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# Thermomechanically-controlled Processing for Producing Ship-building Steels

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**ABSTRACT**

The thermomechanically-controlled processing of a newly developed high-strength low-alloy steel has been designed in such a way that the problems, normally faced in producing the quench and tempered steels, have been mitigated and the final product (steel plates) are available in as rolled condition rather than quench and tempered steels.

A low-carbon, low-alloy steel having nickel, chromium, copper, niobium, boron, has been designed for ease of welding, improved weldability over the conventional steels, and responsive to the thermomechanically-controlled processing. A number of laboratory-scale batches of the alloy were made with different combinations of thermomechanically-controlled processing parameters. The different thermomechanically-controlled processing parameters studied include (i) slab-reheating temperature, (ii) deformation above recrystallisation temperature, (iii) deformation below recrystallisation temperature, and (iv) finish-rolling temperature. The thermomechanically-processed steel plates, under certain combinations of thermomechanically-controlled processing parameters, showed excellent combination of impact and tensile properties. In this paper, the microstructure-property correlation has been made to throw light on the type of microstructure required to obtain such superior package of mechanical properties. Further, the optimised laboratory-scale thermomechanically-controlled processing parameters, which were used to process newer batches of the steel made through industrial route, have delivered encouraging results.

**Keywords:** Thermomechanically-controlled processing, TMCP, HSLA steels, high-strength low-alloy steels, steel plates, quench and tempered steels, low-alloy steels, structural applications, ship-building steels, HICC, hydrogen-induced cold cracking, ULCB steels, ultra-low carbon bainitic steels, microalloyed steels

## 1. INTRODUCTION

The importance and usage of steels for structural applications continue to predominate in critical areas in spite of the availability of a new class of polymeric, ceramic, and composite material with improved properties. In many critical/strategic applications, there is an ever-increasing demand of high performance structural steels which are being imported at high cost. The shipbuilding, aerospace, military applications, nuclear power, automobiles, etc are some of the areas of applications which consume huge tonnage of these special steels. Recently, most of the developed nations seek to enforce technology control and embargo on strategic materials. In this scenario, self-reliance and competence building in the indigenous development and use of critical/

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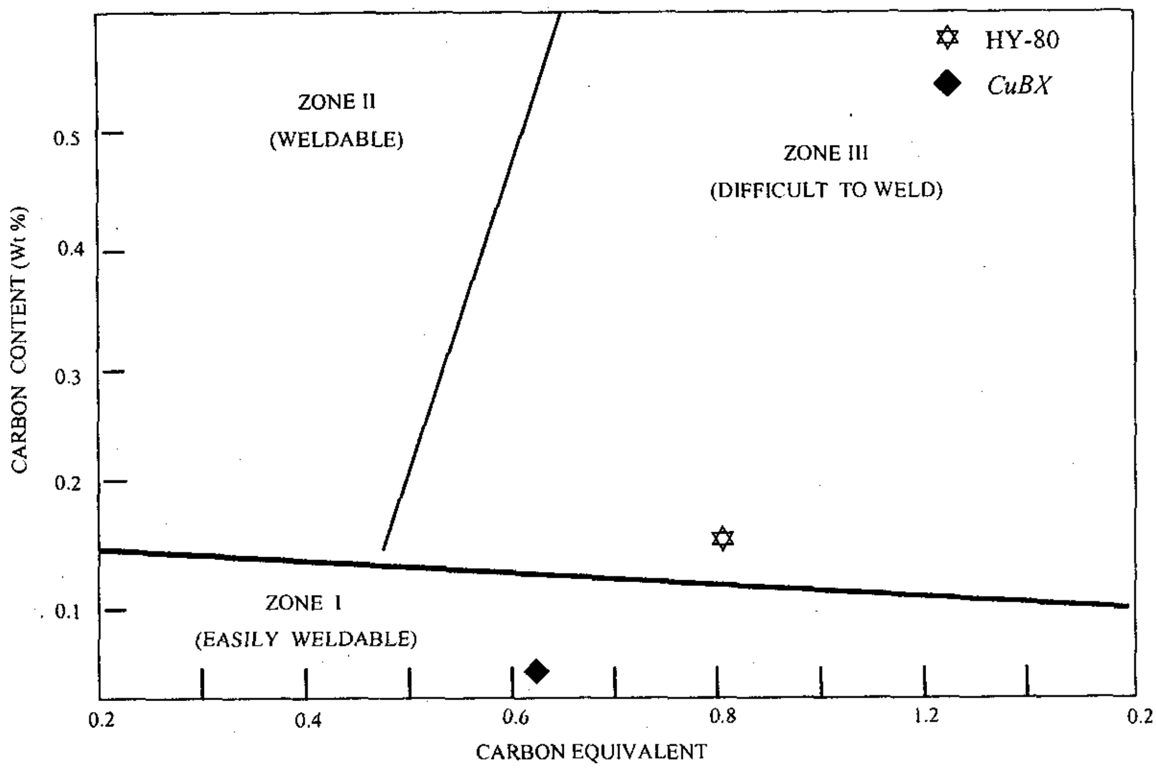
strategic materials for marine applications assume greater significance.

