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The properties and welding of Supraform TM steel

by H.J. DE KLERK* and R.C. DE VILLIERS†

SYNOPSIS

The metallurgical phenomena in the design and manufacture of high-strength low-alloy (HSLA) steels are reviewed, and some aspects of the production of Supraform TM (thermo-mechanically processed) steel are discussed.

The weldability of Supraform TM is related to changes in the microstructure of its heat-affected zone.

SAMEVATTING

Die metallurgiese verskynsels in die ontwerp en vervaardiging van hoë-sterkte-laelegeringstaal (HSLA-staal) word in oënskou geneem en sommige aspekte van die produksie van Supraform TM-staal (termo-meganies verwerkte staal) word bespreek.

Die sweisbaarheid van Supraform TM word in verband gebring met veranderinge in die mikrostruktuur van sy hitte-invloedsone.

Introduction

The development of micro-alloyed steels based on alloy design principles from structure–property relationships is now well documented[1,2]. Several excellent publications[3-9] on the fundamentals of hot working, physical metallurgy, and deformation mechanics of steel have established the science of high-strength low-alloy (HSLA) steels.

Micro-alloyed steels can be defined as low-carbon steels deriving their strength primarily from grain refinement and dispersion strengthening. Other strengthening mechanisms, such as sub-grain forming, dislocation strengthening, cold working, and ageing are used to supplement these two mechanisms in achieving the desired strength. Optimization of the microstructure in this way can lead to high strength with acceptable toughness.

Strengthening Mechanisms in Steel

Grain Refinement

The scientific work of Hall and Petch (1951 and 1953 respectively) on the influence of grain size on the properties of iron and mild steel led to one of the best known structure–property relationships[1,2], namely

$$\sigma_y = \sigma_o + K_y d^{-1/2},$$

where

σ_y is the yield stress
σ_o is the friction opposing dislocation movement
d is the grain diameter
K_y is a constant.

Fig. 1 illustrates the effect of grain size on the yield strength of a carbon–manganese and a carbon–manganese–niobium steel[10].

It is evident that the strengthening effect of grain size as shown by the slope of the two lines in Fig. 1 is the same for the two steels, but the basic strength of the niobium steel is higher than that of the carbon–manganese steel. This is mainly due to dispersion strengthening by fine niobium carbo-nitrides. The grain size effect can also be extended to a much higher level in the niobium steel[11,12].

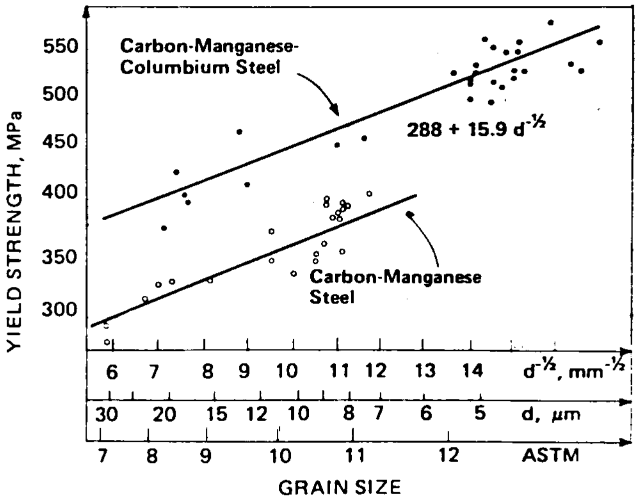

Fig. 1—Experimental Hall–Petch relationships determined by quantitative metallography for carbon–manganese and carbon–manganese–columbium (niobium) steels (after Le Bon and De Saint-Martin[10])

Composition

The easiest way to increase the strength of steel is by increasing the carbon content. In the past this method was used to meet the relatively low level of technological demands on steels. However, with the advent of welding as the major fabricating method, higher carbon contents led to serious problems such as cold cracking. Furthermore, since toughness was the first essential for service performance, pearlite strengthening became unacceptable. The detrimental effect of carbon[13], and hence pearlite, in the microstructure on the fracture toughness can be seen in Fig. 2.

The development of micro-alloyed steels proceeded via 'pearlite-reduced' steels with a carbon content of less than

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0,1 per cent[1]. These had excellent weldability but inferior strength. Controlled rolling provided a means whereby the strength could be increased, but it was only when strong carbide-forming elements were added that the process met with success. Suitable alloying elements are niobium, vanadium, titanium, and aluminium. The nitrides and/or carbides of these elements, which have a limited solubility in iron, inhibit recrystallization and grain growth, and this forms the basis of micro-alloying. Niobium is considered to be the best element to achieve the desired effect.

