

# Lawrence Berkeley National Laboratory

## Recent Work

### Title

DESIGN OF LOW CARBON DUAL PHASE STEELS FOR BARS AND WIRE RODS

### Permalink

<https://escholarship.org/uc/item/5ts090j8>

### Author

Thomas, G.

### Publication Date

1985-10-01



# Lawrence Berkeley Laboratory

UNIVERSITY OF CALIFORNIA

RECEIVED  
LAWRENCE  
BERKELEY LABORATORY

## Materials & Molecular Research Division

14 1986

ARY AND  
SECTION

To be published in the Proceedings of The  
Metallurgical Society/American Institute of  
Metallurgical Engineers Meeting, E.R. Parker  
Symposium, New Orleans, LA, March 2-6, 1986

DESIGN OF LOW CARBON DUAL PHASE STEELS FOR BARS  
AND WIRE RODS

G. Thomas

October 1985

**TWO-WEEK LOAN COPY**  
*This is a Library Circulating Copy  
which may be borrowed for two weeks.*



LBL-20482  
c.2

## **DISCLAIMER**

This document was prepared as an account of work sponsored by the United States Government. While this document is believed to contain correct information, neither the United States Government nor any agency thereof, nor the Regents of the University of California, nor any of their employees, makes any warranty, express or implied, or assumes any legal responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product, or process disclosed, or represents that its use would not infringe privately owned rights. Reference herein to any specific commercial product, process, or service by its trade name, trademark, manufacturer, or otherwise, does not necessarily constitute or imply its endorsement, recommendation, or favoring by the United States Government or any agency thereof, or the Regents of the University of California. The views and opinions of authors expressed herein do not necessarily state or reflect those of the United States Government or any agency thereof or the Regents of the University of California.

**DESIGN OF LOW CARBON DUAL PHASE STEELS FOR BARS AND WIRE RODS****Gareth Thomas****Professor, Department of Materials Science and Mineral Engineering  
University of California  
Berkeley, CA 94720****and****Scientific Director, National Center for Electron Microscopy  
Materials and Molecular Research Division  
Lawrence Berkeley Laboratory  
Berkeley, CA 94720  
USA**

Low carbon dual phase steels can be designed of simple compositions to yield products which are highly cold formable. In this paper specific attention is paid to the production of dual phase steel bars and rods and especially to the potential of the latter for wire drawing which will not require patenting heat treatments.

## Introduction

The terminology "dual phase" has become accepted to define a class of low-carbon low-alloy steels (1) that generally contain two phases, ferrite and martensite or ferrite and bainite, which can be obtained by relatively simple processing involving quenching from the two phase ferrite-austenite field. Strictly speaking, of course, these steels may actually contain more than two phases. If the martensite is low carbon lath martensite, it will contain interlath austenite and if high temperature decomposition of austenite has occurred during quenching, i.e., if bainite is present, there will be bainitic carbides. In microalloyed steels the ferrite phases may also contain the corresponding alloy carbides or nitrides. Thus, whilst the term "dual phase" represents the simple principle of obtaining two phases, the actual microstructures may be multiphase and quite complex. However, in principle these steels can have simple compositions and are economically attractive, and can be designed to provide excellent combinations of cold formability, strength and ductility. Thus they represent an exciting development in steel metallurgy over the past decade (2-5) and the basic physical metallurgy and alloy design aspects have been reviewed recently (6).

Alloy design trends for structural applications have focussed attention on low-carbon, low-alloy steels, thus enabling greater exploitation of their potential for higher strength along with sufficient ductility for a wide range of applications. As an outgrowth of this movement, for example, conventional hot rolled low-carbon steels which were adequate to fill the material requirements for the majority of structural applications for many years are progressively being replaced by high-strength, low-alloy (HSLA) steels. The principal strengthening of these steels is derived from precipitation of finely dispersed alloy carbides and grain refinement (7). However, their overall mechanical properties are not always satisfactory for many applications (e.g., formability).

The search for an alternative has spurred the recent development of duplex ferritic-martensitic (DFM) steels. These are a new class of HSLA steels whose approach to strengthening contrasts markedly with microalloyed HSLA steels in chemistry as well as processing technique, in that the principles of composites have been utilized (6) rather than traditional strengthening methods (8).

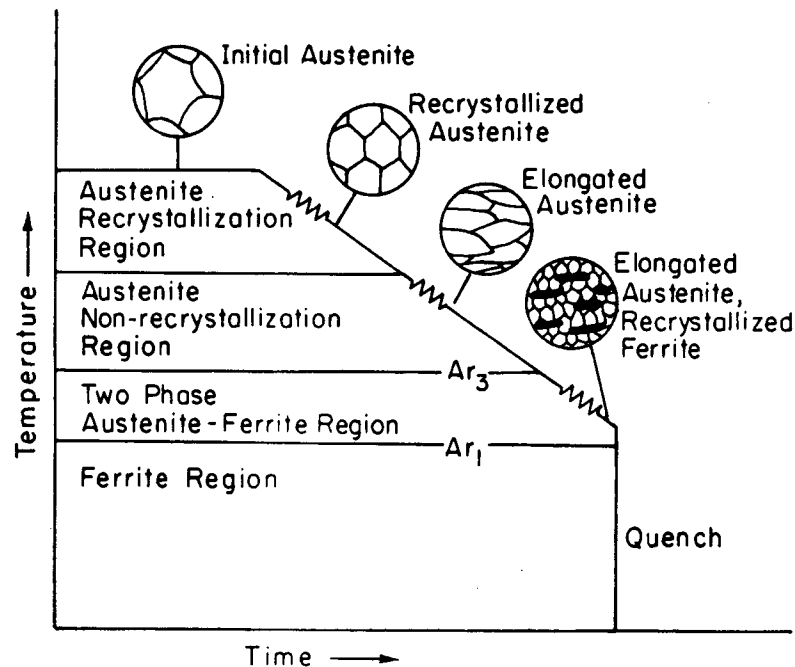
As is now well appreciated, dual phase steels have mechanical properties which are characterized by continuous yielding with high initial work hardening rates, large uniform strains and high tensile to yield ratios. These factors account for their better formability compared to that of ferritic, pearlitic or HSLA steels of similar strengths and allow attractive combinations of strength and ductility to be obtained.

