The use of the superhardenability treatment for the production of high-strength low-alloy steel for hot-rolled plate

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SYNOPSIS

The role of hardenability in the development of high-strength low-alloy (HSLA) steels is discussed. The applicability of the superhardenability treatment, a steelmaking technique for increased hardenability, to HSLA steels is indicated, and the literature on this treatment is reviewed.

Several low-carbon superhardenability HSLA steels were designed by use of published superhardenability multiplying factors. It was found that the hardenability of these low-carbon steels can be increased to the same extent as that of medium-carbon steels. The mechanical properties as determined in normalized bars indicate that the superhardenability treatment can be used successfully in the production of low-carbon hot-rolled HSLA steels.

SAMEVATTING

Die rol van verhardbaarheid in die ontwikkeling van hoëtreksterkte-laelegeringstaal (HSLL-staal) word bespreek. Daar is tekens dat die superverhardbaarheidsbehandeling, 'n staalvervaardigingstegniek vir verhoogde verhardbaarheid, op HSLL-staal toegepas kan word en daar word 'n oorsig oor die literatuur oor hierdie behandeling gegee.

Daar is verskeie soorte superverhardbare laekoolstof-HSLL-staal met gebruik van gepubliseerde superverhardbaarheidsvermenigvuldigingsfaktore ontwerp. Daar is gevind dat die verhardbaarheid van hierdie laekoolstofstaal in dieselfde mate as dié van mediumkoolstofstaal verhoog kan word. Die meganiese eienskappe soos in genormaliseerde stawe bepaal, dui daarop dat die superverhardbaarheidsbehandeling suksesvol op die produksie van warmgewalste laekoolstof-HSLL-staal toegepas kan word.

Introduction

High-strength low-allow (HSLA) steels have been defined in various ways and, according to one such definition[1], these are steels with yield strengths over 275 MPa, the alloying additions being designed to provide specific desirable combinations of properties such as strength, toughness, formability, weldability, and resistance to atmospheric corrosion. HSLA steels according to this definition are limited to steels in the hot-rolled, cold-rolled, or normalized conditions, and quenched and tempered steels are excluded.

The origin of HSLA steels can be traced back to 1900, when Williams[2] noted that copper-bearing steels exhibited increased corrosion resistance when wet and dried alternately. In 1929, when the alloy contents of these steels were increased to further improve their corrosion properties, it was found that, in addition to having 4 to 8 times higher resistance to atmospheric corrosion, they had high strengths[3]. The increased strength was achieved by the solid-solution strengthening of elements such as copper, silicon, and phosphorus, and the increase in hardenability effected by elements such as chromium, nickel, and phosphorus. The results from these studies led to the commercial production of USS COR-TEN steel in 1933, an HSLA weathering steel that is still very popular today.

Since those early days, several other HSLA steels were developed to meet the changing demands of industry. In most of these developments, use was made of alloying additions to improve the mechanical properties by mechanisms such as precipitation hardening, solid-solution strengthening, and increased hardenability. An increase in hardenability and the resultant refinement of the microstructure were central in the majority of these developments. In most cases, increases in hardenability were achieved by a higher alloying content, but because of the increasing cost and scarcity of alloying elements, the current trend is to substitute cooling control after hot rolling for alloy content. This is, however, not always possible or economical from a production point of view, and a way of increasing hardenability without increasing the alloy content would be very valuable.

Superhardenability Treatment for Increased Hardenability

A recent paper by Brown and James[4], in which they claim that the superheating of 150M36 (En 15) melts resulted in a dramatic increase in hardenability, is important if viewed in the above context. This treatment consists of the superheating of steel melts to temperatures above 1650°C and the making of aluminium additions of 0,05 to 0,1 per cent. If the alloying content and base hardenability of these 150M36 melts were above a certain threshold value, the hardenability could be increased to such an extent that the Jominy curves of the superhardenability melts would be similar to those obtained with steels of much higher alloying content such as En 24 (817 M40). In their studies, a large number of 150M36 melts were prepared with and without superheat and additions of aluminium or titanium, and the results clearly indicate that a significant and repeatable increase in hardenability can be obtained by this treatment. They report that additions of titanium are just as effective as additions of aluminium.

Although this paper clearly indicates that a significant

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increase in hardenability can be obtained, the magnitude of the effect was not quantified because the Jominy test was used exclusively. Owing to the high hardenability of the melts after the treatment, flat Jominy curves were obtained and the extent of the increase in hardenability could not be determined. This lack of quantification, as well as the fact that these studies were mainly limited to 150M36 steels with small variations in chemical composition, has limited the widespread use of this treatment in practice.

