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Microstructural and Fractographical Studies on Quenched and Tempered Armor Steels

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In this study, a boron-alloyed experimental steel has been rolled to form a sheet product, and an austenization treatment has been performed, followed by rapid quenching and then tempering. By austenization, a homogeneous austenite phase is obtained and a martensitic microstructure is formed after rapid quenching. This kind of microstructure exhibits high hardness, but low toughness. Thus, a tempering heat treatment is applied at a proper temperature (200 °C and 600 °C) and various times, in order to provide optimum toughness and to develop an ideal combination of strength and toughness.
After this process, at first the phases formed during the equilibrium state have been studied for the alloy system using ThermoCalc data, and also their amounts have been evaluated. The microstructures acquired after the heat treatments have been examined by light microscopy. The effects of the formed microstructures on mechanical properties are comparatively given by their strength, toughness, and elongation values. Furthermore, the failure mechanism with fractografical examinations has been determined by using a Scanning Electron Microscope (SEM).

The performance of steels depends on the properties associated with their microstructures, that is, on the arrangements, volume fractions, sizes, and morphologies of the various phases constituting a macroscopic section of steel with a given composition in a given processed condition. Because all the phases in steels are crystalline, steel microstructures are made up of various crystals, sometimes as many as three or four different types which are physically blended by solidification, solid-state phase changes, hot deformation, cold deformation, and heat treatment. Each type of microstructure and product is developed into its characteristic property ranges by specific processing routes that control and exploit microstructural changes. Thus processing technologies do not only depend on microstructure, but are also used to tailor final microstructure [1].

Quenched and tempered steels are widely used in automotive, defence, and marine industries. These kind of steels have a martensitic/bainitic structure after rapidly quenching and then gain optimum toughness after a tempering heat treatment. Thus some conditions such as alloy design, quenching, and also tempering are very important to determine the final properties of the product [2].

Steel quenching is defined as 'cooling of steel workpieces at a rate faster than still air'. The cooling rate has to be so fast that austenite mainly transforms within martensite and bainite ranges. In industrial applications, such steels are not convenient due to their low toughness after rapidly quenching. In these cases the optimum mechanical properties, for example high strength, hardness, and also toughness, can be obtained by performing a tempering heat treatment at proper temperature and time conditions [3-6].

Figure 1 summarizes the mechanism of hardness variations in a pure iron-carbon martensite structure during tempering heat treatment at 100-700 °C for 1 hour compared with steels where different carbide forming elements are alloyed. In general, carbon atoms within martensite structure joins with iron atoms after quenching, and then iron carbides (Fe_3C) form during tempering, referred as cementite precipitation. During this kind of precipitation, the hardness of steel decreases as a result

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of transformation from martensite to ferrite structure. However various carbide precipitates such as MC, M_2C, and M_7C_3 can develop with addition of some carbide forming alloying elements, which have a strong affinity to carbon at high tempering temperatures. As a result, an increase in hardness known as secondary hardness can be determined. This characteristic hardness increase is achieved by a complex precipitation hardening. As seen, the formation temperatures of formed carbides increase with increasing carbide stability [7-9].

Secondary hardness precipitates form preferentially upon dislocations and lath boundries in a lath-type martensitic matrix. Maximum value of secondary hardness can be obtained by precipitation of nanometer sized, rod shaped semicoherant M_2C and also disc shaped semicoherant MC at 540-560 °C (the dashed line Mo+Cr+V in Figure 1). On the other hand Cr-enriched M_3C and M_7C_3 precipitate during high temperature tempering of steels. While the number density of M_2C is two times of MC carbides at the initial stage of overtempering and mainly determines the basis of matrix hardness, this kind of precipitate develops rapidly and its number density decreases analogous to its low thermodynamical stability during tempering. This change is shown for a high speed steel sample tempered at 600 °C in Figure 2 with images taken by Field Ion Microscope (FIM). Finally the relationship between the amount of dispersed fine carbide in matrix and hardening potential is given in Figure 3. As observed, existing linear relationship has a compatibility on several steel based materials [10, 11].

