THERNINECHANICAL PROCESSING FOR TAILORING PROPERTIES IN STEELS

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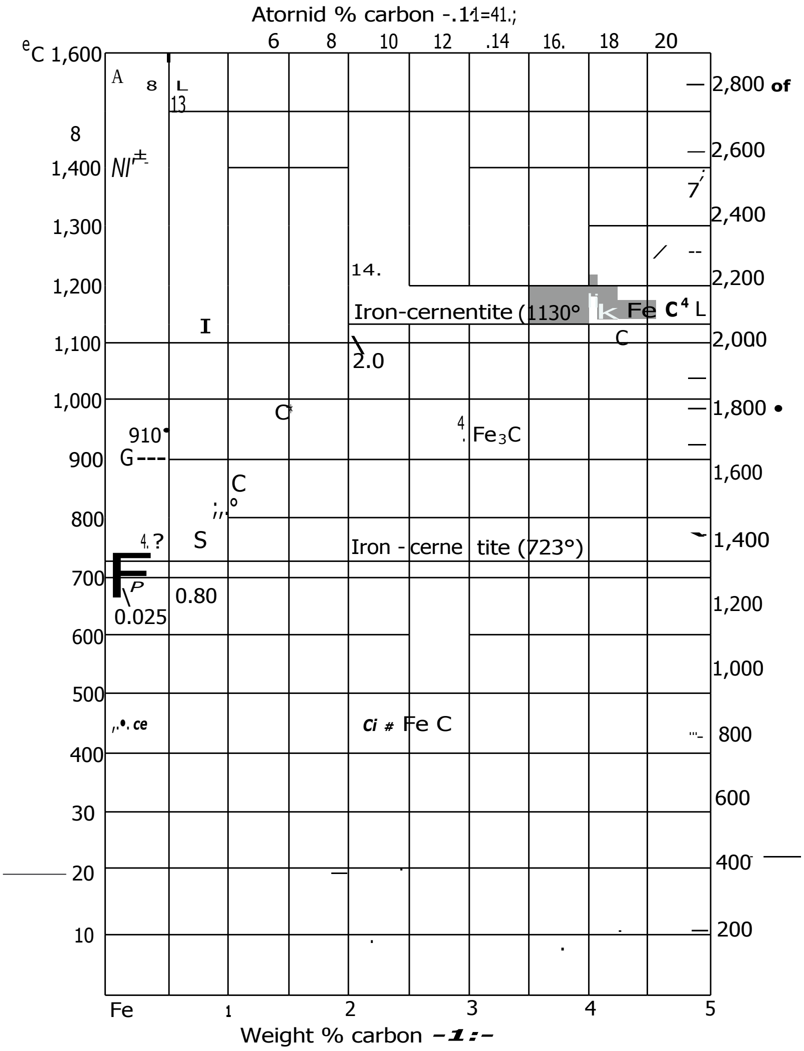
PRELUDE

A look into the simple iron-carbon equilibrium diagram (Fig.1) which is a graphical representation of different phases as a function of temperature and composition shows that a variety of steels with any desired combination of phases can be produced"by employing suitable techniques of heat treatments, mechanical processing or a combination of these. Thus, this diagram acts as a tool in producing the required degree of phases like ferrite, austenite cementite etc in the microstructure to give the expected properties. Three horizontal lines in the diagram represents the three invariant transformations signifying the co-existence of phases in equilibrium at fixed temperature, of which the lower one is of our concern.

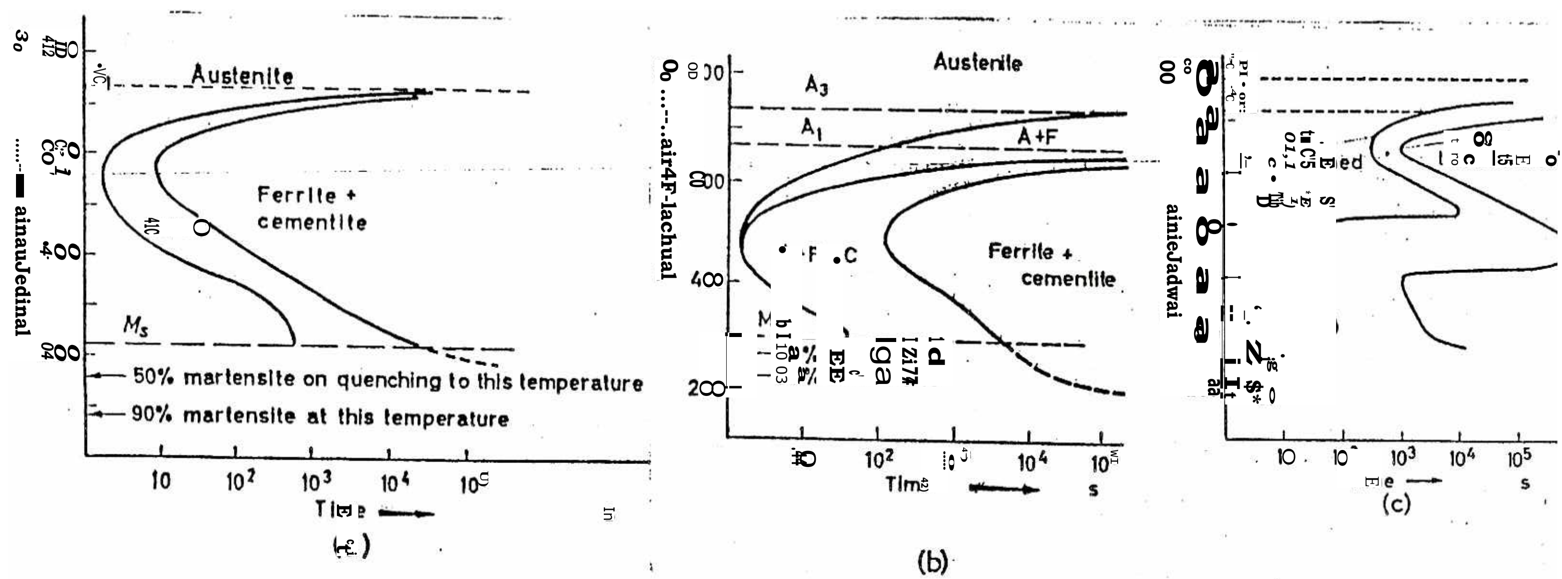
Another tool of great significance is the isothermal transformation curves (Fig.2) (Time, Temperature, Transformation curves) which produces the transformed phases like ferrite, pearlite, and bainite etc when steel is controlled cooled or held at a fixed temperature (isothermal holding) till the desired phases in the proper proportions are produced by diffusional nucleation and growth process, whereas martensite is formed by diffusionless shear mechanism. Needless to mention that an intentional addition of alloying elements (>3%) to the Fe -C system bring about marked changes in the formation, morphology and distribution of different phases. These alloying elements can form solid solution, dissolve in the carbides or form new intermetallic compounds giving a variety of properties by shifting of the I Ti' curve, raising or lowering the horizontal critical lines temperature or separating the pearlite and bainite nose in curve.



(b)

Equilibrium diagram for iron-carbon systems (from Mehl and Hagcl, *Progress in Metal Physics* 6, by courtesy of Pergamon Press)

FIG.1.



. TTT curves for (a) eutectoid, (b) pro-eutectoid, and (c) low alloy (e.g. Ni/Cr/Mo) steels (after *Metals Handbook*)

FIG. 2

Traditionally, steel is hot-deformed above the stable austenitic region by one of the many hot-working or shaping operations, to reduce steel to some required useful shape and size. Separate heat-treatment is also conventionally practiced for properties improvement through microstructural changes.

The thermomechanical processing (TMP) in the real sense started after the catastrophic brittle failure of welded structures of the Liberty ship in deep sea during World War II. This prompted several investigators to introduce Al-killed steel having fine grained microstructure. Later during the sixties at BISRA in England, pearlite free steel, pearlite reduced steel and ultimately microalloyed steel were developed which possessed high strength and ductility coupled with good weldability which was lacking in the earlier materials. During the Seventies high strength low alloyed steels (HSLA) containing micro additions and controlled rolling opened up a new area in TMP and started appearing as a technologically viable and economically suitable material processing. The process has been further developed after the introduction of on-line accelerated cooling following controlled rolling and this has revolutionised TMP in mass production. TMP has become versatile and is inevitable not only to plate but also to strip, bar and section mills.

THERMOVECHANICAL TREATMENT (TMT)

TMT can be defined as a process which combines mechanical working and heat-treatment at high temperature and simultaneous insitu phase transformation with the object of achieving fine, refined micro-structure which enable to impart unique combination of enhanced strength, toughness. This may involves:

(a) Deformation of austenite prior to its transformation,
(b) Deformation during transformation of austenite, and
(c) Deformation on fully transformed austenite.

The Schematic diagrams are shown in Fig.3.