The major source of strengthening in the DFM structure arises from the presence of inherently strong martensite as a load carrying constituent in a soft ferrite matrix which supplies the system with the essential element of ductility. The resulting mixture is analogous to that of a composite but which can be obtained solely by heat treatment or thermal-mechanical processing. The concept of fiber-composite strengthening is thus useful in qualitatively understanding dual phase steels.

Undoubtedly, the occurrence of DFM aggregates is one of the oldest phenomena in the history of martensitic transformations in steel since these mixed microstructures are produced as an unavoidable consequence of

incomplete austenitization and/or lack of sufficient hardenability. In this regard, earlier investigators, e.g., Herres and Lorig (9) considered two phase aggregates to be undesirable microstructural features rather than potential strengthening devices in low carbon steels. The idea of the DFM structures as possible materials of technological interest was recognized by Cairns and Charles (10) in 1967 who produced controlled microstructures of elongated regions of martensite in a ferrite matrix either synthetically or by a combination of cold deformation and rapid re-heating. However, this technique suffers from various disadvantages and the resultant mechanical properties were no better than could be obtained from suitably heat-treated alloy steels. In contrast, Grange (11) utilized the phase transformation occurring in the two phase ( $\alpha$ +V) range in Fe-C phase diagram to obtain fibrous mixtures of martensite and ferrite in various carbon steels by thermo-mechanical treatments. Again, the processing technique is complex and only limited success in improving mechanical properties was achieved.

There is no doubt that the fuel crisis and pressure on the automotive industry fuel economy spurred the initial development of dual phase steels for flat rolled sheet products. Owen (12) has given a review of the dual phase steel potential for automotive industry and no attempt will be made to review this again here. Very recently, the potential of dual phase steels for applications other than flat rolled products, e.g., line pipe, bars, rod, wire, etc., has been realized because these steels can be designed to optimize the ever conflicting property requirements of strength and ductility.



XBL 852-5872

Figure 1 - Schematic of the controlled rolling treatment. (1) Rough rolling in austenite recrystallization region. (2) Intermediate rolling in austenite non-recrystallization regions. (3) Finish rolling (e.g. in a rod mill no twist block).

Thin foils for transmission electron microscopy of the processed steels were prepared by first chemical thinning slices from the bars to a thickness of 0.002 inch (0.05 mm). After 3 mm discs were punched from the slices, they were polished to electron transparency with a twin jet electropolisher using a solution of 75 grams  $\text{CrO}_3$ , 400 ml  $\text{CH}_3\text{COOH}$  and 21 ml of distilled water. Longitudinal sections of the wires were prepared by grinding two sides of the wire to make a thin strip 0.002 inch (0.05 mm) thick and as wide as the original wire diameter. This strip was then polished by the window technique in a solution of 0.5 kg anhydrous sodium chromate and 3.8 l glacial acetic acid using a polishing potential of 25 to 35 volts. Smaller wires were thinned without masking the edges of the strip. The thin foils were examined with a Philips EM 301 transmission electron microscope at an accelerating voltage of 100 kV.

Tensile testing was done on all 5.5 mm bars using standard methods. The wires were tested in 3.5 inch (89 mm) lengths with 1 inch (25.4 mm) gauge lengths. The strain rates were 0.005 to 0.02/minute. Wires larger than 0.526 inch (1.34 mm) in diameter were tested in self tightening wedge grips. In this case the gauge diameter was carefully reduced by 10 pct using 600 grit SiC, to prevent failure in the grips. Smaller diameters were tested in screw tightened grips. The jaws were lined with aluminum strips to prevent failure in the grips. The end of the wires were roughened with coarse emery paper to prevent slipping. The tensile properties are shown in Tables III and IV.

Table I. Alloy Compositions (wt%) and  
Phase Transformation Temperatures

| Alloy | C     | Si   | Mn   | P     | S     | Fe  | $A_{c_3}$ (°C) | $A_{r_3}$ (°C) |
|-------|-------|------|------|-------|-------|-----|----------------|----------------|
| A     | 0.08  | 1.89 | 0.32 | 0.004 | 0.004 | bal | 1030           | 940            |
| B*    | 0.084 | 1.05 | 1.62 | 0.016 | 0.021 | bal | 850            | 760            |
| C     | 0.08  | 0.25 | 1.08 | 0.004 | 0.005 | bal | 830            | 740            |

\* Commercial Welding Rod: samples provided by Stelco, Canada.  
Received as 5.5 mm rod.

#### Cold Drawing of DFM Bars to Wire

Laboratory drawing trials involved starting with rolled 5.5 mm rod or by machining larger diameter, rods (Table I) to 5.5 mm dia. and then drawing without coating in a hand operated drawing machine using 6°-8° semi-die angle conical carbide and diamond dies lubricated with a Dupont Vydax Freon-Teflon dispersion. The reduction in area per pass started at 35% and then was gradually reduced to 20%. This technique differs from conventional drawing of pearlitic steel because of the high initial work hardening behavior of DFM steels. If the reductions are too small central burst cracking can occur (13). Table II indicates the drawability of 5.5 mm wire rods produced from these steels by the controlled rolling process of Fig. 1, except for alloy B which was processed by heat treatment (13).

Although the principles behind the alloy design of dual phase steels has been described earlier (2-6) the main emphasis here is to produce maximum cold formable steels. The formability depends upon morphology volume fraction and type of martensite and in the case of cold wire drawing, from rods, which is conventionally done on pearlite steels (containing up to 0.8% C for high strength) which are also of course composites, the hard phase cementite imposes a drawing limit which normally requires "patenting" heat treatments to maintain further drawing. This process can be obviated in dual phase steels provided the second phase is tough, lath martensite. On the other hand wire drawing is especially difficult if hard particles of twinned martensite (or of course inclusions from the steel making process) are present (13,14). Thus it is necessary to control composition and processing to achieve about 20% lath dislocated fibrous martensite in a fine grained and continuous equiaxed ferritic matrix in order to achieve maximum drawability. This paper discusses some of our recent results of such research on DFM bars and rods for wire drawing.

