JP&M Engineering

Journal of Petroleum and Mining Engineering

Processing and Evaluation of Ferritic-Bainitic Multi-phase Steel

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Abstract

Production of bainite-ferrite multi-phase steels on industrial CSP scale provides approach to improve mechanical properties along with large cost reduction opportunity. In terms of hot-rolling conditions, chemical composition, and microstructure; characterizing ferritic-bainitic steels and their difference from ferritic-pearlitic steels regarding properties and cost becomes of a great importance to understand the optimum conditions for production. This work is aiming to improve mechanical properties of steels by producing bainite-ferrite steel through industrial scale runs on the 6 stands hot strip mill (HSM) with at EZDK steel using low alloy Al-killed steel. The effect of hot rolling coiling temperature and alloy content on final properties was studied. In addition to characterizing differences between ferrite-bainite steels and conventional ferrite-pearlite steels. Ferrite-bainite steel has been successfully produced in different conditions, showing up to 27% (~100 MPa) improvement in yield strength, 8% (42 MPa) in tensile strength, 13% in total elongation, and 11% (27 J) in impact toughness. Vanadium microalloying has proven to have almost no effect on mechanical properties of ferrite-bainite steels upon its rising from 0.045% to 0.063% which is not the case in ferrite-pearlite steels. Ferrite-bainite steel offers a great margin for alloy cost saving compared to ultra-low carbon microalloyed steels (up to 56%) through replacing chemical strengthening with phase transformation strengthening. We expect this new approach to significantly reduce the cost of production without sacrificing the quality.

Keywords

Bainite; Multi-phase steel; CSP; HSM; Microalloying,

Introduction

In order to respond to the requirements of steels to exhibit a good combination of high strength and good impact toughness a new generation of low carbon microalloyed steels has been developed [1]. Increasing carbon concentrations may impair the impact toughness and weldability of steel. In order to compensate the loss of strength due to low carbon content, the alloy design philosophy is based on the advanced use of cost effective microalloying elements (MAE), such as niobium (Nb), titanium (Ti), vanadium (V) and boron (B) in conjunction with moderate levels of other alloying elements, such as manganese (Mn), silicon (Si), chromium (Cr), molybdenum (Mo) and copper (Cu) [2]. The sophisticated use of abovementioned combinations of microalloying and alloying elements in conjunction with low carbon content can lead to steels with yield strength ranging from 500 MPa [3] even up to 900 MPa [4]. These alloying and microalloying elements contribute to the increase in strength via microstructural refinement, precipitation hardening and solid solution strengthening as well as strengthening through microstructural modification [4].

Generally, high strength steels are processed via thermomechanically controlled processing (TMCP). In the last stage, accelerated cooling can be applied to refine the resulting ferrite grain size or to suppress the formation of polygonal ferrite and facilitate the formation of lower-temperature transformation products such as different types of bainite [5].

Microstructures of these steels are often complex, consisting of mixtures of different ferrite morphologies, and therefore, wide combinations of mechanical properties can be achieved by controlling them. The bainitic transformation is one of the most complex and disputed phase transformations in steels and all microstructures found may not exhibit the typical bainitic type of transformation [6].

Multiphase steel production is considered one of the most reliable ways to improve steel mechanical properties; it can improve both strength and ductility.

Moreover, it has a great impact of reducing production cost of certain steels via lowering their alloying content and replacing the precipitation hardening with phase-transformation strengthening.

Ferrite-bainite multiphase steels are believed to offer these benefits in addition to easier production conditions than in case of other multiphase steels (e.g. dual phase). Production of ferrite-bainite steel was not carried out before using the compact strip production plant (CSP) at EZDK Steel and this work introduces three industrial production trials done successfully.

The aim of the present work is to improve mechanical properties by using bainite transformation to produce ferrite-bainite steels through industrial runs at EZDK Steel CSP mill. Mechanical properties improvement is planned to be done by phase-transformation via lowering hot rolling coiling temperatures and comparing the results with similar chemical compositions ferrite-pearlite steels.

Experimental Work

Steel Chemical Composition

The considered 5 scenarios represent 3 ferrite-bainite steels and 2 ferrite-pearlite steels, all of them are Aluminium-killed, and their chemical compositions are shown in table 1.

Table 1 Trials chemical composition (in wt %).

Trial	Type	C	Mn	V	S	P
T-A	Ferrite-bainite	0.168	0.83	0.015	0.002	0.008
T-B		0.184	0.54	0.063	0.005	0.009
T-C		0.194	0.54	0.045	0.003	0.008
FP-01	Ferrite-pearlite	0.190	0.83	0.017	0.003	0.008
FP-02		0.191	0.54	0.060	0.003	0.008

The chemical analyses were done using optical emission spectrometer (model: ARL 3460-2342).

Production Sequence

This work was carried out using the hot strip mill (HSM) containing 6 stands F1 – F6 at EZDK Steel (Alexandria, Egypt) as shown in Figure 1 [7].

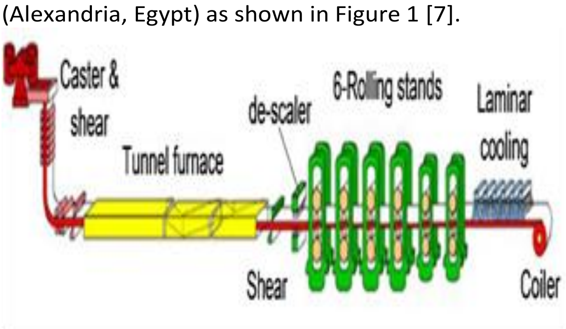

Figure 1 General layout of CSP.

