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Invited keynote paper to be presented at the
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Gareth Thomas

September 1982

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DESIGN OF STRONG, DUCTILE, DUPLEX LOW ALLOY STEELS

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ABSTRACT

Strength, toughness, and ductility are undoubtedly the most important properties specified for structural steels unless the application involves aggressive environments. The major difficulty in optimising these properties 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. This is true in the majority of cases when relatively inexpensive alloying and processing are sought for a practical alloy development.

This paper is aimed at summarising our current alloy design programmes at Berkeley which utilise the concept of two-phase steels as a means of optimising these mutually exclusive properties. The underlying principle here is to utilise composites whereby the advantages of the second phase are optimised while the less desirable features of this phase are simultaneously mitigated by the presence of the other constituent phase. The size, distribution, shape and volume fraction of the second phase critically control the mechanical behaviour of the dual phase systems. As a consequence, these structures offer a degree of metallurgical flexibility that is absent in single phase structures or in many precipitation strengthened systems, for attaining optimum sets of mechanical properties.

Examples are presented here of martensite/austenite (~ 2-5%) mixtures designed for optimum combinations of high strength, toughness, and wear properties in medium carbon steels, e.g. for mining and agricultural applications, and martensite/ferrite (~ 80%) structure for high strength, good formability and improved low temperature ductility in low carbon steels. Applications to sheet, line pipe, and wires will be demonstrated for the latter class of steels.

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. This concept has been the basis for the alloy design programme for nearly two decades now at Berkeley, a programme which optimises ranges of mechanical properties through the characterisation and subsequent manipulation of the microstructure and chemical composition. Three different classes of steels have been utilised in the programme, representing three of some of the most important uses of steel in industry today:

- i) Structural steels as used in mining and agricultural industries, and in defence applications.
- ii) Steels used in the transportation industries, and
- iii) Steels used as rods, tensile wires and pipelines

These categories require high-strength for load-bearing, and high toughness and ductility to resist the propagation of cracks through the material and to ensure good formability. For mining and agricultural applications, wear and corrosion resistance are significant properties also.

Conventional manipulation of mechanical properties through microstructural control has shown that high strength and high toughness are often conflicting requirements, as an increase in strength is often accompanied by a reduction in the ductility and fracture toughness. Typical microstructural effects that show this trend are solute strengthening, precipitation hardening and

dislocation hardening. Nevertheless, two methods of microstructural control are available that can result in an increase in both properties--grain refinement and phase transformation through heat treatment. The principles of composites (viz mixing ductile and hard but tough phases with coherent interfaces) have formed the basis for the alloy design programmes we have developed. The careful selection of heat-treatment temperature, time and path has led to the development of the different classes of steel; these steels exhibit high strength and toughness as a result of a refined microstructure that is a composite of a hard, strong phase and a softer, tough phase. In the case of the structural steels, the steels are medium carbon alloys with a composite microstructure of austenite and martensite (or bainite), the major phase, whereas in the case of the steels used in transportation, rods, wires and in pipelines, they are lean, low-carbon steels with a composite microstructure of ferrite (major phase) and martensite (or bainite).

STRUCTURE-PROPERTIES

Strong, Tough, Wear-Resistant Medium Carbon Structural Steels. High strength structural steels are used extensively for components such as aircraft landing gear, missiles, rocket casing, armour plate and other defence applications. Moreover, where such steels have high hardness and consequently good abrasive resistance, they are used in mining and agricultural operations--e.g. as buckets and in comminution and mineral-processing operations. The limiting factors in the use of high strength steels is their fracture toughness, and in wet environments, their corrosion and wear.

As mentioned above, high-strength structural steels also require toughness and ductility to resist crack propagation and to ensure good formability for ease of fabrication. All too often, steels in use today have arrived on the

commercial market through a process of trial and error. Almost all such steels could benefit from increased levels of toughness.

With this object in mind, a systematic study of the relation between martensite microstructures and mechanical properties utilising a series of Fe-C-X high purity, vacuum-melted experimental alloys (where X is the substitutional solute) has been under way at Berkeley since 1964 (see Rao and Thomas, 1980; Thomas, 1978 for review). These steels have been termed "Qua-tough" steels. The martensite transformation, if controlled so that the inhomogeneous shear component occurs by slip rather than by twinning, is the most efficient method of producing a high density of dislocations, uniformly distributed, in a fine-grained microstructure (Thomas, 1977). The dislocations are a necessary component for both strength and toughness. Whether twinned or dislocated slip martensite occurs is determined mainly by the composition of the steel, in particular the carbon content through its effect in the transformation start temperature, M_s and on the strength of the martensite. The composition must be regulated to maintain $M_s > 200^\circ\text{C}$; this sets the carbon limit to about 0.3wt%. As a result of detailed analyses by electron microscopy and diffraction, retained austenite has been detected, although its presence was not expected. In addition it has been found that this retained austenite, if stable, promotes toughness. Consequently, the microstructure which appears to result in optimum mechanical properties, is a duplex austenite-martensite structure in which martensite laths are surrounded by thin films of retained austenite, see Fig. 1.