Deformation of Austenite Prior to Transformation -- High Temperature Thermomechanical Treatment (HINT)

In the high temperature 'IMP, the steel is deformed in the recrystallisation temperature range above Ar_3 to ensure that the complete grain refinement of austenite occurs before subsequent transformation to ferrite, pearlite, bainite or martensite. This technique can be applied to low carbon steel and C-Mn steels. A typical flow-sheet of_ the HTMT is given below: Austenitisation at 1150°C to 1200°C

Cooling to a temperature just above Ar

Hot deformation of nearly reduction 30% at this temperature

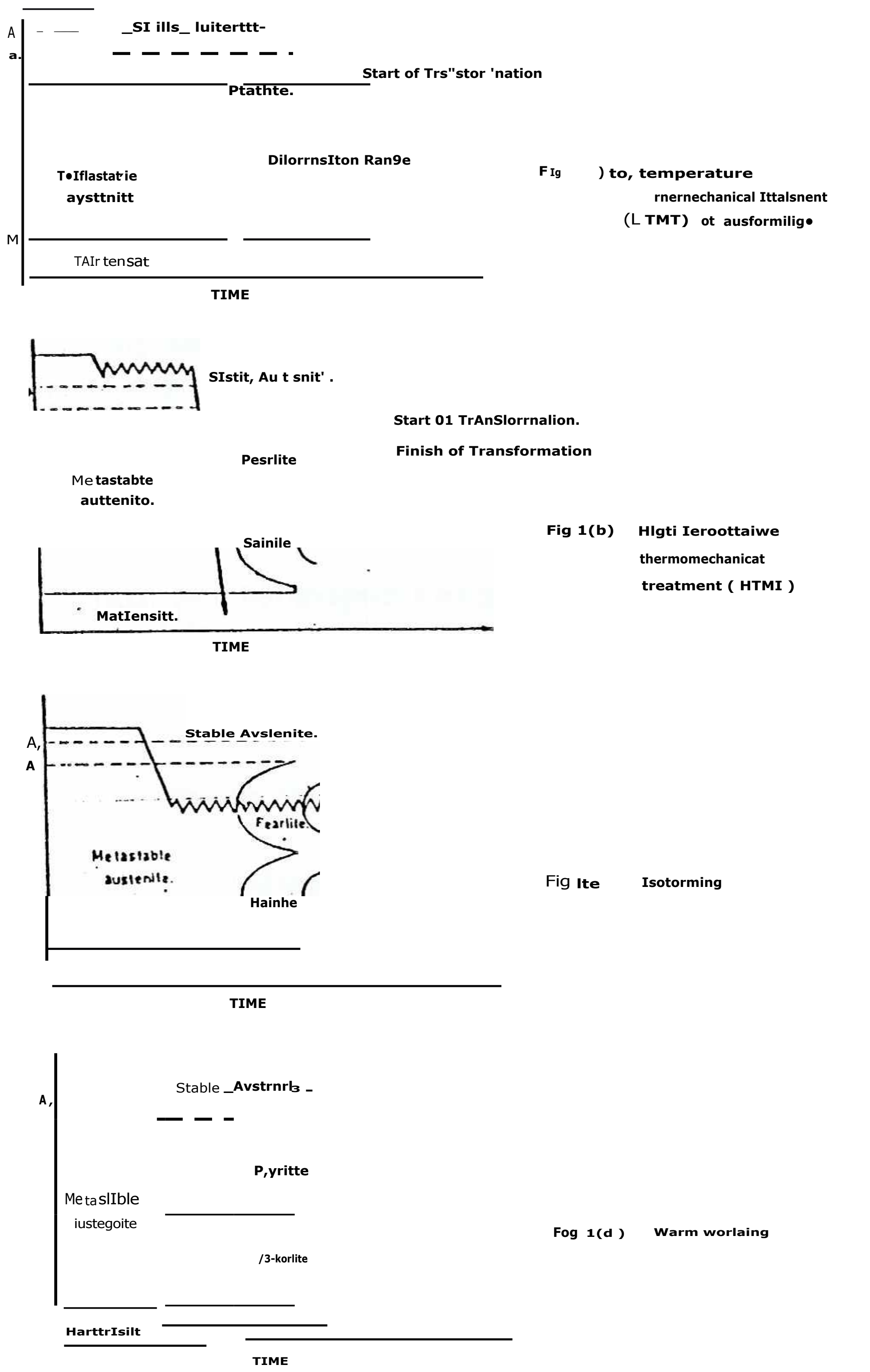
Rapid quenching in oil or water

Low temperature tempering

The improvement in properties in this process is due to the fine martensite formed through the transformation of elongated, strain hardened austenite grains and form fine distribution of carbide on tempering.

Deformation During Transformation -- Low Temprature lhermomechanical Tratment (LTMT)

In LTMT the deformation is given below the critical temperature (Ar) in the metastable austenite region and within the incubation period before transformation in the deep knee of the TiT curve. This process is known as Ausforming and is a combination of deformation and



Schematic t•)rr•iltatiiiii of thenimmechatlical ttratment in relation to a TIT d'or3l11.

FIG.3

tempering. The processing methodology is shown in the following flow-sheet:

Heating to Ar temperature

Quenching to temperature range of 450°C to 650°C.

Deformation in excess of 70% reduction to strain hardend austenite

Quenching to produce a fully martensitic structure

Tempering at low temperature or double tempering

Ausforming is generally applied to steels containing high alloying elements (e.g 4340, H-11, UHS-126, tool steels, HSS, spring steels etc). The alloying elements shift the transformation curves to the right of the Ta diagrams increasing the deep bay of metastable austenite and even separates the pearlite and bainite curves. As such a longer time for deformation is available. Due to this reason low C and C-Mn steels are not suitable to LTMT.

Ausformed steels have high strength with good toughness, but technologically ausforming is not employed due to high power consumption, high rate of roll wear, high cost of alloying elements and high cost of double tempering.

Deformation of Fully Transformed Austenite (Isoforming)

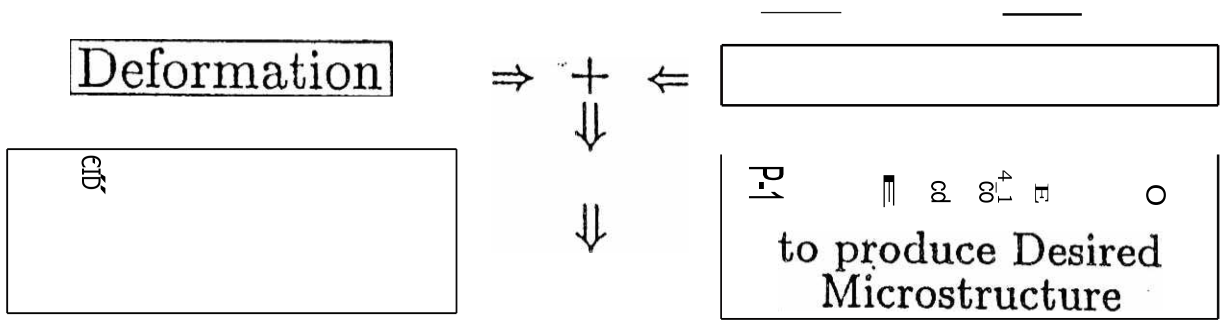
In case of isoforming treatment continuous deformation of steel is given to the austenite in the metastable state. In this process pearlite transformation is superseded and quenched in the intermediate temperature range of 550°-750°C in the region of pearlite nose. Deformation in excess of 60% reduction during the complete transformation is allowed and air cooled. Because of the low temperature (below recrystallisation temperature) deformation, austenite gets strain hardened and fine carbides from austenite,precipitate in the ferrite matrix. This uniform dispersion of cementite in fine spheroidal form improves properties particularly in notch toughness.

When austenite is fully transformed to ferrite and pearlite as in the normalised condition, a 10% reduction by deformation in the temperature region of 400°C above MS temperature produces dynamic strain ageing. The above processing although improves the strength, but reduces the fracture toughness markedly. On the other hand deforming a fully transformed martensite at relatively high temperature but below recrystallisation temperature (Ar_1) improves the strength properties depending on "the degree of deformation and on softening effect due to short time high temperature tempering. This treatment increases the strength and toughness as compared to the conventional process of quenching and tempering.

CONTROLLED ROLLING

Controlled rolling is a special type of high temperature TMP adopted for microalloyed low carbon and carbon-manganese steel by reducing the ferrite grain size and to enhance both strength and toughness in the as hot-rolled condition. Broad Features of Thermomechanical Processing is given in Fig.4. It is well known that the strength of steel can be increased by increasing its C and Mn content but increase in C and Mn drastically reduces several other vital properties like toughness, fatigue resistance and mainly weldability. Increasee in carbon-manganese ratio influences the weldability. Moreover, high alloyed steels are costly and its fabrication is cost intensive process and difficult.

Combined strength and toughness in the steel have been achieved by reducing carbon level below 0.1 in plane carbon steel, C-Mn steel



- Reduced Process Steps
- Energy Saving
- Dynamic Use of Dislocation Substructure
- Modified Nucleation/Growth Behaviour
- Reduced Alloy Content/Carbon Equivalent
- Improved Combination of Strength/Toughness

: Broad Features of Thermomechanical Processing and by microalloying with Nb, V and Ti. In high strength and low alloy (HSLA) steel, these alloying elements form solid solution and carbides and nitrides. These carbides and nitrides, precipitate at high temperature during deformation in the austenite region. Strengthening of the microalloyed steel occurs mainly by the following mechanism:

(i) Refining of the austenite as well as ferrite grain
(ii) Solid solution strengthing
(iii) Precipitation hardening
(iv) Transformation hardening
(v) Dislocation lockup

This is clearly depicted in Fig.10. The simplified rolling schedules and that of TMT are shown in Fig.5.

