PRODUCTION AND PROPERTIES ON PIPELINE STEELS WITH

NON-POLYGONAL FERRITE MICROSTRUCTURE

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Abstract

Flows, and subsquently pressures, in gas pipelines are being steadily increased in an effort to reduce costs. This may indicate a demand for Grade X80 pipe in the not too distant future. In this regard, pipeline steels with non-poligonal ferrite microstructure will certainly come to assume greater importance, because their mechanical properties are intrinsically better than those with ferrite-pearlite microstructures.

The manufacture of Grade X80 pipe requires a careful examination of the contribution made by control rolling and conventional heat treatment and by on-line treatment such as direct quenching and accelerated cooling (which have yet to be developed for full-scale works application).

Development of Gas Pipeline Steels Over the Last Thirty Years

Radical changes occurred in the design principles of welded structures during the first few decades of this century. Where mechanical strength was concerned, it was gradually accepted that the yield strength was a more important requisite than tensile strength. Other properties, such as toughness and weldability also came to be taken into consideration, since these give a truer picture of the materials, in respect to the conditions under which they are used.

Consequently, the trend was towards a lower C content (Figure 1) the mechanical properties being controlled by increasing the amount of Mn. To give some idea of the evolution that has occurred, it is interesting to note that in 1945 the API Specification required a maximum of only 1.5 percent Mn with a C content of up to 0.3 percent. But, by the Sixties, it was common to use an Mn-C ratio of about 6:1 (with up to 0.21% C) (2) for Grade X60 pipe, while in present-day pipeline steels the ratio is as high as 30:1 or over (with C even below 0.06%).

Meanwhile, it has become common practice (3) to deoxide steel by adding Al (Figure 2). This ensures a general improvement in mechanical properties, also through the precipitation of AlN and the consequent reduction in the amount of free N in solution. Knowledge of the mechanisms involved was acquired only after the work done in 1949 by Beeghly (4), who developed a method for the quantitative analysis of AlN. In those years it was also demonstrated that refining of the ferritic grain plays a basic role as regards yield strength and the ductile–brittle transition temperature (5,6).

The effect of Al addition in refining the microstructures, known on a purely empirical base since 1935 (7), was thoroughly exploited by the introduction of normalization (Figure 3), which made possible the production of Grade X52 steels with a transition temperature of below 0°C.

Subsequent use of microalloying elements such as Nb, V and Ti permitted the production of structural steels with better mechanical properties. Research (8,9) revealed the twofold effect of adding microalloy elements which precipitate carbides and nitrides, (namely, refining of the ferritic grain and precipitation hardening). The first of these two mechanisms – similar to that which occurs on the addition of Al – permits improvements in yield strength and the ductile–brittle transition temperature, while the second results in a further improvement in mechanical strength and in a higher transition temperature (Figure 4). The different solubilities in austenite of the compounds of the microalloy elements (Figure 5 and Table I) with C and N, led to development of normalized steels up to Grade X60 containing V and/or Nb, with adequate levels of N, and subsequently to the technique of control rolling, particularly by exploiting the interesting effects of Nb. It is not necessary here to follow the evolution of control rolling. This technique was applied widely by industry towards the mid Sixties (11), although the first trials were run some ten years earlier (16). It is important to emphasize, however, that control rolling made it possible to produce pipes up to Grade X70 which are suitable also for Arctic use (transition temperature below –50 C). Control rolling initially resulted in the further development of Nb and V steels with ferrite–pearlite microstructure. Subsequently, pipeline steels with polygonal and acicular ferrite structure were produced by control rolling.

Acicular ferrite steels derive from the research on bainitic steels initiated by Irvin, Pickering and co-workers (29–32) and continued on laboratory scale with low–carbon steels (12–21), followed by industrial development in Europe, North America and Japan (22–28).
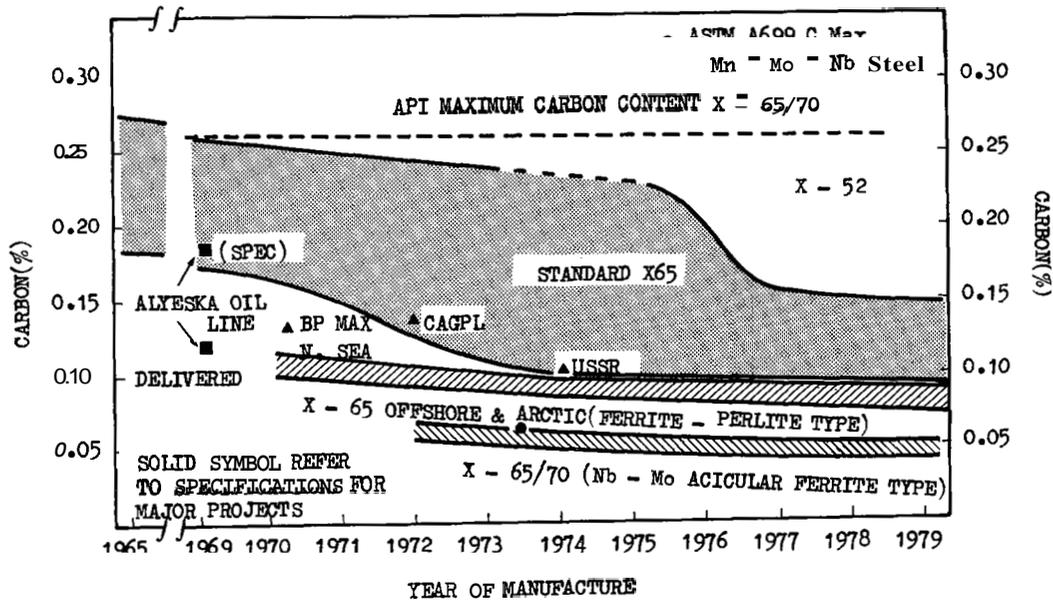
Figure 1. Changes in carbon content of large diameter pipe during the past decade (1).
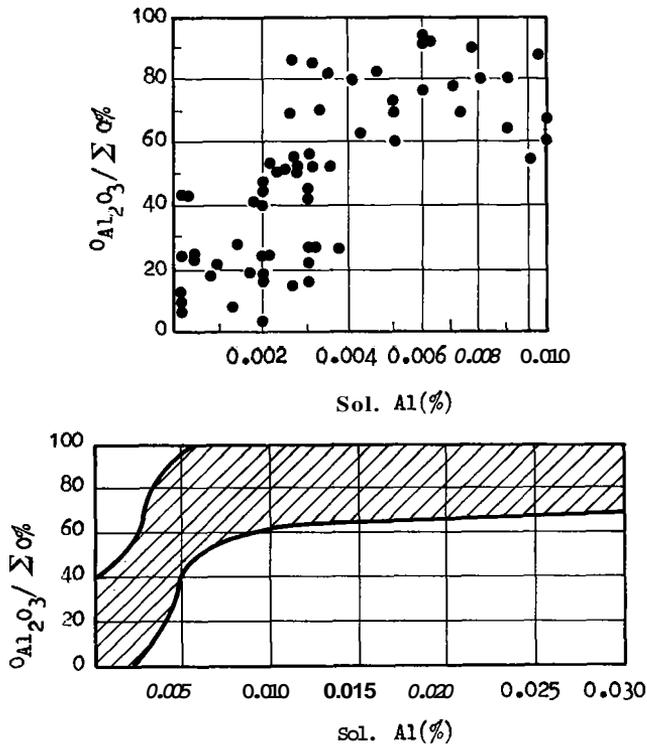
Figure 2. Effect of aluminum in steel on the compositions of oxide inclusions (3).