The excellent combinations of strength and toughness exhibited by the qua-tough steels can be seen even in the untempered condition, as shown in Figs. 2-3. Tempering in the 300-400°C temperature range leads to tempered martensite

embrittlement (TME), and the drop in toughness is associated with the decomposition of the retained austenite to interlath carbide. Recent studies (Jhingan, 1982; Sarikaya and others, 1981; Thomas and Sarikaya, 1981) have shown that the retained austenite is remarkably stable, the activation energy for austenite decomposition being ~ 74 kcal/mole, however the way in which austenite stabilisation occurs is not uniquely known. In an attempt to clarify the importance of chemical stabilisation of the austenite, the local chemical composition of the region adjacent to the austenite/martensite interface has been examined. Using transmission electron microscopy (lattice imaging and convergent beam microdiffraction), and atom-probe spectroscopy, it has been possible to show that the austenite is substantially enriched in carbon with a large peak in the carbon concentration found at the austenite/martensite interface even in the as-quenched condition (see Fig. 4). Carbon depletion has occurred in the martensite matrix, although the distribution is non-uniform due to autotempering. Although chemical stabilisation is obviously a factor in austenite stabilisation, mechanical stabilisation and the size effect (Easterling and Tholen, 1980), i.e. the retained austenite exists as thin films 50-250Å thick, also play important roles.

In addition to the microscopy experiments described above, in-situ fracture studies have been performed using thin foils of quenched, and quenched and tempered steels, in a high voltage electron microscope equipped with a tensile stage (Thomas and others, 1981). Fracture appears to occur exclusively within the martensite laths (depleted with respect to carbon). In the tough condition, fracture is preceded by localised dislocation slip, work-hardening, thinning down (necking) and hole formation. Packets of laths unfavourably oriented, do not participate in fracture initiation. In the tempered condition, slip and failure depends on carbide morphology. It is suggested that the fracture process depends primarily on the

plastic deformation characteristics of the martensite laths, which in turn are affected by the microstructure.

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 of the experimental qua-tough steels have been analysed to avoid the possibility that an increase in toughness is obtained at the expense of fatigue properties. The results have shown that variations in fracture toughness (K_{IC}) from $65 \text{ MPam}^{1/2}$ to $198 \text{ MPam}^{1/2}$ (induced by varying Mn alloying element and by 200°C tempering), appear to have little effect on the "mid-range" of crack growth rates exceeding 5×10^{-6} mm/cycles. 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 as-quenched and 200°C tempering (Kung and Thomas, unpublished).

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 programme, both sliding and abrasive wear behaviour have been examined and measured for the qua-tough steels (Kwok, 1982; Salesky and Thomas, 1981). The results show that the duplex martensite/austenite microstructure of these steels exhibits good wear-resistance in both categories of wear, and are superior to many, commercial alloys that are described as wear-resistant alloys (see Fig. 5). Sliding wear behaviour was found to be beneficially affected by tensile strength, hardness, fracture toughness, and grain-refinement, whereas abrasive wear was found to be positively affected by only the first two factors. More recently, the sliding wear behaviour of these steels has been further enhanced by laser surface hardening. Using a 500 watt, continuous CO_2 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. In addition to these studies, a more fundamental study has been carried out into the mechanisms of wear, specifically sliding wear (Salesky, 1982). The results suggest for several iron-based materials, tested in air, wear particles are formed by crack initiation and propagation along the boundary of dislocation cells present in a deformed region found adjacent to the wear surface (see Fig. 6). Oxidation then occurs during separation of the platelet from the parent material. 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.

Finally, the corrosion properties of some of the experimental steels have been examined and compared to commercial alloys as well as the dual phase steels discussed in the following section. The data obtained are shown in Table 1. The 4wt% chromium steels give good corrosion resistance as the table shows, and have greater corrosion resistance than 4340, a structural steel of similar properties (data obtained at Battelle, Northwest). Once again the qua-tough steels show superior mechanical properties over equivalent commercial alloys.