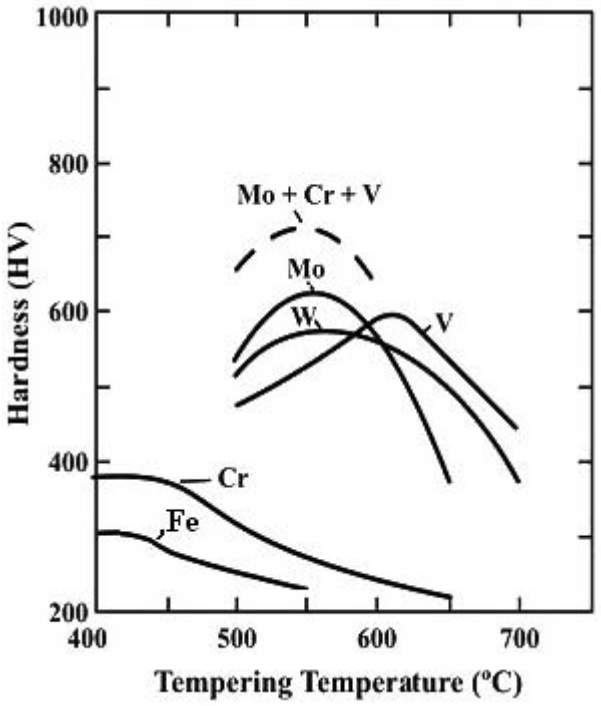


Figure 1. Tempering characteristics of carbides formed by iron and carbide forming elements

Chemical composition (wt.-%)									
C	Mn	Si	P	S	Cr	Mo	(Nb+V+Ti)	Co	B
0.25	0.10	0.10	0.016	0.017	1.5	0.5	0.20	4.5	0.003

Table 1. Chemical composition (wt.-%) of the alloy used in the study

Experimental

In this study, microstructural and fractographical changes in a quenched and tempered steel with a given composition have been investigated by performing different heat treatments. The alloy composition used in experiment is given in Table 1. Roles and details of alloying elements have not been examined in this study, but boron was added at ppm level to promote the bainitic transformation, whereas chromium and molibdenum were selected for precipitation hardening, and also niobium, vanadium, titanium were added as microalloying requirements. Cobalt was prefered for developing a good matrix toughness. The amount of carbon was kept low to provide a good weldability.

Steel was cast using a medium frequency furnace, a model of AEG with a capacity of 550 kg. After casting, plates sized 500 × 550 × 55 mm were rolled eleven times, and after deformation a sheet with 1500 × 600 × 5 mm size was obtained.

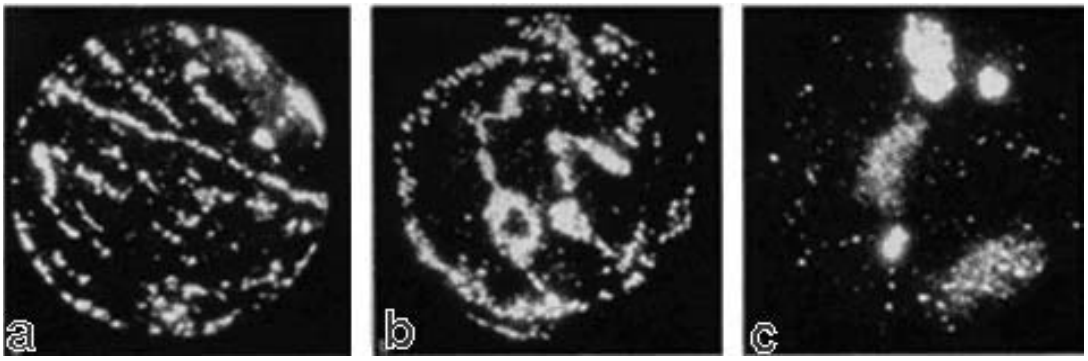


Figure 2. Characterization of typical secondary hardening carbides by using FIM (Field Ion Microscope), coarsening of secondary hardness carbides (light spots). Tempering is performed at 600 °C, a) 100 minutes, b) 300 minutes, c) 3000 minutes