### Experimental Procedure

#### Production of Dual Phase Steel Bars and Rods

In order to be economical it is desirable that the dual phase steel should be produced directly on line in a hot rolling mill although much research has also been done on heat treatment processes (2-6). In the present research program a rolling schedule has been developed to simulate that in a rod or bar mill, but modified as illustrated schematically in Fig. 1. The important features are to control the microstructure by finishing rolling in the ( $\alpha$ + $\gamma$ ) phase field, i.e., just below the  $A_{r3}$  temperature such that unrecrystallized elongated austenite and fine recrystallized ferrite are formed prior to the water quench. In this way a final microstructure of fibrous martensite in a fine equiaxed ferrite matrix can be attained. Examples are shown in Fig. 2. The usual starting size was 2.5 cm dia. and the rolling schedule was done so as to achieve a final rod diameter of 5.5 mm. Such a process is achievable commercially in typical bar and rod mills. In the latter case, however, due to the high speed of a modern mill (100 m/sec), temperature control can be difficult and large volumes of water are required.

In order to finish rolling in the ( $\alpha$ + $\gamma$ ) field the rolling temperature is determined primarily by the composition. Representative compositions of some of the steels used (see ref. 6 for background) and such temperatures are given in Table 1. A commercial welding rod was also studied (alloy B). Since this rod was supplied as 5.5 mm diameter in size control rolling could not be performed. So this alloy was heat treated to produce dual phase structures as described previously (13).

One advantage of the silicon steels is their higher  $A_{r3}$  temperatures. The Mn dual phase steels require fairly low finishing temperatures, but are easily continuously cast.

#### Metallography and Mechanical Testing

Specimens for scanning electron metallography were cut from the processed bars. They were mounted, hand ground and polished using 1 micron diamond paste. The microstructural features were revealed by etching in a 2 pct Nital solution and observed on an ISI DS-300 scanning electron microscope operated at 25 kV.



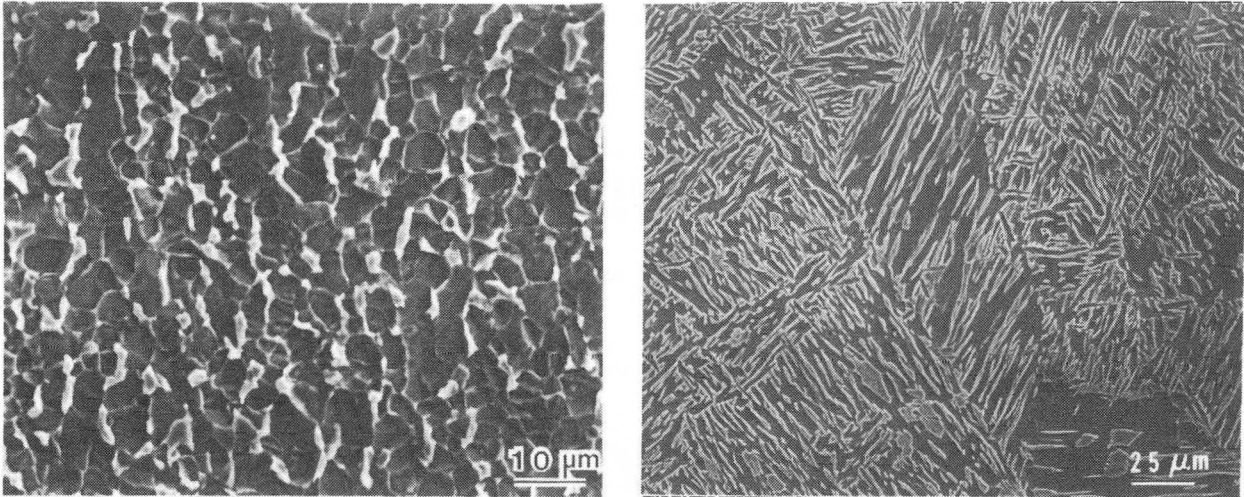


Figure 2 - Scanning electron micrographs of dual phase wire rods, (a) alloy A1, (b) alloy B. XBB 850-8429 XBB 852-1507A

### Results

#### Potential of Dual Phase Steel for Bars

As can be seen from Table III, and from the literature (1-6), a range of yield and tensile strengths are possible in DFM steels depending on compositions and processing. In addition, DFM steels have much greater ductilities, excellent low temperature fracture properties (15) (e.g., DBTT  $< -100^{\circ}\text{C}$ ) and better fatigue resistance (16) at strengths superior to HSLA or microalloyed steels. The tensile properties of the as-processed steels (bars, rods) are summarized in Table III. Data show that these properties follow the law of mixtures, viz., the strength increases while the ductility generally decreases with increasing the martensite volume fraction. Fig. 3 compares the tensile properties for a dual phase steel (such as alloy A) with yield strength of 80 ksi (550 MPa), referred to now as Fy80 with UTS of 120 ksi (830 MPa) with that of a typical mild steel of 60 ksi (415 MPa) now referred to as Fy 60.

Table II. Drawability

| Specimen | Heat Treatment | Vol.Pct. Martensite | Min. dia. (in.) | Total True Strain |
|----------|----------------|---------------------|-----------------|-------------------|
| A1       | CHR            | ~20%                | $< 0.0105$      | $\times 6.05$     |
| A2       | IQ             | ~20%                | $< 0.0105$      | $\times 6.05$     |
| A3       | IQ             | ~40%                | 0.0136          | 5.54              |
| A4       | IQ             | ~50%                | 0.0284          | 4.07              |
| B        | IQ             | ~30%                | 0.0136          | 5.54              |
| C        | CHR            | ~20%                | $< 0.0105$      | $\times 6.05$     |

'CHR' refers to the controlled hot rolling treatment  
 'IQ' refers to the intermediate quenching treatment  
 (see ref. 6)

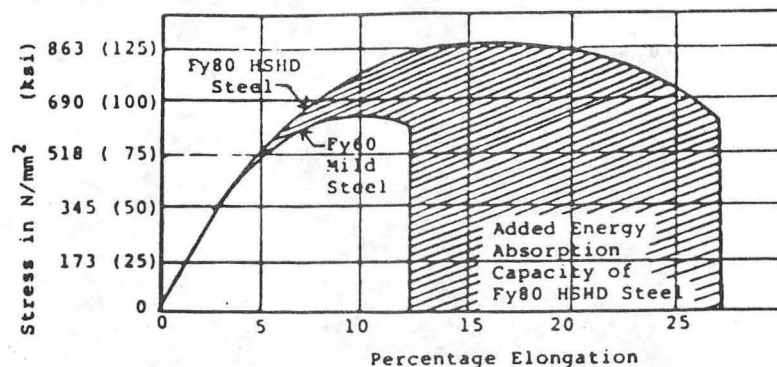
Table III. Tensile Properties  
(as heat treated)