$^0Al_2O_3$ ∎ ∎ ∎ ∎ ∎ ∎ ∎ ∎ ∎ Oxygen as Al_2O_3

$\sum 0$ ∎ ∎ ∎ ∎ ∎ ∎ ∎ ∎ ∎ ∎ Total oxygen
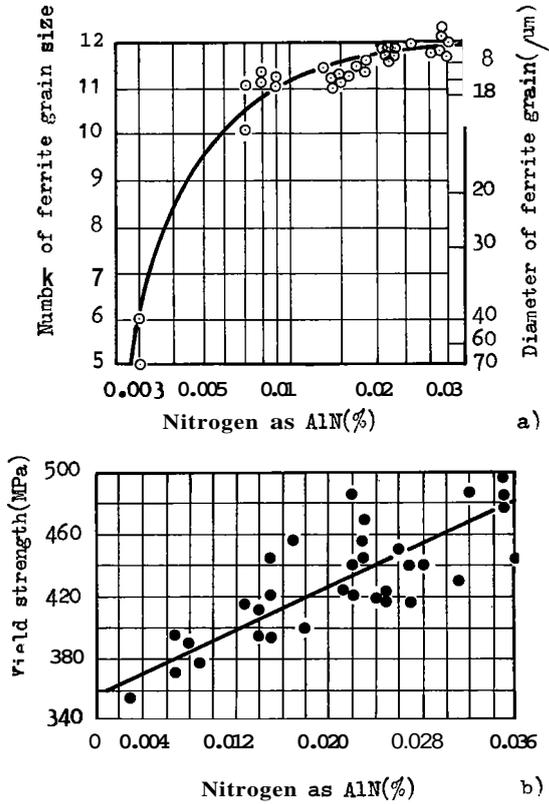
Figure 3. Effect of aluminum nitride on:
(a) ferrite grain size of normalized low C steel.
(b) yield point of normalized low C low Mn steel
(C 0.13 – 0.20%, Si 0.30 – 0.60%, Mn 1.30-1.66%) (3).
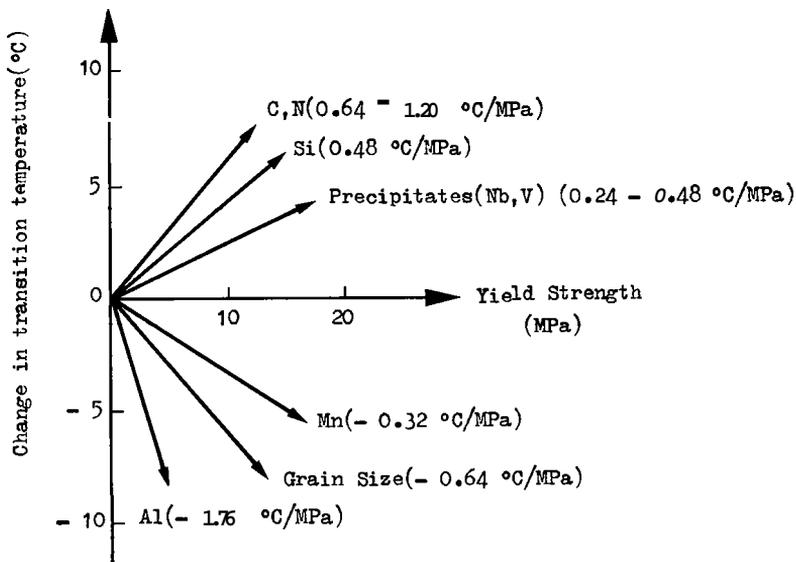
Figure *4*. Strength – toughness vector diagram.
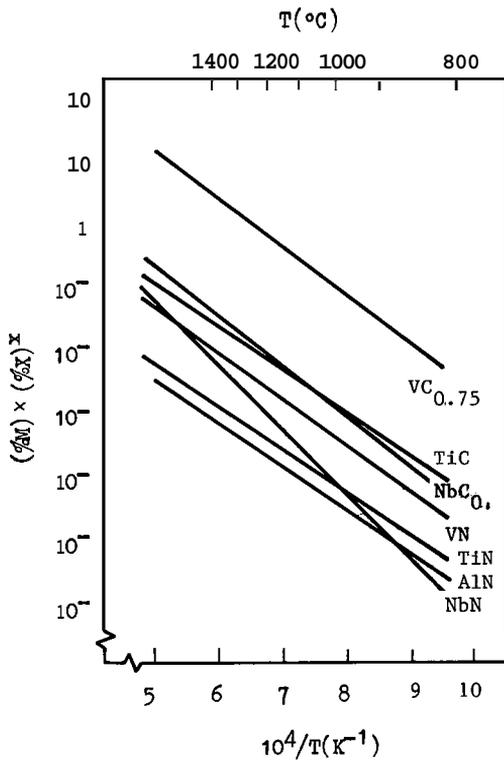
Figure 5. Solubility of carbides and nitrides in austenite (33).

Table I. Solubility of carbides and nitrides in austenite (33).

Carbides	$\log (\% M) \times (\% C)^y$	Nitrides	$\log (\% M) \times (\% N)$
$NbC_{0.9}$	$-\dfrac{7520}{T} + 3.11$	NbN	$-\dfrac{10230}{T} + 4.04$
$VC_{0.75}$	$-\dfrac{7650}{T} + 5.51$	VN	$-\dfrac{7070}{T} + 2.27$
TiC	$-\dfrac{7000}{T} + 2.75$	TiN	$-\dfrac{7000}{T} + 1.51$
		AlN	$-\dfrac{7400}{T} + 1.95$

Thus, future demand from potential users of pipe of higher grade than those now in use (≥ X75) could mean a substantial increase in the research and development of acicular steels produced by control rolling and/or heat treatment.

Non-Polygonal Ferrite Microstructures

Low carbon non-polygonal ferrites fall in the more general class of bainitic steels. There is no unit criterion for the classification of such microstructures at the present time (34-39). In fact, the characteristics of the bainitic transformation may be likened both to transformations of the martensitic type and to those of the thermally activated type.

However, the most useful classification criterion appears to be that based on the presence and the morphology of the carbides. In this way, it is possible to sub-divide the low-carbon bainites obtained by isothermal trans-formation into three types corresponding to three transformation temperature ranges (45-47) (Figure 6). Basically this classification also holds good for the continuous cooling transformations, albeit with the complication of the simultaneous presence of various microstructural components (Figure 7).

In the case of the higher transformation temperatures, phases are obtained which contain only very limited quantities of carbides or no car-bides at all, and with a high density of dislocation. Thus, substructures are formed consisting of packets of lamellae (co-variant packets) separated by low-angle boundaries (Figure 8), or cell structures resulting from the reorganization of the dislocations (Figure 9). In general, these structures (known as acicular ferrite) are associated with islands of high-carbon martensite (C ~ 0.4%) and residual austenite (MA constituent) (41, 42) (Figure 10). The acicular ferrite and MA constituent taken together are generally referred to as granular or massive bainite, (41, 43, 44) or Type I bainite (45-47).