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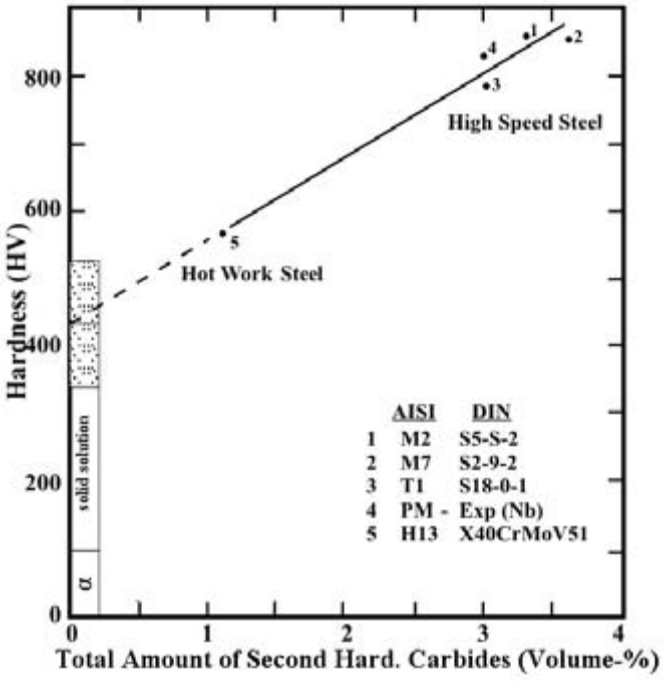


Figure 3. The relationship between the amount of fine-dispersed secondary carbides (vol.-%) and hardening potential of the matrix

Prescribed phases in calculations
liquid δ-Fe (bcc) γ-Fe (fcc) α-Fe (bcc) graphite Fe_3C M_2B MC M_2C M_6C $M_{23}C_6$
Formed phases after solidification
liquid δ-Fe (bcc) γ-Fe (fcc) α-Fe (bcc) M_2B M_7C_3 $M_{23}C_6$ MC

Table 2. ThermoCalc data

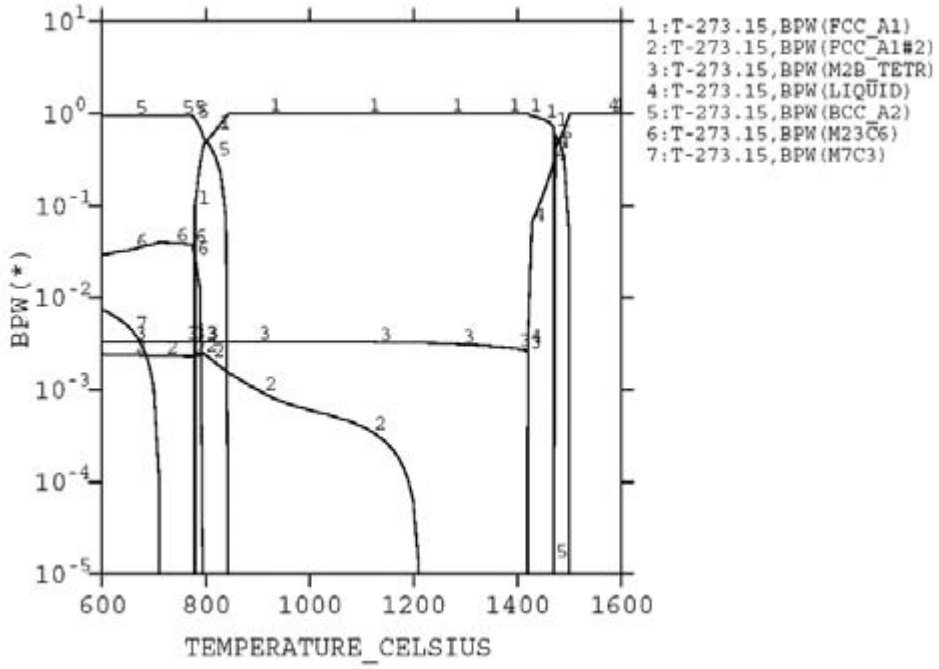


Figure 4. The change of the amount of phases during solidification of the experimental alloy at equilibrium conditions

