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November 1988

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**Control Rolling and Cooling for On-Line Production
of Strong, Tough Steels**

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OF STRONG, TOUGH STEELS

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Abstract

Strong tough steels have been designed to produce microcomposite structures of fine grained auto-tempered lath martensite with untransformed austenite films between the laths. Such structures can be produced on-line in a hot mill by controlled rolling and cooling. The steels are based on Fe/Cr/Mn/C compositions which can be air hardenable, or for lean compositions, water quench hardened with Nb microalloying. The steels have excellent mechanical properties that are superior to current HSLA steels and require no subsequent heat treatment following rolling and cooling. Successful commercial plate trials have occurred.

1. INTRODUCTION

There are current needs to optimise and conserve raw materials and energy, such that physical metallurgists and engineers are challenged to improve the mechanical properties of engineering alloys, in particular steels, and also to minimise high capital costs of new processing facilities. The major difficulty in optimising strength, toughness, and ductility in alloys comes from the fact that strength is usually inversely related to toughness and ductility. The increase in the former is achieved at the expense of the latter and vice versa, as illustrated in fig. 1. This is true in the majority of cases when relatively inexpensive alloying elements and processing are sought for practical alloy development. This concept has been the basis for the alloy design program for nearly two decades now at Berkeley (see e.g. refs. 1-7), a program which optimises ranges of mechanical properties through control of microcomposite microstructures. The steels considered here are those where the primary applications are for structures, mining, and in defense. These applications require steels with high strength for load bearing, and high toughness and ductility to resist crack propagation and to ensure good formability. For mining and agricultural applications, wear and corrosion resistance are significant properties also (8-11).

2. Alloy Design, Processing & Structure - Property Relations

It is important to recognize the differences between crack initiation (C_v) and crack propagation (K_{IC}) toughnesses. These properties often appear to relate inversely, as when considering grain size effects and austenitising temperature to obtain all carbides in solution (12,13). In designing for toughness against crack propagation, high K_{IC} values are required. A simple relation exists between critical flaw size a_c , stress σ , and K_{IC} as follows:

$$a_c = \frac{1}{1.2\pi} \left[\frac{K_{IC}}{\sigma} \right]^2$$

In fig. 3, this relationship is plotted for three steels having a similar yield strength but with different K_{IC} values. For a critical crack size of 0.25 cm, it is clear that the steel with K_{IC} of 100 MPa $\sqrt{m}$ can tolerate such a crack up to the yield limit, whereas a 4340 or EN24 steel can only tolerate such a crack at stresses only about 30% of its yield strength. Thus, the aim of the research has been to raise the toughness levels of structural steels for yield strengths $\sim$ 1200 MPa to produce them on-line in an economical manner. In the present work the steels can all give at least 100 MPa $\sqrt{m}$ K_{IC} values. In our alloy design program, the principles of composites (viz, mixing ductile and strong, tough phases with

coherent or semi-coherent interfaces) have been utilized (1) with the development of medium carbon Fe-Cr-Mn HSLA steels which can be processed to give high strength and toughness in a refined microstructure. The latter are composites of a strong, tough auto-tempered dislocated lath martensitic phase (major phase) and hence $M_s > 200^\circ\text{C}$ and a softer, tough austenite phase as illustrated in Fig. 2. The advantages of further refinement with niobium have also been investigated as recently outlined (14). The current efforts of our research have paid attention to relating laboratory experiments to possible plant operations and this is the topic of the present paper.

