İSTANBUL TECHNICAL UNIVERSITY ★ INSTITUTE OF SCIENCE AND TECHNOLOGY

OPTIMIZATION OF ALLOYING ELEMENTS AND PROCESS PARAMETERS FOR EXTRA DEEP DRAWABILITY PROPERTY OF COLD ROLLED AUTOMOTIVE STEELS

Ph.D. Thesis by

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ABBREVIATIONS

IF	**:** Interstitial Free
CAL	**:** Continuous Annealing Line
CGL	**:** Continuous Galvanizing Line
CPL/TCM	**:** Continuous Pickling and Tandem Line
CQ	**:** Commercial Quality
DQ	**:** Drawing Quality
DDQ	**:** Deep Drawing Quality
SRT	**:** Slab Reheating Temperature
PF	**:** Polygonal Ferrite
CT	**:** Coiling Temperature
SCWE	**:** Secondary Cold Work Embriddlement
SEM	: Scanning Electron Microscope
EDS	: Electron Dispersive Spectroscope
CE	: Carbon Equivalent
BOF	: Bazic Oxygen Furnace
Ti*	: Excess Titanium
Nb*	: Excess Niobium
RTS	: Radian Tub Soaking
GJC	: Gas Jet Cooling
RQ	: Roll Quench
OA	: Over Aging
FJC	: Final Jet Cooling

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SYMBOL LIST

r	:Anisotropy coefficient
r_m	:Average anisotropy coefficient
γ	:Austenite
α	:Ferrit
σ_y	:Yield Strength
σ_t	:Tensile Strength
A_{r3}	:Transformation temperature from γ to α
d	:Ferrit grain size
ppm	:parts per million
mpm	:meter per minute
wt	:weight ton
ζ	:Zeta

OPTIMIZATION OF ALLOYING ELEMENTS AND PROCESS PARAMETERS FOR EXTRA DEEP DRAWABILITY PROPERTY OF COLD ROLLED AUTOMOTIVE STEELS

SUMMARY

Over recent years, new Interstitial Free (IF) steel grades containing ultra low levels of carbon and nitrogen have been adopted to fabricate car body parts. IF steels can provide very high levels of formability.

Elimination of interstitials damaging formability property is achieved through careful control of the steelmaking process by the addition of titanium and/or niobium to react with carbon and nitrogen to form precipitates. In this study, by optimizing Ti and Nb contents in the steel, to obtain good results in mechanical properties especially high r values are aimed.

Precipitation control during hot mill processing is the key in determining the grain size of the hot rolled material. Hot rolling parameters like slab reheating temperature, finishing temperature and coiling temperature are kept constant in this work.

Crystallographic texture is also fundamental in producing exceptional deep drawability. Continuous casting, rolling and annealing operations must also be controlled to optimize the development of preferred precipitation and texture. High reductions, up to 85 %, in the cold mill are the key to deep drawability. Annealing temperature is instrumental in determining final drawability. Annealing time at annealing temperature is also an important factor enabling grain growth to occur, optimizing formability.

As a summary, this work considers the optimization of deep drawing properties by optimization of chemical composition at the steelmaking process and optimization of cold rolling parameters like cold reduction, annealing temperature and soaking time but assuming hot rolling parameters like slab reheating temperature, finishing temperature and coiling temperature were optimized before this study. At the end of study, highest deep drawability ratio mentioned in the literature is achieved in ERDEMIR for automotive market by optimizing chemical composition and cod rolling process parameters. With modified Ti-Nb IF steel that has been cold rolled 80 % and annealed at 870 ˚C the highest r value of 3 was achieved. For Ti-Nb IF steel and galvanized Ti-Nb IF steels, increasing Ti* ratio by approximately 50 % (0.0099 to 0.0148) results in an increase of r values from 1.8-2.0 to 2.3-3.0.

DERİN ÇEKİLEBİLİRLİK ÖZELLİĞİ İÇİN SOĞUK HADDELENMİŞ OTOMOTİV ÇELİKLERİNDE PROSES PARAMETRELERİ VE ALAŞIM ELEMENTLERİNİN OPTİMİZASYONU

ÖZET

Son dönemlerde, otomotiv gövde sacları üretiminde çok düşük oranda karbon ve azot içeren arayer atomsuz çelik kaliteleri geliştirilmiştir. Arayer atomsuz çelikler çok yüksek şekillendirilebilme özelliğine sahiptirler.

Şekillendirilebilme özelliğini olumsuz etkileyen arayer atomlarının eliminasyonu, çelik üretim prosesinde titanyum ve/veya niobyum ilave edilerek karbon ve azotla reaksiyonu sağlanarak çökeltiler oluşturulması yoluyla sağlanır. Bu çalışmada, Ti ve Nb miktarlarının optimizasyonu sağlanarak mekanik özelliklerin iyileştirilmesi özellikle de yüksek şekillendirilebilme yeteneğini gösteren değerlerin elde edilmesi amaçlanmıştır.

Sıcak Haddeleme prosesinde çökelti kontrolü sıcak haddelenmiş malzemede tane büyüklüğünü belirleyen en önemli husustur. Slab ısıtma fırını sıcaklığı, ikmal sıcaklığı ve sarma sıcaklığı gibi sıcak haddeleme parametreleri bu çalışmada sabit tutulmuştur.

Kristalografik yönlenme kabul edilebilir derin çekilebilme özelliği için temel teşkil etmektedir. Sürekli döküm prosesi, haddeleme ve tavlama prosesleri kabul edilebilir kristalografik yönlenme ve çökelti elde edilebilmesi için kontrol edilmelidir. Yüksek ezme miktarları soğuk haddeleme prosesinde yüksek derin çekilebilme özelliği elde edilmesi için önemli proses parametresidir. Tavlama sıcaklığı en son derin çekilebilme özelliği elde edilmesi için önem arz etmektedir. Tavlama zamanı ise derin çekilebilirliği optimize eden tane büyümesini sağlayan önemli bir parametredir.

Özetleyecek olursak, bu çalışma maksimum derin çekilebilirlik özelliği elde edilebilmesi için çelikhane prosesinde gerekli optimum kimyasal kompozisyon değerleri ile soğuk haddeleme prosesinde soğuk ezme oranı, tavlama sıcaklığı ve tavlama süresi gibi parametrelerin oluşturulmasını içermektedir. Çalışma sonucunda, Erdemir'de kimyasal kompozisyon parametreleri ve soğuk haddeleme proses parametreleri optimize edilerek literatürde belirtilen en yüksek derin çekilebilme özelliği değerine sahip çelik saclar otomotiv pazarının kullanımına sunulmuştur. Modifiye edilmiş Ti-Nb IF çeliğinde % 80 deformasyon oranında ve 870 °Ctavlama sıcaklığında en yüksek r değeri olan 3 değeri elde edilmiştir. Ayrıca, Ti-Nb IF çeliği ve galvanizli Ti-Nb IF çeliklerinde artık Ti oranının (Ti*) % 50 oranında (0.0099 dan 0.0148'e) artırılması ile r değeri 1.8-2.0 aralığından 2.3-3.0 aralığına kadar yükselmiştir.

1. INTRODUCTION AND AIM OF THE STUDY

In industrially developed countries, steel has become an integral part of life and nowhere is this more evident than in the automobile. As the automotive industry grows ever stronger, stricter demands are placed on the steel supplier for competitively priced sheet with closely controlled surface quality, gauge tolerances, and uniform mechanical properties.

The overall trend in development is towards high strength steels that allow weight saving through down gauging although there remains some applications of formable steels for the automotive industry is the interstitial free (IF) steels and it is this area on which the present study will concentrate.

IF steels are highly formable due to their very low carbon and nitrogen contents, typically < 0.0030 % C and < 0.0040 % N (all compositions are weight percentages unless otherwise indicated) that confer low yield strengths and high resistance to thinning.

Because the carbon is combined as precipitates rather than in solid solution, IF steels are also non-aging. These characteristics are ideal for deep drawn parts; in fact, only IF steels are able to meet the stringent formability requirements.

The possibility of improving the cold formability of sheet steel by using interstitial free steels has been well known for almost three decades. Starting in the early 1980's in Japan a remarkable growth in the production of IF steels was observed and the tonnage has more than doubled every three years. Similar trends have also been noted in other industrialized countries.

This steady increase is continuing worldwide and the annual production reached about 15 million tonnes in 1995. This remarkable development is supported by the fact that modern steel making technology via the degassing process, sometimes in combination with oxygen blowing, allows the mass production of steel with very low interstitial contents, with levels of carbon and nitrogen below 30 ppm for each interstitial element.

Such low interstitial levels need comparably small additions of stabilizing elements, thus reducing the alloy costs in this steel type, but the breakthrough in the adoption of IF steels was mainly influenced by other factors:

1) The continuous annealing of sheet, which allows for time and cost savings to be made in comparison to the batch annealing process, has become more common.

Furthermore, the more homogeneous conditions in this process reduce the variation in mechanical properties over the length and width of the sheet and also from coil to coil. When using interstitial free steels an excellent Lankford value and also an improved stain hardening potential is obtained, independent, if the batch annealing or continuous annealing process is applied.

2) Continuous annealing is also the standard processing route in hot dip galvanising lines for coating steel sheet with zinc or other metals. The usage of such corrosion-resistant sheet products for the automotive industry is growing very rapidly and many new production lines have been installed worldwide.

These lines do not normally allow for cementite precipitation after recrystallization (so called overaging part in continuous annealing lines).

Since the interstitial free condition guarantees naturally no aging, IF steels are therefore also favorable for galvanising lines as well as for other strip processing routes.

Steel producers first establish their processing parameters according to their mill conditions, and then realize commercial productions according to relevant standards.

Eregli Iron and Steel Works Co. (ERDEMIR), which is the unique integrated flat steel producer of Turkey, carries out R & D studies to achieve further improvement in product quality and commercially produces high grade products for automotive steels.

By objective of this work is the optimization of deep drawing properties by optimization of chemical composition at the steelmaking process and optimization of cold rolling parameters like cold reduction, annealing temperature and soaking time but assuming hot rolling parameters such as slab reheating temperature, finishing temperature and coiling temperature were optimized before this study.

At the end of this study, the highest deep drawability ratio mentioned in the literature is achieved in ERDEMIR for automotive market by optimizing chemical composition and cod rolling process parameters.

2.CHEMISTRY AND HOT ROLLING OF IF STEELS

2.1 Interstitial Free Steels

Interstitial free (IF) steels are state of the art steels, which have exceptionally good deep drawing properties. These steels typically have a total C+N content of less than 50 ppm and contain microalloy additions of Nb and/or Ti. Processing is normally carried out in two stages: (1) production of hot band by hot rolling, accelerated cooling, and coiling; and (2) cold rolling, annealing and coating.

IF steels are used for the production of extra deep drawing grades, especially when processed in a hot dip galvanizing (CGL) or continuous annealing line (CAL). These steels are part of a trend towards ultra clean steels and sometimes other elements, such as sulfur and phosphorus, are also held to low levels in IF grades, depending upon the application [1]. Elimination of interstitials is accomplished by adding sufficient amounts of carbide/nitride forming elements, such as Ti and/or Nb to completely tie up C and N. In the final product these interstitials are in the form of precipitates of Ti and Nb.

A brief history of the development of deep drawable cold rolled sheet steel at Kawasaki Steel Corporation is presented in Figure 2.1. Until the 1970s; only the batch annealing process was used for sheet steels for automotive application. A change in chemistry from rimmed steel to Al-killed steel, accompanying the process change from ingot casting to continuous casting, and decarburizing by open coil annealing improved formability, revealing the limitation of the batch annealing process. Moreover, when IF steel was used in batch annealing, the product did not show formability better than that of open coil annealed rimmed steel, due to the limitation on the annealing temperature in the batch annealing process.

In the 1970s, a continuous annealing process for sheet steel was developed. It was proved metallurgically and experimentally that the formability obtained using conventional low C steel did not satisfy the requirements of customers. At that time, Kawasaki Steel therefore decided to focus on the development of IF steel for the Continuous Annealing Line (CAL) process. In the early stage of the development of IF steel, many disadvantages existed, and the percentage of IF steel production was not high.

By the end of the 1980s, Kawasaki Steel had completed the development of an innovative degassing system, Kawasaki Top Blowing Process (KTB) which enabled the production of ultra low carbon steels (ULC Steels) at a reasonable cost, and an excellent annealing facility, the Kawasaki Continuous Annealing Line (KM-CAL), which enabled high temperature annealing of soft material. Although it took several years to overcome the many metallurgical difficulties of IF steel, all the problems have now been completely solved. Both customers and the company are satisfied with the IF steel product and production process, and IF steel has become a major cold rolled product at Kawasaki Steel. IF steel now has a rather long history and is available in a wide variety of types, nd hence an appropriate understanding is necessary to avoid misinformation and misuse of IF steel.

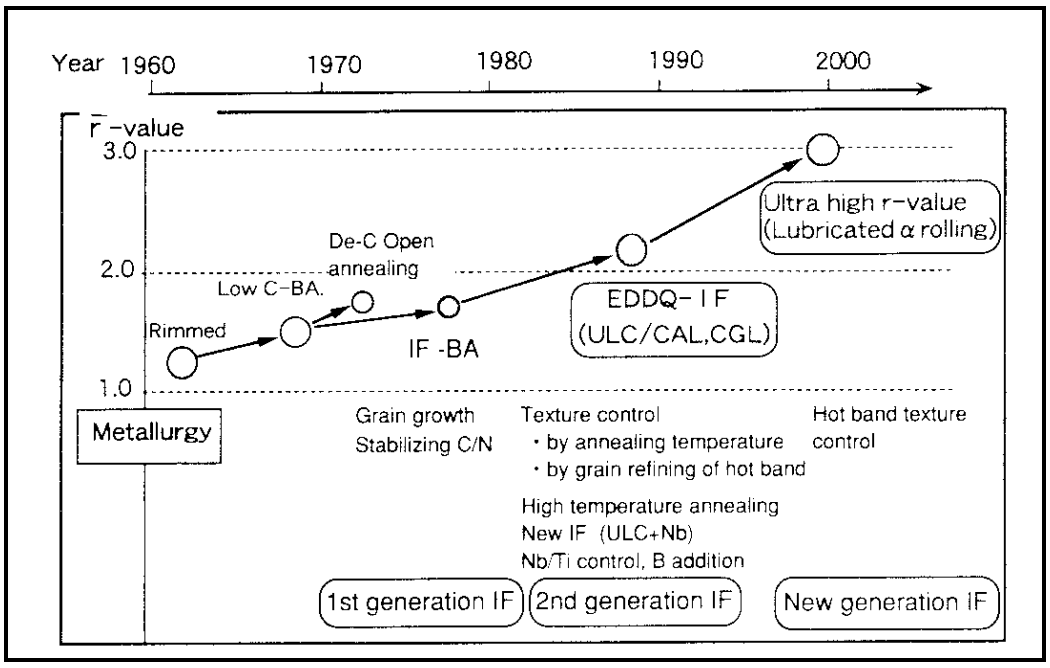


Figure 2.1:History of the development of deep drawing cold rolled sheet steel[1].

A tremendous amount of IF steel is currently being produced by many steel manufacturers all over the world. Because of its extremely superior cold formability, IF steel is applied to various cold rolled and hot rolled products. Its production rate will keep growing in the future.

2.1.1 General Properties of IF Steels

High formability and low yield strength are required for complicated press forming and shape fixability in cold rolled sheet steels. Interstitial free (IF) steels, which are ultra low carbon steels with the strong carbonitride formers such as Ti and Nb in excess of equivalent to carbon, nitrogen and sulfur contents, are extensively applied to various parts such as automotive panels because of their excellent deep

drawability and non aging property. The main automotive applications of formable IF steels are the rear floor pan, the spare wheel well, and the front and the rear door inners (Figure 2.2). In recent years, the demand to reduce the automotive body weight has become greater because of environmental problems, in addition to the improvement of fuel efficiency [2].



Figure 2.2:An automotive door inner produced from Interstitial Free Steel [2].

The advanced steelmaking technology makes it possible to reduce C content less than 30 ppm. In order to improve formability of such ultra low C steels, it is necessary to fix solute C and N through thermo mechanical treatments by adding carbonitride formers such as Ti or Nb. Since the significant reduction of C content results and reduction of either Ti or Nb content, the precipitation behavior is greatly different in terms of the solution or precipitation temperature comparing with conventional ultra low C steels [3].

IF steel sheets generally exhibit a strong {111} or {554} recrystallization texture after cold rolling and annealing. Three mechanisms have been proposed to explain texture development in IF steels, (a) scavenging effect, (b) effect of precipitates and (c) effect of solute X (Ti, Nb). Both mechanisms (a) and (b) are considered to be significant. The existence of solute X itself is not regarded as an important factor because {111} textures are developed even under very small amount of solute X.

The effect of carbide forming elements into three groups with respect to content of carbon and carbide forming elements X;

1) C> 0.005 wt %, X/C > 1

Precipitation of carbides occurs at the high temperature range in γ phase and the influence of hot working conditions on the mechanical properties is small.

2) C=0.003 – 0.005 wt %, X/C = 1

The mechanical properties strongly depend on hot working conditions because carbides such as NbC and TiC start precipitation during hot rolling. High reduction and high speed hot rolling decrease the number of fine carbide depositing after hot rolling, thus the mechanical properties of the steel sheet are improved.

3) C<0.003 wt %

Precipitation of carbides occurs during hot rolling in γ phase. However, grain size of hot bands effectively decreases by the effect of solute niobium. Therefore, the effect of carbide forming elements in solute becomes important in this range [4].

From the research results, in order to obtain no aging property, strong recrystallization texture, and coarse ferrite grains during IF steel production, it is indicated that the following attention should be paid.

(1) Trying to reduce C and N content and other elements such as O, S, P (with lower oxide inclusion contents) in steelmaking. Generally, C<50 ppm, N<30 ppm, the addition of Ti or Nb must be reasonable.

(2) During hot rolling process, and just after finish rolling process are the most important for higher r and n values.

(3) Adopting high reduction of cold rolling as well as high temperature annealing. Based on the degree of influence, the proper order is steelmaking, annealing, hot rolling and cold rolling, as shown in Table 2.1. Therefore, to obtain super formable IF steels, attention at every step of the whole technological process should be paid.

Table 2.1:Key points of Interstitial Free steel production.

The degree of influence	Steelmaking	Hot rolling	Cold rolling	Annealing
Key points Controlled	1.ULC 2.Microalloying 3.Purification	1.Refinement of ferrite grains 2.The coarsening of precipitates	1.High reduction rolling	1.Homogeneous and coarse recrystallization grains 2.Developing recrystallization texture

IF steel is of the best deep drawability among the deep drawing steel sheets developed till now, of which r values lie in the range of 1.8 – 2.5. However, further improvement of r value of IF steel sheet is still needed. The reasons are that the newly developed integrated press forming of automotive parts requires higher formable grade IF steel sheet and the application of high strength steel sheet in

automobile needs further improvement of r value because the increase of strength of the sheet will deteriorate the drawability of the sheet. Figure 2.3 shows the properties of various grades of steel sheets. The target grade of super formable steel sheet is the super EDDQ in the figure 2.3, the highest grade in term of formability with r > 2.5 and n > 0.25. Therefore, super EDDQ grade IF steels have great attraction on auto industry [5]. In figure 2.3, CQ means commercial quality, DQ means drawing quality, DDQ means deep drawing quality, EDDQ means extra deep drawing quality, Super EDDQ means super extra deep drawing quality.

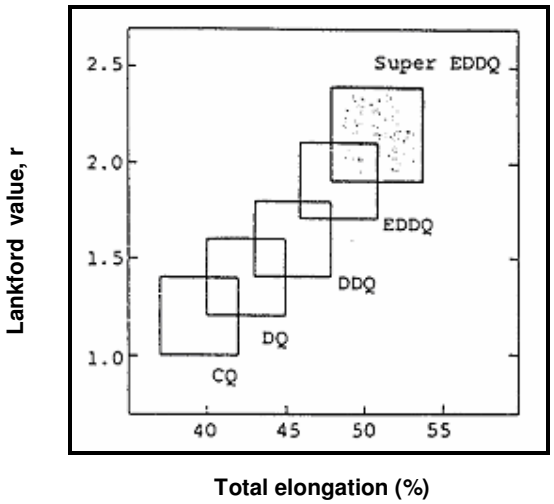


Figure 2.3:The properties of various grades of steel sheets [5].

2.2 Types of IF Steels

Cold rolled steel sheets were formerly classified by their r value and total elongation (El) into four grades, CQ (commercial quality), DQ (drawing quality), DDQ (deep drawing quality) and EDDQ (extra deep drawing quality). Recently, super EDDQ grade of 2,5 in r value and over 50 % in elongation was requested due to severe forming requirements. IF steels, extra low carbon steels alloyed with Nb and/or Ti, were known to be the best candidate for producing such high formable cold rolled steel sheet [6].

2.2.1 Ti IF Steels

In the case of Ti added IF steels, the carbon, nitrogen and sulphur elements form precipitates of different types, such as TiN, TiS, $Ti_4C_2S_2$ and TiC. The other alloying elements (Mn,P) can also form precipitates of MnS and Ti-Fe-P type. All these precipitates lead to a complete elimination of the nitrogen, carbon and sulphur elements from the solid solution. In order to obtain good properties, especially high formability and deep drawability, attention should be paid to each step of the

technological process so as to optimize the microstructure. In particular, it is important to control the precipitate characteristics (type, amount, size and distribution), as they strongly influence the properties [7]. In Ti IF steels, Ti combines with N and S prior to scavenging C. Thus, the minimum Ti level for Ti IF steels is given by the following relationship in wt. % [8].

$$Ti = 3.42N + 1.5S + 4C \quad \textbf{(2.1)}$$

To clearly distinguish between the total Ti, Ti available to combine with C and excess Ti, the following terminology is used. Ti refers to the total Ti content of the steel. Ti_{eff} refers to the Ti available to combine with C after combining with N and S. Ti* refers to the excess Ti over and above the amount required to stabilize C, N and S. Ti_{eff} and Ti* are given by the following relationships (in wt. %) [9]:

$$Ti_{eff} = \text{Total Ti} - 3.42N - 1.5S \quad \textbf{(2.2)}$$

$$Ti^* = \text{Total Ti} - 4C - 3.42N - 1.5S \quad \textbf{(2.3)}$$

Figure 2.4 exemplify the effect of Ti content on the r-value and ductility of Ti-IF steels. There appears to be general agreement among various authors (e.g Ref. 10,11 and 12) that to obtain good properties the Ti content must exceed that required to tie up all N, S, and C, i.e.Ti^*, must be greater than zero.

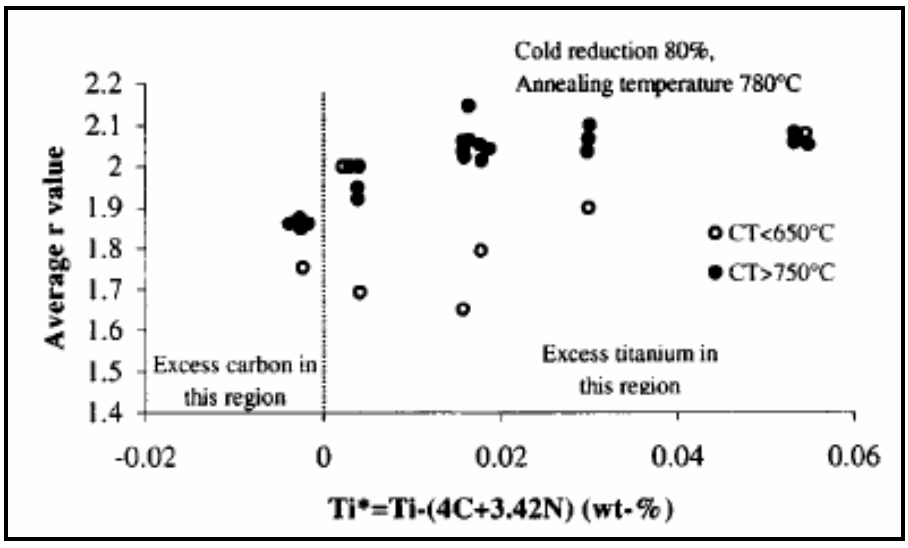


Figure 2.4:Effect of excess Ti (Ti^*) and coiling temperature on r-value in Ti-IF steels[10].

The level of excess Ti (Ti^*) required to obtain good properties depends upon the coiling temperature (CT). As shown in Figures 2.4, at coiling temperatures above about 750 ˚C, r value reaches a maximum at approximately 2,05 for Ti^* levels of 0,015 %.

However, at lower coiling temperatures, < 650 ˚C, r value increases with Ti^* such that a value of 2,05 is reached only at approximately 0,055 % Ti^*.This difference in behavior may be a consequence of the solubility of Ti at these two temperatures. Interstitial free steels with similar processing histories can have different r values, depending on the variations in chemistry, because of differences in the type, size, and dispersions of precipitations. However, as kinetics dominate, it is difficult to obtain equilibrium mole fractions of certain precipitates, notably $Ti_4C_2S_2$, under "normal" mill processing conditions, leading to substantial deviation from equilibrium predictions [13].

2.2.2 Ti-Nb IF Steels

Titanium-Niobium IF steels are able to encapsulate, in the main, the advantages of both Ti only and Nb IF steels while minimizing the disadvantages. This combination provides the best mix of properties to make an ideal substrate for both hot dip coating and galvannealing operations. If insufficient Ti is added such that the level of addition falls below that required for stabilization, TiN forms first and any S that is left behind after the formation of TiS is consumed as MnS. If the steel is stabilized by the insufficient addition of Ti and/or Nb then excess C exist in solid solution and bake hardenable steel may be produced. Excess Ti content can be linked to the incidence of surface streaking; the frequency of appearance of this type of defect should be reduced (as compared with Ti IF) in dual stabilized Ti-Nb IF steels. The optimum Ti and Nb levels required to achieve the highest r values have not been clearly identified although the benefits of Nb in solid solution have been endorsed by Tither et al [14]. Solute Nb residing near or at grain boundaries is known to be beneficial in producing a coating integrity in the final coated product. The Ti-Nb IF steels are less susceptible to cold work embrittlement and exhibit better spot weld fatigue characteristics than Ti only steels.

With respect to the effect of either Ti or Nb addition on mechanical properties, some investors found the following characteristics. Ti added steels show some advantages comparing with Nb added steels; the re-crystallization temperature is low, the mechanical properties such as total elongation and Lankford value are excellent even when low temperature coiling treatment is employed. However, the planar anisotropy of Ti added steels are somewhat higher. Furthermore, Ti is substantially active to S, N and O as well as C; it is not easy to control the Ti content in the optimum range with respect to C content.

On the other hand, Nb added steels have such advantages that the planar anisotropy can be improved and that providing bake hardenability and controlling aging properties are much easier because Nb forms solely NbC in ordinary process.

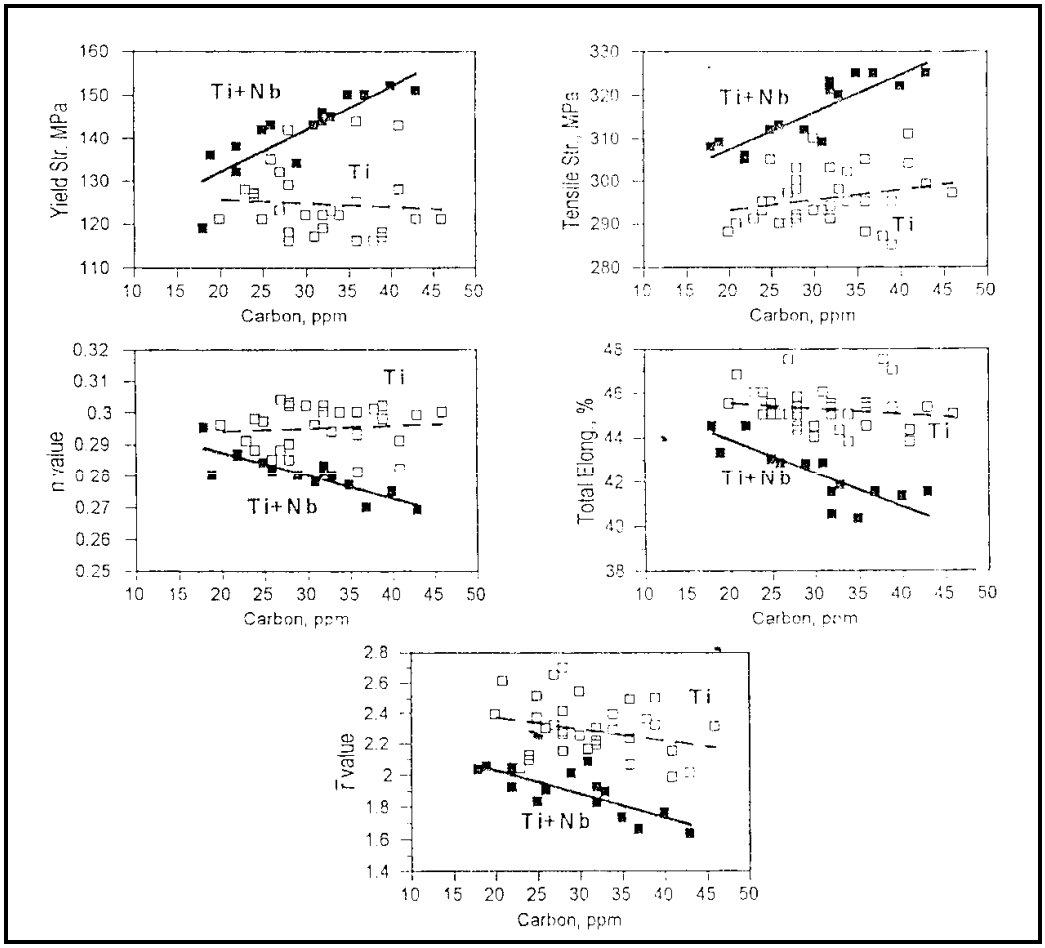


Figure 2.5:Effect of C content on properties of Ti and Ti+Nb IF steels [16].

To avoid these defects associated with the single addition, the combination of both Ti and Nb was recently proposed so that the both advantages could be obtained at the same time [15].

In the Ti-Nb IF steel, primarily intended for exposed, galvannealed applications to circumvent surface problems (such as wood grain and white streaks) Ti and Nb additions are made as follows [16]:

$$N \leq Ti\ (\text{wt.\%}) \leq 3.4N + 1.5S\ ;\ Nb\ (\text{wt.\%}) \geq 7.75C \qquad \textbf{(2.4)}$$

Low carbon and nitrogen contents (generally less than 50 ppm) are preferred in order to keep the Ti and Nb additions to low levels. The carbon content also affects the mechanical properties as illustrated in Figure 2.5 [16]. In TiNb grade, lowering of the C content improved forming properties (total elongation, n and r values) while lowering the yield and tensile strengths.

The data represents substrate properties for material processed as follows: hot mill coiling temperature: 620 – 680 ºC, soaking temperature on hot dip galvanising line:

800–830 ºC. In the Ti grade, excess Ti level was 0.02–0.03 %, while in the TiNb grade, excess Nb (i.e., Nb in excess of that required to stabilize C) level was less than 0.02 %. The mechanical properties are plotted as a function of ferrite grain size in Figure 2.6.

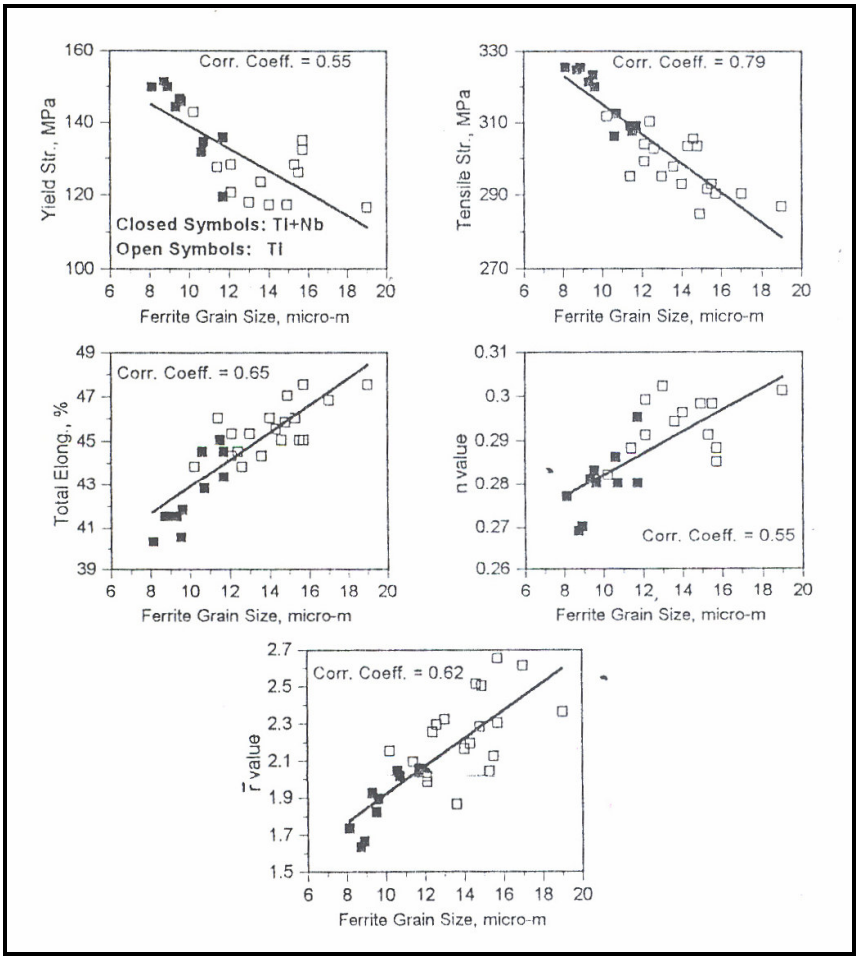


Figure 2.6:Relationships between mechanical properties and ferrite grain size for Ti and Ti+Nb steels (closed symbols: Ti+Nb, open symbols: Ti) [16].

Data for both Ti and TiNb grades are included. Good correlations (R2 = 0.60 to 0.80) were observed, with the forming properties (total elongation, n and r values) increasing with coarsening grain size, accompanied by decreases in both yield and tensile strengths. Finer ferrite grain sizes (8 to 12 µm) were encountered in the TiNb grade compared to the Ti grade (12 to 20 µm). These differences in grain size are presumably related to differences in the size of precipitate particles: finer precipitates inhibit grain boundary migration to a greater extent. Comparisons of the precipitates in Ti and TiNb steels indicate that for a given hot mill coiling temperature, a finer precipitate size distribution is present in the Ti-Nb steel; this is attributed to finer NbC particles in the Ti-Nb grade [16].

2.2.3 Nb-IF Steel

In Nb IF steels, Nb reacts with C and N to form carbonitride precipitates. Aluminum, added during the steel killing process, also reacts with N to form AlN. However, the solubility product of AlN compared with that of NbCN leads to preferential precipitation of AlN at a higher critical temperature, reducing the amount of N available for precipitation as NbCN.

The solubility products are dependent on bulk chemistry, temperature, and precipitate composition.

The amount of solute Nb in a Nb IF steel is given below by the formula (2.5)

$$Nb^{*} = Nb - 7{,}75\ C - 6{,}65(N\text{-}Al_{acid\ sol}/1{,}93) \qquad \textbf{(2.5)}$$

Where $Al_{acid\ sol}$ is the amount of Al available for the formation of AlN, found by the acid dissolution method whereby the acid dissolves AlN but not Al_2O_3.

Figure 2.7 indicates that the Nb IF steel has a cold work embrittlement transition temperature of -55 ˚C; in comparison, the value for the Ti IF steel is -30 ˚C. The solute Nb residing at grain boundaries in the Nb IF steel introduces site competition for those elements, such as P, that are detrimental to cold work embrittlement resistance.

This enables lower ductile to brittle transition temperatures to be realized in Nb IF steels, which generally exhibit poorer elongation and r values than Ti IF steels (Figure 2.8).