The more traditional upper and lower bainites are obtained at the lower transformation temperatures.

Tensile strength of non-polygonal ferrite

The type of transformation which results in non-polygonal (or bainitic) ferrite produces a high density of dislocations ($\sim 10^9$ mm^{-2}) which endows it with inherently better tensile strength than polygonal ferrite (transformation strengthening (50)).
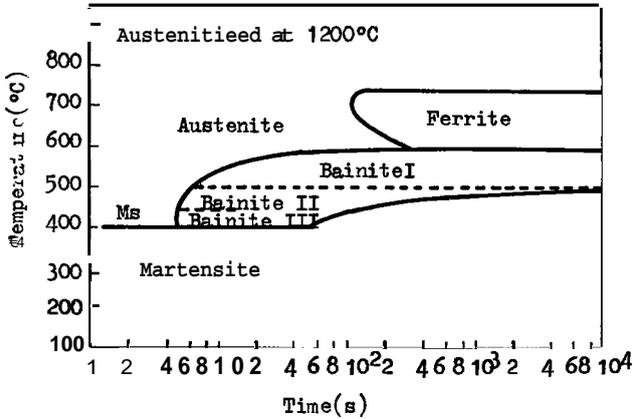
Figure 6. The schematic representation of the temperature
ranges where 3 types of bainite formed in the
Cu–Ni–Cr–Mo–V steel during isothermal transfor-
mation *(45)*.
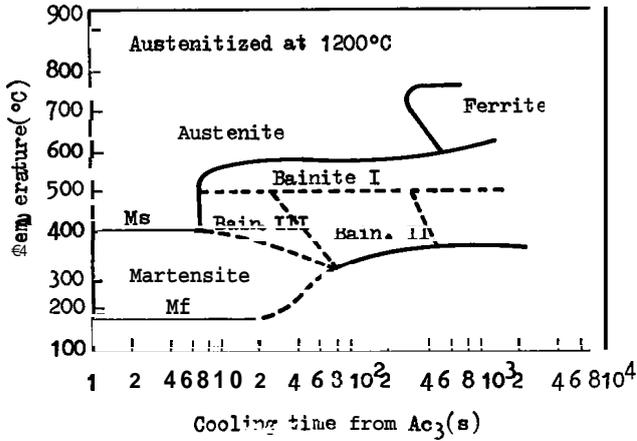
Figure 7. The schematic representation of the temperature
ranges where 3 types of bainite formed in the
Cu–Ni–Cr–Mo–V steel during continuous cooling *(45)*.
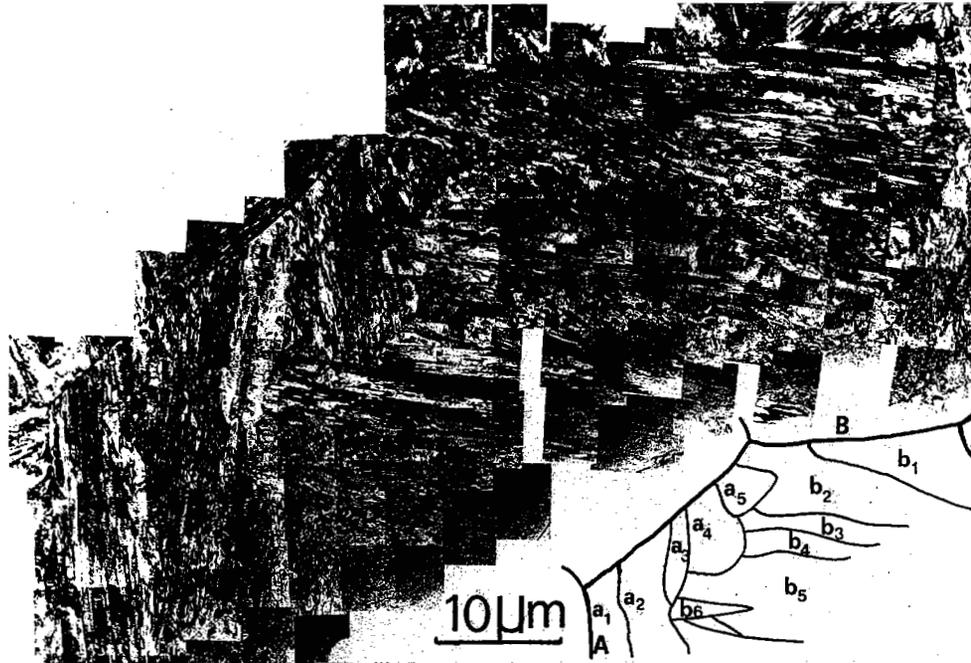
Figure 8. Typical aspect of the bainite derived from the transformation of coarse grained austenite ($d_{AG} \sim 50 \mu m$). The boundaries of the co-variant packets are schematically reproduced (0.025% C-2% Mn-3% Cr steel (40).
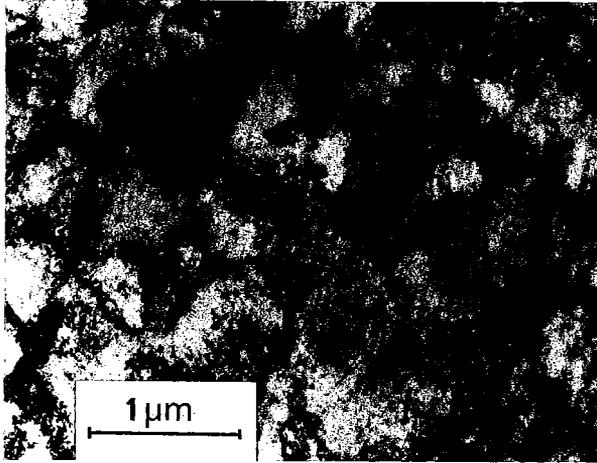
Figure 9. Example of cell structure in the bainite of a
0.925% C-2% Mn-3% Cr steel with a very fine
austenite grain *(40)*.
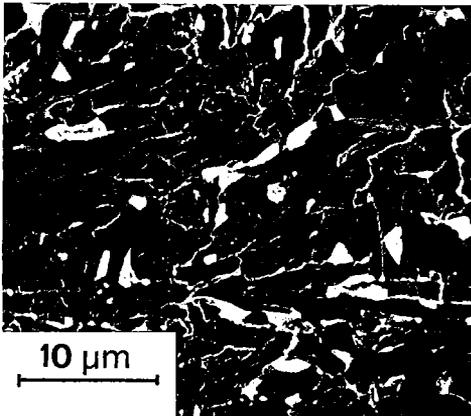
(a)
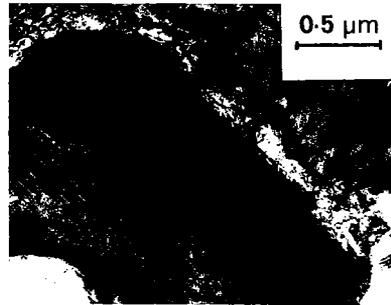
(b)

Figure 10. Observation of MA constituent under the scanning
electron microscope (a) and transmission electron
microscope (b). (0.02% C-2% Mn-0.3% Mo-0.002% B steel).