The Cr range varies from 2-12% with the high Cr values being designed for improved corrosion resistance, e.g., in precious metal mining conditions where the environment is very acid. Strength and hardness is determined mainly by carbon content and Mn which is more economical than Ni helps in the stabilization (1,6) of the interlath austenite, in addition to adding hardenability. Thus, desulphurizing should be effective to avoid Mn sulphides, not only for toughness requirements but also to allow Mn to do its function. Molybdenum, where necessary (cost factors), is also beneficial for strength and temper resistance (2). For HSLA steels with leaner compositions, e.g., Fe/2Cr/1Mn/0.25C microalloying with Nb is beneficial (14) in not only refining austenite grain size but by increasing hardenability as is evident from the CCT diagram of fig. 5. The steels are designated QT n where the number n refers to the Cr content (n=2,4,8,10, etc.) Solution treatments of $\sim 1150^\circ\text{C}$ are suitable for these Fe/Cr/Mn/C steels since iron and chromium carbides are not difficult to dissolve (unlike other alloying elements, e.g., Ti). For the Nb bearing steels soaking at 1200°C is needed. The subsequent rolling-recrystallisation processing allows a refined microstructure to be achieved. Figure 6 shows schematically the processing procedures developed for the air and water quench hardenable steels (QT 10 and QT 2 + Nb). Unlike the control rolling accelerated cooling methods used for low carbon (ferritic) steels, the object of the present work is to finish rolling just above the austenite recrystallisation temperature, so as to obtain fine grained austenite and thus fine packet martensite on subsequent quenching.

With the increasing potential application for structural steels in a "dynamic" structure (where the resistance to fatigue plays an important role), such as machinery in the mining industry, the fatigue properties are also significant. The latter are now being analysed to avoid the possibility that an increase in K_{IC} toughness is obtained at the expense of fatigue properties. Results on QT⁴ alloys have shown that variations in fracture toughness (K_{IC}) from 65 MPam^{1/2} to 198 MPam^{1/2} with Mn content and tempering conditions, appear to have little effect on the "mid-range" of crack growth rates exceeding 5×10^{-6} mm/cycles (fig. 4). In addition, the experimental steels have better fatigue resistance in the medium crack growth range than commercial steels AISI 4340 and 300M for the same heat-treatment conditions such as in the as-quenched or 200°C tempered states, as shown in fig. 8. The determination of the fatigue limit of the leaner alloys, e.g., QT 2 with and without Nb is in progress and will not be discussed further here.

Many of the uses of structural steels require not only good strength-toughness characteristics, but also good wear corrosion resistance. As part of the alloy design program, both sliding and abrasive wear behavior have been examined and measured for our microcomposite steels (9-11). The results show that the duplex martensite/austenite microstructures exhibit good wear-resistance in both categories of wear, and are superior to many commercial alloys that are described as wear-resistant alloy. More recently, the sliding wear behavior of these steels

has been further enhanced by laser surface hardening. Using a 500 watt, continuous CO₂ laser, localised, rapid heating and quenching can be confined to the surface of the alloys, producing a hard, grain-refined microstructure to depths of 500 microns and hence, a two-fold increase in wear resistance (17). Consequently, materials with a fine-scale microstructure and a high yield strength in the absence of large undeformable particles, appear to be the most wear resistant.

In order to meet production hot mill plant requirements, plants with no accelerated cooling facilities will require steels with higher alloy compositions to be air hardenable in order to obtain the desired microstructures (fig. 2), e.g., Fe/10Cr/1Mn/0.2C (18). Such a steel is air hardenable to ~40 cm thickness. For plants with water cooling capabilities typical quench rates are ~30°C/sec. In this case leaner alloys with microalloying, e.g., 0.02% Nb (14) with Fe/2Cr/1Mn/0.25C compositions are suitable for final thicknesses up to 25-30 mm. In the latter case, by controlling the finish roll temperature (optimum is 900°C) and cooling (quench) rate, with arresting the quench above M_f, (to allow auto tempering) it is possible on a commercial plate mill to achieve the composite microstructures and properties on-line without subsequent tempering. This approach is important for many plate applications and represents potential cost savings compared to post rolling heat treatment procedures. Of course, subsequent tempering will always raise the toughness. The specific procedures adopted depend on the properties and applications desired.