In a subsequent study, Mostert and Van Rooyen[5-7] attempted to clarify these issues. A range of forty-nine casts with widely different chemical compositions and base hardenabilities, including steels with much higher base hardenability than those studied previously, was prepared with (1650°C) or without (about 1550°C) superheat and additions of aluminium.

In order to quantify the hardenability increase, they made use of three hardenability tests: the standard Jominy test[8], an extended Jominy test[9] for slower cooling rates, and an air-hardenability test[10] for even slower cooling rates. Vickers-hardness testing was used to indicate the actual hardenability, and these values were expressed as D_1 (50 per cent martensite). The actual hardenability values were compared with the values calculated from the chemical composition.

The typically large differences in observed hardenability behaviour of the conventional and superhardenability casts are illustrated in Fig. 1. In Fig. 2 the actual hardenability of the superhardenability-treated and conventional melts are compared with their base hardenability (calculated from the chemical composition and the austenite grain size).

The superhardenability results of Fig. 2 were subsequently presented as superhardenability multiplication factors (the ratio of the actual to the predicted values), as in Fig. 3. These multiplication factors for the super-
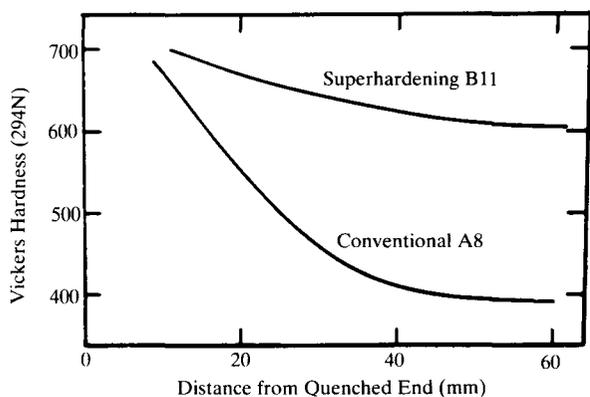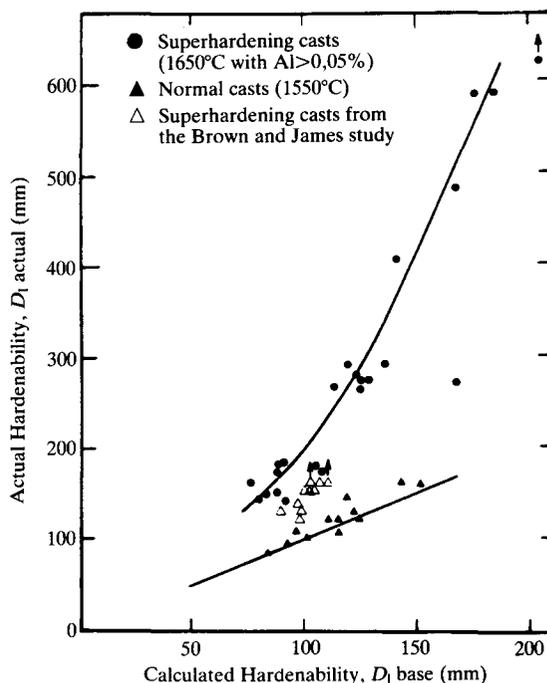

Fig. 2—Hardenability-test results of superhardenability and conventional casts[5]
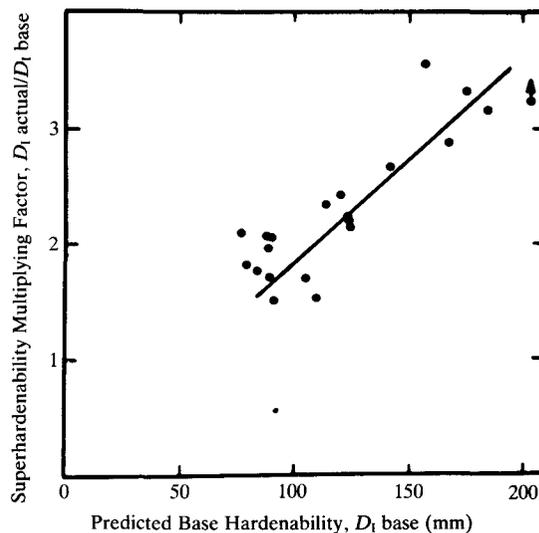
Fig. 3—Superhardenability multiplication factors as a function of base hardenability[5]