The temperature at which the transformation starts (Bs) has a decisive influence on the microstructure. If it is lowered by means of alloy elements, or by higher cooling rates, the dislocation density increases (i.e., there is refinement of the average dimensions of the cell or lamella substructure) which results in an increase in 0.2 proof stress $(\sigma_{0.2})$ and in tensile strength (σ_m) (48, 50, 51). The $\sigma_{0.2}$ value depends on the width of the lamellae or the diameter of the cells (d); opinions differ as to the type of functional relation. Gladman et al. (52) report dependence of the $K_1 d^{-1/2}$ type, while according to Langford & Cohen (53), it is of the $K_2 d^{-1}$ type. In any case, it seems certain that the mechanism operation is not of the Hall-Petch type, which holds good for polygonal ferrite; this postulates that the plastic deformation, produced in a grain which has yielded, propagates because of the stress field produced by the pile-up of dislocations on the grain boundaries. Another notable contribution to mechanical strength, due to the particular type of transformation, is attributable to the presence in solid solution in the ferrite matrix of a fair quantity of interstitial elements (C and N) (40).

Gladman et al. (52) attempt to explain the various contributions to $\sigma_{0.2}$ by means of a regression equation:

$$\sigma_{0.2}(MPa) = 88 + 37\ (\%\ Mn) + 83\ (\%\ Si) + 2900\ (\%\ N_{free}) +$$
$$15.1\ (d_L^{-1/2}) + \sigma_P + a_P \tag{1}$$

where d_L = dimension of lamellae (mm)

$\quad\sigma_P$ = precipitation strengthening as per the Ashby-Orowan model

$\quad\sigma_P$ = strengthening due to random dislocations.

On the other hand, other authors tend to isolate the more marked contributions by not considering the others separately. For instance, Buzzichelli & Mascanzoni (40), following the procedure proposed by Speich & Swann (54), have obtained the following with very low-carbon bainitic steels (C ≤ 0.05%) containing 2 percent Mn and 3 percent Cr:

$$\sigma_{0.2}(MPa) = 30 + 1900\left\{\%\ (C + N)_{free}\right\}^{1/2} + K\ d^{-1} \tag{2}$$

where (55) d = dimension of lamellae or cells (mm)

$$K = \begin{cases} 1.70 \times 10^{-1}(mm^{-1})\ \text{for cell structures} \\ 1.22 \times 10^{-1}(mm^{-1})\ \text{for lamellar structures} \end{cases}$$

Toughness of non-polygonal ferrite

While the tensile strength of low-carbon steels with an acicular micro-structure depends on the average size of the substructure (lamellae or cells), toughness appears to be goverened by the size of the bainite packet or covariant packet (Figure 11 and Table II). In fact, the other parameters proposed as toughness indexes (such as the unit crack path (UCP) or the
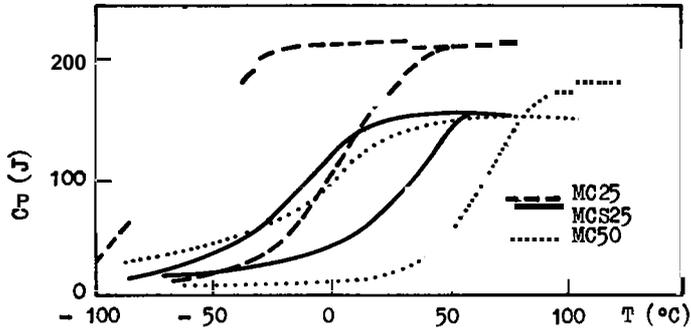
Figure 11. Effect of bainitic packet refinement on Charpy-V transition curves for three steels. Three sets of curves refer to packet sizes respectively given in Table II.

Table II. Results of variable-temperature mechanical tests.

Material	Slow bending test			Charpy test		Uniaxial tensile test (room temperature)
	$d_B^{-1/2}$, $(mm^{-1/2})$	σ_{cf} (MPa)	NDT (°C)	$d_B^{-1/2}$, $(mm^{-1/2})$	FATT 50% (°C)	$\sigma_{0.2}$ (MPa)
MC25	9.2	1850	-166	8.9	15	*550*
	11.2	2370	-198	12.4	*-50*	620
MCS25	9.1	1690	-140	9.1	40	550
	12.9	2410	-205	12.1	-5	580
MC50	10	1850	-60	9.1	75	700
	14.4	2680	-196	16.9	0	750
MCS50	11.2	2240	-170	–	–	–
	14.4	2670	-190	–	–	–

effective grain size (measured on the surface of the fracture) seem to be directly correlated with the average dimensions of the covariant packet (46, 57–62) (Figure 12).
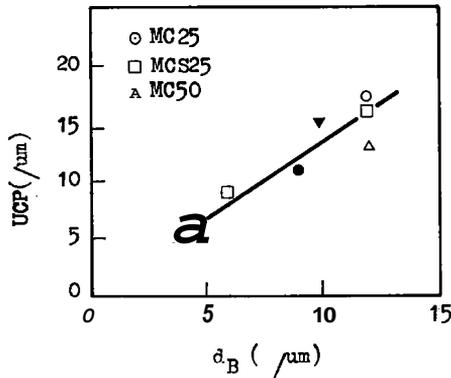
Figure 12. Relationship between unit crack path and bainite packet size. Points ● and ▼ refer respectively to 2% Mn–3% Cr steels with 0.045% and 0.03% C.

The factors permitting the bainite packet to be refined, with consequent improvement in the toughness of the material are: the lowering of the trans– formation temperature (which seems, however, not to have a very marked effect), and especially, refining of the austenitic grain size (32, *49*, 57, 63, 64). In this regard, the following empirical relation has been proposed (40):

$$\overline{d}_{pB} = 1.4(\overline{d}_{GA})^{0.55}$$

(3)

where $\overline{d}_{pB}$ and $\overline{d}_{GA}$ represent the average dimensions of the bainite packet and the austenitic grain, respectively (assuming the latter to be equiaxed).

The transition temperature, measured in the Charpy–V impact tests (ITT) can be expressed by a relation of the type (52):

$$ITT(°C) = -19 + 44 \ (\% \ Si) + 700 \ (\% \ N_{free})^{1/2} +$$
$$0.26 \ (\sigma_P + \sigma_d) - 11.5 \ (d_{pB}^{-1/2})$$

(4)

where a_P has the same meaning as in equation **1**

σ_d is the strengthening due to random dislocations (σ_ρ of equation 1) and to those which form the low–angle boundaries

d_{dB} is the average diameter of the bainite packet.

Another factor having a decisive influence on the toughness of the acicular or bainitic microstructure **is** the presence of carbides or islands of martensite (MA) and also their size and distribution *(49)*.

815

Regarding acicular ferrite, one of the basic morphological characteristics distinguishing it from upper bainite is the lack of any trace of the boundaries of the prior austenitic grains. Since the latter constitute preferential sites for the precipitation and segregation of impurities, their absence (according to some authors (12, 13)) seems to confer an inherently higher toughness to the acicular ferrite.

Alloy design of non-polygonal ferrite steels

In the case of gas pipeline plate with a ferrite-pearlite microstructure produced by control rolling it is necessary to consider the effect of the individual alloy elements on such matters as the following in order to find economical chemical compositions which ensure a good combination of strength, toughness and weldability:

- Growth of austenitic grain in the heating phase preceding rolling (Al-Ti-Nb).
- Recrystallization of the austenite (Nb-Ti-V) during rolling.
- Solid solution strengthening (Mn, Si, etc.).
- Precipitation strengthening (Nb-Ti-V).
- Inclusional control (Zr-Ce-Ti-Ca).