The structural steels used for the construction of naval surface vessels, strive for high performance properties in terms of strength, toughness, weldability, and corrosion resistance. Such requirements assume still greater importance for structural steels used for building hulls of submersibles. The structural steels used for this purpose are quench and tempered, medium carbon, low-alloy steel like HY-80, HY-100[1]. The strength is normally achieved by adding carbon and other alloying elements. The strength and toughness requirements are maintained by heat treatment process, namely hardening, followed by tempering at appropriate temperature for a required duration. Indigenous production of steels for naval application is extremely difficult in the present-day scenario due to non-availability of infrastructural facilities for rolling wide and long plates and their heat treatment by roll quenching or pressure quenching routes.

Another disadvantage of the conventional high-strength steels is the poor weldability due to higher carbon equivalent[2]. This places the steels unfavourably wrt weldability as depicted in the Graville diagram (Fig. 1). To counter this deterioration in weldability, stringent welding conditions like controlled pre-heat, post-heat, interpass temperature, heat input, etc are required to be imposed and very low diffusible hydrogen in the weld has to be maintained through proper conditioning of weld consumables. This escalates the fabrication cost of these traditional quench and tempered steels.

A continuous search for better steels with more attractive combination of mechanical properties and improved weldability has evolved new generation of steels, such as high-strength low-alloy (HSLA) steels, ultra-low carbon bainitic (ULCB) steels, and other microalloyed steels. These new generation of steels have been developed as candidates to replace the traditional HY steels, particularly in naval context. These steels achieve their best




C.E. = $C + (Mn + Si)/6 + (Cr + Mo + V)/5 + (Ni + Cu)/15$

**Figure 1. Graville diagram**

combination of properties using both the advanced alloy design concept and thermomechanical processing. This paper gives an insight into the research and development efforts to develop and produce better quality, high-strength structural steels by thermomechanically-controlled processing route, mitigating the problems faced in using the present-day quench and tempered steels for marine applications.

## 2. ALLOY DESIGN

The alloy design involved improving weldability by reducing carbon content and carbon equivalent. It was also intended to make the steel more responsive to thermomechanical processing. This was done by appropriate alteration of grain-coarsening temperature, austenite recrystallisation temperature ($T_{RXN}$), $A_{r3}$ (austenite à ferrite transformation) temperature, and shape of CCT diagram by suitable combination of microalloying additions. The approach to alloy design[4-6] is depicted in Fig. 2

## 3. THERMOMECHANICALLY-CONTROLLED PROCESSING

The thermomechanically-controlled processing produces a material with superior characteristics by controlling the deformation and the temperature of deformation during the hot-rolling processes, which were originally designed to obtain the desired external shape of the product.

This type of processing saves energy in the manufacturing of the products by minimising or altogether dispensing with the post-deformation heat treatment procedure, thus increasing the productivity. Additional cost-saving also occurs from the point of view of cost of alloying elements as this process allows leaner chemistry to be used without sacrificing mechanical properties.

The purpose of thermomechanically-controlled processing of steel is to exploit the effect of plastic deformation above and below the recrystallisation-stop temperatures on the microstructure of austenite in such a way as to develop a most favourable, fine-grained microstructure on transformation to achieve improved mechanical properties. The grain refining is a technique, which improves toughness and maintains, if not improves, the strength level of the steel. The most important parameters of thermomechanically-controlled processing are the slab-reheating temperature, deformation above $T_{RXN}$, deformation below $T_{RXN}$, finish-rolling temperature, and cooling rate.