Dual-Phase, Low-Carbon Steels. During the past several years, dual-phase ferrite-martensite steels have received increasing attention (Davenport, 1981; Kot and Morris, 1979; Kot and Bramfitt, 1981), through their possession of characteristic microstructural features which combine high strength and ductility with excellent formability. The strengthening principle of the dual-phase structure involves the incorporation of inherently strong martensite, as the load-carrying constituent, in a soft, ferrite matrix, which supplies the system with the essential ductility. Thus these alloys can be considered equivalent to fibre composites, with the advantage of producing coherent phases by solid-state phase transformation (Fig. 7a). The alloy design principles followed at Berkeley have been described in

detail by Thomas and Koo (1979) and will not be repeated here. However, it is emphasised that the properties of dual-phase steels are largely dependent upon three factors:

- (i) morphology--size, shape and distribution of both the ferrite and martensite, (e.g. Fig. 7b).
- (ii) the properties of the two phases, and
- (iii) the volume fraction of martensite (or bainite).

Of these factors, morphology plays the most important role in controlling ductility (Kim and Thomas, 1981). It has been shown that a coarse dual-phase structure has very poor tensile elongation ductility. This is due to cleavage crack initiation and propagation within the ferrite matrix where the maximum stress concentration takes place (Fig. 8). However, when the microstructure is refined, the degree of stress concentration decreases, and fracture is initiated by void nucleation and growth around the ferrite/martensite interfaces. These voids, once initiated, propagate within the ferrite matrix without fracture occurring within the martensite (Fig. 9). The flow strength, on the other hand, appears to be less sensitive to morphology, and depends approximately, and in an empirical fashion, on the "law of mixtures" given below,

$$\sigma_c = \sigma_M V_M + \sigma_\alpha (1 - V_M)$$

where σ_M and σ_α are the strengths of martensite and ferrite respectively, and V_M and $(1 - V_M)$ are their corresponding volume fractions. In general, there is a linear increase in strength as a function of volume fraction of martensite. However, in some alloy systems (Hoel and Thomas, 1981) which contain carbide-forming elements such as Nb and Mo, the strength decreases or remains constant with

increasing martensite volume fraction; this represents an anomaly in the usual "law of mixtures" behaviour. If the volume fraction of martensite has been increased by raising the annealing temperature within the ($\alpha + \gamma$) field, then two ways are possible wherein the strength of the steel can be reduced. The first is by a decrease in the carbon content of the ferrite, which leads to a decrease in the density of carbide precipitation in the ferrite (Fig. 10), and the second is by a decrease in the carbon content of the martensite, thus lowering its strength. The overall effect, which depends on the relative magnitude of these factors, can result in a negative slope in the "law-of-mixtures" equation (Eq. 1), if the strength of the ferrite changes significantly. (For more details, see Thomas and Koo 1979.) Thus, the complex interactions of parameters such as morphology, properties and volume fraction of constituent phases must all be controlled when designing dual-phase alloys. When this is achieved, excellent mechanical properties can be realised, with high strengths attainable for relatively low carbon alloys (see Fig. 11). Moreover, the alloy compositions are quite simple, with no need for microalloying elements, and are therefore inexpensive, and by controlling the composition, considerable flexibility (i.e. strict heat treatment is not needed) in intercritical annealing temperature is possible (Thomas and Koo, 1979).

The main principles of alloy development for high formability as used in sheet, rods, wire, etc. can be summarised as follows:

- (1) The ferrite should contain mobile dislocations, and should therefore be 'clean' and free from extensive pinning.
- (2) The percentage of carbon in the martensite should not exceed 0.3wt%, in order to obtain a strong, tough martensite (as illustrated for the 0.3C steels in the previous section).
- (3) A fine distribution of martensite (or bainite) particles should be present

in the ferrite matrix; the optimum volume fraction for good ductility and formability is ~20% martensite, but for high strength, without the high formability requirements, a greater percentage of martensite can be used. Thus the properties of a dual-phase steel can be "tailor-made" to suit the need, by the appropriate heat-treatment (Fig. 7a).

Utilising these principles several ranges of alloys have been developed, starting with a simple Fe/Si/C alloy, which has the fibrous structure shown in Fig. 7b, and has excellent properties (Fig. 11). The figure summarises our data in comparison with data for some high strength low-alloy (HSLA) steels that contain expensive microalloying additions. Although interest in dual-phase steels has been generated largely by the recent fuel crisis and its impact on the need for weight savings in transportation systems, i.e. for flat sheet products, the potential application for dual-phase alloys are broad indeed. These alloys form a new class of strong, ductile steels for applications requiring tensile strength levels of ~100ksi for pipeline, and up to 400ksi for tensile wires, rods and bars. These properties can be obtained by cold wire drawing without the necessity for intermediate patenting heat treatments (Nakagawa (1982). Strengths up to 400ksi are indeed remarkable for such simple steels with less than 0.1wt% carbon.