Solidification Modelling with ThermoCalc

The transformations generally occur for a given temperature and time. For a certain alloy composition, many transformations and also stable phase ranges formed after or before transformations can be obtained by using thermodynamical data. ThermoCalc results have been used to determine the transformations, stable phase ranges, and also the amounts of the phases at equilibrium conditions. None-equilibrium solidification is formed after the rapidly quenching of steels. In this study transformations, formed phases, and also their amounts have been examined under equilibrium and none-quilibrium conditions. Table 2 shows the prescribed phases and formed phases at equilibrium conditions for the alloy used in this experiment. On the other hand Figure 4 represents the relationship between the amount of phase per weight and temperature for the experimental alloy at equilibrium conditions. In this figure, curve numbered 4 refers to the liquid phase of the steel. It is obvious that when the temperature value within the x-axis is considered, liquid phase is stable at 1550 °C and above temperatures, and as a result solidification begins from 100% liquid. On y-axis, 10° corresponds to vol-% 100. By decreasing the temperature below 1550 °C, δ-Fe (bcc) is obtained. In addition to this formation, the austenite phase (fcc) forms as a result of peritectic reaction (liquid + δ → γ) at 1400-1500 °C. Primary M_2B phase forms at approximately 1420 °C, and its amount is about vol-% 0.4 compared to the overall composition. The amount of primary M_2B phase has no discernible change with decreasing temperature. Eutectic secondary MC carbide phase forms from liquid in addition to austenite along interdendritic zones at 1210 °C, and its amount is about vol-% 0.3. This reaction can be shown as liquid → γ + MC. Solid state reactions form after reaching 850 °C, and as a result of this reaction ferrite (fcc) forms. Following this formation, $M_{23}C_6$ forms at 800 °C, and its amount is about vol-% 4, M_7C_3 type carbide phase forms at 700 °C, and its amount is about vol-% 0.8. All transformations are completed at 600 °C, and finally the percent distribution of the phases are 94.5% ferrite, 0.4% $M_2B_{primary}$, 0.3% $MC_{secondary}$, 4% $M_{23}C_{6secondary}$, 0.8% $M_7C_{3secondary}$ at room temperature.

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Heat Treatments and Mechanical Properties

The heat treatments applied in this study consist of austenizing, quenching, and tempering, respectively. Austenization is a kind of heat treatment that gives a single phase of austenit (γ) in steels. After rapidly quenching of steel a martensitic/bainitic microstructure can be formed, and then tempering is applied to gain an optimum strength and toughness combination. Table 3 shows the applied heat treatments and the mechanical properties of samples after heat treatments. After performing austenization and then quenching, a tempering heat treatment was applied on samples at 200 °C to provide the aging of martensite. However a higher temperature such as 600 °C was selected to gain toughness and also secondary hardness to steel.

As can be seen in Table 3, a higher toughness has been obtained with tempering at 600 °C for 2h. This is a result of dissolved martensite compared to tempering at 200 °C.

Microstructural Characterization

Figure 5 represents general microstructures of samples after quenching. As seen, steel has a matrix between martensite and bainite according to the cooling level.

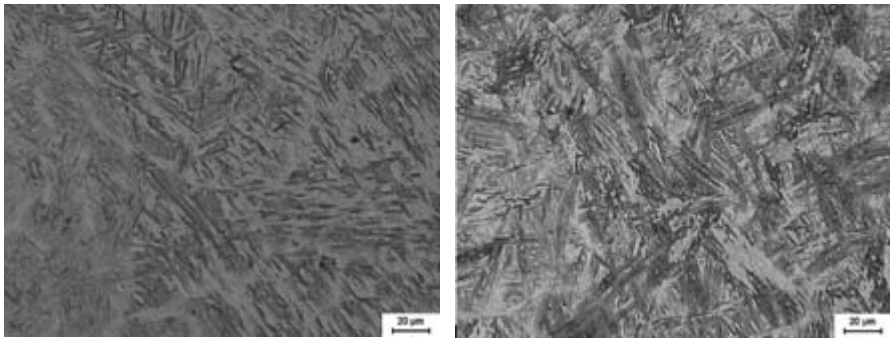
Some microstructural examples of experimental steel, subjected to various heat treatments, are given in Figure 6 by light microscope. All microstructures have a bainitic characteristic. Bright fields refer to the ferrite phase which can be etched easily with etching reagents. On the other hand tiny dark colored points refer to very fine precipitates within the matrix. As a result of rapid solidification, segregation can develop in the matrix. These formations have negative effects on the physical and also mechanical properties of the materials. Figure 7a shows segregation regions in experimental steel subjected to a heat treatment series of 1000 °C/1 h + 200 °C/1 h.