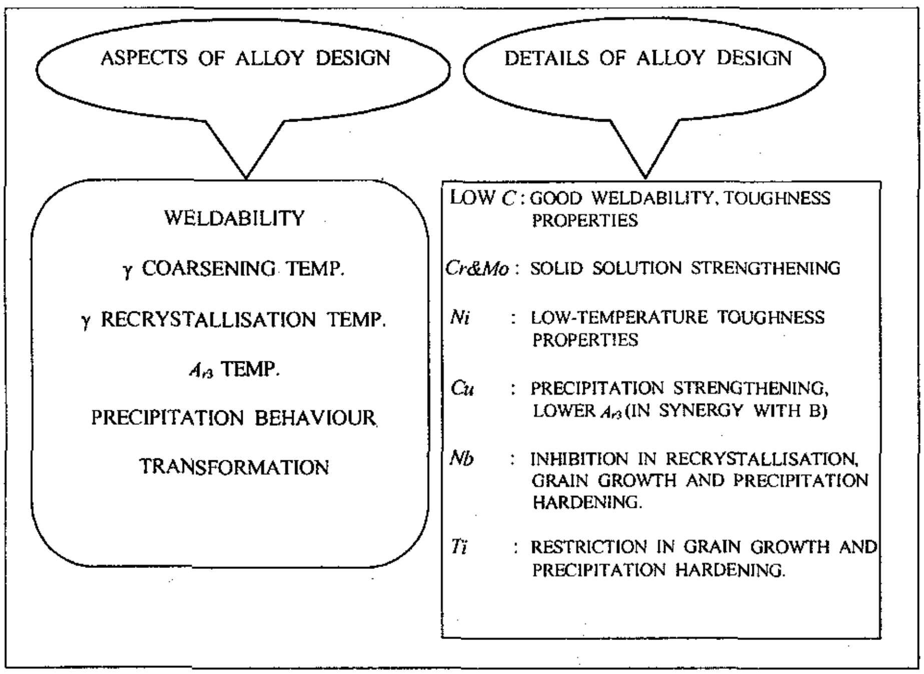



**Figure 2. Alloy design approach**

**Table 1. Chemical composition of the alloy**

| Element | Wt (%) |
|---|---|
| *C* | 0.040 |
| *Mn* | 1.500 |
| *Ni* | 1.200 |
| *Mo* | 0.300 |
| *Cu* | 1.200 |
| *Cr* | 0.500 |
| *Nb* | 0.020 |
| *B* | 0.002 |
| *Ti* | 0.020 |

## 4. EXPERIMENTAL PROCEDURE

The nominal chemical composition of the alloy (*CuBX*) used in the present investigation is given in Table 1. This steel was prepared in a vacuum induction furnace of 50 kg capacity. The ingots were homogenised and forged to an intermediate size, and finally thermomechanically-processed to 12.5 mm and 25.0 mm plates. The processing schedule is shown in Fig. 3.

The melts were cast in vacuum to cylindrical ingots of diameter approx. 150 mm and height of approx. 360 mm. The ingots were cooled down in vacuum. The ingots, after radiography and hot-top cutting for the elimination of casting defects, were further homogenised at 1100 °C. The ingots were forged at 1050–1100 °C to slabs. The slabs were subsequently thermomechanically-processed to steel plates of thickness 12.5/25.0 mm.