3. Results and Discussion

Table 1 summarises some representative properties that have been obtained for a range of compositions optimised using the criteria described above. On-line rolling data are compared to laboratory results. Figures 7,a,b, show the strength-toughness-property relationships for the complete range of alloys studied. For on-line controlled rolling and cooling, figure 8 shows the marked effect of finish roll temperature on toughness. This temperature should be > 870°C, i.e., the recrystallisation temperature for this austenite. Typical grain sizes (prior austenite) are in the range ASTM 7-9. Figure 9 shows the microcomposite structure. The results are encouraging as the properties attainable on-line, in commercial plants are very good, and appear to be attainable at reasonable cost, without complex processing routes and to be tunable to specific applications. The mechanical properties are also well matched to other properties, e.g., good corrosion resistance at high strengths and toughnesses are expected for the QT 10 series of steels, (4), as well as wear resistance. Good weldability has already been demonstrated (14).

The stability of the microstructure determines the final range of applicability of any material. In the present series of steels, autotempering allows formation of intralath carbides without the interlath austenite decomposing to carbide, which, if this occurs, leads to temper martensite embrittlement. Then this structure very closely resembles upper bainite which of course must be avoided, so that the steels must have enough hardenability. The steels are temper resistant to the range 250°C-400°C depending on composition (6,7).

A model for the good toughness has been proposed earlier (7) and is illustrated in fig. 9. The coherent or semi-coherent austenite-lath martensite interface, as well known from the K-S crystallographic relationships (1), allows slip to cross both phases so large plastic zones can occur ahead of growing cracks. If, however, there is no "buffer" film of austenite, but rather interlath carbides, a small plastic zone is possible and hence, low K_{IC} values. Thus, the path of

growing cracks will be determined by the morphology of the structure (fig. 9). Mechanical stability of the austenite is also important. Silicon has been found to destabilize austenite near high stress fields, e.g., at crack tips. The austenite then transforms to high carbon twinned martensite which is also embrittling. Thus, silicon is not recommended as a deoxidiser, but rather Al. Also, since austenite has a high solubility for many elements, e.g. H₂, these steels may be quite tolerant to impurities.

Finally, by increasing carbon it may be possible to achieve very much higher combinations of hardness and toughness than are available at present.

Summary

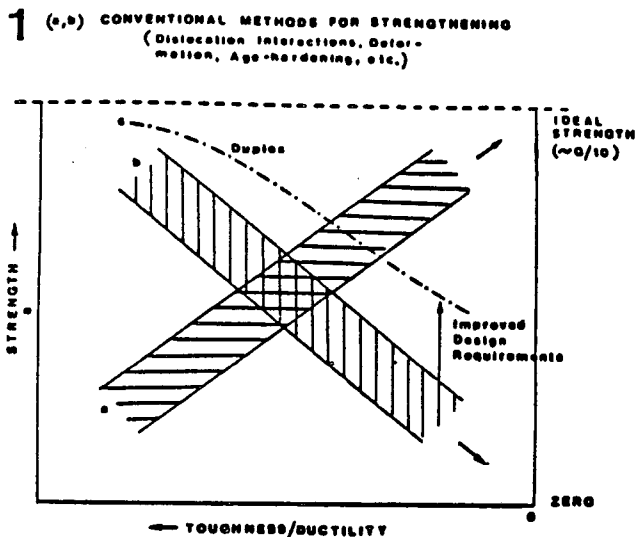
This research has indicated that microstructural control of steels is possible on-line in a modern hot mill so as to achieve microcomposites of martensite-austenite in a fine packet morphology. The benefits of Nb additions to leaner alloys has been demonstrated. Many commercial applications of these principles of the alloy design processing criteria appear to be feasible, and the results are attractive when compared to other existing steels (fig. 10) which may also be much more expensive, (e.g., maraging).