In the case of steels with a microstructure that is wholly or partly acicular ferrite and/or bainite, another two closely interconnected variables come into play, namely:

- Bainitic hardenability understood as the possibility of obtaining acicular microstructures at the cooling rate that can be used.

- Bainitic transformation temperature (Bs) which, as seen earlier, is of decisive importance as regards mechanical properties.

Table III indicates the alloy elements normally used for gas pipeline steels with non-polygonal ferrite microstructure. More recent investigations have also covered such alloy elements as V, Ti, Cr and B.

The effect of some alloy elements on bainitic hardenability is expressed quantitatively by the formula (49):

$$B\ (°C) = 830 - 270\ (\%\ C) - 90\ (\%\ Mn) - 37\ (\%\ Ni) - 70\ (\%\ Cr) - 83\ (\%\ Mo) \tag{5}$$

Table III. Typical composition ranges of non poligonal ferrite steels.

Elements	wt %
C	0.06 max
Mn	1.6 - 2.2
si	0.10 - 0.40
Al	0.04 max
S	0.010 max
P	0.015 max
Mo	0.25 - 0.40
Ni	0.35 max
Nb	0.03 - 0.10
N	0.01 max

B is a very effective element as regards hardenability, only a few ppm (10 to 30) being needed in solution to produce very marked effects (65-71). B acts only **on** the rate of nucleation of the proeutectoid ferrite, since segregation of the B atoms on the boundaries of the austenite grain results in a decrease in the free energy available for they → a transformation. Thus, in the CCT diagrams it is possible to observe a marked shift of the ferritic nose towards the slower cooling rates. B plays a very effective synergic role in combination with Mo (72, 73) and **Nb (74).**

Si and V favor the formation of polygonal ferrite (13, 14) instead.

Steels up to Grade *X70* for Gas Pipelines

As noted earlier, in the last decade there has been a major development of Grade *X70* steel pipe production for gas transmission lines. This develop- ment has received much of its impetus from the discovery of enormous deposits of natural gas in remote areas with very severe environmental conditions, such as the Arctic regions of North America and Siberia. There has also been an increase in pipe diameters (up to *56")* with working pressures of up to *75* atm to boost the flow rate, and hence reduce transmission costs. The low operating temperatures involved, however, require very tough steels that are highly resistant to the propagation of ductile fracture. This has led to a demand for pipes with Charpy-V values of *40* and *60* **J** down to *-40* **C**, measured by the transverse test.

With the development and greater understanding of control rolling it has become possible to obtain the desired properties from conventional ferrite- pearlite steels.

Inhibition of austenite recrystallization below 1000 C *(75)* ensured by Nb and to a less extent by Ti and V (associated in some cases with control of austenite grain size in the reheating prior to rolling) permits extremely fine ferritic grain associated with good strength and toughness properties to be obtained (average diameter up to *5 - 6,*um). Use has recently been made also of the ferrite work hardening that is obtained by adopting rolling cycles with marked reduction in mixedy $+\alpha$ phase *(76-79)* to raise mechanical strength even further.

The alternative route followed for the production of Grade *X70* pipelines is to use low-carbon steel with a mixed polygonal and acicular ferrite microstructure. Unlike pearlite, this latter phase exerts a twofold action on the tensile properties of the steel right from the very first stages of plastic deformation. The acicular ferrite participates directly in the plastic deformation and the associated small volume fractions of constituent MA, that are also present, are generally sufficient to produce continuous yielding **(18, 42, 80, 81).** As a result of the first characteristic, the yield strength of the steel is governed by the yield strength of the poly- gonal and acicular ferrite and by the relative volume fractions, as indicated by the relation (40):

$$\sigma_Y = f^p \sigma_{YP} + (1 - f^p) \sigma_{ya} \tag{6}$$

where f^p = volume fraction of polygonal ferrite

σ_{YP} = yield strength of polygonal ferrite

σ_{ya} = yield strength of acicular ferrite

Steels containing acicular ferrite with a high work hardening rate in the first stages of plastic deformation are of particular interest for making pipes by the UOE process (Figure 13). During this type of cold forming the plate is subject to bending deformations (U and O forming); the inner part of the pipe being subjected to compression and the outer part to tension. In the case of O-forming, moreover, there is also a slight compression of the whole thickness of the pipe wall. Finally, during expansion (E), all the layers of the pipe wall are under tension usually of 1.2 to 1.5 percent. During UOE forming, therefore, the Bauschinger effect is exerted to a varying extent in the outer and inner layers of the pipe: the ensuing decrease in yield strength in steels containing acicular ferrite is, however, counter-balanced by the work hardening produced during expansion.

The effect of 'this work hardening is such as to compensate also for the decrease in yield strength that occurs (again owing to the Bauschinger effect) in the cold flattening of the tensile test specimens cut transver-sally from the pipe, as per API 5LX Standard. Therefore, the yield strength of pipes with an acicular microstructure is generally greater than that of the plate, as proved in the laboratory (82-85) and in the industrial practice (22) (see Figure 14).

The use of low-carbon contents (down to a minimum of 0.05%), such as those now attainable with the BOF, ensures lower ductile-brittle transition temperatures and higher Charpy-V energy values (as proved by Smith et al (15) (see Figure 15)).

Further improvements in the ductility can also be achieved by lowering the S content (see Figure 16), which leads to a reduction in the volume frac-tion of non-metallic inclusions. The use of lanthanides has also helped improve plate toughness. The best known of the lanthanides is Ce which is commonly used in a ratio of 1.5 - 2.0 with S to obtain the lowest Charpy energy anisotropy (see Figure 17). An industrial method of inclusions control has also been developed (86) which involves injecting fine-sized calcium silicon and calcium carbide directly into the ladle by an argon lance. This permits reduction of both the S and the O contents (Figure 18) and results in spheroidal inclusions, as does the Ce treatment.