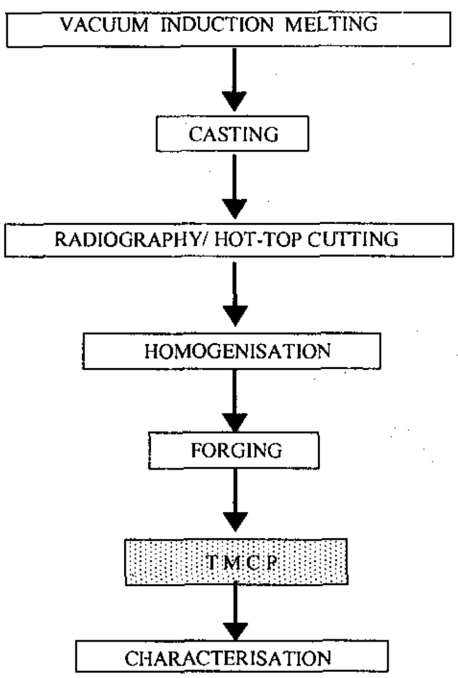



**Figure 3. Processing schedule**

A typical thermomechanically-controlled processing schedule is shown in Fig. 4. During thermomechanical processing, different parameters studied were: (i) slab-reheating temperature (SRT), (ii) percentage deformation, and (iii) finish-rolling temperature (FRT). The range of slab-reheating temperature used was 1000-1200 °C. The amount of deformation above $T_{RXN}$ was varied from 20 per cent to 40 per cent and the amount of deformation from 60 per cent to 80 per cent was given below $T_{RXN}$. The total deformation was varied from 65 per cent to 80 per cent. The finish-rolling temperature also varied within a range 700-900 °C in steps of 50-100 °C. Ten number of laboratory-scale melt and subsequent thermomechanically processing was taken up to optimise the mechanical properties.

The steel plates after thermomechanical processing were characterised in terms of tensile properties, impact toughness values, and microstructure. Impact testing was carried out mainly at –40 °C. The optical metallography was done to characterise and compare the microstructure of different plates.

Subsequently, few industrial melts (eg, 1 ton, 5 ton, and 25 ton ) were also taken and the properties were assessed. The dimensions of the plates processed through laboratory-scale were approx. 12.5 mm × 600 mm × 750 mm from a single melt. In case of industrial melts, from one melt, number of plates of following dimensions were processed:

(i) 12.5 mm × 900 mm × 1000 mm

(ii) 25.0 mm × 900 mm × 1000 mm

## 5. RESULTS & DISCUSSION

In thermomechanically-controlled processing, (Fig. 4), reheating of the slabs is primarily done to enable the slabs to be deformed plastically to

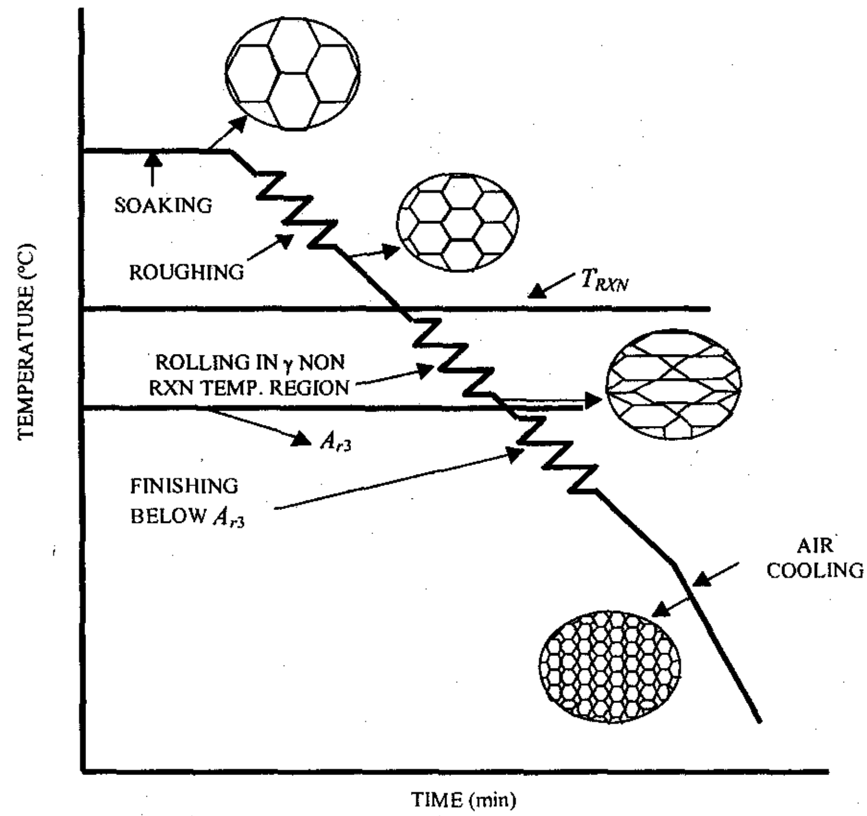



**Figure 4. Typical thermomechanically-controlled processing schedule**

predetermined thickness. Roughing refers to the first phase of plastic deformation, generally above the recrystallisation-stop temperature ($T_{RXN}$). Here, the γ is refined by repeated deformation and static recrystallisation. Further, plastic deformation can be given to the plates below $T_{RXN}$, ie, in the γ non-recrystallisation temperature region. At this stage, the austenite gets continually elongated and broken without any recrystallisation of γ occurring. The deformation at this stage substantially increases the nucleation rate at γ-grain boundaries and within the γ grains. The intragranular nucleation of α is one of the most important aspects of thermomechanically-controlled processing. The deformation of the plates may still be continued below $A_{r3}$ temperature (γ → α transformation temperature), ie, in the two-phase (γ + α) region. At this stage, the ferrite gets deformed with the formation of substructures. At the same time, γ gets further work-hardened. During cooling after deformation, austenite transforms to equiaxed ferrite grains, while deformed ferrite changes into subgrains.