Acknowledgement

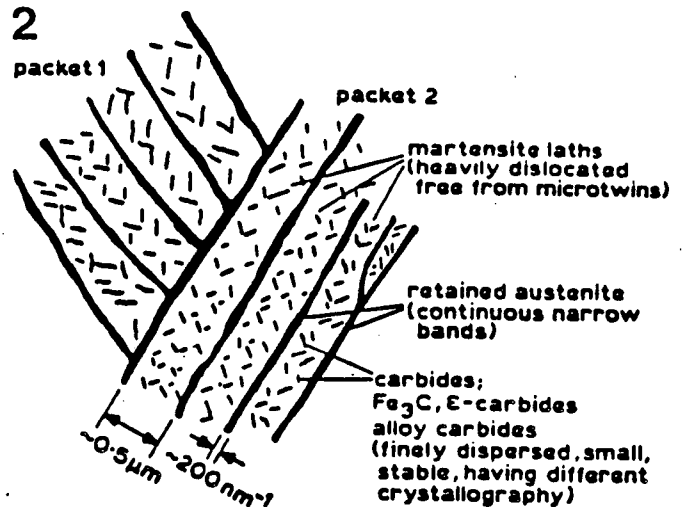
This work was supported by the Director, Office of Energy Research, Office of Basic Energy Sciences, MCSD Division, of the U.S. Department of Energy, under Contract No. DE-AC03-76SF00098. Thanks are due to all the researchers and graduate students who have been involved in this project for the past 15-20 years. Steels have been provided by various companies, notably Allegheny-Ludlum, ISCOR, South Africa, Posco, S. Korea. The cooperation of ISCOR and Dillingen (West Germany) for plant trials, and for making their data available is very much appreciated.

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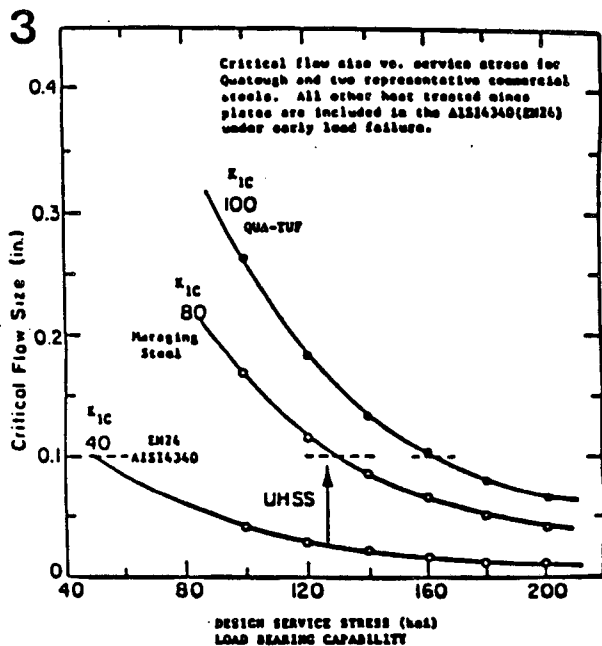
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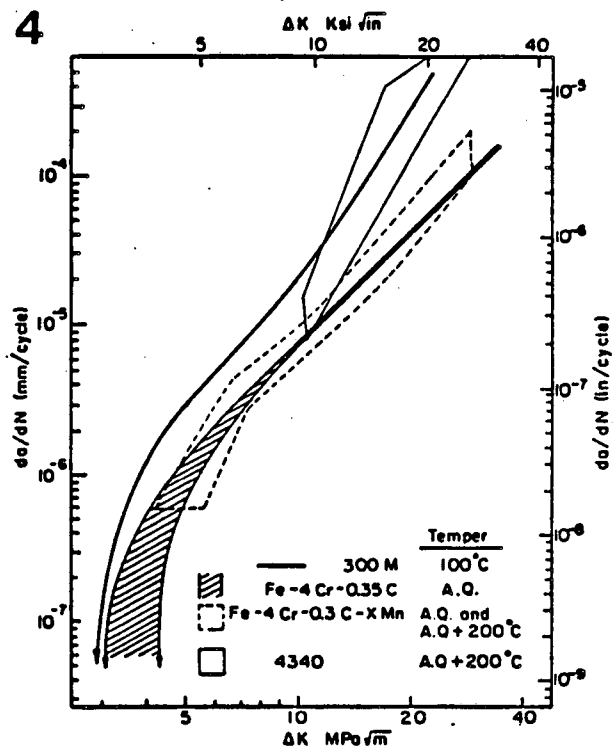
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Fig. 1 Schematic diagram illustrating conflicting properties of strength and toughness arising from conventional strengthening methods. By using microcomposite structures toughness can be improved.