Fig. 1—Typical Jominy curves for superhardenability and conventional casts[5] (for compositions see Table I)

TABLE I

CHEMICAL COMPOSITION OF CASTS A8, B11, C9, and C11

| Cast | Melt temp. °C | Composition, % (by mass) | | | | | | | | | | | | ASTM grain size |
		C	Mn	P	S	Si	Ni	Cu	Cr	Mo	V	Nb	Al	
A8	1550	0,46	1,09	0,02	0,03	0,19	0,13	0,16	0,94	0,06	0,04	0,01	0,03	8
B11	1650	0,44	1,08	0,02	0,03	0,28	0,16	0,15	0,94	0,06	0,04	0,02	0,08	8

hardenability treatment permit quantitative calculation of the hardenability of a steel manufactured by this treatment. The following conclusions were drawn from Fig. 2.

1. Although the actual hardenability of the conventional melts was predicted fairly accurately from the chemical composition, the actual hardenability values after the superhardenability treatment were much higher than the hardenability calculated from the chemical composition.

2. The increase in hardenability (the ratio of the actual to the calculated hardenability) due to the treatment was found to depend on the base hardenability and was much greater in the higher-alloyed steels. This was a novel finding, and indicates that extremely high hardenability can be obtained by the superhardenability treatment of fairly highly alloyed steels, thus making the through-hardening of very large components possible.

3. The increase in hardenability was similar for steels of similar base hardenability, even when there were large differences in the chemical composition. This observation suggests that predictable increases in hardenability can be obtained in a wide range of steels.

Superhardenability Treatment for HSLA Steels

The effective and predictable way in which the super-hardenability treatment can cause large increases in hardenability without major alloying additions makes it eminently suitable for the production of HSLA steels requiring high hardenability.

One such application is that of as-hot-rolled martensitic–bainitic HSLA steels. The fairly large section sizes, combined with the low-carbon content (0,1 per cent) required for good impact and weldability properties in this type of steel, would require very large alloying additions for sufficient hardenability in conventional steels. For example, the well-known 0,1 C–2,25 Cr–1 Mo steel type does not achieve sufficient hardenability for this purpose[11] (Fig. 6). However, the application of the superhardenability treatment to suitable low-carbon Mn-Cr steels should permit the production of low-cost as-hot-rolled martensitic steels. Because no previous work is known to have been done on low-caron superhardenability steels, a number of such steels were produced on a laboratory scale.

From the results of previous work[5] (Fig. 3), a steel type for this application was designed. From Fig. 3 it follows that, if a steel with a base hardenability of 100 mm (D_1 50 per cent martensite) were used, superhardenability D_1 values of about 180 mm could be expected. An H value (severity of quench) of 0,02 was accepted as appropriate for air cooling, and conversion[12] to the actual air-hardenable diameter indicated that successful superhardening of this steel would permit air-hardening of bars of 30 mm diameter. The carbon level was fixed at 0,1 per cent, and the base hardenability increased to the 100 mm D_1 level with additions of manganese and chromium. These elements were decided upon because of the very cost-effective way in which they increase hardenability[13]. From CCT diagrams[14] of highly-alloyed conventional low-carbon steels, it was expected that through-hardness values of Hv 400 could be

found in the as-rolled 30 mm bars, in combination with ambient Charpy energy values of about 45 J.

Vacuum-induction melts (2 kg in mass) were subsequently prepared (the analyses are given in Table II) and forged into 30 mm rounds. Alloying elements were added to the melt and, in the case of the superheated melts, the melts (with alloying elements added) were kept at 1650°C for 2 minutes. Extended Jominy tests were performed after austenizing at 940°C, and Vickers hardness (294 N) curves were determined (Fig. 4). Metallographic examination of the ground flats indicated that in all cases the non-martensitic constituent was lower bainite. The hardness of a 50 per cent martensite–50 per cent lower bainite mixture for these steels was obtained from published CCT curves[14] of steels with appropriate compositions that had been transformed to a martensitic–lower bainitic structure. The D_1 values (50 per cent martensite) were then obtained from the hardness curves.

The data on cooling rates for the extended Jominy test[15] were used in the conversion of the specific distances on the test bars to D_1 values. Austenite grain sizes were determined on the quenched ends of the test samples by the use of an etchant consisting of aqueous picric acid and hydrochloric acid[16]. The average grain size of ten fields was determined at a magnification of 500 times by use of the standard ASTM rosette, and these values were then converted to ASTM grain sizes. All the test samples were cut and examined through the section to ensure that decarburization had not influenced the results.