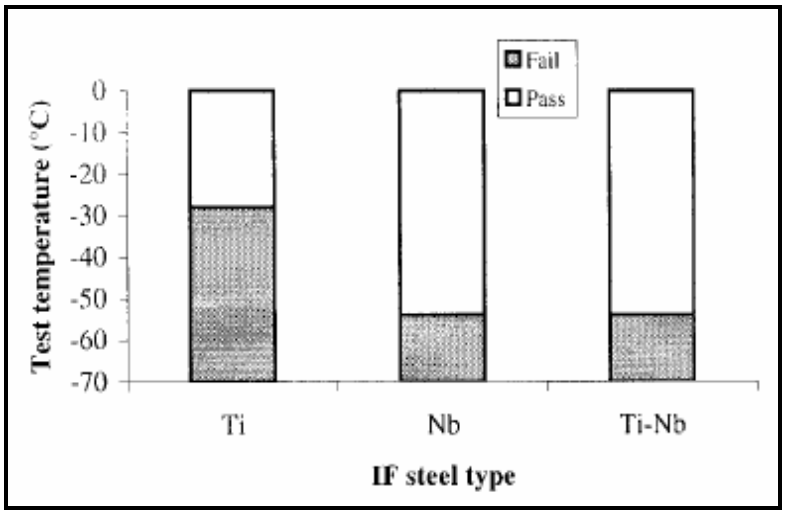


Figure 2.7:Secondary cold work embrittlement transition temperatures for Ti,Nb, and Ti-Nb IF steels [21].

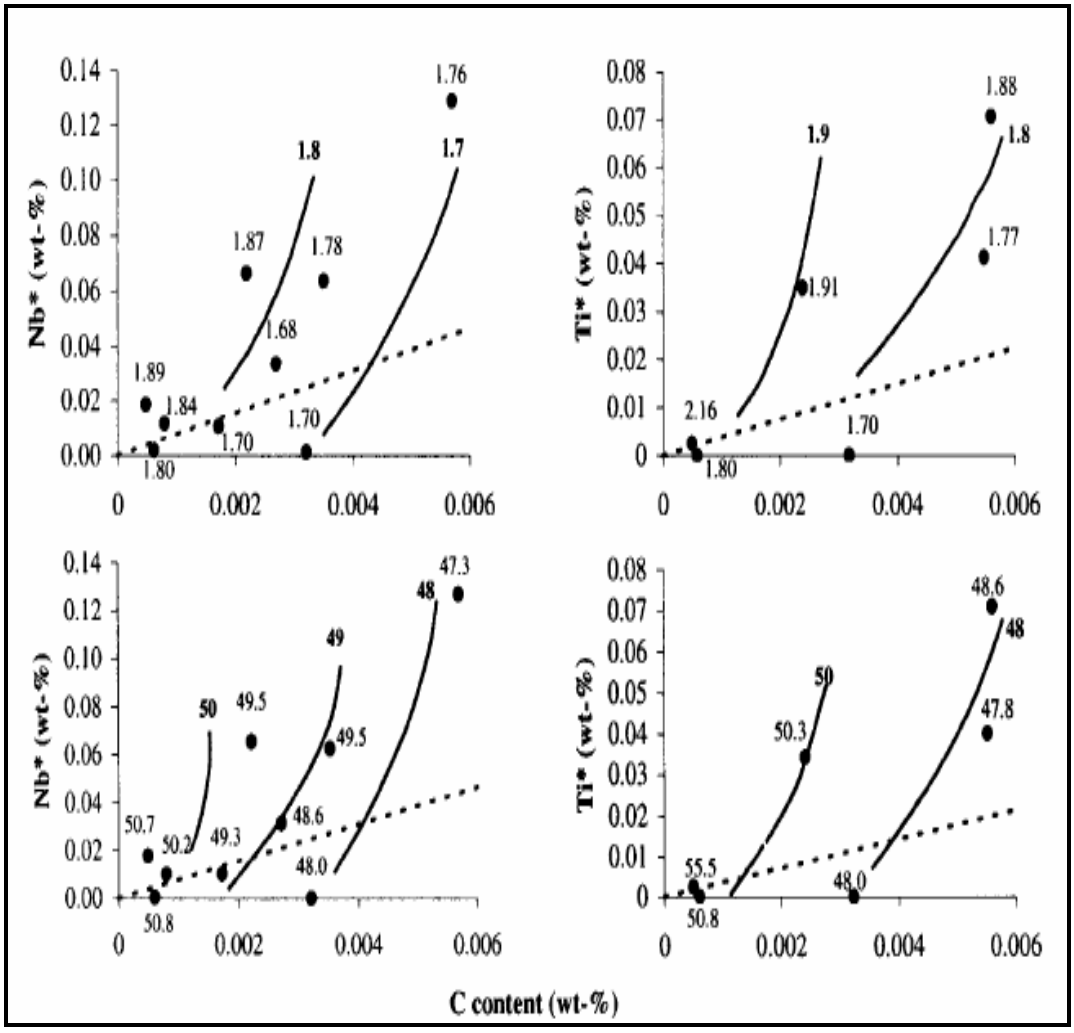


Figure 2.8:Effect of C content and effective amounts of Nb and Ti on r value (top) and total elongation (below) [22].

Niobium increases yield strength by precipitation hardening; the magnitude of the increase depends on the size and amount of precipitated niobium carbides (Figure 2.9). The usual niobium addition is 0.02 to 0.04%. Strengthening by niobium is 35 to 40 MPa (5 to 6 ksi) per 0.01% addition.

This strengthening was accompanied by a considerable impairment of notch toughness until special rolling procedures were developed and carbon contents were lowered to avoid formation of upper bainite.

In general, high finishing temperatures and light deformation passes should be avoided with niobium steels because that may result in mixed grain sizes, which impair toughness. Niobium steels are produced by controlled rolling, recrystallization controlled rolling, accelerating cooling, and direct quenching.

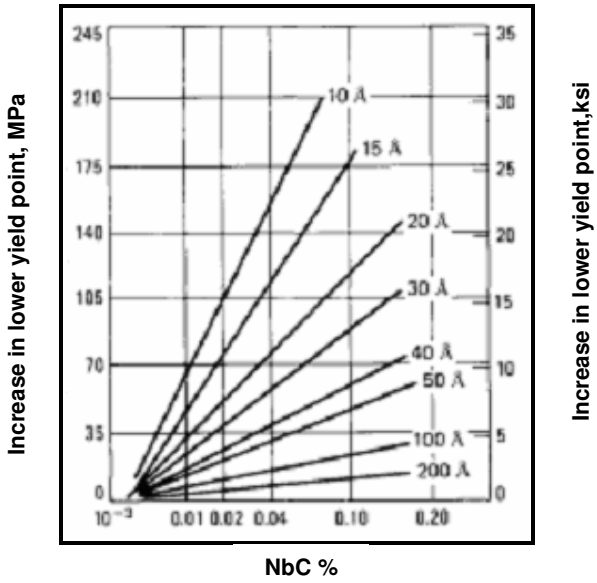


Figure 2.9:Effect of Niobium carbide on yield strength for various sizes of niobium carbide particles [22].

The recrystallization controlled rolling of niobium steel can be effective without titanium. Also, much less niobium is needed, and niobium-titanium steels can be recrystallization controlled rolled at higher temperatures.

2.3 Chemistry of IF Steels

The effective decarburization of molten steel is one of the most important factors for the IF steel, because carbon content determines the properties of final products and production cost related to the contents of Ti and Nb. Recently, decarburization has been carried effectively in a vacuum degasser, where the carbon content in a converter is approximately equal to that of aluminum killed steel which is not decarburized [3].

With regards to alloy composition, there has been a great improvement in steelmaking in order to allow, for instance, production of steels with low interstitial elements carbon, nitrogen and with low sulfur and low phosphorus contents. The steelmaking improvements have allowed the production of ultra clean steels for automotive applications. Vacuum degassed interstitial free sheet steels have very high formability, especially for parts requiring good deep drawability. The high formability is achieved by lowering the interstitial element content to very low levels in steelmaking and by additions of stabilizing elements such as titanium or niobium which combine with the N and C not removed by the steelmaking process. At the end, the chemical composition of IF steels must be adjusted in order to satisfy the different requirements that are put on modern sheet steels [23].

2.3.1 Carbon and Nitrogen

In modern steelmaking numerous chemical elements are controlled to levels of 50 ppm and below. This control has led to improvements in the mechanical characteristics, e.g. creep resistance, at room temperature and elevated temperatures, improved toughness and fracture mechanics at low temperature, better hot and cold formability, weldability and corrosion resistance. Interstitial carbon is a particularly important alloying element and it plays an important role in defining the road map of the steel industry into the next century. In order to achieve extra deep drawing quality for car bodies and beverage can fabrication, ultra low carbon steels are needed. For the best results the carbon as well as nitrogen contents needs to be lowered to levels of 10 to 15 ppm and fixed by titanium additions in order to produce an interstitial free steels [17].

The concentrations of residual interstitial nitrogen and carbon in steels is not only a crucial qualification standard for interstitial free steels, but also a process control factor in the production of a wide variety of improved steels.

Carbon and nitrogen contents have a deleterious effect on the formability of steel. In Figure 2.10, average r value as a function of carbon content for a number of steels is given [18].

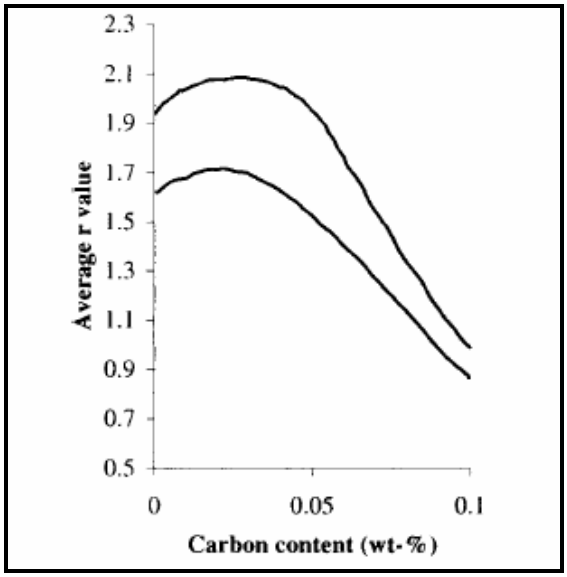


Figure 2.10:Effect of carbon content on r values in steel [18].

In Figure 2.10, Fukuda [18] showed the effect of carbon and cold rolling deformation on the development of average r values. The high r-values are associated with the formation of beneficial recrystallization textures, which are produced after hot deformation, coiling, cold rolling and annealing. A critical fundamental observation in this processing sequence is the fact that carbon atoms in interstitial solid solution

degrade beneficial {111} recrystallization texture and unfavorable {110} and {100} components [18].

The carbon content of the interstitial free steels is below that of the solubility limit of carbon in bcc ferrite at its annealing temperature, therefore, no pearlite forms in these steels. These steel products can withstand deeper draws with less breakage than other grades of steel sheet, especially in stamping operations. Such plasticity allows large stampings with extreme contours to be produced in a single operation, which is economically attractive in the production of automobile bodies [24].

Further improvement of the formability of IF steel sheets has since been achieved through the optimization of chemical composition and processing conditions, such as decreasing the carbon content and the slab reheating temperature, refining the hot band grain size, increasing the annealing temperatures [25].

The average r value of IF steel is significantly improved by refining the hot band grains, by decreasing the finish rolling temperature in large thermal drop, to temperature even below the Ar_3 transformation temperature, during hot rolling process. However, the average r values of sheets hot rolled almost isothermally at a finish rolling temperature below the Ar_3 transformation temperature deviate detrimentally, when compared with hot band of the same grain size.

Figure 2.11 shows the effect of carbon content on the r value and it improves with the decrease in carbon content and reaches about 2.6 at 0.001% C. The decreament of r value with increasing carbon content over 0.003 % C is larger in a 0.30 % Mn steel than in a 0.02 % Mn steel. This is thought to be a result of the copresence of carbon and manganese since titanium content is too small to stabilize carbon completely as TiC precipitates over 0.003 % C. The increase in the solute carbon making complex with manganese atom leads to the suppression in development of {111} recrystallization texture component [26].

Nitrogen is also a critical interstitial element, but at the low levels present in IF steels is considered to be effectively removed from interstitial solid solution by aluminum nitride or titanium nitride formation which usually takes place at a relatively higher temperature than carbides.

As a result, major attention is focused on the removal of carbon whose concentration in the solid solution matrix is determined by the precipitation reactions that occur during solidification and subsequent thermo mechanical processing [27].

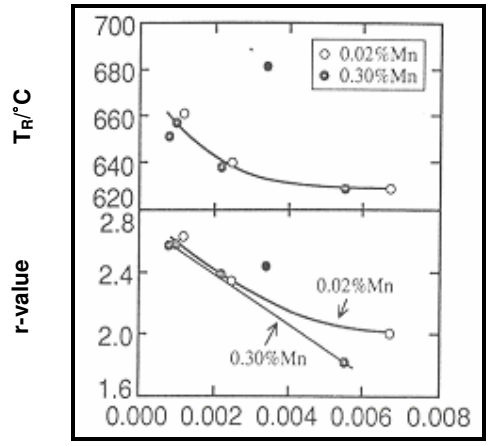


Figure 2.11:Effect of carbon content on ŕ-value and half recrystallization temperature [26].

The effect of nitrogen content on r value and half recrystallization temperature in 0.002 % C, 0,05 % P, 0.001 % N, 0.025 % Ti steels is similar to that of sulfur content as shown in Figure 2.12. With increasing nitrogen content, r-value deteriorates, while half recrystallization temperature remains constant [26].

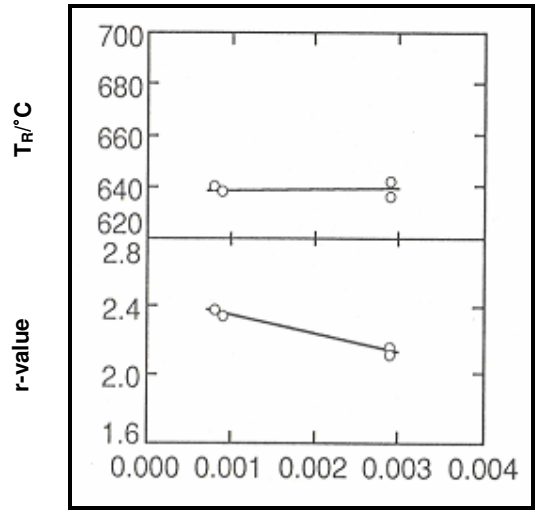


Figure 2.12:Effect of nitrogen content on ŕ and half-recrystallization temp.[26].

2.3.2 Titanium and Niobium

In IF steels, the deep drawing quality can be obtained by adding Ti or Nb to stabilize solute atoms (C and/or N). When IF steel is first developed, the normal C level was from 0.005 to 0.010 %, requiring a higher addition of carbide (nitride) forming elements than with recent IF products. In the former case, the precipitation of TiC or NbC occurred in a relatively higher temperature region of the γ phase. Therefore, the effect of hot rolling conditions on the mechanical properties of cold rolled and annealed steels was not significantly large. In contrast, when the carbon

content is reduced to 0.003% or less for improvement of press formability, the temperature range for TiC or NbC precipitation shifts to a markedly lower temperature in the γ range, or even into the ferrite region. The mechanical properties of the IF steels are then strongly affected by both hot rolling and annealing treatment. When the C content is lowered to 0.003 % or less, Nb forms finer precipitates than Ti, leading to a considerable increase in the recrystallization temperature. This in turn explains why Nb steel shows somewhat poor press formability compared to Ti added steel, when other processing conditions are the same. However, if both the hot rolling and continuous annealing conditions as well as the steel chemistry, are optimized, a small addition of Nb can induce to a remarkable degree the characteristic improvement of the properties, such as the planar anisotropy of r value seen in IF steels. Such improvement cannot be easily obtained in Ti added steels [26].

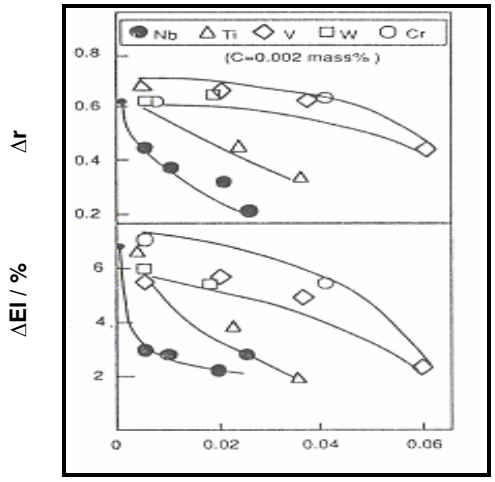


Figure 2.13 Planar anisotropy of mechanical properties plotted against effective amount of carbide forming elements [26].

Figure 2.13 shows the effect of the alloying elements Nb, Ti, V, W and Cr on the planar anisotropy of the r value and elongation in 0.002 % C cold rolled and annealed steels.

Neither V, W nor Cr causes any significant change in the planar anisotropy of the r value and elongation. On the other hand, even an extremely small amount of Nb significantly reduces planar anisotropy. Small additions of Nb stabilize solute C in the hot rolling stage more easily than other elements, as shown in Figure 2.14, where aging index (AI) is a parameter of the quantity of solute C content. Furthermore, the grain size of the hot band is also reduced by Nb addition.

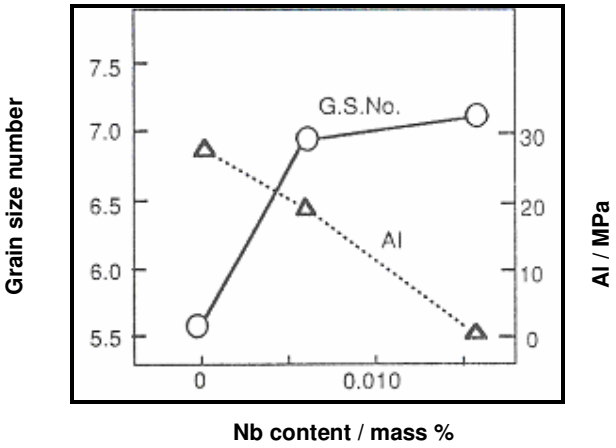


Figure 2.14:Grain size number and aging index (Al) of hot rolled sheets plotted against Nb content [28].

During cold rolling, the {110}<001> orientation, which indicates large planar anisotropy, is often formed in a deformation band inside grains when solute carbon is present. Therefore, both grain refining and stabilizing of the solute atoms in the hot band by Nb addition are very effective for the improvement of planar anisotropy in IF steels [28]. The addition of small amount of a Nb to Ti added steel may lead to further improvement of average properties as well as their planar anisotropy in IF steels. The effect of Nb on the r value and elongation in Ti added steel is shown in Figure 2.15. It is clearly observed that high average r value, and elongation and small planar anisotropy can be obtained at the same time when 0.005 to 0.010 % Nb is co added with Ti. As a result, Nb plays an important role in improving press formability in Ti steel. Furthermore, low Ti and Nb coadded steel is also suitable as the substrate material of galvannealed sheet when improved surface quality is required. Thus, Ti and Nb coaddition is widely adopted by many steelmakers as the basic chemistry of IF steels [28].

Increasing coiling temperature to 700 ºC markedly increased the r value of steel with insufficient Nb addition. Formation of Nb-C precipitation clusters was suggested that stabilize the solute C and assist in the development of a strong {111} recrystallization texture. The r value of fully stabilized steel also increased with increasing coiling temperature as a result of precipitate coarsening. A dissolution reprecipitation mechanism is proposed for the development of an elongated grain structure and strong {111} <110> texture in steel with lean alloy addition [29]. Micro alloying of ULC steels with titanium promotes improved texture as a result of the removal of free carbon atoms as titanium carbonitrides and ferrite matrix cleaning [30].

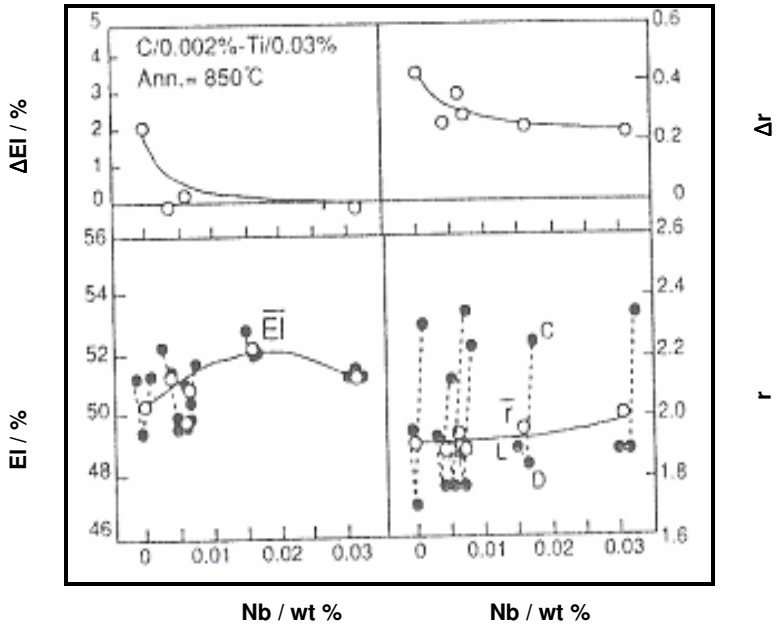


Figure 2.15:Effect of Nb addition to Ti-based sheet steels on planar anisotropy of elongation and r-value [28].

Microalloying with Nb leads to finer ferrite grains. Addition of phosphorus to these steels favors ferrite grain coarsening and practically leads to matching of ferrite grain size independently of titanium and/or niobium presence.

The difference between the sizes of as rolled ferrite grains in the cases of micro alloying with titanium, niobium and phosphorus in different combinations proved to be definitely related to sizes of titanium and niobium carbonitride dispersed particles which are generated during hot rolling [30].

Micro alloying of ULC steels with niobium promotes an increase of the recrystallization initiation temperature as a result of fine particle formation in as rolled steels, an improved texture as a result of ferrite matrix cleaning of free interstitials and a decrease in planar anisotropy as a result of ferrite grain size refining [30].

2.3.3 Sulphur

Lowering the sulphur content in IF steel retards the precipitation of fine titanium carbides, owing to the possibility of prior precipitation of $Ti_4C_2S_2$, thus increasing the propensity for recrystallization and grain growth [18].

When coupled with a slab reheat temperature of ≤1150˚C, the molar fraction of TiC precipitates is reduced thus providing an effective method of encouraging recrystallization and grain growth in these steels [31].

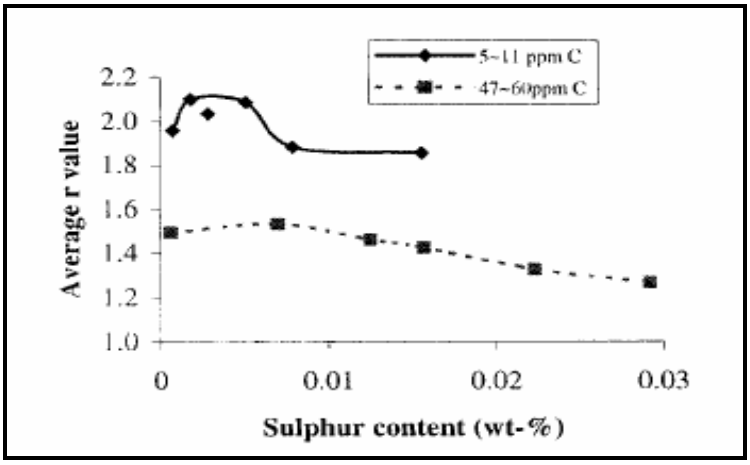


Figure 2.16:Effect of S content on r value of 0,043 Ti-0,008 Nb IF steel, slab reheated at 1100 ˚C [31].

Figure 2.16 shows that optimization of the sulphur content relative to the steel chemistry is necessary in achieving the highest elongation and r values. The maximum in the figure corresponds to sulphur contents that are equiatomic with carbon.

2.3.4 Sulphur and Manganese

When sufficient quantities of manganese and sulphur (>0.08 and >0.001% respectively) are present in IF steel, they will combine to form MnS. These precipitates have a strong influence as they act to refine the grain structure on the austenite to ferrite transformation. At manganese contents <0.005%, in the presence of adequate amounts of carbon, sulphur, and titanium, TiS and $Ti_4C_2S_2$ are able to form.

2.4 Deep Drawability Property of IF Steel

R value is a most important value to judge the drawability of steel sheets. In conventional production of steel sheets for deep drawing applications, high r-value is obtained by means of accurate texture control of steel sheet during cold rolling and annealing process. In recent decades, the development of IF steel satisfies the demand of car manufacturers due to its high r value and excellent deep drawing properties. With the development of the automobile industry in recent years, the designed shape of car panel parts is getting more and more complex and car panel parts tend to be successfully punched in one time. So the new technology to improving deep drawing properties of steel sheets further attracts more and more attention of material researchers and car manufacturers [32]. The key to enhanced formability of IF steels is the formation of favorable cube on corner re-crystallization textures. Microstructures with these textures having high r values defined as follows:

$$r = \varepsilon_w/\varepsilon_t = \ln(wi/wt) / \ln (t_i/t_t) \quad \textbf{(2.6)}$$

The r-values are a function of grain orientation, and the anisotropy of r values is incorporated into average terms, r or r_m, as follows:

$$r = r_m = (r_0 + 2r_{45} + r_{90}) / 4 \quad \textbf{(2.7)}$$

where r_0 = r value determined in the rolling direction, r_{45} = r value determined at 45 degree to the rolling direction, r_{90} = r value determined in the transverse direction. Another measure of formability, related to earing, is the planar anisotropy defined as follows:

$$\Delta r = (r_0 - 2r_{45} + r_{90}) / 2 \quad \textbf{(2.8)}$$

The carbon content of steels and the coarsening of carbide in hot band by high temperature coiling have great effects on the improvement of the deep drawability. The next important element coming after carbon is manganese. The r value decreases with increasing manganese content. In the case of continuously annealed sheets, manganese gives a secondary effect due to a change in carbide morphology besides its own effect. This means that carbide becomes harder to be coarse as manganese content increases even if coiled at high temperature, and gives an unfavorable effect on the deep drawability [33]. Deep drawability also varies by cold reduction ratio**.** Figure 2.17 shows the relation between the r-value and cold reduction ratio in continuous annealed sheets.

In the case that carbon content is 0.04 %, the r-value reaches the peak when the cold reduction ratio is 80 %. In the case that carbon content is 0.02%, the r-value increases linearly with increasing in cold reduction ratio and the peak reduction ratio shifts the vicinity of 90 %.

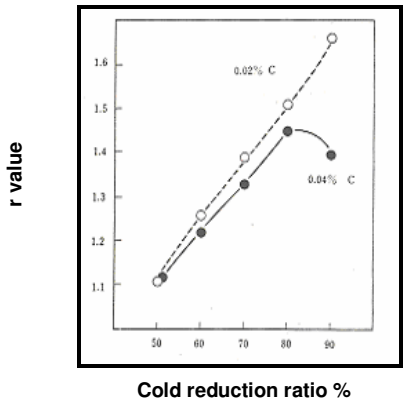


Figure 2.17:Influence of cold reduction ratio on the r-value of continuously annealed steel sheets [33].

In the case of annealing by rapid heating, however, this phenomenon is mainly caused by a change in carbide morphology. This corresponds to the fact that carbide is up to coarsen as its carbon content becomes lower. Figure 2.18 shows the effect of annealing temperature on the r value. The increment of the r-value by high temperature annealing is due to the grain growth.

Accordingly, the materials containing the small amount of carbon and nitrogen, which means the few existence of carbide and nitride as the obstacles for grain growth by high temperature annealing. The deep drawability can be improved by proper combination of the various factors described above. In the case that high deep drawability, such as about 2.0 in the r value, is required the steels containing titanium or niobium, which are known as interstitial free steel, is generally used [33].

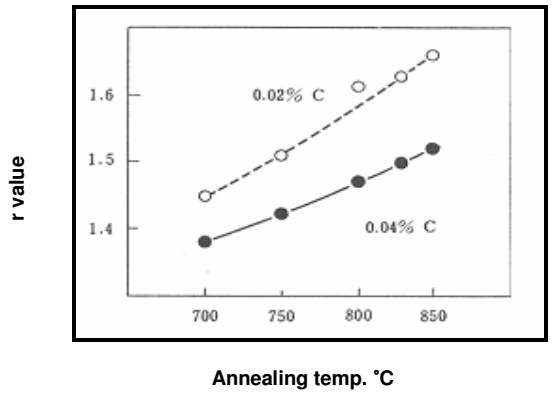


Figure 2.18:Influence of annealing temperature on the r-value of continuously annealed steel sheets [33].

Good ductility and non aging behavior are favorable characteristics that make IF steel so attractive. However, the element that perhaps contributes most to the popularity of IF steel is its excellent deep drawability. For hot dip galvanized and galvannealed applications the ultra low carbon IF steel is superior to the Al killed low carbon steel in terms of aging and deep drawability. The deep drawability of these steels, which is dependent on both chemistry and thermo mechanical processing, is due to the preferred crystallographic texture of the sheet that results after cold deformation and annealing. Onset and completion of re-crystallization have been reported to occur at considerably longer times in cold rolled IF steels than in other low carbon steels, giving more time for recovery processes. However, very little information is available concerning the kinetics of the recovery and re-crystallization processes occurring during isothermal and continuous heating annealing treatments [7].

IF steels develop strong {111} re-crystallization texture after annealing, which gives rise to high r value (Lankford value) and good deep drawability. It is well known that solute carbon is detrimental to the development of {111} texture. Therefore, the

addition of a sufficient amount of strong carbide former, such as Nb or Ti to stabilize C is essential to the production of steel sheets with satisfactory properties. Hot rolling parameters, including slab reheating temperature, finishing temperature, coiling temperature, and hot rolling speed and reduction also have a strong effect on the deep drawability of IF steels. These effects have been related to the size and distribution of the precipitates and grain size of hot band [8].

Several mechanisms have been proposed to explain the development of a strong {111} re-crystallization texture in Nb or Ti stabilized extra low carbon steel. Most researcher [8] thought that the removal of interstitials from solid solution is responsible for improved {111} annealing texture. Some other researchers [8] considered that the carbide controls the {111} texture or carbonitride precipitates through an interaction with the re-crystallized grain. Furthermore, the presence of solute Nb in ferrite has also been suggested to provide a suitable condition for {111} texture development [8].

A recrystallization texture of IF steel often bears distinctive features. Most researchers [34] have reported that IF steel re-crystallization textures characterized by a very narrow and intense distribution of grain orientation about the <111>//ND fibre as well as (111)<112>, (111)<123> have been proposed as position of maximum intensity of the re-crystallization texture of IF steels. If there is something like a typical re-crystallization texture, it is reasonable to assume that there is also something like a typical re-crystallization mechanism. The resemblance between IF re-crystallization textures produced with a range of composition and processing factors gives reason to believe that the mechanisms that are responsible for this texture are also relatively little influenced by composition and processing factors. Since most recrystallization texture is measured after a soaking at high temperature, the phenomenon that is considered to be typical for IF steel is not only re-crystallization but also includes grain growth. Re-crystallization itself already consists of several phenomena, namely recovery, nucleation and grain growth of re-crystallized grains in the deformed matrix [34].

On the other hand, the ferrite matrix of heavily cold rolled IF sheet steels re-crystallizes during annealing to a polycrystalline structure with a strong (111)<110> re-crystallization texture, producing high values of the average normal anisotropy ratio, average r value, which is associated with the high formability of ultra low carbon IF steels. The deep drawing characteristics of IF sheet steels, in terms of the measured normal anisotropy ratio from a tensile test, are strongly dependent upon the development of strong {111} re-crystallization texture during in line annealing. Furthermore, during the process of removing interstitial elements by using

microalloying elements, precipitates could be left in the matrix that may adversely affect the re-crystallization process. The excess Ti in the matrix is far less effective than Nb in retarding the re-crystallization process, but this characteristic is considered to a distinct advantage in the design and processing of Ti stabilized interstitial free steel for extra deep drawing grades [26].

2.5 Effects of the Processing Conditions

2.5.1 Hot Rolling Conditions

Once the favorable texture components are established in the hot band, subsequent cold rolling and annealing merely strengthens the texture. If, however, the hot band texture is not at its most favorable, more variability in the properties of final cold rolled and annealed product will be experienced and such high r-values will not be achievable.

The hot rolled grain size has a significant effect on the resulting texture developed during cold rolling and annealing [36]. Although a large final grain size is desirable for a high r value, initially, a fine hot band grain size is preferable. This implies that the final texture after annealing originates from the hot band. The final annealed texture arises from the nucleation and growth of grains during recrystallization. It becomes evident why a fine hot band grain size is desirable when considering possible nucleation sites for recrystallization and grain growth. Dasarathy [37] reported that when hot band grain size (in a rimming steel in this particular case) was varied over the range from ASTM 5-6 to ASTM 12-13, no significant difference in the cold rolled texture was observed; however, on recrystallization (by means of a short anneal at 700 ˚C), the annealing textures differed widely. The {111} ND fibre required for deep drawability was strengthened appreciably in the fine grained material: by contrast, the {111}<110> orientation detrimental to formability was intensified in the coarse grained material.

In the hot strip mill, the major variables for controlling the mechanical properties of IF steel are reheat, finishing, and coiling temperatures. Controlling of the hot band metallurgical properties is fundamental if high r values are to be achieved in the final cold rolled and annealed product.

2.5.2 Slab Reheating Temperature and Duration

During slab reheating, dissolution of precipitates occurs; this is the first stage in determining the final hot band texture and grain size. Figure 2.19 shows that at slab reheat temperatures of ~1250 ˚C carbide and carbonitride precipitates may almost

completely dissolve while others such as TiN and TiS are relatively stable. At a slab reheat temperature of 1000 ˚C, all but the finest precipitates remain.

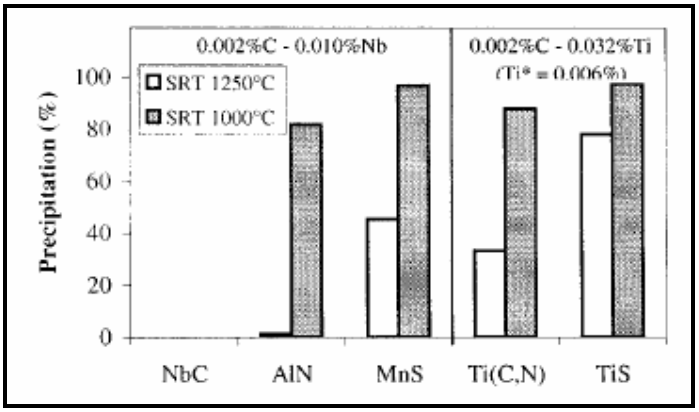


Figure 2.19:Effect of slab reheat temperature (SRT) on dissolution of precipitates in Ti and Nb IF steels [38].

Many authors have studied the effect of slab reheat temperature (SRT) on r value [38]. Some concluded that the effect of SRT was small, whereas others felt in to be a significant variable which could be used to further increase formability.

A composite graph (Figure 2.20) of the effect of SRT on r value has been complied from the data of a number of authors [40, 39]. Figure 2.20 shows that overall the effect of SRT on r value is small and in most cases negligible until SRT is reduced below about 1150 ˚C. Satoh et al.[40,41] found a less dramatic effect than Sanagi et al [39] whereas Gupta et al.[42] observed no increase in formability. This may be due to the specific chemistry involved as Satoh et al.[41] discovered.

Figure 2.21 shows the interaction between SRT, alloying addition, and resultant r-values. A lower SRT has a beneficial effect on r-value at lower levels of alloying addition and a smaller effect at higher alloying levels; additions of Ti having a greater effect than Nb for the same change in reheat temperature.