In the present investigation, the thermomechanically-processed steels, under certain combinations of thermomechanically-controlled processing parameters, showed excellent combination of tensile and impact properties at the room temperature, [Tables 2(a) and 2(b)]. The impact properties of the alloy, in longitudinal direction, at subzero temperature were also very high [Table 2(b)]. The following properties obtained refer to the samples drawn from the steel plates having 12.5 mm thickness.

The microstructure of this batch of steel with such superior combination of strength and toughness showed non-equiaxed and fine-grained ferrite (Fig. 5). The grain size was uniform and varied within a narrow band. The average grain size was ~3.45 μ. These microstructural features were obtained using the following thermomechanically-controlled processing parameters (Fig. 6):

(a) Slab-reheating temperature: 1000-1050 °C for 2 h.

**Table 2(a). Tensile properties of the alloy**

| Property | Tensile properties | |
|---|---|---|
| | Longitudinal | Transverse |
| 0.2 % Yield strength (MPa) | 620 (621, 627,616,616) | 600 (622,594,596,587) |
| Tensile strength (MPa) | 756 (754,758,755,759) | 739 (744,734,738,739) |
| Elongation (%) | 24 (24, 24, 25, 23) | 26 (25, 23, 28, 28) |
| Reduction in area (%) | 67 (70, 66, 70, 62) | 73 ( 75, 69, 76,72) |

(b) Deformation above the recrystallisation temperature : 20-25 per cent

(c) Deformation below the recrystallisation temperature: 75-80 per cent

(d) Finish-rolling temperature: 775-800 °C

(e) Total deformation: 65-70 per cent.

**Table 2(b). Impact properties of the alloy**

| Temperature | Impact toughness (J) |
|---|---|
| Room temperature | 272 (278, 270, 268) |
| – 40 °C | 265 (256, 270, 269 |
| – 92 °C | 87 ( 100, 100, 60 ) |

### 5.1 Slab-reheating Temperature

A low slab-reheating temperature of 1000-1050 °C was selected due to the following reasons:

- Controlled dissolution of carbides/ carbonitrides[7]
- Restriction of γ-grain growth by the undissolved carbides/carbonitrides, and
- Uniform fine γ-grain prior to roughening, which strongly influenced the formation of a uniform grain size distribution in the final microstructure.

The complete dissolution of the precipitates (with higher finish-rolling temperature) resulted in the increased precipitation during rolling and subsequent cooling. It although resulted in higher strength due to increased precipitation, but was accompanied by a loss in toughness of the alloy.

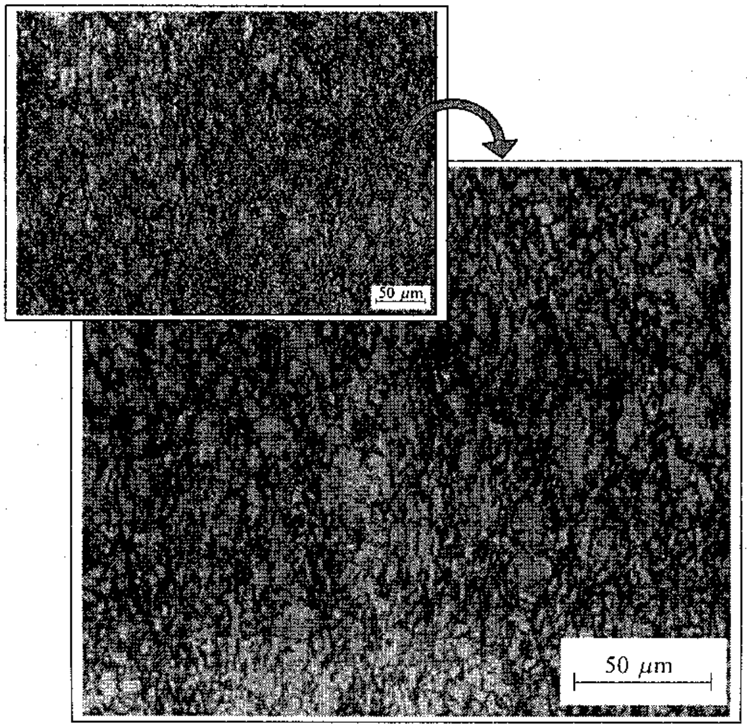
### 5.2 Deformation above Recrystallisation Temperature