The multiplying factors of Kramer, Siegel, and Brooks[17] were used in the calculation of the base conventional hardenability from chemical composition. The actual hardenabilities of the experimental melts are compared with the calculated values in Fig. 5 and Table II.
The following conclusions can be drawn from Fig. 5.
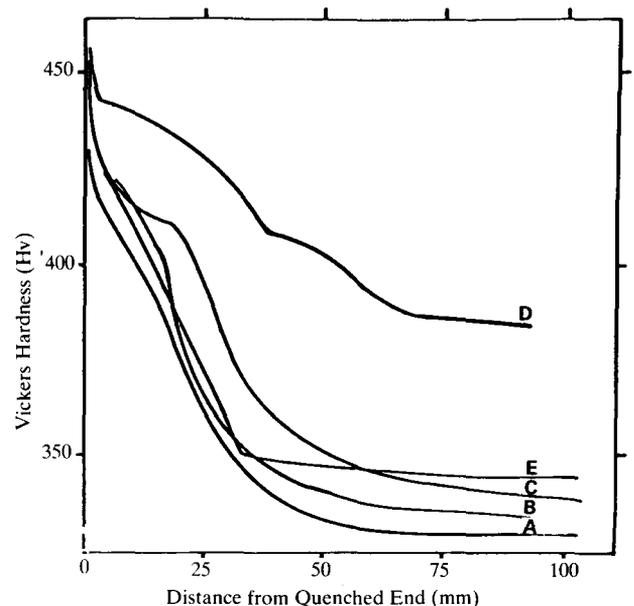(1) The good agreement between the actual and the calculated hardenabilities of the conventional melts indi-

Fig. 4—Extended Jominy curves for experimental steels

TABLE II
CHEMICAL COMPOSITIONS AND HARDENABILITIES OF EXPERIMENTAL CASTS*

Melt	Max melt temp. °C	Composition, %									ASTM grain size	Calculated D_I	Actual D_I
		C	Mn	P	S	Si	Cr	V	Al	Ti			
A	1650	0,106	2,02	0,011	0,019	0,25	1,51	0,043	0,063	0,045	9,4	74	99
B	1550	0,123	1,95	0,009	0,024	0,22	1,58	0,29	0,006	0,016	6,9	96	99
C	1550	0,132	2,11	0,010	0,020	0,25	1,55	0,040	0,005	—	7,8	96	111
D	1650	0,138	2,19	0,010	0,023	0,35	1,55	0,046	0,011	0,039	9,9	99	180
E	1650	0,138	2,13	0,010	0,025	0,29	1,51	0,054	0,029	0,076	9,8	91	96

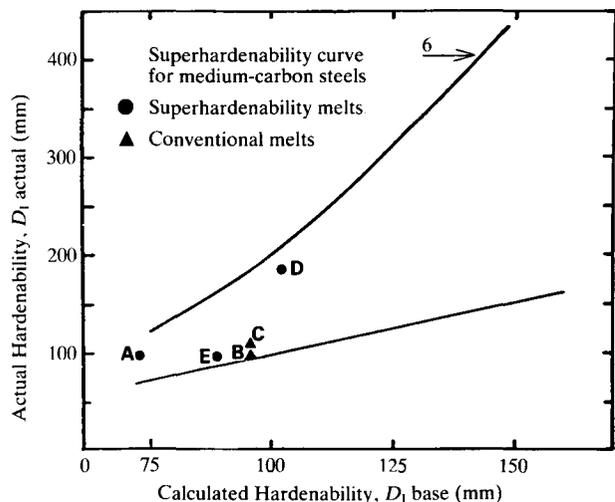*The Ni, Cu, and Mo contents are all smaller than 0,02 per cent.

Fig. 5—Hardenability test results for low-carbon superhardenability and conventional casts

cates that the calculation system used is accurate for these low-carbon Mn–Cr steels.