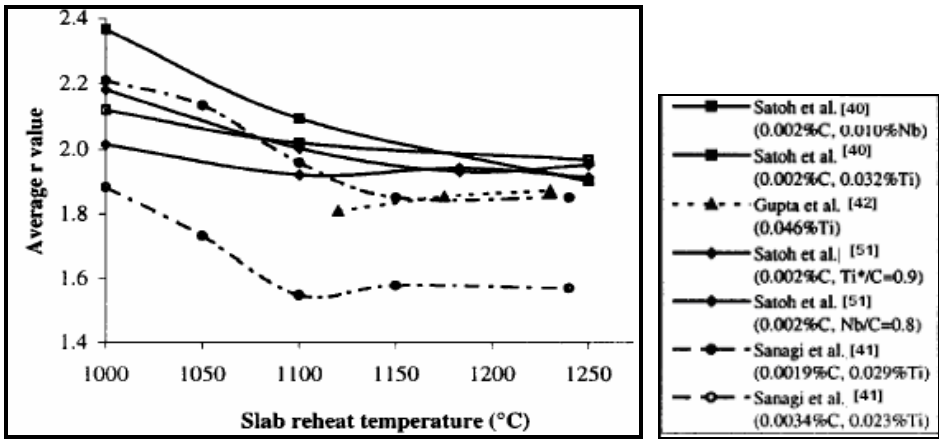


Figure 2.20:Composite graph of effect of slab reheat temperature on r [40].

Katoh et al. [16] showed that the beneficial effect of a low SRT in IF steel was greater for higher contents. Sanagi et al. [39] reported the effects of SRT and coiling temperature on precipitates in the hot band and mechanical properties in cold rolled and annealed sheet of Ti IF steel. They concluded that precipitation of $Ti_4C_2S_2$ in the hot band, achieved by using a low SRT and high coiling temperature, improved r value. Moving to lower SRT is difficult for some rolling mills due to roughing mill load limitations and problems in maintaining the finishing temperature above the austenite to ferrite transformation temperature.

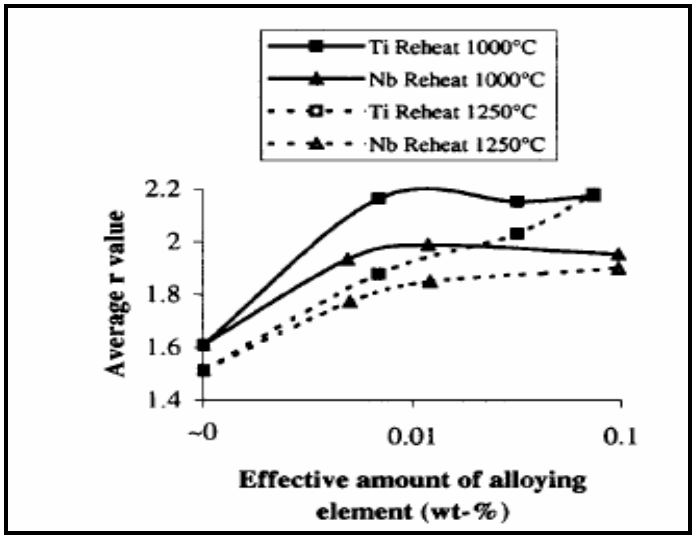


Figure 2.21:Effect of alloying element content and slab reheat temperature in hot rolling on r value of continuously annealed IF steel sheet (ferrite decomposition temperature 880 ˚C) [41].

Overall it can be said that while a low SRT may be preferable it is not essential for processing IF steels unless temperatures less than about 1150 ˚C are employed. A lower SRT is especially favorable with low levels of the alloying additions Ti and Nb as this retards the dissolution of TiC and NbC. Thus it can be said that the extent of the effect of SRT on r value is dominated by the dissolution and reprecipitation of carbides and nitrides which in turn is controlled by the level of C, N, and carbonitride forming elements in the steel. From the available literature, the ability to reduce the SRT from 1250 to 1100 ˚C would allow an increase in r-value of approximately 3 to be realized [41].

2.5.3 Roughing

De Ardo [43] states that the most favorable texture components for developing a high r value are strongest when the reheated austenite grain size is large, initial pass strain is heavy (>50 % reduction), and when the finishing temperature is below the austenite recrystallization temperature. This indicates that the evolution of the most beneficial texture components are closely related to the formation of

deformation bands in the austenite grains. Conditions are result in preferred {111}<110> textures in the hot band. The first roughing pass for conventionally rolled IF steel is generally no greater than 30 % and therefore any crystallographic textures set up in the steel during the roughing or finish rolling stages are “diluted” on the austenite to ferrite transformation, resulting in a hot band with little or no preferred crystallographic texture.

2.5.4 Finishing Temperature

The effect of hot rolling conditions such as finishing temperature, hot rolling reduction, cooling regime, and rolling speed on the properties of IF steel are complicated by the fact that often the effect exerted is related to the influence of both SRT and steel composition. The effects of various finishing temperatures at hot rolling on r value have been intensively investigated over the years.

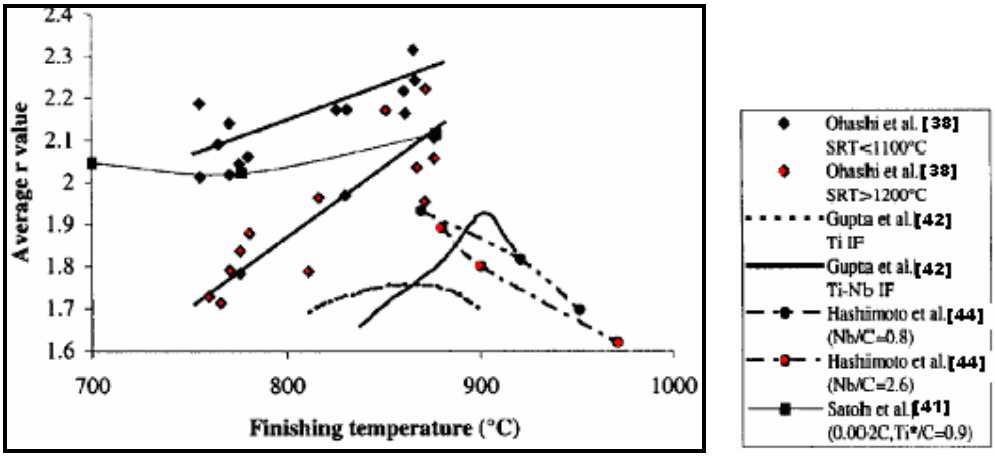


Figure 2.22:Composite graph of effect of finishing temperature on r value [42,44-46].

Figure 2.22 summarizes some of the data showing the effect of finishing temperature on r-value.

Work carried out by Gupta *et al.*[42] provided the most extensive view of the effect of finishing temperature for both Ti and Ti-Nb IF Steels. They found that for both steel types, r value increases to a maximum and then begins to decrease. Referring to Figure 2.22, it can be observed that the effect of finishing temperature is more pronounced in the Ti-Nb IF steel than the TI IF steel. The maximum r value developed in the Ti-Nb IF steel is greater than that of the Ti IF steel by almost 0,1, the maxima occurring at 900 ˚C a mixed grain structure of small and large grains consisting of a grain size distribution is formed. On further lowering the finishing temperature, uncrystallized grains appear in the micro structure; this is indicative of rolling taking place in the austenite-ferrite or ferrite regions. The same is true for the

Ti IF steel rolled below 870 ˚C. The difference in the recrystallization temperatures of the two steels results from the addition of Nb suppressing recystallization through grain boundary pinning by fine precipitates.

Many authors have observed a continuous decrease in r value with increasing finishing temperature [45, 46]; data generated by Hashimoto et al. [44] also exhibit this type of trend (Figure 2.22). Hashimoto studied Nb only IF steels, so it is feasible that for the chemistry and hot rolling practice used, the finishing temperatures studied did not coincide or over lap with the Ar_3 temperature thus explaining the absence of the maximum as observed by Gupta et al [42].

Satoh et al.,[41] Ohashi et al.[38] (Figure 2.22), and Known et al.[46] observed a continuous increase in r value with increasing finishing temperature, the opposite of that observed by Hashimoto et al.[44]. At a reheat temperature of 1200 ˚C Ohashi et al.[38] observed a drop in r value of ~0,1 over the temperature range 850-800 ˚C. Over the same temperature range, a drop in r value of ~0,05 is experienced at a reheat temperature of 1100 ˚C, thus showing the beneficial effect of low slab reheat temperature on r value. The range of finishing temperatures examined implies that rolling took place in the ferrite region so the ferrite to austenite transformation temperature may not have been encountered. The data seem to suggest that only the left hand side of the pilot is visible as compared with that of Gupta et al [42].

It can be seen that the height and position of the maximum in the graph of finishing temperature versus r value is determined by a combination of the interstitial content, the richness of the Ti and Nb additions, the SRT, and the hot rolling conditions. The height of the peak is determined also by degree of cold reduction imparted to the steel and the maximum soak temperature reached during annealing. The performance of the final product in terms of formability is a complex interrelationship between all the variables from steelmaking to annealing.

It is recognized that r value increases with decreasing hot band grain size, the finest grain size being realized when the finishing temperature is close to the austenite to ferrite transformation temperature, thus suggesting once again that maxima should be observed in all trends if a sufficiently wide range of finishing temperatures is examined.

It is well known that both a fine initial hot band grain size and a large degree of cold rolling reduction are required to achieve the highest r-values.

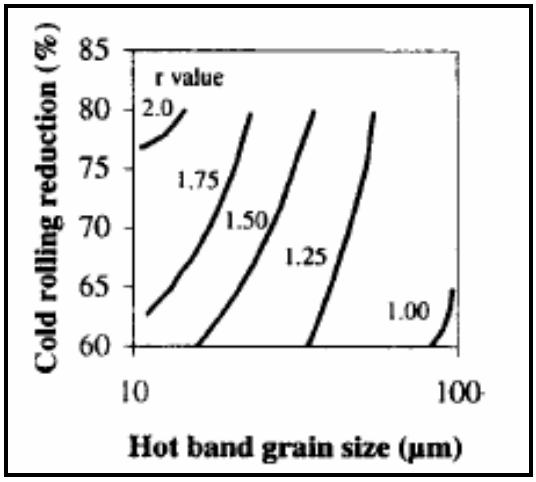


Figure 2.23:Effect of initially grain size and cold rolling reduction on r-value of IF steels [28].

From Figure 2.23 it can be seen that to obtain the optimum final texture for the best drawability, the hot band grain size should be held below 15 µm (ASTM 9-9,5), preferably around 10 µm (ASTM 10-10,5), and cold rolling reduction should be as high as 80 %. The hot band grain size is dependent on the precipitate distribution, which at this stage of processing is dependent on reheat temperature, rolling loads, speeds, and temperature profile through the hot mill. A number of authors [47-49] have found that hot rolling with high reductions and high speeds increases r value and elongation in some steels (Figure 2.24); this is related to precipitation becoming large and sparse at high reductions.

After finding different r values for various kinds of steel at a given grain size, Takahashi et al. [48] proposed that it was not only the small initial hot band grain size but also attaining grain growth through the preferential nucleation of recrystallized grains. Recrystallization and grain growth during annealing is linked to the pinning efficiency of precipitates, finer precipitates having greater pinning effect, limiting r value in the final product. . Subramanian et al.[50] examined the precipitate distributions in two steels with comparable a sparse dispersion of coarse precipitates exhibits a stronger intensity of {111} texture and higher grain boundary mobility than a steel with a similar chemistry and a dense dispersion of fine precipitates. The overriding controlling factor was the effect of the precipitates on grain boundary mobility during recrystallization and grain growth. Therefore a fine balance must be struck between optimization of the hot band grain size and optimization of precipitation through careful control of the hot rolling conditions.

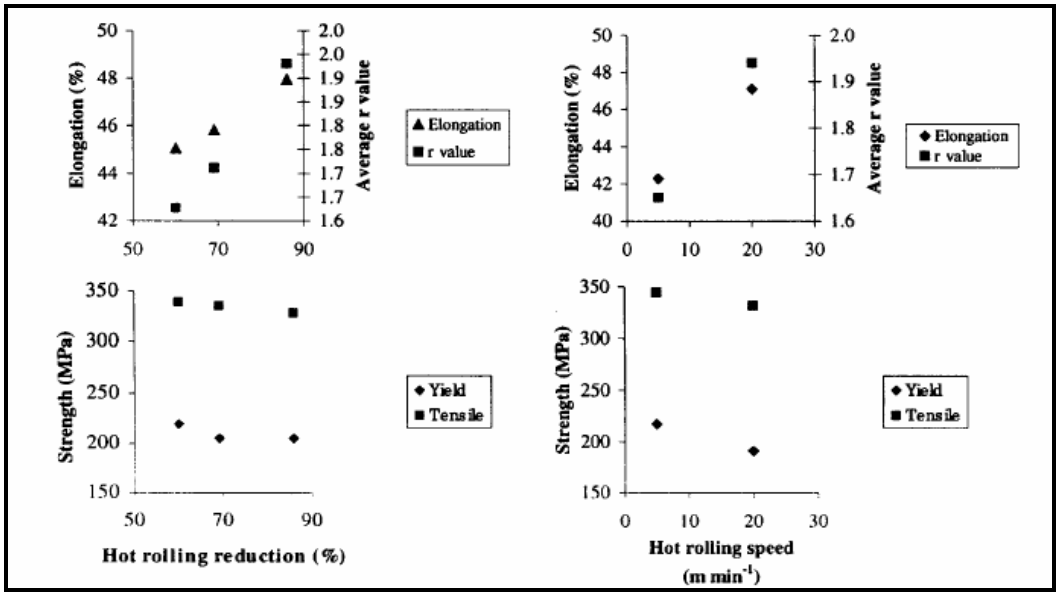


Figure 2.24:Effect of hot rolling reduction and hot rolling speed on mechanical properties of cold rolled and annealed (830 ˚C for 45 s) steel sheet [49].

Irie et al.[51] identified three groups of Ti and/or Nb steels whose behavior in terms of the hot rolling conditions could be distinguished:

(i) for fully stabilized steels containing > 0,0050 %C, precipitation occurs high in the austenite region, minimizing the effects of hot working on the properties of such steel. Steels of this type would have been the early IF steels where extensive vacuum degassing was not available to reduce interstitial contents to a lower level.

(ii) for < 0,0030% C, precipitation occurs after hot rolling during the austenite to ferrite transformation and so very little effect of hot rolling is observed and high r values are obtained. All data in Figure 2.23 should conform to this statement since all steel chemistries contained < 0.0030%C. If precipitation occurs after hot rolling then the curve from work of Gupta et al.[42] would be very shallow, thus explaining the small difference in r value with varying hot rolling condition.

(iii) for 0,0030-0,0050 %C and metal /C =1, precipitation occurs during hot rolling and so much steels are very sensitive to hot working conditions. This statement is more in agreement with Gupta et al. [42] than (ii); the sensitivity of the steel is dependent on the gradient of the slopes leading to the maximum and is composition dependent.

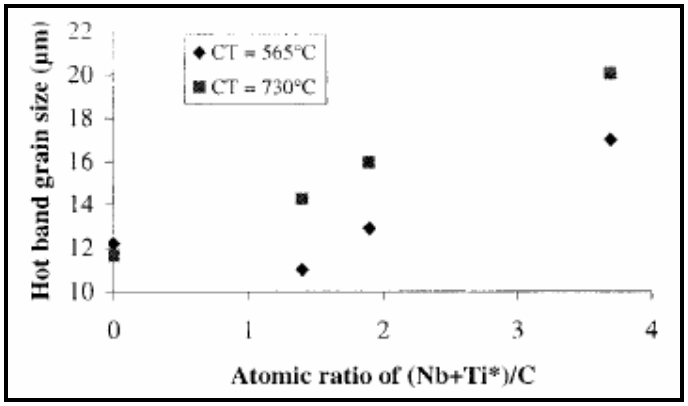


Figure 2.25:Variation in hot band grain size with amount of stabilizer and coiling temperature (CT) on series of experimental IF steels [39].

As can be seen from Figure 2.25, as the amount of stabilizing element increases, so too does the grain size. This may be due to the precipitation of Ti(C,N) occurring at higher temperatures due to the relative "richness" of the addition; thus the precipitates have a greater opportunity to coarsen, enhancing grain boundary mobility which results in a larger hot band grain size [52-53].

2.5.5 Cooling Rate

The cooling rate after hot rolling determines the amounts of C and N in solution in interstitial free steels prior to cold rolling. This in turn, affects the solute C and N levels present during annealing process, which are further modified by the cooling history after annealing. The cooling rate influences the amount of C and N in solution by determining the amount of carbonitride that can precipitate.

For the range of cooling rates employed in accelerated cooling of sheet, the transformation products are either polygonal ferrite (PF) or massive ferrite (MF). Studies of precipitation in normally processed hot band report large (100 to 1000 nm) TiN,TiS, and $(TiNb)_4C_2S_2$ and small (10 to 100 nm) TiC and (TiNb)C precipitates. The MC precipitates often form as coprecipitates on preexisting nitride, sulfide, or carbosulfide particles [44].

One possible way to refine the hot band grain size could consist in the application of accelerated cooling after finish rolling combined with appropriate coiling temperature to minimise grain growth. No major difference in grain size can be observed from the significant increase in cooling rate 15 ºC/sec/. to 400 ºC/sec/. and this independently of the coiling temperature. Mechanical properties of the hot rolled specimens are not very sensitive to the cooling rate after hot rolling. The insensitivity of the hot band grain size relative to the cooling rate is explained by a typical transformation mechanism of the ULC steels. It can be supposed that, taking into account the very pure character of the steel, a kind of nucleation site saturation occurs even for

relatively low cooling rates. Considering such nucleation site saturation, the final grain size will depend only on the amount of nuclei which depends in its turn on the austenite grain size just before transformation. The hot rolling finishing temperature is therefore considered to be much more important for the final hot band grain size than for the cooling rate after hot rolling [45]. The lowering of the coiling temperature does not lead to a reduction of the grain size for the applied cooling rate.

Reducing the onset time for cooling is not only advantageous for the grain refinement of ferrite, but also weakens the dependence of the finishing temperature on grain refinement. Since the fluctuation of the temperature in a hot strip is unavoidable in practice, the reduction of the onset time for cooling is an effective and practical means for reducing the fluctuation in microstructure and consequently in the mechanical properties of the product. The increase in the final reduction refines the grain size of ferrite regardless of the onset time of cooling. The influence of the cooling rate on the grain refinement of hot bands is relatively small. The increase of the cooling rate from 40 to 120 ºC/sec achieves a modest grain refinement of ASTM G.S. No. 0,5. The reason lies in the rapid progress of transformation which does not allow a large overcooling [44].

High cooling rate of hot band of Ti alloyed IF steel after hot rolling finishing and after coiling can cause the presence of carbon in fine cementite or Cottrell segregation as a result of carbon aging ratio is sufficiently high. This carbon effects on properties of hot band through increasing of yield strength together with appearance of yield point elongation, and by ferrite grain refining. This carbon also hinders growth of grains during continuous annealing and thus impairs formability of cold rolled annealed steel.

2.5.6 Coiling Temperature

Many authors [51, 52] believe that the mechanical properties of Ti added steels vary little with coiling temperature, making steels with stoichiometric Ti/N ratios robust to such processing variables. This is attributed to the stability of precipitates in the steel, particularly coarse TiN.

In Ti-Nb steels, however, both Kino et al.[53] and Ito et al. [54] found that a higher coiling temperature resulted in being observed when Nb/C<1.

Increasing coiling temperature resulted in higher r-values and elongation while yield and tensile strength decreased; the effect on elongation is particularly strong as seen in Figure 2.26. A high coiling temperature favours precipitate growth resulting in coarse, widely spaced precipitates that are advantageous for ferrite recrystallization and grain growth during annealing, both of which are highly

desirable in the formation of the favorable texture required for deep drawing applications.

Precipitate growth is especially important in Nb containing steels where very fine Nb carbides and carbonitrides are likely to form which have a deleterious effect on formability.

Increasing the coiling temperature has the beneficial effect of reducing the recrystallization temperature (Figure 2.27) as a result of this the hot band grain size increases, which adversely affects r value development as previously seen in Figure 2.23.

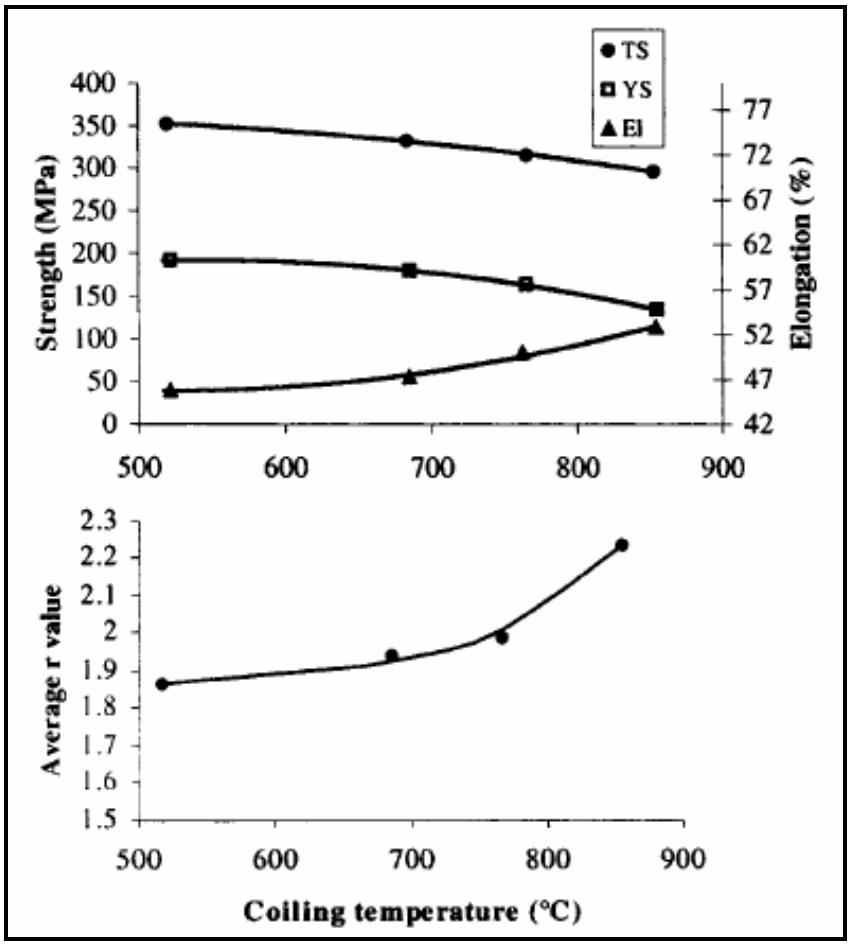


Figure 2.26:Effect of coiling temperature on elongation (El), yield strength (YS), tensile strength (TS), and r value of mill processed IF steel containing 0,003 wt % C (Nb/C=1,7, continuously annealed at 830 ˚C [44].

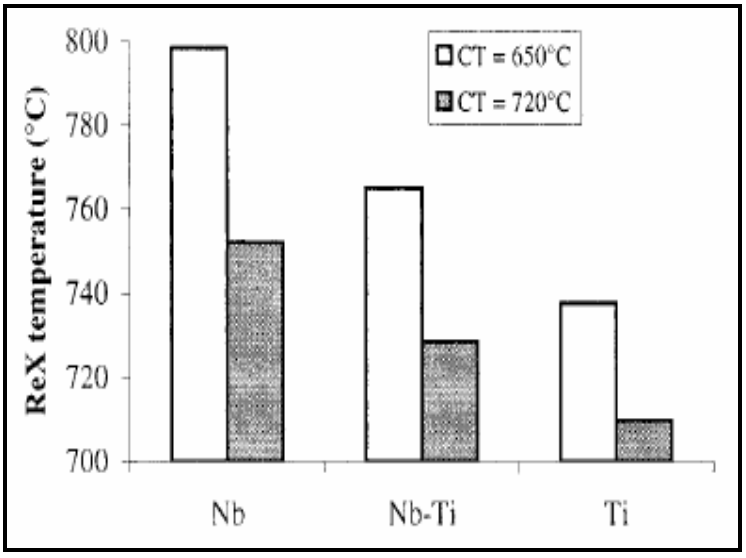


Figure 2.27:Effect of coiling temperature (CT) on recrystallization temperature of Ti, Ti-Nb, and Nb IF steels [55].

If the prior processing and chemistry within reasonable limits (i.e. excess Ti, finishing temperature close to but above A_3), acceptable r values should be produced at all coiling temperatures.

If, however, prior processing and chemistry are not within specification, then a high coiling temperature may be beneficial in “recovering” the product. The higher coiling temperature allows more time for the removal of C from solid solution through the growth of C containing precipitates of Ti carbosulphide and Ti carbides in Ti and Ti-Nb IF steels. The disadvantages of employing a high coiling temperature are twofold [56];

(i) There is a greater change of increased micro structural differences due to the variation in heat transfer rates between the inner and outer laps and the bulk of the coil which then translate into property variations in the cold rolled and annealed product.

(ii) There may be more extensive scaling that requires a more effective scale removal or pickling system to be adopted before cold rolling.

Low coiling temperatures (<600 ˚C) promote the formation of fine, closely spaced precipitates due to the driving force for precipitate growth being lower.

These inhibit grain growth and recrystallization and retard the development of the favourable textures required for deep drawability but maintain a preferred fine grain size.

3. COLD ROLLING AND ANNEALING OF IF STEELS

There are two main functions of the cold rolling process: to produce the required gauge for the customer; and to optimize the r value of the material through the production of a high energy deformed structure from which recrystallization of grains with favourable textures can be nucleated and grown during annealing. Figure 3.1 illustrates the effect of degree of cold reduction on r value for three types of IF steel [21]. It is widely known that strength, ductility and strain hardening index (n) change very little with the degree of cold reduction. A maximum cold reduction of approximately 90 % results in the best r value in all three steels. In reality, cold reductions of 90 % are impractical due to the limitations of hot rolled gauge and cold rolling mill power, hence, reductions are limited to a maximum of about 80 %. The effect of cold reduction is consistent between three steel types. Directional r value is also affected by the degree of cold reduction; the r_{90} value tends to be higher than the r value in the remaining directions except at very high cold reductions (>90 %) where r_{45} becomes more dominant.

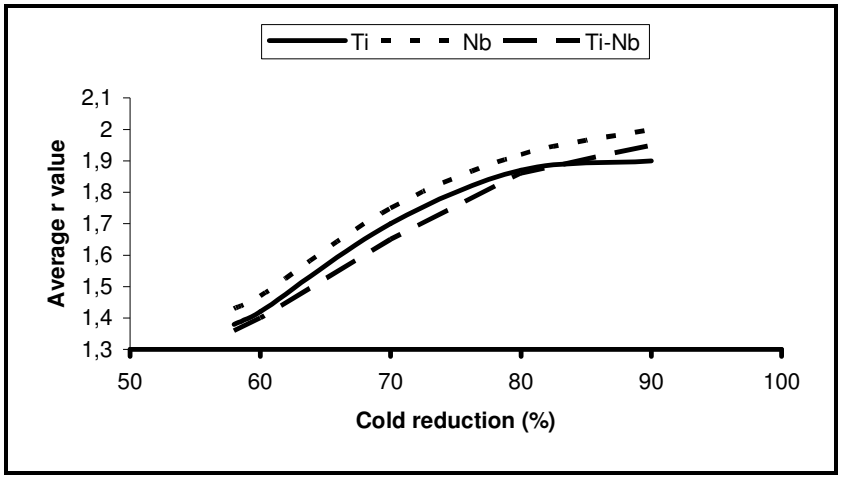


Figure 3.1:Effect of cold reduction on r value of Ti, Ti-Nb, and Nb IF steels[21].

Due to the high coiling temperatures associated with the production of IF steels on the hot mill, a reduction in HCl pickling efficiency can be anticipated. The introduction of a tension leveller at the entry of the pickle line has been successful overcoming this effect. The majority of IF steel production is aimed for automotive applications where a high level of formability is essential. In particular, the r value

(index of deep drawability) is required to be above 2.0 in order to resist fracture when the sheet is subjected to severe deformation.

The cold rolling process is characterized by complex and non uniform plastic deformation. As a result of cold rolling, residual stresses develop in the material. Both the elastic and plastic behavior of iron is to some extent anisotropic. Therefore, besides developing residual stresses, a cold rolled piece of iron will also possess a certain texture. Texture is characterized by the existence of preferred orientations in a polycrystalline material.

The formation of dislocations and their glide mechanisms are responsible for the development of texture as well as residual stresses. The formation of dislocations, however, also has the effect of strengthening the cold worked product. This work hardening increases with increasing plastic deformation. Work hardening is known to disappear continuously during hot working of materials. The processes involved are dynamic recovery and recrystallization, where the latter takes place at higher temperatures than the former.

The temperatures at which both dynamic recovery and recrystallization may take place decrease with increasing dislocation density due to increasing plastic deformation. The influence of dynamic recovery and recrystallization on the texture of the rolled material will be different. While the dynamic recovery will give rise to a polygonization in the individual crystallites only, the recrystallization will actually form completely new crystallites through a nucleation and growth process.

Therefore, unlike recovery the recrystallization may lead to a texture that is different from the non recrystallized material.

It is well known that texture control is essential for manufacturing deep drawing steel sheets. This texture control has usually been performed in the cold rolling and annealing processes.

The texture of IF steels, and their formability depends on five distinct metallurgical mechanisms, each of them influences the texture produced in a particular stage of processing. In the order of occurrence during rolling, these mechanisms are;

i) austenite deformation

ii) austenite recrystallization

iii) the austenite to ferrite transformation

iv) ferrite rolling

v) ferrite recrystallization.

When IF steel textures are being discussed, attention is generally focused on stages iv) and v) above .i.e. on the individual effects of cold rolling and of annealing. Much less importance is attached to the first three mechanisms, which are the ones that determine the hot band texture. Nevertheless, the latter plays a key role in the formation of annealing textures through its influence on the preferred orientations developed during cold rolling [57].

Increased amounts of cold reduction result in lower elongations and n values, higher yield and tensile strengths, and higher r values at cold reductions ranging from 65 % to 80 % for continuously annealed Ti IF steel. The r values for low carbon steels typically peak at a cold reduction near 70 % whereas r values for IF steels peak near 90 %.

Cold rolled sheet steels were formerly classified by their r value and total elongation into four grades, CQ (commercial quality), DQ (drawing quality), DDQ (deep drawing quality) and EDDQ (extra deep drawing quality).

3.1 ANNEALING OF IF STEELS

Texture control and hence control over r value is achieved through primary recrystallization although the mechanism controlling the formation of recrystallization textures is not fully understood. This reflects the presence of complex interactions involving such factors as the solute carbon and nitrogen levels, the presence or absence of fine precipitates, and hot band grain size [9]. Typical annealing treatments produce some degree of grain growth following the completion beneficial in general, so control of second phases which may impede the process is an important aspect of practical texture control [17].

It is believed that a dense dispersion of fine precipitates does no impair the recovery and nucleation of recrystallized grains with {111} orientation. However, the pinning force exerted by the particles on the grain boundary retards the growth of these grains. With the retardation in the growth of grains with the {111} orientation, competitive growth of grains with other orientations, notably {100}, occurs, resulting in a decrease in r value. The greater the proportion of {111}/{110} orientations, the higher the r value that results. After cold rolling, annealing is employed to produce the recrystallized microstructure necessary for drawabilty. Ray et al.[36] have summarized the steps that lead to the formation of annealing textures in low and extra low carbon steels as follows:

(i) when recrystallization starts, because of the high stored energy, and high driving force for recovery, the first nuclei assume the {111} orientation

(ii) these nuclei initially grow at the expense of deformed regions of the same orientation

(iii) once they have reached a critical size, they grow at the expense of adjoining deformed zones to which they are related by <110> rotations

(iv) nuclei of lower stored energy orientations, such as {112} and {110}, begin to appear. Recrystallization is essentially finished when the nuclei, which have now become grains, occupy the entire volume of the material.

Within the literature, there is debate as to whether the strong development of the {111} type orientations during annealing is dominated by the nucleation or growth of grains with favourable textures.

During continuous annealing, r value increases with increasing annealing temperature (Figure 3.2) until the ferrite to austenite transformation temperature reached, whereupon r value decreases due to the production of a mixed grain structure of small and large grains. Figure 3.2 also shows the relative insensitivity of Ti IF steels to coiling temperature, unlike Ti Nb IF steels, as discussed in the section of “Coiling Temperature”.

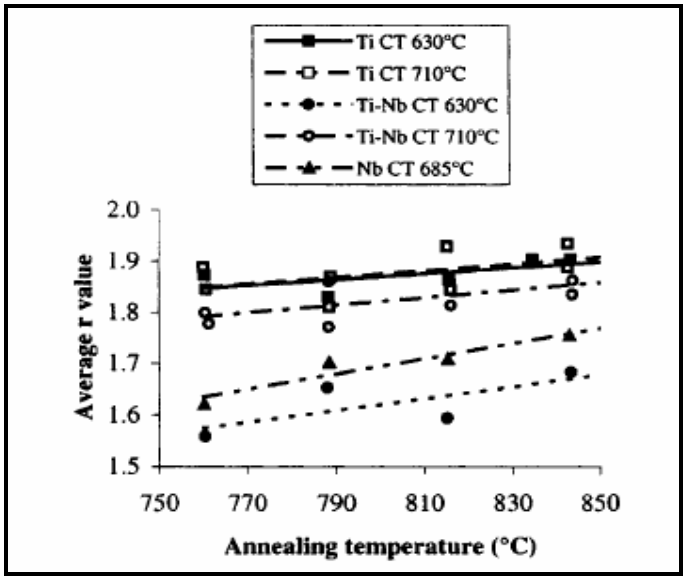


Figure 3.2:Effect of continuous annealing temperature on r value of Ti, Nb, and Ti-Nb IF steels [42]. (CT=Coiling Temperature)

The final crystallographic texture of the steel and hence good drawability is due to the degree of deformation during annealing. High r values and elongations are favoured by the large recrystallized grain size produced during annealing; this is true for IF steels produced by both batch and continuous annealing routes (Figure 3.3). In general terms, the annealing of cold rolled low carbon sheet involves:[36]

(i) partial dissolution of carbides, during heating, liberating some carbon that diffuses into the matrix

(ii) recovery

(iii) recrystallization of the deformed ferrite

Interstitial free steels subject to continuous annealing must be designed so that recrystallization and grain growth is possible in a short period of time, to accommodate the high heating rates and short soaking times employed in commercial practice. The formation of carbide, nitride, and carbonitride precipitates is thought to contribute to the reduction in the time required for recrystallization and grain growth to take place.

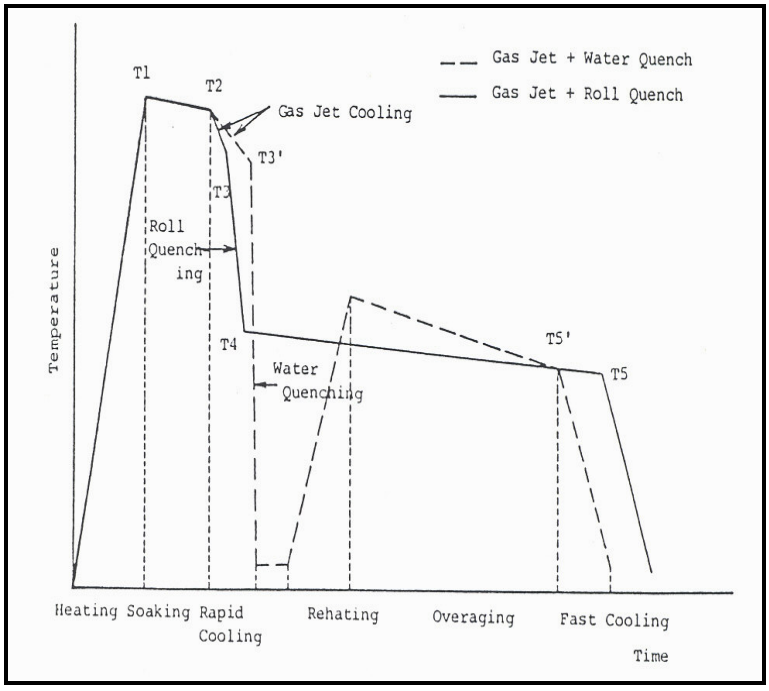


Figure 3.3:Schematic Annealing Cycles of CAL Process [36].