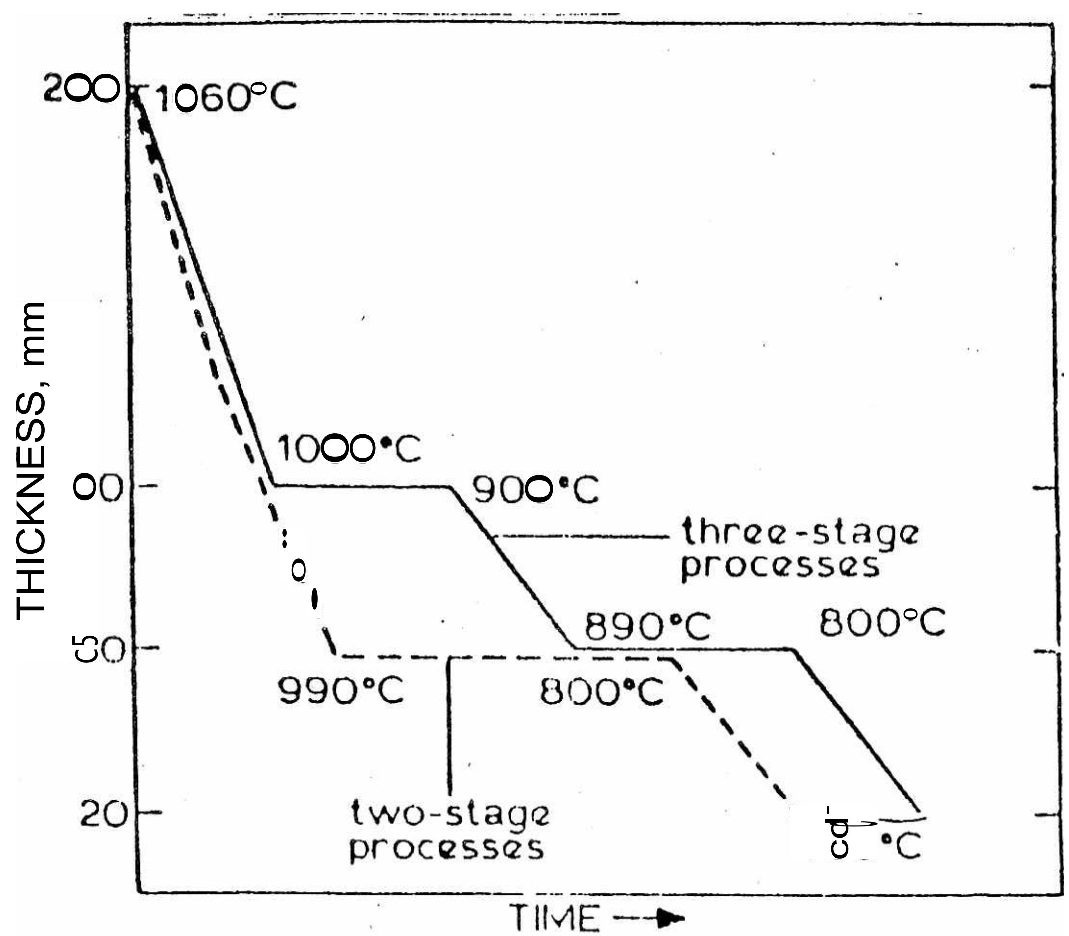
The fundamental difference between conventional hot rolling and controlled rolling lies in the fact that in the former ferrite nucleates exclusively at grain boundaries and in the latter, nucleation is made to occurs both at grain interior and grain boundaries, which lead to a great difference in the ferrite grain size between the two, process as illustrated in Figs.5(b) and 5(c).

Controlled rolling technique consists of three basic steps:

Step 1 : Deformation in the recrystallisation region at high temperature,

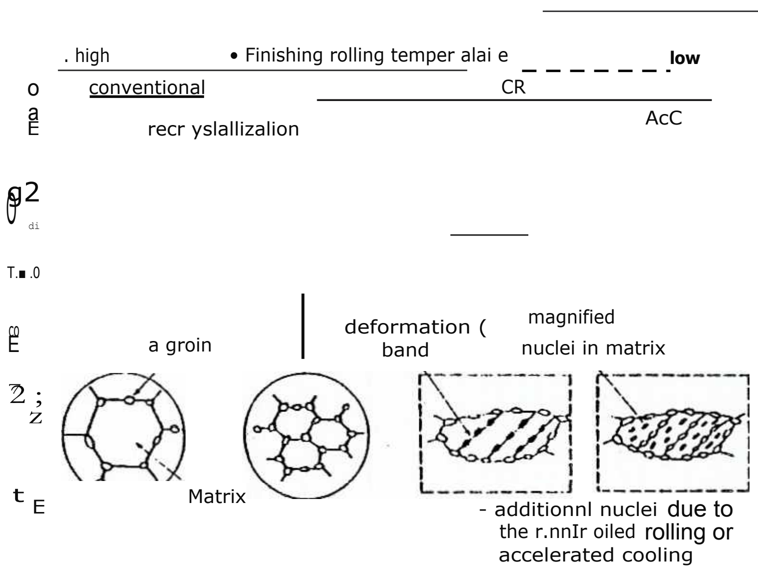
Step 2 : Deformation in the non-recrystallisation region at low temperature range, just above Ar_3, and;

Step 3 : Deformation in the austenite-ferrite two-phase region.



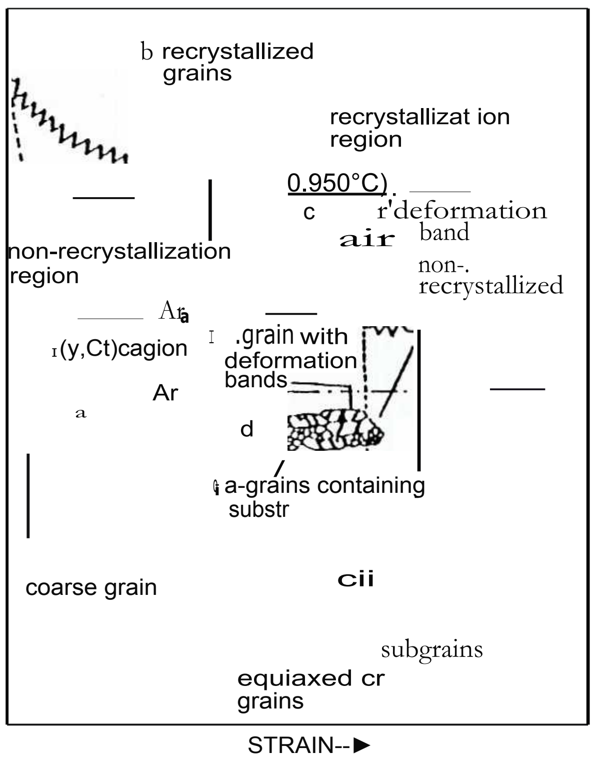
17 Simplified rolling schedules; conve al two-stage process and new three- process; after Lafrance *el al.*[129]

FIG.5a



Schematic illustration of the microstructural changes in the austenite during hot rolling, and in the partially transformed microstructure, in order to show where ferrite nuclei occur during the γ→α transformation.

FIG.5b



Schematic illustration of three stages of controlled-rolling process and change in microstructure with deformation in each stage; after Tanaka[43]

FIG.5c

Step 1 - Deformation in the recrystallisation region at high temperature

Conventional hot deformation of steel is associated with simultaneous restoration processes. In austenite the restoration is by recrystallisation but in case of ferrite it is by the recovery process. The flow curve characteristics of high temperature deformation strongly depends on temperature and strain rate. In hot deformation flow stress increases, reaches a peak and falls to a steady stress value with further increasing in strain. During this steady state condition both strain hardening and work softening occur by annihilation of dislocations.

In controlled rolling two kinds of restoration processes are associated with the hot deformation; namely, (i) dynamic recrystallisation which occur during deformation and (ii) static recrystallisation which occur during intervals of deformation and after the deformation is completed. These are shown in Fig.6. Static recrystallised grains are stable. Micro-addition of Nb in HSLA steel, mainly controlls the hot deformation characteristic in the recrystallisation region. Niobium raises Ar_3 temperature and retards recrystallisation (Fig.11) in two ways, by

(i) a solute drag effect at high temperature,
(ii) strain-induced precipitation of Nb (C/N), at low temperature.

Austenite grain size below 15-20 micron is not possible to achieve by conventional hot deformation in the recrystallisation region. This limits the ferrite grain refinement which nucleates at austenite grain boundaries. To achieve fine austenite grain, critical amount of deformation is required in multipasses.

The most important point in controlled rolling is to know in advance the recrystallisation stop temperature. This temperature is raised with the increase of solute content, mainly due to Nb, compared to V, Al and Ti. Hot deformation can reduce transformed ferrite grains to 7-10 um. Nb(C/N) particles which remains undissolved at rolling temperature enhanced austenite grain refinement as shown in Fig.8 and Fig.9.

A plane 0.07 C, 1.4 Mn steel, recrystallises down to Ax_3 temperature (790°C) but by adding 0.03 to 0.05 wt.% Nb the recrystallisation temperature is raised to 960-1000°C because of grain boundary pinning by Nb(C/N) and solute drag effect. Thus hot deformation in recrystallised region only reduces austenite grains and simultaneous refining of ferrite grains by transformation and nucleation. Hence, initial austenite grain size should be finer prior tb rolling, is the pre-condition of controlled rolling. The hot-deformation at low reheating temperature, but above Ar_3, along with higher deformation in the last pass, is an important pre-requisites.