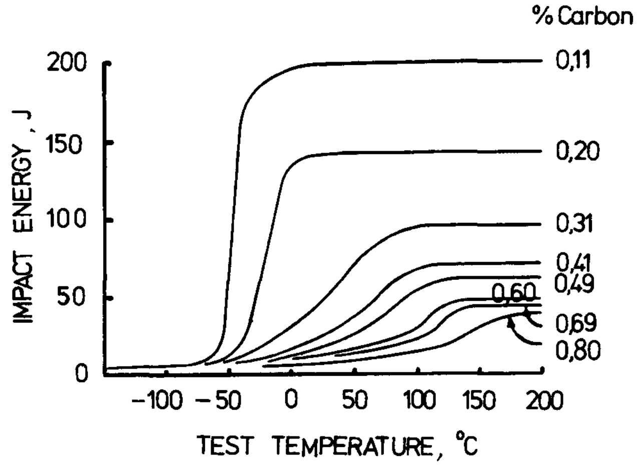

Fig. 2—The effect of carbon, and hence pearlite, content on the impact energy and transition temperature of ferrite–pearlite steels (after Burns and Pickering[1])

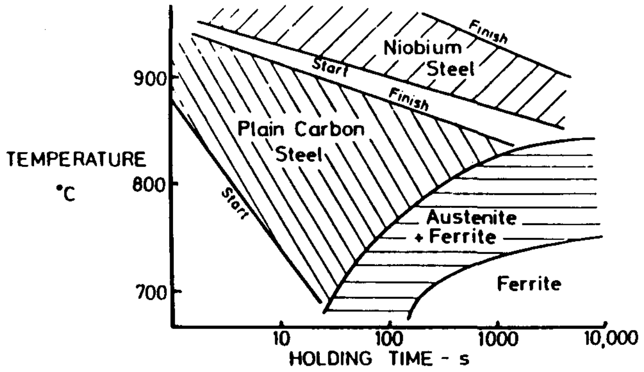

Fig. 3—Effect of niobium on the recrystallization of austenite after a reduction of 50 per cent (after Pickering[11])

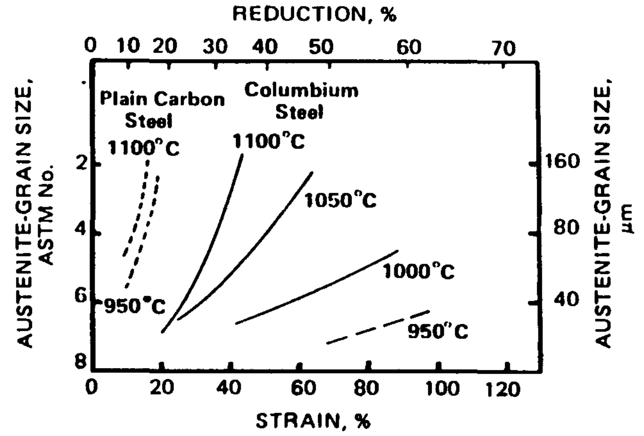

Fig. 4—Effect of deformation temperature and initial grain size on the critical amount of deformation required for the completion of recrystallization in plain carbon and columbium (niobium) steels (after Tanaka *et al.*[12])

Figs. 3 and 4 illustrate the effect of niobium on the recrystallization behaviour of plain carbon and niobium steels[11,12].

Controlled Rolling

The ferrite grain size after hot rolling is determined by the austenite grain size and the transformation temperature. Control of the austenite grain size and of the extent of deformation of the austenite prior to transformation to ferrite is achieved by controlled rolling.

Hot working is usually done between 1250 and 1050°C without temperature control. Normal hot rolling of carbon steel takes place in this temperature range. The temperature during the finishing passes is usually high enough for easy recrystallization and grain growth of the deformed austenite[2]. Controlled rolling of normal carbon steels in the range of 1050 to 800°C made it possible for a ferrite grain size of 10 µm to be obtained, instead of the normal 40 to 200 µm[3]. Careful control of the temperature–deformation history of the steel was necessary for the desired results to be achieved[10,14].

In a niobium-containing steel, complete recrystallization takes place only at much higher deformation and at lower temperatures during cooling. Since higher deformation leads to a finer grain size after recrystallization, a very fine austenite grain size can be obtained. Grain sizes smaller than 1 µm are possible.

Fig. 5 predicts the changes in microstructure during the controlled rolling of a carbon–manganese steel and a niobium steel[15].

The structural changes that take place during controlled rolling have been summarized by Honeycombe[2] as follows:

(1) hot deformation of the steel in the austenitic condition

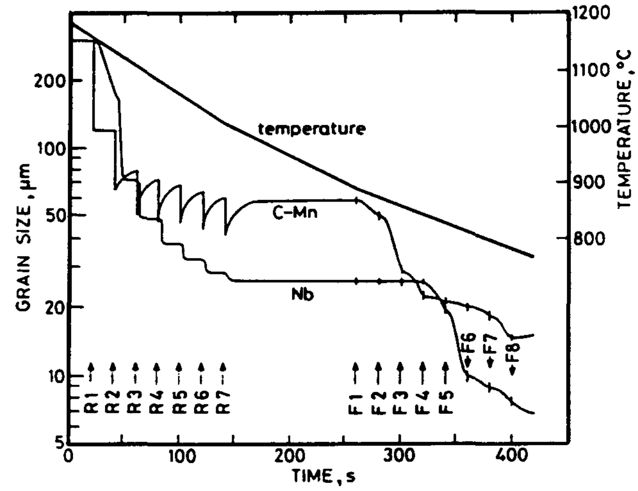

Fig. 5—Changes in austenite grain size predicted during a simplified schedule for the controlled rolling of 20 mm plate from a 200 mm slab of carbon–manganese steel with a niobium content of 0,04 per cent (after Leduc[4])
R = rolling pass F = finishing pass