Molten steel flows from a ladle, through a tundish into the mold. Once cast in the mold, the molten steel freezes against the water-cooled walls of a bottomless copper mold to form a solid shell. The mold is oscillated vertically in order to discourage sticking of the shell to the mold walls. Below the mold exit, the thin solidified shell acts as a container to support the remaining liquid, which makes up the interior of the strand. After the center is completely solidified (at the "metallurgical length" of the caster), the strand is cut with shear into slabs with fixed thickness (=54 mm) but different width according to the required final product (in this work the slabs were deformed to a final sheet thickness of 2.7mm (T-B, T-C, FP-02) and 3.0mm (T-A, FP-01)).

The function of the tunnel furnace is to attain the required temperature before the rolling process and to obtain an even temperature distribution over the slab. The rolling process starts with the roughing stand F1 and ends with the finishing stand F6. The parameters controlling this process will be described in the next sub-section. After rolling, sheets are cooled to reach the required coiling temperature. There are several strategies for cooling after last stand: early cooling, late cooling or free defined cooling. Therefore, control of the laminar cooling is necessary to obtain the required coiling temperature and accordingly the required properties.

Rolling Parameters

Automatic measurements and calculations of different rolling parameters were carried out for each of the six stands.

Temperature Measurement and Control

Pyrometers are utilized to measure the temperature after every stand and at the coiler. There are two types of pyrometers used in the production line:

- The first one is used at every stand (type SYSTEM4-M1) with measurement range (600~1600 °C) and accuracy of ±(0.004 ×measured temperature) [8].
- The second one is used at coiler (type SYSTEM4-M2) with measurement range (300~1100 °C) and accuracy of ±(0.0025x measured temperature) [8].

Both utilize a silicon cell detector and operate at short wavelengths around 1~1.6 µm where emissivity errors are minimized. In addition, they have a fast response time of 5 ms and minimum target diameter of 0.9 mm [8].

The finishing deformation temperature at last stand and coiling temperature can be easily controlled by setting the required temperature by the operator in the main pulpit of HSM. Accordingly, the system automatically determines and adjusts the rolling line to obtain these temperatures by controlling the rolling speed, inter-stand cooling operation and the amount of cooling water.

Mechanical Tests

In this work, the considered production trials have been evaluated by the tensile test. Samples for the tensile tests were taken from the sheets after discarding the un-cooled and off-gauge parts.

Tensile test samples were machined from longitudinal direction according to the standard EN10002-1 for quasi-static tensile with test length of 43.4 mm and width at the test length of 20 mm.

Three samples for every case were tested and the average of the 3 samples values was calculated. Tensile tests were carried out using the tensile testing machine Model Zwick/Roll (250 KN) (accuracy ±1 % in yield and tensile strength and ±2 % in elongation). Impact test was conducted on small size impact samples (10 mm height, 2.5 mm thick, 55 mm length and 2mm V-notch depth) according to the standard EN 100045-1 (accuracy ±1 %). Three samples for every case were also tested and the average of the 3 samples values was calculated. Charpy impact testing machine type RKP 300/450 was used for determining the impact toughness.

Microstructure Examination

Samples were taken from sheets cross section and perpendicular to rolling direction to examine the microstructure and to measure grain size.

The procedure of specimens' preparation consists of cutting, mounting, grinding, polishing and etching. After cutting the specimens, mounting of specimens is usually necessary to be easily handled. Surface layers damaged by cutting were removed by grinding. Mounted specimens were ground using SiC abrasive paper in different grades (320~1200). Samples were water cooled during grinding and rinsed after each grinding stage. Polishing discs were covered with soft cloth impregnated with abrasive alumina particles (0.05 μm) and an oily lubricant.

Etching is used to reveal the microstructure of the metal through selective chemical attack. It also removes the thin, deformed layer introduced during grinding and polishing. The used etchant was Nital (2% nitric acid) [9].

Following the specimen preparation, the specimens were investigated using Olympus optical microscope, equipped with a digital camera model AXIOCAM MRC5. Microstructure images were taken at different magnification and measurement of grain size as well as grain area was carried out according to ASTM 112.

Scanning electron microscopy SEM (Type JEOL model JSM 5410) and Field emission scanning electron microscope, FESEM, model quanta SEG 250, were used to deeply examine and investigate the microstructures at different magnifications.

Results and Discussion

Processing parameters

This is the first time to produce ferrite-bainite steels at EZDK.

Ferrite-bainite steels may be a possible substitution for normal ferrite-pearlite steels used in many applications.

The processing parameters of the production of the considered 5 runs are illustrated in table-2.

In the ferrite-bainite production, coiling temperatures were chosen to be well below the calculated *Bs* temperature of each steel composition.

Figures 2, 3 and 4 illustrate cooling regimes of trials T-A, T-B, and T-C.

The bainite starting temperatures for the processed steels were found in the range of 597 to 608 ^{o}C, Table 2.

Online production of ferrite-bainite steels was mainly through adjusting coiling regime by controlling water flow in order to reach temperatures below Bs.

For normal ferrite-pearlite steels, coiling temperature is set to be 620 ^{o}C to get ferritic-pearlitic structures.

Table 2 Processing parameters in the industrial scale production trials.

