Research Article

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Journal of Heavy Metal Toxicity and Diseases
ISSN 2473-6457

2020
Vol. 5 No. 3:1

DOI: 10.21767/2473-6457.10030

A Review Paper on the Development of Dual Phase Steel

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Received on: August 4, 2020; **Accepted on:** August 27, 2020; **Published on:** September 15, 2020

Citation: Hussain Dawki, Andrew Gamal, Omar El-Kolaly (2020) A Review Paper on the Development of Dual Phase Steel. J Heavy Met Toxicity Dis Vol.5 No.3:1

Abstract

Ever since introducing the concept of Advanced High Strength Steels, Dual Phase Steel has proven to be the brightest candidate amongst the group. Owing to the simple microstructure, easy and economical production method, and its excellent strength-ductility balance, Dual Phase Steels are already being utilized in manufacturing light weight parts in the automotive industry. However, there are many scientific aspects of the subject that are still not explored. This along with the greater demand in the last few decades for initiating the use of light, strong and economically efficient steels led to wider research efforts to gain a better understanding of Dual Phase Steels. In this work, the advancements of research on Dual Phase Steel development are reviewed, with focus on the microstructural development via changes in composition and thermal treatment.

Keywords

Dual Phase, Intercritical Annealing, Isothermal, Martensite, Ferrite

Introduction

Nowadays, it is more important than ever for automotive industries to produce lower density metals with sufficient strength-ductility balance for the purpose of weight reduction, and in turn to minimize the energy consumption of vehicles to meet the environmental regulations imposed. Lightweight materials used in car manufacturing are extremely expensive. Therefore, it cannot be used when producing commercial, reasonable priced cars. This has introduced a great demand for the development of new light weight and strong materials. Metals are a great contender, with special emphasis on steels for their economic value, availability, and other desirable properties during manufacturing, such as formability, machinability, and castability. Dual phase steels are thus a very suitable and efficient material, fulfilling all the aforementioned aspects. In this work, various features of the Dual Phase Steel were discussed, hence DP steels are produced by a rather simple thermomechanical processing route, along with micro alloying, resulting in a simple ferrite martensite microstructure that gives rise to properties that are very acceptable industrially. Desirable properties of DP steels include and not limited to, a high UTS, along with high early strain hardening and a low initial yielding stress. In addition, this steel grade exhibits a homogenous plastic flow and good mechanical behavior.

Scientifically speaking, the maximum potential of DP steels has not been reached yet through the entire publications proposed. Various possibilities are still available, for instance the microstructure developed through the process and the mechanical properties resulted from it. Consequently, DP steel contradicts the complexity of the microstructure of other steel types. Processing route is essential, because of different constituents may be created, such as retained austenite, pearlite, bainite, carbides, and acicular ferrite. Moreover, composition and process selection can affect the volume fraction of phase, grain size, and ferrite/martensite morphology.

Chemical Analysis

It is mandatory to identify the main elements which are added to the composition of dual phase steel, therefore Pushkareva et al. [1] has specified the primary elements to be added into the composition, such as C, Si, Mn and Cr. Dual Phase steels are produced with high formability and suitable strength prospect, that's why low carbon steels are used. Because of its applications in automobile body panels, tin plate, and wire products, the design of composition should consider the low amount of C and Mn. [2]

M. Soliman and H. Palkowski et al. [3] have reported the response of bake hardening on dual phase due to its importance in automotive industry. Chemical analysis was preformed using glow discharge spectrometry (GDS). The reason of using the combustion method was accurate measurement of C and N. The results were 0.075 C, 0.007 N, 1.02 Mn. 0.433 Cr, 0.031 Al, 0.059 Si, 0.005 S, and 0.02 P.

After reviewing previous work [4], the concluded average for each element can noticed in Table 1.1:

C	Si	Mn	Cr
0.075 – 0.25	0.059 – 0.4	0.4 – 1.9	0.195 – 1.15

Table 1.1 Average Chemical Compositions

One of the main reasons to determine the chemical composition for DP steel besides the resulted properties and creation of the desired phases and constituents, to calculate the transition temperatures on the phase diagram to be able to design the thermal cycle by applying the chemical composition on the empirical equations. [5]

$$\begin{aligned}\text{Trzaska} \Leftrightarrow A_1(K) &= 1012 - 22.8\text{C} - 6.8\text{Mn} + \\ &+ 11.7\text{Cr} - 15\text{Ni} - 6.4\text{Mo} \\ &- 28\text{Cu} \\ \text{Park} \Leftrightarrow A_3(K) &= 1228 - 350\text{C} - 25\text{Mn} + 5 \\ &+ 106\text{Nb} + 100\text{Ti} + 68\text{Al} \\ &- 33\text{Ni} - 16\text{Cu} + 67\text{Mo}\end{aligned}$$

Microstructure and Morphology

Dual Phase Steels are low-carbon steels having a microstructure consisting of 10-40 volume percentages of hard martensite phases along with a soft ferrite matrix. Nonetheless, the morphology of the main desired phases, namely martensite and ferrite, and the formation of secondary phases are mainly dependent on the heat treatment and thermomechanical processing routes. [6]

A typical process to produce DP steel using continuous annealing technique involves heating a steel sample with ferrite-pearlite microstructure to reach intercritical temperatures and form ferrite austenite mixture, and then by accelerated cooling to form martensite phase from austenite. [7]

In a previous work [8], three 0.16C–1.38Si–3.20Mn DP steel samples were prepared with varying the intercritical annealing temperature (780 C to 820 1C) to increase the volume fraction of austenite, which in turn will transform into martensite upon quenching. SEM results of the heat-treated samples are shown in Fig.1.1. The resulting microstructures showed the expected increase in the volume fraction of martensite with increasing the annealing temperature, along with an increase in martensite grain size. It is worth mentioning that micro alloying elements like chromium, molybdenum, vanadium, etc., are usually added to dual phase steels to enhance properties. Their effects are continuously and extensively studied in many works; however, they were not added on the samples discussed above as to not influence micro structural phases and morphologies apart from the investigated conditions.

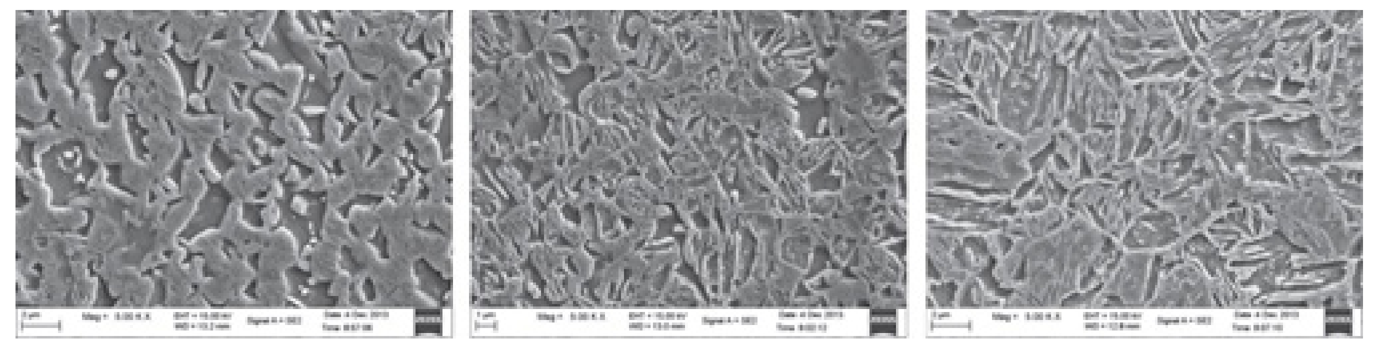

Figure 1.1 SEM results of the heat-treated samples.

Apart from the basic processing technique to produce a DP steel sample, for the same given composition, alterations in the intercritical annealing temperature, the annealing duration, and the mechanical processing if carried out (hot or cold rolling) all influence the microstructure via changes in grain size of the produced phases, volume fractions, distribution and morphology.

For instance, in a previous work [9] aiming to investigate ultrafine grain sized DP steel, a sample was produced by cold rolling an intercritically annealed sample and then subjecting it to a short three-minute intercritical annealing followed by water quenching. The microstructure produced included a very fine-grained martensite structure embedded in a ferrite matrix, unlike a control sample that was not cold rolled, and which revealed a coarse-grained structure

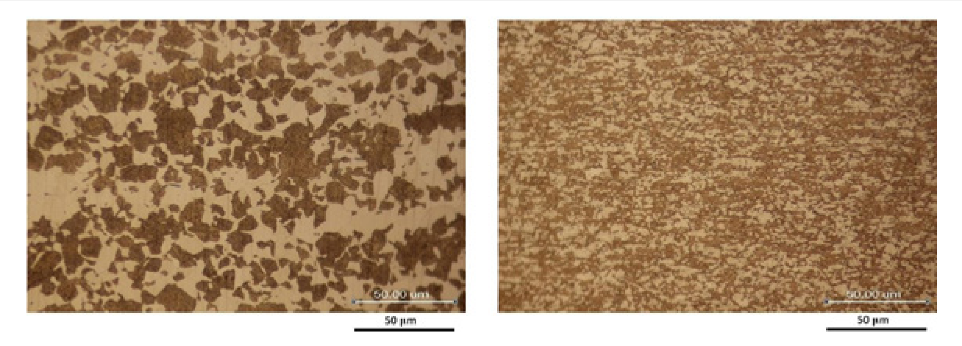

Figure 1.2 Coarse-grained structure.

(Fig.1.2).

Furthermore, the thermo-mechanical route chosen gives rise to the resulting phase distributions. A previously studied example includes the effect of the order of Ferrite formation step in the sequence and that it gives rise to martensite distribution, i.e., if a fully austenized low carbon steel sample goes directly through the partial ferrite transformation through annealing and subsequent quenching, this results in the formation of isolated martensite that are uniformly distributed in the ferrite matrix. On the other hand, if the same fully austenized sample was quenched to a full martensite structure and cold rolled.