(2) recrystallization of the austenite
(3) grain growth of the austenite
(4) precipitation of nitrides and carbo-nitrides in the austenite
(5) transformation of the austenite to ferrite ($\gamma \rightarrow \alpha$)
(6) precipitation of alloy carbides during that transformation
(7) precipitation of alloy carbides in the ferrite.

The optimum condition at the mill exit is a deformed, fine-grained austenite. At the finishing temperature, some ferrite is present and is also deformed. If the finishing temperature is too low, partial recrystallization of the ferrite can lead to a mixed grain size, which gives poor toughness. The optimum finishing temperature is about 800°C, at which the best toughness and strength are obtained.

After the mill exit, the cooling rate must be controlled because the transformation to ferrite should take place at the lowest temperature if the finest equi-axed ferrite is to be obtained. At too low a temperature, highly dislocated lath ferrite with poor toughness is obtained. The final optimum microstructure would be an equi-axed, fine-grained precipitation-strengthened ferrite with a low dislocation density. The pearlite content should be kept to a minimum that is consistent with the required properties.

The precipitation of carbides during and after the transformation in the ferrite is of minor importance for micro-alloyed steels such as Supraform TM. Most of the niobium carbides are already precipitated when the rolling ends.

Supraform TM

Properties

The need for a hot-rolled, high-strength, low-alloy steel with improved formability led to the development of the Supraform TM grades. (The letters TM stand for thermo-mechanically processed). These are reduced-pearlite steels with a carbon content that is usually between 0,05 and 0,11 per cent, while the sulphur content is usually less than 0,006 per cent (Table I). Specifications that cover this type of steel are ASTM A715 and SEW092. The low carbon content and inclusion shape control give them excellent formability for the strength level. Non-strain-ageing properties are imparted to the steel by aluminium treatment. At present niobium is used as micro-alloying element.

Supraform TM is currently produced only as a strip-mill product so that the width is limited to a maximum of 1800 mm. Decoiling considerations limit the thickness to 10 mm, although 12 mm may be produced. Production parameters impose a minimum thickness of 2 mm for the lower- and 2,5 mm for the higher-strength grades.

Typical minimum values for the total elongation in a tensile test are given in Table II, which conveys an idea of the ductility, and thus the formability, in operations such as press-brake bending and stretch forming. The main strengthening mechanism in HSLA steels, i.e. grain-boundary strengthening, also favours higher ductility. This implies that the formability of HSLA steels is better than that of pearlite-strengthened carbon steels at the same strength level.

TABLE I
CHEMICAL COMPOSITION OF SUPRAFORM TM STEEL (LADLE ANALYSIS, %)

C	Mn	Si	P	S	Al	Nb*	V*	Ti*
0,12 (max)	1,40† (max)	0,50 (max)	0,035 (max)	0,035 (max)	0,02 (min)	0,01 0,08	0,20 (max)	0,20 (max)

*Any combination of these elements to give the indicated values of yield strength
†1,5% max. for Grade TM500

TABLE II
MECHANICAL PROPERTIES OF SUPRAFORM TM STEEL*

Grade	Tensile strength MPa	Yield strength (min) MPa	Elongation (min), %		Mandrel diameter for 180° bend test° t
			$t^{\dagger}$ 3 mm‡	t 3 mm§	
TM340	420/540	340	20	18	0,5^4
TM380	450/590	380	20	15	0,5
TM420	480/620	420	20	16	0,5
TM460	520/670	460	20	15	1,0
TM500	550/700	500	18	15	1,0

* Tensile test to BS 18 Part 2 or 3
† t = material thickness
‡ Gauge length 80 mm
§ Gauge length 200 mm
° Bend test: the sample specimens shall be free of cracks on the outside

Applications

At present Supraform TM is used mainly in the automobile and transport industries. It should also have applications in the line-pipe and construction industries wherever a high ratio of strength to weight, combined with fine grain for toughness, is required. It is particularly suitable where parts are pressed or cold formed. The strength of Supraform TM grades is shown relative to other structural steels in Table III and Fig. 6.

Welding of Supraform TM

Supraform TM has excellent weldability owing to its low carbon content. The low sulphur and phosphorus contents prevent hot cracking in the weld metal.