Fig. 2 Schematic of desired microcomposite microstructure of autotempered packet lath martensite and continuous, stable, interlath martensite.



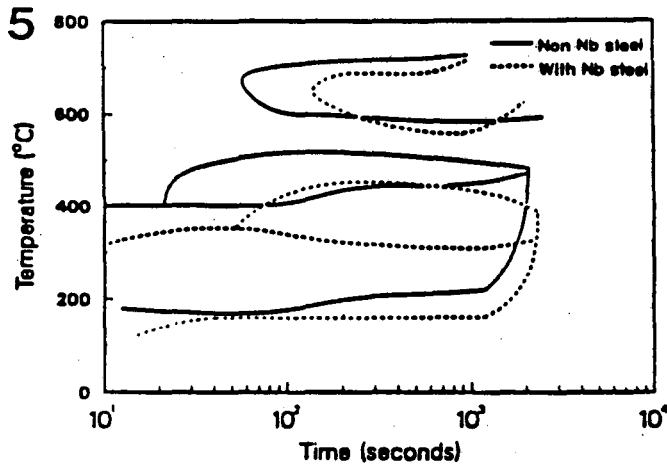
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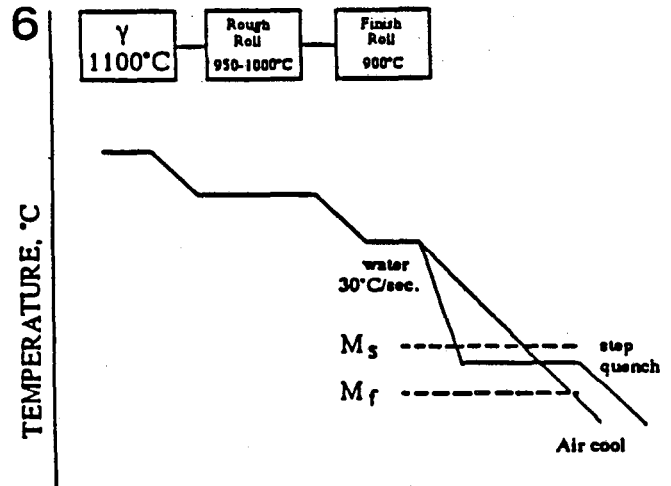
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Fig. 3 Plot of critical flow size vs. design stress for steels of similar yield strength.

Fig. 4 Fatigue crack growth data for QT 4, AISI 4340, and 300 M steels.