Step 2: Deformation in the non-recrystallisation region at low temperature range above Ax_3

Deformation of previously recrystallised austenite below its recrystallisation temperature is most important in controlled rolling as it significantly increases the number of ferrite nucleation sites. The main microstructural changes are grain elongation because of low temperature deformation above Ar_3 temperature and additionally introduction of deformed bands into the grains. Band density increases with the increase in the amount of deformation.

These elongated grains as well as the deformation band act as interfaces, and therefore, the total interfacial area per unit volume increases to give rise to more potential and highly densed sites for ferrite nucleation on and arround the grains and on the banded areas. Thus the increase in the number of ferrite nucleation sites depends on the increase in cumulative rolling deformation in the non-recrystallisation region. Effect of cumulative strain (Rolling reduction) in recrystallisation region and non-recrystallisation region on austenite grain size is shown in Fig.7.

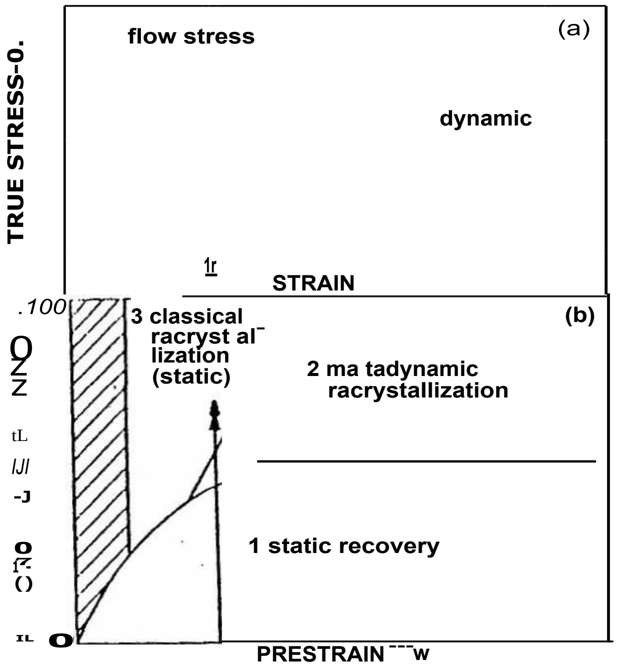
In between recrystallisation region and non-recrystallisation region, partial recrystallisation results softening effect of austenite, but work-hardening accumulates because of roll passes. Mixed grain size is produced during partial recrystallisation. Nb(C/N) precipitation retards recrystallisation but mainly by solute drag mechanism. The mixed grain structure which develops during interpass holding time must be avoided, it can not be eliminated by latter stages in rolling. Continuous rolling schedules in the whole temperature range be applied to avoid delay.

Ferrite grain is controlled by under-cooling when austenite transformed to ferrite at higher reduction. High yield strength and high value of toughness is related with fine ferrite grain size strengthened by various means.

Step 3: Deformation in austenite-ferrite two phase region

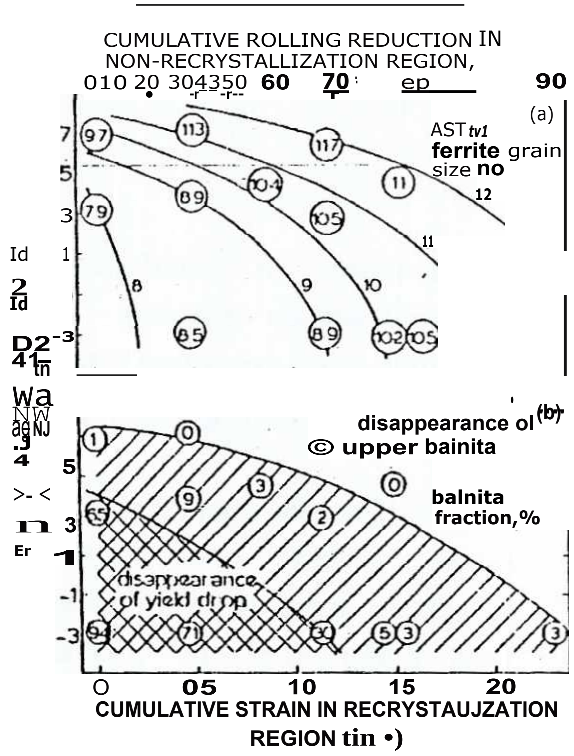
Hot deformation of 60-7070 reduction in the non-recrystallised region fragmentates the fine austenite grains achieved after deforming in the recrystallised region.

This simultaneous effect of refinement as well as deformation in the two-phase region develops deformation bands, austenite grains and sub-structures in ferrite grains. Increase in deformation, the yield strength increases, while impact transition temperature (ITT) and absorbed energy decreases. Deformation in two phase region also produces a mixed equiaxed ferrite grain and cold worked polygonal subgrains.



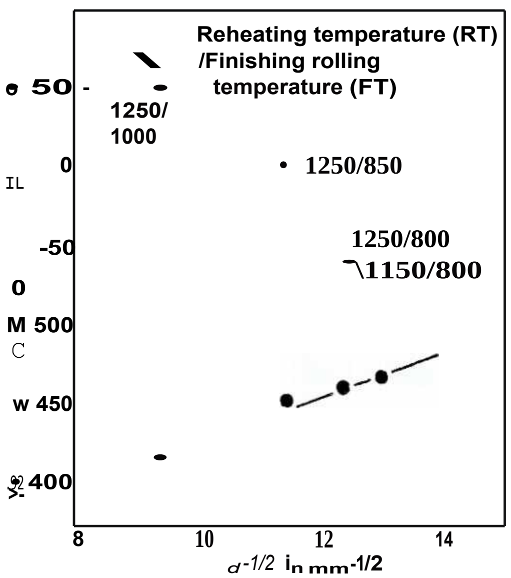
Schematic representation of a stress—strain curve and *b* of inter-relation between three softening mechanisms and of dependence on strain of softening proportions attributable to each mechanism; hatched area is forbidden zone; after Djaic and Jonas[4]

FIG.6



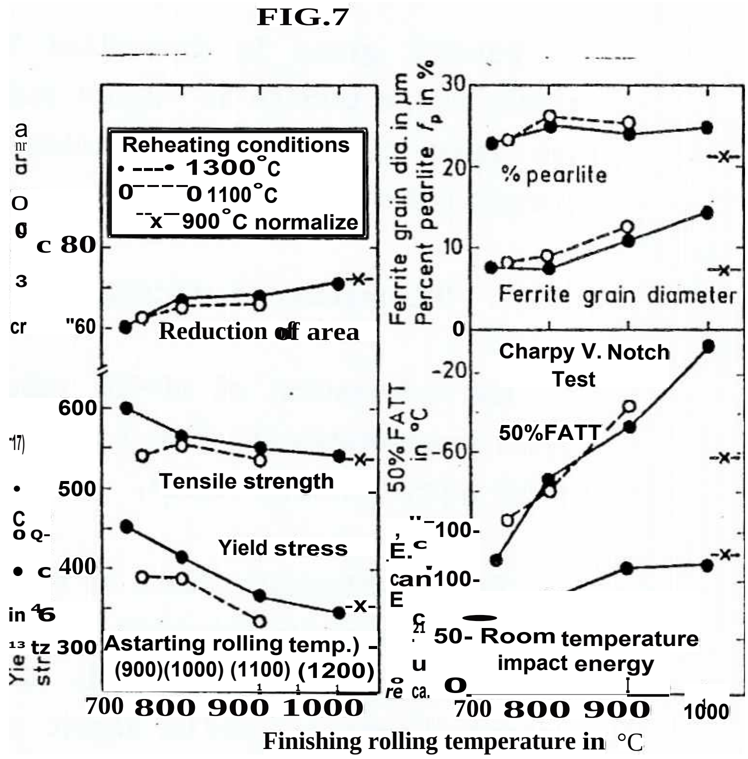
Effect of recrystallized austenite grain size and total reduction in non-recrystallization region onOferrite grain size and upper bainite fraction in Mn stee ; after Sekine and Maruyama[41]

FIG.7



,Grain size dependence of the yield stress and FAIT (Nlatsubara et al., 1972) for a 0.14 wt.% C; 1.3 wt.% Mn; 0.034 wt.% Nb sled. The steel was rolled in the laboratory from 60 mm thick to 12 mm thick at the reheating temperatures (RT) and finishing rolling temperatures (FT) indicated *(d* is the ferrite grain size in mm).

FIG.8



•. The effect of the . finishing rolling temperature on the microstructure and mechanical properties **of a 0.18 wt.% C, 1.36 wt.% Mn** steel (kozasti, **1972). The** steel was rolled in the laboratory using 9 passes each of 20% reduction over a temperature range of 200' C.