| Specimen | Heat Treatment | Vol.Pct. Martensite | Y.S. ksi(MPa)   | U.T.S. ksi(MPa) | $e_u$ (%) | $e_T$ (%) |
|----------|----------------|---------------------|-----------------|-----------------|-----------|-----------|
| A1       | CHR            | ~20%                | 79.2<br>(546)   | 119<br>(820)    | 14.4      | 25.5      |
| A2       | IQ             | ~20%                | 69.6<br>(480)   | 117<br>(807)    | 12.4      | 22.7      |
| A3       | IQ             | ~40%                | 83<br>(572)     | 130.5<br>(900)  | 9.9       | 19.9      |
| A4       | IQ             | ~50%                | 91<br>(627)     | 135.8<br>(936)  | 8.7       | 18.6      |
| A5       |                | 100%                | 147.2<br>(1015) | 172.6<br>(1190) | 3.5       | 12.3      |
| B        | IQ             | ~30%                | 81<br>(558)     | 130<br>(896)    |           | 20        |
| C        | CHR            | ~20%                | 54<br>(372)     | 97<br>(669)     | 18.2      | 30        |

Lin et al. (17) have pointed out the advantages of the Fy80 steel for engineering construction applications such as reinforcing bars and, when drawn, as prestressed tendons particularly in view of the higher strengths at superior ductilities to the existing standards for conventional steel bars at Fy60. Furthermore, due to the low carbon levels the DFM steels Fy80 are easily butt or lap welded and have excellent corrosion characteristics (see Fig. 4). The latter results are particularly attractive for tendons used in environments where saline corrosion is dominant. In addition, savings in weight of up to 25% can be effected in construction applications whenever force controls the design. For example, 7.5 sq. ins. of Fy80 rebars will supply the same force as 10 sq. ins. of Fy60 rebars whether working stress or UTS is used in the design. On the other hand, Fy80 rebars will require approximately 2% more steel to meet the requirements of bond or anchor development. Thus, overall one could anticipate an economic savings of about 20% if dual phase Fy80 steel were to replace conventional Fy60 bars.

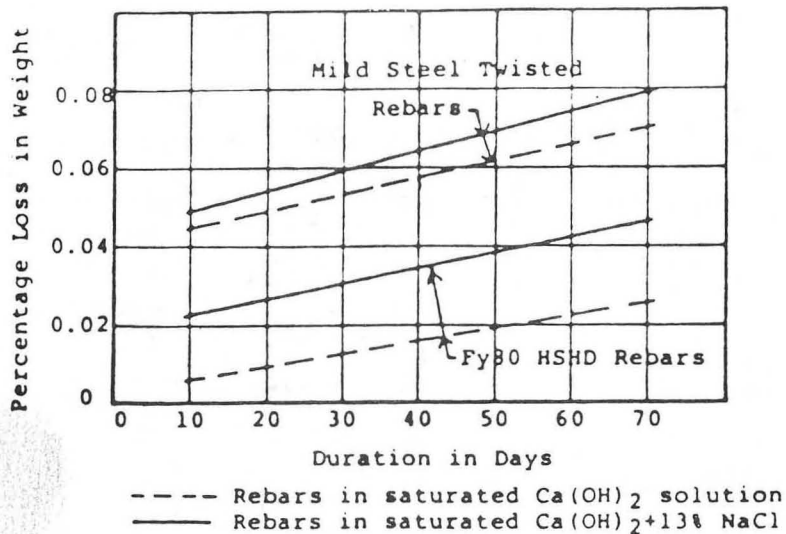
#### Structure and Properties of Drawn Wire

Typical starting microstructures are shown in Fig. 2 in which the desired fibrous lath martensite is present at about a volume fraction of 20% vol.%. Figure 5 shows a typical transmission electron micrograph of the dual phase structure, consisting of lath martensite and ferrite. The



XBL 8510-4275

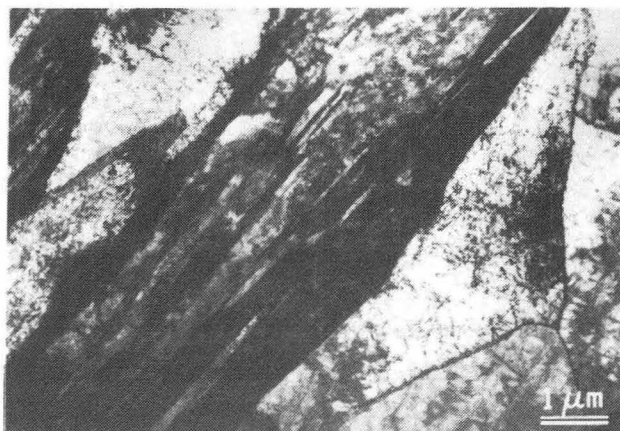
Figure 3 - Tensile Properties of DFM and conventional steels showing added energy absorption capacity of Fy80 HSHD steel.