Upon heating, the martensite will completely recrystallize into ferrite, and austenite will nucleate at the grain boundaries, giving rise to chainlike network (Fig 1.3) martensite grains upon quenching. [10]

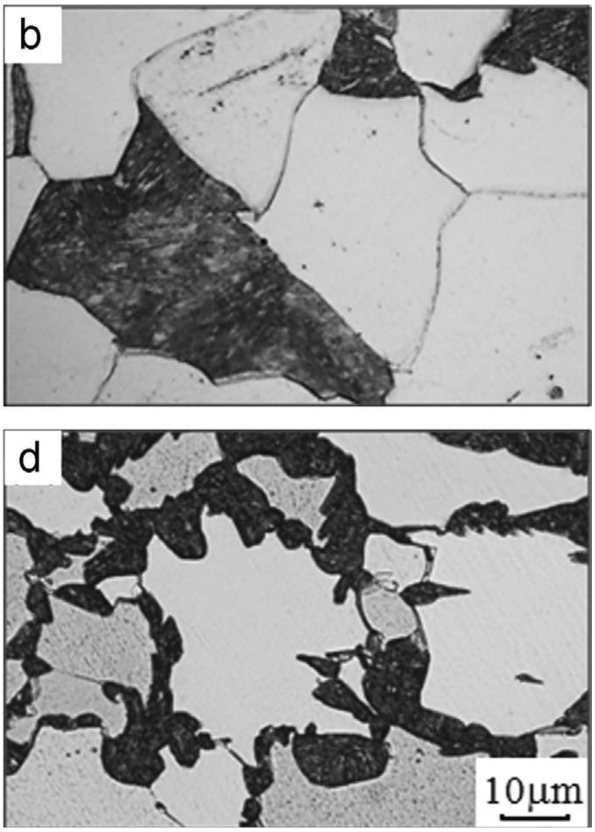

Figure 1.3 Coarse-grained

Properties

- Continuous yielding behavior (no defined yield point);
- A low 0.2 percent offset yield strength;
- A high tensile strength;
- A high work-hardening rate;
- High uniform and total elongations;
- high early-stage strain hardening. [11]

Both Hot and Cold-rolled DP steels offer an incredibly advantageous combination of low yield, high-tensile strength, easy cold working, and weldability due to their ferrite- martensite imbued lattice microstructure. The carbon content of dual-phase steels enables the formation of martensite at practical cooling rates, which increases the hardenability of the steel. Generally, higher carbon will promote a stronger steel and a higher fractional percentage of martensite. In DP steels the soft Ferrite phase is generally continuous, giving these steels excellent formability. When DP deforms the strain is concentrated in the lower strength Ferrite phase surrounding the hard islands of Martensite, which creates the very high initial work hardening rate exhibited by these steels. Due to high-strain hardenability, dual-phase steels also have a high-strain redistribution capacity. This means improved drawability as well as finished part mechanical properties (yield strengths) that are higher than the initial blank. DP steels also have a bake hardening effect that is an important benefit over conventional HSLA type materials. The bake hardening effect is the increase in yield strength resulting from elevated temperature aging created by the curing temperature of the paint bake cycle. DP grades are currently being produced from 500 to 1200 Mpa minimum tensile strength with 5-35% total elongation. [12]

Applications

In automobile Dual-phase steels are often used in the following [13]:

- DP300/500 Roof Outer, Door Outer, Body Side Outer, Floor Panel
- DP350/600 Floor Panel, Hood Outer, Body Side Outer, Cowl, Fender, Floor Reinforcements
- DP500/800 Body Side Inner, Quarter Panel Inner, Rear Rails, Shock Reinforcements
- DP600/980 B-Pillar, Floor Panel, Engine Cradle, Seat Rails
- DP700/1000 Roof Rails
- DP 800/1180 B-pillar upper

Economy

Dual Phase steel is trending through industrial aspects, especially in automotive industries. Many companies around the world are producing different grades for DP Steel. The need for this type of steel relies on The mechanical behavior of Dual Phase steels, which is characterized by continuous yielding, low yield strengths, high work hardening rate, and high stretch formability, characteristics that make Dual Phase steels excellent choice for demanding automotive applications. The strong bake hardening behavior of DP steels add significant strength to a part after paint baking. The low carbon contents of DP steels make them a friendly option for all welding processes including resistant spot, resistant seam, arc, and laser methods. All these features make Dual Phase steel a great choice for demanding automotive applications. [14]

Variant companies produce dual phase steel as main production line such as:

1. Thyssenkrupp
2. ArcelorMittal
3. AKSteel

4 U.S. Steel Košice

Although several products are in service with known mechanical properties that shown in Table 1.2 [15]:

Grade	Yield Strength Mpa	Tensile Strength Mpa	Elongation n in 50mm %	n-value calculated between 4% and 6% strain	n-calue calculated between 10% strain and uniform elongation
CR330Y590T-DP	330-430	590-700	≥21	≥0.18	≥0.14
CR440Y780T-DP	440-550	780-900	≥15	≥0.15	≥0.11
CR590Y980T-DP	590-740	980-1130	≥11	n/a	n/a

Materials and Methods

Effect of Composition

The target behind adding other elements into the composition of dual phase steel is to reach suitable hardenability. The adjustment of hardenability causes a delay in the transformation time of austenite to ferrite or pearlite that reduces cooling rates. Mechanical properties are not affected by adding various alloying elements such as (Cr, Mo, Ni) otherwise, it is significantly changeable with respect to the amount of Carbon inside the composition especially in tempered martensite. So that any mechanical enhancement shows great dependency on the percentage of carbon through the specimen.[16] As a general basis there are two methods in which alloying elements reduce the rate of austenite decomposition. They are being reduced either by the growth rate or the nucleation rate of ferrite, pearlite or bainite. Hardenability is limited basically by the rate of formation of pearlite at the nose of the C curve in the TTT diagram. To understand the effects of alloy elements on pearlite growth it is essential to differentiate between austenite stabilizers (Mn, Ni, Cu) and ferrite stabilizers (Cr, Mo, Si). Austenite stabilizers decrease A1 temperature, while ferrite stabilizers have the typical opposite effect. All these elements are dissolved by substitution in austenite and ferrite phases.

Effect of adding alloying elements

Through Figures 2.1 and 2.2, it is clearly shown the effect of alloying elements (C, Mn) on the transformation kinetics. It also demonstrates the effect of increasing percentage of and how Mn significantly depresses the A1 temperature. Generally, increased

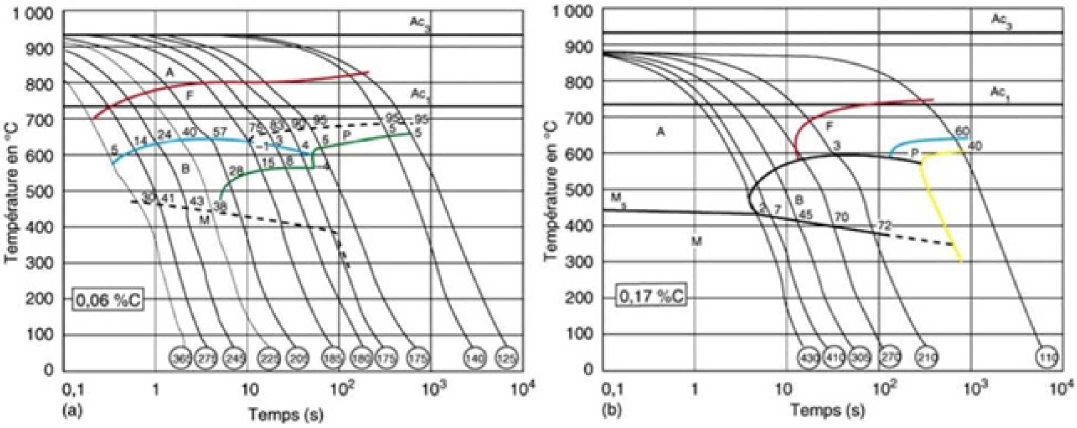

Figure 2.1 Effect of C on CCT diagram

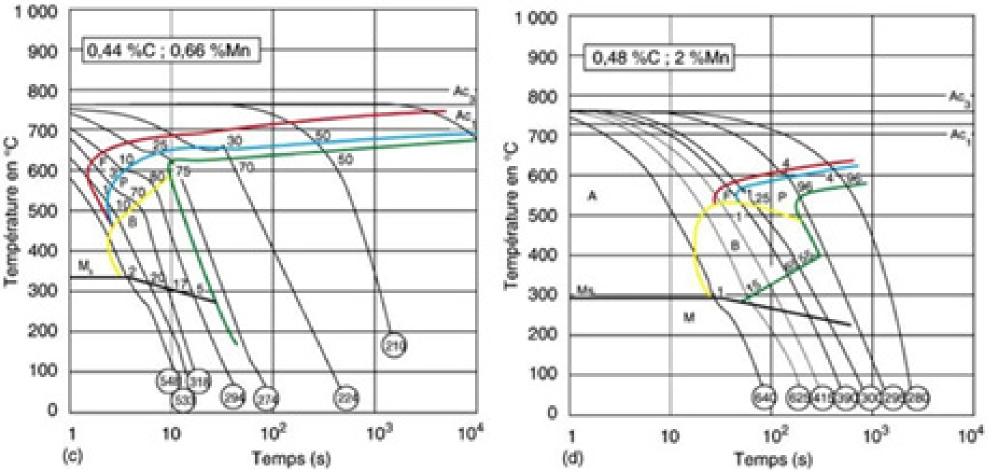

Figure 2.2 Effect of Mn on CCT diagrams

amounts of alloying elements lower the temperature of starting different transformations and reduce the reaction rates. [17]

The numbers on the cooling curves represent the volume fractions transformed in the corresponding range, while the circled numbers at the bottom of these curves are the final Vickers hardness values measured on the specimens. [18]

The role of alloying elements

There are typical alloying elements which are (C, Mn, Si, Cr). Each one of them has its own effects which are summarized in Table. [19]

Alloying Element	**Influence**
C	Determines the amount of second phase (at a given temperature)
	Strong hardenability agent
Mn	Austenite forming element, decreases C solubility in ferrite
	Retards pearlite and bainite formation, increases the transformation temperatures
	Regular solid solution hardening
	Increase the hardness of tempered martensite by retarding the coalescence of carbides, and thus provides a resistance to grain growth in the ferrite matrix
Si	Raises transformation temperatures
	Increases the activity of carbon and hence promotes the ductility of ferrite
	Inhibits carbide formation, particularly Fe3C at the ferrite-martensite interface
	May also contribute to the solid solution strengthening
	Negligible effect on hardenability
	Refines microstructure
	Increases the hardness of tempered martensite by inhibiting the conversion of epsilon carbide to cementite
Cr	Austenite forming element, decreases C solubility in ferrite
	Strong carbide-forming element
	Retards pearlite and bainite formation, increases the transformation temperatures
	Increases hardenability
	Increases connectivity of martensite
	Increases hardness of tempered martensite through alloy carbides formation

Table 2.1 Influence of alloying

During intercritical annealing Mn is portioned between the ferrite and the austenite phases, which results a higher Mn content in the austenite phase that provides further enhancement of the hardenability. [20]

Mn and Si influence the aging behavior of ferrite. The activity of carbon in iron is reduced by Mn and thus reduces the driving force for precipitation relative to binary Fe-C alloys. Mn also forms the metastable ε-carbide, which is more easily nucleated than the cementite. Si has the opposite effect, raising the activity of C and promoting the formation of the metastable carbide. [21]

Effects of alloying elements on austenitising

Alloying elements influence on the austenization through these general consequences:

1. Adding ferrite stabilizer and carbide former alloy elements lead to rise in austenitising temperature.

2. Longer austenitising time is required when alloy or mixed carbides are present because alloy carbides especially Mo take much longer time for dissolution, that's why, higher temperatures of austenitising are a must.[22]

Effects of alloying elements on ferrite phase

Transformation of austenite to proeutectoid ferrite kinetically is retarded generally due to presence of austenite stabilizers or ferrite stabilizers that most probably cause a solute drag migrating interphase boundaries and the rate of diffusion of carbon. [23]

Effects of alloying elements on martensite phase

Martensite phase volume fraction and quantity depends on several factors such as temperature, chemical composition, degree of deformation and the rate of cooling. Elements which increase the hardenability to ease martensite formation are Cr, Mo, Ni, Mn and B.