3.1.1 Soaking Temperature and Soaking Time

Osawa et al [58] found that the hardness decreased within increase of soaking temperature for a given composition of Ti stabilized IF steel. The grain size number became small with the increase of soaking time. However, this tendency was not clear at high temperature soaking. In the case of 900 ºC soaking, the change of grain size number was small within short time soaking. However, as soaking time increase, grain size number decreased at soaking for 420 s to 1800 s. The r-values were abruptly increased by soaking for longer than approximately 420 s because of full re-crystallization. On the other hand, the r values steadily increased with increasing annealing temperature. For longer times of over 420 s, however, a

dramatic increase of r-value resulted [58]. For the Continuous Annealing Line (CAL) Technology, the mechanical properties are largely affected by soaking time, and this is caused by ferrite grain growth. As to soaking for 30 s, soaking is insufficient. But if a soaking time is longer than 40 s, changes in these properties are small, and the longer the soaking time, the lower yield strength and the higher r value gradually. Since the soaking time directly affects the furnace length and line speed, minimization in soaking time is desirable in view of processing merits. Judging from the above results, permissible mechanical properties are obtained even for the soaking time for 40 sec, a soaking for longer than 40 s, as a matter of course, causes further improvement of the mechanical properties [59].

3.1.2 Cooling Rate

In case of water quenching method, CQ grade material annealing profile is, after heating and soaking at 700 ˚C, cooled down to 560 ˚C by gas jet and then water quenched. The cooling rate is about 1 to 2 ˚C/sec. Then it is reheated to 400 ˚C, and after being given over-aging treatment, it is fast cooled from 290 ˚C to less than 160 ˚C. The over-aging time is about 1 min. In the case of roll quenching method, CQ grade material is, after heating and soaking at 700 ˚C, cooled to 600 to 550 ˚C by gas jet then cooled by cooling rolls to 400 ˚C with the cooling rate of 100 ˚C to 400 ˚C/sec. Then over-aging treatment is accomplishment within 2 to 3 min. until cooled to outlet's temperature about 320 ˚C in over-aging zone. As IF steels are used for DQ, DDQ, EDDQ, there is no metallurgical restriction on the cooling process after they are annealed at 770 ˚C, 800 ˚C and 820 ˚C. Over-aging treatment is not required, but they may go through the over-aging zone and be subjected to an over-aging treatment, if the production facility requires this [59].

3.1.3 Hot Dip Galvanising

In an effort to combat corrosion, most automotive manufacturers are now using coated steels in the production of automobile body panels. One of the main criteria is that once coated, the material must still be amenable to press forming without any deterioration in product quality. It has been reported both at Inland Steel Company [47] and elsewhere [60] that the r value of galvannealed steel is reduced by ~0.2 and elongation by ~1 % in comparison with the uncoated product. Thus, steels that were used previously in the uncoated condition, once coated, may perform less well during the press forming operation. The reduction in press performance is due to the properties of the coating and the substrate/coating interface. A drop in both r value and elongation is unavoidable but may be minimized to the relationship between the drop in r value and the r value of the uncoated steel is linear, the drop being higher

at higher r values and less at lower r values (Figure 3.4). The value at which the drop in r value becomes negligible approaches unity for substrates with an Fe-Zn (galveanneal) coating and 1.4 for Ni-Zn (Nizec) coated substrates if the lines on Figure 3.4 are extrapolated. Thus, the specification of the substrate material before galvannealing must be higher than that required in the finished product to compensate for the deterioration in properties on coating.

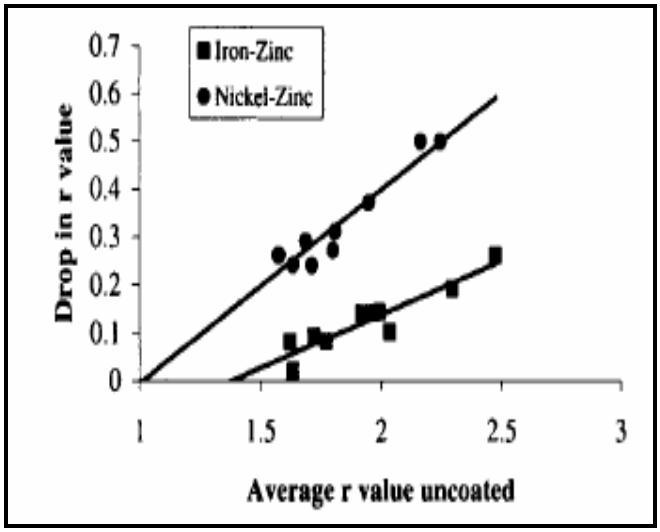


Figure 3.4:Relationship between substrate r values and drop in r value that occurs after applying Fe-Zn and Ni-Zn alloy coatings [61].

The corrosion resistance and micro structural characteristics of zinc coated steels depend heavily on the hot dip coating practice employed. Coating performance in the pres depends on the crystallographic texture of the coating, the coating weight, and the phases present in the coating. For example, galvannealed steel sheet with a coating weight of 60 g/m^2 that is used in the manufacture of corrosion resistant automotive body panels can be affected by flaking and powdering problems in the press if the phase composition of the coating is incorrect. Both substrate chemistry and the rate of Fe-Zn alloy formation during galvannealing and hot dip galvanizing have an effect on the phase composition of the coating which in turn affects the powdering / flaking performance of steel.

Generally, rates of Fe-Zn alloy formation during hot dip galvanizing and galvannealing are accelerated by the steels, C does not segregate to the grain boundary as it would in Al killed steels, instead it is tied up as TiC precipitates. The grain boundaries are therefore relatively free from any segregated C and Fe diffusion can occur relatively unhindered. The Ti-Nb IF steels make better substrates for hot dip coating and galvannealing than Ti IF steels, due to a more uniform coating section taking place which results in greater uniformity in the coating structure. The rate of Fe-Zn alloy formation is accelerated as excess Ti content

increases and the effect of grain size on Fe-Zn formation is marked. IF Nb were absent, as in Ti IF steel substrates, the Fe-Zn reaction would proceed at such a rate as to produce “outburst” phenomena, that is, an uncontrolled reaction of the Zn in the coating and the Fe in the substrate. The presence of Nb at or near the grain boundaries enables regulation of the Fe-Zn coating reaction by slowing it down. Figure 3.5 illustrates the improvement in powdering resistance with Ti-Nb IF as substrate [62]. As can be seen, greater coating weights can be achieved for Ti-Nb substrates for the equivalent powdering resistance of Ti only substrates.

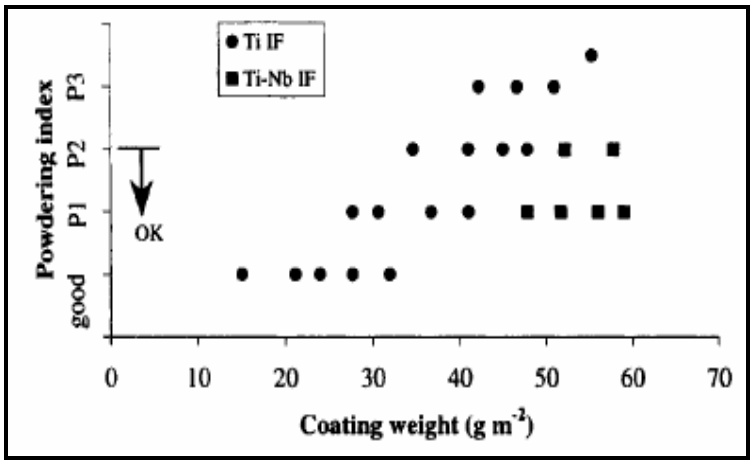


Figure 3.5:Powdering evaluation of galvannealed steel sheets using 45 ˚ V notch bending test: Ti IF steel 0,003%C, 0,057 %Ti; Ti-Nb IF steel 0,003 %C, 0,014%Ti, 0,018 %Nb; 0,10 %Al in Zn bath [62].

When the average Fe content in the coating is too low, the coefficient of friction is high due to the presence of ζ phase ($FeZn_{13}$) which leads to flaking and/or insufficient material flow during forming [63].In contrast, when the average Fe content in coating is too high, the coating tends to peel off into powder during forming due to the presence of Fe_3Zn_{10} adjacent to the steel substrate [63].

3.1.4 Skin pass Elongation of IF Steels

A relatively light cold rolling operation that may be used on cold rolled and some coated steel such as galvanized for improving mechanical properties of steel. Temper rolling of sheet steel improves surface finish, alters mechanical properties and reduces the tendency of the steel to flute during fabrication.

Consequently, the optimum skin pass elongation in actual production is to be changed according to chemical composition, hot rolling condition, cold reduction and continuous annealing conditions etc., and optimum skin pass elongation should be determined after consideration on the factors for each product grade [59].

4. ERDEMIR IF STEEL PRODUCTION

4.1 Steel Production

4.1.1 Basic Oxygen Furnaces (BOF)

The existing BOF shop capacity is 3 million tons per year and certain amounts of scrap. Slag makers and lime are added into the hot metal obtained from the three converters each with a capacity of 120 tons (Figure 4.1). The hot metal coming from the blast furnace is transferred to the BOF furnaces [68]. Then the hot metal is stirred through special pipes blowing pure oxygen into it so that steel is produced in accordance with existing international standards. Before the BOF process takes place, the hot metal is transported to the desulphurization facility. After blowing oxygen into the BOF furnaces, the steel is deoxidized by adding Al or Si elements that have high oxygen affinities [69].

The micro alloyed steels contain a little amount of alloying elements such as V, Ti, Nb and Mo elements. These types of steel contain 0.60-1.60 % Mn elements to guarantee the required steel quality.

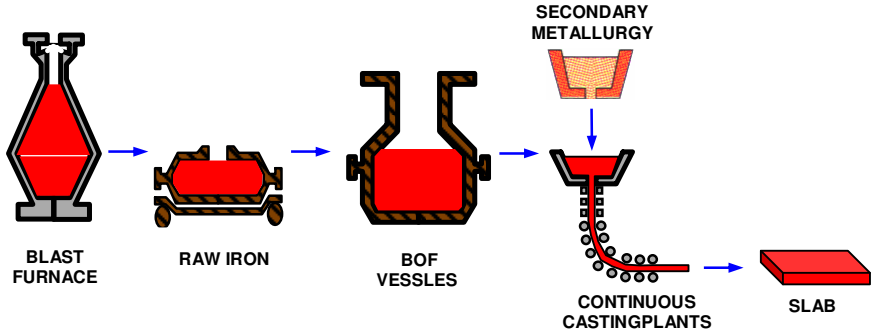


Figure 4.1:Steel production and continuous casting processing diagram [70].

4.1.2 Secondary Metallurgy Process

In order to be able to produce high quality steel in Secondary Metallurgy Facilities, three processes of a ladle furnace are necessary: gas removing under vacuum, chemical heating, and alloying facility at a secondary metallurgy plant.

In the caster furnace, liquid steel is heterogenic in its chemical composition and temperature distribution. Therefore, liquid steel is stirred with argon gas at the secondary cast facility station in order to obtain a homogenous steel composition and temperature.

Al and Si elements are added in the chemical heating station. Further, spherical processing with wreckage flotation for inclusion growth and Calcium (Ca) is added in this station. In order to have enough wreckage spherical processes, the amount of Ca element should be 20-30 ppm. The ERDEMIR Chemical Heating and Alloying Station Facility properties are listed in Table 4.1 [71].

Table 4.1:ERDEMIR Chemical Heating and Alloying Facility properties [71].

ERDEMIR Chemical Heating and Alloying Facility (IR-UT)	
Start-up	December 1995
Capacity	120.0000 tons/year
Type	IR-UT
Snorkel inner/outer diameter	1.2 / 1.7 m
Heating rate with Al/Si	7.0 / 4.0 ℃/minute
Oxygen lance height	6.7 m
Oxygen lance stroke	5.8 m
Snorkel weight	7 tons
Number of alloying bunkers	6
Degusting	Dry type (with nomex bags)

Special steel has a small amount of C, N and H elements that are processed in the RH-Vacuum plant. Liquid steel is processed in the vacuum boiler, which has vacuum pressure of 1 millibar. Vacuum degassing of the steel is necessary to get to very low carbon levels required for IF steels, for example, smaller than 0.005 % C. The most common type of degasser currently in use is the circulation-type degasser. It has two refractory tubes or snorkels that immerse in the liquid steel inside of the ladle. A vacuum is applied in the chamber above the snorkels, and argon gas is injected into the steel through a number of injection points located in one of the snorkels. The argon gas locally reduces the density of the steel and causes an

upward movement of the steel, which subsequently re-enters the ladle through the other snorkel leg. In this way, a re-circulation flow is established. The technical properties of the ERDEMIR RH Vacuum Degassing Facility is shown in Table 4.2 [71].

Table 4.2:ERDEMIR RH Vacuum Degassing Plant Properties [71].

ERDEMIR RH Vacuum Degassing Facility	
Start-up	March 1995
Pumping capacity (at 67/ 6,7/ 0,67mbar)	2400, 1000, 600 kg/hour
Type	RD-KTB
Heat size	120 tons
Vessel height	7.8 m
Vessel / snorkel inner diameter	708 / 0.5 m
Argon lift gas flow	1200 Nlt/minute
Number of lift gas ports	12
Water ring pumps	3
No, of steam ejectors/intercondensers	3/2
Vessel preheating	Natural gas (target 1470 ^{o}C)
Circulating rate	60 tons/minute
Number of alloying bunkers	16
Vessel design	Quick change type
Snorkel	Welded type

4.1.3 Continuous Casting Plants

Liquid steel is continuously cast into moulds by slab casters and then solidified to produce slabs. There are 4 slab casters in the steel shop.

Continuous casting process is an advanced technology used in the production of semi-finished steel. The liquid steel that is produced in the BOF and filled into the ladle is continuously loaded into a copper mould. The steel is passing through the mould and then cooled with water to solidify as a slab. The slab is then sent to stock

areas after being cut to lengths of 6, 9 or 12 meters. The technical properties of the Continuous Casting Plants at ERDEMİR are shown in Table 4.3 [71].

Table 4.3:Technical properties of the Continuous Casting Plants at ERDEMİR [71].

Continuous Casting Plants				
	ERDEMİR			
Continuous Casting Plants	No.1	No.2	No.3	No.4
Start-up/modern.	Sept. 1978/nov.'00	May 1982/Dec.'00	October 1993	July 1994
Production capacity	550 000 tons/year	550 000 tons/year	850 000 tons/year	850 000 tons/year
Machine type	Curved	Curved	Curved	Curved
Heat size	120 tons	120 tons	120 tons	120 tons
Mould type	Curved	Curved	Straight	Straight
Mould length	800 mm	800 mm	900 mm	900 mm
Slab section width limits	750-1 300 mm	1 000-1 600 mm	750-1 650 mm	750-1 650 mm
Slab thickness limits	200 mm	200 and 300mm	140 and 200 mm	140 and 200 mm
Casting speed range 200mm.	0.8-1.3 m/minute	0.8-1.3 m/minute	1.0-1.6 m/minute	1.0-1.6 m/minute
For CCM 2 300 mm, CCM 3 and 4 140mm	-	0.3-0.5 m/minute	1.2-2.2 m/minute	1.2-2.2 m/minute
Speed range (average)	1.1-1.2 m/minute	1.1-1.2 m/minute	1.1-1.4 m/minute	1.1-1.4 m/minute
Casting time per heat	53-80 minute	47-73 minute	25-70 minute	25-70 minute
Oscillator type	Crank motion	Crank motion	VAI quick change	VAI quick change
	Dual eccentric	Dual eccentric	Single eccentric	Single eccentric
Oscillation stroke	0-12 mm	0-12 mm	0-16 mm	0-16 mm
Oscillation frequency	25-125 /minute	25-125 /minute	60-300 /minute	60-300 /minute
Tundish capacity	21 tons	21 tons	25 tons	25 tons
Tundish liquid steel level	935 mm	935 mm	1 000 mm	1 000 mm
Flow control from tundish	Stopper	Stopper	Slide gate	Slide gate
Mould level control	Automatic (Co 60)	Automatic (Co 60)	Automatic (Co 60)	Automatic (Co 60)
Number of strands	1	1	1	1
Machine radius	9.8 mt	9.8 mt	8 m	8 m
Straightening type	Single point	Double point	Continuous	Continuous
Metallurgical length	18.9 mt	18.9 mt	25 m	25 m
Roll design	Split roller	Split roller	Split roller	Split roller
Length of cut slabs	3.5-12 m	3.5-12 mt	2-12 m	3.5-12 m
Secondary cooling	Air mist	Air mist	Air mist	Air mist

4.2 Hot Rolling

4.2.1 Hot Rolling of IF Steels

The slabs are either produced at the Continuous Casting Plants at ERDEMIR or imported. After their surface was scarfed at the stock field, the slabs are loaded into the slab reheating furnaces to get them to a hot rolling temperature. There are two furnaces available. The first two furnaces have capacities of 110 and 320 tons per hour, respectively.

In the Hot Strip Mill no 2, the slabs are rolled through seven stand finishing mill facilities with a maximum thickness of 1.5 mm and a maximum width of 1525 mm (Table 4.4). The whole production line is computer controlled including an automatic width control, a sliding system with F5-F7 stands, a roll bending system, a hydraulic automatic thickness control, and an automatic jumping/stepping control in the coiler [72].

Table 4.4:Specifications of the Hot Strip Mill no 2 [72].

Units	Specifications
Slab Yard	60.000 tons stock capacity
Slab Reheating Furnaces	2 furnaces, each 318 tons/hour capacity
Slab Sizing Press	Continuous Type, max 250 mm sizing
Rougher	RM1 2-Hi, RM2 4-Hi
Coilbox	For 12 m slabs
Finishing Mill	7 stands, AGC, roll shifting and bending systems
Coilers	2 coilers, hydraulic, Automatic Jump Control (AJC)

4.3.Cold Rolling

The production flow consists of five separate lines in the conventional cold rolling mill which are pickling, cold rolling, cleaning line, annealing and temper rolling processes. But upto date cold mills consist of continuous lines such as Continuous Pickling and Tandem Cold Mill (CPL-TCM), Continuous Annealing Line (CAL) and Continuous Galvanising Line (CGL). Important improvements in product quality and productivity are thus achieved with these lines. The Cold Mill No 2 has 3 main processing lines and has high technology configuration.

4.3.1.Continuous Pickling / Tandem Line (CPL/TCM)

Hot rolled coils are taken from the hot mill exit conveyor and then transported to the pickling / tandem line by coil carrier trucks and then cooled down to ambient temperature. Continuous Pickling and Tandem Cold Mill consists of equipments like welding machines, tension levelers, pickling tanks, rinse tanks, dryers, side trimmers and tandem mill (Figure 4.2) [73].

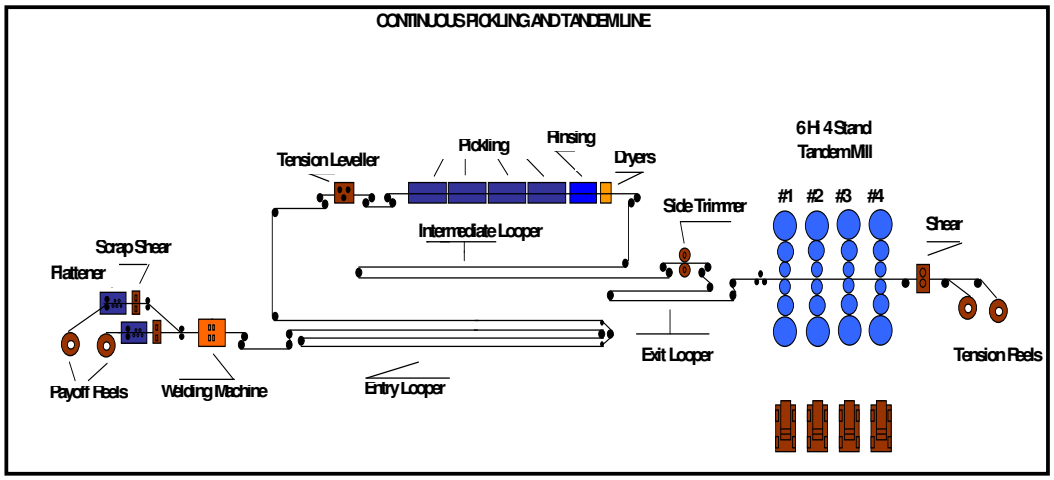


Figure 4.2:Continuous Pickling / Tandem Line Process Diagram [73].

After all these processes are completed, the coils are cold rolled down to the required thickness set by customer and then stocked at intermediate stock yard. After program is given by the Production Planning Department , those programmed coils sent to related lines. They are mostly transferred to the Continuous Annealing Line (CAL) and the Continuous Galvanizing Line (CGL). Some of the coils are also sold just as they are as full hard material. Those full hard coils are used at customers lines as semi finished material. The technical properties of the Continuous Pickling and Tandem Line are shown in Table 4.5 [73].

Table 4.5:Technical properties of the Continuous Pickling and Tandem Line [73].

Technical Properties of Continuous Pickling and Tandem Line

Production Capacity	1.500.000 tons/year	***Inlet Speed***	480 mpm
Length of the Line	300 m	***Pickling Speed***	240 mpm
Coil Weight	28 tons	**Pickling Outlet Speed**	330 mpm
Strip Width	650 – 1550 mm	**Rolling Outlet Speed**	1250 mpm
Inlet Thickness	1.80– 4.50 mm	**Rolling Type**	4 Stand – 6 HI
Outlet Thickness	0.40 – 2.00 mm	**Working Roll Diameter**	475 / 405 mm
Coil Outer Diameter (Inlet)	1295 – 1905 mm	**Intermediate Roll Diameter**	550 / 480 mm
Coil Inner Diameter (Inlet)	762 mm	**Back Up Roll Diameter**	1350 / 1250 mm
Coil Outer Diameter (Outlet)	1200 – 1850 mm	**Total Rolling Power**	20.500 kW
Coil Inner Diameter (Outlet)	508 mm		

4.3.2 Continuous Annealing Line (CAL)

The ERDEMIR NKK-CAL Process is a continuous annealing process, which was originally developed and industrialized by adopting the water quenching method for the production of cold rolled steel sheets (Figure 4.3). This process has not only made it possible to manufacture deep drawing quality cold rolled steel sheets but also enabled to produce the new products, which could not be manufactured by the conventional batch annealing process.

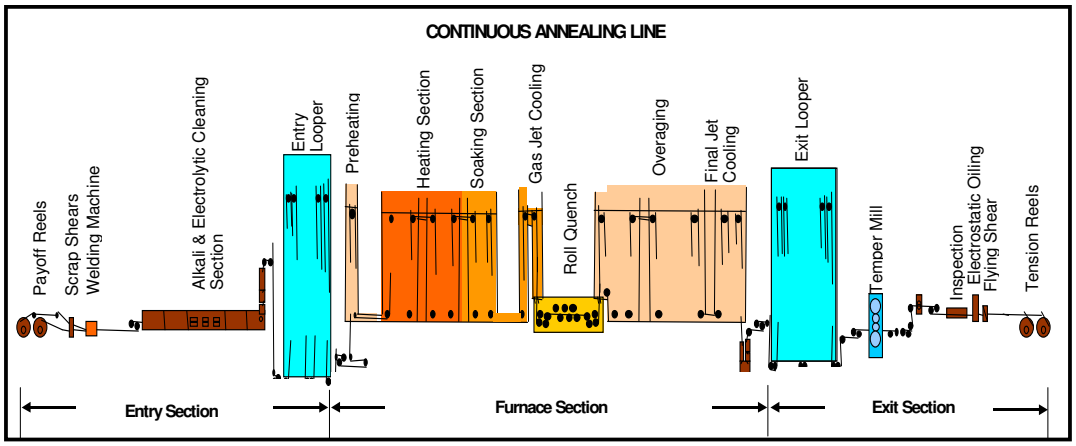


Figure 4.3:Continuous Annealing Line at ERDEMIR [74].

Cleaning, annealing and tempering processes are continuously performed at this line. Instead of Batch Annealing, the coils are not coiled and annealed as strips. This gives a uniform heating to the strip. [74].

4.3.2.1 Principles of ERDEMIR-CAL-Process

Cold sheets have a wide range of applications as represented by uses in automobile bodies and electrical appliances. In most cases, sheets undergo forming processes like press forming, roll forming or bending. Therefore, good ductility and deep drawability are very important properties of drawing quality steel sheets.

For the production of steel with good formability, the following factors and requirements are very important:

a) To provide a steel with a suitable chemical composition and to reduce detrimental impurities.

b) To enlarge the ferrite grain sizes as far as possible within a certain range.

c) To reduce the solute carbon in the ferrite thus rendering the material milder and simultaneously preventing hardening that causes strain aging.

d) To minimize nonmetallic inclusions in the ferrite structure.

e) To form the crystal orientation for raising the Lankford value.

Factors a, b, d and e can be even changed in the conventional continuous annealing process by slightly changing the chemical composition and the hot rolling conditions.

> However, it is not possible to completely precipitate solute carbon in the conventional continuous annealing process. And this factor has a considerable effect on the formability, even if all other factors are favorable. Residual solute carbon has a decisive effect on cold sheets.

4.3.2.2 Process and Features of CAL Heat Cycle

The technique developed by NKK Corporation is based on rapid quenching and short period over-aging. The heat cycle of CAL is as follows [Figure 4.4].

a) The heating and soaking are basically similar to those in conventional continuous annealing lines.

b) By rapid cooling down below 400 ˚C through water quenching or roll quenching methods, supersaturated solute carbon is obtained.

c) Reheating to 350 ˚C to 400 ˚C and holding at this temperature in the roll quenching method, causes the generation of innumerable carbide precipitation of solute carbon. Over-aging at this temperature permits almost the complete precipitation of solute carbon within a short time.

d) Cooling to less than 160 ˚C and discharge from the furnace [74].

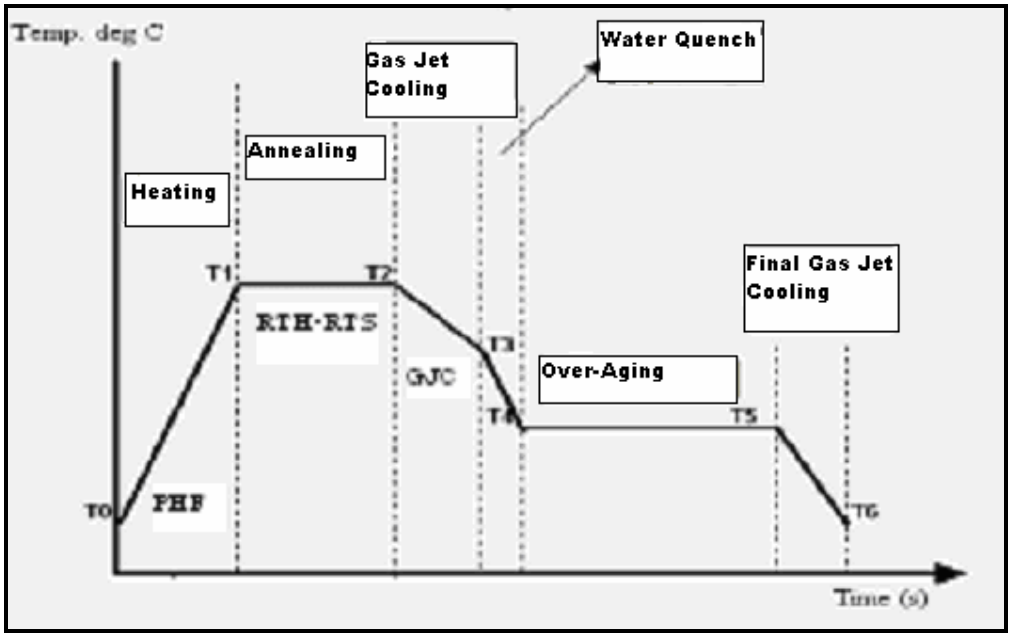


Figure 4.4:The Heat Cycle of the Continuous Annealing Line [74].

4.3.2.3 CAL Heat Cycle

Table 4.6 shows the representative continuous annealing cycles in ERDEMIR-CAL-Process corresponding to each product grade.

In the Roll Quenching method, CQ grade material is, after heating and soaking at 700 ˚C, cooled down to 600 to 550 ˚C by gas jet and then cooled by cooling rolls to 400 ˚C with the cooling rate of 100 ˚C to 400 ˚C/sec.

Then the over-aging treatment is completed within 2 to 3 min. until cooled down to outlet's temperatures of about 320 ˚C in the over-aging zone.

As IF steel is used for DQ, DDQ or EDDQ, there is no metallurgical restriction on the cooling process heat treatment or soaking at 770 ˚C, 800 ˚C and 820 ˚C.

The over-aging treatment is not required, but steel may go through the over-aging zone with some over-aging treatment, if the production facility requires it [65].

Effects of processing factors in ERDEMIR-CAL Process on the mechanical properties of products are basically the same except some differences caused by the primary cooling rate.

The relationship between processing factors and product qualities are summarized as follows [74].

Table 4.6:Continuous Annealing Cycles in ERDEMIR-CAL-Process corresponding to each product grade [74].

Product Grade	Strip Temperature (°C)	Time (Sec.)

	Radiant Tube Soaking	Gas Jet Cooling	Roll Quench	Over Aging	Final Jet Cooling	Radiant Tube Soaking	Over Aging
CQ	700 ± 30	550 600 ± 25	400 ± 40	320 ± 30	≤ 160	≥ 40	≥ 150
DQ-1	770 ± 30	550 600 ± 25	400 ± 40	320 ± 30	≤ 160	≥ 40	≥ 150
DQ-2	770 ± 30	-	-	-	≤ 160	≥ 40	-
DDQ	800 ± 30	-	-	-	≤ 160	≥ 40	-
EDDQ	820 ± 30	-	-	-	≤ 160	≥ 40	-
HSS-45	800 ± 30	550 ± 25	400 ± 40	320 ± 30	≤ 160	≥ 40	≥ 150
HSS-60	800 ± 30	550 ± 25	400 ± 40	320 ± 30	≤ 160	≥ 40	≥ 150

In the preheating section, the strip is heated before given to the annealing furnace. The idle exhaust gases (HNX) are used in this section. Burning the natural gas in the RTH1, RTH2 and RTS radiant tubes forms the exhaust gas, then sucked by two exhaust fans and blown upon the strip to prevent any heat losses. To prevent oxidation the strip surface is held at 160 ^{0}C. Oxidation occurs very quickly, if the temperature is over 160 ^{0}C so that the exhaust gas is mixed with cold air for balancing the temperature. After that, the gas is blown upon the strip surface by jet fans. The maximum value of the gas is 400 ^{0}C. After completing this process, the exhaust gas is taken outside by a top fan.

In the Radiant Tube Heating (RTH) section the strip is heated to a maximum temperature value shown in the temperature values in Table 4.6. The Radiant Tube Heating section is grouped in two sections as RTH 1 and RTH 2. In RTH1, a strip is passed through 13 times, in RTH 2 it is 12 times. The heating process takes place with 'U' and 'W' shaped radiant tubes. These tubes work under negative pressure. They have a manual closing valve, gas control valve and exhaust gas control valve in all zones. Their total numbers are 256 tubes in RTH 1 and 119 tubes in RTH 2. Natural gas is used in the radiant tubes. Their capacities are 85.000-120.000 Kcal/h.

In the Radiant Tube Soaking section, the annealing temperature is held constant to obtain wanted mechanical properties of the material. A strip passes 9 times this section as well as 24 tubes. The heat capacities of these tubes are 82.500 Kcal/h.

In the Gas Jet Cooling section, there are two parts that each strip passes once. A strip is cooled in this section before passing the roll quench section. The section is divided into 5 zones. There are several cooling fans in each zone that have various speeds. The furnace atmosphere gas (5-7 %H – 95-93 %N: HNX) is sucked in by a fan and cooled by heat exchangers. Then the cooled gas is blown upon a strip by tubes. There are 2 heat exchangers that are cooled by water.

In the Roll Quenching (RQ) section, a strip goes through 6 rolls with water passing through them. The strip temperature is decreased to a required value by controlling the pressure of the rolls. In roll quenching processes, any pickling equipment is not

necessary and the final temperature of a strip after the quenching can be adjusted to that of an over-aging temperature so that no reheating is required. Regarding manufacturing of mild steel sheets, a roll quenching temperature should be maintained between 500 to 600 ˚C to have better ductility and aging properties.

In the Over-Aging (OA) section, the temperature is about 400 ℃. A strip goes through this section about 2 to 3 minutes. The section is divided into two groups; OA1 has 17 strip passes and OA2 has 15 strip passes. Electrical radiant tubes do the heating. Cooling is done in the OA2 section by two fan systems circulating HNX. There is no cooling system in the OA1 section.

In the Final Gas Jet Cooling (FJC) section, a strip is cooled down to 160 ℃ in order to prevent surface oxidation. It contains 48 jet coolers. A strip goes through 6 rolls and is then cooled down. The total capacity of fans is 15.000 m^3/h, and the heat exchangers have a total capacity of 120.300 Kcal/h.

In the Final Water Quench (FWQ), a strip is cooled down to 40 ℃ by spraying water onto the strip surface. Then the water is dried by an air dryer.

The typical features of the water quench method are as follows:

a) High cooling rate (about 1000 ˚C/sec) makes it possible to reduce an over-aging time (about 1 minute).

b) High strength steel sheets of 350 to 1500 Mpa in tensile strength with a wide variety of grades are easily manufactured by quenching temperature control.

c) When manufacturing mild steels, the quenching temperature should be controlled with proper temperatures to realize a better balance of ductility and aging properties.

In the Skin Pass Mill section, the 4 high skin pass mill is used with an 1100 mm diameter backup roll and a 460 mm diameter work roll. The work rolls are textured with an Electrical Discharge Texture (EDT). The maximum roll force within the roll gap is 1000 tons. The maximum bending forces are 30 tons. Tension within the mill area is up to 38 tons applied by four bridle rolls at the entry side of the mill.

The control system of the mill allows strip processing with constant roll force or elongation control via roll force or tension. Tension measuring units are placed at the mill's entry and exit side. A tension leveler is located at the bridle rolls at the entry side.

4.3.3.Continuous Galvanizing Line (CGL)

For over a century, zinc has enhanced the longevity and performance of steel. Zinc coatings provide the most effective and economical way of protecting steel against corrosion. Zinc-coated (galvanized) steel offers a unique combination of positive features. The key ones for residential construction are [75]:

- High strength, determined by the steel substrate,

- Formability, a key feature for roll forming coated steel sheets,
- Light weight of steel framing and roofing as compared to competitive materials,
- Corrosion resistance for long life durance and esthetic appearance,
- Recycling of scrap material and after end-of-life demolition, and
- Low cost, competitive with all construction materials of matching quality.

For these reasons, galvanized steel is an ideal material for a multitude of building applications. In the residential construction market, galvanized steel has particular and cost-effective applications in framing, roofing, rain ware (gutters and downspouts), ductwork (heating/cooling and venting) and household appliances.

Galvanized steel resists corrosion. Since in industrialized nations, at least 4% of galvanized steel production is lost to corrosion each year, the trend in last years has been more towards customer and manufacturer needs for increased protection by higher contents of zinc and additional applications of zinc-coated steel [72].

The continuous hot-dip coating process for steel-sheet products is a widely used method. It was originally developed 50 years ago for galvanizing (zinc coating). Now it is also applied to other metals than steel sheets, and the early practitioners would hardly recognize the coating lines of today. It has become a very sophisticated, technically advanced operation, enabled by the availability of advanced electrical/mechanical control systems. Originally, products were used for applications that did not require high product quality or formability. Today, the consumer industries are using hot-dip coated products for the most demanding applications; items such as automotive hoods, fenders and doors. Not only do these applications require a high degree of surface quality, but they also demand a high degree of formability. Hot-dip coated sheet is produced by CGL in thicknesses from 0.25 mm to 4.30 mm [75].

4.3.3.1.The Continuous Galvanizing Process (CGL)

In the ERDEMIR Galvanizing line that started in the middle of 2001, the hot dip galvanizing process is used, which is one of the two main continuous galvanization technologies applied today. In this line, CQ, DQ, DDQ, EDDQ, BH, and HSLA steel qualities are produced. Besides, five different processes (cleaning, annealing, galvanizing, tempering, and chromation) are continuously done on the same line.

In the Galvanizing line, not only galvanized sheets are produced for different markets like building, construction, and electrical goods, for example, but also high quality galvanized sheets are made for the automotive industry and similar sectors. The annual capacity of the facility is 250,000 tons (Figure 4.5) [75].