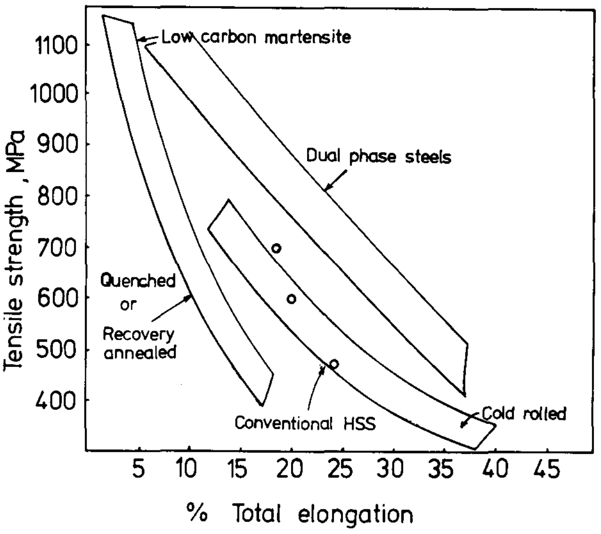

Fig. 6—Tensile strength and total elongation of commercially available high-strength steels (after Manganon[17])

TABLE III

STRUCTURAL STEELS CLASSIFIED ACCORDING TO TYPE AND YIELD STRENGTH
(PREPARED BY ISCOR MARKETING DEVELOPMENT)

	STRUCTURAL STEEL									
	PLATE				SHEET, COILS AND PLATE 12 mm AND THINNER		SHEET AND COILS			
							HOT-ROLLED		COLD-ROLLED	
Yield stress mPa	As-rolled and normalised	As-rolled with copper (0,20 – 0,35%)	Weather resistant	Quenched and tempered	Weather resistant	Improved formability	Standard grades	Standard grades with copper	Weather resistant and standard grades	Pressing grades
700 600 500 400 300 200 100	BS4360 – 55 BS4360 – 50 BS4360 – 43	BS4360 – 50B BS4360 – 43A	COR-TEN B BS4360 – WR50B1	ROQ-tuf AD690 AE690 AF690 ROQ-tuf C550 ROQ-tuf C500 ROQ-tuf C410	COR-TEN A COR-TEN A-F BS4360 – WR50A1	SUPRAFORM TM500 SUPRAFORM TM460 SUPRAFORM TM420 SUPRAFORM TM380 SUPRAFORM TM340 SUPRAFORM 290 SUPRAFORM 250 SUPRAFORM 220 SUPRAFORM 190	BS1449 HR50/35 BS1449 HR45/25 HR37/23	BS1449 HR50/35 BS1449 HR45/25	Sheet COR-TEN A Coil BS1449 CR37/23	SUPRAFORM CR340 SUPRAFORM CR300 SUPRAFORM CR260 SUPRAFORM CR235