Andrew's formula demonstrates the effect of adding alloying elements on Martensite start temperature in steels: [24]

$$M_s=539-423C-30.4Mn-12.1Cr-17.7Ni-7.5Mo-11Si$$

Effect of Micro-alloying elements

The behavior of the individual micro alloying elements classifies them in two groups:

(a) The mildly carbide-forming elements, or the elements that do not form carbides.

(b) The strong carbide-forming elements. Generally, the non-carbide forming elements are also γ stabilizers,i.e. they expand the γ field, and the carbide-formers are α stabilizers, i.e. They reduce the γ phase field. Any of the elements in solid solution in α strengthen the ferrite matrix in steel. They contribute differently to hardening, and differ also in the extent to which they reduce plasticity in adding a certain increment to strength [25] P, Si, Mn, Ni, Mo, V, W and Cr, by order of importance, increase hardness of pure Fe. The non-carbide-forming elements in steel are almost wholly dissolved in the ferrite either in pearlitic steels or in the structures of tempered martensite. Ni, Si, Al and much of Mn in steel are dissolved in the ferrite regardless of the C content. The carbide-forming elements are to be found extensively in ferrite only when the C content is insufficient to combine with them. The excess C required to substantially withdraw the element from ferrite is less, the stronger the carbide-forming tendency. The mildly carbide-forming elements Cr and Mn are generally divided between carbide and ferrite, even with relatively high C content. The principle of partition of these carbide-forming elements follows the same general trend when the steel is heated to form austenite, except that the solubility of these elements may be high in austenite and, since the C itself is often largely dissolved, there is frequently no carbide phase left. Only with the stronger carbide-forming elements or in high-C steels are persistently insoluble carbide particles left at high heating temperatures. V, Nb and Ti stabilize C even at relatively low concentration. A classification of substitutional micro alloying elements according to their carbide-forming tendency and their effect on the γ field.

B segregates to austenitic grain boundaries and interacts with dislocation networks. The nose of ferrite is thus driven back to the right on CCT diagrams, i.e. the incubation time is increased. The B effect is only observed when B is present in solution;

- Ti is added to keep the B in solution by forming TiN and avoiding the formation of BN precipitates. The Ti content must be at least 3.4 N to fully stabilize the N. Ti also contributes to the strengthening
- Nb causes a strengthening due to NbC or Nb(C,N) precipitate formation. It also retards the recrystallisation and the grain growth by forming fine carbides. This results in finer grains. In combination with B, the growth of largeFe23 (CB) 6 precipitates at γ grain boundaries is avoided
- Mo is a ferrite stabilizer. It reduces the diffusivity of C in γ. Mo is responsible for a pronounced solute drag like effect (SDLE) due to its segregation to the$\alpha-\gamma$ interface, therefore decreasing the mobility of the $\alpha-\gamma$ interface. Moreover, the ferrite and pearlite formations greatly but does not influence the bainite formation. It is also a solid solution strengthening element
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	γ-stabilizers	γ-stabilizers
Carbide formers	Mo, Cr, V, W, Ti, Nb, Zr	Mn, Co
Non carbide formers	Si, P, Al	Ni, Cu

Table 2.2 General trends of substitutional elements in the Fe-C binary system

Effect of adding Niobium

Generally, in steels, Niobium is widely used for the control of structure development factors as austenisation, phase transformation, precipitation behavior, and grain size. In DP steels, the microalloying of Nb is quite common for its strengthening effect mainly through grain refinement accompanied with some precipitation hardening according to the thermomechanical process conditions.[26] This strengthening mechanism have made Nb among the first choices for microalloying in DP steels as other microalloying elements, Titanium for example, use precipitation hardening, which has a deteriorating effect on ductility (Fig 2.3). It is worth mentioning that in DP steels, Nb grain refinement effect is on both the primary ferrite phase and the secondary martensite phase, all of which have a positive effect on both strength and ductility. In addition, the amount of Nb needed is less than the amounts of Titanium and Vanadium, both commonly used in DP steels, to achieve the same increase in strength. The strengthening mechanism by which Nb functions are further elaborated.

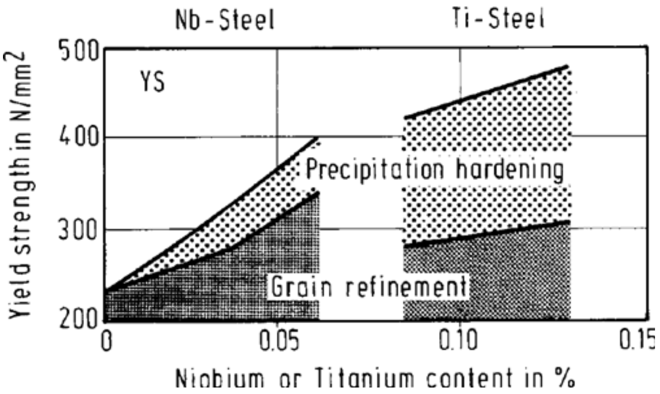


Figure 2.3 Strengthening mechanisms in microalloyed HSLA sheet

As for precipitation hardening, [27] Nb has the potential of precipitating into very fine particles during the annealing process. This has the potential of a significant strengthening effect. To induce grain refinement, Nb retards the recrystallization during austenite to ferrite transformation, and during hot rolling process. Thus, more nuclei are available resulting in a finer microstructure, and in turn, this grain size refinement is still observed in an annealed DP steel sample. In addition, the refinement of grains is accompanied with an increase in homogeneity (Fig 2.3) in the distribution of phases. This is especially important in DP steels to avoid band formations of martensite in the centerline of steel sheets, consequently leading to better formability.

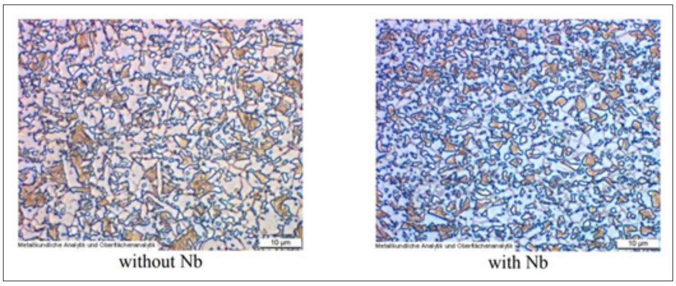

without Nb
with Nb

Figure 2.4 Influence of Nb microalloying on the microstructure of a DP600 steel

The effect of Nb is not limited to refinement of ferrite and martensite grains.[28] The kinetics of phase transformation after intercritical annealing is highly effected by the addition of Nb, namely, it enhances the formation of ferrite at a high cooling rate. this is double effective, as it both increases the volume fraction of ferrite and reduces martensite, but with more available carbon the formed martensite is harder. therefore, for the same strength, a DP steel can be produced with less brittle martensite and more ferrite enhancing elongation.

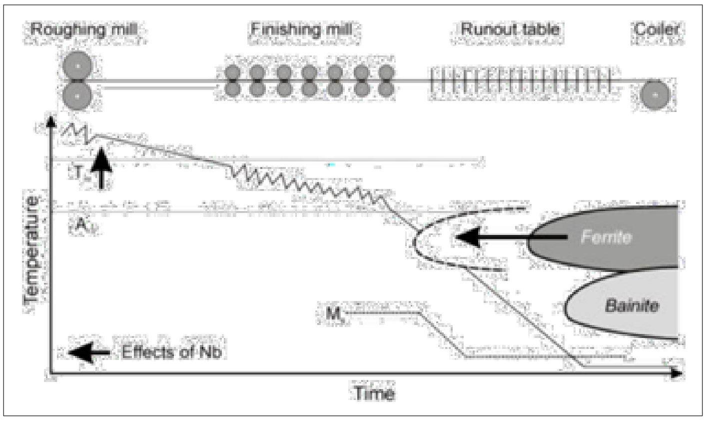


Figure 2.5 Effect of Nb on hot strip processing of DP steel

Mechanical properties are greatly enhanced with the addition of Nb (Fig 2.5). As a result to the changes in microstructure discussed earlier, enhanced mechanical properties arise with the addition of Nb. Tensile and yield strength can increase significantly, up to even 200Mpa with no great loss in elongation. Furthermore, advantages in mechanical behavior can also be noticed. These include a higher resistance against splitting during forming operations and an increased range in the annealing cycle that leads to a better control of steel properties. Also strengthening by Nb can allow for decreasing the C content to below 0.15%, improving steel weldability to produce a better ductile weld seam. Finally, by reducing martensite, there is a less risk of Heat affected zone softening arising in the microstructure, which can develop due to tempering of martensite.

Effect of Thermal Treatment

It is important to stress on the role of different heat treatment processes, which can vary from process to another. It affects the microstructure distribution and the presence of various phases which can be created during solidification. In addition of that, the manipulation of cooling rates and isothermal holding times can significantly change the properties and the microstructure. In the next example, rate of cooling through austenite-to-ferrite transformation point is accelerated, shown by the P-S-K line in the Figure 3.1. The cooling rate is increased in pearlitic steel to about 200° C per minute, which generates a DPH of about 300, and cooling at 400° C per minute raises the DPH to about 400. Thus, the formation of a finer pearlite and ferrite is obtained. [29]

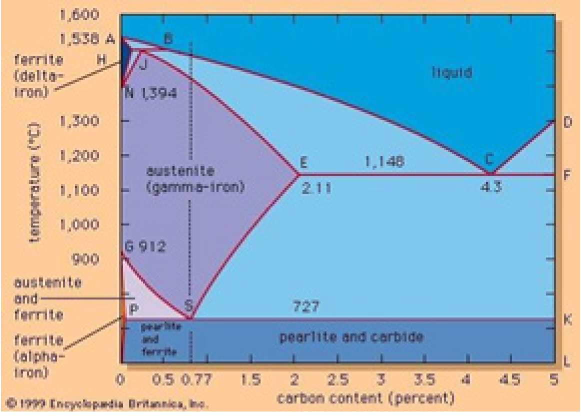


Figure 3.1 Fe-C Phase Diagram

Effect of isothermal holding temperature and time

Isothermal process is a thermodynamic process in which the temperature of a system remains constant.[30] So that, isothermal holding is used to increase amount of certain phase at any region on the TTT diagram. Many investigators have studied the effect of varying heat treatment process on microstructure and mechanical properties. [31]

Ziyong Hou et.al. [32] Have studied the effect of isothermal holding temperature at 250, 300, 350, and 400°C on microstructure and mechanical properties of multiphase steel. Thus, it was discovered that ferrite, bainite, martensite and retained austenite are formed because of adding small percentages of Cr, Nb and Mn with applying isothermal temperature holding.