(2) Although the chemical composition of melt D is very similar to melts C and E, the hardenability curves (Fig. 4) indicate that much greater hardenability was obtained for melt D. As indicated in Fig. 7, the size of the superhardenability increase in this low-carbon steel was similar to that found in the medium-carbon steels previously investigated (0,3 to 0,6 per cent carbon). The fact that the superheated melts A and E showed much smaller hardenability increases than melt D can be attributed to the lower D_I base values of melts A and E, which seem to be close to the threshold limit. The results therefore indicate that the threshold limit (that value of D_I base below which no hardenability increase can be obtained by the treatment) for these low-carbon steels is very similar to the 75 mm value observed for medium-carbon steels. These results imply that equivalent superhardenability increases can be obtained for steels with large differences in carbon content as long as the D_I base values are equivalent. This observation should be considered in conjunction with the previous conclusion[18] that large differences in the content of major alloying elements, such as manganese and chromium, did not result in different superhardenability behaviour as long as the D_I base values were similar. From these observations it seems likely that steels with the same base hardenability will show similar increases in superhardenability irrespective of chemical composition.

It should be noted that the above conclusions are based on only three low-carbon superhardenability melts at this stage and should therefore be regarded as tentative.

(3) The superhardenability multiplying factor system developed for medium-carbon steels could be used to accurately predict the hardenability of low-carbon steel D.

(4) If the hardenability curves of low-carbon (Fig. 4) and medium-carbon steels (Fig. 1) are compared, it is seen that, although the increase in D_I values is similar for steels with similar D_I base values, the actual increase in hardness at the air-cooled end of the Jominy bar is much larger in the medium-carbon steels (100 to 200 Hv) than in the low-carbon steels (50 Hv). This observation is important from a practical viewpoint, and can be attributed to the fact that, for the achievement of a D_I base value greater than the threshold value, larger alloying additions have to be made to a low-carbon than to a medium-carbon steel. Therefore, the non-martensitic constituent of a conventional (1550°C melt temperature) low-carbon steel with D_I base of 100 mm is lower bainite with a relatively high hardness, compared with the upper bainite–pearlite structure of an equivalent medium-carbon steel. Upon superhardening, the medium-carbon steel exhibits a large increase in hardness at the air-cooled end of the Jominy bar due to the large difference in hardness between martensite (superhardenability steel) and pearlite (conventional steel). The relatively small increase in hardness in the low-carbon steel is due to the small difference in hardness between the martensitic (superhardenability) and bainitic (conventional) structures in question.

Mechanical Properties

The values for the hardness and 20°C Charpy energy of steels were determined after various rounds had been air-cooled from an austenitizing temperature of 940°C. The results (Table III) indicate that very good through-hardness and Charpy values were obtained in the 22 mm superhardenability steel. It should be kept in mind that the impact properties of similar production-scale steels should be better than those obtained in these 2 kg melts. This combination of mechanical properties makes this type of

steel very suitable for applications where relatively high hardness, good through-hardness, and good impact strength are required. The fact that these properties can be obtained in large section sizes in the as-rolled condition makes this type of superhardenability HSLA steel very attractive from a production point of view. The industrial applications of such a steel would include plate for wear conditions in the mining and agricultural fields. The low-carbon content should ensure good weldability, and the low cost would make it very competitive in this field.

Fig. 6—Extended Jominy curves for steels (see Table V)

TABLE III

COMPARISON OF THE HARDNESS AND AMBIENT CHARPY ENERGY OF NORMALIZED ROUNDS

Melt	Section size mm	Hardness 1 mm beneath surface Hv	Centre hardness Hv	Charpy energy at 20°C J
D	22	420	394	42
C	25	358	347	—
D	30	386	381	40
C	30	346	344	29

Relative Hardenability and Cost

The real attractiveness of this type of steel can be appreciated only when its hardenability and cost are compared with those of conventional steels. For this purpose, the hardenability curve of superhardened steel D is compared with that of conventional Roq Tuf and $2\frac{1}{4}$ Cr–1 Mo steels[11] in Fig. 6. The total alloying cost per 160 t melt was calculated and compared using the prices of alloying element as at Pretoria in June 1983[19] (Table IV).

From Table IV it is evident that superhardenability melt D offers a very good combination of hardenability and cost compared with the two conventional steels.

Other Types of HSLA Superhardenability Steel

Although the experimental work described in this paper was concentrated on superhardenability HSLA hot-rolled martensitic–bainitic steels, the treatment could be used to produce other HSLA steel types requiring high hardenability, such as hot-rolled dual-phase steels. The attractive features of this treatment should ensure that many superhardenability applications in the HSLA steel field will be found.