A 20-25 per cent deformation (in two passes) above $T_{RXN}$ was found to be optimum from the point of view of:

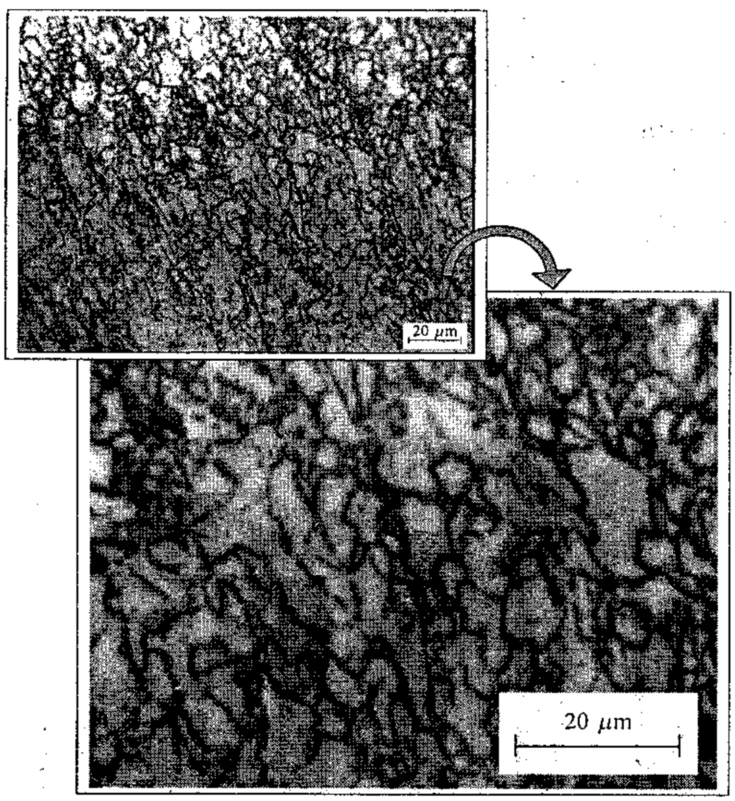
- A deformation more than 20-25 per cent wa: observed not to contribute further refinemen of γ-grains beyond a certain limit[8] of 10-15μ
- Since the slab-reheating temperature is low the temperature regime available for deforming γ in recrystallisation temperature regime i: narrow (~50-100 °C). Two passes, each o 15-16 per cent deformation, and totallin 20-25 per cent was practically possible to b applied within this temperature regime.
- Moreover, it was also intended to have highe percentage deformation below $T_{RXN}$ to achiev the maximum benefits of deformation belo $T_{RXN}$.

### 5.3 Deformation below Recrystallisation Temperature

Continued deformation of the fine recrystallise γ-grains below the recrystallisation temperature leac to pancaking of the γ-grains. The γ deformed belo $T_{RXN}$ had various defect structures, such as elongate austenite grain boundaries, twin boundaries, ar highly dislocated region. The more the γ was flattene the finer the grain size formed after the γ to transformation, as the defect structures, which a associated with the flattened γ, provides nucleatir site for the α formation. The alloy was design in such a way that it was possible to obtain a wid γ non-recrystallisation temperature[9] region (~2( °C) and it was possible to give a deformation about 75-80 per cent below $T_{RXN}$, in four passe The copper has less effect on the reduction on t γ to α transformation temperature. At the sar time, boron is also effective in preventing ferr formation. The reduction of γ to α transformati temperature has been observed to be more pronounc




(a)




(b)