Trial number	T-A	T-B	T-C	FP-01	FP-02
Type	Ferrite-bainite			Ferrite-pearlite	
Finish rolling temp., ^{o}C	880	888	890	900	900
Coiling temp., °C	450	504	520	620	620
Cooling Rate (CR) ,°C/s	22	40	37	25	25
Reduction @ F1, %	54	60	60	54	60
Reduction @ F2, %	46	47	47	46	47
Reduction @ F3, %	39	40	40	39	40
Reduction @ F4, %	34	36	36	34	36
Reduction @ F5, %	27	24	24	27	24
Reduction @ F6, %	34	19	19	22	19
Final thickness, mm	3.0	2.7	2.7	3.0	2.7
Bainite start temp. °C	599	608	607	597	607
Average grain size, μm	6	4.5	5.5	8	7
Yield strength, MPa	457	467	470	360	456
Tensile strength, MPa	572	594	592	530	560
Elongation, %	26	26	27	23	25
Impact toughness at 0 ° C, J	257	267	273	250	243

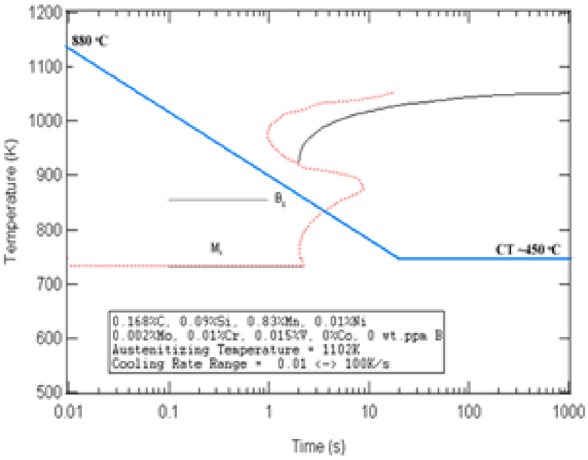

Figure 2 Coolinh regime of T-A.

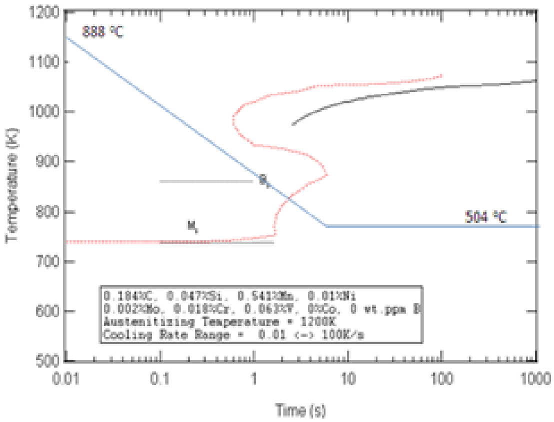

Figure 3 Cooling regime of T-B.

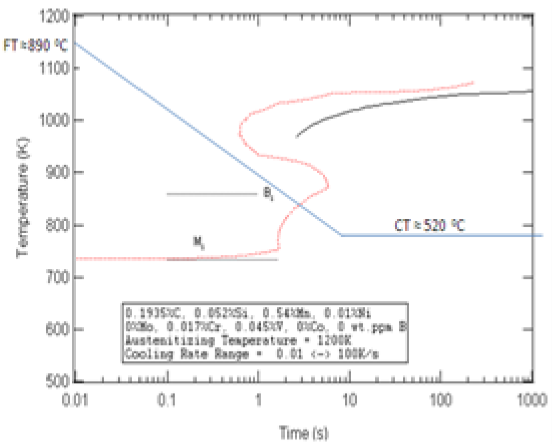

Figure 4 Cooling regime of T-C.

Microstructure evaluation

Among the five on plant-scale trials to produce different hot-deformed low-carbon steel with mainly variation in vanadium content as indicated in the experimental part; three trials have been carried out as an attempt to produce a mixed ferrite and bainite microstructure without any isothermal holding (i.e. continuous cooling following the hot deformation part of the processing). Two average cooling rates of ~ 20 ^{o}C/s and 38 ^{o}C/s are used with a lower coiling temperature of 450 ^{o}C associated with the lower cooled trials when a minimum vanadium content is present in the steel composition, as shown in Tables 1 and 2. The second and third trials (T-B) and (T-C) contain 0.063 and 0.045 wt% V respectively, but both cooling rates, coiling temperatures as well finishing rolling temperatures are the same as indicated in table-2.

The microstructures observed by light microscope and scanning electron microscope (SEM) are shown in figure 5 (a&b) of the low-vanadium trial (0.015 wt% V). The microstructure showed a mixed microstructure dominating by ferrite appears white (Fig. 5a). In addition to a second phase of bainite morphology, with the ferrite is typical quasi-polygonal ferrite. This type of ferrite is generally characterized by displaying irregular boundaries and to contain a dislocation substructure, appears as faint boundaries inside the ferrite grains. Such type of quasi-polygonal ferrite microstructure is observed with all three trials' microstructures, at higher microalloyed vanadium content, as shown in figures 6 and 7.

In general, the amount of bainite obtained is observed to vary with both cooling rate as well as vanadium content. For the present relative high carbon content steel ~ 0.180 wt% C and high cooling rate during processing it is expected that the quasi-polygonal ferrite will form first during austenite transformation, the residual austenite is then enriched not only by carbon, but also by other alloying elements. The enriched matrix will subsequently enhance the bainite formation and possibly martensite transformation could take place during continuous cooling; similar suggestion has been examined in a sheet-steel simulation processing [10].