Generally sulphur tends to segregate in steels. Manganese has a good reaction capability with sulphur to form MnS, which is a kind of non-metallic inclusion and effects the behaviour of steel during its rolling and failure. Figure 7b shows an example of primary MC carbide

Mechanical Properties	Heat Treatments			
	1000 °C/½h +200 °C/1h	1000 °C/1h + 600 °C/1h	1000 °C/1h + 600 °C/2h	1000 °C/1h + 600 °C/3h
Tensile Strength (MPa)	1726	1345	1280	1310
Yield Strength (MPa)	1592	1258	1128	1210
Elongation (%)	8.3	6.9	10.1	6.1
Hardness (HV)	414	385	388	301
Impact Toughness* (J/cm²)	36.6	52.6	59.0	50.1

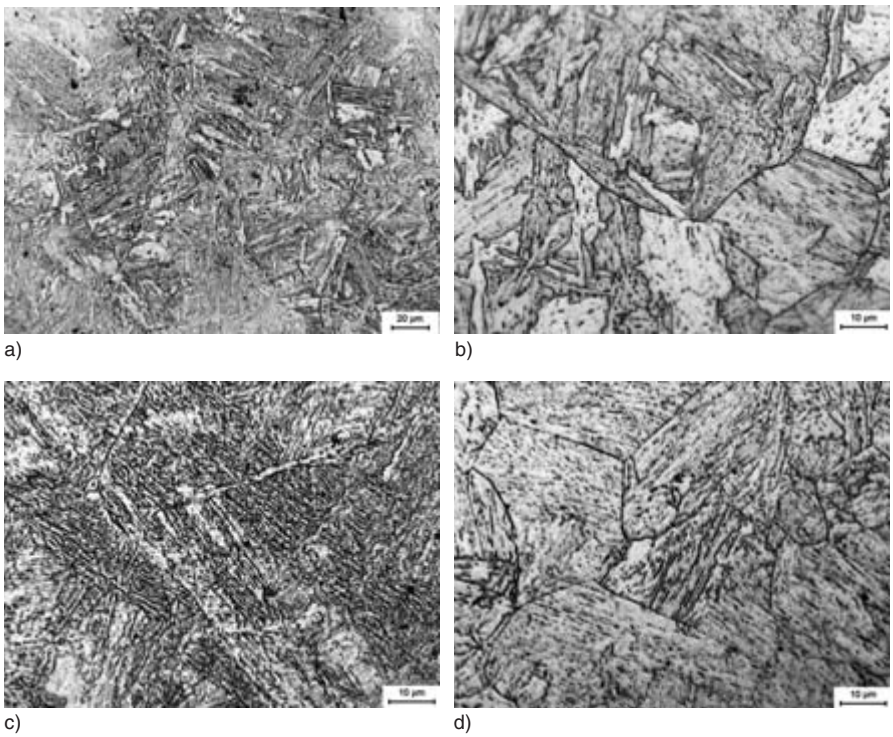
*) Impact tests have been performed at room temperature

Table 3. Applied heat treatments on experimental steel and changes on mechanical properties



a) b)

Figure 5. Microstructures of experimental steel after quenching by light microscope, a) and b) lath-type bainite and martensite, etched with nitale

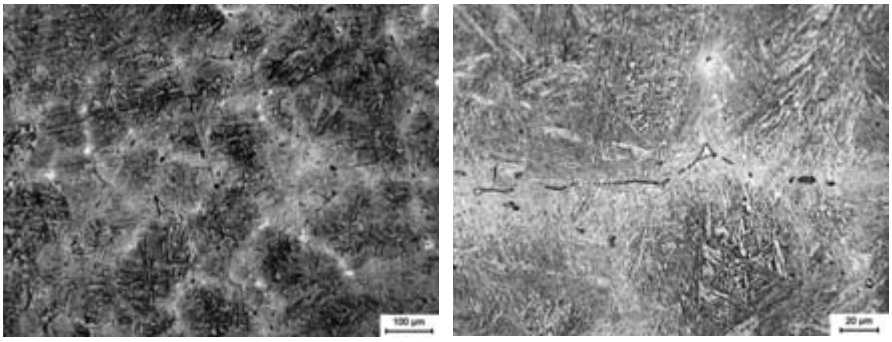


a) b) c) d)

Figure 6. Some microstructure examples of experimental steel subjected to various heat treatments, a) 1000 °C/30 min. + 200 °C/1h, b) 1000 °C/30 min. + 600 °C/1h, c) 1000 °C/1h + 200 °C/1h, d) 1000 °C/1h. + 600 °C/1h