**Figure 5. Microstructure of sample showing toughness ~265 J at – 40 °C**

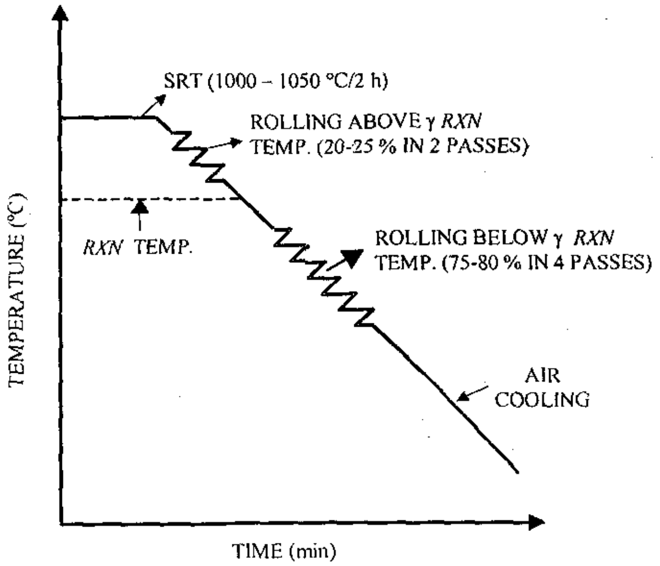



**Figure 6. The optimised thermomechanically-controlled processing schedule.**

when both copper and boron are present, ie, a synergism exists between copper and boron. It is reported that the above transformation temperature is lowered by nearly 110 °C with furnace cooling when ~1.8 per cent copper is added to boron-steels[9].

### 5.4 Finish-rolling Temperature

A finish-rolling temperature of 775-800 °C was found to be optimum as

- A lower finish-rolling temperature (~700 °C) resulted in the formation of mixture of acicular ferrite/non-equiaxed ferrite, and bainite.
- A higher finish-rolling temperature (900-850 °C) resulted in coarse bainitic microstructure with mixed grain sizes.
- An intermediate finish-rolling temperature (800-775 °C) with a heavy deformation in the last pass resulted in fine-grained, non-equiaxed ferritic microstructure with a uniform grain size distribution. The formation of such a microstructure may be a result of increased precipitation of *Nb(CN)/NbC*, acceleration of $\gamma \rightarrow \alpha$ transformation during and/or transformation.

A combination of superior strength and superior toughness properties, in a high-strength low-alloy steel, depends primarily on its final microstructure. Refinement of grain size results in improvement in both strength and toughness[4,5]. A fine ferrite grain constitutes a smaller dislocation path than a coarse ferrite grain. This exerts a positive effect on both the strength and toughness. The presence of a very fine grain size (average 3.45 $\mu$), Fig. 5, in the steel is indicative of both high-strength and high toughness.

Concurrently, the shape of the grains also plays a dominating role on the final properties of the