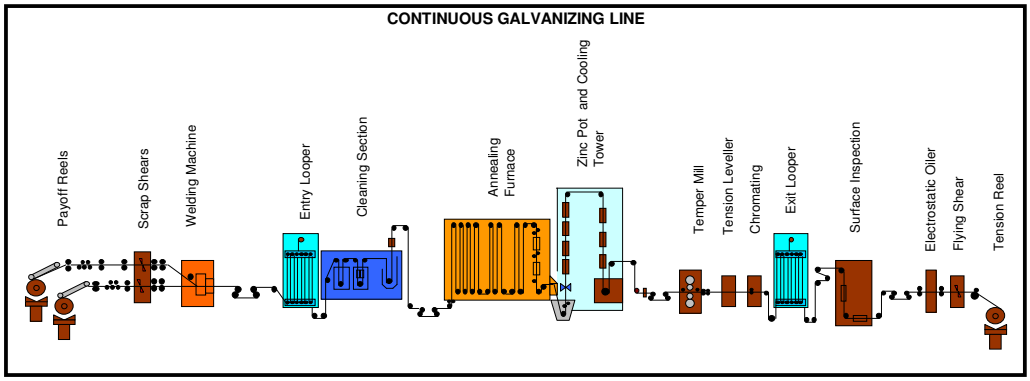


Figure 4.5:ERDEMIR Continuous Galvanizing Line (CGL) [75].

The line was planned to produce high surface quality sheets for the automotive industry. Further, the CGL was designed to be highly flexible in terms of production sequences as well as metallurgy. And environmental aspects are very important. In the Radian Tube Heating section a strip is heated up to a predetermined temperature, and the steel casing is lined with fire-resisting insulation material.

This section is a vertical type furnace with a self supportive gas tight construction reinforced with section steel support beams and back stay, adopting heating equipment by radiant tubes. Natural gas is used for the main and pilot burners. Combustion air is pushed into a recuperator installed at the lower part of the radiant tube, which is used after preheating by combusted exhaust gas. Combusted exhaust gas is sucked into the plenum chamber by an exhaust gas blower and then diffused outside.

The Soaking section holds a strip at the predetermined temperature for the required time after the strip was heated up in the radiant tube heating section. Unlike the radiant tube heating section, the soaking section does not require a heating equipment to heat up the strip. An electrical tube heater is installed to compensate heat loss from furnace walls and retain the strip at the wanted temperature. Tube heater and controller are divided into two zones to control the temperature in each zone.

The heating equipment is an electric heating device in form of a tube heater and controller installed around the furnace walls, dividing the heaters in two zones.

The Slow cooling section slowly cools a strip coming from the soaking section to a predetermined temperature using gas jet coolers. Atmospheric gas is cooled by the jet coolers and then sprayed on the strip through nozzles to cool the strip. After cooling the atmospheric gas is sucked again into the heat exchanger and cooled. To ensure a wanted strip temperature at the delivery section with a minimum loss of

temperature, tube heaters are installed as heating equipment in the soaking section. Besides, bottom and top rolls are made to be roll chambers, while the tube heaters are installed to control the chambers' temperature.

The jet cooler equipment takes advantage of the circulation of atmospheric gas. A wind box is placed along the strip pass line and fixed on the sidewalls of the furnace. A circulation gas blower, heat exchanger and a duct are fixed on the deck, and the duct is connected to the furnace body to be free from the thermal expansion of the furnace body. Manuel dampers are installed in the circulation duct in order to adjust the flow rate of the atmosphere cooling gas supplied into both parts of the wind box [76].

The Rapid cooling section cools a strip to the predetermined zinc coating temperature. This section uses a high cooling rate gas jet cooler and the wind box method. Besides, a motor damper and a manual damper are installed as circulation ducts. The cooling equipment consists of 2 units and is equipped with a shift and support roll to stabilize the pass.

Each shifting and support roll is able to shift at the time of threading. A circulation gas blower, heat exchanger and a duct are fixed on the deck, and the duct is connected to the furnace body to be free from the thermal expansion of the furnace body.

For the Zinc Pot, the 220 tons zinc pot of the line was designed to limit variations of bath level, temperature and composition. Due to the galvanic production the width and length of the pot is not so high. Two inductors, 500 kW each, are attached to the pot. Each inductor can be separately controlled. The bath level is measured by a laser system with very small tolerances. A bath analysis is continuously done. The wiping system is used to adjust the coating thickness. Both air and nitrogen is implemented. To switch between air and nitrogen is possible during a line run. In the Pot Cooling section, a 46 m high tower can be found. In it, 6 air jet coolers are installed that can be regulated individually or automatically via temperature controls. The final cooling is done in a water quench followed by squeezing and drying.

In the Skin Pass Mill and Tension Leveler section, 4 high skin pass mills are used with a 940 mm diameter backup roll and a 500 mm diameter work roll. The work roll can be semi-automatically changed during line run.

The work rolls are textured with Electrical Discharge Texturing (EDT). For an efficient cleaning of backup and work rolls, a high pressure cleaning system is applied. Both the high pressure cleaning system and wet tempering system use de-mineralized water. All water is collected in dumps.

Subsequently to the temper mill, two pairs of wringer rolls are installed to limit the amount of moisture on the strip surface. For final drying of the strip surface, a dryer is used after the chromating section.

The maximum roll force within the roll gap is 700 tons, and maximum bending forces are 30 tons. Tension within the mill area is up to 30 tons applied by four bridle rolls at the entry side of the mill.

The intermediate bridle is also equipped with 2 rolls, which allow a proper tension decoupling between the skin pass mill and tension leveling area. Tension measuring units are placed at the mill's entry and exit side. A tension leveler is located at the bridle rolls at the entry side.

The control system of the mill allows strip processing with constant roll force or elongation control via roll force or tension.

In the Chromating section, the subsequent chromatizing process provides temporary corrosion protection. The liquid is applied to a strip surface via spreader bars and pressed by wringer rolls.

A steam-heated dryer carries out drying of the surface. The whole post treatment section is designed to ensure the most efficient use of chemicals and to prevent environmental pollution. Therefore, the entire enclosures as well as the chemical tank are equipped with a suction system [72].

5. EXPERIMENTAL PROCEDURE

The aim of this study was to improve the deep drawability property (anisotropy coefficient) by optimizing alloying elements and process parameters of cold rolled annealed and galvanized IF steels used for the production of automotive steel sheets.

5.1 Materials/ Alloy Design

This experimental study consists of two stages. In the first stage, three different chemical compositions were examined: Those steels were identified named as Titanium-Stabilized IF Steel, Titanium-Niobium Stabilized IF Steel for Continuous Annealing Line (CAL) process, and Titanium-Niobium Stabilized IF Steel for Continuous Galvanising Line (CGL) process.

By using the first stage study results, new grades with modified compositions were then produced and investigated.

The chemical compositions of examined steels were determined according to excess Ti and excess Nb amounts and the carbon equivalence (CE) equation given below.Excess Ti and Excess Nb are the most effective factors in interstitial free steels. [77].

$$CE = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Si + Ni + Cu)}{15} \quad \textbf{(5.1)}$$

Excess Ti = Ti-(4C+3.42 N+1.5S) **(5.2)**

Excess Nb = Nb-7.75C **(5.3)**

The nominal compositions of six different IF steels utilized in this work are given in Table 5.1. In the Table 5.1.also Erdemir grades are given.and they were named according to production route.

Three of them were produced at Continuous Annealing Line and last three of them were produced at Continuous Galvanising Line.

Table 5.1:Chemical compositions of the steels used in this study.

Process	CAL			CGL		
Type of Steel	**Ti-IF Steel**	**Ti-Nb IF Steel**	**Modified Ti-Nb IF Steel**	**Galvanized Ti-Nb IF Steel**	**Galvanized and Modified Ti-Nb IF Steel**	**Galvanized Low N content Ti-Nb IF Steel**
Steel Code	**CAL1**	**CAL2**	**CAL3**	**CGL1**	**CGL2**	**CGL3**
Erdemir Grade	**7118**	**7318**	**7316**	**1315**	**1315**	**1315**
C %	0.003	0.003	0.0027	0.003	0.003	0.003
Mn %	0.098	0.07	0.0857	0.107	0.125	0.153
P %	0.007	0.005	0.0063	0.006	0.008	0.009
S %	0.007	0.005	0.0043	0.006	0.007	0.007
Si %	0.004	0.005	0.0044	0.003	0.004	0.005
Al %	0.037	0.034	0.0391	0.037	0.043	0.04
Cu %	0.012	0.012	0.0204	0.013	0.021	0.018
Cr %	0.009	0.01	0.0094	0.014	0.016	0.01
Ni %	0.014	0.016	0.0213	0.025	0.016	0.007
Mo %	0.005	0.002	0.0011	0.001	0.002	0.003
V %	0.003	0.001	0.0008	0.001	0.001	0.002
Nb %	-	0.009	0.0134	0.008	0.014	0.006
Ti %	0.067	0.029	0.0393	0.045	0.037	0.042
N ppm	37	39	38	41	41	22
O ppm	25	23	41	33	39	0
CE %	0.025	0.019	0.022	0.027	0.03	0.034
Excess Ti %	0.032	-0.0038	0.0090	0.0099	0.0004	0.0148
Excess Nb %	-	-0.0142	-0.0075	-0.0152	-0.0092	-0.0172

5.2 Material Processing

5.2.1 Continuous Casting Study

Continuous casting at the ERDEMIR plant produced the slabs except Low N Content Galvanized Ti-Nb IF Steel Grade (CGL3). Low N Content Galvanized Ti-Nb IF steel (CGL3) was imported as slab in the thickness of 200 mm. At Erdemir plant, heats with targeted compositions were produced in the steelmaking shop, and 27 castings were done at the Continuous Casting Plant (Table 5.2). The slabs were 200 mm in thickness, 1200 mm in width and 12000 mm in length, and a total of 65 slabs were produced for the experiments.

Table 5.2:Number of slabs and casting produced at ERDEMIR plants.

	Ti-IF Steel	**Ti-Nb IF Steel**	**Modified Ti-Nb IF Steel**	**Galvanized Ti-Nb IF Steel**	**Modified and Galvanized Ti-Nb IF Steel**
Steel Code	**CAL 1**	**CAL 2**	**CAL 3**	**CGL 1**	**CGL 2**
Number of Slabs	12	16	12	16	9
Number of Casting	5	7	5	7	3

5.2.2 Hot Rolling Study

Hot rolling of the continuous cast slabs was made at the second hot rolling mill of ERDEMIR. The process parameters of the mill are given in Table 5.3.

Table 5.3:The hot rolling mill process parameters.

Hot Rolling Temp. Design	**CAL**			**CGL**		
	Ti-IF	Ti-Nb IF Steel	Modified Ti-Nb IF Steel	Galvanized Ti-Nb IF Steel	Galvanized and Modified Ti-Nb IF Steel	Galvanized Low N content Ti-Nb IF Steel
Steel Code	CAL 1	CAL 2	CAL 3	CGL 1	CGL 2	CGL 3
Slab Reheating Temperature, ℃	1200 ±10	1200 ±10	1200 ±10	1200 ±10	1200 ±10	1200 ±10
Finishing Temperature, ℃	940 ±10	940 ±10	940 ±10	940 ±10	940 ±10	940 ±10
Coiling Temperature, ℃	700 ±10	720 ±10	720 ±10	730 ±10	730 ±10	730 ±10

After reheating the slabs in the slab furnace to a temperature of about 1200 ˚C in 153 minutes, they are sent through the Roughing Mill at a temperature of 1185 ˚C for the first time to remove surface scales.

This is repeated 5-7 times until the slab temperature cools down to 1100 ˚C. To reach the goal of getting the right temperature, the slabs are kept outside of the Finishing Mill. Then they are rolled down to the final thicknesses from 4.50 mm to 2.70 mm in this mill as hot rolled coils at a temperature of about 940 ˚C and after this, their temperatures drop down to 700 ˚C - 730 ˚C on the Cooling Table. Slabs those are hot rolled to various thicknesses are listed in Table 5.4.

Ohashi et.al explained [38] that slab reheating temperatures of nearly 1250 ˚C may almost completely dissolve carbide and carbonitride precipitates while others such as TiN and TiS are relatively stable. At a slab reheat temperature of 1000 ˚C, all the finest precipitates remain [38].

5.2.3 Cold Rolling

The hot rolled coils are then transferred to Cold Mill No 2, and cold rolled to various thicknesses given at Table 5.4. In the scope of this study, the trial procedure of the trial coils was prepared according to the ERDEMIR Cold Mill processing conditions. Firstly, the trial coils were processed at the Continuous Pickling and Tandem Cold Mill.

After tension leveler applied for breaking the oxides on the coil surface before the pickling section, the oxide layers were removed by pickling in HCl acid processing tanks and then strip is side trimmed according to customer requirements. After side trimming strip is cold rolled down to various thicknesses by giving 47-82 % reduction in thickness at the tandem cold mill section.

After completing the Continuous Pickling and Tandem Cold Mill lines, some of the coils were sent to Continuous Annealing Line for annealing process and the other part of the coils were galvanized at the CGL.

All of the process parameters like the cold deformation ratio or process temperatures were taken from the process computers. Exit thicknesses of the coils after cold rolling and the corresponding cold deformation ratios are shown in Table 5.4.

Table 5.4:Cold rolling deformation ratios and exit thicknesses.

HotRolling Exit Thickness (mm)	Cold Rolling Exit Thickness (mm)						Cold Deformation Ratio (%)
	Ti-IF Steel	Ti-Nb IF Steel	Modified Ti-Nb IF Steel	Galvanized Ti-Nb IF Steel	Modified Galvanized Ti-Nb IF Steel	Low N content Galvanized Ti-Nb IF Steel	
	CAL 1	CAL 2	CAL 3	CGL 1	CGL 2	CGL 3	
4,5	0.8	0.8	0.8	0.8	-	0.8	82
	1.5	1.5	-	1.5	-	-	67
	1.6	-	-	-	-	-	64
	-	-	-	1.7	-	-	62
	1.8	-	-	-	-	-	60
	-	-	-	1.9	-	-	57
	2	2	-	2	2	-	55
4,2	1.5	1.5	-	1.5	1.5	-	64
4	0.7	-	-	-	-	-	82
	0.8	-	0.8	-	-	-	80
	0.9	-	-	-	-	-	77
	1	-	-	-	-	-	75
	1.1	-	-	-	-	-	72
	2	2	-	2	-	-	50
3,6	0.8	0.8	0.8	0.8	-	0.8	77
3,8	1.5	1.5	-	1.5	1.5	-	60
	2	2	-	-	2	-	47
3	0.8	0.8	-	0.8	-	0.8	73
2,8	-	-	-	-	-	0.8	71
2,7	0.8	0.8	-	0.8	-	-	70

5.2.4 Continuous Annealing

Cold rolled Ti IF (CAL 1), Ti-Nb IF (CAL 2) and Modified Ti-Nb IF (CAL 3) steels coils were annealed at 830ºC -850ºC and 870ºC in the Continuous Annealing Line. For each deformation ratio, 6 coils were annealed. The annealing process parameters are shown in Table 5.5.

Table 5.5:Continuous Annealing Process Parameters.

Annealing Temperatures	830-850-870 ˚C
Annealing Times	50 secs. to 220 secs.

The heat treatment profiles of the examined steels in the CAL are shown in Figure 5.1

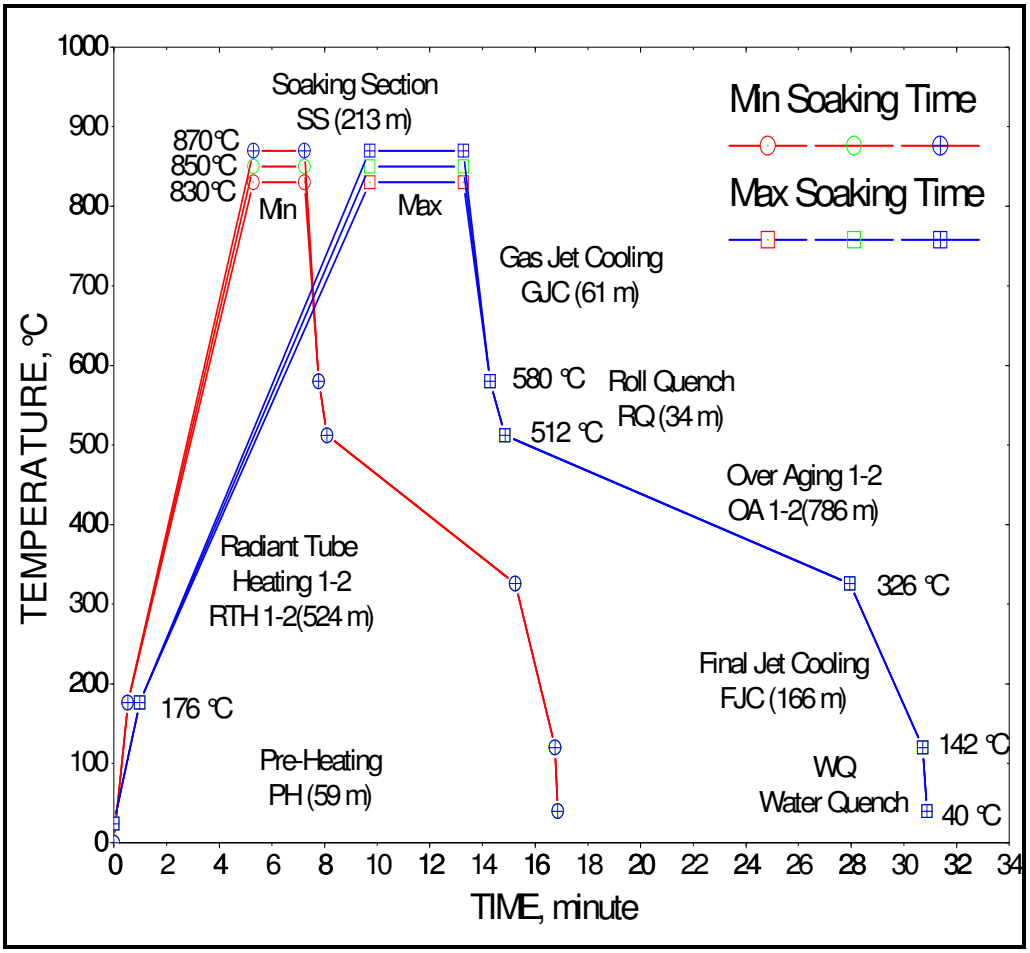


Figure 5.1:The annealing cycles of the steel samples in the Cont. Annealing Line.

5.2.5 Continuous Galvanizing

Cold rolled Ti-Nb IF (CGL 1), Modified Ti-Nb IF (CGL 2) and Low N Content Ti-Nb IF (CGL 3) steels were processed at the Continuous Galvanizing Line. Firstly, the coils were electronically cleaned by the Alkali treatment process and annealed at 830-850 and 870 ºC temperatures.

Then the hot dip process passing through a 220 ton zinc pot at 460˚C galvanized the coils. The heat cycle process parameters are shown in Figure 5.2.

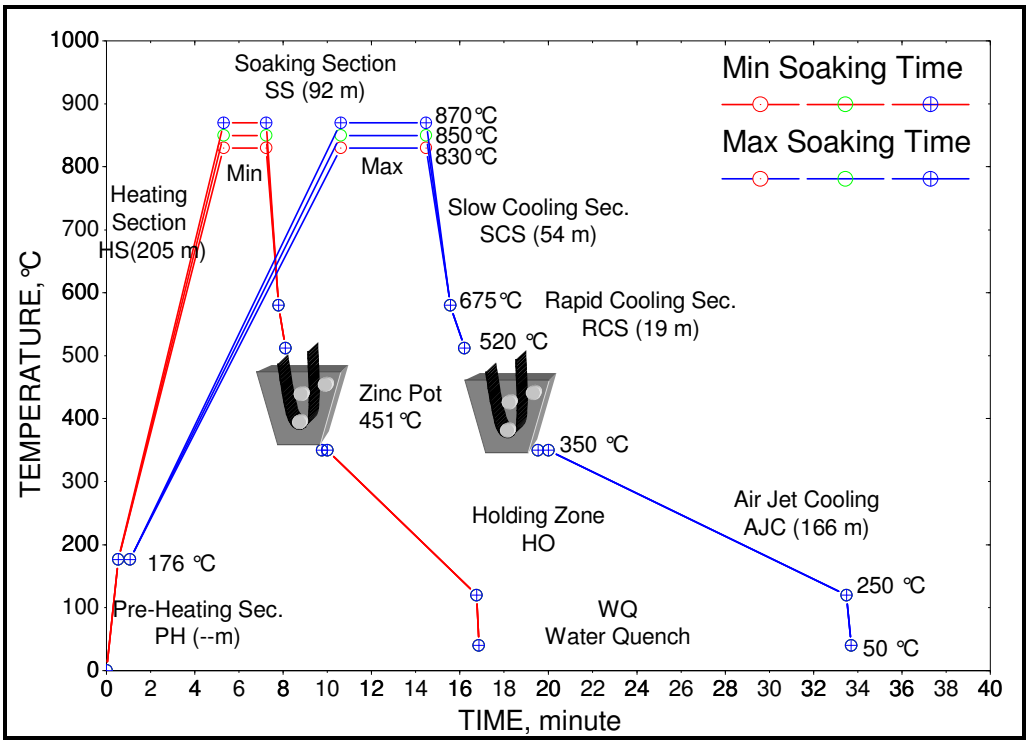


Figure 5.2:The annealing cycles of the steel samples at the Cont. Galvanizing Line

5.3 Experimental Techniques

5.3.1 Tensile Tests

Yield strength, tensile strength, and elongation at fracture, strain hardening exponent and anisotropy coefficients were the main mechanical properties determined by tensile tests.

Tensile specimens were cut along rolling direction (0°), transverse direction (90°) and at 45° to the rolling direction. Samples were taken from the middle section of each coil.

Tests were conducted on a Zwick Z-100 testing machine. All tests were made at a strain ratio of 2% Lo/min. The dimensions of the tensile specimens are shown in Figure 5.3, which were processed according to DIN EN 10002-1 Type 2 standards.

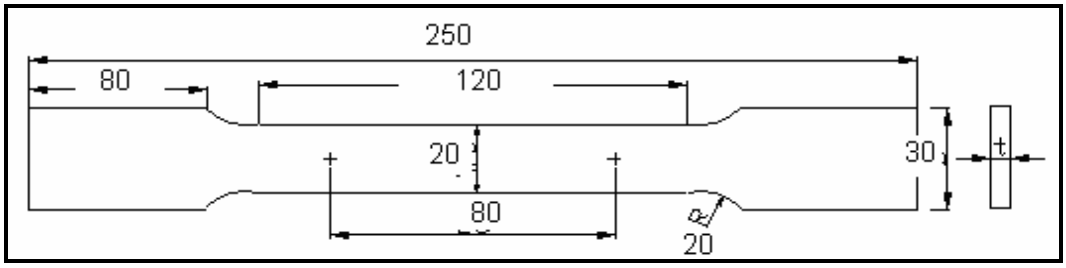


Figure 5.3:Dimension of the tensile specimens.

For each direction (0°-45°and 90°), anisotropy coefficients (r) were calculated for strains of 6,8,10,12,14,16 % according to the following ;

$$r = \frac{\ln \dfrac{bo}{bo - \Delta b}}{\ln \dfrac{(\mathrm{Lo} + \Delta \mathrm{L}) * (\mathrm{bo} - \Delta \mathrm{b})}{\mathrm{Lo} * \mathrm{bo}}} \qquad \textbf{(5.4)}$$

where

r : 0°- 45°- 90° anisotropy

b_0 : Initial specimen width

Δb : Reduction in width

L_o : Initial gauge length

ΔL : Extension

Averages of mechanical properties (including anisotropy coefficients) were determined as [77]

$$X_{average} = \frac{[X_0 + (2xX_{45}) + X_{90}]}{4} \qquad \textbf{(5.5)}$$

where

X_0 : Average mechanical property for direction of 0 °,

X_{45} : Average mechanical property in direction of 45 °, and

X_{90} : Average mechanical property in direction of 90 °.

5.3.2 Optical Microscopy

Optical microscopy was used for the microstructural examinations and grain size measurements of the samples. Microstructural examinations were conducted on the cross sections perpendicular to the rolling direction. The samples were grounded with rough, moderate and fine grinding discs and polished using 9–6–3–1 μm water based diamond suspension. The polished samples were etched twice using 3 % nital solution for 15 seconds.

Microstructures of materials (ferrite grain size, carbide type and distributions) were investigated with a LECO 500 optical microscope, Clemex Quantitative Image Analysis System and Clemex Vision Software Ver.3.

5.3.3 Scanning Electron Microscopy (SEM) Analysis

Scanning Electron Microscopy (SEM) was used for the determination of the microstructure and carbide distribution of the samples. The Scanning Electron Microscope was a JEOL 5600 JSM with an attached EDS system.

5.3.4 Texture Analysis

In this study, texture analysis was conducted with a RIGAKU D-Max Ultima X-Ray Diffractometer with a pole figure attachment. The first step of the pole figure analysis was a sample preparation study. Test samples were prepared from annealed samples, punched to a 4 mm diameter and then metallographically prepared. The next step was data collection from an X-Ray diffractometer. The XRD measurements of samples were carried out in the range of 20-90 2θ degrees with 0.02 steps. Then pole figure measurements were done with different α and β angles in {110}, {111} and {200} diffraction lines.

5.3.5 Continuous Annealing Simulation Studies

In order to determine the effect of the annealing temperature and soaking time on the grain size of the samples, the simulation experiment was performed with the Gleeble-3500 Thermal-Mechanical Test and Simulation Machine at the ERDEMIR Research and Development Simulation Laboratory (Figure 5.4).

Samples taken in the rolling direction from Ti IF and Ti-Nb IF steel sheets (Figure 5.5) were prepared for the simulation study.

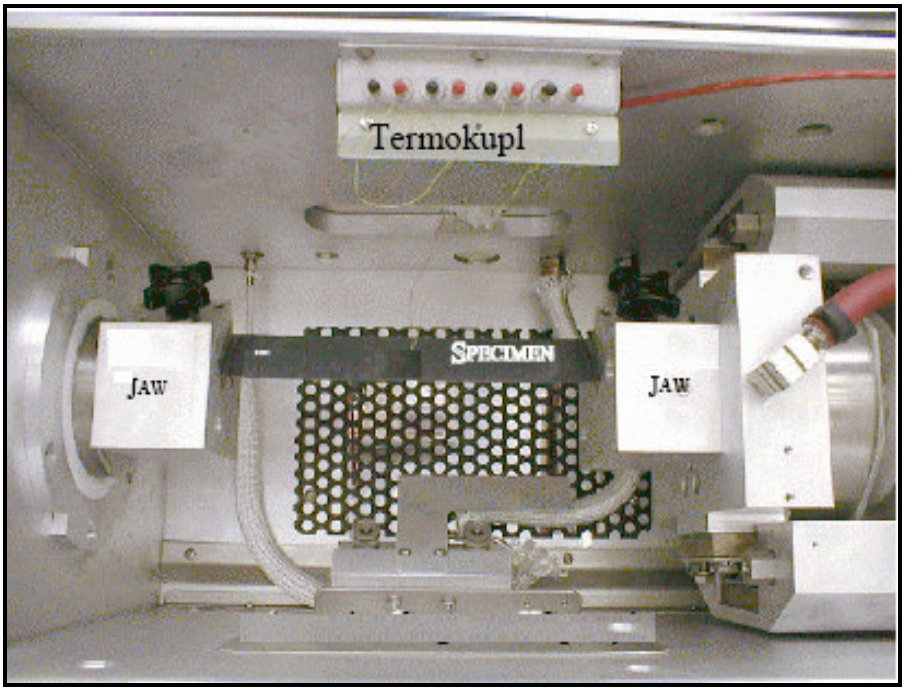


Figure 5.4:Gleeble-3500 tool cabinets.

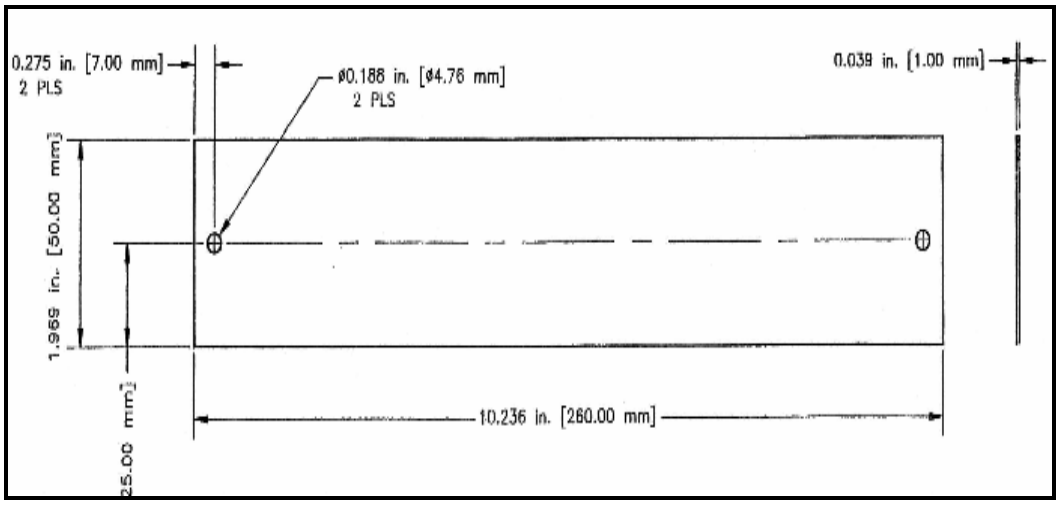


Figure 5.5:Sample used for the Gleeble-3500 tool cabinet.

In order to prevent oxidation of the samples, the Gleeble-3500 device was vacuumed up to 2 torr, and Argon gas was given to the system and then vacuumed to 2 x 10^{-5} torr with a diffusion pump.

The annealing cycle of the simulation experiment is shown in Figure 5.6. The soaking times were 50, 100 and 150 sec. respectively and the soaking temperature was varied from 830 to 870 ˚C.

The samples were heated at 4 ˚C / second from 400 to 870 ˚C and then cooled down to room temperature by a water quench system.

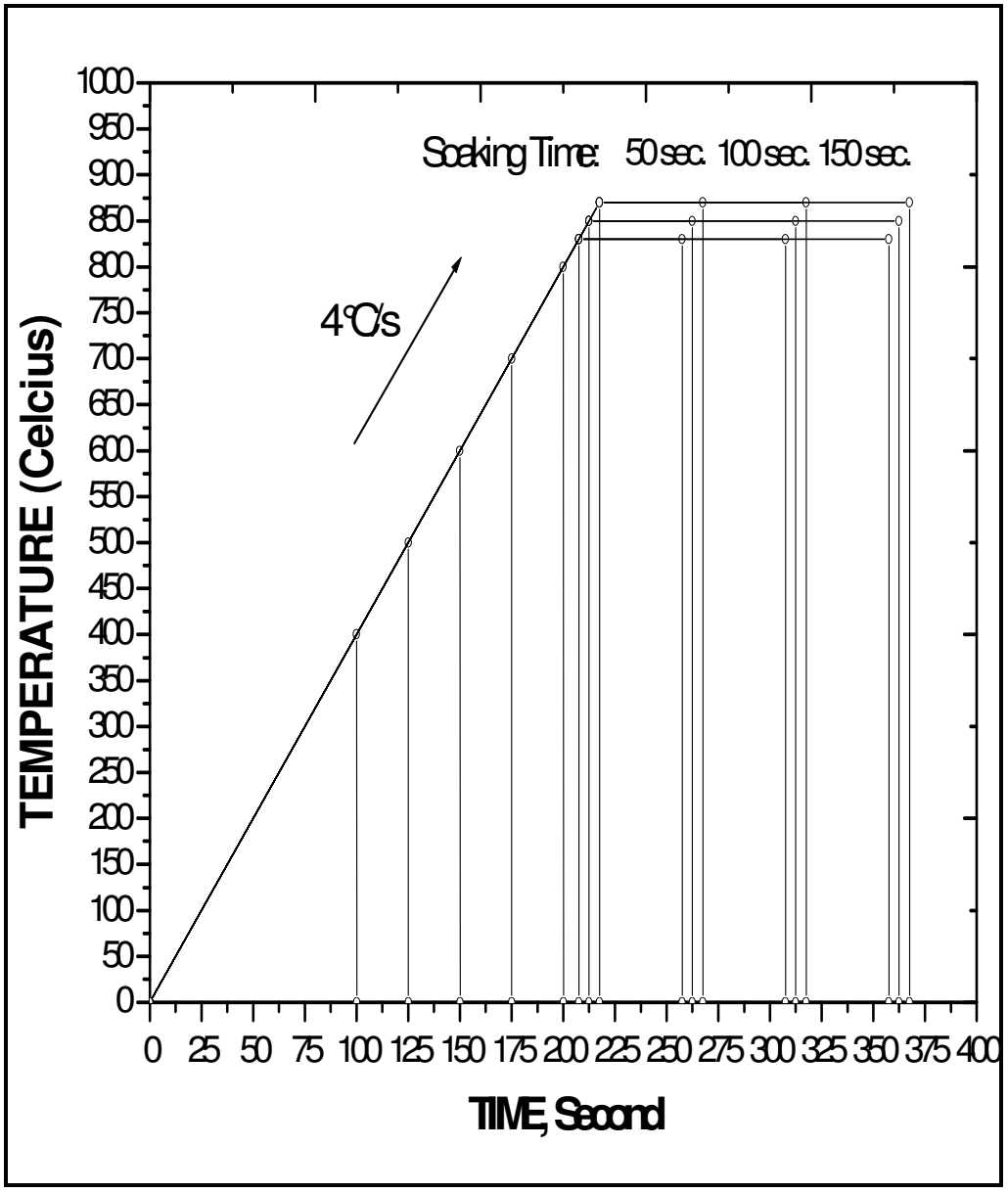


Figure 5.6:Temperature cycle of the simulation experiment.

The samples used in the simulation experiment were metallographically prepared and then examined by microscope.

Microstructures of the samples and grain sizes were examined with the LECO-500 type optical microscope combined with an image analysis system.

6. EXPERIMENTAL TEST RESULTS

6.1 The effect of Process Parameters on Mechanical Properties

In this section, the effects of soaking temperatures and cold deformation ratios on the mechanical properties of CAL and CGL steels are presented. Slabs are rolled up to 4.5 mm thickness in the hot rolling process and then cold rolled down to various thicknesses. Cold rolling was applied to those steels with a reduction ratio between 47-82 % and then they were continuously annealed at three different temperature ranges at CAL and CGL (820-839 ˚C, 840-859 ˚C, 860-880 ˚C). The graphs to show the effects of processing parameters on the mechanical properties of cold rolled steel sheets annealed at CAL and CGL are given in Appendix A and B.

Tensile test results for maximum drawability obtained Ti IF (CAL 1), Ti-Nb IF (CAL 2) and modified Ti-Nb IF (CAL 3) steels are given in Table 6.1 Firstly, the grades 7118 (CAL 1), 7318 (CAL 2) and 7316 (CAL 3) which were cold rolled and then annealed at CAL (given in Appendix A and B) were investigated. It became obvious that at a temperature range of 820 – 870˚C, the yield strength, the ratio between yield strength and tensile strength, and the elongation values decrease, whereas the average anisotropy coefficient (r value) and the tensile strength increase with increasing cold deformation ratio.

Fukuda [18] explained the effect of cold rolling reduction on the development of r values. The high r values are associated with the formation of beneficial re-crystallization textures that are produced after hot deformation, coiling, cold rolling and annealing. A critical fundamental observation in this processing sequence is the fact that carbon atoms develop in interstitial solid solution degree beneficial for {111} re-crystallization textures and unfavorable for {110} and {100} components.

For the Ti IF steel (the grade 7118-CAL 1), the anisotropy coefficient increases with increasing cold deformation ratio. For a 55 % cold deformation ratio the r value was 1.8, and with increasing ratios from 55 % to 80 % and increasing annealing temperatures from 850 ˚C to 870 ˚C, the r value raised to 2.7. The r value of Ti-IF Steel (grade 7118-CAL 1) was found to be between 2.20 - 2.70. It has the highest deep drawability value among the investigated grades in this part of study. Osawa and his friends [57] found that increased amounts of cold reduction result in lower elongations and n values, higher yield and tensile strengths, and higher r values at cold reductions ranging from 65 % to 80 % for continuously annealed Ti-IF steel (CAL 1) which is in agreement with the result of this study.