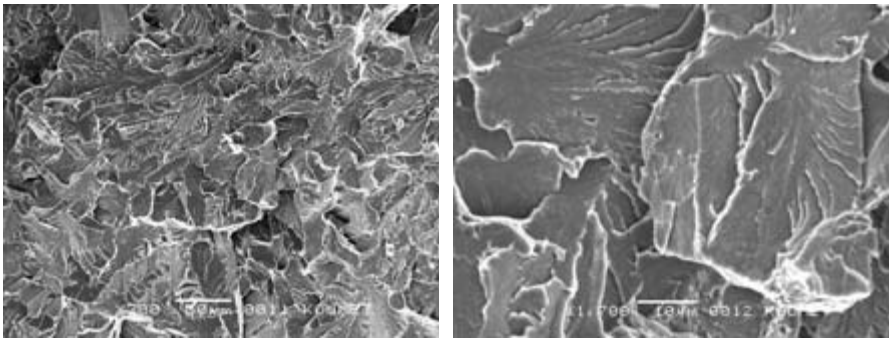
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a) b)

Figure 7. Microstructural examples showing segregation regions, a) and b) Heat treatment: 1000 °C/1h + 200 °C/1h, etched with nitale, LM



a) b)

Figure 8. Fracture surfaces of experimental steel failured by cleavage rupture at different magnifications, a) and b) Heat treatment: 1000 °C/30 min. + 200 °C/3h

which is formed during cooling and remains in the matrix due to inadequate austenization temperature or insufficient time to be dissolved totally. In general, the austenization heat treatment is performed after CW, and all phases are desired to be dissolved to form a single austenite phase. Subsequently proper quenching and tempering are applied on materials in order to gain optimum mechanical properties.

Fractographical Examinations

The aim of fractography is to analyze failured objects. Fractographical examinations present a good relationship between the reasons of formed fracture surface topography and basic rupture mechanism. Fractographical examinations presented in this study have been made by using scanning electron microscope (SEM) due to its high resolution. In general SEM examinations are an important part of failure analysis.

Fractures consist of crack formation (crack nucleation) and crack growth. Many parameters, such as type of loading and the sensitivity of materials to crack play an important role in failure. Fractures in engineering alloys can occur by a transgranular (through the grains) or an intergranular (along the grain boundaries) fracture path. There are essentially four basic fracture modes: dimple rupture, cleavage, fatigue, and decohesive rupture [6]. In this section the fracture surface characteristic and some of the mechanism associated with the fracture modes (especially dimple rupture and also cleavage) are presented and illustrated.

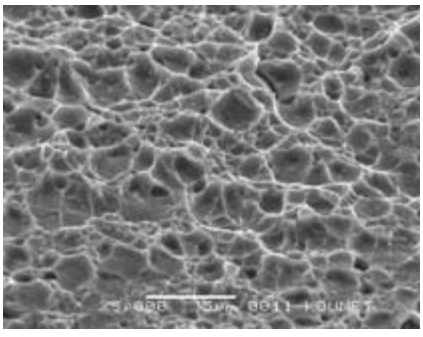


Figure 9. Fracture surfaces of experimental steel failured by dimple rupture, heat treatment: 1000 °C/1h + 600 °C/2h

Cleavage is a low-energy fracture that propagates along well-defined low-index crystallographic planes known as cleavage planes. Figure 8a represents a type of cleavage rupture of the examined experimental steel samples subjected to a heat treatment series of 1000 °C/30 min. + 200 °C/3h. This figure illustrates an example of intergranular rupture, and whereas Figure 8b shows the cleavage planes of Figure 8a at a higher magnification. As a result of low temperature tempering, the matrix consists of a relaxed martensite and a large number of dispersed fine carbide precipitates. These precipitates add high strength and hardness to the matrix but adversely affect the deformation capability of steel.

When overload is the principal cause of fracture, most common structural alloys fail by a process known as microvoid coalescence. The microvoids nucleate at regions such as second-phase particles, inclusions, grain boundries and dislocation pile-ups. As the strain in the material increases, the microvoids grow, coalesce, and finally develop a continuous fracture surface [6]. Ductile rupture has a characteristic appearence like dimples under SEM. Figure 9 represents an example of dimple rupture of examined experimental steel which is subjected to a heat treatment series of 1000 °C/1h + 600 °C/2h. Depending on increased tempering temperature and time, the toughness of the steel increases and as a result the plastic deformation capability of the steel improves.