The effect of mixing small amounts from these elements is the enhancement of hardenability and the features of microstructure. On the other hand, volume fraction of RA has increased that increased the ductility especially high carbon enrichment accompanies stability of RA phase. Also, hard martensite and the dispersion of fine carbides improve the strength. [33]

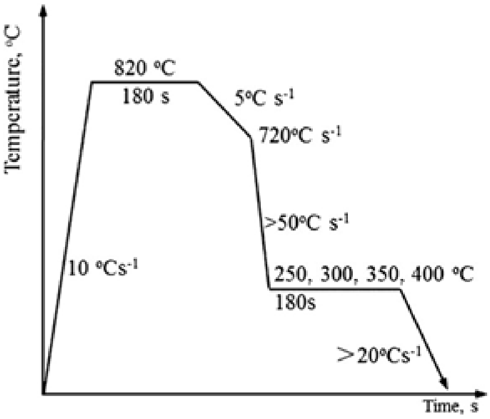


Figure 3.2 Schematic for the intercritical annealing and isothermal holding

From another prospective, M. Balbi et.al.[34] studied the effect of holding time for 5, 15, 30, 60, and 120 min at intercritical temperature (830°C) after step quenching from 930°C for 20 min on the microstructure and mechanical properties of a ferrite-martensite dual phase steel. The results showed that holding time and proportion of the produced phases have no relevance in determining the strength of the samples. In addition of that, the mechanical properties of the obtained samples showed similar values to the commercial dual phase steel.

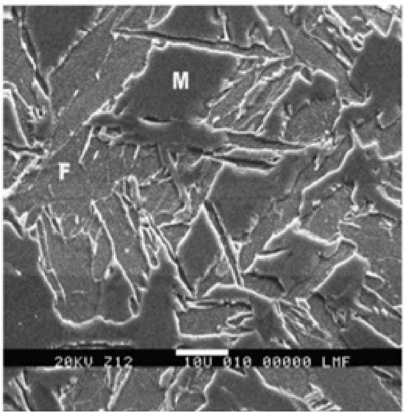


Fig 3.3 Morphology of 120 min isothermal holding

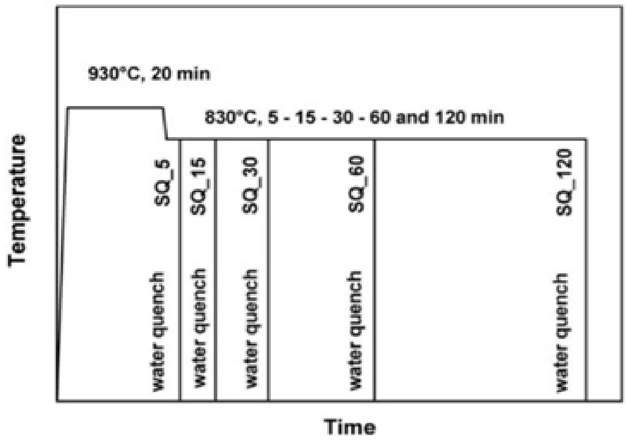


Figure 3.4 Thermal Cycle of the samples

The holding time could not be the same duration of previous studies. E. Fereiduni.[35] has investigated the methodology of improvement the mechanical properties for AISI4140 DP Steel under tough ferrite formation by following specific thermal cycle, in which the samples are annealed to austenite region at 860°C for 60 min then step quenching is applied at 600°C 20-55s . Then the process is finalized with tempering stage at 600°C for 30 min.

The produced microstructure has shown a transformation from hard continuous grain boundary to soft polygonal/quasi-polygonal appearance. As well hardness has been affected in variant ways due to variation of holding time, but in general microhardness values have dropped with increasing holding time from 20 s to 45 s at 600°C. The reason for microhardness results is the formation of tough grain boundary ferrite in conjunction with martensite which cause higher hardness response.

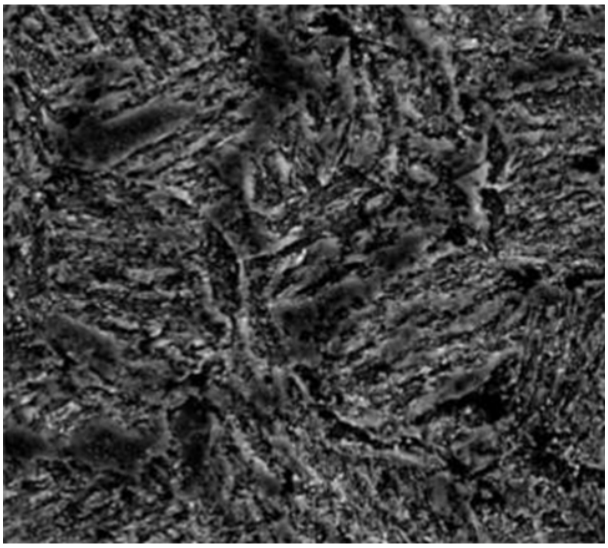


Figure 3.5 DP morphology after tempering

An interesting investigation was performed by Z.P. Xiong et.al.[36] The study was targeting to observe the effect of isothermal holding at 650°C for 100-900 s. It was found that the kinetic of ferritic formation is rapid during the first 100 s, after that rate of formation for ferrite phase decrease with increasing the holding time.

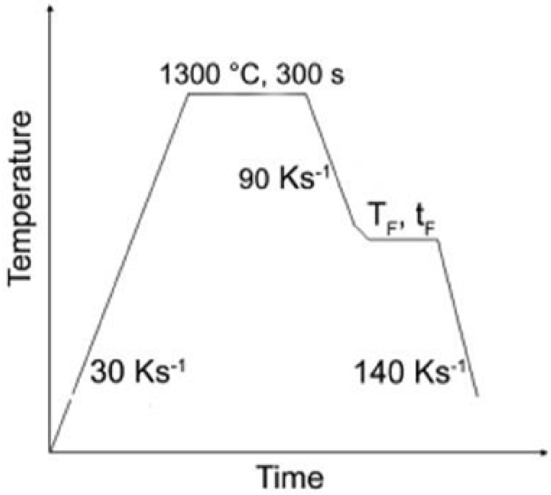


Figure 3.6 Heat Treatment Cycle

The microstructural examination of a 0.08C-0.81Si-1.47Mn-0.03Al (wt.%) DP.

Steel delivered by the research facility simulation of strip casting has indicated that higher holding temperatures and longer holding times are beneficial to avoid Widmanstatten ferrite and also result in a homogeneous distribution of martensite.

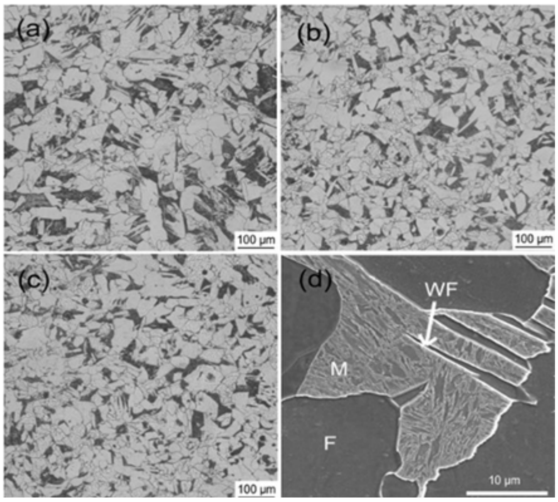


Figure 3.7 Morphology after heat treatment

Bake Hardening effects on Dual Phase Steel

Bake hardening was used on High Advanced Strength Steels for the purpose of increasing the yield strength without changing the thickness of specimen. Firstly, it was developed and applied on malleable low carbon steels to improve its resistance to dentation. Nevertheless, researchers investigated and reported to increase the yield strength by up to 100 MPa by applying Bake hardening with pre-straining to 10%.[37]

The mechanism of strengthening can be divided into three stages for multi-phase or dual phase steels:

1- Formation of Cottrell atmosphere around the mobile dislocations in ferrite. 2- The precipitation of low temperature carbides in ferrite.

3- Formation of clusters and particles in martensite/ bainite.

In addition of that, previous studies have shown that the dislocation substructure formed in polygonal ferrite because of martensite transformation during processing of the DP steel, and the microstructural changes, during pre-straining affect the BH behavior. Hence, there are four major factors, that can affect the bake-hardening behavior: (i) the amount of carbon in solution, (ii) the formation of the dislocation substructure in ferrite, (iii) differences in the kinetics of strain ageing of the phases and (iv) the effect of strain partitioning between soft and hard phases during straining.[38][39]

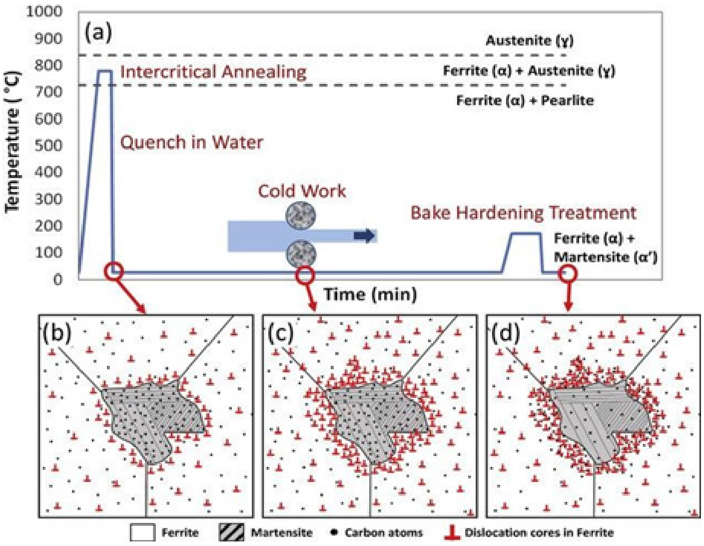


Figure 3.8 Schematic for Bake Hardening effect

Evaluating the bake hardening effect is standerdized with DINEN10325:2006[40], That calculates a value using index called BH index. The average value for DP Steel is 30-60 MPa as BH Index. Following this eqaution according to the noted standard for pre-strain 2%.

$BH_2 = Re_{L,t}$ (or$Rp_{0.2t}$)-$Rp_{0.2r}$

Where $Re_{L,t}$—is the yield strength of lower point when the stress-strain shows a sharp yielding point for the sample after baking; $Rp_{0.2,t}$—the yield strength of the crossing point of 0.2% offset line with the tensile test curve when the stress-strain does not show a sharp yielding point; $Rp_{0.2,r}$—the yield strength of the relevant pre-strain for the sample before baking.

Chun-fu Kuang et.al.[41] studied the effect of pre-straining between 1% to 8% and bake hardening at 300°C on microstructure and mechanical of dual phase steel.

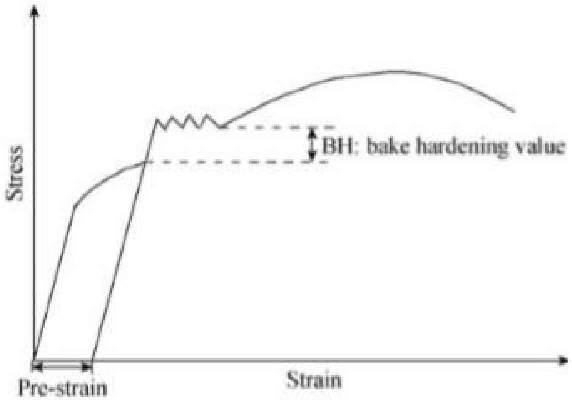


Figure 3.9 Measuring the BH value

It was discovered that the maximum values of BH were between 0% to 1% of pre-straining due to the increase of the Cottrell atmosphere amount, on the other hand, between 1% to 8% pre-straining causes reduction in BH value, because of the unpinned dislocations. Also, increasing bake hardening temperature from 100 to 210°C, led to slightly increase in BH value due to the formation of Cottrell atmosphere and the precipitation of carbides in ferrite. However, higher baking temperature in the range from 210 to 300°C causes a significant improvement in the BH value because of the precipitation of carbides in both the ferrite and martensite phases. Increasing Baking time from 10 min to 100 m enhances BH value.