1 Plates   2 "U" Forming   3 "O" Forming   4 Expansion
              (Bending)   (Compression)   (Tension)
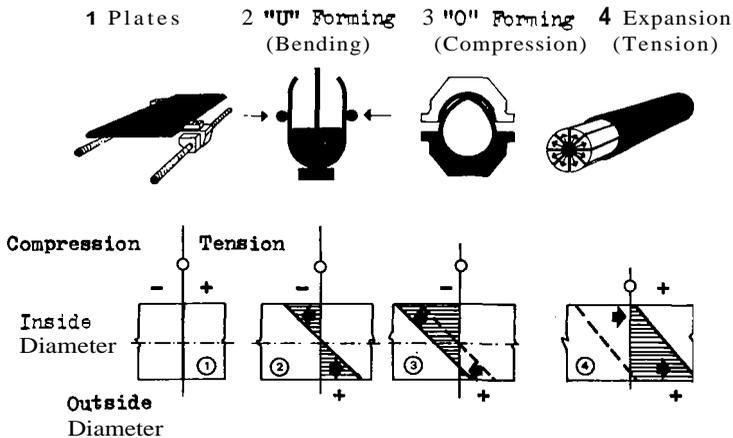
Figure 13.   Steps in U-O-E process of pipe making and their
              influence on strain distribution in the inside
              and outside layers of the pipe wall.
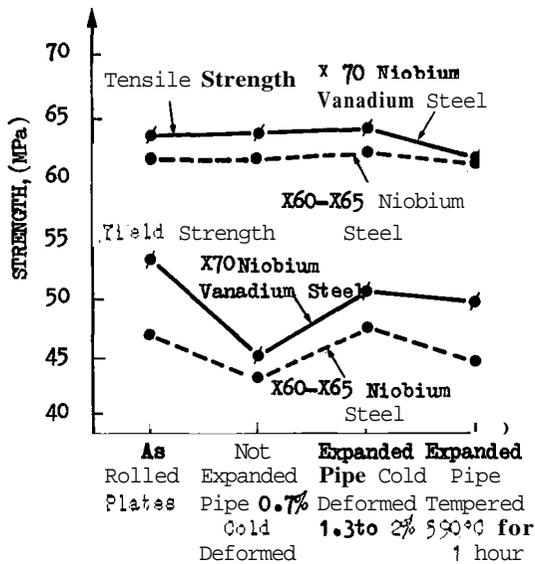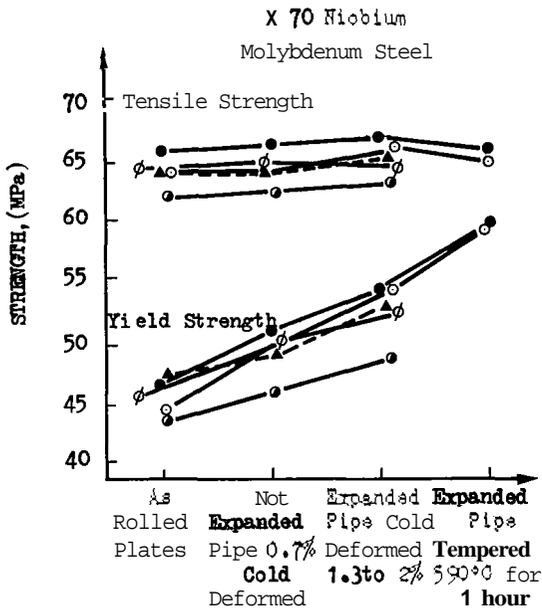
Figure 14. Effect of cold forming and tempering on the transverse tensile properties of commercial niobium, niobium-vanadium and molybdenum-niobium steel described in Table IV (22).

Table IV. Average composition (wt %) and quantities made of recently produced pipeline steels *(22)*.

Pipe diameter (m)	Symbol	Thickness (mm)	C	Mn	Si	Al	S	P	Ni	Mo	Other	Carbon eq. *	Number of Heats
1.4	●	17.5–20.5	0.153	1.43	0.27	0.034	0.016	0.017	0.036	-	-	0.39	7000
1.2	◰	14.5	0.115	1.59	0.34	0.027	0.007	0.014	0.032	-	V 0.073	0.395	600
1.2	●	17	0.06	1.61	0.29	0.040	0.009	0.015	0.044	0.32	-	0.39	65
1.2	o	14.5	0.065	1.60	0.28	0.043	0.009	0.015	0.040	0.28	-	0.39	60
1.4	◑	30	0.06	1.62	0.26	0.040	0.006 to 0.012	0.019	0.045	0.30	Ni 0.26	0.40	3
1.4	◒	30	0.06	1.75	0.26	0.040	0.006 to 0.012	0.019	0.045	0.30	Ni 0.26	0.42	2
1.4	x	18	0.06	1.60	0.36	0.037	0.009	0.014	0.047	0.29	-	0.39	15

All steels produced in *320* ton heats, either ingot or continuously cast. The *30* mm pipe was only ingot cast and mish-metal treated. Steel with (X) was produced in test quantities only from continuous casting. The symbols above were used for steels in Figure *15*.

* $Ceq = C + \dfrac{Mn}{6} + \dfrac{Cr + Mo + V}{5} + \dfrac{Cu + Ni}{15}$

ONE
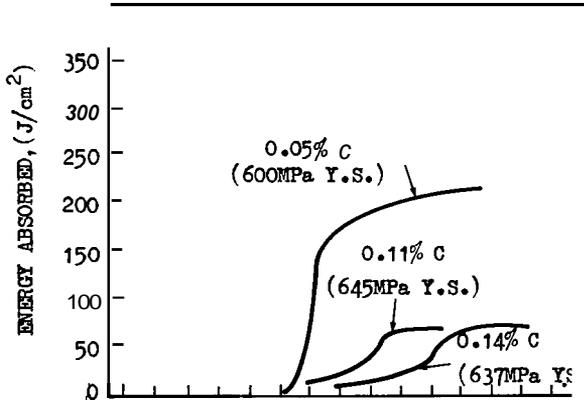
Figure 15. Impact resistance in transverse direction of 12.7 mm–thick aged plates of 1.8% Mn–0.25% Mo–0.08% Nb steels at three levels of carbon content (14).
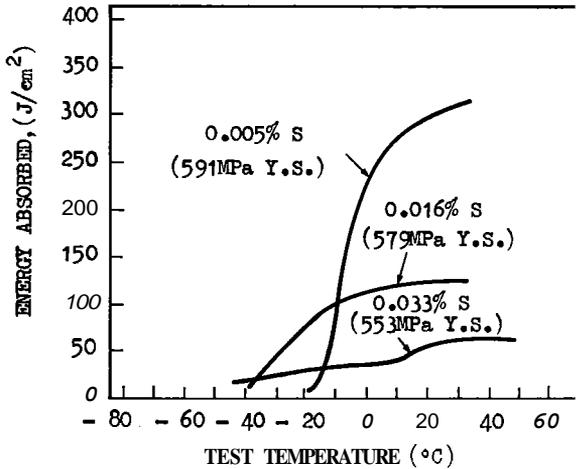
Figure 16. Impact resistance in transverse direction of 12.7 mm–thick aged plates of 0.05% C–1.86% Mn–0.24% Mo–0.07% Nb steel at three levels of sulfur content (14).
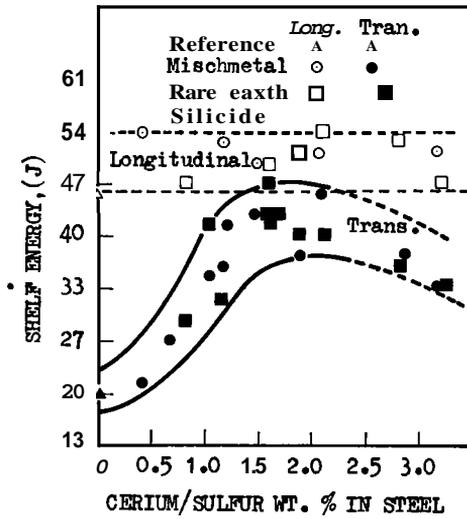
Figure 17. Relationship between shelf energy determined on longitudinal and transverse Charpy 1/2 size V-notch specimens and cerium-to-sulfur ratio. Each data point is the average of two tests (87).
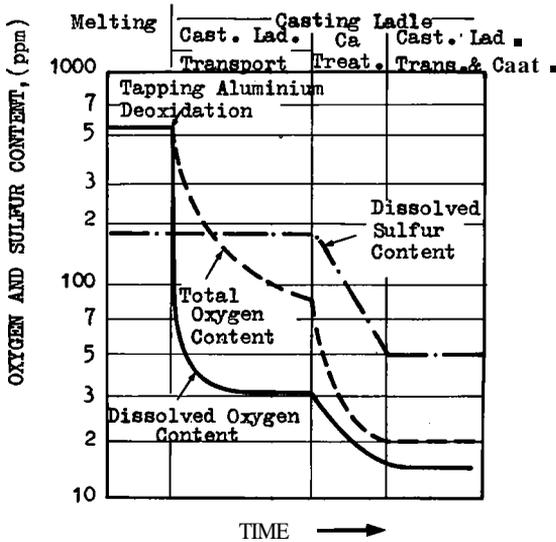
Figure 18. Variation of the oxygen and sulfur contents of the steel during calcium treatment in the ladle (86).