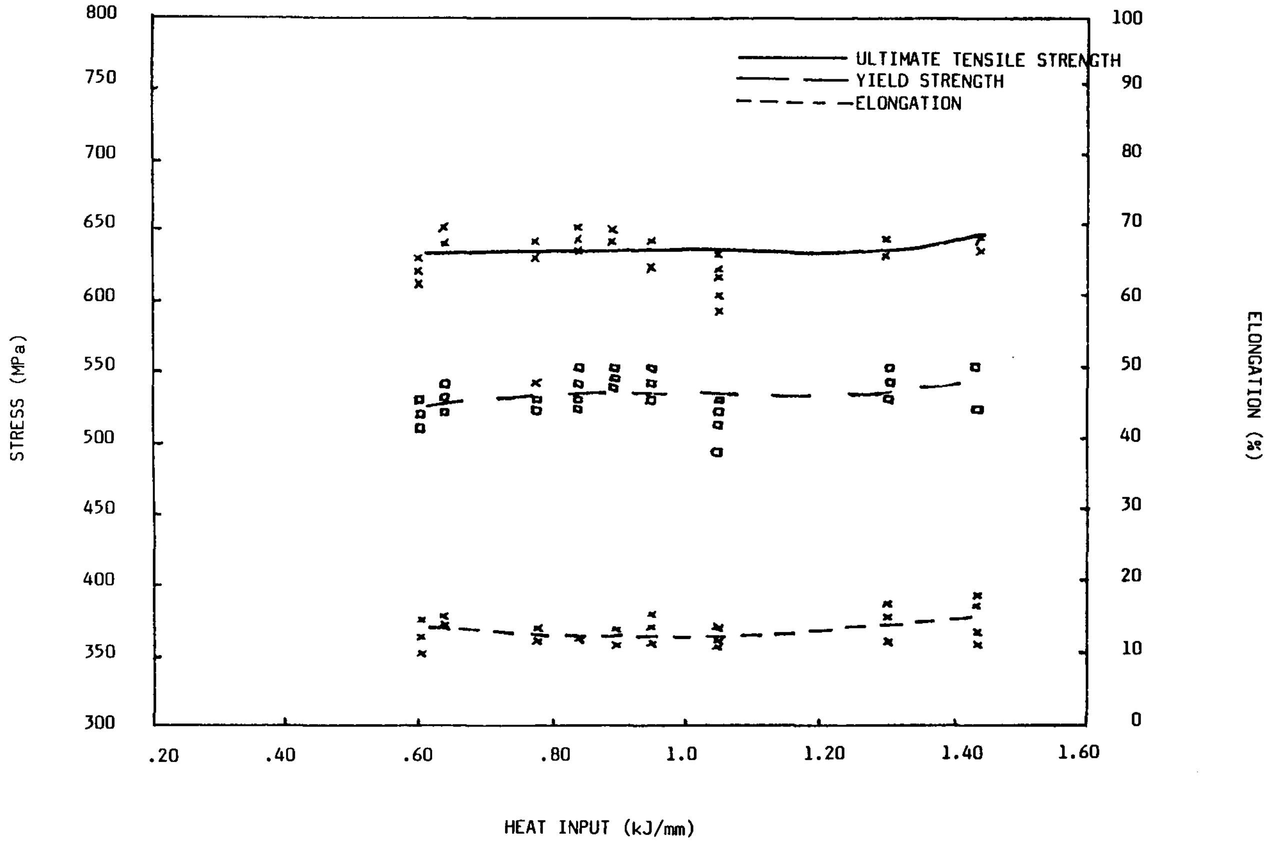

Fig. 7—The effect of heat input on the tensile properties of Supraform TM 460, 5 mm thick, C.E. = 0,32 Transverse weld properties ('normal' heat input = 0,8 kJ/mm, lowest joint efficiency = 0,94)

$$\text{C.E.} = \%\text{C} + \frac{\%\text{Mn}}{6} + \frac{\%\text{Cu} + \%\text{Ni}}{15} + \frac{\%\text{Cr} + \%\text{Mo} + \%\text{V}}{5}$$

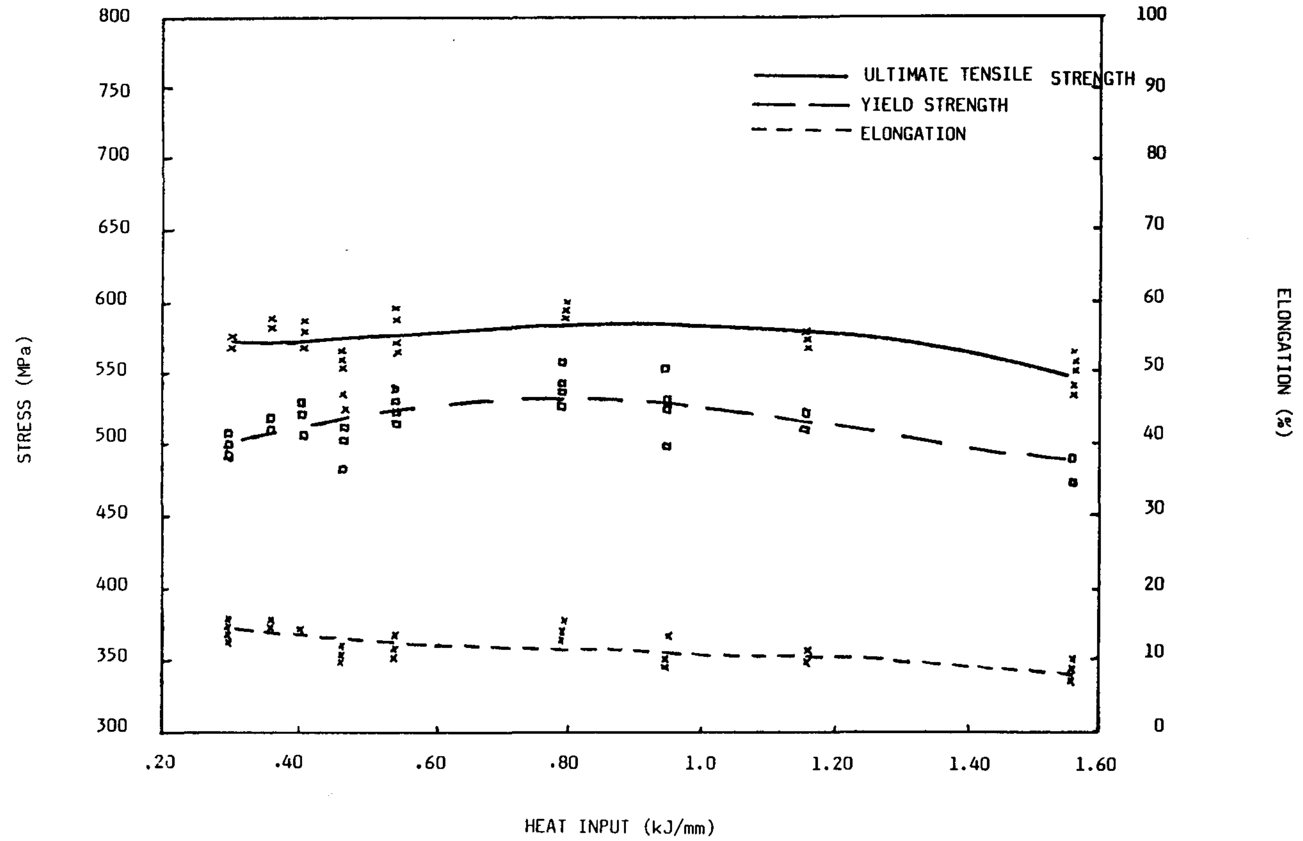

Fig. 8—The effect of heat input on the tensile properties of Supraform TM 380, 3 mm thick, C.E. = 0,23 Transverse weld properties ('normal' heat input = 0,6 kJ/mm, lowest joint efficiency = 0,96)