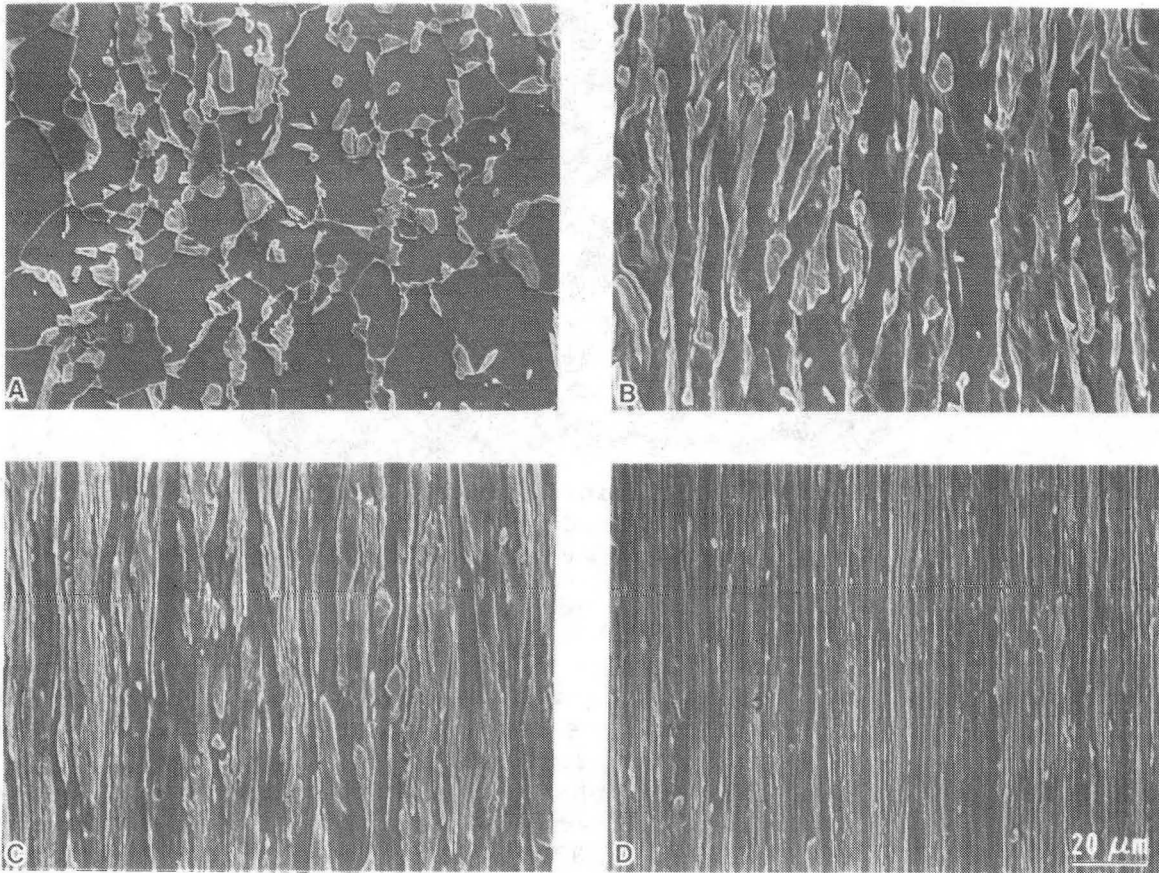
XBL 8510-4274  
 Figure 4 - Corrosive loss for mild steel and Fy80 HSHD rebar. Courtesy of R.H.G. Rau

progression of the dual phase structure as the reduction in area by wire drawing was increased to 97% is shown in Fig. 6. These micrographs clearly show that the lath (dislocated) martensite deforms continuously with the ferrite matrix, and there is no observable void formation at the interface between the two phases. After drawing, a very high dislocation density is developed as can be seen by the transmission electron micrograph of Fig. 7. It is obviously quite difficult to distinguish between the ferrite and martensite. This structure is similar to that of drawn pearlite and ferrite.

The drawing limit for all the steels corresponds to the maximum drawing strain the wire can sustain before failure at the die. Generally, all the dual phase steels produced by the controlled rolling process and the intermediate quenching treatment could be drawn to large strains ( $\epsilon > 4$ , i.e. >98% reduction in area) and high strengths (tables 3,4) without any patenting (intermediate) heat treatments or fracture. The drawability limits for the different specimens are listed in Table 2. A comparison of the three intermediate quenching treated silicon containing alloys shows that the drawing limit of the specimen with the highest volume fraction of martensite is the lowest, as might be expected.



XBB 850-8427  
 Figure 5 - Transmission electron micrograph of initial wire rod showing lath martensite in ferrite matrix developed in the specimen A2.



XBB 851-851

Figure 6 - Scanning electron micrographs of dual phase steel wire as a function of reduction in area (R.A.) by cold drawing (A) 0% R.A. ( $\epsilon=0$ ), (B) 70% R.A. ( $\epsilon=1.2$ ); C 88% R.A. ( $\epsilon=2.1$ ); (D) 97% R. A. ( $\epsilon=3.6$ ). Alloy A.

Table IV. Tensile Strengths of Drawn Dual Phase Wire. ksi (MPa)

| Wire dia.<br>in. (mm) | Total<br>True Strain | A1            | A2            | A3              | B             | C             |
|-----------------------|----------------------|---------------|---------------|-----------------|---------------|---------------|
| .217 (5.51)           | 0                    | 119<br>(820)  | 117<br>(807)  | 130.5<br>(900)  | 130<br>(896)  | 97<br>(669)   |
| .174 (4.42)           | 0.44                 | 168<br>(1158) | 170<br>(1172) | 179.3<br>(1236) | -             | -             |
| .119 (3.02)           | 1.2                  | 205<br>(1413) | 201<br>(1386) | 215<br>(1482)   | 208<br>(1434) | 176<br>(1214) |
| .0764(1.94)           | 2.09                 | 223<br>(1538) | 217<br>(1496) | 233<br>(1607)   | 235<br>(1620) | 195<br>(1345) |
| .0526(1.34)           | 2.83                 | 245<br>(1689) | 245<br>(1689) | 260<br>(1793)   | -             | -             |
| .0362(0.92)           | 3.58                 | 280<br>(1930) | 271<br>(1869) | 287<br>(1979)   | 293<br>(2020) | 252<br>(1738) |
| .0195(0.495)          | 4.82                 | 340<br>(2344) | -             | -               | -             | 307<br>(2117) |
| .0172(0.437)          | 5.07                 | -             | 345<br>(2379) | 369<br>(2544)   | -             | -             |
| .0136(0.345)          | 5.54                 | 380<br>(2620) | 375<br>(2586) | 395<br>(2724)   | 393<br>(2710) | 357<br>(2462) |
| .0105(0.267)          | 6.06                 | 407<br>(2806) | 403<br>(2779) | -               | -             | 380<br>(2620) |