Steels used for pipeline must have high strength as well as high toughness (low ductile-to-brittle transition temperature). The alloy design principle for pipeline application is that high toughness can be obtained primarily by grain size control, while maintaining a dual-phase structure to simultaneously obtain high strength. Controlled rolling of an Fe/1.5Mn/0.06C alloy, followed by direct quenching produces a duplex microstructure in which upper bainite particles are

uniformly distributed within the fine, ferrite matrix (Kim and Thomas, 1982). The mechanical properties of this ferrite-bainite steel, in the as-hot-rolled condition, far exceed the property specifications for arctic pipeline. A further advantage of this ferrite-bainite steel is that there is an abrupt increase in strength during pipe-forming due to its high work-hardening rate. This clearly demonstrates the potential use of dual-phase ferrite-bainite structures in pipeline applications (Table 2).

In conclusion, from the standpoint of their superior properties, and simplicity in composition and processing, dual-phase steels show great promise, not only for the transportation industry, but for general structural applications, e.g., pipeline, wires, rods and bars. Their advantages include economy of material and enormous flexibility wherein a single alloy can be manipulated to provide a wide range of products and properties. The main limitations of low carbon dual phase steels for general applications are material dimensions, i.e. sufficient hardenability must be achieved (~ 0.5 in thickness/diameter) to obtain martensite/bainite on quenching from the ($\alpha + \gamma$) field, and problems of microstructure control if welding must be done after the dual phase treatment.

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REFERENCES

- Davenport, A. T. (Ed.), (1979). Formable HSLA and Dual-Phase Steels. A.I.M.E., New York, New York.
- Hoel, R. H., and G. Thomas (1981). Scripta Metall., 15, 867.
- Jhingan, A. (1982) Ph. D. Thesis. University of California, Berkeley, California.
- Kim, N. J., and G. Thomas (1981). Metall. Trans., 12A, 483.
- Kim, N. J., and G. Thomas (1982). Metall. Trans., (in press).
- Koo, J. Y., and G. Thomas (1977). Metall. Trans., 8A, 525.
- Kot, R. A., and J. W. Morris, Jr. (Eds.), (1979). Structure and Properties of Dual-Phase Steels. A.I.M.E., New York, New York.
- Kot, R. A., and B. L. Bramfitt (Eds.), (1981). Fundamentals of Dual-Phase Steels. A.I.E., New York, New York.
- Kwok, S. (1982) M. S. Thesis. University of California, Berkeley.
- Nakagawa, A. (1982). Ph.D. Thesis in progress, University of California, Berkeley.
- Rao, B. V. N., and G. Thomas (1980). Metall. Trans. 11A, 441.
- Salesky, W. J. and G. Thomas (1981). Wear, 75, 21.
- Salesky, W. J. (1982). Ph. D. Thesis. University of California, Berkeley.
- Sarikaya, M., G. Thomas, J. W. Steeds, S. J. Barnard and G. D. W. Smith (1981). A.S.M. Int. Conf. on Solid-Solid Phase Transformations, Pittsburgh, PA.
- Thomas, G. (1977). In R. I. Jaffee and B. A. Wilcox (Eds.), Fundamental Aspects of Structural Alloy Design. Plenum Press, New York. p. 331.
- Thomas, G. (1978). Metall. Trans., 9A, 439.
- Thomas, G., and Koo, J. Y. (1979). In Structure and Properties of Dual-Phase Steels, A.I.M.E. New York, p. 183.

Thomas, G., and M. Sarikaya (1981). A. S. M. Int. Conf. on Solid-Solid Phase Transformations, Pittsburgh, PA (in press).

Thomas, G., M. Sarikaya, S. J. Barnard and G. D. W. Smith (1981). Inst. of Metals (London), Proceedings Liverpool Conference on Advances in Physical Metallurgy of Steels (in press).

Table 1

CORROSION* DATA FOR LOW CARBON QUA-TOUGH AND
AND SOME COMMERCIAL STEELS

<u>Specimen</u>	<u>Condition</u>	<u>Weight Loss</u> <u>Mg/Decimeter²</u>	<u>MmPY</u> <u>Mm Per</u> <u>Year</u>
Fe-2Si-0.1C	As-received	2091.6	1.70
	Duplex Ferrite- Martensite	1956.4	1.59
AISI 1010	As-received	2077.2	1.68
	Duplex Ferrite- Martensite	1946.9	1.58
Fe-4Cr-0.3C-2Mn	As-quenched	1173.5	0.95
	200°C Tempered	988.5	0.80
Fe-4Cr-0.3C-5Ni	As-quenched (Coarse grained)	926.1	0.75
	As-quenched (Grain refined)	774.5	0.63
AISI 4340	As-received	--	2.79.
AISI 410 (Stainless steel)	As-received	--	3.05

* 15% NaCl, 100 ppm CO₂ (pH 4.4) at 250°C
(Courtesy of Battelle Northwest and R. Clark).