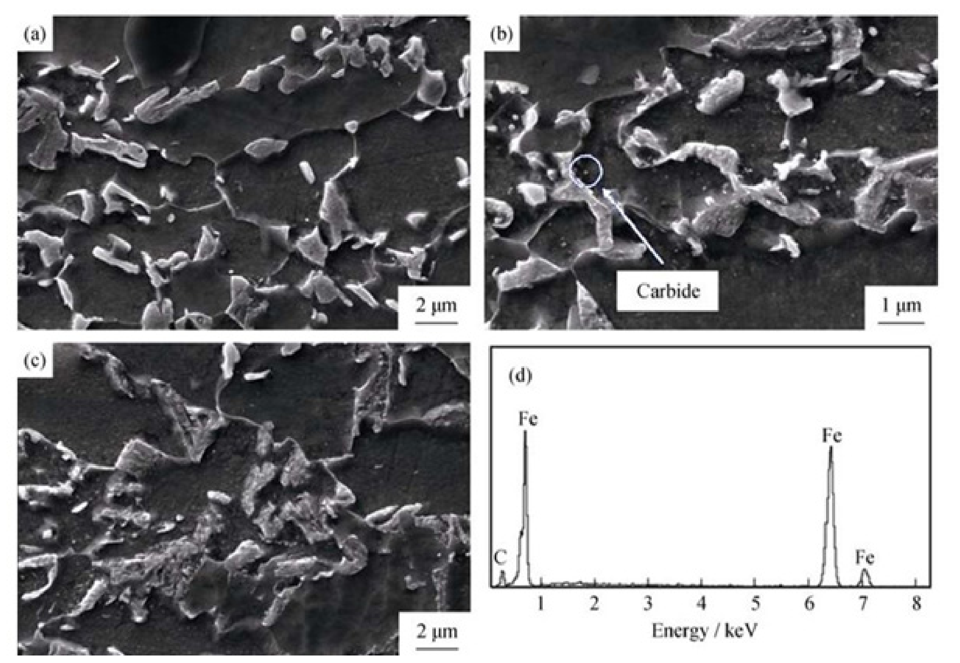


Figure 3.10 Effect of baking temperature on the microstructure evolution of the DP steel: (a) 100°C, (b) 250°C, and (c) 300°C; (d) EDS spectrum of the specimen showing the presence of fine carbides

Mehdi Asadi et.al.[42] has run out an investigation of the influence of load paths whether it is uniaxial or biaxial loading, on the microstructure and mechanical properties of dual phase steel. The drawn conclusions are:

- Type of the load path bake hardening conditions and degree of pre-straining are controlling the local mechanical strength of the BH.
- Higher strength value is noticed for biaxial condition at the same level of pre-strain.
- A significant higher strength increased is present after baking at 170 and 240°C for 20 min. This increment is higher for the baking condition with 240°C and 20 min.

Mustafa Türkmen et.al.[43] directly studied the effect of bake hardening at 180°C for 10-160 min, after 4% tension pre-strained. On the different morphologies whether fibrous or bulky of high martensitic dual phase steel by varying its volume fraction. To obtain fibrous martensite two step quenching is performed. For instance, in that research, the specimen was normalized at 980°C for 30 min then quenched, followed by heating at 820°C for 30 min then quenched in cold water. From another aspect, The bulky martensite is produced by annealing at 980°C for 30 min then cooled in air till 820°C and kept at this temperature for 30 min, at the end the specimen is quenched in cold water.

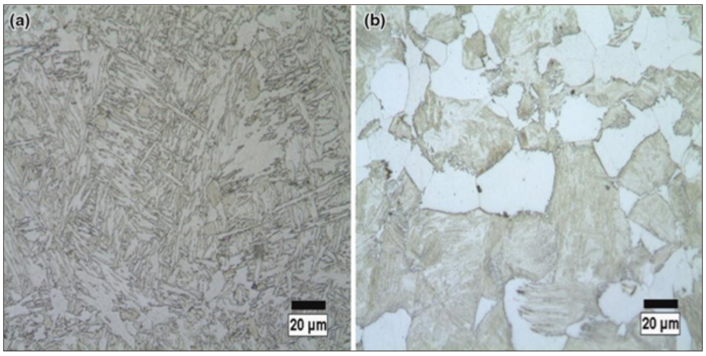


Figure 3.11 Microstructures of dual-phase carbon steel intercritically annealed at 775 °C: a) fibrous martensite, b) bulky martensite

Thus, the effect of BH of the specimens at 180°C for 20 min with fibrous and bulky martensite increased LYS and UTS, but. decreased the total elongation and reduction in area. However, fibrous morphology offered the best balance between the strength and ductility as desired for dual-phase steels. On the other hand, when the BH time increased to 160 min.

LYS and UTS of fibrous and bulky martensite decreased. Nevertheless, total elongation and reduction in area increased. Because of The formation of carbides inside the martensite during the over-ageing contributes to the decrease in the strength. So, a correlation between martensite volume fraction and BH effect could be concluded. Increasing martensite volume fraction increases the BH values because of stronger dislocation pinning at ferrite martensite interface.

Effect of Austenization

To obtain dual phase steels, there are several processes that can be used, such as intermediate quenching, intercritical annealing and step quenching [44].

In the intermediate quenching (IQ) heat treatment, dual phase steels are reheated to α+γ phase region and then quenched from the austenization temperature. IQ process results in dual phase microstructure of fine fibrous martensite and ferrite. During the reheating, a partial austenization occurs and a microstructure of austenite and ferrite is obtained. After quenching, the austenite transforms to martensite, while ferrite remains in final microstructure. Austenite is formed by nucleation and growth mechanisms during the partial austenization [45]. Various nucleation site for the austenite from initial martensitic microstructure is possible such as prior austenite grain boundary, carbide precipitations in prior austenite grain boundary, packet and block boundaries and lath boundary [46]. Lath boundaries are low angle boundaries and their interfacial energy is relatively low. Thus, it is not a preferred nucleation site for austenite.

Nakada et al. [47] concluded that the partial austenization temperature affects the nucleation site for austenite. At the relatively higher temperatures, it becomes 40% of total austenite that nucleates at prior austenite grain boundary, while it is slightly seen at lower temperatures. Nucleation site also affect the austenite morphology, acicular austenite grains are related to the nucleation at lath boundaries, while granular austenite grains are formed at prior austenite grain boundaries. Hara et al. [48] claimed that nucleation behavior of austenite is related to heating rate and higher heating rate results with granular austenite which nucleates at prior austenite grain boundaries while lower heating rates results with acicular austenite. After rapid nucleation of the austenite with carbon diffusion, it continues to grow with carbon and manganese diffusion. Diffusion rate of Mn in austenite is 3 times slower than in ferrite.

Therefore, the growth mechanism goes on slower than the nucleation mechanism when the substitutional elements start to contribute the partitioning between austenite and ferrite [49]. IQ method provides a preferable scattered, homogenous, and fine martensitic final microstructure, along with refining the matrix. This is mainly the reason why dual phase steels produced by IQ method microstructure is known to have a fibrous microstructure.

Ersoy et al. [50] concluded via SEM photomicrograph of sample IQ950-5 (Fig.19) showed after the microstructural investigations that two types of martensite morphology can be found in microstructure as martensite island and martensite fibers.

During the partial austenization and decomposition of initial martensite microstructure, the nucleation of austenite occurs along the lath boundaries as well as in martensite laths, blocks, and packets. This results in a distribution of fine fibrous martensite in the ferrite matrix as indicates with black arrows in Fig.3.12. As mentioned before, prior austenite grain boundaries are also the preferred nucleation sites for austenite. Thus, martensite islands are also found in the resulting microstructure.

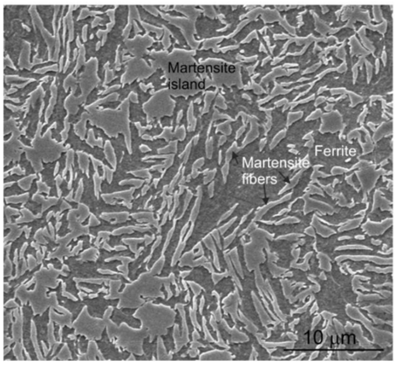


Figure 3.12 SEM photomicrograph of sample IQ950-5

It is concluded that the nucleation site determines the morphology of martensite. On the nucleation at the prior austenite grain boundary, martensite grows without encountering a barrier and grows easily compared with the one nucleate at martensite laths, blocks, and packets. In other words, nucleation at prior austenite grain boundary results with coarse martensite islands, while finer and irregular distributed martensite fibers occurred on the nucleation at martensite laths, blocks, and packets.

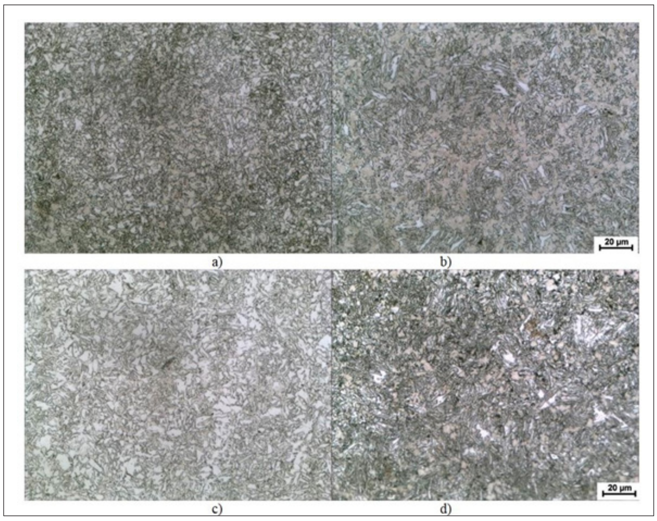


Figure 3.13 I.M micrographs of samples (a) IQ900-5 (b) IQ950-5 (c) IQ1000-5 and (d) IQ1000-15

The effect of austenization temperature on the size distribution of martensite fibers and islands was studied and found that the average size is hardly affected with the austenization temperature and measured for 900 °C, 950 °C, and 1000 °C is 1.570 µm, 1.428 µm, and 1.426 µm, respectively (Fig.3.13, a, b, and c). The size of martensite fibers is measured between 0.2 to 2 µm. Martensite islands have very low frequency and their size measured between 2 to 10 µm. According to this knowledge, the amount of martensite islands decreases with increasing austenization temperature. This may be due to higher amount of prior austenite grain boundaries at lower austenization temperatures. With the increasing austenization time from 5 min to 15 min (Fig.3.13, d), it was also similarly observed that the amount martensite islands decreased.

The effect of Cooling Rate

The cooling rate parameter is a very important factor in any metallurgical process, especially in DP steels production. As it affects greatly the kinetics of transformation between phases, it is of great importance to investigate the cooling rate effect in DP steels, and the way it induces microstructural changes when cooling rate is varied during the production process, especially in the final quenching step.