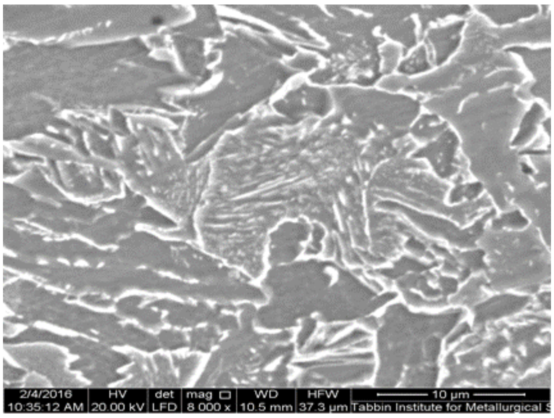

Figure 5 (a) light micrograph of T-A showing bainite laths; (b) corresponding scanning electron micrograph.

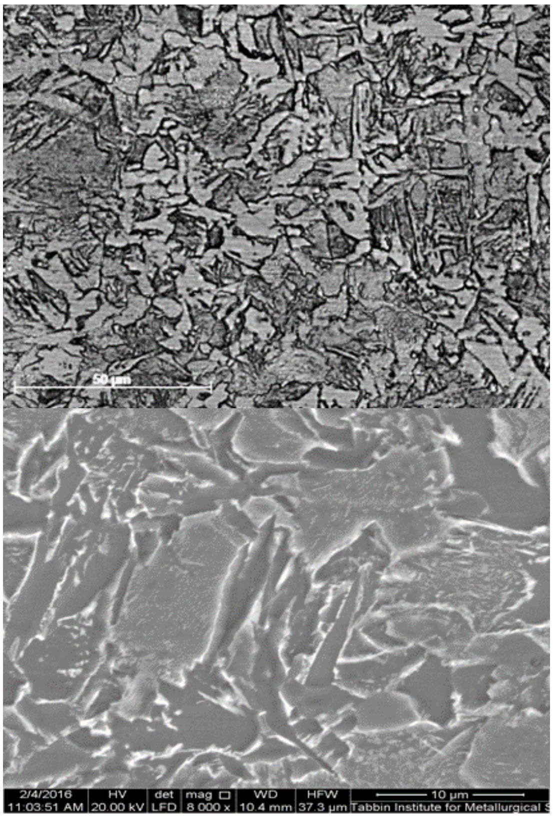

Figure 6 (a) light micrograph of T-B showing bainite laths; (b) corresponding scanning electron micrograph.

Bainite morphology can be either one or more in one single microstructure and could be lamellar and/or granular bainite as shown in figures 6 and 7, where a higher rates of cooling (40 & 37 ^{o}C/s) and coiling temperatures (504 & 520 ^{o}C) at different vanadium contents are used for these two trials. The formation of granular bainite has been reported to take place at a lower-temperature and slower rate of cooling [11]; this would mean that granular bainite transformation takes place at later stage by regard to ferrite transformation and even can occur during coiling of steel. However, a further investigation is needed to determine the transformation sequence for such complex microstructure.

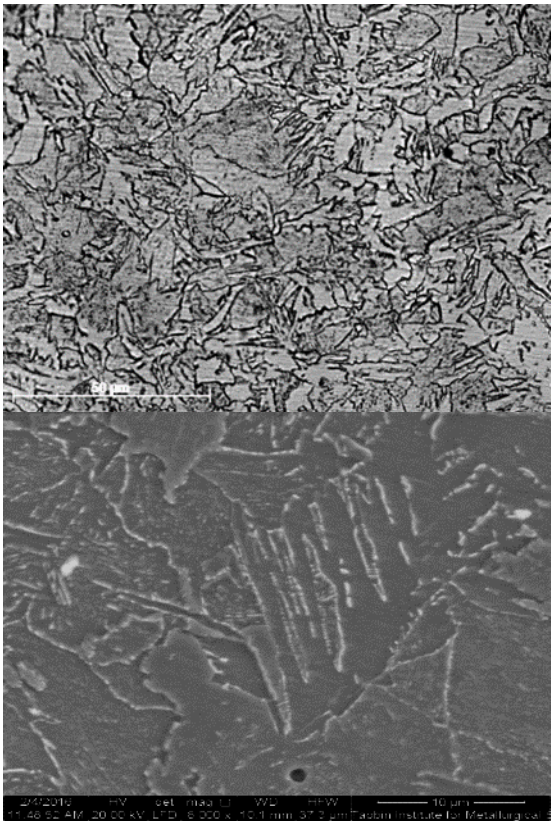

Figure 7 (a) light micrograph of T-C showing bainite laths; (b) corresponding scanning electron micrograph.

The possible formation of vanadium carbide or carbo-nitride precipitates depends not only on V-content but also on the C and N contents of steel. It is argue that vanadium does not directly precipitates in austenite and the precipitate formation can be enhanced if nitrogen is present in the steel or by plastic deformation [12] (strain induced precipitation). Further, once the precipitates are formed they will be the main factor governing the formation of Intergranular ferrite. However, the exact mechanism by which vanadium additions are enhancing the nucleation rate of ferrite is not quite known. The effect of increasing vanadium content on the steel bainite morphology is not pronounced through the SEM observations in the present work. An added difficulty could be related to the formation of very small size VC or V(C,N) precipitates due to continuous cooling used for processing of steels in trials B and C, where would be less times for coarsening. Moreover, the processes of vanadium precipitates can have an important role in preventing the softening of microstructure during cooling from coiling temperature to room temperature. The effectiveness of precipitates in preventing recovery of the formed bainite in hot-strip steel has been reported and can maintain the steel strength [13].