FIG.9

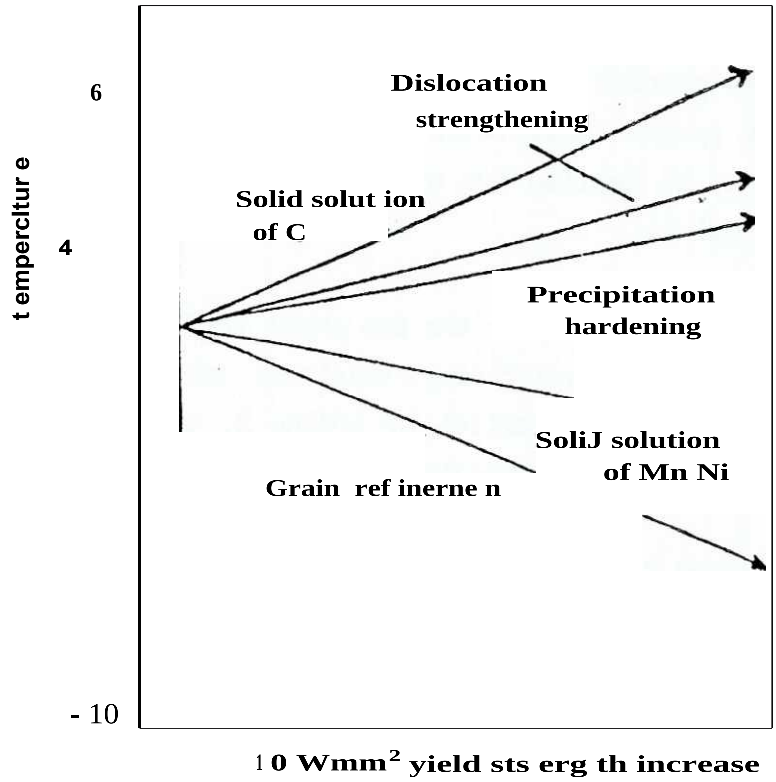
Fine ferrite transformed grains then pinned down austenite subgrains and prevent grain growth. It has been seen that deformation in austenite region produces soft grain alone, but deformation in two phase region produces a duplex structure of soft and hard substructure. This difference in microstructure between the two controlled rolling condition bring about a difference in mechanical properties between the two.

As no recrystallisation takes place in two phase region, only recovery process due to pricipitation of substructure leads to a fine ferrite grain. Controlled rolling after assessing Ar temperature precisely is carried out either in single austenite region or in two phase region.

But deformation in the two phase region produces crystallographic texture which causes through thickness embrittlement. This is avoided by inclusion shape control by adding Zr or Ca which do not flattened during rolling but spheroidised, being very hard.

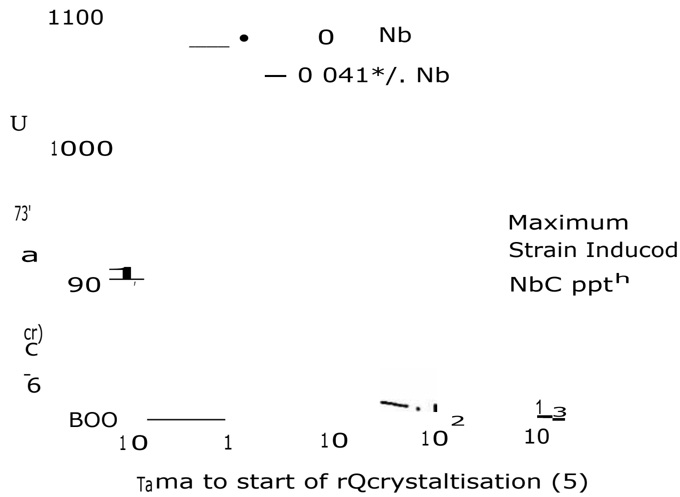
Accelerated Cooling in TIC' (TMCP)

Continuous on line accelerated cooling (Fig.12) after controlled rolling of steel has been recently adopted for further reducing the grain size and thereby increase strength and ductility. Thermomechanical Processing (TMCP) is the most modern technology developed by lowering the transformation temperature, in combination to controlled rolling and accelerated cooling. This is an accepted technology where chemistry, reheating temperature, hot rolling schedule and accelerated cooling are optimised and extensively applied for plate hot strip, rod, sections etc. Accelerated cooling is carried out at the rate of 10°C/s through transformation temperature range of 750-500°C from above Ar_3 temperature till the end of the transformation. Further air cooling to room temperature additionally



1}ifferent strengthening mechanisms and their effect on transition temperature

FIG.10



Effect of Nb on the retardation of the recrystal illation of aus teni to during hot working.

FIG.11

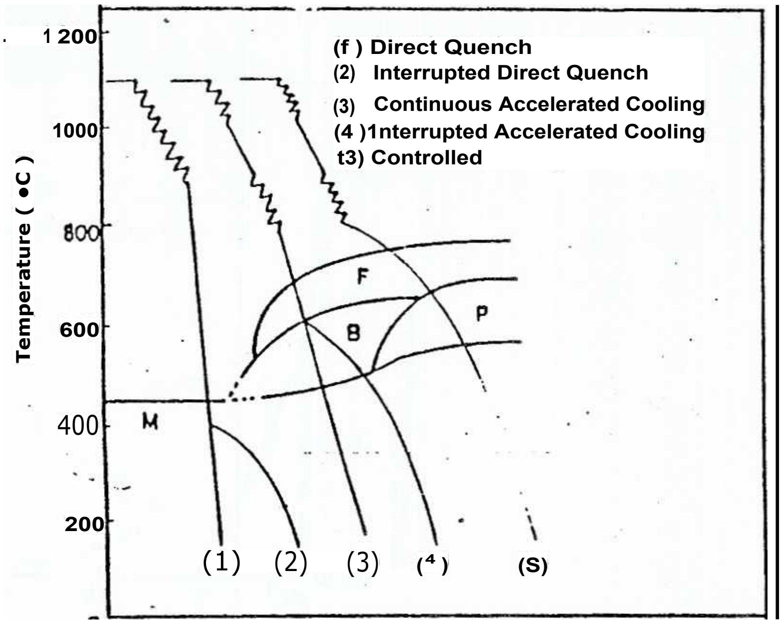
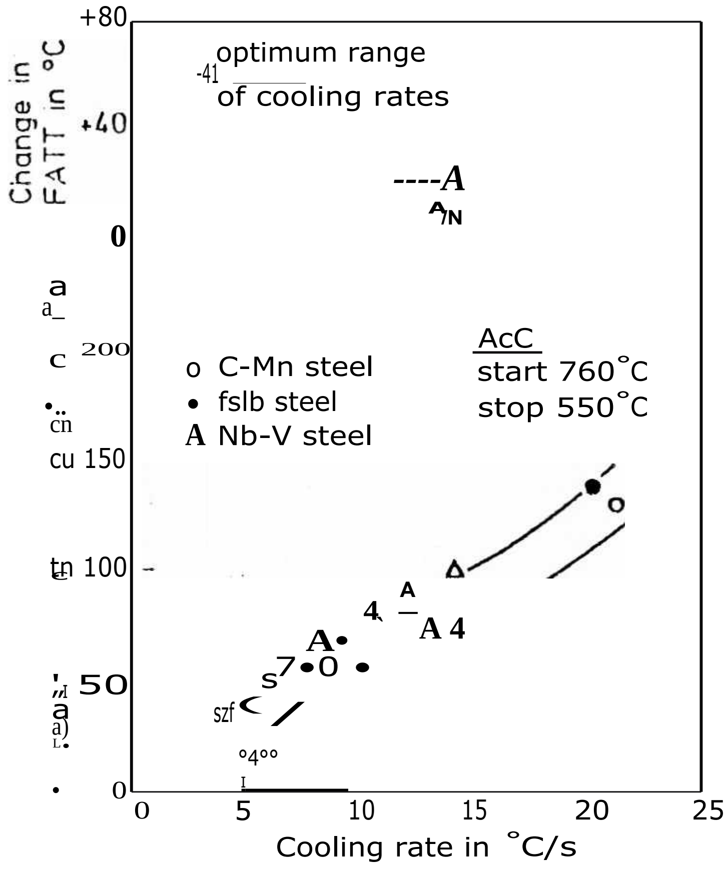


FIG.12



The effect of accelerated cooling after hot tolling on **the** tensile itength and **FATE (Tsukada** et al., 1982).

FIG.13

refines ferrite grains and depresses Ar_3 temperature. The microstructure of TMCP steel consists of fine ferrite and finely dispersed bainite which ultimately transforms to ferrite. This structural change improves strength and toughness. Further, fine dispersion of carbide/nitride enhances the strength tremendously. The effect of different accelerated cooling systems, on transformed phases as well as the property enhancement are shown in Fig.13. A typical controlled rolling and accelerated cooling in for a plate mill is shown in Table 1.

Thermomechanical Processing (D4T) for Wire Production

(a) Stelmore Process:

The rolling of rods to wire is well known industrial practice, a new technology of TMP was developed designated as Stelmore Process Fig.14. The process involves controlled cooling of hot rolled rod with particular chemistry and the processing schedule to obtain the expected final properties as given in Table 2. Thus Stelmore is basically a high temperature thermomechanical process extensively used for drawing of high carbon steels. The development of suitable cooling cycle in rod mill equipped with Stelmore cooling system can give high carbon steel wire with properly refined pearlitic structure with minimum interlamellar spacing. Reduction of area even upto 98% reduction can be achieved. The parameters of importance in Stelmore process are, laying temperature, conveyor speed and fan setting angle etc. It is possible to select the optimal thermal cycles to a particular wire rod depending on chemical composition and diameter. The key factor is to have a transformation temperature leading to shortest transformation (   ) time. The process details and properties are given in Fig.14. Conventional lead patenting produces high quality wire and is replaced by hot water quenching process 'Tempcore' shown in Fig.15.

Typical controlled rolling (CR) and accelerated cooling (AcC) conditions for a thick plate, in a 0.12 wt.% C; 1.4 wt.% Mn; 0.025 wt.% Nb steel.