Welding tests on two grades of Supraform TM show that the steel is weldable under a wide range of heat input. Welded joints had strengths only slightly lower than the base metal with a joint efficiency of 95 per cent. Figs. 7 and 8 show the results of tests on the 460 and 380 grades, and the changes that occur during fusion welding are illustrated in Fig. 9.

An area of importance is the high-temperature zone, where austenite grain growth occurs. Work published[16] on grain growth shows that the maximum grain size in the heat-affected zone (HAZ) is limited by the presence of niobium in the steel. The austenite grain size for given conditions of heat input, preheating temperature, plate thickness, and position in the HAZ is much smaller for a niobium-containing steel than for an ordinary carbon–manganese steel. This effect is illustrated in Figs. 10 and 11, which show the grain growth for two such steels.

The expected loss in strength in the region of austenite grain growth does not occur. It is restored by the high cooling rate, which causes the austenite to become transformed to fine lath ferrite with high dislocation density. Some niobium carbide precipitation strengthening can also occur. Fig. 12 is an electron-micrograph of such a microstructure.

A slight loss in toughness occurs in the HAZ because of the change from the optimized structure. The following mechanisms for embrittlement in the HAZ may play a part:

(a) grain growth in austenite

(b) grain-boundary precipitation of cementite, as shown in Fig. 13

(c) a change from equi-axed to lath ferrite

(d) a change in pearlite morphology (Fig. 13).

All these are thought to be of minor significance on thin plate such as the thicknesses in which Supraform TM is produced. However, they could cause problems on thicker HSLA plates, where it is more difficult to achieve the desired properties.

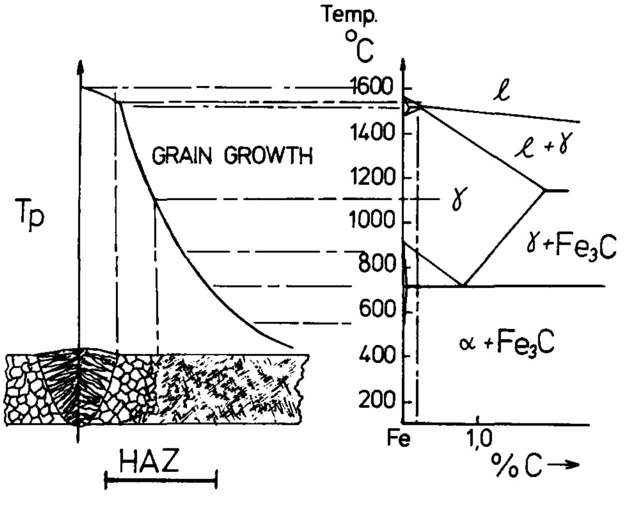

Fig. 9—Schematic diagram showing the changes occurring in the heat-affected zone (HAZ) of a low-carbon (0,15 per cent by mass) steel weld (after Ashby and Easterling[16])

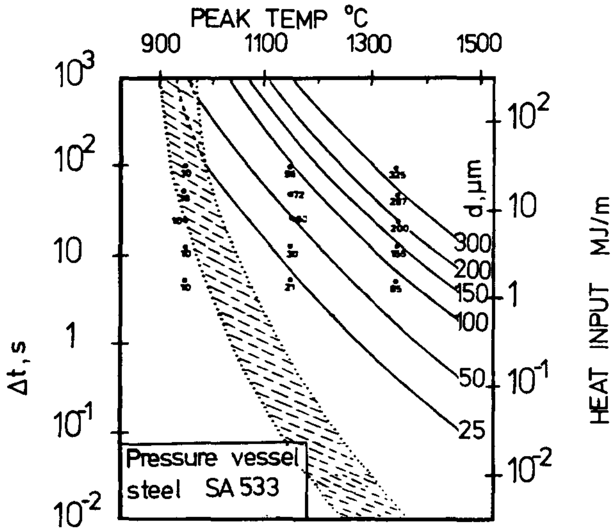

Fig. 10—Grain growth in a pressure-vessel steel — data of Klumpes for thick plate (after Ashby and Easterling[16])
t = time to cool from 800 to 500 °C in seconds