Following the hot-deformation in all the trials the microstructures in general did not show any of the typical morphologies of the high deformed structure, which can be an indication that recrystallization either dynamic or static must have been taking place during processing. The austenite dynamic recrystallization can be envisaged to take place during high strain deformation at high temperature of the early stage of hot-deformation. On the other hand, the interpasses times between the six-deformation stands, however; it could be only a few seconds but it is considered quite enough for static recrystallization to occur, a total of interpasses time was estimated to be between 12 ~ 15 seconds for online six stands. The formed ferrite from the austenite transformation might undergo a dynamic recrystallization due to continuous accumulated strain before reaching the final stand. The presence of VC or V(C, N) precipitates will act as pinning for grain boundaries movement (i.e. it retards the recrystallization process). This retardation effect is likely to be dependent on the amount of precipitates present in either austenite or ferrite matrix.

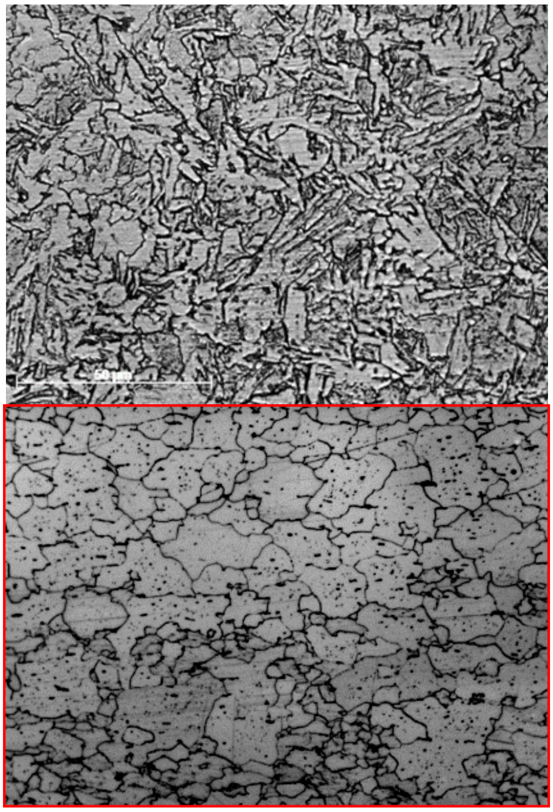

Figure 8 Comparison between light micrographs of (a) T-A and (b) FP-01.

The same steel compositions used in trials T-A and T-B were used to produce another on plant-scale microstructure of ferrite and pearlite. The total strain during hot-deformation was the same and the finishing rolling temperature is slightly higher at 900 ^{o}C; however, coiling temperature is raised up to 620 ^{o}C replacing the lower one at 500 ^{o}C for bainite-ferrite steel, as in table-2.

The produced microstructures for two different vanadium microalloying are shown in figures 8 to 11. And it is clear that by only changing the hot-rolling conditions a different microstructure can be obtained.

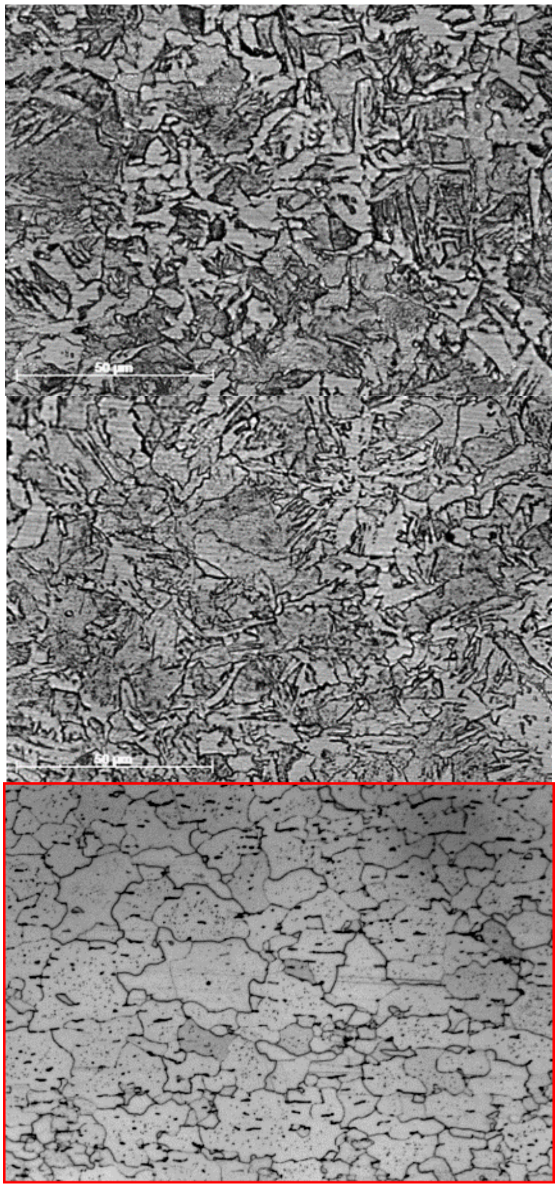

Figure 9 Comparison between light micrographs of (a) T-B, (b) T-C, and (c) FP-02.