Type[a]	CV	CR-I	CR-II
Reheating temp. (°C)	1200	1150 1100	1100
Slab thickness (mm)	220	[illegible]	[illegible]
No. of roughing passes	9	[illegible]	[illegible]
Temp. range of roughing (°C)	1100 - 1000	[illegible]	[illegible]
Delay time (s)	25	[illegible]	[illegible]
Finishing start temp. (°C)	1020 1000	[illegible]	[illegible]
Finishing start thickness (mm) (Total CR reduction %)[b]	100	[illegible]	[illegible]
No. of finishing passes	6	[illegible]	[illegible]

0.8

85

120

[illegible] vy CR [illegible] ss) × 100

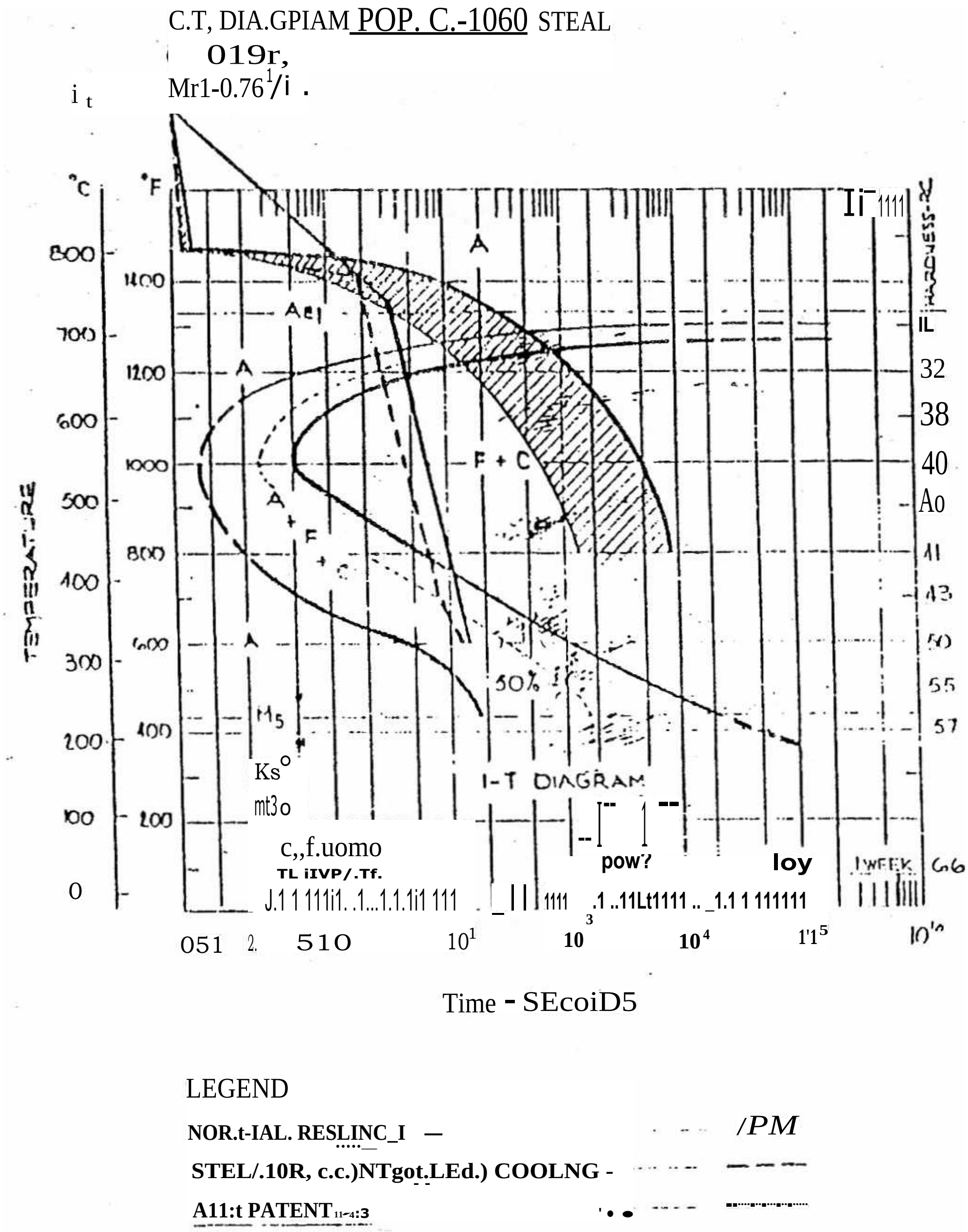

°C
°F
800
700
600
500
400
300
200
100
0
1400
1200
1000
800
600
400
200
TEMPERATURE
A
F + C
50%
I-T DIAGRAM
1 WEEK
32
38
40
41
43
50
55
57
66
Time - Seconds
LEGEND

FIG.14

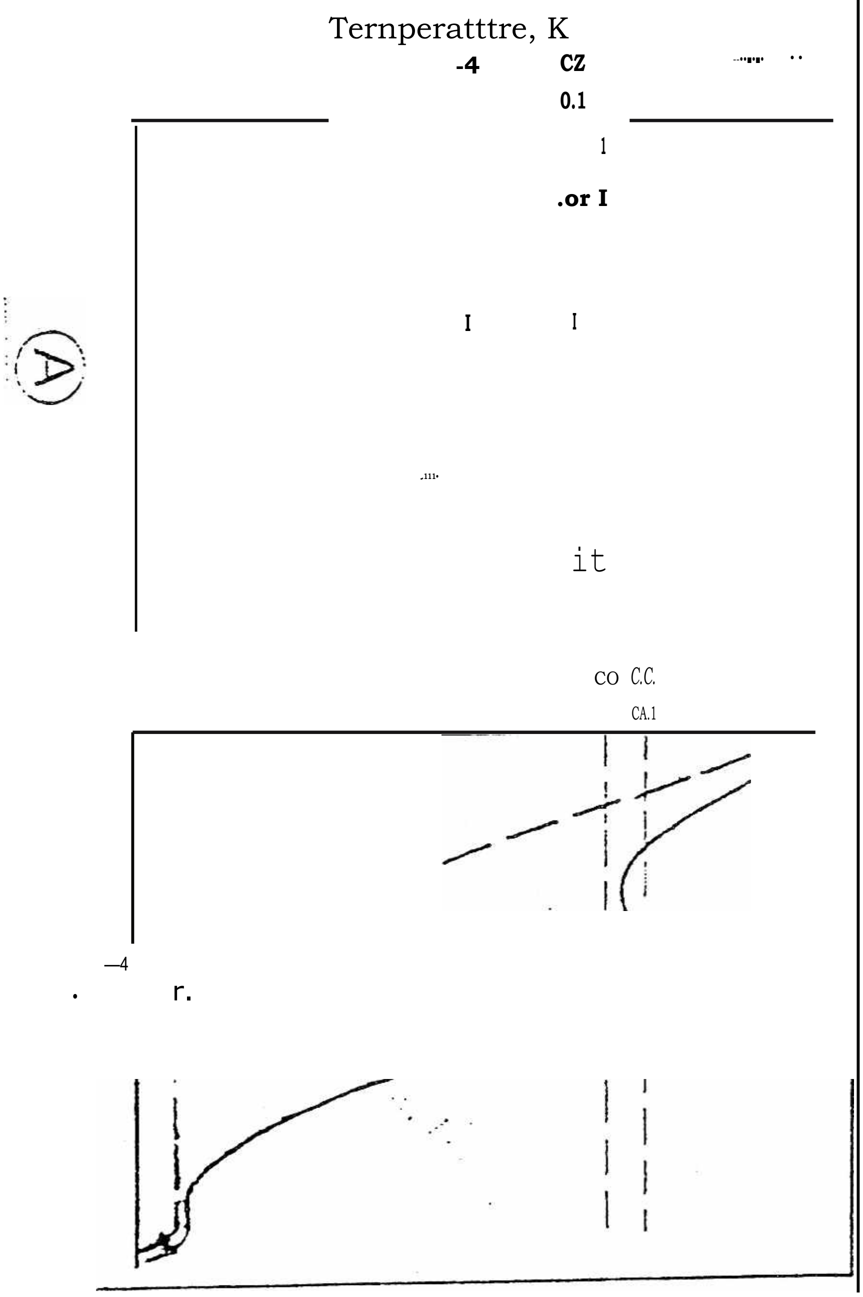
235/245

235/245

297

328

65C M.B. SPRING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(b) Tempcore Process:

A further modification, using controlled cooling after controlled rolling as described earlier, is the Tempcore process. In this process fast cooling by pressurised water as the bar leaves the last rolling stand partially transforms the surface layer to martensite, the centre remain austenite. The bar is then air cooled so that surface martensite being tempered, then on cooling bed the austenite core transform at a longer temperature to fine ferrite-pearlite structure. Thus improve strength and drawability shown in Table 3.

Effect of Alloying Elements

Alloying elements which improve strength, toughness and weldability in the microalloyed steel are most desirable. Strength and toughness is improved by grain refinement, toughning ferrite grain, by solid solution or precipitation where as weldability is improved by controlling carbon equivalent.