The highest r value of Ti-IF steel (the grade 7118-CAL 1) was obtained for a deformation ratio higher than 73 % and at an annealing temperature of 870 ˚C with

no variation in mechanical properties along the whole coil. Anisotropy coefficients or deep drawability factors seem to be increasing with an increasing annealing temperature and deformation ratio.

The carbon equivalent value was calculated to be between 0.018–0.031 %, and excess titanium was between 0.033 – 0.042 percent. It was also found that the nitrogen content was between 32 – 39 ppm in Ti IF steel (CAL 1), and the strain hardening coefficient was between 0.22–0.24. The other mechanical properties such as yield strength, tensile strength and elongation values do not show any systematic variation with respect to the annealing temperature.

In the Ti-Nb IF steel (the grade 7318-CAL 2), the highest r value was obtained at a cold deformation ratio higher than 73 %, and the lowest r value was obtained at an annealing temperature of 830 ˚C. The strain hardening coefficient was found to be 0.23 and here as well, no systematic variation in mechanical properties for different annealing temperatures was detected.

With modified Ti-Nb IF steel (the grade 7316-CAL 3), the r value was between 2.3 – 3.02, which is a relatively higher r value than for the Ti-Nb IF steel (grade 7318-CAL 2). Modified Ti-Nb IF steel (the grade 7316-CAL 3) could be classified as extra deep drawing quality. The r value around 3.0 was obtained with the highest annealing temperature of 870 ˚C and the highest cold deformation ratio of 80%. Tokunada and Yamada observed [21] that a maximum cold reduction of approximately 80 % results in the best r value in IF steels.

With modified Ti-Nb IF steel (the grade 7316-CAL 3), the carbon equivalent percentage was between 0.022 – 0.024 and the excess titanium between 0.006 – 0.009. The difference between Ti-Nb IF and modified Ti-Nb IF steel was only slightly higher for an increased carbon equivalent and excess niobium, but had a much higher increase in excess titanium.

The mechanical properties of modified Ti-Nb IF steel (the grade 7316-CAL 3) were nearly the same as the ones of the Ti-Nb IF steel (the grade 7318-CAL 2), except the elongation. The tensile test results of Ti-IF Steel, Ti-Nb IF Steel, and modified Ti-Nb IF steels are shown in Table 6.1.

Table 6.1:Tensile test results for maximum deep drawability obtained in Ti IF, Ti-Nb IF and modified Ti-Nb IF steels.

Material	**Max. r Value**	**Mechanical properties corresponding to max. r values**	**Metallurgical Parameters corresponding to max r value obtained**

		Yield Strength (MPa)	Tensile Strength (MPa)	Elongation at Fracture %	σ_T/σ_Y	n	Annealing Temp. (˚C)	Def. Ratio %
7118 Ti IF Steel	2.71	156	271	43.0	0.57	0.23	872	73
7318 Ti-Nb IF Steel	2.66	159	286	48.0	0.55	0.23	853	73
7316 Modified Ti-Nb IF Steel	3.02	164	288	48.4	0.56	0.22	870	80

The effect of the deformation ratio and annealing temperature on the mechanical properties of hot dip galvanized IF steels are shown in Appendix B in Figure B.1-3.

The mechanical properties of galvanized Ti-Nb IF (CGL 1) steel are shown in Appendix B in Figure B.1, which were annealed at 830 ˚C and hot dip galvanized. It was obvious that the r and n values increased with an increasing cold deformation ratio, but no important differences in elongation, yield and, tensile strengths or yield/tensile strength ratio were observed.

The lowest yield strength and highest r value were obtained for low N content galvanized Ti-Nb IF steel (CGL 3) with a 77 % cold deformation ratio and an annealing temperature of 845 ℃. Galvanized Ti-Nb IF steel (CGL 1) had the worst mechanical properties at an annealing temperature of 830 ˚C. Carabajar[1] reported that with increasing nitrogen content; the r value deteriorates while the re-crystallization temperature stays constant which agrees with this result of the present study.

For galvanized (the grade 1315-CGL 1), modified galvanized (the grade 1315-CGL 2), and low N content galvanized (1315-CGL 3) Ti-Nb IF steel samples annealed at 840 - 860℃, the yield strength and yield/tensile strength ratio decrease, whereas the tensile strength and r value increase with higher deformation ratios. In contrast to galvanized Ti-Nb IF steel (CGL 1), low nitrogen galvanized Ti-Nb IF steel (CGL 3) has the best formability properties. In addition, r values for all samples increase with increasing annealing temperature.

Calculations for galvanised Ti-Nb IF steel (the grade 1315-CGL 1) show that the carbon equivalent value was between 0.025 – 0.034 %, excess titanium between 0.007–0.016 %, and excess niobium-0.014 %. The nitrogen content was found to be between 34– 41 ppm.

For all samples, which were annealed between 860 - 880℃ and then hot dip galvanized, by increasing the deformation ratio, the yield strength and yield/tensile ratio while the elongation increases. But there was also a slight increase in the r value and tensile strength. Low N content Ti-Nb IF steel (the galvanized grade 1315-CGL 3) has the highest formability value, and galvanized Ti-Nb IF steel (the grade 1315-CGL 1) has the worst mechanical properties at 865 ℃. However, galvanised Ti-Nb IF steel had the lowest yield strength and the highest r value with a 67 % cold deformation ratio.

The effect of processing variables of galvanised Ti-Nb IF steel (the grade 1315-CGL 1), Modified Galvanised Ti-Nb IF steel (the grade 1315-CGL 2) and galvanised Low N Ti-Nb IF steel (the grade 1315-CGL 3) samples on the total elongation, r and n values were changing due to the chemical composition and zinc coating layers.

Variations in mechanical properties of coated samples might happen because of the interaction between coating layers and base metal.

Calculations for galvanized Modified Ti-Nb IF steel (the grade 1315-CGL 2) show that the carbon equivalent value was between 0.028 – 0.035 %, excess titanium between 0 – 0.002 %, and excess niobium between (-0.006) – (-0.009) %. The nitrogen content was between 37 – 40 ppm.

In table 6.2, the process parameters and mechanical properties are listed. Hutchinson[28] observed that high average r and elongation values can be obtained at the same time when 0.005 to 0.010 % Nb and Ti are co-added. As a result, Nb plays an important role in improving press formability in Ti steel.

It has been reported both at Inland Steel Company [47] and Tanaka and friends [60] that the galvanizing reduced ~0.2 % of the r value and ~1% of the total elongation in comparison to the uncoated product.

The tensile test results of galvanized steels are shown Table 6.2.

Table 6.2:Tensile test results for maximum drawability obtained in galvanized 1315 Ti-Nb IF, galvanized 1315 modified Ti Nb IF, galvanized 1315 low N content Ti-Nb IF steels.

Material	Max. r Value	Mechanical properties corresponding to max. r values					Metallurgical Parameters corresponding to r value obtained	
		Yield Strength	Tensile Strength	Elongation at Fracture	σ_Y/σ_T	n	Annealing Temp.	Def. Ratio

		(MPa)	(MPa)	%			(ºC)	(%)
1315 Ti Nb IF	2.2	159	294	44.0	0.54	0.23	865	77
1315 Modified Ti Nb IF	1.7	168	288	46.2	0.58	0.22	845	67
1315 Low N Content Ti-Nb IF	2.4	162	299	48.1	0.54	0.23	852	77

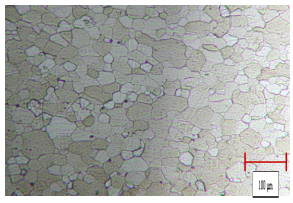
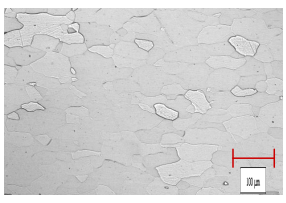
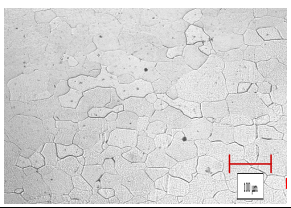
6.2 Effect of Process Parameters on Microstructure

The microstructure study was done with LECO 500 optical microscope for CAL and CGL steel samples. In general, the microstructures show typical ferrite structures with different grain morphologies. It can be observed that the grain size of ferrite increases with increasing annealing temperatures which is an expected result.

The effect of cold deformation and annealing temperature on microstructure of Ti IF (CAL 1), Ti-Nb IF (CAL 2), and Modified Ti-Nb IF steel (CAL 3) samples produced in CAL and Galvanized Ti-Nb IF (CGL 1), modified galvanized Ti Nb IF (CGL 2), low N content galvanized Ti-Nb IF steel (CGL 3) samples produced in CGL are shown in Figure 6.1 and Figure 6.2, respectively.

Figure 6.1 and Figure 6.2 present the size and the aspect ratio of the ferrite grains with respect to cold deformation ratio. There is a tendency that the grain size is getting smaller with an increasing deformation ratio at a constant annealing temperature.

Among the modified Ti-Nb IF steel, one of them which was cold rolled 80 % and annealed at 870˚C, had the largest grain size of 44.3 µm, whereas Ti IF steel (grade 7118-CAL1) that was cold rolled 73 % and annealed at 870˚C, had the smallest grain size of 18.5 µm (Table 6.3).

Cold Reduction Ratio	Annealing Temperature		
	820-840 ºC	840-860 ºC	860-880 ºC
Ti IF Steel			
47%			

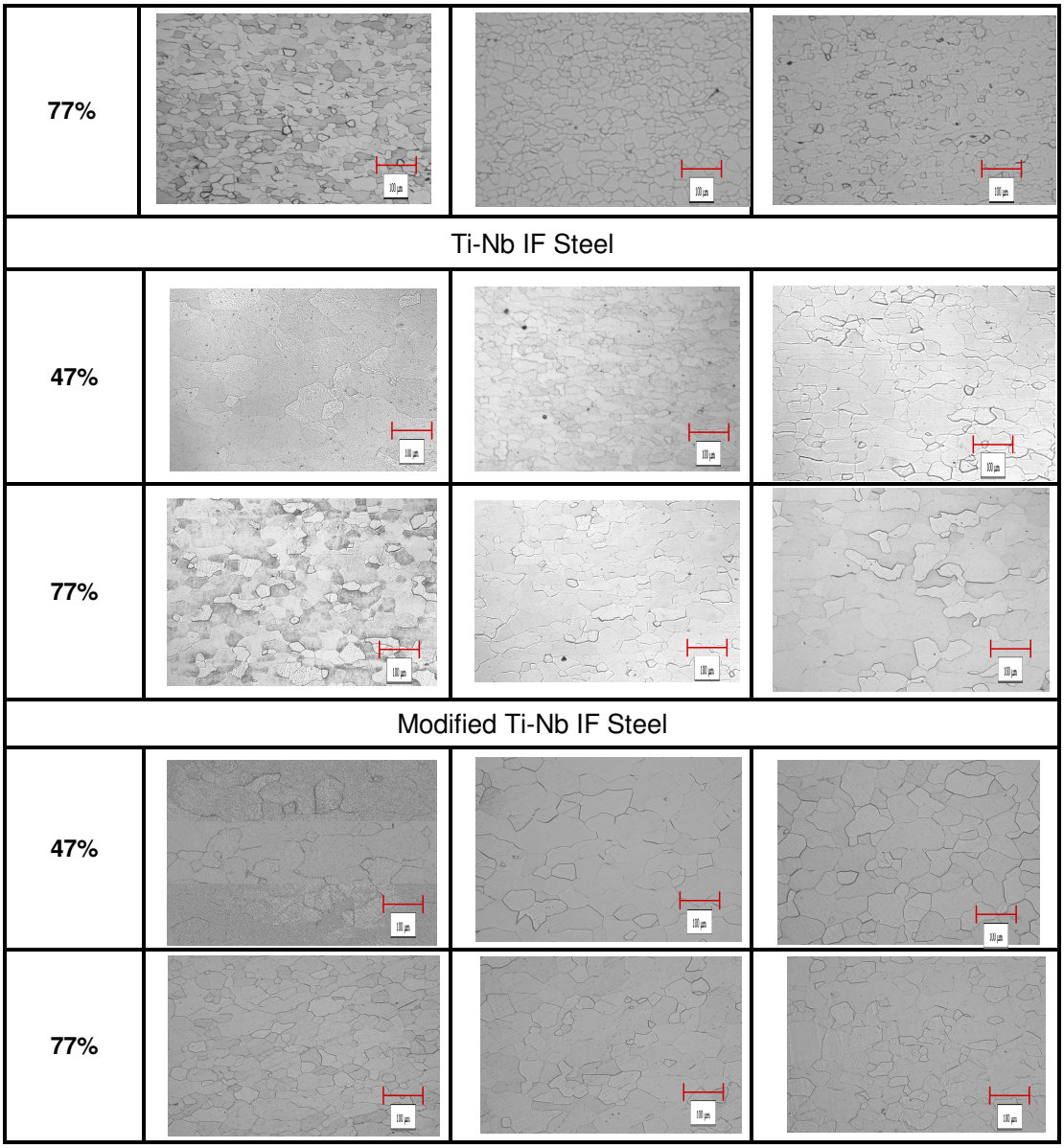
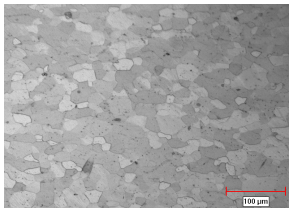
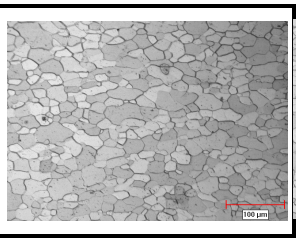
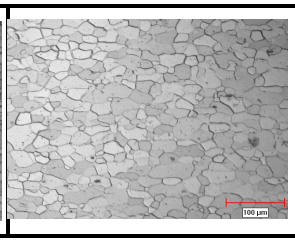


Figure 6.1:Effect of the cold deformation ratio on microstructure of the annealed Ti IF,Ti-Nb IF and modified Ti-Nb IF steel sheets.

	Annealing Temperature		
	820-840 ºC	**840-860 ºC**	**860-880 ºC**
Galvanized Ti-Nb IF Steel			
47%			

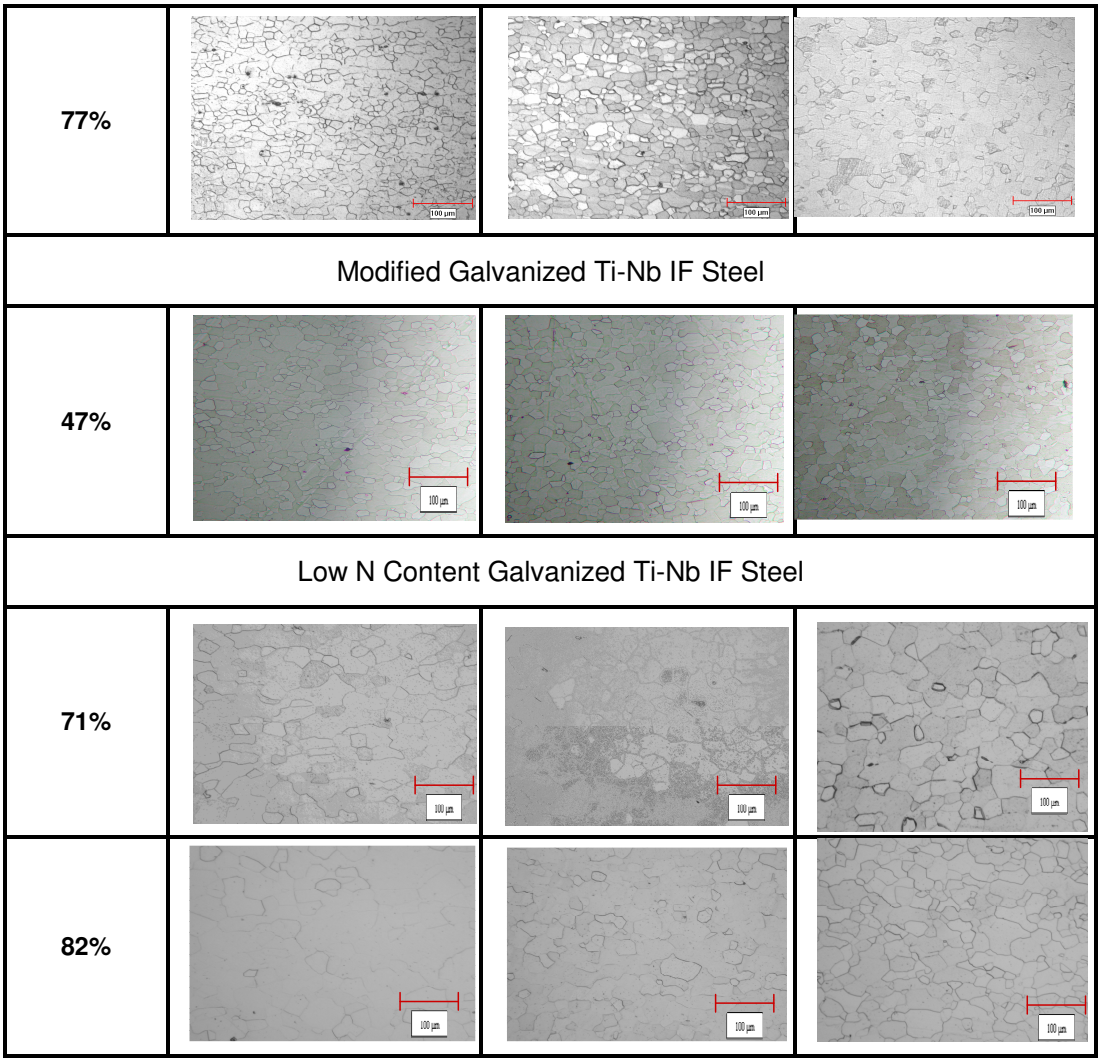


Figure 6.2:Effects of the cold deformation ratio on microstructure of the galvanized steels; galvanized Ti-Nb IF, modified galvanized Ti-Nb IF, low N content galvanized Ti-Nb IF sheets.

Among the samples of low N content galvanized Ti-Nb IF steel (CGL 3) one of them that was cold rolled 77 % and annealed 850 ˚C had the largest grain size of 33.6 µm. Galvanized Ti-Nb IF steel (grade 1315-CGL 1) that was cold rolled 77 % and then annealed at 865 ˚C, had the smallest grain size of 19.4 µm (Table 6.4).

Table 6.3:The median grain sizes with corresponding process parameters which have highest r values in Ti IF, Ti-Nb IF and modified Ti-Nb IF steels.

Material	Max. r Value	Metallurgical Parameters corresponding to max. r Value		Chemical Composition Corresponding to max. r Value			Microstructure	
		Annealing Temp. (℃)	Def. Ratio (%)	CE	Ti* %	Nb* %	d	W/L Aspect Ratio
Ti-IF	2.82	872	73	0.025	0.032	-	18.5	0.80
Ti-Nb IF	2.02	853	73	0.019	-0.038	-0.014	19.6	0.85
Modified Ti-Nb IF	3.02	870	80	0.022	0.009	-0.007	44.3	0.78

Katoh and friends observed strong correlations between growing forming properties (total elongation, n and r values) with a coarsening grain size, and decreasing yield and tensile strength values [16]. The median grain sizes with corresponding process parameters with higher r values are shown in Tables 6.3 and 6.4.These precipitates are mainly Ti (C-N) particles.

EDS analyses taken from these particles indicated that they are Ti (C-N), TiN, (TiNb)C and $Ti_4C_2S_2$ particles, which were precipitated on the ferrite grain boundaries as shown in Figure 6.3.

Hutchinson and also Kırmanoglu reported [17-78] that for Ti added IF steels, the carbon, nitrogen and sulfur elements form precipitates of different types, such as TiN, TiS, $Ti_4C_2S_2$ and TiC [17].

These types of particles originate from Ti and Nb based alloying elements, which were added to the steel to prevent the formation of free carbon as well as to control process parameters, such as the slab reheating temperature, finishing temperature, coiling temperature, cold deformation ratio, annealing temperature and soaking time.

Table 6.4:The median grain sizes with corresponding process parameters which have highest r values in galvanized Ti Nb IF, modified galvanized Ti-Nb IF, low N content galvanized Ti-Nb IF steels.

Material	Max. r Value	Metallurgical Parameters corresponding to max. r Value		Chemical Composition Corresponding to max. r Value			Microstructure	
		Annealing Temp. (℃)	Def. Ratio (%)	CE	Ti* %	Nb* %	d	W/L Aspct Ratio
Galvanized Ti Nb IF	2.2	865	77	0.027	0.099	-0.015	19.4	0.76
Galvanized and modified Ti Nb IF	1.7	845	67	0.030	0.004	-0.009	22.3	0.92
Galvanized low N Content Ti-Nb IF	2.4	852	77	0.034	0.014	-0.017	33.6	0.94

The process parameters of the steels produced in that study investigated on SEM are given in Table 6.5.

Table 6.5:The process parameters of the ULC IF steels produced in the study.

Code	Hot Rolled Thickness (mm)	Cold Rolled Thickness (mm)	Deformation Ratio %	CAL/CGL Soaking Temp. (˚C)	Soaking Time (Sec.)
Ti IF	3.6	0.8	77	854	100
Ti-Nb IF	4.0	2.0	50	830	210
Ti-Nb IF	3.6	0.8	77	827	61
Modified Ti-Nb IF	4.2	1.5	64	841	100

Matsudo and et.al. explained [33] that incrementing of the r value for high temperature annealing is caused by grain growth. Consequently, materials with a small amount of carbon and nitrogen cannot prevent further grain growth for high temperature annealing [33]. In addition, innumerable fine precipitates, which are mainly TiN and TiC, can be observed in these steels as shown in Figure 6.3.

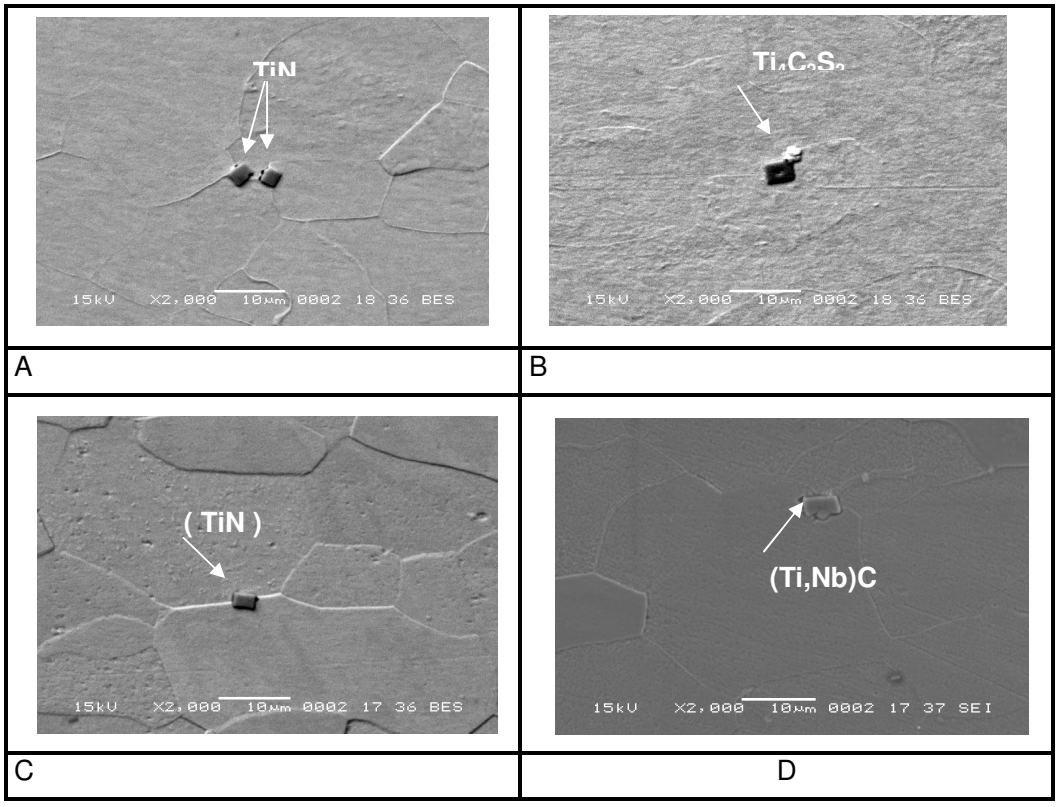


Figure 6.3:SEM micrographs of the selected steels of a) Ti IF b) Ti-Nb IF c) Ti-Nb IF d) modified Ti-Nb IF.

6.3 Texture Analysis Results

In Figure 6.4, X-Ray diffraction (XRD) analysis is given for the sample Modified Ti-Nb IF steel.

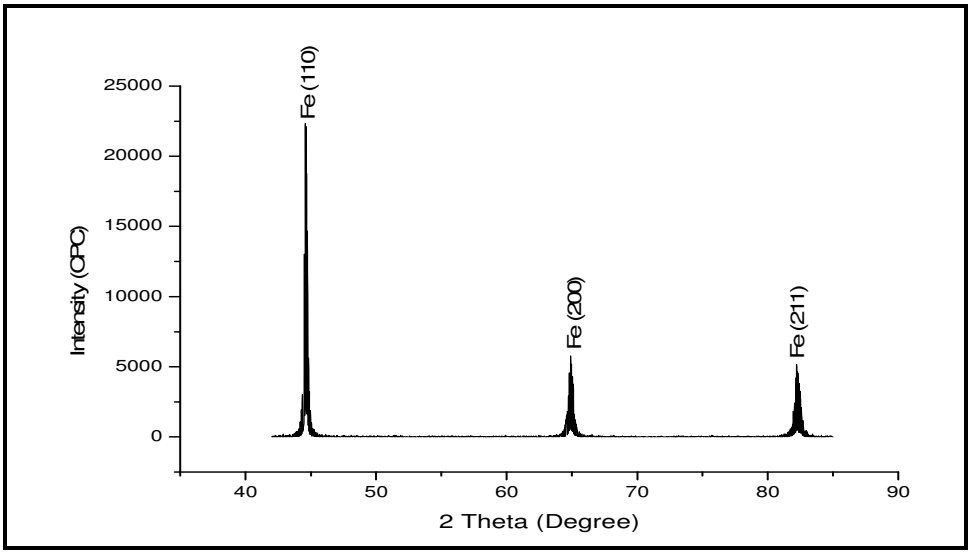


Figure 6.4:XRD analysis of the modified Ti-Nb IF steel (Sample No 5).

It can be seen that the material had three diffraction planes, which were (110) - (200) and (211). Since there were three different diffraction angles on the Modified Ti-Nb IF steel (CAL 3) samples, there was no strong reflection in the plane texture structure of the material. The strongest reflection is seen on the (110) plane, which means the strongest grain orientation on this plane.

Liv, Eloot and his friends explained that in order to get good drawability property, {111} / (110) reflections must be the strongest [79-80].

Strongly textured structure was only found in one crystallographic plane, the Modified Ti-Nb IF steel sample which had stronger orientation in the 110 direction. Similar XRD results were obtained from the other samples.

The process parameters of Modified Ti-Nb IF steel samples (CAL 3) used in texture analysis are given in Table 6.6.

Table 6.6:The process parameters and mech. properties of modified Ti-Nb IF steel

Sample No	Grade	Thickness (mm)	Cold Deformation ratio (%)	Annealing Temp (˚C)	Soaking Time (sec)	Yield Strength (MPa)	Tensile Strength (MPa)	r value	n
1	Modified Ti-Nb IF steel	0.9	77	864	70	163.3	293	2.08	0.23
2	Modified Ti-Nb IF steel	1.0	75	870	70	165.8	289.4	2.30	0.24
3	Modified Ti-Nb IF steel	1.8	60	864	170	179.7	288.9	1.58	0.22
4	Modified Ti-Nb IF steel	2.0	55	866	210	173.5	281.6	1.95	0.22
5	Modified Ti-Nb IF steel	0.8	80	870	110	164	288	3.02	0.22

When analyzing the reflections on the 110 and 200 planes for sample No 1 of Table 6.6 during the pole figure analysis shown in Figure 6.5, it was observed that a stronger color density on the 110 plane together with more red color which means that the material was strongly oriented on that region.

However, crystallographic distributions for all planes existed in the same plane. Pole figure investigation of sample no:5 of Table 6.6 which has r=3 value is shown in Figure 6.6. Figure 6.6 revealed that there is a good orientation in [100] direction. However, there is some un-oriented grains in [110] direction. This may be due to the local grain orientation.

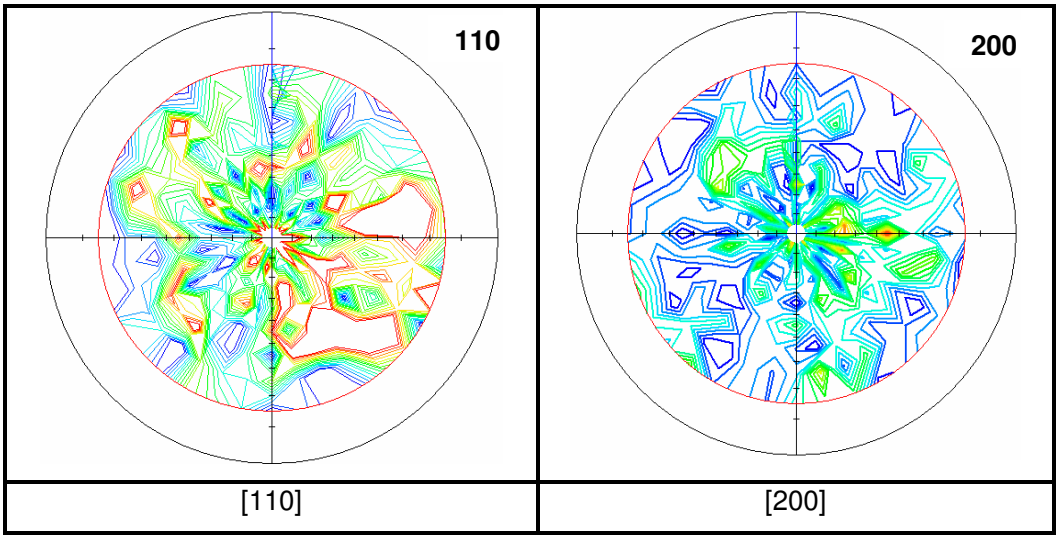


Figure 6.5:XRD Pole figure analysis of (110) and (200) planes for the modified Ti-Nb IF steel (Sample no: 1).

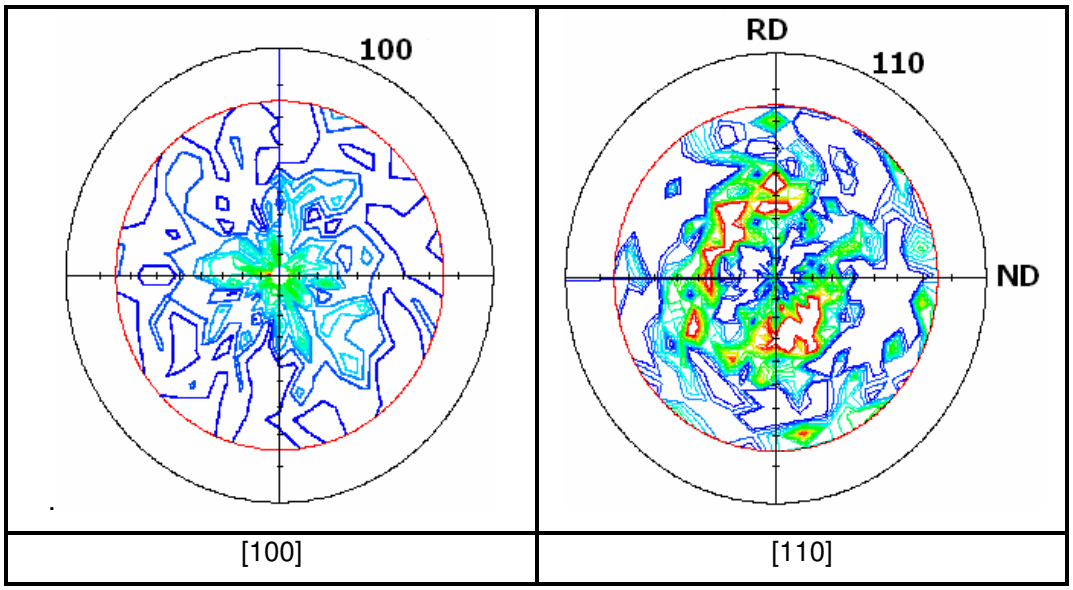


Figure 6.6:XRD Pole figure analysis of (100) and (110) planes for the modified Ti-Nb IF steel (Sample no: 5 which has r=3 value).

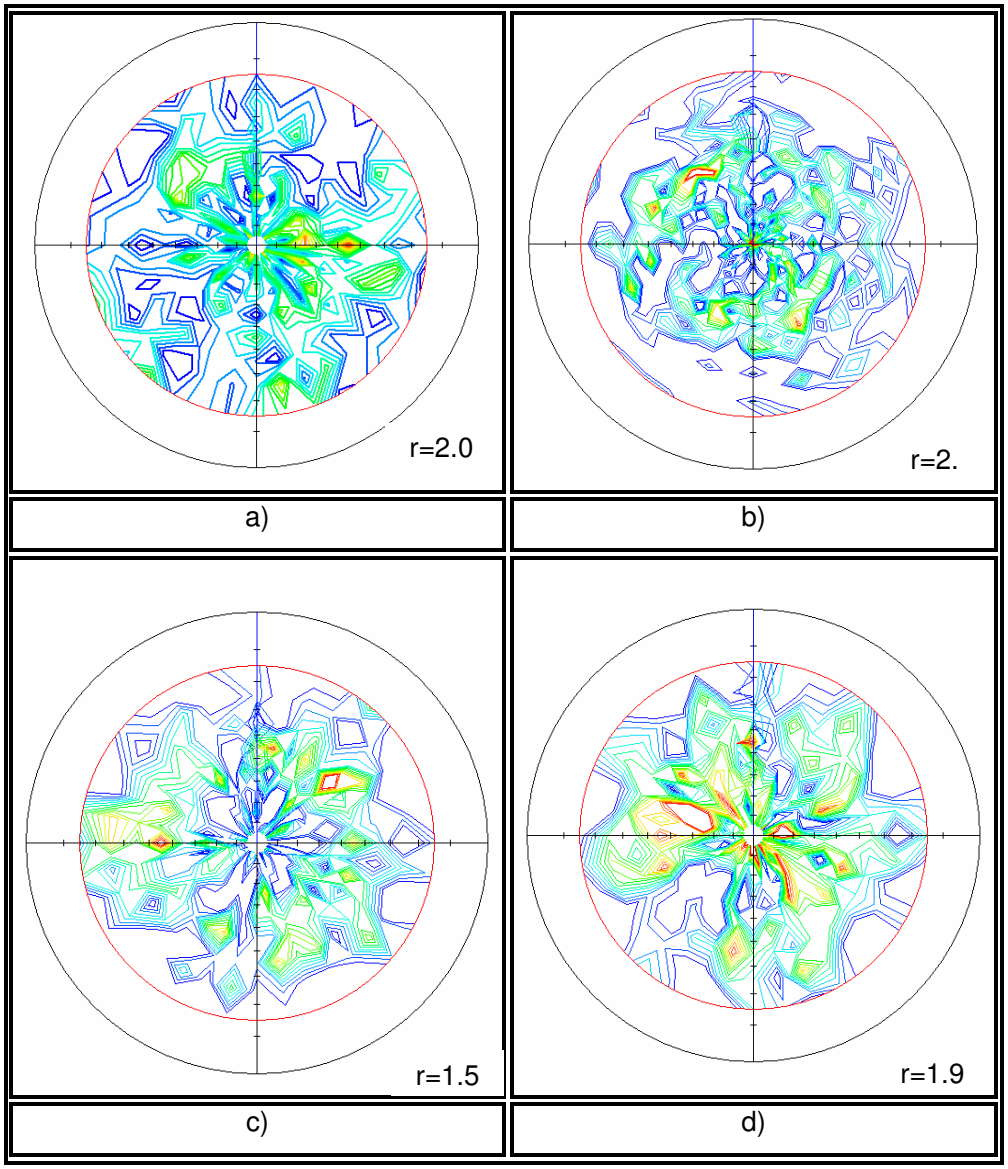


Figure 6.7:Pole figure analysis of modified Ti-Nb IF steel (grade 7316-CAL3) which process parameters given in Table 6.6 for (100) plane:

a) Sample No 1 (thickness: 0.9 mm, 77 % reduction ratio)
b) Sample No 2 (thickness: 1.0 mm, 75 % reduction ratio)
c) Sample No 3 (thickness: 1.8 mm, 60 % reduction ratio)
d) Sample No 4 (thickness: 2.0 mm, 55 % reduction ratio).