The observed microstructure is mainly dominated by polygonal ferrite matrix with a second phase of pearlite concentrated essentially at grain boundaries. There is no much difference in microstructure which could be related to different vanadium contents. However, a major difference from ferrite bainite microstructure is that the ferrite grain size is larger in ferrite-pearlite steel. This could be related to higher coiling temperatures and lower cooling rates which are favorable conditions for grain coarsening. The microstructure also shows the presence of sub-structure boundaries that could be a result of a recovery process enhanced by the remaining high dislocation density following the hot-deformation stage. However, the vanadium which is known to retard the pearlite transformation has no evidence to have such effect on the obtained steel microstructure.

Mechanical properties

General discussion

Table-3 summarizes mechanical properties of the considered runs.

Comparing similar chemical compositions of ferrite-bainite and ferrite-pearlite steels (i.e. T-A with FP-01 and T-B, T-C with FP-02) showed significant improvement. Improvement reaches 27% in yield strength, 8% in tensile strength, 13% in elongation, and 11% in impact toughness.

Table 3 Mechanical properties of the considered runs.

Trial	Thickness Mm	Yield MPa	UTS MPa	Elongation %	Impact toughness at 0 ^{o}C Joule
T-A	3.0	457	572	26	257
T-B	2.7	467	594	26	267
T-C	2.7	470	592	27	273
FP-01	3.0	360	530	23	250

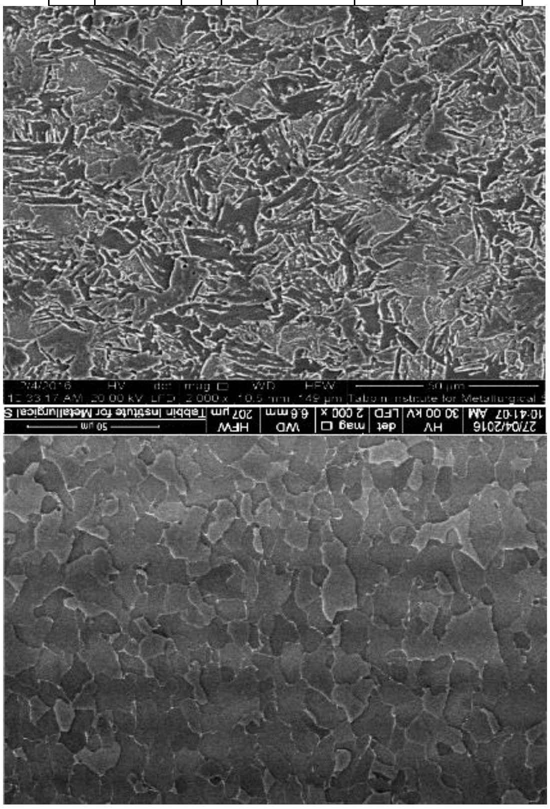

Figure 10 Comparison between scanning electron micrographs of (a) T-A and (b) FP-01.

The yield strength of low carbon steel can in general be raised by microalloying elements additions, the improvement is mainly related to ferrite grain refinement, both solid solution and precipitation strengthening. Furthermore, more strengthening can be also provided through controlled austenite transformation to very fine structure ≈ 1µm such as martensite and/or bainite with accompanying high dislocation density in the ferrite matrix, due to their displacive type of transformation. Therefore, a base bainite microstructure produced during continuous cooling following hot-rolling can be considered as a powerful method to reach a relative high yield stress with good toughness.

As indicated before, the present produced ferrite microstructure steel can be driven into polygonal ferrite (PF), acicular and/or granular ferrite, bainite ferrite using the processing controlled parameter such as cooling rate and coiling temperature. All the transformed microstructures come from a mixed diffusion and shear transformation mode and strained the ferrite matrix due to the displacive nature of transformation and the polygonal ferrite occurs only through diffusion mechanisms. Most of microalloying elements strengthening additions contribute to strengthening through the carbide/carbonitride precipitation and/or segregation to boundaries pinning their movement. However, the exact percentage of their contribution to the total strengthening is not that clear.

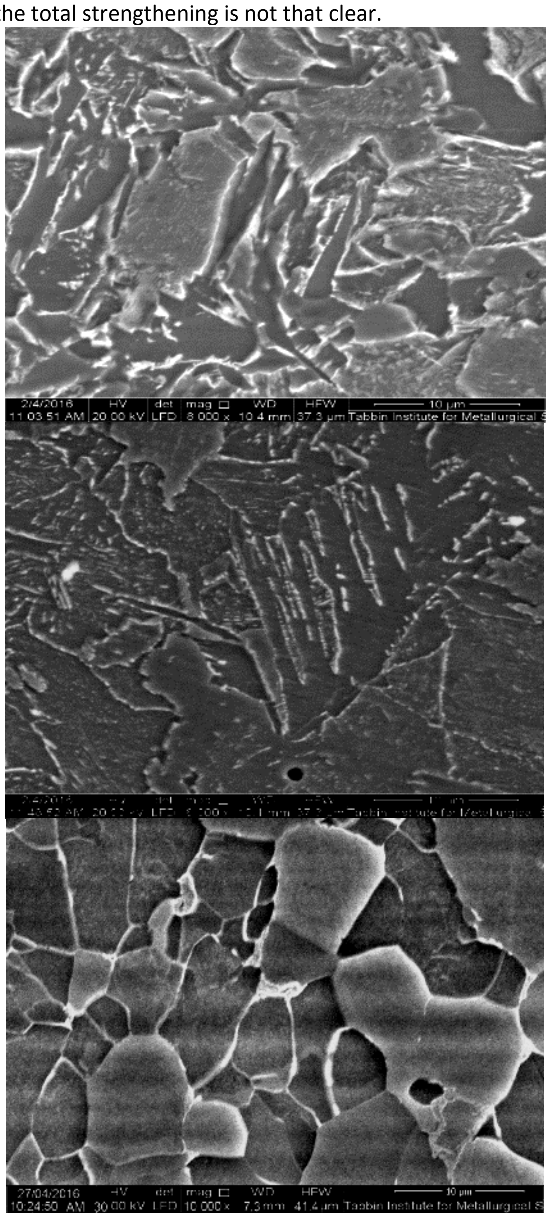

Figure 11 Comparison between scanning electron micrographs of (a) T-B, (b) T-C, and (c) FP-02.