Segregation regions within materials generally are not desired due to their negative effects on the mechanical properties of materials. In cast conditions, interdendritic regions have a high density of segregated element phases which have been formed during dendritic solidification. These kind of regions have a low plastic deformation capability similar to grain boundaries, and exhibit a brittle characteristic. Figure 10a illustrates a fracture surface of the tensile experiment sample which is undergone a heat treatment series of 1000 °C/1h + 600 °C/2h indicating a tough matrix and an elongated inclusion like MnS in the rolling direction. Interdendritic regions have flat surfaces demonstrating brittle characteristics. Dendritic regions have a great toughness and thus indicate ductile characteristic contrary to interdendritic regions after fracture. Figure 10b shows the transition of brittle-ductile

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regions in the same sample where inclusions have a weak bonding capacity to the matrix and failed prefentially by cleavage. On the other hand matrix has a higher toughness level which illustrates dimples when tensile is applied to it.

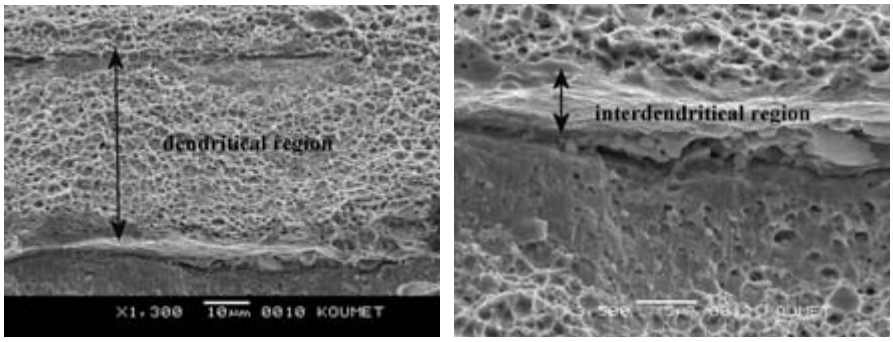
Examples of fractographical examinations are given in Figure 11 showing fracture surfaces of the tensile sample along the rolling direction, and in Figure 12 showing fracture surfaces of the impacted sample obtained perpendicular to the rolling direction. When all fracture surfaces are examined, it is obvious that desired fracture characteristic is obtained in the precipitation strengthened matrix. Fractures develop without any plastic deformation through inclusions like MnS, and this explains that how brittle rupture can occur. Figure 11 supports this kind of concept as MnS lies parallel to the rolling direction and causes failure in tensile sample. Consequently, fractures occur in the same direction. However fractures form vertically to inclusion lines in impacted sample, having an appearance of dark circular dimples surrounding inclusions.

Conclusions

In this study, microstructural and fractographical examinations of an armor steel have been carried out. Heat treatment series, consisting of austenizing, quenching, and then tempering, respectively, have been selected to observe differences between microstructures, fracture surfaces, and also mechanical properties of samples.

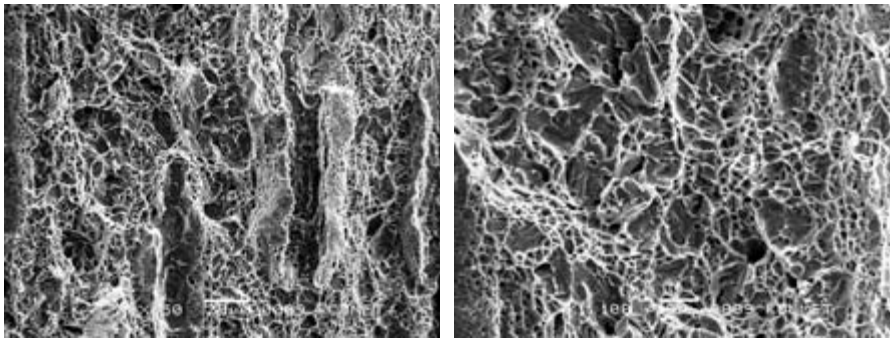
Formation of phases, and phase distribution in matrix throughout temperature range have been examined with a given composition at equilibrium condition by using the Thermocalc programme. All of the phase formations, including peritectic reaction such as liquid + $\delta \rightarrow \gamma$, and eutectic reaction such as liquid $\rightarrow \gamma$ + MC between and in interdendritic region, can be obtained by thermodynamical calculations.

After applying austenization heat treatment, a martensitic/bainitic microstructure has been obtained by quenching. Tempering heat treatment was performed to gain optimum mechanical properties. Maximum tensile and also yield strength have been observed in the sample subjected to a heat treatment series of 1000 °C/30 min. + 200 °C/1h.