The pole figure analysis on the [100] plane is shown in Figure 6.7 for the Modified Ti-Nb IF steel (CAL 3) grade material, which has different thicknesses. As seen in this figure, the reflection is randomly distributed.

This shows that no strongly oriented texture structure exists in the material depending on the thickness / deformation ratio. This might be due to an insufficient deformation ratio, and/or the soaking temperature applied to the material. Similar textural behaviors were observed for the other samples.

6.4 Continuous Annealing Line Simulation Results

A continuous-annealing simulator was used to simulate annealing cycles. The re-crystallization characteristics of Modified Ti-Nb IF steel (grade 7316-CAL 3) were investigated and are shown in Figure 6.8. As can be seen in this figure, grains were getting larger with an increasing annealing temperature and time.

At annealing temperatures lower than 700 ℃, recrystallization did not occur. Therefore, no visual grain exists in samples that were annealed at 400 – 700 ˚C.

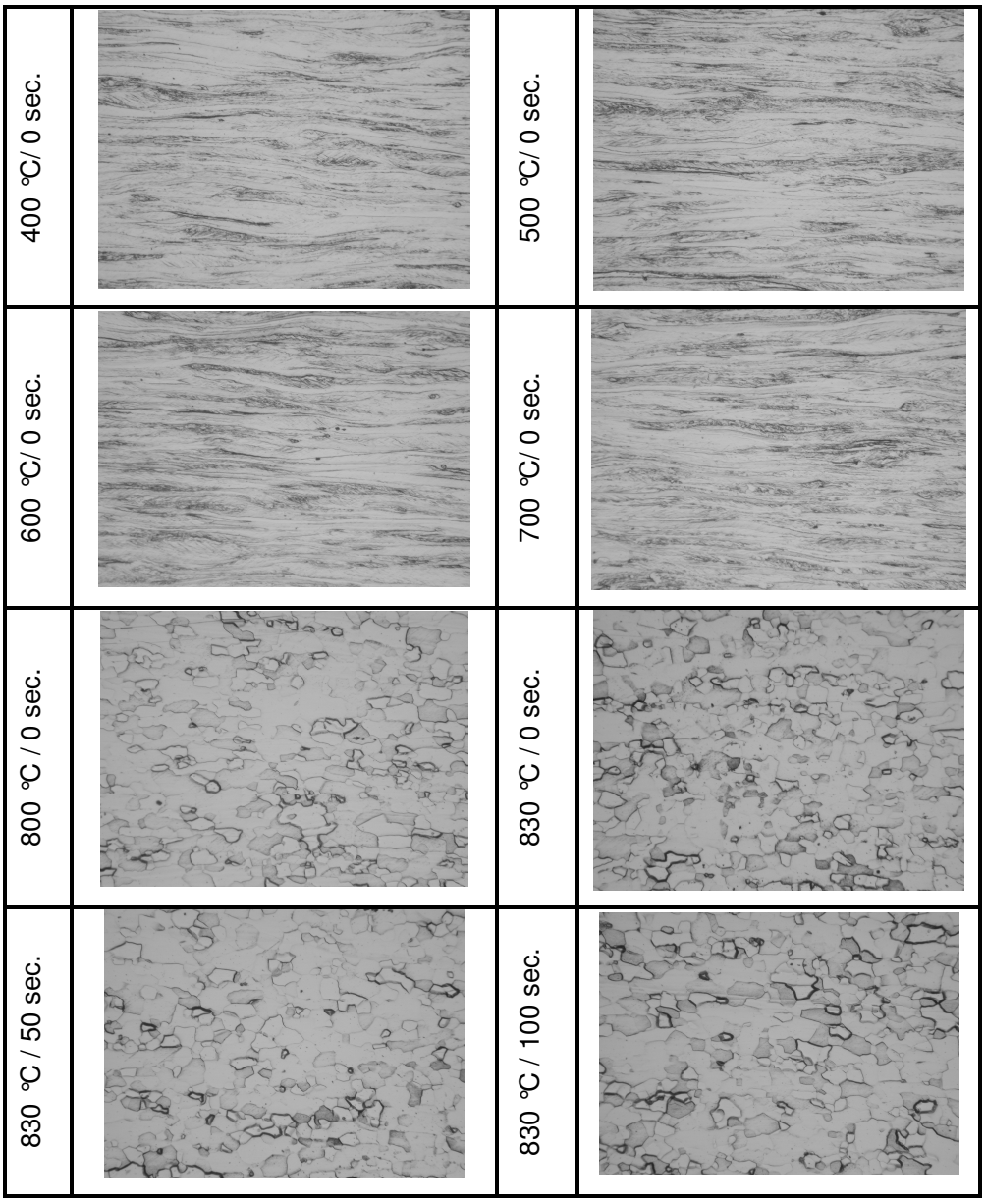
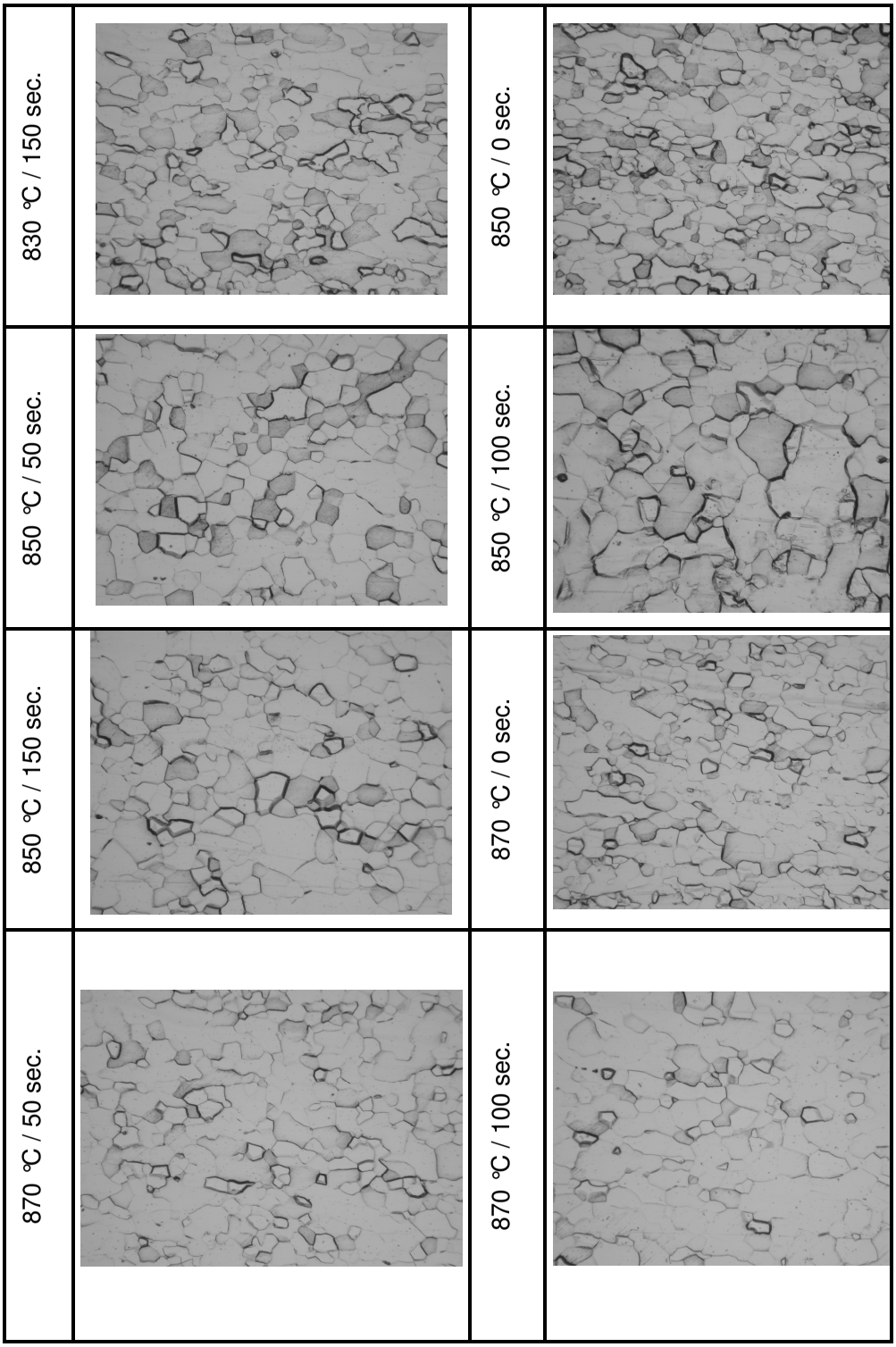


Figure 6.8:Structural changes in modified Ti-Nb IF steel (CAL 3) annealed at different temperatures and times (to be continued).

Figure 6.8:Structural changes in modified Ti-Nb IF steel (CAL 3) annealed at different temperatures and times.



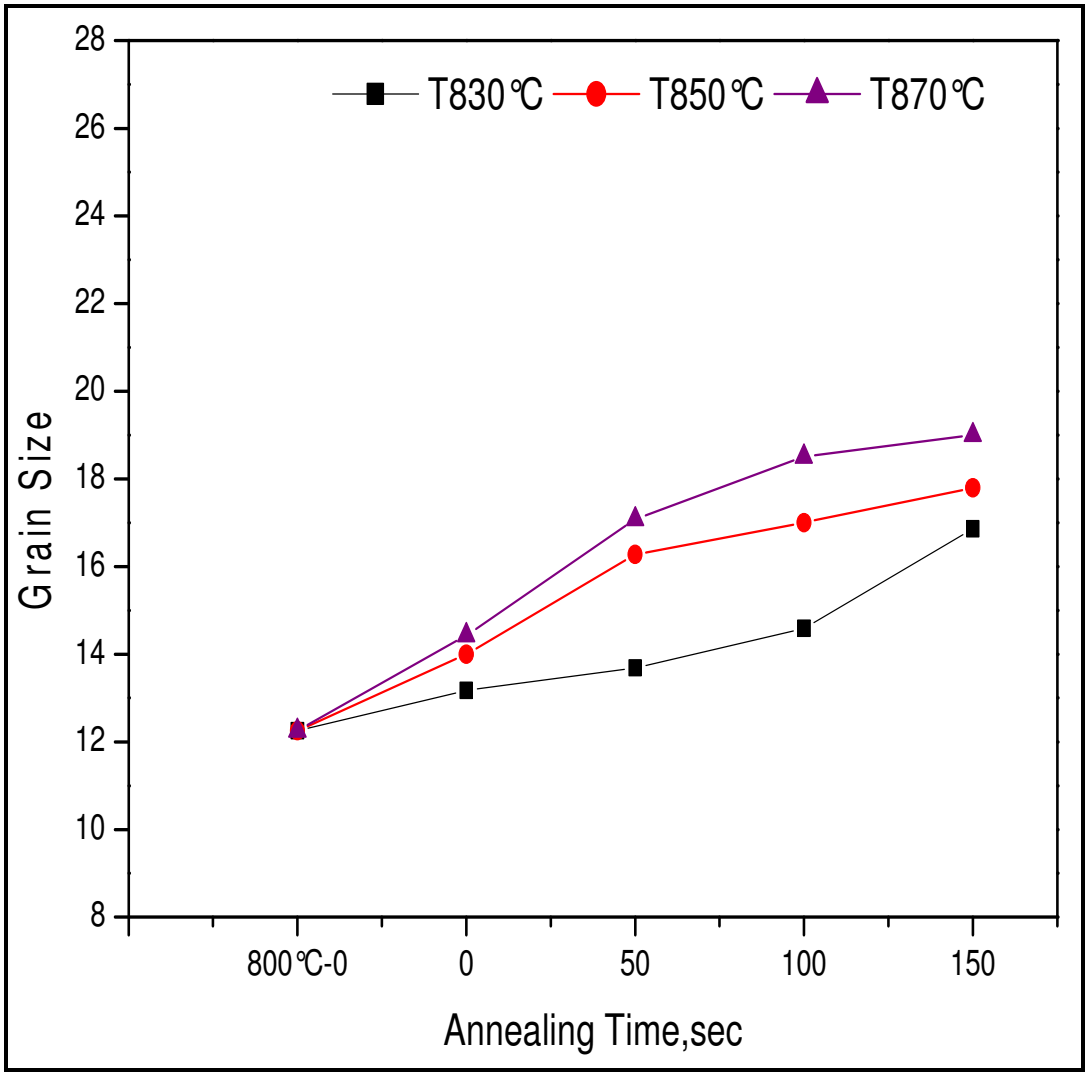


Figure 6.9:Effects of annealing temperature and time on grain size (in μm) for 7316 grade-CAL 3 modified Ti-Nb IF steel samples.

Figure 6.9 shows the effect of annealing temperatures and times on the grain size for 7316 grade-CAL 3 modified Ti-Nb IF steel samples. Re-crystallized structures were obtained when the specimens were annealed between 830 ˚C and 870 ˚C for 0, 50, 100, and 150 sec. The influence of soaking time on the grain size was recognized in cases of annealing below 830 ˚C.

The grain size changed between 12.2 μm and 19 μm and was increasing with an increasing soaking time and temperature. Katoh and friends [10] explained that finer grain sizes (8 to 12 μm) were encountered in the Ti-Nb grade compared to the Ti grade (12 to 20 μm).

7. GENERAL RESULTS

The following general results were obtained in this study which is done to optimize the alloying elements and process parameters for extra deep drawability of cold rolled automotive steels.

1) With Modified Ti-Nb IF steel (the grade 7316 -CAL 3) sample that has been cold rolled 80 % and annealed at 870˚C, the highest r value of 3.02 was achieved, which is mentioned as the maximum value in the literature.

2) For Ti-Nb IF steel (the grade 7318-CAL 2) and galvanized Ti-Nb IF steels (the grade 1315-CGL 1), increasing the excess Ti by approximately 50 % (0.0099 to 0.0148) results in an increase of r values from 1.8-2.0 to 2.3-3.02.

3) For Ti IF steel (the grade 7118 -CAL1) with 82% cold deformation ratio and 870 ˚C annealing temperature had the lowest yield strength and the highest r value in of 2.82.

4) In order to get higher r values, the cold deformation ratio should be much higher than 70 %, and annealing temperatures should be about 870 ˚C.

5) Yield strength, the ratio between yield strength and tensile strength, and elongation values decrease while the anisotropy coefficient (r) and tensile strength increase with an increasing cold deformation ratio at a certain annealing temperature range.

6) The lowest r value was obtained at the Ti-Nb IF steel (the grade 7318-CAL 2) which is cold rolled 47 % and annealed at 830 ˚C.

7) The highest r and n values were obtained in the annealing temperature range between 840˚C and 860˚C for a Ti IF steel (the grade 7118-CAL 1) grade which was cold rolled 77 %. The highest n value was obtained for Ti IF steel (the grade 7118-CAL 1) which was cold rolled 77 %.

8) The optimum annealing temperature was 850 ˚C for Ti-Nb IF steel (the grade 7318-CAL 2) at 73% deformation ratio, and the optimum annealing temperature was 870 ˚C for modified Ti-Nb IF steel (the grade 7316-CAL 3) at 80 % deformation ratio.

9) Optimum formability was obtained at 865℃ with a 80 % deformation ratio for galvanized Ti-Nb IF steel (the grade 1315-CGL 1), 845℃ with 67% deformation

ratio for modified galvanized Ti-Nb IF steel (grade 1315-CGL 2) and 850℃ with 77 % deformation ratio for low N content galvanized Ti-Nb IF steel (the grade 7316-CGL 3).

10) The effect of soaking time on the grain size was large when the soaking happens just above the re-crystallization temperature. In the continuous annealing process, the soaking time is limited to within a few secs. Thus, the soaking temperature is significant for controlling the mechanical properties of stabilized ultra low carbon IF steels.

11) There was a tendency that the grain size was getting smaller with an increasing deformation ratio at constant annealing temperatures. Among the modified Ti-Nb IF steel, one of them which was cold rolled 80 % and annealed at 870˚C, had the largest grain size of 44.3 μm, whereas Ti IF steel (grade 7118-CAL 1) that was cold rolled 73 % and annealed at 870˚C, had the smallest grain size of 18.5 μm. Among the samples of low N content galvanized Ti-Nb IF steel (CGL 3) one of them that was cold rolled 77 % and annealed 850 ˚C had the largest grain size of 33.6 μm. Galvanized Ti-Nb IF steel (grade 1315-CGL 1) that was cold rolled 77 % and annealed at 865 ˚C, had the smallest grain size of 19.4 μm.

12) In texture analysis, there were three different diffraction angles on the Modified Ti-Nb IF steel (grade 7316-CAL 3), but there was no strong reflection in plane texture structures on the material. The strongest reflection was seen on the (110) plane. There is no oriented texture structure on the material depending on the thickness / deformation ratio.

13) In the CAL Simulation Study, recrystallized structrures were obtained when the specimens were annealed between 830-870 ˚C for 0, 50, 100 and 150 secs. The influence of soaking time on the grain size was recognized below annealing temperature of 830 ˚C. The grain size increased with an increasing soaking time and soaking temperature.

8. FUTURE STUDIES

1) In the future, galvannealed and galvanized steels can be compared in order to examine the deep drawability property.
2) Limit Forming Diagram studies can be done for the CAL and CGL samples.
3) As mentioned in the literature, studies about hot rolled conditions are very important for texture growth. May be that is the reason for the weak texture observed in the CAL and CGL samples. This subject should be investigated.
4) Kinetics of recovery processes in the CAL/CGL processing should be researched in the future.

REFERENCES

[1] **Carabajar S., Merlin J., Massardier V and Chabanet S.,** 2000. Precipitation evolution during the annealing of an interstitial free steel**,** *Materials Science and Engineering Proceedings* **A281** , 132-142

[2] **Llewellyn D.T. and Hudd R.C.,**1998. Steels: Metallurgy and Applications, 3rd edn., 39., London, Butterworth-Heinemann

[3] **Obara T.and Sakata K.,**1998. Development of metallurgical aspects and processing technologies in if sheet steel**,** *39TH MWSP NF.PROC.,ISS*,**XXXV**, 307

[4] **Kıtamura M., Tsukatani I. and Inoue T.,**1994. Effect of carburizing after recrystallization on formability and bake hardenability of ultra low carbon ti-added cold-rolled sheet steels, *ISIJ International*, **34**, **1**, 115

[5] **Jeong W.C.,**2000. Strength and formability of ultra-low-carbon ti-if steels*, Metallurgical and Materials Transactions A***, 31A**, 1305

[6] **Alan W. Cramb. and Byrne M.,**1990. Steelmaking and casting practices for high quality interstitial free steels, metallurgy of vacuum degassed steel products edited by R.Pradhan, 3-25

[7] **Hawbolt E.B. and Mukunthan K.,**1996. Modelling recovery and recrystallization kinetics in cold rolled ti-nb stabilized interstitial-free steel, *Metallurgical and Materials Transactions* A **27A**, 3421

[8] **Matsuoka S., Sakata K., Satoh S., and Kato T.,** 1994. Effect of hot-rolling strain rate in the ferrite region on the recrystallization texture of extra-low c sheet steel, *ISIJ International*, **1**, **34**, 77

[9] **Tsukatani I.,** 1998. Effect of hot band micro structural factors on recrystallization texture in ti-added ultra-low carbon cold-rolled sheet steel, *ISIJ International*, 625.

[10] **Katoh H., Takechi H., Takahashi N. and Abe M.,** 1985. Cold rolled steel sheets produced by continuous annealing, pp.37-60, Technology of Continuously Annealed Cold Rolled Sheet Steel, R. Pradhan ed., TMS-AIME, Warrendale, PA.

[11] **Hayakawa H., Takahashi N., Shibata M., Kawano A. and Matsuda M.,** 1987. Effects of ti addition and hot and cold rolling condition on the recrystallization behaviour during continuous annealing of very low carbon cold rolled steel sheet, pp.109-116, *Tetsu-to-Hagane*, **73(2).**

[12] **Goodman S.R., Mould P.R. and Siple J.C.,** Effects of composition and processing on the recrystallization behavior and tensile properties of continuous annealed titanium –containing interstitial free steel sheet, pp.167-184, ibid

[13] **Subramanian S.V., Prikry M.I, Ulabhaje A., and Balasubramanian K.,** 1991. in interstitial free steel sheet: processing, fabrication and properties, (ed. L.E.Collins and D.L.Baragar), 15-38; Ottawa,CIM/ICM.

[14] **Tither G., Garcia C.I., and Deardo A.J.,** 1994.:Proc.*Int.Forum on Physical metallurgy of if steels*, Tokyo, Japan, Iron and Steel Institute of Japan, 293-322.

[15] **Wilshynsky-Dresler D.O., Matlock D.K., and Krauss G.,** 1995. *ISS Mech. Work. Steel Process. Conf.*, 33**,** 927-940**.**

[16] **Katoh H., Takechi H., Takahashi N., and Abe M.,** 1985 in Technology of continuously annealed cold-rolled sheet steel, (ed.R.Pradhan), 37-60; Warrendale PA, AIME.

[17] **Hutchinson B.,** *Mater.Sci.Forum*, 1994. 157-162, 1917-1928

[18] **Tsunoyama K., Sakata K., Obara T., Satoh S., Hashiguchi K., and Irie T.,** 1988. in Hot and cold rolled sheet steels, (ed. R.Pradhan and G. Ludkovsky), 155-165; Warrendale, PA. TMS.

[19] **Mizui N.** 1998. Proc. Int.symp. on Modern lc and ulc sheet steels for cold forming, processing and properties, Aachen, Germany, (ed. W.Bleck), 169-178.

[20] **Tither G. and Stuart H.,** 1995. in HSLA steels 95, (ed. L.Guoxun et al.), 22-31; Beijing, Chinese Society for Metals.

[21] **Tokunaga Y. and Yamada M.**:Method for the production of a cold rolled steel sheet having super deep drawability, *US Patent* 4, 504-326.

[22] **Tsunoyama K., Satoh S., Yamazaki Y., and Abe H.,** 1990. in Metallurgy of vacuum degassed steel products, (ed. R. Pradhan), 127-142; Warrendale, PA.TMS

[23] **Bleck W., Bode R. and Hahn F.J.,** 1990. Interstitial free steels processing, properties and application*, Metallurgy of Vacuum Degassed Steel Products,* edited by R.Pradhan, 73-90

[24] **Wilshynsky D.O., Matlock D.K. and Krauss G.**, 1990. Microstructure and mechanical properties of interstitial free steel sheet steels with different stabilizing additions*, Metallurgy of Vacuum Degassed Steel Products* edited by R.Pradhan, 247 – 261

[25] **Mendoza R., Alanis M., Alvarez-Fregoso O. and Juarez-Islas J.A.**, 2000. Processing conditions of an ultra low carbon / ti stabilized steel developed for automotive applications , 771

[26] **Parker B.A., and Hirsch Y.C.,** 1983. Advances in physical metallurgy and applications of steels*, The Metals Society,* London, 26

[27] **Asamura T.,** 1998. Recent development of modern lc and ulc sheet steels in japan, *International Symposium Modern LC and ULC Sheet Steels for Cold Forming: Processing and Properties Conference Book* ***1***, Aachen, Germany,

[28] **Hutchinson W.B., Nilsson K.I. and Hirsch J***.,* 1990. Annealing textures in ultra low carbon steels, *Metallurgy of Vacuum Degassed Steel Products* edited by R.Pradhan, 109 – 125

[29] **Mendoza R., Alanis M., Alvarez-Fregoso O.and Juarez-Islas J.A.,** 2000. Processing conditions of an ultra low carbon / ti stabilized steel developed for automotive applications , 771

[30] **Pradhan R.,** Cold rolled interstitial free steels: A Discussion of some metallurgical topics*,* Bethlehem Steel Research Labs. USA, 167-168

[31] **Mizui N.**:Proc.Int.Symp.on Modern lc and ulc sheet steels for cold forming: processing and properties, Aachen Germany, 1998, (ed. W.Bleck), 169-178

[32] **Takechi H.**, 1994. Metallurgical aspects on interstitial free sheet steel from industrial viewpoints*, ISIJ*, pp 1,3-6

[33] **Matsudo K., Shimomura T., Osawa K., Kinoshita M., Ono S. and Nozoe O.,** 1938. Production of deep drawing quality cold rolled steel sheets by the nkk continuous annealing line process (NKK-CAL Process), *Nippon Kokan Technical Report Overseas* **No.38**, 14-16

[34] **Vanderschueren D., Yoshinaga N., and Koyama K.,** 1996. Recrystallization of ti if steel investigated with electron back-scattering pattern (EBSP), *ISIJ International*, **Vol.36**, **8**, 1046

[35] **Mendoza R., Huante J., Alanis M., Gonzales-Rivera C., and Juarez-Islas J.A.,** 2000. Processing of ultra low carbon steels with mechanical properties adequate for automotive applications in the as-annealed condition, *Materials Science and Engineering* **A276**, 203-209

[36] **Ray R.K., Jonas J.J., and Hook F.J.**: *Int. Mater. Rev.*, 1994. **39,** (4), 129-172.

[37] **Dasarathy C.** A review of recent observations on the deformed, recovered and recrystallized states iron and low carbon ferrite steels, *Report* SM/668/A, British Steel Corp., 1973

[38] **Ohashi N., Irie T., Satoh S., and Hashimoto O.,** 1981:Development of cold rolled high strength steel sheet with excellent deep drawability, *SAE Paper* 810027, Society of Automotive Engineers, Warrendale, PA, USA.

[39] **Sanagi S.** Curr. Adv. *Mater. Proc.*, 1990. (3), 1768-1771.

[40] **Satoh S., Obara T., Takasaki J., Yasunda A., and Mishida M.,** 1985: *Kawasaki Steel Tech.Rep.* **, 12**

[41] **Satoh S., Obara T., Nishida M., and Irie T.,** 1985. in Technology of continuously annealed cold-rolled sheet steel, (ed. R.Pradhan), 151-166; Warrendale, PA, AIME

[42] **Gupta I., Parayh T., and Shang I.T.,** 1988. Hot and cold rolled sheet steels, (ed.R.Pradhan and G.Ludkovsky), 139-153; Warrendale, PA, TMS

[43] **IF Steel Times**, May 1998. (Basic Metal Processing Research Institute, University of Pittsburgh, PA, USA).

[44] **Hashimoto O., Satoh S., Irie T., and Ohashi N.,** 1982. Proc. Int. Conf. on Advances in physical metallurgy and applications of steel, *Book* 284, 95-104; Warrendale, PA, TMD.

[45] **Sato S., Obara T., Takasuki J., Yasuda A., and Nishida M.,** 1984. *Kawasaki Steel Tech. Rep.*, 1984, **(16),** 273-279.

[45] **Furuno Y., Kawano A., Sayanagi S., Matsuda M.**, **H.Hayakawa, and M.Shibata.,** 1987. *Tetsu-to-Hagane (J.Iron Steel Inst. Jpn)*, **72,** S637.

[46] **Kwon O., Kim G., and Chang R.W.,** 1990. in Metallurgy of vacuum degassed steel products, (ed. R.Pradhan), 9-12; 1990, Warrendale, PA, TMS

[47] **Gupta I. and Bhattacharya D.,** 1990. in Metallurgy of vacuum degassed steel products, (ed.R.Pradhan), 43-72; Warrendale, PA, TMS

[48] **Takahashi N., Abe M., Akisue O., and Kato H.,** 1982. in Metallurgy of continuously annealed sheet steel, 51; Warrendale, PA, and TMS

[49] **Obara T., Satoh S., Nishida N., and Irie T.,** 1984. Scand, *J.Metall*, 1984, **13,** 201-213.

[50] **Subramanian S.V., Prikryl M., Gaulin B.D., Clifford D.D., Benincasa S., and O'reilly I.,** 1994. *ISIJ Int.*, 1994, **34, (1),** 61-69.

[51] **Irie T., Satoh S., Yasuda A., and Hashimoto O.,** 1982. in Metallurgy of continuous annealed sheet steel, (ed.B.LBramfitt and P.L.Mangonon),155-171; Warrendale, PA, TMS.

[52] **Bakkaloğlu A.,** 2001. IF çeliklerinin özellikleri üzerinde proses parametrelerinin etkisi, *I nci Demir-Çelik Sempozyumu*, 695-697, Kdz.Ereğli

[53] **Kino N., Yamada M., Tokunaga Y., and Tsuchiya H.,** 1990. in Metallurgy of vacuum degassed steel products, (ed.R.Pradhan), 197-214; Warrendale, PA.TMS.

[54] **Tanaka H., Ogura Y., Sekiya T., and Yamada M.,** 1985. *Bull. Jpn. Ins. Met.*, **24,** (4), 316-318.

[55] **Yamada M., and Tokunaga Y.,** 1985. *Trans.Iron Steel Inst. Jpn.*, **25,** 272

[56] **Bleck W., Bode R., and Hahn F.J.,** 1990. *Thyssen Tech.Ber.*, (1), 69-85.

[57] **Hashimoto N., Yoshinaga N., and Senuma T.,** 1997. Texture evolution of if steel due to recrystallization, *ISIJ International*, 617

[58] **Osawa K., Satoh S., Obara T., Katoh T., Abe H. and Tsunoyama K.,** 1990. Recrystallization behavior of extra low c cold rolled sheet steels during continuous annealing*, Metallurgy of Vacuum Degassed Steel Products,* edited by R.Pradhan, 181 – 196

[59] **NKK Corporation**, 1995. *Technical Information on NKK-CAL-PROCESS*, **Vol.1**, 77-98

[60] **Tanaka H., Y.Ogura, K.Yoshioka, T.Hirose, H.Yamada, and M.Sako.,** 1987 *NKK Tech.Rep* (Overseas), **(48)**, 21-28.

[61] **Hudd R.**, 1997. Welsh Technology Centre, British Steel Strip Products, Port Talbot, UK, *personal communication*.

[62] **Toki T., Oshima K., Nakamori T., Saito Y., Tsuda T., and Hobo Y.,** 1993. in The physical metallurgy of zinc coated steel, (ed.A. R.Marder), 169-180; Warrendale, PA. TMS.

[63] **Kato C., Koumura H., Uesugi Y., and Mochizuki K.** 1993. in The physical metallurgy of zinc coated steel, (ed. A.R.Marder), 241-250; Warrendale, PA, TMS

[64] **Y.Maehara, M.Mizui, and M.Arai,** 1991. in Interstitial free steel sheet: processing, fabrication, and properties, (ed. L.E.Collins and D.L.Baragar), 135-144; Ottawa, CIM/ICM.

[65] **Matas S.J.,** Properties of ultra-low-carbon (interstitial free) steels.

[66] **Yan B.,** 1995. ISS *Mech. Work.Steel Process. Conf.*, 1995, **33,** 101-114

[67] **Bergengren Y. and Melander A.,** 1992. *Int. J. Fatigue*, **14,** 154-162.

[68] **Gündüz O.,** 1999. Sürekli döküm proses parametrelerinin makroinklüzyon oluşumuna etkileri ve gerçek zamanlı bir kalite tahmin modelinin oluşturulması, *Doktora Tezi*, İ.T.Ü. Fen Bilimleri Enstitüsü, İstanbul

[69] **Çayır E.,** 1996. Optimization of energy consumption in iron and steelmaking process by use of second law of thermodynamics, *Ph.D.Thesis*, İ.T.Ü. Fen Bilimleri Enstitüsü, İstanbul.

[70] **ERDEMİR**, 2001. *Ürün Kataloğu*, Kdz. Ereğli.

[71] **Koç, M.,** 2002. Çelik Üretim Prosesi, Erdemir, Kdz.Ereğli

[72] **Tozlu, İ.** 2002. *1.Sıcak Haddehane Tanıtım Kitabı*, Erdemir, Kdz.Ereğli.

[73] *Instruction Manual for Tandem Mill* (CPL-TCM) for Ereğli Demir ve Çelik Fabrikaları T.A.Ş**,** 1996, Hitachi Ltd., **1**, Tokyo

[74] **NKK Corporation**, 1995. *Technical Information on NKK-CAL-PROCESS*, **1**, 17-20

[75]**Nippon Steel Corporation**, 2000. *Instruction Manuel for Continuous Galvanizing Line for ERDEMIR*, Tokyo

[76] **Riha, G., GuggeNberger J., Kaiser M.,** 2004. The New Galvanizing Line No. 3 at voestalphine Stahl GmbH, Galvatech "04 *Conference Proceedings*, 173-184

[77] **Kayalı E.S., Çimenoğlu H., Baydoğan M., Akoy M.A.,** 2005. Erdemir üretimi sacların anizotropi katsayılarının geliştirilmesi, *Sonuç Raporu*, Kdz. Ereğli

[78] **Kırmanoğlu I**., (2001) "IF çeliklerinin üretim prosesleri ve derin çekme özelliklerinin araştırılması", YTÜ, *Bitirme Tezi*, 1-78

[79] **Liu Y., Sun J., Zhou L., Tu Y., Xing F., Guo Y.and Tong Q.,** 2003. "Experiment of investigation of deep drawing sheet texture evolution", *Journal of Material Processing Technology*, **140**, 509-513.

[80] **Eloot K., Okuda K., Sakata K.and Obara T.,** 1998. "Texture evolution during cold rolled steel with a strong {111} hot rolling and re-crystallization band texture", *ISIJ International*, **38** No. 6, 602-609.

APPENDIX A: CAL LINE RESULTS

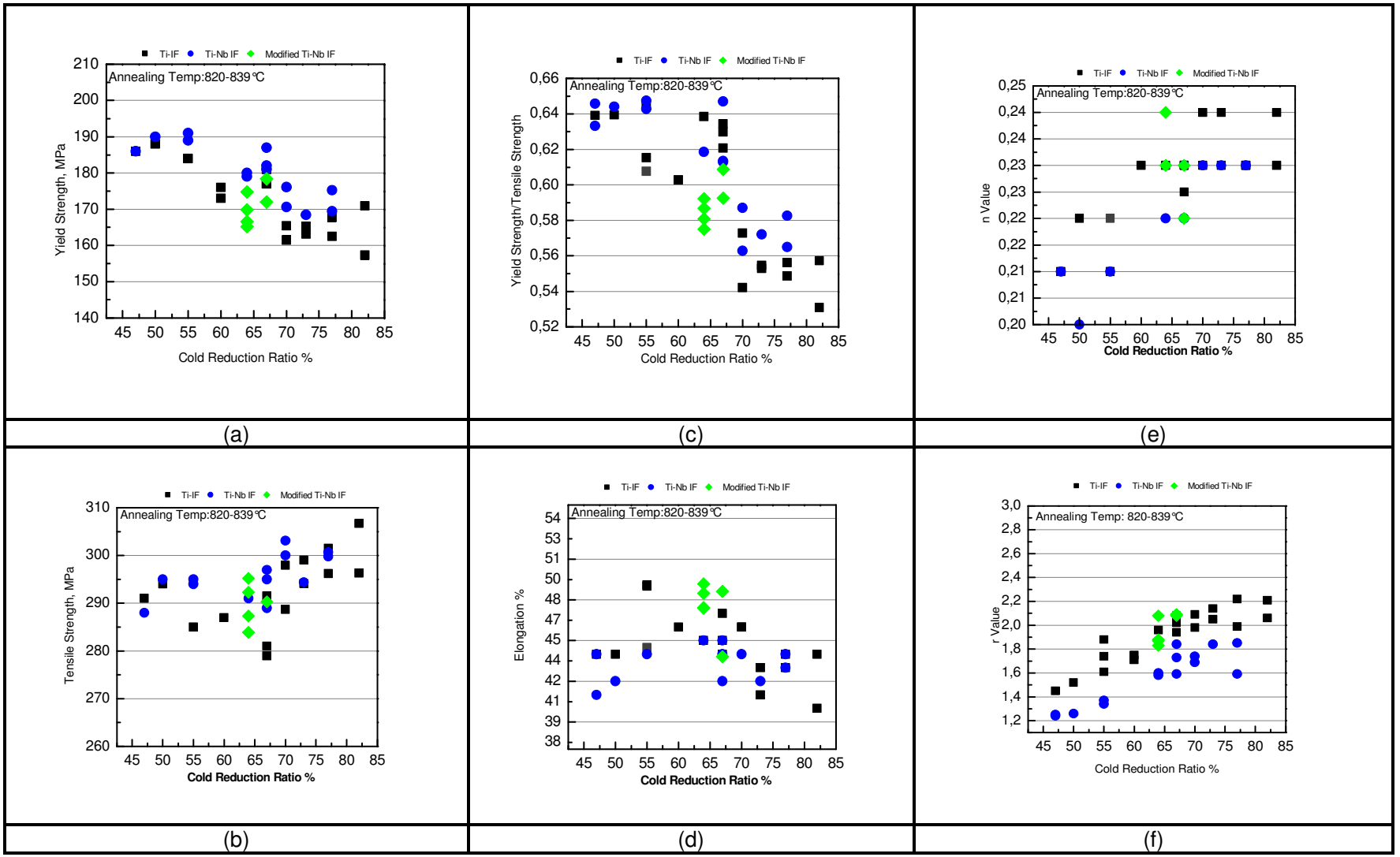


Figure A.1:Effects of the cold reduction ratio on a) Yield Strength b) Tensile Strength c) Yield Strength/Tensile Strength d) % Elongation e) n value f) r value of the annealed sheets at 830 ºC.