Control rolling

As previously stated, the transmission capacity of gas pipelines has to be increased for economic reasons. This is achieved principally by raising the working pressure. In a few years, this trend is certain to lead to a demand for pipes of X80 grade or greater. Steels with a partly or wholly acicular microstructure are thus destined to become of ever–increasing importance, since it seems unlikely that the strengths required can be ensured by ferrite–pearlite steels.

To achieve the necessary mechanical properties, the most direct route is to establish appropriate chemical compositions that enable the objective to be reached solely by control rolling. However, there are some extremely negative aspects involved, as is evident. Not the least of these is the need to use steels with compositions that are richer and more costly, thus de-creasing or nullifying the economic advantages deriving from the use of a less–costly manufacturing process such as control rolling. A marked re-straint in the search for new compositions also regards the need to ensure good weldability. This greatly limits the use of large amounts of elements such as Mn, Mo, Cr and Ni that are of particular importance because of their hardening effect, in order to obtain materials with a sufficient quantity of acicular phase.

Considerations such as these have led, in recent years, to increased interest and in–depth research into the use of post–rolling heat treatments, particularly those which can be performed on–line, since these appear the most promising.

Quench and tempering of plates and pipes

Conventional quench and tempering treatment (Q.T.) after rolling (un-controlled) and natural cooling to room temperature consists in reheating to a temperature slightly in excess of Ac_3 (usually 900 – 950°C) followed by quenching in water and then tempering at sub–critical temperature. This is, therefore, a costly process and productivity is low. These disadvantages, moreover, are not counterbalanced by the lower cost of the compositions which can be used. In fact, with the type of heating involved, it is not possible to dissolve the carbides formed by the microalloy elements (except in the case of V). Consequently, precipitation hardening cannot be conveniently exploited. On the other hand, the use of higher reheat temperatures would cause a marked increase in austenite grain size with a consequent fall–off in toughness (89). The required strength levels must, therefore, be achieved essentially through the production of a microstructure consisting mainly of martensite. This involves the use of elements such as Mo, Ni and Cr in considerable quantities, together with relatively high C contents (0.13 to 0.22%).

Until recently, the Q.T. process has been the only way of manufacturing pipes having yield strengths of up to 700 MPa and above with good impact properties despite these disadvantages.

Q.T. treatment of the pipes directly, rather than the plate, has opened up the way for great progress in this field. First, it avoids problems that can arise during the cold forming of high–strength plate which has already

been treated. Then, there are metallurgical advantages due to the fact that the pipeheating cycle, which is often very rapid because of the use of induction heating system, enables very fine microstructures to be attained. However, the greatest advantage is the marked improvement in mechanical properties, especially the toughness of the weld both in the fusion and in the heat affected zone (96).

The good toughness of Q.T. plate, as measured by the Charpy–V test, corresponds to a sufficiently low 50 percent FATT values. Full–scale bursting tests run by the CSM and Italsider show, instead, that the behavior of these materials is less favorable as regards resistance to ductile fracture propagation. Indeed, the results show that propagation did occur on Q.T. pipes with Charpy–V energies sufficient for arrest, calculated according to the various current theories (CSM–Italisider, Battelle, AISI, British Gas) based on tests performed on control rolled pipes (Figure 19). As postulated by Bonomo et al. (97), this means that to arrest ductile fracture in Q.T. pipes higher CV values are needed than for controlled rolled pipes.

It is, therefore, for Q.T. materials, to determine CV values capable of ensuring arrest of ductile fracture or to find a laboratory test of more general validity which is also applicable to materials of this kind.
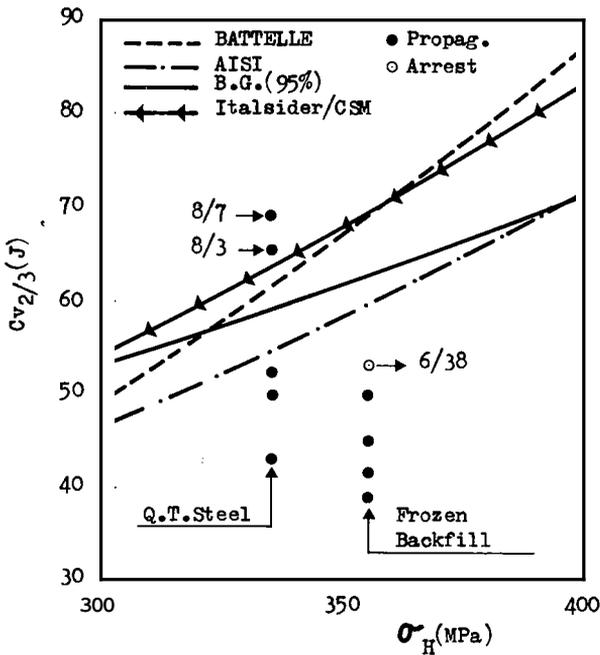
Figure 19. Frozen backfill and Q.T. material results as compared with the current formulae (97).

Direct quenching

Many of the negative aspects typical of Q.T. treatment can be overcome by quenching the control rolled plate directly as it leaves the finishing mill (Direct Quenching). The first laboratory tests on D.Q. were performed in the mid Sixties by the British Steel Corporation, mainly on C-Mn steels microalloyed with Nb, V, Ti and Cu (C less than 0.10%) (89–91). Subsequently, this type of treatment was investigated elsewhere; in France by IRSID (92), and in Italy by Italsider (88) and the CSM (93).

The advantages of D.Q., compared with the previous type of treatment, are evident as far as productivity and energy saving are concerned. However, there are also marked advantages of a purely metallurgical nature (Figure 20), since D.Q. enables very good combinations of strength and toughness to be obtained with relatively poor and economically advantageous chemical compositions, even without having to temper the plate provided that very high cooling rates are adopted (90 – 100 C/s in the middle).

Unlike the conventional method of treatment, D.Q. permits optimal use to be made of all the advantages of control rolling, particularly the possibility of exploiting precipitation hardening due to the microalloy elements (Nb, V and Ti) and of obtaining a fine, highly–deformed austenite grain which, after quenching, produces microstructures of the acicular ferrite, traditional bainite or martensite type (depending on the composition and the cooling rate), characterized in any case by very fine substructure and consequently endowed with good impact properties.

The best mechanical properties appear to be obtained by using finishing–rolling temperatures slightly higher than Ar_3 (about 30 C) with steels of the C–Mn–Nb and of the C–Mn–V type (94). For instance, in the case of 13 mm thick plate having a composition 0.13 percent C, 1.25 percent Mn and 0.025 percent Nb, Preston & Mitchell (94) report the following results in the untempered state: tensile strength = 922 MPa, 0.2 percent proof stress = 740 MPa, T_{27J} = – 90 C.