The relation between strengthening mechanism and responsible microstructure is usually associated with the changes in yield and ultimate tensile stress as well as yield stress to ultimate stress ratio (YS/UTS=YR).; thus yield ratio (YR), is equivalent to work hardening rate. For many steel applications the optimized YR values are predetermined and related to a specific microstructure [14]. However, it is impossible to state which microstructure has the lowest YR value, since this ration can be function of both yield strengthening and work hardening rate.

The tensile test results of the different production trials can be classified into two types according to the microstructure: 3-trials with a dominant ferrite-bainite and 2-trials dominated by ferrite-pearlite microstructure, the tensile and impact toughness results are shown in table 3. The variation of both Mn and V contents in the steel compositions is shown in table 1.

The Vanadium Effect

In order to evaluate the role of vanadium addition in steel properties, a comparison between trials (T-B) and (T-C) is used, since all the processing parameters are similar, mainly cooling rate and coiling temperature but they have different vanadium contents of 0.063 and 0.045 wt% respectively. The results in table-3 show very close values of YS, UTS and YR ratio for both steel types with different vanadium contents. A slight reduction on the impact values with increasing vanadium content.

These results confirm the microstructure finding in the previous section, where both steels have a similar basic microstructure of ferrite and bainite with almost the same morphology and vanadium addition did not produce any effective changes with steel structure as well as the strength. At a lower vanadium content of 0.014 wt%, but both cooling rate and coiling temperature are reduced for trial (T-A), it resulted in a lower yield and ultimate tensile strength and lower impact values compared with the previous two trials with higher vanadium contents, table-3. However, the microstructure of this trial shows also the presence of bainite structure in a ferrite matrix, but it is obvious that reducing the cooling rate and coiling temperature increased the amount of polygonal ferrite which can be responsible for the observed mechanical properties reduction. These change in microstructure were not accompanied by a change in the work hardening rate, since YR ratio is around 80% compared to 79% for the higher vanadium steels. Furthermore, the reduction of cooling rate with coiling temperature is known to increase the amount of polygonal ferrite on the expense of bainite morphology [11] which is confirmed by the present results, it remains questionable is it the absent role of vanadium in preventing the recovery of bainite structure during coiling which is responsible for trial (T-A) strength reduction.

It has been generally argued that the strain effect produced the bainite displacive transformation in ferrite matrix will almost or totally recovered in the microstructure during cooling from coiling to room temperature. Moreover, the softening effect resulted from recovery process can be to large extent prevented if vanadium is added to the steel. The vanadium solute will segregate partly or totally to grain boundaries and causes pinning their movement and could as well interact elastically with dislocation in matrix and pin them preventing dislocation rearrangement as necessary movement for recovery from taking place, such mechanism has been proposed by several authors [11], [15].

The role of vanadium in steel is generally very complex it can form the V(C,N) precipitates as indicated before, such precipitates can contribute to the bainite steel strength, if present in a reasonable quantity. This type of precipitates are reported to form inside the ferrite structure with a very small size and been impenetrable hard particles which strength effect will occur by Orowan mechanism which is particle-distance dependent. [16]

Ferrite-pearlite microstructure effect

As indicated in table 2, the difference between trial (T-A) and (FP-01) is mainly the coiling temperature (450 and 620 ^{o}C respectively) with low cooling rate of around 25 ^{o}C/S for both. Such difference resulted in two completely different microstructures, at higher coiling temperature a polygonal ferrite dominating matrix is obtained with second phase precipitates, pearlite structure confined mainly at the boundaries. On the other hand, with a low coiling temperature, a ferrite bainite structure is obtained as indicated before. The observed difference in mechanical properties, as shown in table-3, is quite expected and the low strength values are associated with polygonal ferrite microstructure. Furthermore, lower YR of 68% is obtained for this softer ferrite-pearlite microstructure indicating a lower work hardening rate compared to trial (T-A) ferrite bainite steel. These results show that coiling temperature is very much effective in controlling the type of the steel microstructure relatively to vanadium addition effects. However, using the same processing conditions as for the producing (FP-01), but with increasing the vanadium content from 0.017 to 0.060 wt% for trial (FP-02), resulted in similar polygonal ferrite and pearlite microstructure, the high vanadium steel showed a relative smaller ferrite grain size, which is possibly the main contributor to the observed increase in strength between the two trials (FP-01 & FP-02). It is expected that refinement of ferrite grains in the higher v-steel is related to the effect of V(C,N) precipitates formation which can effectively limit possible ferrite grain coarsening following the high coiling temperature. It has been demonstrated that precipitation strengthening of vanadium increases significantly with carbon content of the steel [17], which can be another strengthening mechanism.