Table 2
PROPERTY SPECIFICATIONS FOR ARCTIC PIPELINE AND TYPICAL MECHANICAL
PROPERTIES OF THE AS-HOT-ROLLED Fe/1.5Mn/.06C STEEL
WITH AND WITHOUT COLD WORKING

	UTS MN.m ⁻²	Charpy shelf energy	
		Transverse	Longitudinal
Arctic Pipeline	585.7	92 J. at -60°C	136 - 170 J. at -60°C
Fe/1.5Mn/.06C (as hot-rolled)	642.2	149 J. up to -100°C	183 J. up to -100°C
Fe/1.5Mn/.06C (with 2% cold reduction)	649.0	129 J. up to -100°C	170J. up to -150°C

Fig. 1 - Bright-field (a) and dark-field (c) TEM micrographs of Fe-3Cr-2Mn-0.3C steel in the as-quenched condition revealing the general appearance of laths and retained austenite. (b) Shows the selected area electron diffraction pattern (After Sarikaya and Thomas (1981), courtesy of A.S.M.)

Figs. 2 and 3 - Mechanical properties summary showing toughness to strength relationship in the experimental 0.3%C alloys and equivalent commercial alloys: Fig. 2: CVN impact energy vs. tensile strength; Fig. 3: Plane strain fracture toughness (K_{1C}) vs. tensile strength (After Sarikaya, Rao and Thomas)

Fig. 4 - (a) Field ion-atom probe analysis of C, Cr and Mn of Fe-3Cr-2Mn-0.5Mo-0.3C after tempering at 200°C for 1 hr; (b) Atom probe analysis of C in Fe-4Cr-0.34C alloy following quenching 0.5mm dia. sample in iced water. (After Sarikaya and others, 1981)

Fig. 5 - Histograms of relative wear resistance of experimental and commercial alloys: (a) sliding wear resistance; (b) two-body abrasive wear resistance. (Courtesy of W. J. Salesky and S. Kwok)

Fig. 6 - Electron micrograph showing abrasive worn quatough steel (below arrow). The martensite laths follow the wear direction. The arrow shows the interface above which nickel has been plated to preserve the worn surface. (Courtesy of W. J. Salesky)

Fig. 7a - Sketches to show principles of dual phase steels. The volume fractions of ferrite and austenite (and ultimately martensite or bainite) depend on carbon content and intercritical annealing temperature. Several transformation paths are indicated.

Fig. 7b - Optical micrograph showing the fibrous ferrite-martensite composite microstructure with prior austenite grains in a dual-phase Fe-Si-0.1C steel. (After Thomas and Koo (1979), courtesy of A.I.M.E.)

Fig. 8 - Scanning electron micrograph of step-quenched coarse dual-phase structure showing the cleavage cracks developed in ferrite. M: martensite; F: ferrite. The stress axis is normal to the micrograph. (After Kim and Thomas (1981), courtesy of A.I.M.E.)

Fig. 9 - Scanning electron micrograph of intercritically annealed fine globular structure showing crack propagation in the ferrite matrix. The stress axis is normal to the micrograph. (After Kim and Thomas (1981), courtesy of A.I.M.E.)

Fig. 10 - Transmission electron micrograph showing the presence of fine homogeneously nucleated precipitates in the ferrite phase of a Mo-bearing dual-phase steel. These precipitates strengthen the ferrite phase. (After Hoel and Thomas (1981), courtesy of Scripta Met.)

Fig. 11 - strength vs. tensile ductility for Fe-2Si-0.06C dual phase steel compared to HSLA steels. Notice the superior properties and higher tensile to yield ratios (high formability) typical for low carbon dual phase steels (After Thomas and Koo (1979), courtesy of A.I.M.E.)

XBB 796-7880

Fig. 1

XBL 8010-6229

Fig. 2

XBL 8010-6228

Fig. 3

XBL 818-10992

Fig. 4

XBL7911-13211

Fig. 5a

Fig. 5b

XBL 813-8719

Fig. 6

XBB 827-5916

XBL79I-5511

Fig. 7a

XBB 798-10688

Fig. 7b

XBB 797-9314

Fig. 8

XBB 804-4089

Fig. 9

XBB 828-7052

Fig. 10

Fig. 11

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