XCG 8710-11438

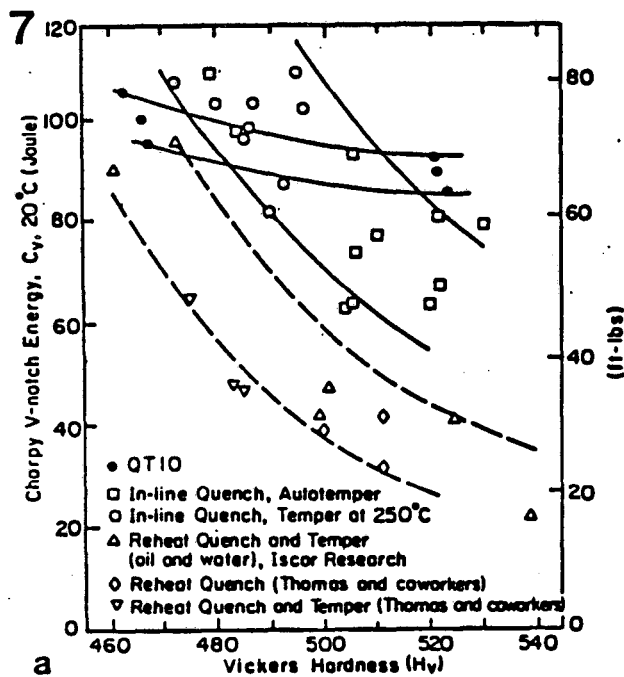


Recommended processing for on-line hot mill products such as plates, rounds, sections. If subsequent processing is required for W.Q. alloys steels can be normalized from finish rolling temp., processed and then Quench 900°C and temper 225°C.

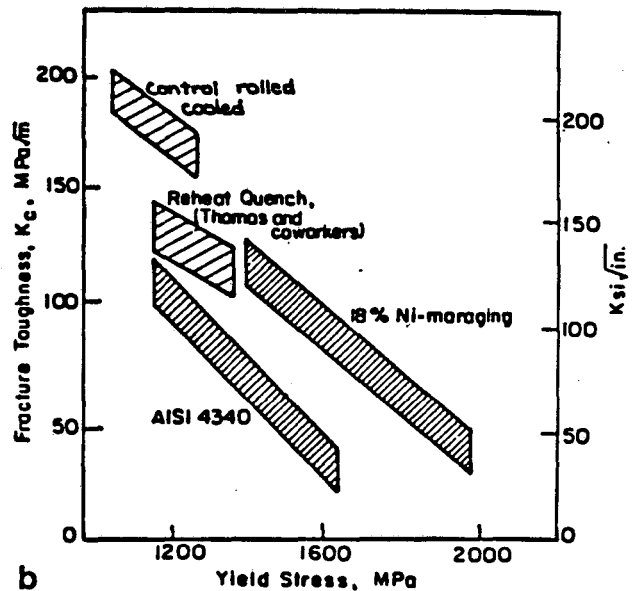
XBL 8810-3730

Fig. 5 CCT diagrams for QT 2, with and without Nb. (0.02%).

Fig. 6 Schematic showing processing routes for on-line production of microcomposite steels.

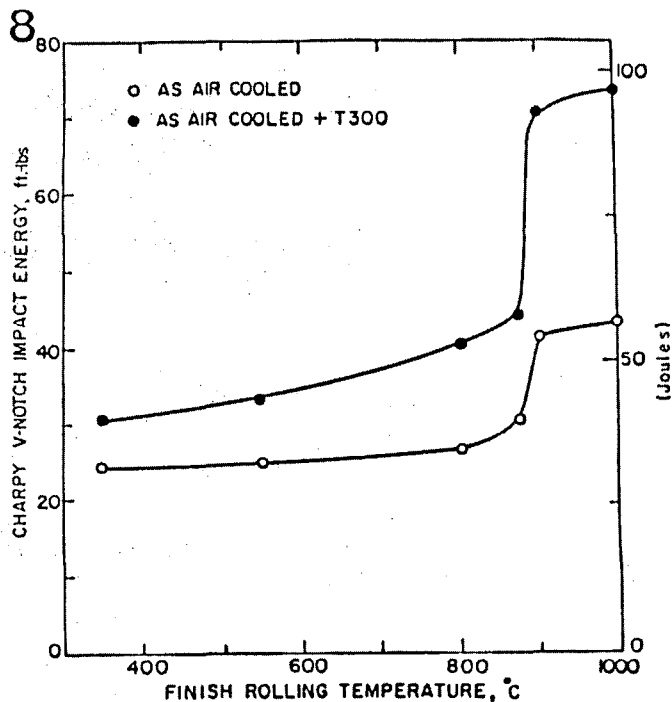


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XBL 859-6625A

Fig. 7 Summary of data for charpy (a) and K_{IC} (b) properties of QT steels.