Bainite strengthening in hot-rolled band

Low carbon bainite can provide high strength to steel through various strengthening mechanisms; a) The fine bainite laths, b) The displacive type of bainite transformation is accompanied by a high dislocation density generated in the ferrite matrix and c) The main ferrite matrix strengthening through solid solution hardening in addition to possible second phase precipitation.

Recently, several attempts were carried out in order to quantify the different strengthening mechanisms for microalloyed steels with complex microstructure. A simple summation is a simplistic case to predict the total strengthening, marked deviation of the linearity in strengthening has been observed for various mixed microstructures [18]. The largest effect on the yield strength on the bainite microstructure comes from grain size and grain boundaries misorientation. However, an effective grain size for acicular ferrite, bainite and/or martensite is difficult to be measured with a high degree of confidence.

In order to make a quantitative analysis of the various strengthening mechanisms contribution to yield strength for the various examined microstructure steels, the well-known empirical equation [19] is used to calculate the strengthening contributions from both chemical composition of ferrite bainite steel and the grain size :

$$\Delta\sigma_c = 15.4\big(3.5 + 2.1[Mn] + 5.4[Si] + 23[N_f] + 1.13d^{-0.5}\big) \quad (1)$$

Where [Mn], [Si], and [N_f] are in weight percent and the free nitrogen dissolved in ferrite respectively, and d is the average grain size in mm, $\Delta\sigma_c$ in MPa, and the effect of vanadium microalloying addition was assumed negligible. The average grain size measured is shown to vary for the different bainite ferrite structures, to be between 7 to 5 µm. the estimated value for the largest d=7 µm is $\Delta\sigma_c = 284\ MPa$ and when a small d value of 5 µm is used it gives $\Delta\sigma_c = 322\ MPa$. These results show that grain size is effective in the steel strengthening. On average, the $\Delta\sigma_c$ contribution to strengthening will be estimated as 303 MPa. If this value is compared to the lowest measure yield strength of the 3-trials for the bainite ferrite structure steel which is 457 MPa, table 3, a difference of around 154 MPa stress will be then Mn accounted for the estimation by equation 1.

Increasing in dislocation density results in a strength increment, because further dislocations movement will become more difficult due to dislocations intersecting. The bainite ferrite contains relatively higher dislocations densities compared to polygonal ferrite microstructures, since some of dislocation substructure formed during austenite deformation can be transmitted to bainite. The displacive growth of bainite accompanied with the formation of dislocations lead to additional strengthening [19]. The well-known relation between dislocation density and strength increment is given by the following equation (equation-2) [20]:

$$\Delta\sigma_d = m\alpha\mu b\rho^{0.5}$$

Where $\Delta\sigma_d$ is the dislocation contribution to yield strength, m is Taylor factor for polycrystal, α is a geometrical factor depends upon the type of dislocation intersection, taken here to be 0.5π for first dislocation [21] and m=2.733 for bcc crystals, µ is the shear modulus and ρ is the dislocation density.

More recently, Takahashi and Bhadeshia have indicated that for low alloy steels the dislocation density is transformation temperature dependent [22], where the dislocation density seems to increase with decreasing transformation temperature.

Numerous measured and estimated values have been reported in the literature for dislocation density and the $\Delta\sigma_d$ value varies between 108 to 415 MPa mainly for ferrite bainite structure under continuous cooling [23]. However, a reliable dislocation measurement or estimation should take in consideration the effect of bainite morphology features effect on mechanical properties as well as bainite volume fraction and chemical composition of the steel.

The possible formation of V(C,N) and carbide precipitates is another possibility for increasing the strength as indicated before. The fine V(C,N) precipitates can effectively contribute to the steel strength if it has only high volume fraction [24]. Recently, the carbonitride vanadium precipitates are proven to be more effective in retarding the recovery of dislocations in bainite ferrite and, to a lesser degree, as precipitate strengthening [11]. This has been indicated to be due to the small size of the precipitates and their location at the dislocation substructure which take place in prior austenite grain boundaries.

From the above discussion it is possible to conclude that among the different strengthening mechanisms in ferrite bainite microstructure, the effect of solute addition and grain size and/or bainite lath size are the most effective in strengthening followed by the accumulated dislocations and their interaction and finally the second phase precipitates effect.

Conclusions

The current work has focused on the metallurgical aspects related to the production of ferrite-bainite low-carbon steel and its difference from the conventional ferrite-pearlite steel. 5 trials were carried out on a hot deformation processing line to assess conditions for ferrite-bainite production and characterize the difference from similar chemical composition ferrite-pearlite steels. The following conclusions were made:

- Ferrite-bainite multiphase steels could be produced at EZDKCSP mill by utilizing cooling water to attain a coiling temperature below the *Bs* temperature.
- The microstructures of the produced ferrite-bainite 3 trials shows bainite phase mainly in the form of laths.
- Mechanical properties have been improved in ferrite-bainite steels (up to 27% (~100 MPa) improvement in yield strength, 8% (42 MPa) in tensile strength, 13% in total elongation, and 11% (27 J) in impact toughness).
- Vanadium microalloying has proven to have almost no effect on mechanical properties of ferrite-bainite steels upon its rising from 0.045% to 0.063% which is not the case in ferrite-pearlite steels.
- Ferrite-bainite steel offers a great margin for alloy cost saving compared to ultra-low carbon microalloyed steels (up to 56%) through replacing chemical strengthening with phase transformation strengthening.

Acknowledgment

The authors express their great thanks and appreciation for EZDK Steel Co. for their cooperation and help in executing the steel production trials.

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