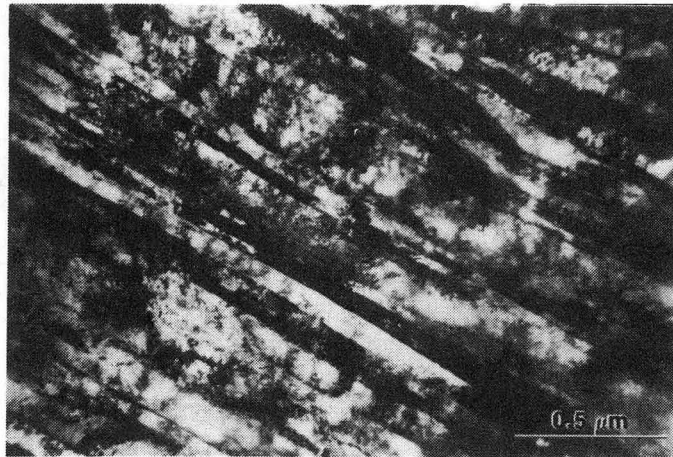
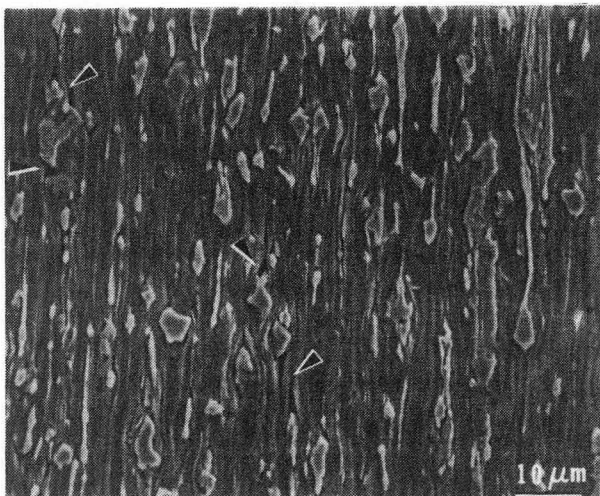


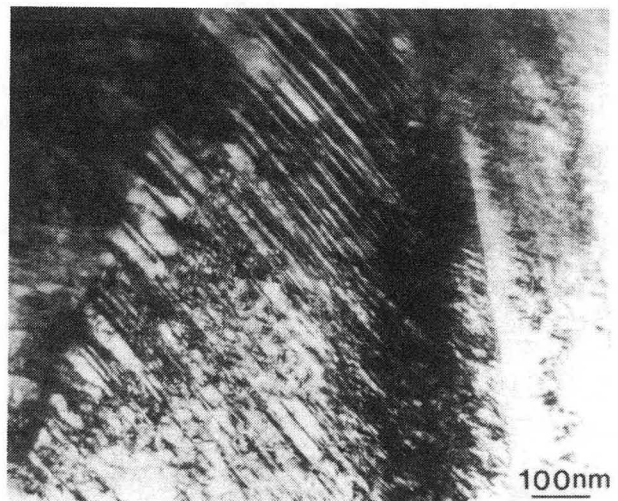
Figure 7 - Transmission electron micrograph of dual phase steel wire after a total drawing strain of 3.6. Specimen A2. XBB 823-2272

Figures 9-12 show drawing data for some of the steels investigated. It can be seen that very high tensile strengths are achieved even though the carbon content is low (compare to conventional 0.7%C wire, Fig. 12). From Table IV it is seen that dual phase steel wire rods can be drawn to a total true strain  $\epsilon=3.6$  i.e., 97% reduction in area for a strength level of 270 ~ 300 ksi (1860 ~ 2070 MPa) such as needed for bead wire or pre-stressed wire tendons for concrete, or drawn to  $\epsilon=6$ , i.e. 99.8% reduction in area for high strength levels of 380 ~ 400 ksi (2626 ~ 2760 MPa), required for example for tire cord applications.

In order to achieve successful wire drawing to tire cord size and strength, metallurgical control is essential. For example, the effect of martensite on drawability is particularly important if the microstructure contains plate (twinned) martensite rather than lath martensite. This can



XBB 858-5974



XBB 850-8428

Figure 8 (a) - Scanning electron micrograph of highly deformed ferrite around non-deforming martensite particles. ( $\epsilon=3.6$ ). Notice the void formation near the martensite-ferrite interface. (b) TEM micrograph showing (in initial wire rod) the presence of twinned martensite - the undesirable phase for wire drawing.

happen due to (a) too low a quench temperature, or (b) inefficient cooling or (c) carbon segregation (6). These factors cause the carbon content of austenite to exceed 0.4%wt which on subsequent transformation leads to high carbon plate martensite. The latter is non-deformable leading to microvoids and coalescence and subsequently to shear failure (Fig. 8).

#### Discussion

The results shown in Fig. 9-12 indicate that the flow stress of dual phase steel wires follow the empirical equation initially proposed by Embury and Fisher (18), viz.,

$$\sigma_f = \sigma_i + \frac{k}{\sqrt{r_0}} \exp\left\{\frac{\epsilon}{4}\right\} = \sigma_i + \frac{k}{\sqrt{r_0}} \sqrt{\frac{D_0}{D}}$$

where  $\sigma_i, k$ : constant

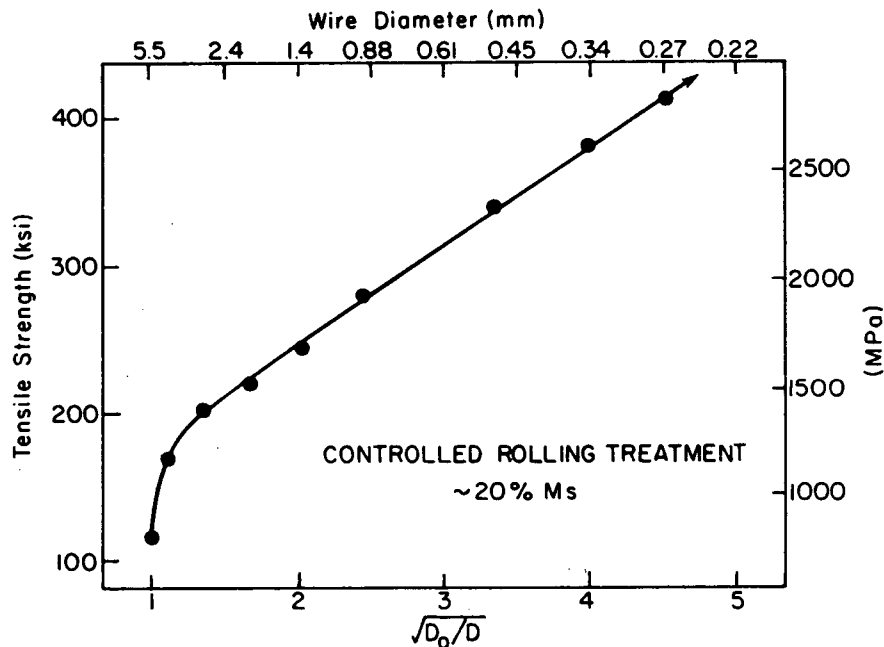
$\sigma_f$  : flow stress of wire after being drawn to a strain  $\epsilon$

$D_0$  : initial wire rod diameter before drawing

$D$  : wire diameter after being drawn to a strain  $\epsilon$

$r_0$  : mean spacing of dislocation barrier of wire rod

It is apparent from the above equation that to obtain higher strength wire with a given deformation,  $r_0$  should be as small as possible. That is, the effective grain size (the scale of the DFM structure) should be refined. This means optimising recrystallisation of ferrite during final rolling (Fig. 1).