Marion Calcagnotto et al. [51] investigated the effect of cooling rate. Low carbon samples were heated at an intermediate heating rate of +20 K/s to 730°C, held for 1 min and then cooled at rates between –140 K/s and –20K/s. It was found that during cooling, ferrite grows at the expense of austenite, which is accompanied by an increase in the austenite carbon content. Thus, the higher the cooling rate (i.e. shorter cooling time between intercritical annealing temperature and room temperature), the higher the martensite fraction and the smaller the ferrite grain size. This tendency is clearly revealed in Fig.3.14.

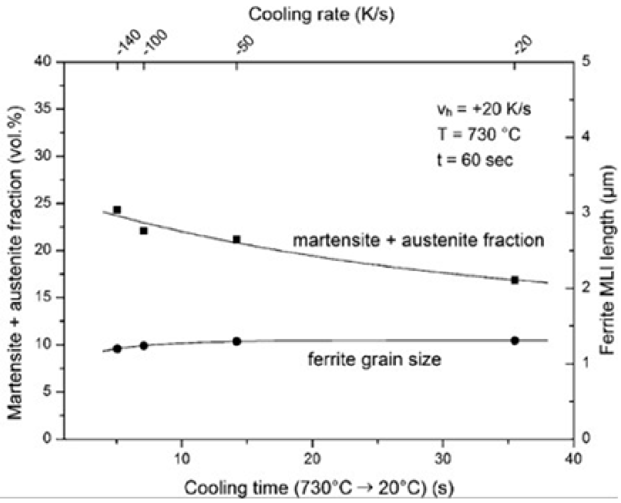


Figure 3.14 **I**nfluence of the intercritical annealing parameter (cooling rate) on martensite + austenite volume fraction and ferrite grain size

The micrographs in Fig.3.15, (g) and (h), further show the reduction in martensite fraction when the cooling rate is lowered. Comparing these micrographs, one can see that lowering the cooling rate leads to a finer martensite distribution, i.e. the martensite island size is reduced and the percolation of martensite, which occurs occasionally in the rapidly quenched DP steel, is impeded.

In the sample cooled at –20 K/s (Fig. 22, h), some bainitic areas are visible (white arrow), indicating that–20 K/s falls slightly below the critical cooling rate for complete austenite-to- martensite phase transformation. A high martensite fraction again coincides with a more equiaxial ferrite grain shape.

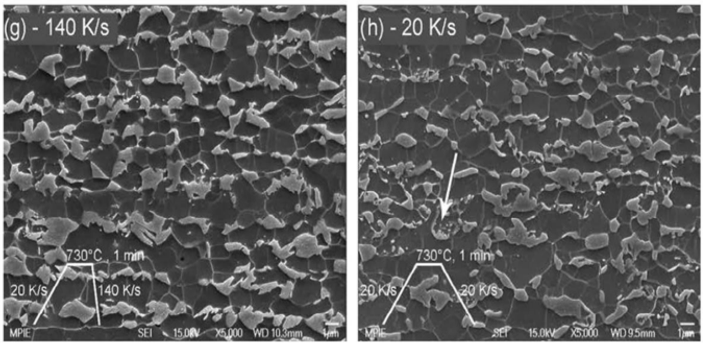


Figure 3.15 Influence of cooling rate on micro-structure evolution

As austenite is enriched in carbon during slow cooling, more effective austenite stabilization was anticipated. Indeed, the highest retained austenite fraction (1.7 vol.%) is present in the sample cooled at 20 K/s. The hardenability of the steel is fairly high, even at a moderate cooling rate of 20 K/s, the main part of the second phase fraction is martensitic. Furthermore, the process of carbon enrichment in austenite during slow cooling (due to the growth of ferrite at the expense of austenite) is more effective when the diffusion distances are small, as it is the case in ultrafine grains. Thus, the higher carbon content of the remaining austenite leads to an improved hardenability.

Time does not lead to significant changes in ferrite grain size, whereas austenite grows continuously. This indicates that during isothermal holding, ferrite grain growth continues. However, ferrite grain coarsening is balanced by the continuous growth of austenite into the adjacent ferrite. The limited mobility of ferrite grain boundary described above contributes to the stability of the ferrite grain size.

With the onset of austenite formation, ferrite grain growth is slowed down. In contrast to ferrite, the martensite island size increases significantly with increasing holding temperature and holding time, whereas it decreases with decreasing cooling rate.

Also from another perspective, the most general manufacturing method of DP steel is the continuous annealing method in which sheet steel is heated to ferrite – austenite phase and is subsequently cooled to that some fraction of austenite phase transforms to martensite. the properties of the sheet steel manufactured by this method are influenced by heat treating condition and in particular by cooling condition. The cooling rate required for the formation of the dual phase depends on the stability of the austenite phase formed during heating, i.e., chemical composition and heating temperature and time. In general, the cooling rate must be increased with decrease in the stability of the austenite phase.

In addition, Koichi Hashiguchi et al. [52] found that for a steel with a given chemical composition, there is a minimum cooling rate (critical cooling rate) required for obtaining DP structure. This critical cooling rate can be predicted from CCT curve during cooling from ferrite – austenite range.

Fig.3.16 shows the CCT curve of 1.2%Mn – 0.5%Cr steel as an example of calculation. It is clearly seen from this curve that the whole austenite phase formed during heating transforms to martensite at the cooling rate larger than about 200 oC/s. When the cooling rate is less than 200 oC/s a part of the austenite phase changes to ferrite, pearlite and bainite during cooling, so the final fraction of martensite phase decreases.

When the cooling rate is less than 7-3 oC/s, the decomposition of the austenite phase is completed in the high temperature range, and martensite phase cannot be obtained. That is, in the 1.2%Mn- 0.5%Cr steel the critical cooling rate required for obtaining dual phase is about 7-3 oC/s. The critical cooling rate obtained by this calculation is in good agreement with the experimental result.

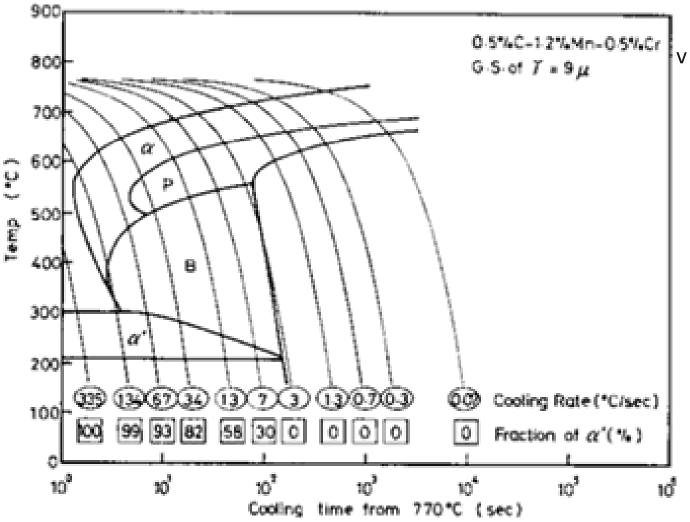


Figure 3.16 calculated CCT curve of 1.2%Mn – 0.5%Cr steel

Fig.3.17 shows the relation between cooling rate and fraction of martensite in cooled condition. The critical cooling rate decreases with the increasing contents of Mn, Cr and Mo. In 1.2%Mn – 0.5%Cr steel, the fraction of martensite phase produced by the critical cooling rate of 5 oC/s is considered to be the minimum amount required for the formation of dual phase steel. In other alloy systems too, the cooling rate required for obtaining this critical fraction of martensite phase is regarded as the critical rate.

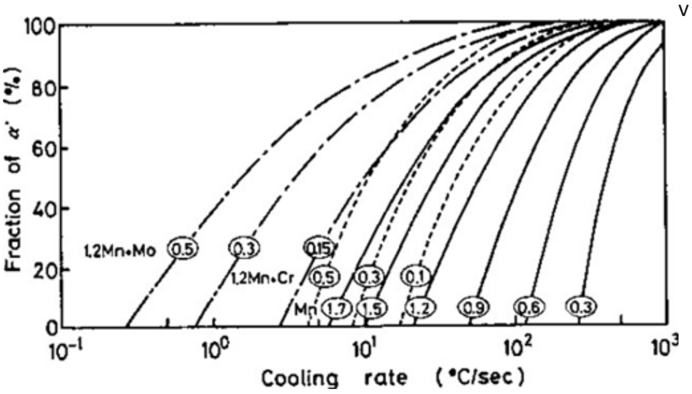


Figure 3.17 Calculated relation between cooling rate and fraction of martensite in prior austenite

Fig.3.18 shows the relation between the critical cooling rate and alloy content as clearly seen in this figure, logarithm of the critical cooling rate decreases in a linear manner of each alloying element on the critical cooling rate converted into Mn content from curves in this figure The effect of Mo on the critical cooling rate is the largest followed by Cr, Mn, Cu, Ni and Si.

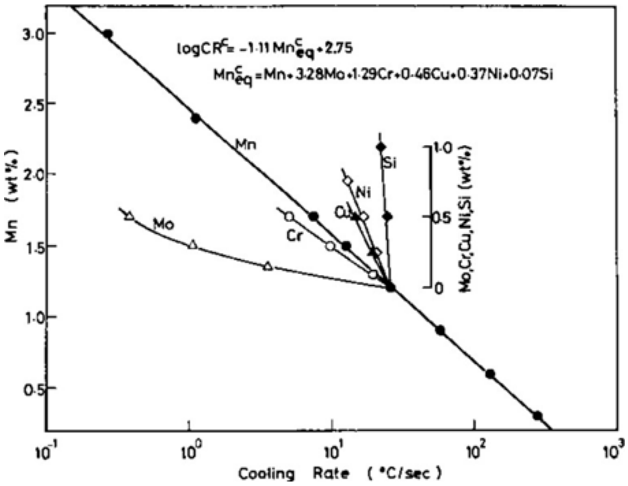


Figure 3.18 Calculated relation between critical cooling rate and alloy contents

The effect of cooling rate on the tensile properties of Cr-added steels is shown in Fig.3.19, The tensile strength increases, and the total elongation decreases monotonically with increasing cooling rate. The critical cooling rate at which the yield point elongation diminishes to zero, decreases with increasing Cr content. At the critical rate both the yield strength and the yield- to-tensile strength ratio (yield ratio) diminish. At any cooling rate both the yield strength and the yield ratio decrease with increasing Cr content. At higher cooling rates, the yield strength increases in a linear manner with cooling rate.