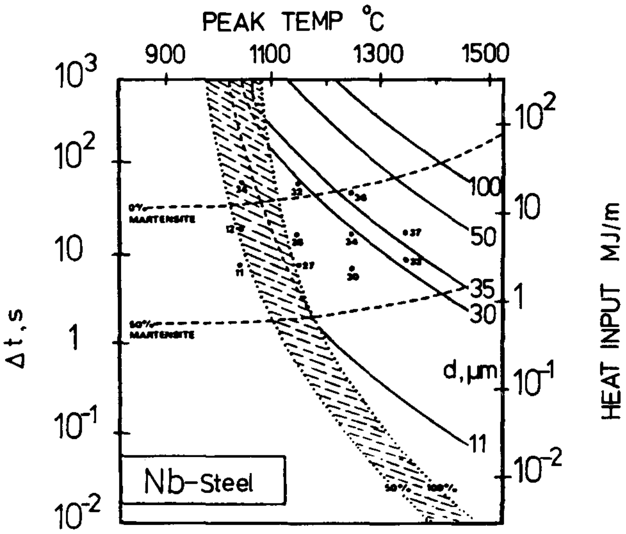

Fig. 11—Grain growth in a niobium–micro-alloyed steel— thick plate (after Ashby and Easterling[16])

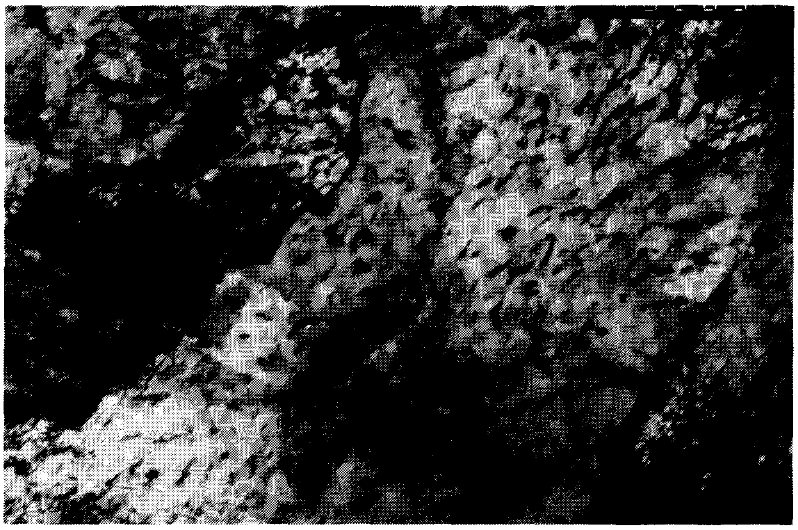

Fig. 12—Thin-foil micrograph of lath ferrite in the HAZ of a weld on Supraform TM 420

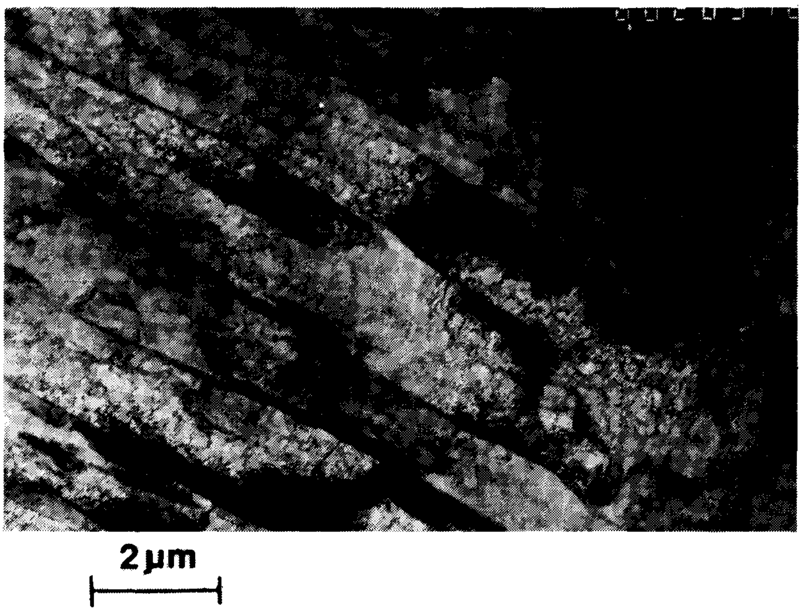

Fig. 13—Thin-foil micrograph of pearlite in the HAZ

Acknowledgements

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President-Elect of IMM

The Council of the Institution of Mining and Metallurgy has unanimously elected Mr W.G. Yuill to be President for the session 1985–86 in succession to Mr P.M.J. Gray.

William Godson Yuill, B.Sc., D.I.C., F.I.M.M., was educated at Falkirk High School and Glasgow University. He graduated in pure geology in 1951 and then joined Siamese Tin Syndicate. At that time, Siamese Tin were initiating the investigations of the Leadhills–Wanlockhead lead–zinc mining district in the Southern Uplands of Scotland.