XBL 852-5870

Figure 9 - Plot of tensile strength of dual phase steel wire as a function of wire diameter. Specimen A1.

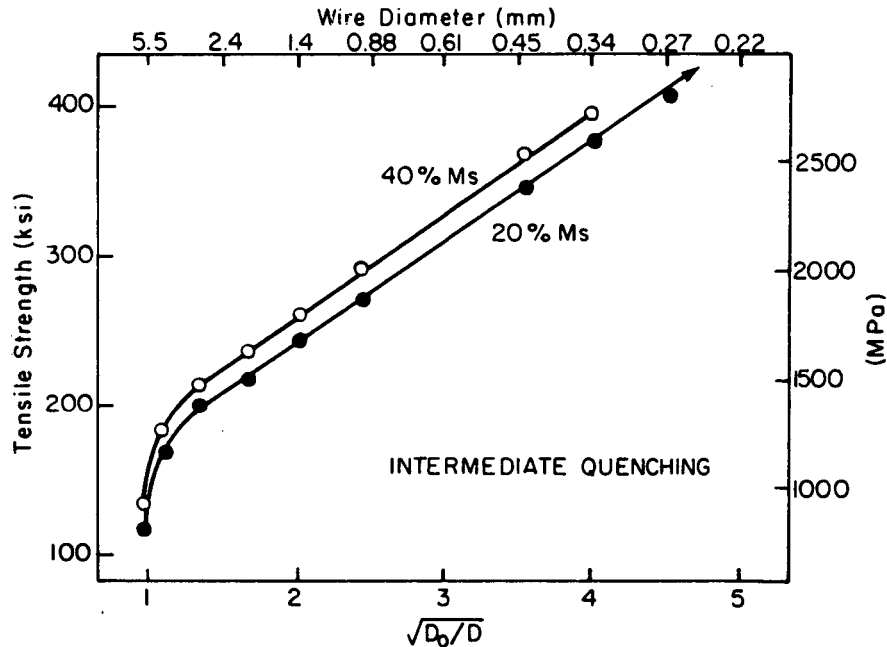


Figure 10 - Plot of tensile strength of dual phase steel wire as a function of wire diameter. Specimen A2,A3. XBL 852-5871

The effect of volume fraction of martensite on the tensile properties of drawn wire is shown in Fig. 11. Whilst this is easily varied by heat treatment (tables II, III) it is difficult to control the volume fraction by the rolling processing because the finish temperature does not affect this parameter appreciably over the range which is needed to obtain lath martensite. What is critical is that the quench be rapid enough to prevent further  $\gamma \rightarrow \alpha + \gamma$  decomposition above the Ms temperature which would result in high carbon plate martensite or bainite formation. As shown in Fig. 8, this is undesirable.

The hardenability of the austenite phase is determined by partitioning of alloying elements, especially carbon, during the time the steel is in the  $(\alpha + \gamma)$  phase field. The application of Energy Dispersive X-ray microanalytical techniques (19) indicates slow partitioning of Mn and only slight partitioning of silicon but it is estimated from CCT diagrams corresponding to the estimated austenite composition below  $A_{r3}$  that water quenching should be effective for producing the desired dual phase steels as rods or bars up to 3/4" diameter.

#### Conclusion

In conclusion, the present research shows that dual phase steels can be designed and processed as new, economical low carbon steels bars and rods for construction applications and as rods for cold drawing into high tensile strength steel wires. Current work indicates wires of tensile strengths up to 400,000 psi can be obtained (figs. 10-12). Potential applications for dual phase steel wire include bead wire, tire cord, wire rope and prestressed concrete (Fig. 13). It is particularly significant that the dual phase steel rods can be drawn to high strength tire cord sizes without patenting heat treatments, and this would be attractive from energy savings and economical reasons. It should be possible to produce wire rods in existing rod mills by adapting the processing procedures and controls outlined in this paper.

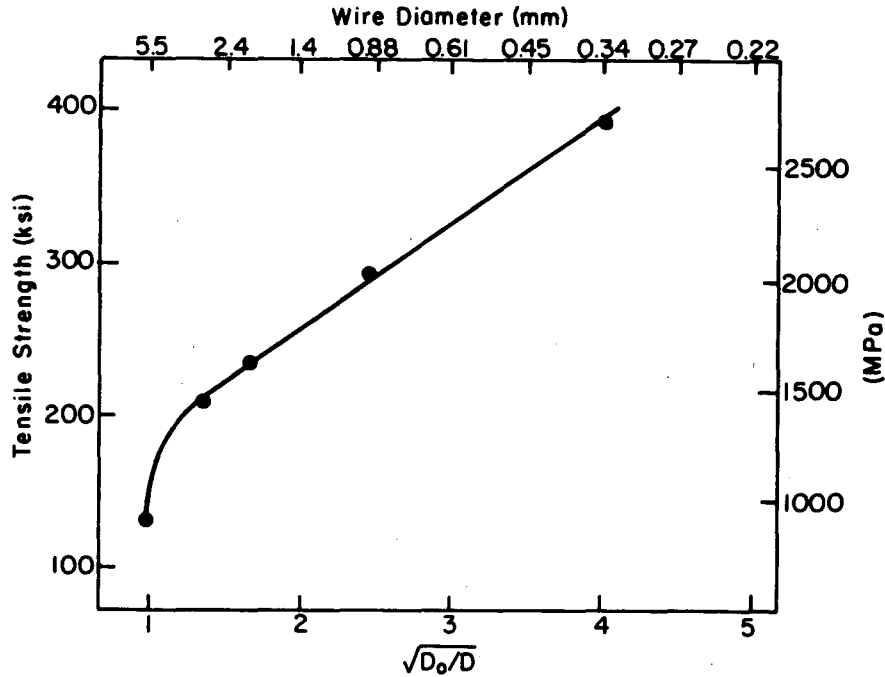


Figure 11 - Plot of tensile strength of dual phase steel wire as a function of wire diameter. Specimen B. Alloy C behaves almost indently to this. XBL 857-6464