APPENDIX A: CAL LINE RESULTS

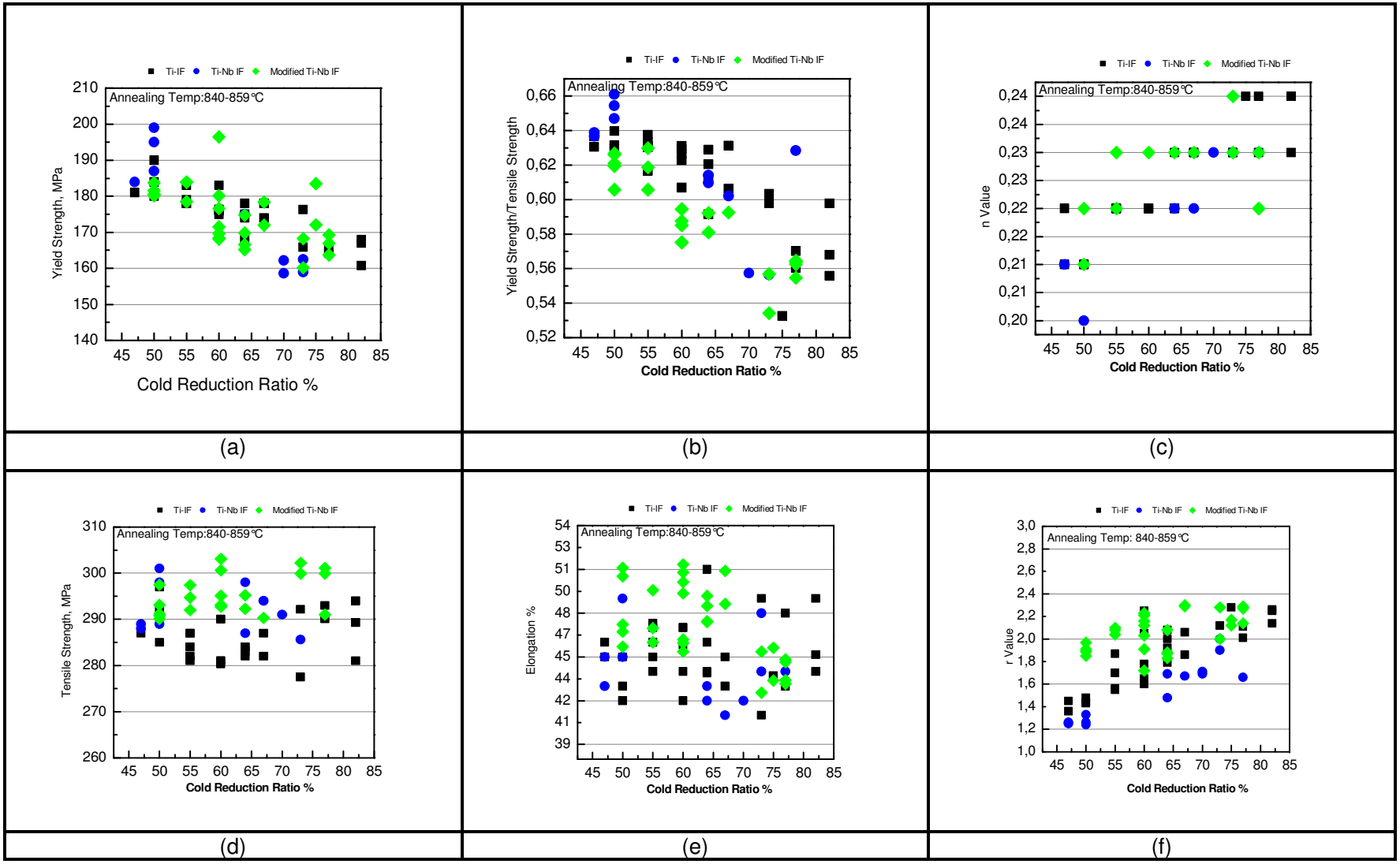


Figure A.2:Effects of the cold reduction ratio on a) Yield Strength b) Tensile Strength c) Yield Strength/Tensile Strength d) % Elongation e) n value f) r value of the annealed sheets at 850 ºC

APPENDIX A: CAL LINE RESULTS

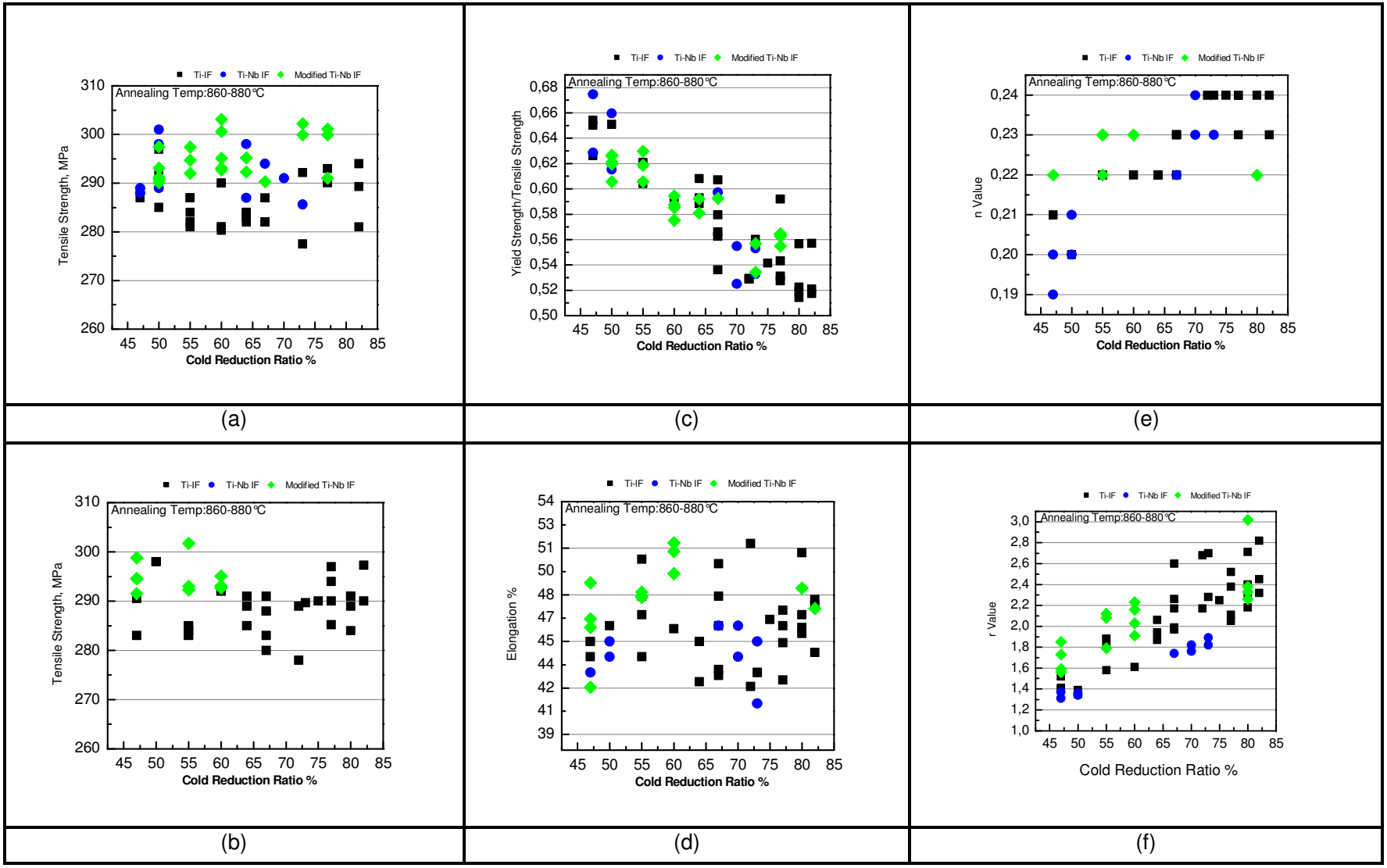


Figure A.3:Effects of the cold reduction ratio on a) Yield Strength b) Tensile Strength c) Yield Strength/Tensile Strength d) % Elongation e) n value f) r value of the annealed sheets at 870 ºC

APPENDIX B: CGL LINE RESULTS

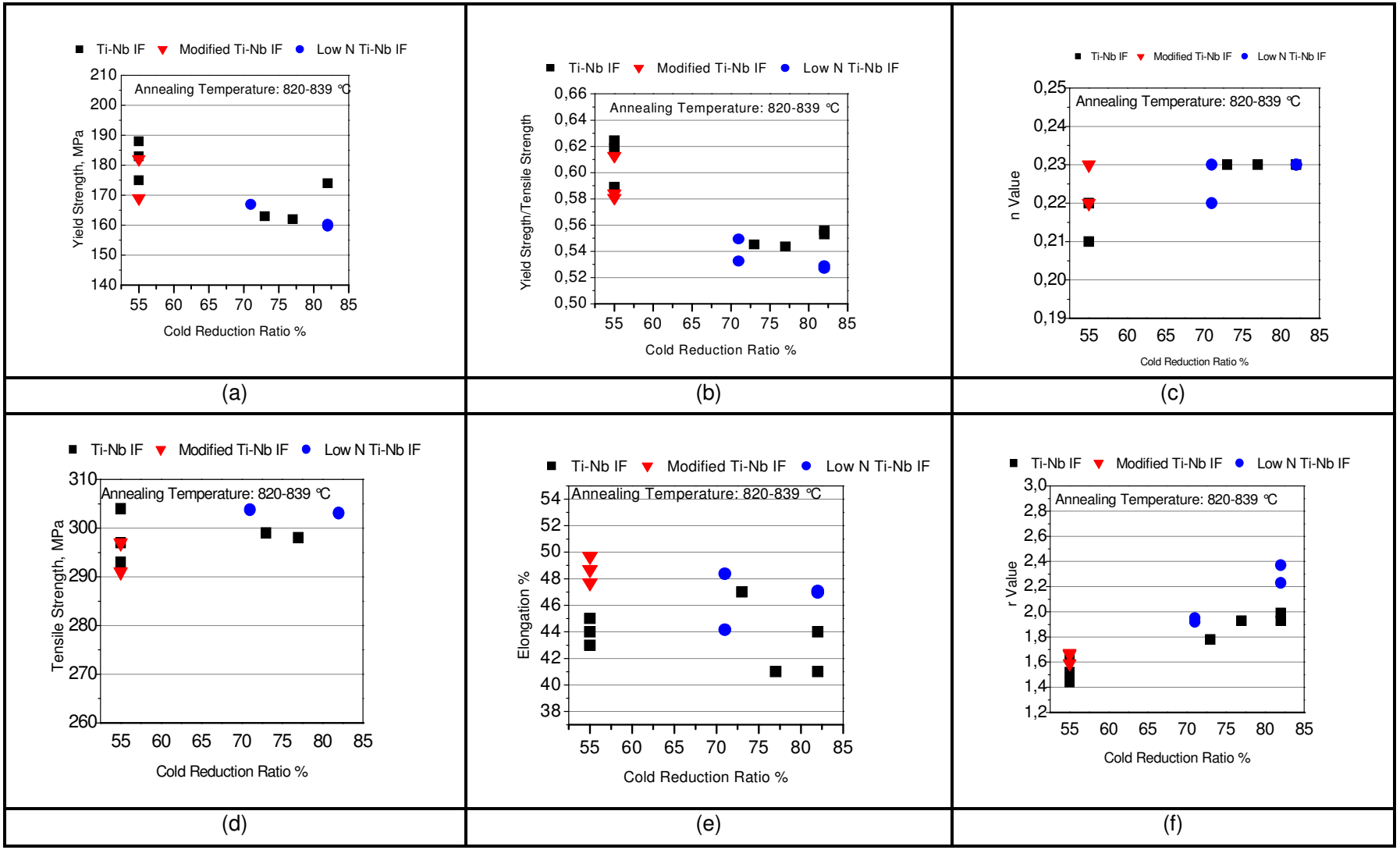


Figure B.1:Effects of the cold reduction ratio on a) Yield Strength b) Tensile Strength c) Yield Strength/Tensile Strength d) % Elongation e) n value f) r value of the annealed and galvanized sheets at 830 ºC

APPENDIX B: CGL LINE RESULTS

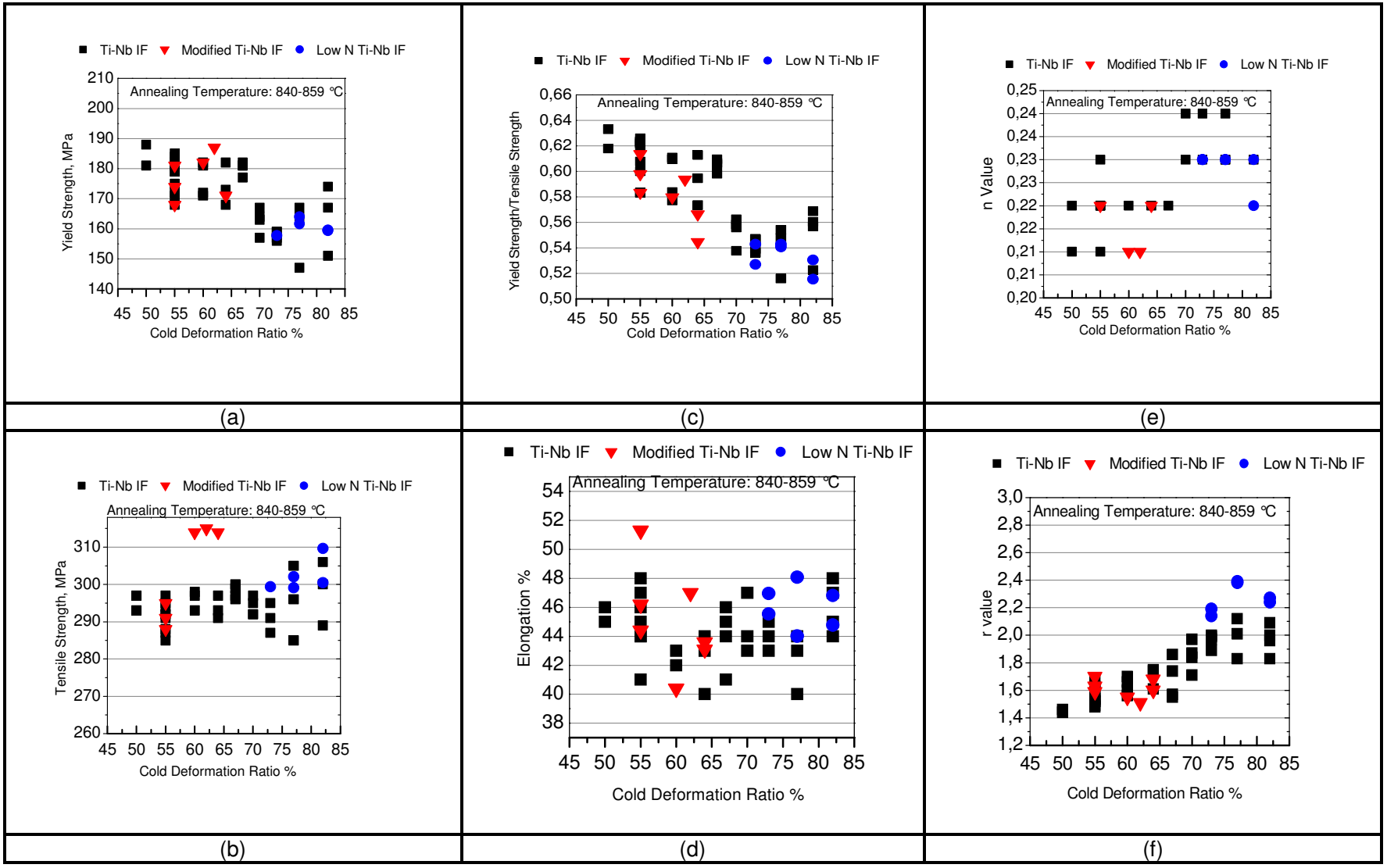


Figure B.2:Effects of the cold reduction ratio on a) Yield Strength b) Tensile Strength c) Yield Strength/Tensile Strength d) % Elongation e) n value f) r value of the annealed and galvanized sheets at 850 ºC

APPENDIX B: CGL LINE RESULTS

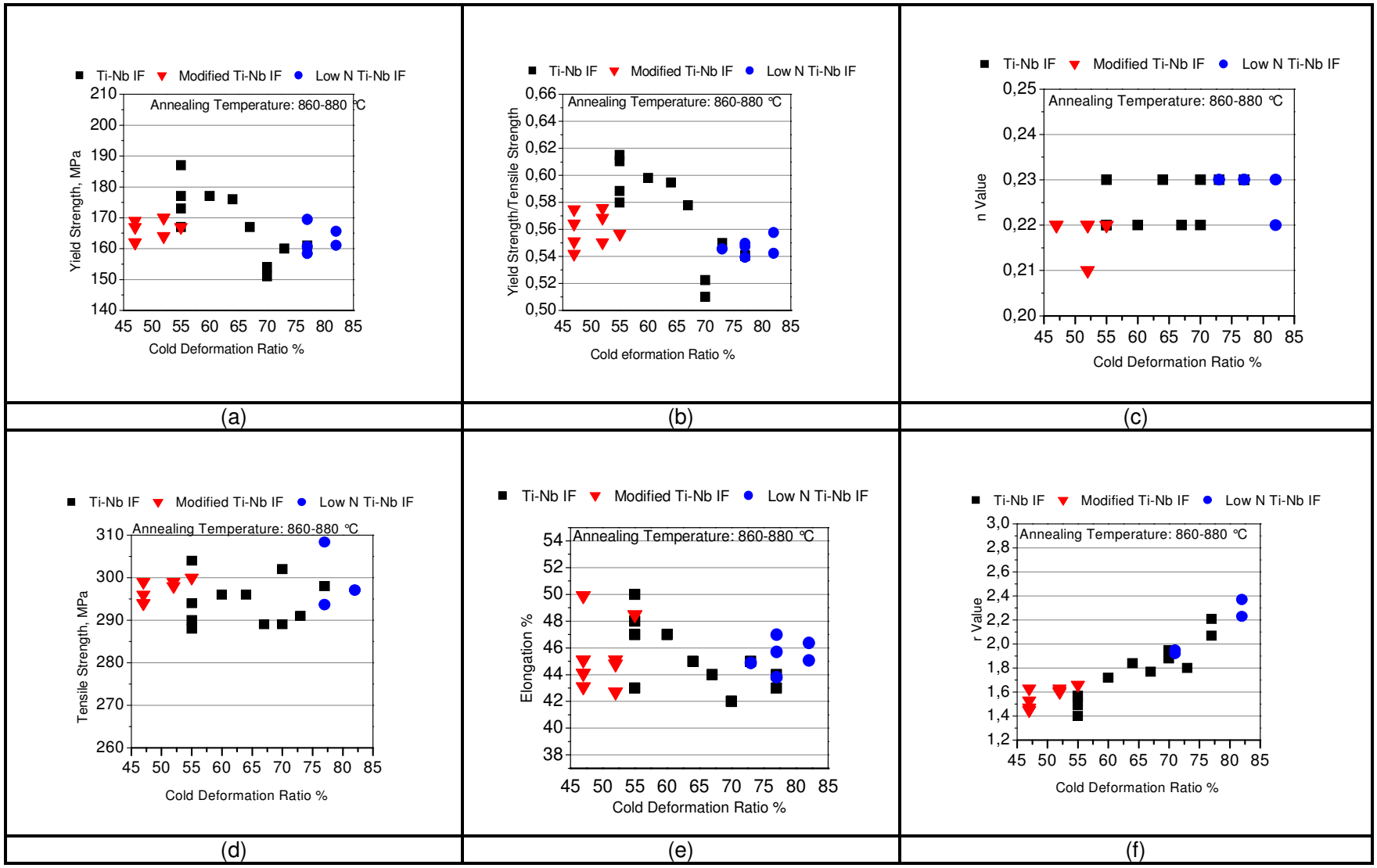


Figure B.3:Effects of the cold reduction ratio on a) Yield Strength b) Tensile Strength c) Yield Strength/Tensile Strength d) % Elongation e) n value f) r value of the annealed and galvanized Sheets at 870 ºC

APPENDIX C: GRAIN SIZE RESULTS

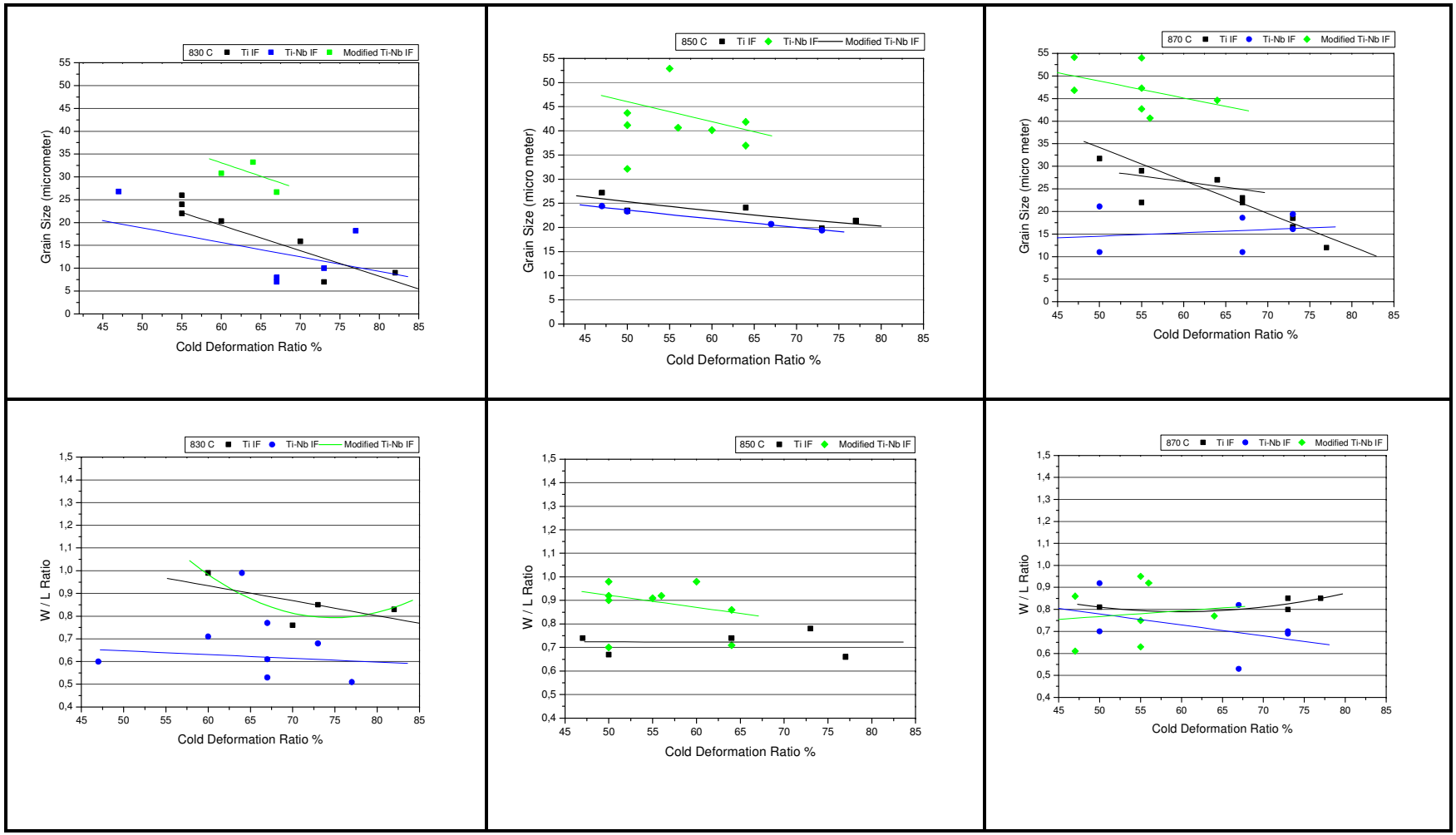


Figure C.1:The effect of cold deformation ratio on the grain size and Width/Length ratio of steels Ti IF, Ti-Nb IF and modified Ti-Nb IF steel at 830˚C,850˚C and 870 ºC

APPENDIX C: GRAIN SIZE RESULTS

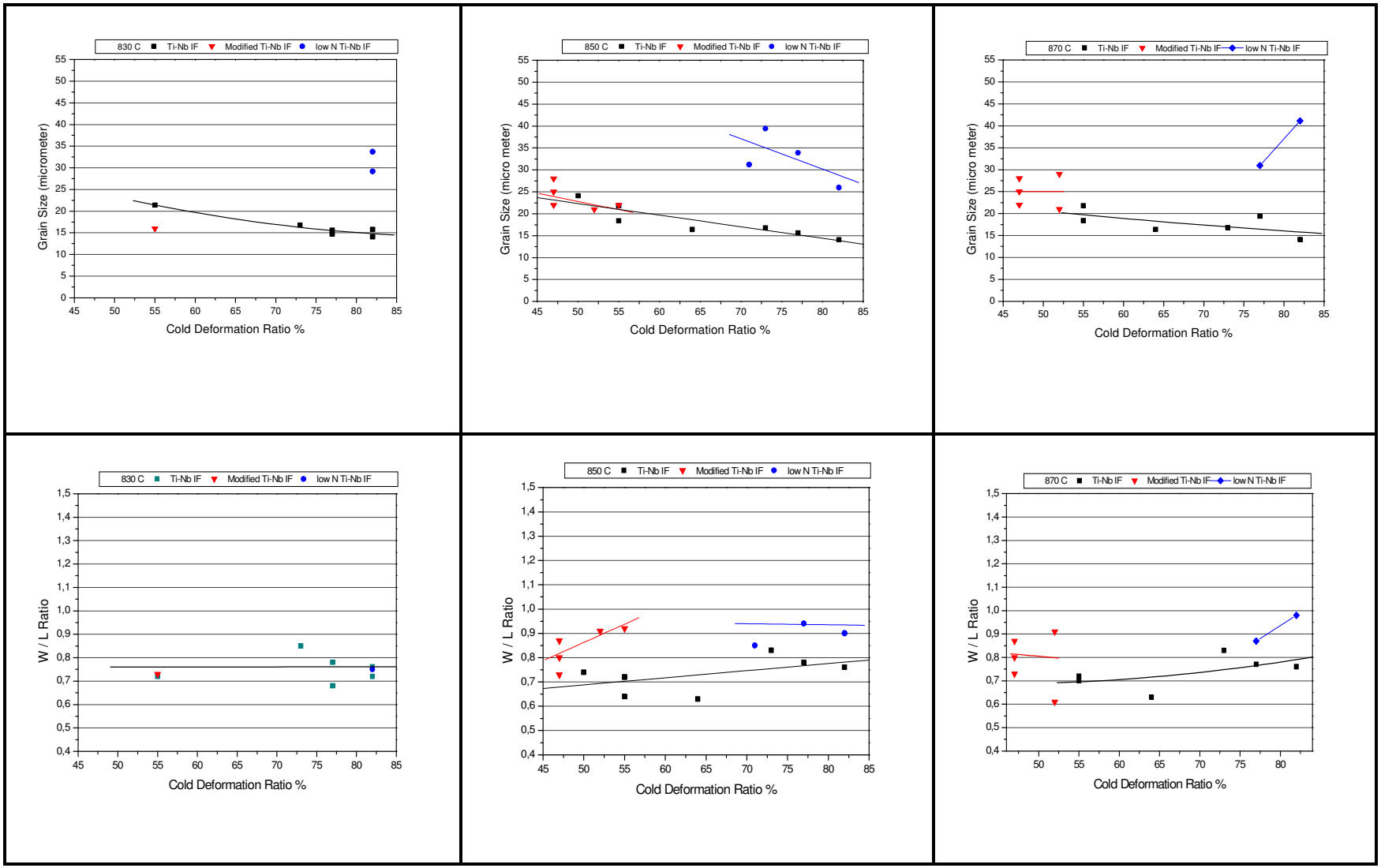


Figure C.2:The effect of cold deformation ratio on the grain size and Width/Length ratio of steels galvanized Ti IF, galvanized and modified Ti-Nb IF, galvanized low N Ti-Nb IF at 830 ˚C. 850˚C and 870 ºC

APPENDIX D: CHEMICAL COMPOSITIONS

Table D.1:Chemical Compositions of galvanized Ti Nb IF, Ti IF and Ti Nb IF steel samples

Döküm No	Döküm Kalitesi	C %	Mn %	P %	S %	Si %	Al %	Cu %	Cr %	Ni %	Mo %	V %	Nb %	Ti %	N ppm	O ppm	Ti* %	Nb* %
1315 (CGL 1)																		
314269	7317-K	0.003	0.137	0.006	0.005	0.004	0.05	0.012	0.018	0.024	0.001	0.0005	0.009	0.047	34	-	0.016	-0.014
314271	7317-K	0.0029	0.092	0.006	0.006	0.003	0.03	0.011	0.016	0.028	0.001	0.0004	0.008	0.042	41	33	0.007	-0.014
314278	7317-K	0.0028	0.089	0.009	0.006	0.004	0.035	0.016	0.01	0.025	0.001	0.0004	0.008	0.05	48	27	0.008	-0.014
314272	7317-K	0.0027	0.118	0.005	0.008	0.003	0.042	0.013	0.014	0.025	0.002	0.0005	0.008	0.05	39	34	0.011	-0.013
314274	7317-K	0.0028	0.108	0.007	0.006	0.004	0.037	0.015	0.013	0.026	0.001	0.0008	0.008	0.046	40	33	0.012	-0.014
314272	7317-K	0.0027	0.117	0.005	0.008	0.003	0.042	0.012	0.014	0.024	0.002	0.0005	0.008	0.044	39	34	0.008	-0.013
314279	7317-K	0.003	0.088	0.005	0.006	0.003	0.025	0.013	0.01	0.022	0.001	0.0006	0.008	0.044	44	36	0.008	-0.015
7318 (CAL 2)																		
333997	7318-K	0.003	0.051	0.006	0.005	0.006	0.032	0.013	0.011	0.018	0.005	0.0014	0.011	0.029	38	15	-0.003	-0.012
333998	7318-K	0.0029	0.073	0.006	0.005	0.005	0.036	0.011	0.01	0.017	0.0009	0.0002	0.001	0.029	40	23	-0.004	-0.021
333999	7318-K	0.003	0.086	0.005	0.003	0.005	0.028	0.01	0.01	0.015	0.0013	0.0007	0.01	0.031	40	-	0.001	-0.013
334034	7318-K	0.0025	0.077	0.005	0.005	0.004	0.038	0.012	0.009	0.017	0.0016	0.0007	0.009	0.029	40	27	-0.002	-0.010
334036	7318-K	0.0029	0.061	0.004	0.005	0.004	0.035	0.012	0.008	0.011	0.0017	0.0003	0.013	0.028	36	26	-0.003	-0.009
7118 (CAL 1)																		
334040	7118-K	0.0028	0.102	0.008	0.007	0.004	0.045	0.012	0.01	0.016	0.0017	0.0012	-	0.066	36	18	0.032	-
334041	7118-K	0.0029	0.102	0.006	0.007	0.004	0.033	0.013	0.01	0.011	0.002	0.0014	-	0.064	40	22	0.028	-
334042	7118-K	0.003	0.07	0.006	0.007	0.004	0.033	0.01	0.008	0.007	0.0016	0.0011	-	0.066	32	15	0.033	-
334044	7118-K	0.006	0.107	0.007	0.005	0.004	0.041	0.015	0.009	0.017	0.019	0.011	-	0.073	39	36	0.042	-
334045	7118-K	0.0029	0.109	0.007	0.008	0.003	0.031	0.01	0.01	0.02	0.002	0.001	-	0.065	38	33	0.028	-

APPENDIX D: CHEMICAL COMPOSITIONS

Table D.2:Chemical compositions of galvanized and modified Ti-Nb IF, modified Ti-Nb IF, and Ti IF steel samples

Döküm No	Döküm Kalitesi	C %	Mn %	P %	S %	Si %	Al %	Cu %	Cr %	Ni %	Mo %	V %	Nb %	Ti %	N ppm	O ppm
CGL	**1315 (CGL 2)**															
434164	7315-K	0.0030	0.1452	0.0087	0.0073	0.0058	0.0475	0.0236	0.0181	0.0225	0.0029	0.0015	0.0139	0.0366	40	37
414551	7315-K	0.0029	0.1176	0.0094	0.0066	0.0038	0.0390	0.0186	0.0135	0.0150	0.0017	0.0006	0.0133	0.0370	47	36
414880	7315-K	0.0025	0.1132	0.0073	0.0080	0.0035	0.0428	0.0219	0.0162	0.0113	0.0020	0.0007	0.0136	0.0362	37	44
CAL	**7316 (CAL 3)**															
418444	7316-K	0.0027	0.089	0.0056	0.0057	0.0041	0.0390	0.0238	0.0131	0.0227	0.0017	0.0008	0.0136	0.0390	40	36
418488	7316-K	0.0027	0.0857	0.0063	0.0043	0.0044	0.0391	0.0204	0.0094	0.0213	0.0011	0.0008	0.0134	0.0393	38	41
418489	7316-K	0.0020	0.0862	0.0065	0.0038	0.0042	0.0440	0.0187	0.0100	0.0210	0.0012	0.0009	0.0153	0.0402	39	-
	7118 (CAL 1)															
434122	7116-K	0.0030	0.1163	0.0060	0.0059	0.0056	0.0600	0.0204	0.0130	0.0209	0.0024	0.0007	0	0.0578	40	0
434124	7116-K	0.0026	0.1157	0.0062	0.0071	0.0041	0.0600	0.0298	0.0129	0.0349	0.0029	0.0012	0	0.0678	40	0
434129	7116-K	0.0026	0.1180	0.0070	0.0048	0.0034	0.0399	0.0148	0.0126	0.0186	0.0026	0.0016	0	0.0622	36	30

AUTOBIOGRAPHY

Dursun Ali YAŞACAN was born in Rize / Pazar at 1966. He had finished Polatlı High School at 1983. He graduated from M.E.T.U Metallurgical Engineering department at 1988 and then finished his master thesis program at the same department at 1992. He started to work Ereğli Iron and Steel Works Co. (ERDEMIR) at 1989. was working as a manager in ERDEMIR Cold Mill No 2 until march of 2006. And he is now working at Borusan Group in Turkey. He has some articles about Total Productive Maintenance (TPM), Hydrogen Production, Preventive Maintenance and Automotive Steels. He is married and has two children.