In 1955 he joined Mackay and Schnellmann (later Mackay and Schnellmann Ltd) as a geologist, and was appointed Managing Director of that organization in 1970 — a position that he still holds. During his years with Mackay and Schnellmann, he has worked on professional assignments in more than forty countries, spending extended periods in Iran, Nigeria, and Liberia. In 1960–61 he undertook post-graduate research at the Royal School of Mines, obtaining the D.I.C. in mining engineering. Most of his professional career has been concerned with the exploration, development, and evaluation of metallic and industrial mineral deposits.

Mr Yuill was elected a Member of the Institution in 1958 and a Fellow in 1965; he was a Member of Council from 1974 to 1981 and since 1982; and he served as Vice-President for the session 1983–84. He has been an active member of several standing committees and Chairman of the Overseas Committee since 1979.

Mr Yuill will assume the Presidency of the Institution at the Annual General Meeting on 16th May, 1985.

Author's reply*

Dr Lloyd believes that the gold supply controls the gold price, and goes on to show that the use of the model results in the price varying from country to country. He correctly implies that this situation is impossible, but incorrectly concludes that the model must be worthless.

The flaw in Dr Lloyd's argument is that the supply does not control the price; the reverse is the case. There is one basic price world-wide, which results in the supply varying from country to country according to circumstances.

For example, an increase in price tends to increase production from the new alluvial deposits in Brazil but to reduce production from old established mines in South Africa. This reduction comes about because these mines lower their pay limits to prolong their lives and to minimize costly selective mining.

The discussion on this paper is now closed: Editor.

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* To Dr P.J.D. Lloyd's contribution to the paper 'The uneconomic production of gold in South Africa' by H.L. Monro, which appeared in the *J. S. Afr. Inst. Min. Metall.*, vol. 84, no. 5. May 1984. pp. 113-124. Dr Lloyd's contribution was published in the *J. S. Afr. Inst. Min. Metall.*, vol. 84, no. 9. Sep. 1984. p. 297.

Recovery of gold and silver

Papers are invited for the International Conference 'Gold and Silver Recovery 1985', which is to be held at Beaver Creek, Colorado, from 2nd to 6th June, 1985. Technical topics for papers should be of practical value in the exploration for, mining, extraction, and recovery of gold and silver. Preference will be given to papers that describe useful new technology and processes, particularly those which have been demonstrated in pilot plants or full-scale commercial operations. Presentations that provoke discussion and participation by the audience are encouraged. The emphasis of this Conference is on useful information for operators and cost assessment that will help reduce the risk of innovations.

The following are examples of topics of interest:

- Exploration and Resource Potentials
  - Biogeochemistry
  - New Analytical Techniques
  - Improved Geological Models
  - Potential New Resources
- Mining
  - Cost-saving Innovations
  - Backfill
  - In-pit Crushing and Conveyors versus Trucks
- Extractive Metallurgy: Refractory Ores
  - Characteristics of Refractory Ores
  - Processing of Refractory Ores
  - Pre-oxidation by Roasting
  - Pressure and Chemical Pre-oxidation
  - Biological Oxidation
  - Treatment of Manganiferous Silver Ore
  - Pressure Leaching
  - Fine Grinding
  - Improved Recovery in Calcine Leaching
  - Carbon-in-leach
  - Non-Cyanide Lixiviants
- Heap Leaching
  - Applicability of Heap Leaching
  - Heap Leaching in Desert Climates
  - Heap Leaching in Cold Climates
  - Heap Leaching in Wet Climates
  - Heap Leaching in High Altitudes
  - Heap Leaching in Environmentally Sensitive Areas
  - Heap Leaching of Refractory Ores
  - Heap-leach Pads and Pond Linings
  - Water Management in Heap Leaching
  - Cost Parameters of Heap Leaching
  - Heap Leaching of Tailings
  - Vat Leaching versus Heap Leaching
- In-situ Leaching
- Carbon-in-pulp and Resin-in-pulp
  - New CIP Concepts and Design
  - Carbon Performance
  - Screens for CIP
  - Agitation for CIP
  - Water-quality Effects in CIP
  - CIP versus Merrill–Crowe Applicability
  - Resin-in-pulp
  - Resin Selection for RIP
  - When Not to Employ CIP
- Carbon Columns and Continuous Ion Exchange
  - New Concepts in Column Design
  - Effect of Water Quality of Carbon Adsorption
  - Selection of Carbon and Resins
  - Gold Recovery from Low-level Effluents
- New Carbon- and Resin-elution Processes
- Carbon Regeneration
- Maximizing Recovery — The Systems Approach
  - Treatment of Tailings and Effluents
  - Gold Recovery from Copper-rich Ores
- Environmental Aspects of Gold and Silver Operations
  - Regulations and Permits
  - Water Management
  - Tailings Management and Design
  - Cyanide Destruction or Recovery
- Placer Operations and Gravity Processes
- New Process for Large Low-grade Oxidized-ore Deposits

Abstracts or outlines should be submitted to Randol International Ltd, 21578 Mountsfield Drive, Golden, Colorado 80401, U.S.A. Telephone: (303) 526-1626; telex 45-885 TAS DVR.