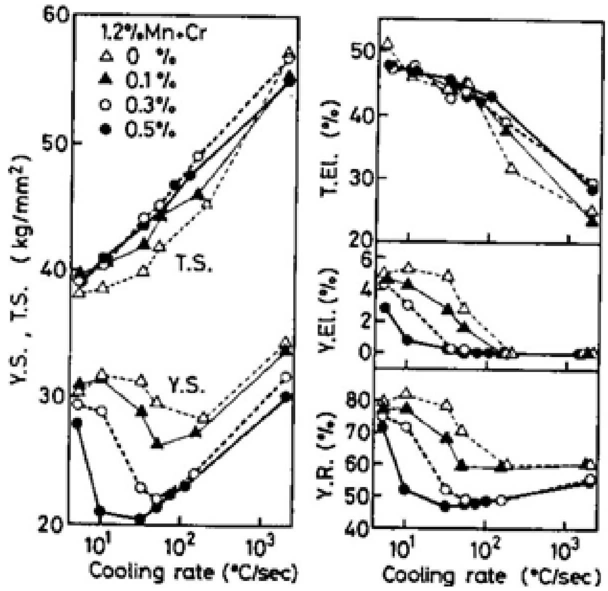


Figure 3.19 Effect of cooling rate on yield and tensile strengths, yield ratio, yield point elongation and total elongation in Cr bearing steels

Fig.3.20 shows optical micrographs of specimens subjected to various cooling rates. When the cooling rate is less than a critical value, the microstructure consists of ferrite and pearlite but when the cooling rated exceeds a critical value the microstructure consists of ferrite grains dispersed with fine martensite islands, regardless of cooling rate. The ferrite-pearlite structure exhibits yield point elongation, a relatively high yield strength and a large yield ratio whereas the ferrite plus martensite duplex structure exhibits continuous yielding, a lower yield strength and a lower yield ratio, with a very good correspondence between microstructure and tensile behavior.

A similar behavior is found for the Mn or Mo added steels, with an increase in Mn or Mo content the microstructure changes from ferrite-pearlite to ferrite-martensite. This change is accompanied by a suppression of the yield point elongation and a lowering of the yield strength and yield ratio.

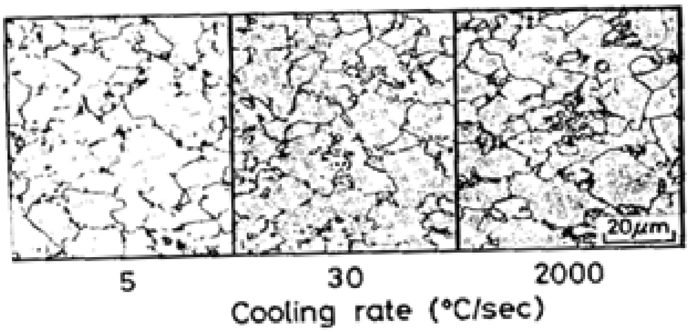


Figure 3.20 Effect of cooling rate on microstructure

Thermomechanical processing route

Apart from effect of the formerly discussed thermal treatment parameters in DP steel development, i.e. isothermal temperature and duration, the influence of the thermomechanical treatment route is essential to the understanding of the resulting properties and structure. In dual phase steels specifically, the thermomechanical treatment parameters include the types and parameters of rolling involved, along with the order of the involved thermal and mechanical processes.

Traditionally in dual phase steel production, hot rolling is carried out immediately after intercritical annealing and followed by quenching. In a previous work, the effect of the amount of reduction in thickness during hot rolling was investigate by E. Ahmad et al. [53] 0.9-C low alloy steel samples were hot rolled to lower thickness to 0%, 20%, 30% and 50%, immediately after all the samples were heat treated at 790 oC for 20 minutes, and the rolling was followed by quenching in iced brine solution. It was concluded that rolling in the intercritical temperature range decreases the hardenability of the steel as it increases the volume fraction of ferrite, this is maybe explained by the fact that the newly formed epitaxial ferrite was allowed to grow from existing ferrite as a result of the increased interfacial area of ferrite due to grain elongation in rolling. Therefore, the volume fraction of austenite decreases leading to less martensite upon quenching and less overall hardenability (Fig.3.21). In addition, it was also noted that the strength increased when the samples were subjected to hot rolling, and this increase can maybe be attributed to the formation of the substructure in ferrite and the fibrous microstructure achieved in which better stress transfer between the matrix and martensite is achieved due to more area of contact.

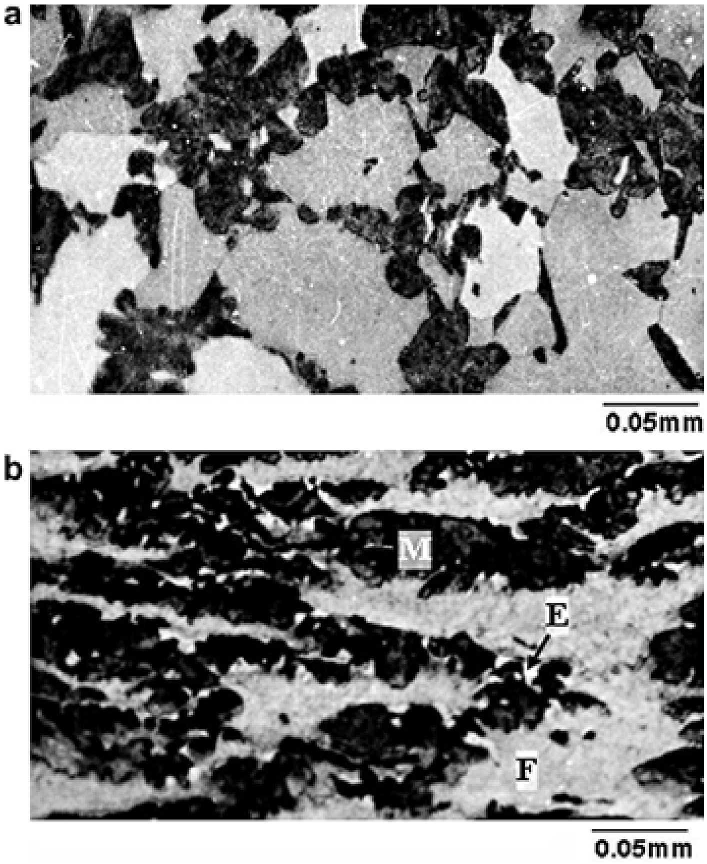


Figure 3.21 The microstructure of: (a) OL and OT with equiaxed ferrite and martensite and (b) 50L and 50T, with elongated fibers. M = martensite, F = ferrite and E = epitaxial ferrite

Another important parameter during rolling is the rolling speed used. A study by S. Serajzadeh et.al [54] has concluded the effects of ranging hot rolling speed on the microstructure of DP steels. It was found that with the increase of rolling speed, i.e. strain rate, ferrite volume fraction increases, and a finer microstructure is achieved (Fig.3.22). Also, an increase in total ductility is noted with increasing the rolling speed via a change in the stress-strain behavior (Fig.3.23) of the samples used. This is mainly attributed to the effect of the strain rate on nucleation sites, that is, the deformation helps introduce more nucleation sites for ferrite allowing for more uniform and finer microstructure. This is particularly true for steel with a high-volume fraction of ferrite.

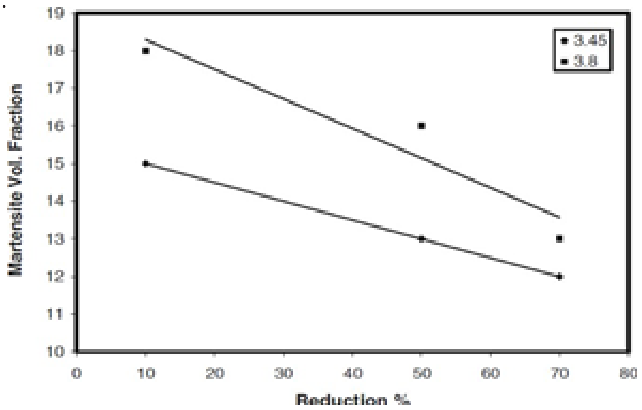


Figure 3.22 Effect of strain rate on martensite volume fraction with respect to reduction

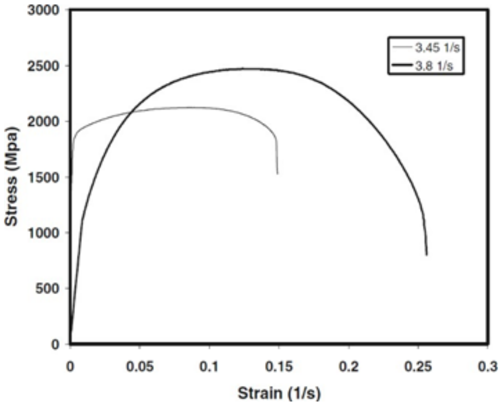
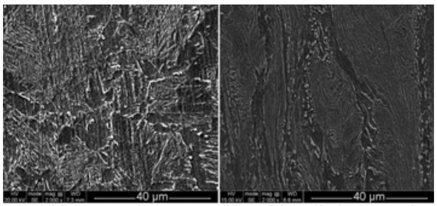


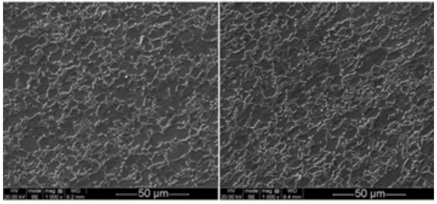
Figure 3.23 Effect of strain rate on stress- strain behavior of dual phase steel

Rolling parameters understanding leads us to the point of understanding the significance of the order of mechanical, thermal, and thermomechanical processes involved in the production of DP steel. This has been investigated in many works mainly from the point of view of the initial processing and microstructure prior to the traditional thermomechanical treatment used in DP production.

For instance, previous works investigated the methods for grain refinement of Dual phase steels using via altering the initial microstructure by processing the steel samples before subjecting them to intercritical annealing and quenching. A work by H. Mirzadeh et.al [55] explored the effect of cold rolling complete martensite samples prior to annealing and quenching, instead of the usually employed ferrite-pearlite initial microstructure. Three Samples of the composition 0.12C–1.11Mn-0.16Si were heat treated. One was intercritically annealed at 820 C and quenched, the other two were fully austenized and quenched to form martensitic structure and then cold rolled up to a 70% reduction in thickness, followed by intercritical annealing at 820 C and 850 C respectively. Both of the prior cold rolled samples exhibited enhancements in tensile strength and strength ductility relationship due to a great increase in grain refinement in comparison with the sample of ferritic-martensitic initial microstructure. The grain refinement potential of this technique can be explained by noticing the microstructure of the cold rolled martensite prior to the treatment (Fig.3.24).



(a) Martensite (b) Cold-Rolled Martensite



(c) DP2 (d) DP3

Figure 3.24 SEM images of the (a) Martensite, (b) Cold-rolled martensite, DP2 steel, and (d) DP3 steel.

The presence of a lamellar structure in the direction of rolling can become the base for formation of finer ferrite grains along with many more austenite nucleation sites that will in turn result in a finer overall structure. As a result, ferrite grains of 5 um or less are observed with fine martensite islands. To elaborate, the prior cold rolled sample annealed at 820 C exhibits 560 MPa and 45% elongation in comparison to 510 MPa and 31% elongation in the sample of ferritic-martensitic initial microstructure.

Similar approach was explored by Y. Mazaheri et.al [56], where duplex ferrite martensite samples were cold rolled to achieve 80% thickness reduction, followed by intercritical annealing and quenching in order to produce ultra-fine-grained dual phase steel (Fig.3.25). It was concluded that the new DP steel (formed by intercritical annealing of cold-rolled duplex microstructure at 790 C for 8 min followed by water quenching) showed an excellent combination of ultra-high strength (1430 MPa tensile strength) and adequate ductility (12.6% total elongation). The developed DP steels showed superior strength–elongation balance (110– 130 J/cm3) in comparison with the commercial modern high strength automobile steels, i.e., DP800 and DP980.