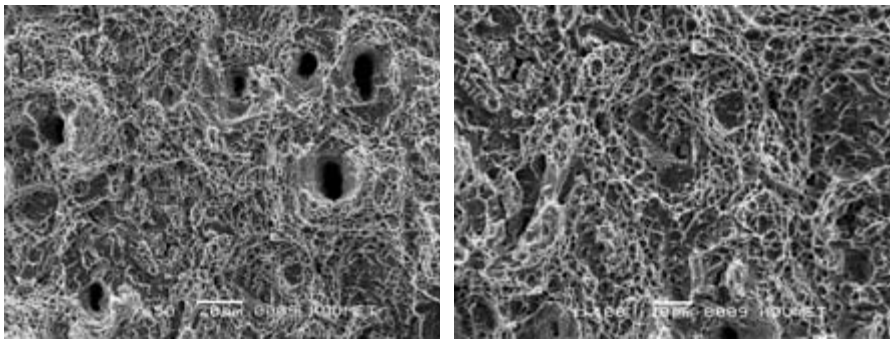
a) b)

Figure 10. Fracture surfaces of tensile sample, a) dendritic region within steel, b) the transition of dendritic region-interdendritic region, heat treatment: 1000 °C/1h + 600 °C/2h



a) b)

Figure 11. Fracture surfaces of tensile sample, longitudinal specimen, heat treatment: 1000 °C/1h + 600 °C/2h



a) b)

Figure 12. Fracture surfaces of impacted sample, transverse specimen, heat treatment: 1000 °C/1h + 600 °C/2h

Maximum hardness determined on the same sample is 414 HV. Cleavage type fractures occur in the hard martensitic/bainitic matrix due to the selection of low tempering temperature and time. As a result of this behaviour the same sample has the lowest impact strength, that is 36.6 J/cm^2.

When tempering temperature is increased, secondary hardness forms in the tempered martensitic matrix giving adequate hardness and toughness. For example, in the sample subjected to the heat treatment series of 1000 °C/1 h + 600 °C/2 h, the matrix has the optimum mechanical properties, while tensile strength and yield strength are 1280 MPa and 1128 MPa respectively. When its fracture surface is examined it is obvious that matrix has uniform dimples, indicating good plastic deformation capability under loading. As a result, the sample has the highest impact strength, that is 59 J/cm^2.

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Dr.-Ing. Alpay Yılmaz, Jahrgang 1973, achieved his Bachelor Degree at the Technical University of Istanbul in 1996, his MSc degree at the Kocaeli University in 2002, and his Dr.-Ing. degree at the Kocaeli University in 2006. From 1996 to 1998 he was working in the Foundary Anadolu Döküm. Then he changed to the Turkish Army and was head of the Departments Foundary and Heat Treatment at the 7th Maintenance Center in Tuzla, Istanbul from 1998 to 2007. Since 2007 he is at the Turkish Land Forces Command, Office of Technical Project & Administration in Ankara, Turkey.

Abstract

Mikrostrukturelle und fraktographische Untersuchungen an abgeschreckten und angelassenen Panzerstählen. Die diesem Beitrag zugrunde liegende Studie befasst sich mit experimentellen Bor-legierten Stählen, die nach dem Walzen austenitisiert, abgeschreckt und angelassen sind. Bei der Austenitisierung hat sich eine homogene Austenitphase eingestellt, die nach dem Abschrecken vollständig in Martensit umwandelt ist. Dieses Gefüge ist sehr hart, jedoch ist seine Zähigkeit niedrig. Deshalb ist eine Anlass-Wärmebehandlung bei 200 °C und 600 °C und unterschiedlichen Zeiten durchgeführt worden, um eine optimale Zähigkeit und damit eine ideale Kombination von Festigkeit und Zähigkeit zu erhalten.
Die Phasen, die sich bei der Abkühlung der gegebenen Stahlzusammensetzung bilden, sowie ihre Mengen wurden für den Gleichgewichtszustand mittels ThermoCalc bestimmt. Die verschiedenen Wärmebehandlungsgefüge sind mit dem Lichtmikroskop untersucht worden. Der Einfluss der Gefüge auf die mechanischen Eigenschaften sind durch den Vergleich ihrer Festigkeit, Dehnung und Schlagzähigkeit angegeben worden. Darüber hinaus sind die Bruchflächen der Proben für den Zugversuch und Schlagbiegeversuch mit dem Rasterelektronenmikroskop untersucht worden, um die Bruchmechanismen zu ermitteln.

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