Solubility of Nb, V and Ti as microalloying elements in plain carbon steel and C-Mn steel is different. Nb is soluble at higher temperature than V and Ti. *As* a complete solid solution all these microalloying elements except Nb, at high temperature of reheating, hardened the austenitic grain. The undissolved microalloying elements begins to precipitates at lower temperature during deformation and recrystallisation, as carbide, nitride and carbo-nitrid at austenite grain boundaries and at dislocation sites, thereby prevent grain growth, and refining the grains. In addition, the microalloying element retard the recrystallisation (Fig.11) and raises the Pir_3 temperature by, keeping some undissolved elements for precipitation in the later stage of austenite to ferrite transformation.

In controlled rolling, micro alloying elements such as Al, N, V, Ti, Nb etc has significant influence in controlling (i) amount of

The quality statistics of single-step hot water quenched and lead patented steel wires

[illegible]	[illegible]	Heat treatment	Properties after treatment: [illegible] MN/[illegible]	[illegible]		Properties after drawing	[illegible] MN/m 2			[illegible]
[illegible]	[illegible]	0.3% solution (366K)	[illegible]		[illegible]		[illegible]	[illegible]	[illegible]	
		[illegible]	[illegible]	[illegible]	[illegible]					
[illegible]	[illegible]	0.5-0.7% solution (358K)	[illegible]	[illegible]	[illegible]		[illegible]	30-45	[illegible]	
			[illegible]	[illegible]	[illegible]			[illegible]	[illegible]	
[illegible]	[illegible]	Boiling water cooling	[illegible]		[illegible]			[illegible]		
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]		[illegible]	40.35 1	[illegible]	[illegible]
		LP	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	
		Normalising	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	
	[illegible]	355K 0.1% drawing powder solution	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]
			[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	
[illegible]	[illegible]	373K [illegible]	[illegible]	[illegible]			[illegible]			[illegible]
		[illegible]	[illegible]	[illegible]	[illegible]		[illegible]			
	[illegible]	[illegible]	[illegible]			[illegible]	[illegible]	[illegible]	[illegible]	[illegible]
		[illegible]	[illegible]		[illegible]		[illegible]	[illegible]		
	[illegible]	Boiling water cooling	[illegible]	[illegible]			I2020			[illegible]
		[illegible]	[illegible]	[illegible]			[illegible]	[illegible]		

Tensile strength, ψ: Reduction of area, δ: Elongation, η_b: Bend number,
Torsion number, **LP: Lead patenting

TABLE.3

13.27

deformation, (ii) recrystallisation temperature, (iii) grain refinement, (iv) prevent grain growth and (v) grain strengthening.

From the reheating stage upto the end of controlled cooling operation, microalloying elements has different role from solid solution hardening to precipitation after quenching and air cooling. Thus microalloying is indispensable in thermomechanical controlled processing of HSLA steel containing 0.03-0.07 Nb.

THERMOMECHANICAL PROCESSING OF ULTRA HIGH CARBON STEEL

(a) Divorce Eutectoid Transformation (NET)

This is one of the unique 'IMP employed for ultra high carbon steel (1-27 C) where heterogeneous network of proeutectoid carbides and coarse lamellar pearlite structure is modified to fine submicron size cementite in ferrite matrix through a combination of hot rolling and warm working followed by high temperature austenitising treatment and isothermal cooling Figs.16 & 17. This ultrafine duplex structure enhances room temperature strength and ductility in inherently brittle ultra high carbon steel. The steel also exhibits superplasticity at about 800°C.

(b) Divorced Eutectoid Transformation with Associated Transformation (DETWAD)

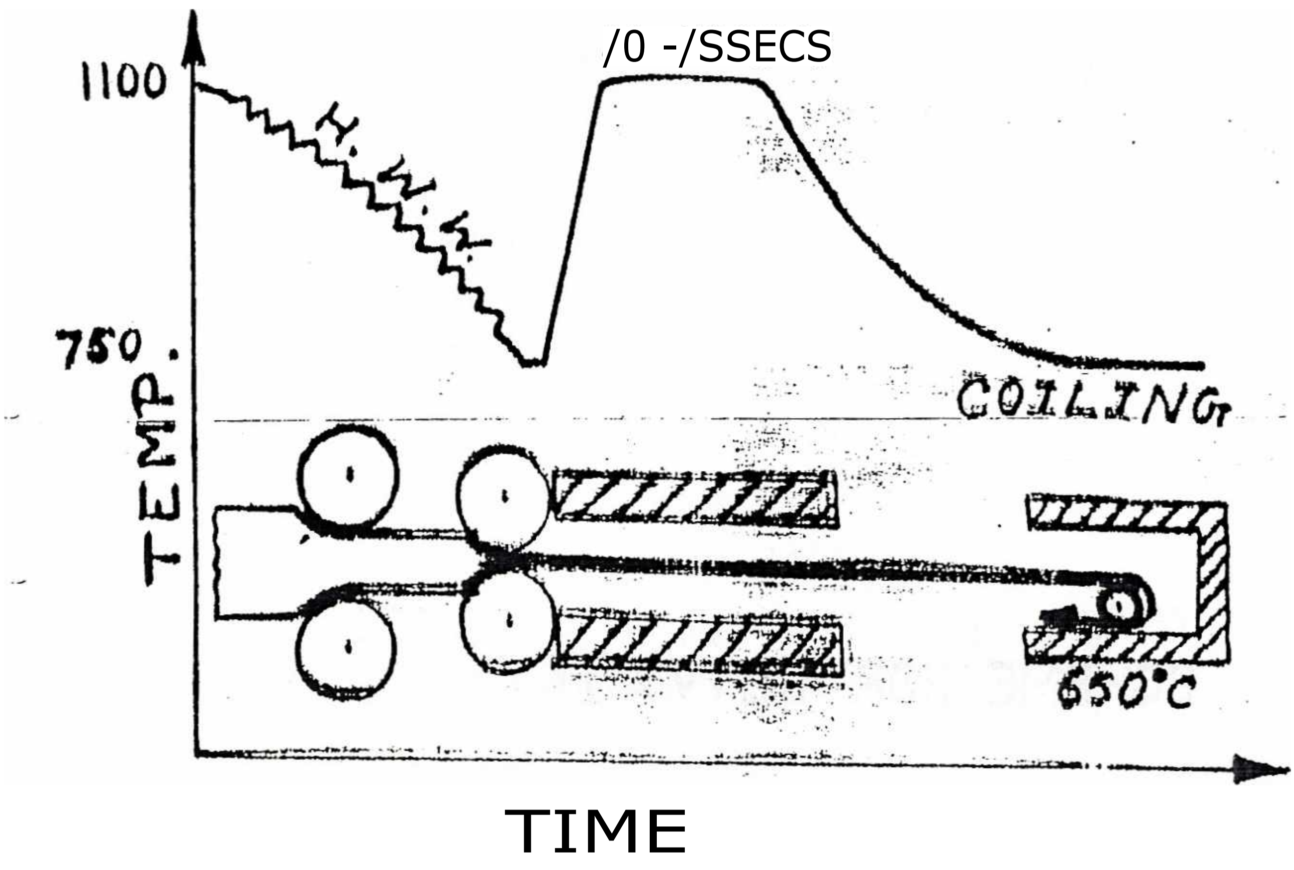
In an alternative process in continuation of the DET the steel is control cooled to recrystallisation temperature. During this period deformation of about 507 reduction in a single or multiple passes is given followed by isothermal heat-treatment. Sub micron size grains of cementite in ferrite matrix is produced with this TMP treatment. Further rolling at about recrystallisation temperature decreases preferentially the ferrite grains and improves the strength, ductility and high temperature superplasticity. The thermomechanical processing route as well as microstructural changes and properties are shown in Fig.18 and in Tables 4 & 5.

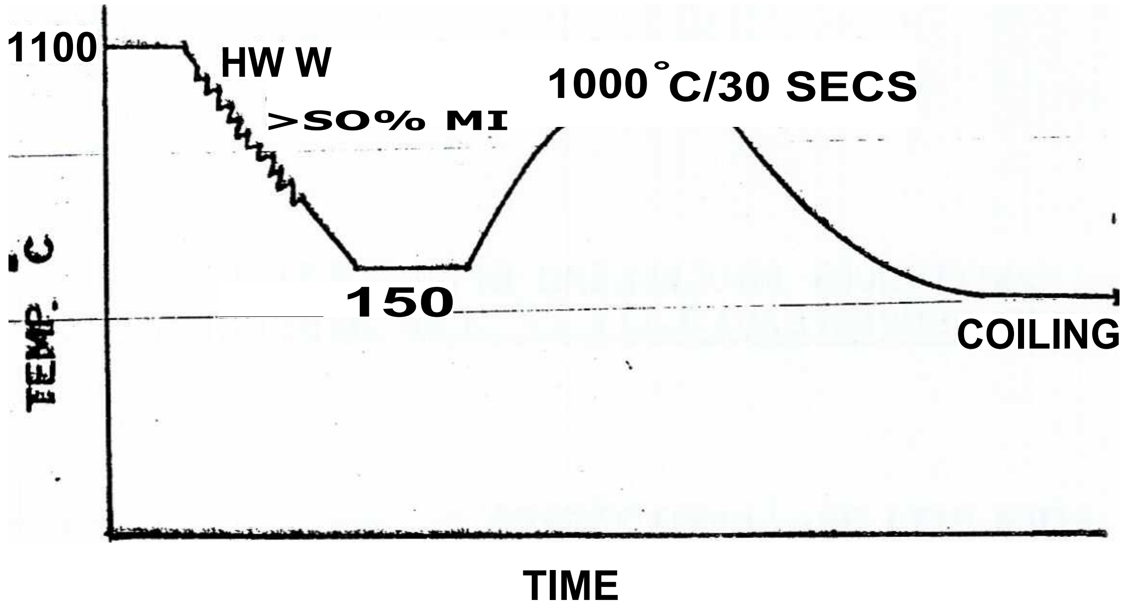
CON T IN UUUS PROCESSING O UFIG ST EEL
att„guiLiii0/14f rnin T I IICKNLS S