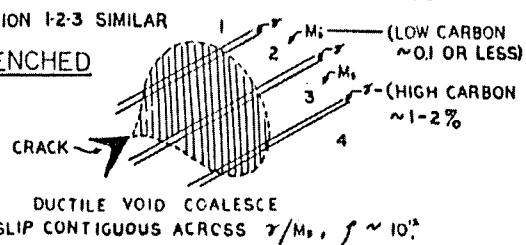


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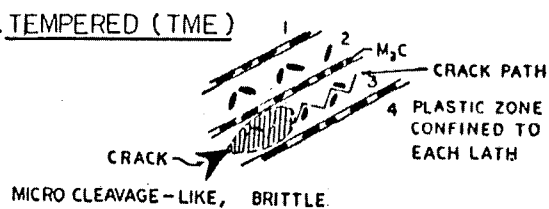
9 FRACTURE IN LATH MARTENSITE

ORIENTATION 1-2-3 SIMILAR

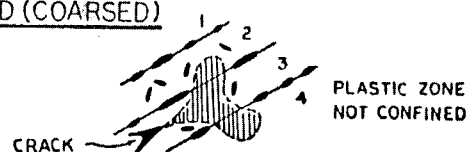
1. AS QUENCHED



2. TEMPERED (TME)



3. TEMPERED (COARSE)



MICRO SLIP

DUCTILE

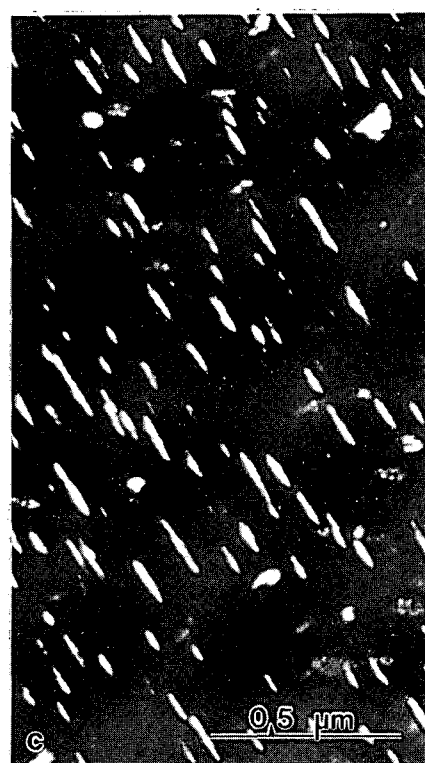
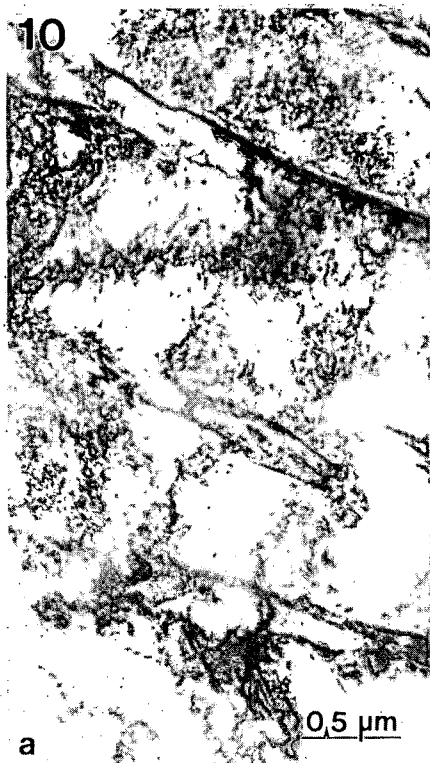
SLIP ON {111} PLANES WITHIN MARTENSITE IS ALSO RESTRICTED DUE TO {110} CARBIDES

PLASTIC ZONE DEFECT BY MICROSTRUCTURE (CARBIDE MORPHOLOGY)

XBL 824-9220

Fig. 8 Effect of finish roll temperature on charpy data for QT 10 steel.

Fig. 9 Schematic to show effect of microstructure on plastic zone and crack paths.