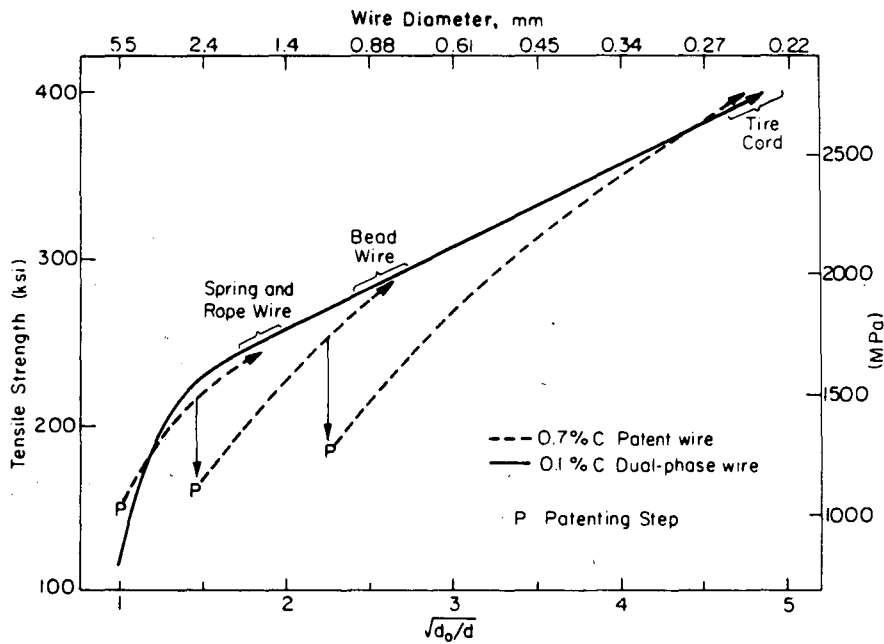


Figure 12 - Comparison of the drawing schedule and resulting tensile strength for dual-phase wire and patented pearlitic wire. (Courtesy of R. M. Fisher). XBL 8210-6760B



### Acknowledgements

This work was supported by the Director, Office of Energy Research, Office of Basis Energy Sciences, Materials Sciences Division, U. S. Department of Energy under Contract No. DE-AC03-76SF00098. The author is grateful to Nippon Kokan K.K., Japan, Stelco, Canada and Arbed Saarstahl, West Germany for providing some of the steels used in this investigation. I am particularly grateful to the many student and staff scientists who have contributed to our dual phase steel research programs, for example, J. Y. Koo, N. J. Kim, A. Nakagawa and J. Ahn. I also thank Dr. R. M. Fisher, Center for Advanced Materials, Lawrence Berkeley Laboratory, Berkeley, CA for helpful discussions, and Mr. R. H. G. Rau of Mukand Iron and Steel Co. for providing the corrosion data in Figure 4.

### References

1. S. Hayami and T. Furukawa, "Micro Alloying '75", Proc. of an Int. Symp. on High Strength Low Alloy Steels held in Washington, D. C., 1975, Union Carbide Corp., 270 Park Ave., New York, NY. (1977). pp. 311-321.
2. A. T. Davenport, Ed., "Formable HSLA and Dual Phase Steels", The Met. Soc. AIME, New York, NY (1978).
3. R. A. Kot and J. W. Morris, Eds., "Structure and Properties of Dual Phase Steels". *ibid* (1979).
4. R. A. Kot and B. L. Bramfitt, Eds., "Fundamentals of Dual-Phase Steels", *ibid* (1981).
5. J. Y. Koo and G. Thomas, Metall. Trans. A, 8A (1977) p. 525.
6. G. Thomas, in Frontiers in Materials Techniques, M. A. Meyers and O. T. Inal, Eds., Elsevier Science Publishers, B.V. The Netherlands, (1985) Chap. 3, pp. 89-123.
7. R. W. K. Honeycombe, Microstructural Design in Low Alloy Steels, Specialty Steels and Hard Materials, N. R. Comins and J. B. Clark, Eds., Pergamon Press, (1983) p. 23.
8. G. Thomas, V. F. Zackay and E. R. Parker, "Strengthening Mechanisms in Metals and Ceramics", J. J. Burke, N. L. Reed and V. Weiss, Eds., Syracuse University Press, (1966) p. 3.
9. S. A. Herres and C. H. Loring, Trans ASM, 40 (1945) p. 775.
10. R. L. Cairns and J. A. Charles J.I.S.I 205 (1967) p. 1044.
11. R. A. Grange, Proc. 2nd Intl. Conf. on Strength of Metals and Alloys, Pacific Grove, California, ASM (1970) p. 861.
12. W. S. Owen, Metals Tech. 7. (1980) p. 1.
13. A. H. Nakagawa and G. Thomas, Metall. Trans. A vol. 16A, (1985), pp. 831-840.
14. J. Ahn and G. Thomas, "Interwire '85" 55th Annual Convention, Nov. 3-8, 1985, Atlanta, GA, in press.
15. N. J. Kim, A. J. Yang and G. Thomas, Metall. Trans. A, vol. 16A (1985) pp. 471-474.
16. V. B. Dutta, S. Suresh, G. Thomas and R. O. Ritchie, Int. Congress of Fracture, ed. I. Le May EMAS, Brazil (1983), pp. 457-466.
17. T. Y. Lin, P. Y. Chow, G. Thomas and R.H.G. Rau, 10th FIP Congress, New Delhi, Feb. 1986, in press.
18. J. D. Embury and R. M. Fisher, Acta Metall., Feb. (1966), pp. 147-159.
19. M. Ohmura, Ph.D. Thesis, UC Berkeley (1985).

This report was done with support from the Department of Energy. Any conclusions or opinions expressed in this report represent solely those of the author(s) and not necessarily those of The Regents of the University of California, the Lawrence Berkeley Laboratory or the Department of Energy.

Reference to a company or product name does not imply approval or recommendation of the product by the University of California or the U.S. Department of Energy to the exclusion of others that may be suitable.

*LAWRENCE BERKELEY LABORATORY  
TECHNICAL INFORMATION DEPARTMENT  
UNIVERSITY OF CALIFORNIA  
BERKELEY, CALIFORNIA 94720*