempered plates at 300 ºC for 20 minutes (6000 X)
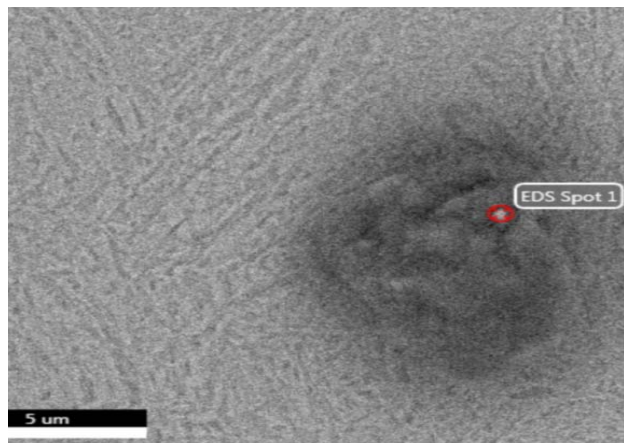
Fig. 12 Cotton wool like carbide precipitate of rolled and tempered
plates at 300 °C for 20min

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