HOT & mum ROLL IN G U P TO 750°C)

1 INQ -AT 1100° C.; VCR 10- _ SEC . _

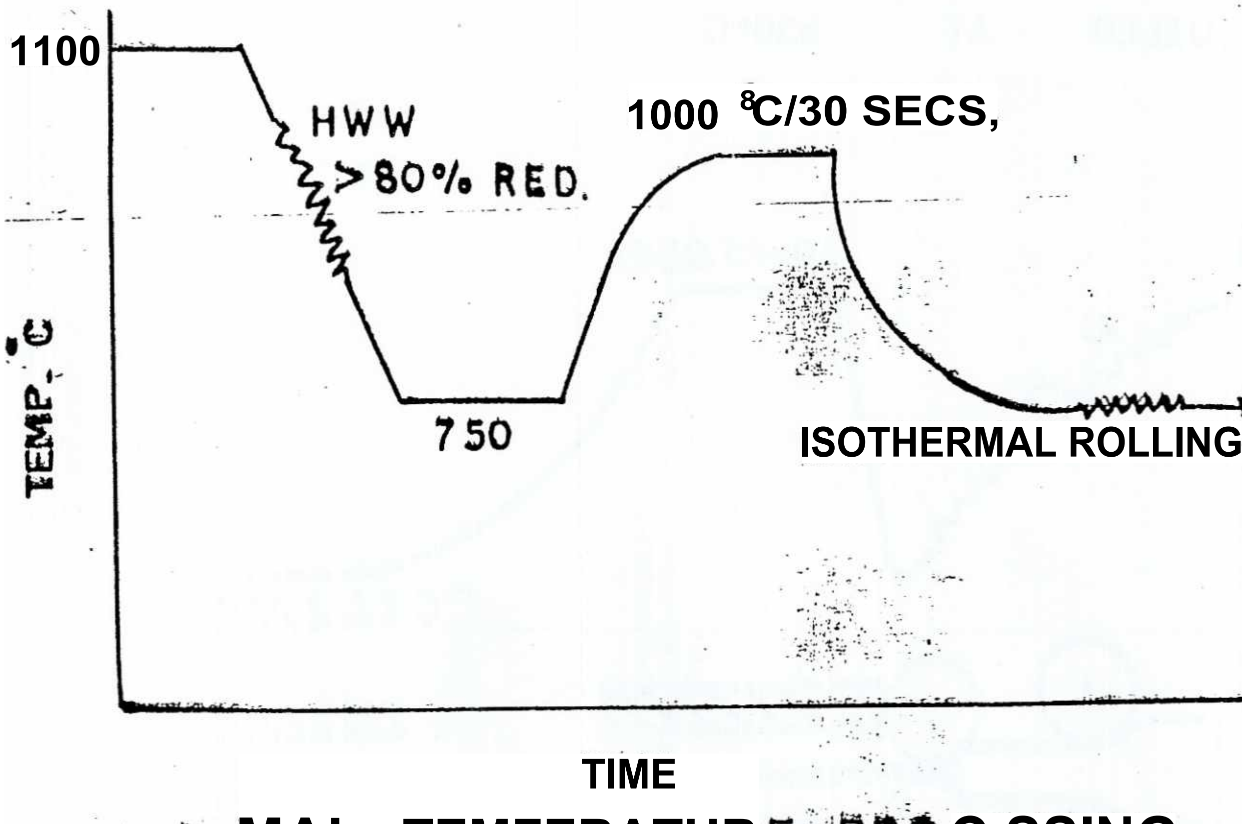
COIL EU AT 6500 C





. TIME -TEMPERATURE PROCESSING
SCHEME FOR DET.

FIG. 17



MAL ′TEMFERATURE PROC SSING
SCHEME FOR DETWAD.

TABLE,4

HIGH TEMPERATURE TENSILE TEST DATA

A = 1.20% Carbon; B = 1.45% Carbon

Sl. No.	Alloy	TMT	% E	ES^{-1}	(MPa		Strain Ratio	
1.	A	DET	400	$1.6x10^{-3}$	40		0.80	
2,	B	DET	500	$1.6x10^{-3}$	42	48	0.77	
3.	B	DET	350	$8x10^{-3}$	58			
4.	A	DETWAD (aircooled)	470	$1.6x10^{-3}$	40	100	0.78	
5.	A	DETWAD Controlled cooled	660	$1.6x10^{-3}$	29	58	0.88	0.5
6.	A	-do-	200	$1.6x10^{-1}$				
7.	A	DETWAD Isothermally rolled after DET.	640	$1.6x10^{-3}$	33		0.80	0.5
8.	B	DETWAD Controlled cooled	700	$1.6x10^{-3}$	25		0.80	0.5
9.	A	QTAD	560	$8x10^{-4}$ -	27	34	0.70	
10.	A	QTAD	150	1.6x10				

TABLE.5

Tensile Test Result nt Room Temperature

Sl.No.	Alloy	TMT	% E 25 mm G.L.	(MPa)
1.	A	DET	16	1280
2.	B	DET	4	1150

THERMOMECHANICAL PROCESSING TO CONTROL OF TEXTURE IN EXTRA-DEEP-DRAWN (EDD) STEEL.

Conventionally, aluminium-killed steel is thermomechanically treated in batch-production of deep-drawn (DD) steel sheet for automobile industries.

Though, now-a-days, on-line, continuous-annealing (CA) has been developed in advanced countries, but quality DD sheet material is, still produced from box-annealing sources. Enhancement of DD to EDD quality with higher value (lower) is only possible by controlling the texture development during cold-rolling and annealing in Al-killed steel.

Low carbon steel (< 0.1 C) with stoicheometric ratio of aluminium and nitrogen, traditionally, hot-deformed in complete austenitic region, where (A1N) remains in stable solid solution. The finishing rolling temperature should be high about 900°C.

With the aid of specially cooling system, the hot-rolled steel is then coiled at low temperature below 600°C in a quick succession.

During the finishing rolling and coiling AIN should not be allowed to precipitate at all and the hot-rolled ferrite grain should be fine (ASTM6) and equiaxed.

On further, cold-rolling, due to prefered orientation of the grain in the rolling direction, 'rolled in' texture developed, mostly in (111) plane along the sheet surface, along with (100) texture. Degree of texture formation increases with increase in reduction (60-90%).

During box-annealing of the cold-rolled steel sheet strong 'annealing texture' (111) further developed simultaniously with AIN precipitation minimising (100) texture.

The A1N precipitates retards the recovery process and during long time annealing, precipitate is completed with recrystallisation and 'pan cake' type grain growth, resulted higher value and low   value.

Generation of cumulative 'rolled in' texture by cold-rolling, and 'annealing texture' during annealing and 'texture-hardening' due to precipitation in through-thickness direction, resist the thinning effect in actual deep-drawing operation.

The most important variables in the processing are,

(i) High hot-finishing temperature,
(ii) Low coiling temperature,
(iii) Amount of cold reduction,
(iv) Rate of heating and cooling from annealing temperature,
(v) Annealing time and temperature,
(vi) A1N precipitate is prevented during coiling, precipitate, is allowed during box annealing, finally
(vii) Annealing atmosphere.

The chemical composition of aluminium killed steel, the processing route and the properties are shown flow-sheet Fig.19 and Table 6.

CONCLUSION

Various techniques of thermomechanical processing of steels have been briefly described here from actual laboratory scale experiments at NML and from open literatures available.

THERMO MECHANICAL PROCESSING OF EDD STEEL

COMPOSITION.

0604C, 02IMh, *0'03Si, 0°03AL,* tool/mit!,

pRoctssiNe.,

loci nun riNISHING. TIMP, vote

1014 COILING, TEMP. 6801

COLD ROLLING 80;4 POS.

NO- STAGE ANNEALING 100°

FIG.19

P	ANmEALING• R 70ec			AR	1- 211/200	G.S. ASTM
A						
60% CR +	*20 ↑Irs.*		1.35	+0.021	2.90	7;5
70% "			1.36	+0.011	2.58	7.5
75% "	•	N	1.63	+0.60	4.71	8.5
80% N	N	N O	1.30	+0.05 .	2.99	8.0
60% N	N	N N	1.13	+0.52	2.47	8.0
70% *	N	0 N	1.33'	+0.37	3.09	8.0
75% N	N	N O	1.25 !	+0.43	3.43	6.5
80% "	N	• N	1.64 '	+0.43	2.60 /	8.0
C						
70% CR +	*30* Prs		1.96	0.67	11.43	7.5
75%			1.66	0.56	5.94	7.0
80%	•	*	2.26	10.83	6r:-55	7.5
70% "	N	N N	1.98	0.74	8.88	7.0
75% "	N	O O	1.82 =	0.56	13.40	7.0
80% "		o N	2.211	0.64	10.27	740 *It•*

TA3LE.6

The most significant development in the field of hot-deformation of low cost microalloyed steel during last decade has been the thermo-mechanical processing with accelerated cooling. This process has been revolutioned plate, section and bar mill in mass production of structural steel with improved combination of properties.

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