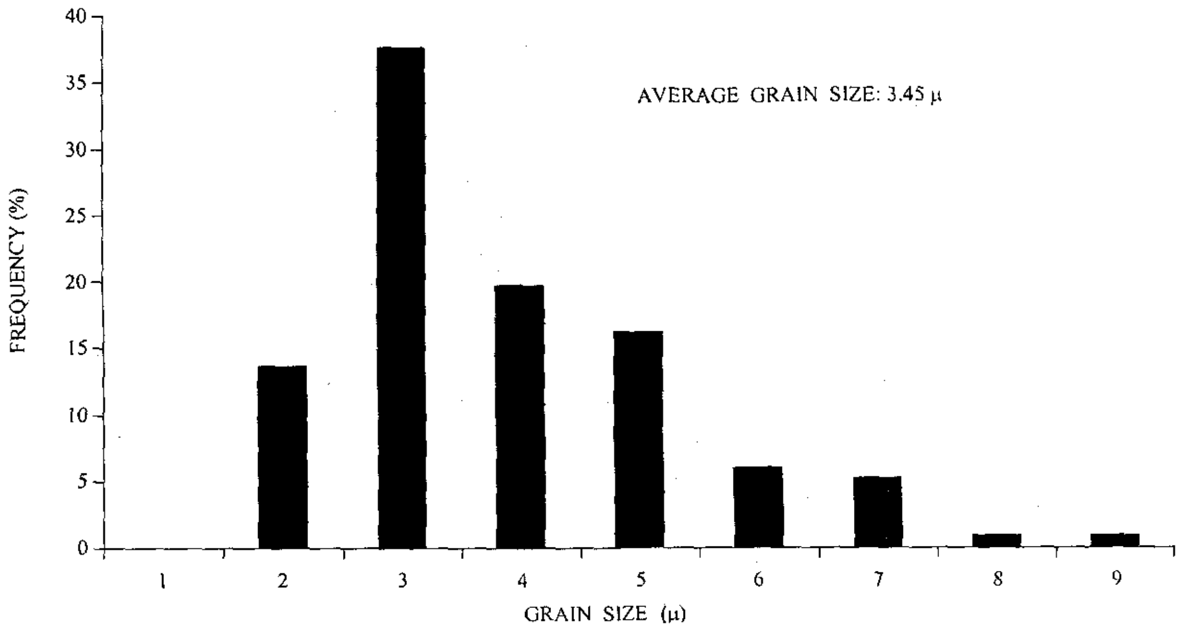



**Figure 7. Grain size distribution of *CuBX* steel**

alloy. A non-equiaxed ferrite grain microstructure with irregular grain boundaries provides a more tortuous path for a crack to propagate compared to that of an equiaxed grain microstructure. This results in absorption of more energy during crack propagation, and thus, enhancing the impact toughness value of the steel. In the present steel, the ferrite microstructure has jagged grain boundaries and the shape of the ferrite grains is off-polygonal, anisotropic, and irregular, (Fig. 5). This type of ferrite is often designated as quasi-polygonal ferrite[10], $\alpha_q$. In this type of microstructure, the trace of the prior austenite grain boundaries is scarcely seen. (Fig. 5), as the $\alpha_q$ grow crossing over the $\gamma$-grain boundary. The $\alpha_q$ phase may nucleate as an allotriomorph at the $\gamma$-grain boundary or other defect but grows mainly in a much lower temperature region. As a result, these ferrite grains exhibit off-polygonal/ irregular/acicular shapes. The presence of this type of ferrite has contributed to the enhancement in strength and toughness values[11]. It is not only the shape and size of grains but also the distribution of grain size which has a significant role to play to affect the mechanical properties.

A mixed grain size, ie, a wide grain size distribution has a deleterious effect on the mechanical properties[12]. The presence of mixed grain sizes has been shown to be responsible for the deterioration in resistance to low-temperature brittle fractures in similar grades of steel[13]. In the present case, the grain size distribution was found to be uniform and narrow (Fig. 7). This also contributed to the enhancement of the mechanical properties.

## 6. CONCLUSION

A new generation of high-strength steel with a combination of superior strength and toughness has been successfully developed through thermomechanically-controlled processing route. The product is available in as rolled condition rather than in quench and tempered condition. The design utilises a novel and innovative concept of grain refining using microalloying elements and thermomechanically-controlled processing as the means of optimising mutually exclusive properties of high strength together with high toughness. The combination of high strength and toughness is realised when the following features of microstructure are obtained:

- Non-equiaxed ferritic microstructure
- Fine grain (< $5\mu$ average size)
- Uniform grain size distribution

The above microstructural features could be produced when the chemistry of the steel was suitably designed as given in Table 1 and the controlled rolling schedule was designed as per the following thermomechanically-controlled processing parameters:

- Suitable alloy design to raise $T_{RXN}$ (~ 950 °C)
- Suitable alloy design to lower $A_{r3}$ ( < 750 °C)
- A low-soaking temperature (~1000-1050 °C) to prevent the growth of austenite grains and also to restrict the complete dissolution of *NbC*/*Nb*(*CN*) precipitates
- A lower amount of deformation (~20-25 %) above $T_{RXN}$ and remaining (75-80 %) deformation below $T_{RXN}$
- A finish-rolling temperature of ~800-775 °C.

Theoretically, the weldability of the developed steel is enhanced due to its low-alloy content accompanied by low-carbon content. Studies were initiated on the hydrogen-induced cold cracking susceptibility of this steel. The results indicated the steel to be less susceptible to hydrogen-induced cold cracking than the conventional ship-grade steels.

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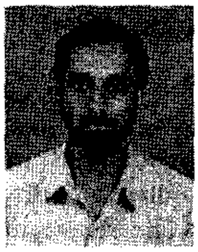

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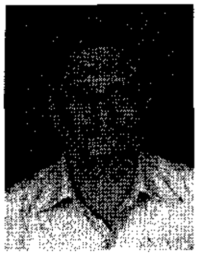

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