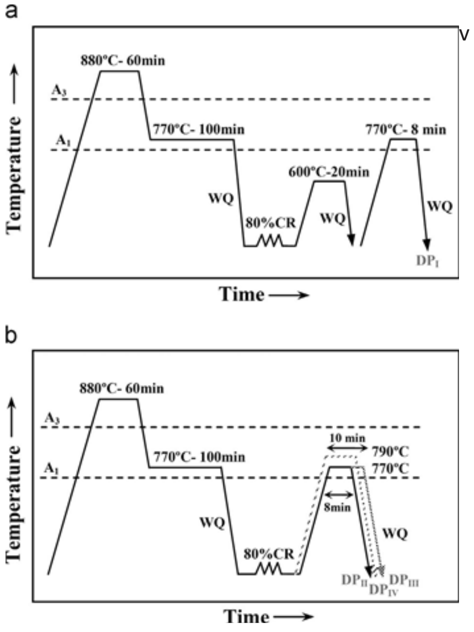


Figure 3.25 Thermomechanical treatments developed to produce

UFG-DP steels A1: start and A3: finish temperature of austenite formation during heating; WQ: water quench; CR: cold rolling.

In the same line of exploring thermomechanical route effect, A. K. Panda et.al [57] previous work explored how changing routes specifically in thermal treatment can influence the product. Fe – 0·08C–1·00Si–1·21Mn–0·02P–0·012S–0·42Cr–0·41Mo steel samples were subjected to two different treatment routes (Fig.3.26). In the first route, the samples were heated up to 1000 C and held for 10 min and then let to cool to reach intercritical annealing temperature (780, 210, and 840), then immediately rolled and water quenched after the rolling. Whereas in the second route, the samples were heated to reach one of the intercritical annealing temperatures and held for the same time (10 min), followed immediately by rolling and quenching. The results indicated that the first treatment route induced more formation of ferrite upon quenching, mainly due to the deformation of austenite as the rolling was carried out after austenizing the samples. In addition, more martensite was formed in the second route samples because of the transformation of ferrite and pearlite phase prior to quenching.

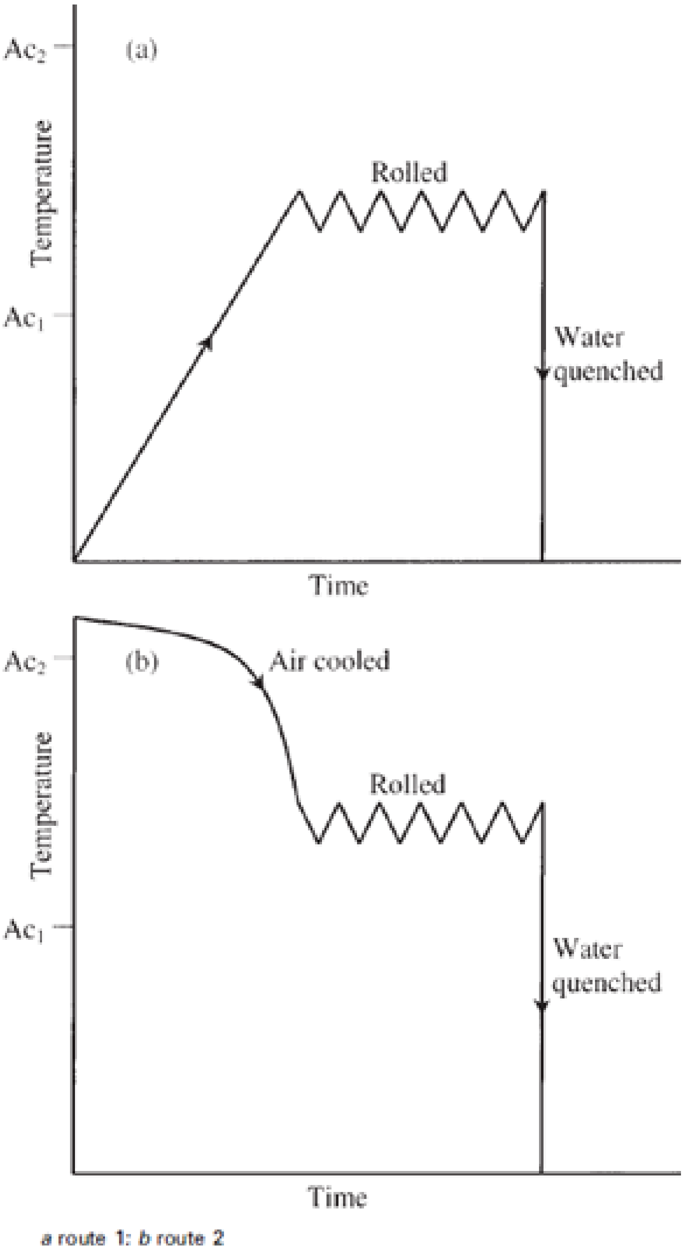


Figure 3.26 a) route 1; b) route2

Relationship of Structure Types and Morphology with Mechanical Properties

Trends in dual phase steel mechanical properties and their relation to structure and morphology are in many ways difficult to observe because of the variety of microstructures in different works, making it hard to compare results given the number of varying factors involved in each case. However, a work by Tasan et al. [58] summarizes some observable trends from various works (Fig.4.1). Noticeable trends included a general increase in both martensite volume fraction and martensite carbon content when grain refinement is carried out. Also, an increase in martensite volume fraction and martensite carbon content increases the strength while decreasing ductility.

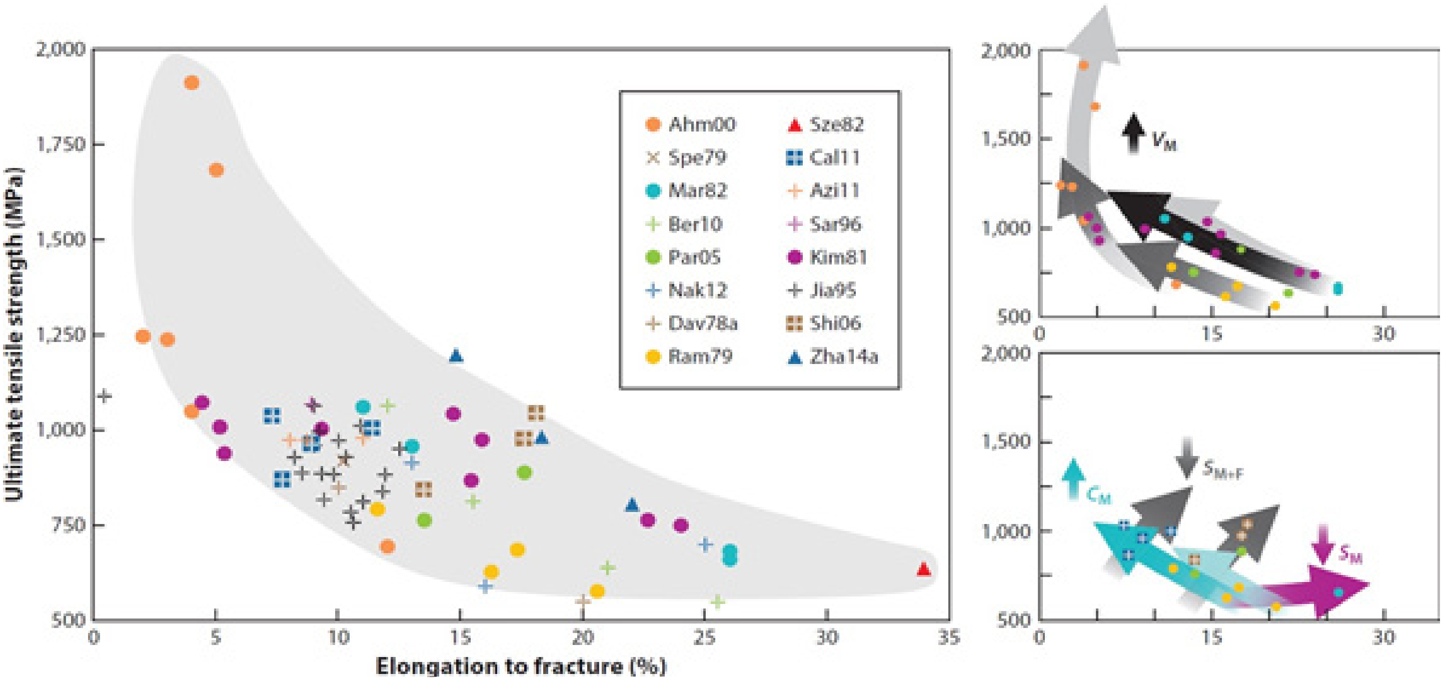


Figure 4.1 The UTS-ETF combinations of all DP steels [49]

Tasan et al. also explored results of one of the few studies discussing mechanical properties trends in DPS using a numerical approach.

The most important parameters of structure-property relationship in regard with DPS, namely, martensite volume fraction and martensite carbon content and the microstructure morphology were investigated by A.P. Pierman et.al [59]. It was deduced that increasing martensite volume fraction gave an increase in tensile strength and also gave rise to a larger yield value, this behavior is an expected one according to the composite model that DPS are explained upon, in that there are hard martensite grains in a soft ferrite matrix. A general increase in ductility was noticed with the increase of martensite volume fraction for the same martensitic carbon content, and similarly increasing carbon content of martensite increases ductility for the same martensite volume fraction.

However, this general trend fails to apply when combining both large volume fractions and large carbon content of the martensite, and it can be explained by damaged induced fracture or softening. It was also concluded that increasing general carbon content increases strength and hardness generally in DPS, however, the carbon content has no effect on the initial yield point, this behavior can be attributed to the fact that flow response of martensite is affected by carbon content.

Regarding morphology, expectedly it was found that at a given volume fraction and composition, morphology significantly influences the mechanical behavior. Also noting that equiaxed microstructures exhibited an increase in strength and a decrease in ductility, when compared to specimen of fine and dispersed elongated microstructure. This is a result of the impediment of ferrite plastic flow by martensite particles that act as a continuous network along grain boundaries.

Conclusion

In this work, we explored the current understanding of the term Dual Phase steel through the latest literature on the subject. The need for light and strong steels was discussed along with the economic importance that gives rise to the evolution of research in Advanced high strength steels like DP. The chemical composition, microstructure, morphology of phases present and the effect of processing, alloying and microalloying were all discussed in that how they affect the final properties. The full potential of DP steels is not yet fully explored. This can be attributed to the number of parameters and changes that can be done during the production thermomechanical processing route, and other factors including altering the micro-alloying elements and compositions, initial microstructure and the thermomechanical route. The effects of using any combination of these factors leads to a change in microstructure as expected in any metal making process. Generally concluding, according to previous works, changes in mechanical properties of DP are mainly due to changing the percentage of phases (mainly ferrite and martensite) and their morphology, more specifically, the effect of grain size and increasing fineness of the microstructure. Increasing Martensite volume fraction increased the tensile strength and hardness of the steel present, while inducing a finer microstructure increased toughness and improved the DP steels strength ductility balance. More research and exploration on the subject are highly encouraged, given all the promising results.

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