XBB 851-274

XBB 851-459

Fig. 10 Electron microscopy showing as cooled structure of QT 10 steel: (a) bright field, (b) dark field- interlath austenite, (c) darkfield- intralath M_3C .

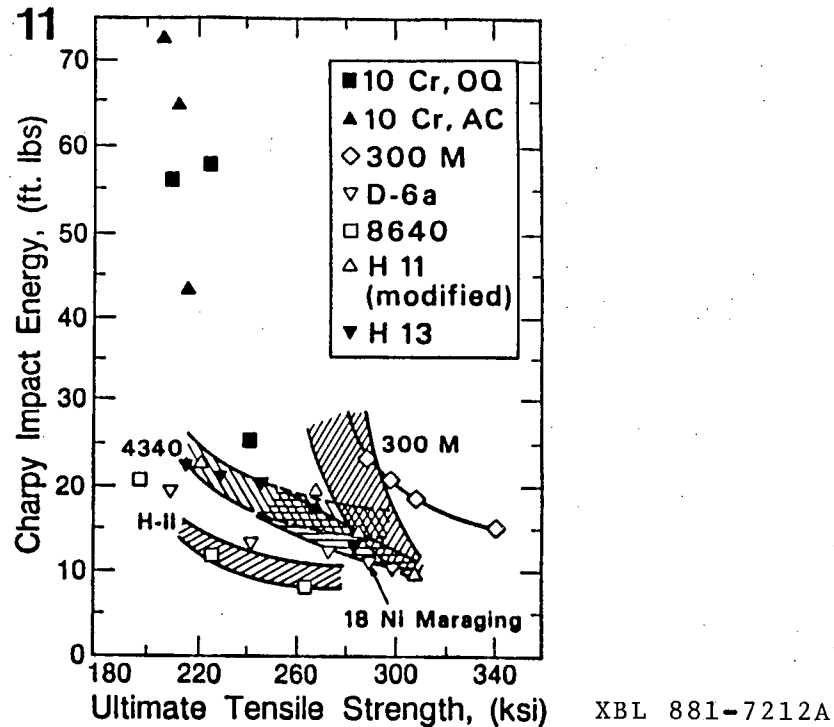


Fig. 11 Comparison of charpy toughness- UTS properties for commercial steels (data from ASM handbook Vol. 1, 9th ed.) and QT 10. Data for QT 2 is similar to QT 10.

Table 1

SOME REPRESENTATIVE MECHANICAL PROPERTIES OF Fe/Cr/Mn/C [QUATOUGH] STEELS

Alloy/Processing	Charpy Impact Energy, Joules (longitudinal)	UTS MPa	Yield MPa	Hardness R_c	K_{IC} MPa.m ^{1/2}	%Elongation	Applications
QT2, Fe-2Cr-.25C-1.2Mn							
QT2	90	1600	1375	45-49	130	13.3	rounds, bars, plate mining, fasteners agricultural eq., etc.
QT2, Q+200°C	120	1450	1111	45	145	13.3	
QT2+Nb(0.02) hot rolled*	92	1450	1192	45	155	15.0	
QT4, As Quenched Fe-4Cr-.3C-2Mn	39	1680	1465	49	ND	3.6	landing gear, struct. steel, ordnance, pressure vessels
Q + T200°C	55	1620	1340	49	196 (calc.)	6.5	
QT10 Fe-10Cr-.2C-1Mn							corrosion and wear resistant structural steel, pressure vessel mining industry.
hot rolled, OQ	25	1685	1400	49	ND	5.0	
OQ+T200°C	68	1515	1350	47	ND	15.0	
OQ+T200°C	55	1550	1260	47	ND	13.5	
hot rolled, AC	57	1485	1375	46	ND	9.0	
AC+T200°C	110	1450	1275	44	ND	16.0	

* Trials at Dillingen Huttenwerke Plate mill, West Germany.

OQ oil quenched ; AC air cooled

QT is the trademark "Quatough" for these steels.

The number after QT refers to the Cr content in wt. %.

ND, Not determined

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