By adopting even lower C values it is possible to obtain even better CV results, though there is some fall off in tensile strength. Plate 12 mm thick having a composition of 0.07 percent C, 1.68 percent Mn and 0.09 percent Nb, produced in the CSM pilot plant (93) has given 50 percent FATT values of – 110 C in the longitudinal direction, with 0.2 percent proof stress and tensile strengths of 570 and 825 MPa, respectively.

Despite the undeniable advantages of D.Q., there are difficulties in its practical application, not the least being the capital investment involved. There is also the fact that to make the best possible use of the advantages offered by D.Q. there must be the least possible delay between the time the plate leaves the finishing mill and the start of quenching, especially when finishing temperatures close to Ar_3 are adopted. This involves a layout problem, owing to the dimensions of the quench plant that must guarantee a high cooling rate for quite thick plate. Moreover, there are engineering difficulties that have not yet been completely solved. To attempt to alleviate these difficulties, studies have been made of special chemical compositions which reduce the adverse metallurgical consequences of the delay prior to quenching (95). However, these compositions contain elements such as Mo, Cr, and Ni and thus they are not as advantageous from the cost aspect. To date, it does not seem as if this problem has been fully resolved.
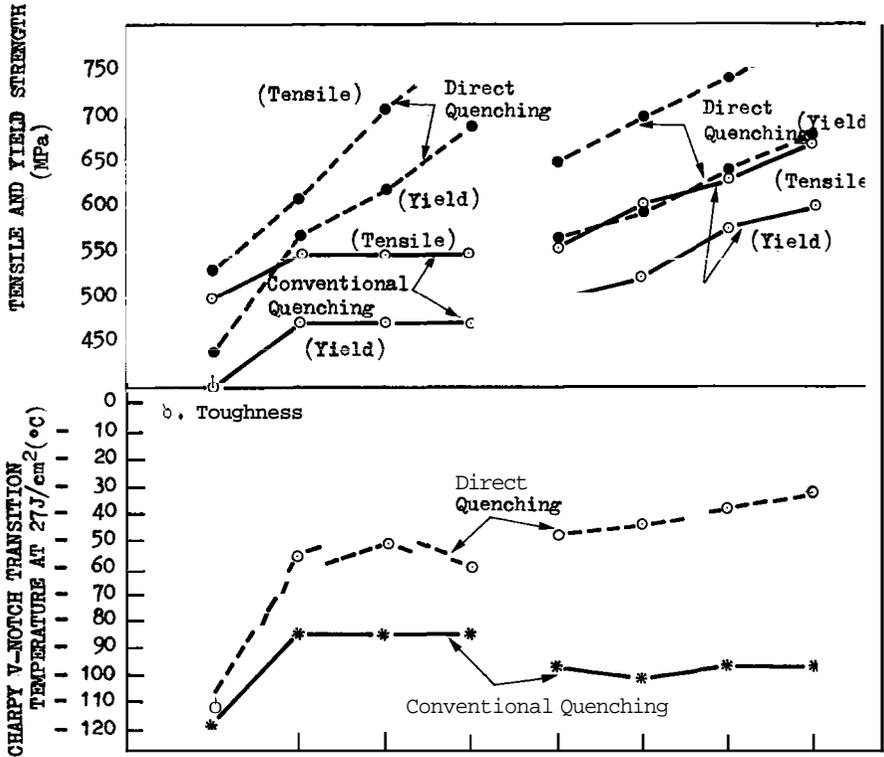
Figure 20.   Comparison of direct and conventional quenching on the
             mechanical properties of two low-carbon steels contain-
             ing varying amounts of microalloying elements.   The base
             composition of this aluminum-killed steel was 0.07% car-
             bon, 1.25% manganese, 0.25% silicon, and 0.13% sulfur.
             The finishing temperature was 900 C.   Each steel was aus-
             tenitized at 910 C for 30 minutes and tempered at 600 C
             for 1 hour.   The quenching rate was 55 C per second for
             the skin and *40 C* for the core, corresponding to 20 to
             25 mm for an industrial roller quench (88).

Accelerated cooling (A.C.) after control rolling and before coiling is usually used on hot-strip lines. The advantages from the metallurgical and productivity aspects are well-known. In recent years, an examination was made of the possibility of extending this treatment to the production of high tensile strength plate for gas pipelines. The literature on this matter, however, is still very limited (96, 98-102). The purpose of A.C. is to improve mechanical properties, especially plate strength, by modifying the microstructure or by refining this through lowering the transformation temperature. The cooling rate generally lies between that typical of air cooling (which varies depending on plate thickness) and 20 - 25 C/s. Ouchi et al. (96), for instance, report that with appropriately chosen process parameters and a cooling rate around 10 C/s, the microstructure of an as-rolled Grade X70 Nb-V steel changes from ferrite-pearlite to partly acicular ferrite, while the strength increases to that typical of a Grade X80 steel (Figure 21), with a slight improvement in toughness, as well.

However, it would appear that the best results are obtained by interrupting the A.C. at a relatively high temperature (450 - 600 C) and then continuing cooling in air. Yet, it seems there is no simple correlation between mechanical properties and the interruption temperature (96, 101). The advantage of interrupting the A.C. at a relatively high temperature is particularly evident with steels containing microalloy elements, especially Nb, because this ensures the precipitation of fine carbide and carbonitride particles which play an important role in the precipitation hardening of the steel.
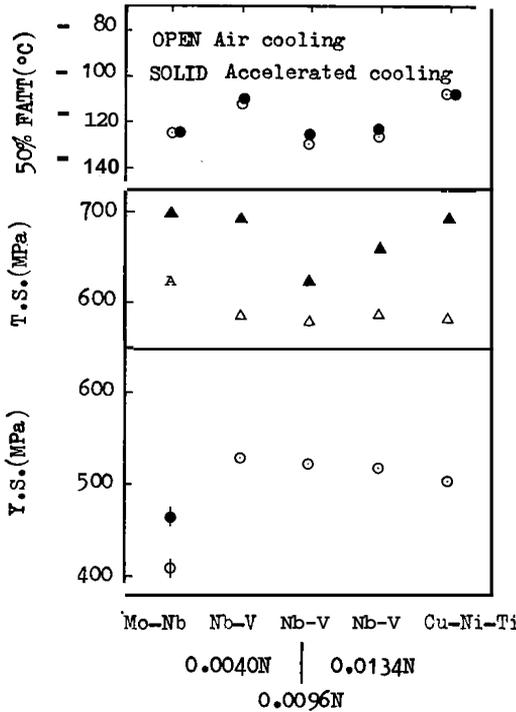
Figure 21.   The effect of the accelerated cooling on the mechanical properties (Plates thickness 20mm) (102).

The use of A.C. after control rolling, therefore, permits the use of steels with relatively low carbon contents for the production of high grade plate. Thus, there is a substantial saving on composition cost with only a slight increase in processing cost, while the capital investment is certainly less than needed for the installation of a direct quenching plant. So far, no A.C. plants for commercial pipe production have been built. There are important design problems to be overcome, such as ensuring even cooling of the whole of the plate surface to avoid deformation and lack of uniformity.

Acknowledgements

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