Conclusion

The results of the hardenability tests on conventional and superhardenability low-carbon steels indicate that similar
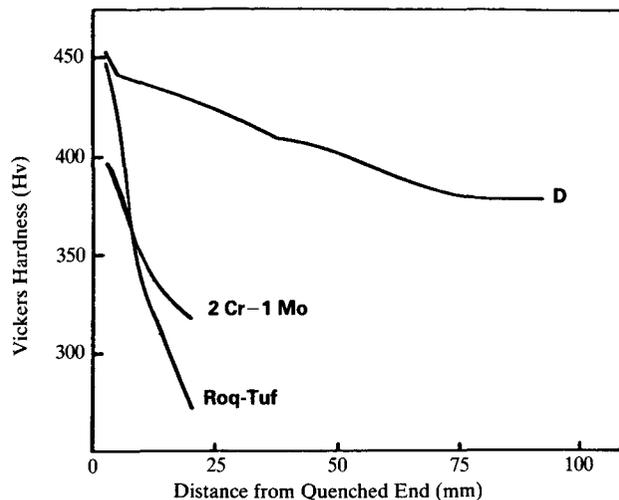
superhardenability behaviour was obtained to that previously found in medium-carbon superhardenability steels. The mechanical properties of the normalized bars indicate that the superhardenability treatment can be used successfully to produce low-carbon hot-rolled HSLA steels.

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TABLE IV

COMPARISON OF THE COST AND HARDENABILITY OF STEELS

Steel	Alloying element, %							Hardenability as D_I (50%) mm	Cost of alloying elements per 160 t melt R
	C	Mn	Si	Ni	Cr	Mo	B		
Superhardenability melt D	0,14	2,2	0,3		1,6			180	8 719
Roq Tuf AE 690	0,17	0,6	0,3	1,3	0,2	0,5	0,001	83	30 168
$2\frac{1}{4}$ Cr–1 Mo	0,08	0,5	0,3		2,3	1,0		—	37 301

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Multi-phase flow

BHRA, the Fluid Engineering Centre, will hold its Second International Conference on Multi-phase Flow in London from 19th to 21st June, 1985. Offers of papers are invited.

Pipelines and equipment associated with the recovery and refining of hydrocarbons often operate with multi-component, two-phase flow, while two-phase or multi-phase flows are incorporated into many processing operations. Departures from normal operating conditions in a range of industries can also give rise to multi-phase phenomena. Designing and operating plant in which multi-phase flows occur present a number of problems: the interaction between the phases is complicated; many design procedures are based on results derived from studies using small-scale models, which cannot be extrapolated with confidence; and reliable prototype measurements are not widely available.

Industries working at very high temperatures and pressures, such as nuclear-power generation, in combination with increased interest in gas-gathering systems and wet-gas pipelines, have stimulated fresh research into multi-phase flows. The Conference will deal with physical and mathematical modelling of liquid–gas and liquid–liquid flows for all these applications, as well as flows containing small amounts of solids.

The Conference will concentrate on the practical design and operation of multi-phase systems including model and prototype studies; the effects of scale-up; comparison of flow prediction and on-site measurement; practical instrumentation for multi-phase flow measurement; and flow through pipeline components, such as bends, tees, and valves.

Prospective authors should submit titles and synopses as soon as possible. Manuscripts will be required by 3rd December, 1984. Further information is available from the Conference Organiser, Multi-phase Flow, BHRA, the Fluid Engineering Centre, Cranfield, Bedford MK43 OAJ, England; telephone (0234) 750422; telex 825059.

Underwater mining

The 15th Underwater Mining Institute is scheduled for 28th to 30th October, 1984, at the Yahara Center in Madison, U.S.A. The technical programme will include presentations on the following:

● Development of Real-Time Analysis Systems for *in situ* Marine Minerals.
● Tectonic Setting of the Juan de Fuca Sulphide Deposits.
● Recent NOAA Studies of Several Vent Sites on the Juan de Fuca Ridge.
● Cobalt-bearing Manganese Crusts from the Pacific: Elemental Content and Extractive Metallurgy.
● Stockwork-like Sulphide Mineralization in Young Oceanic Crust: DSDP Hole 504B.
● Use of Seafloor Sulphide Data in Exploring for Sulphide Deposits on Land.
● The Continuous Seafloor Sediment-assay System: Ground Truth Confirmation.
● New Observations from the Submersible, ALVIN, in Hydrothermal Vent Fields, Southern Juan de Fuca Ridge.
● Progress in Ferromanganese Crust Research, including Cobalt-Rich Crusts.
● New Marine Mining Research and Exploratory Programmes.
● Marine Minerals Research in the Exclusive Economic Zone.
● Phosphate Resource Potential in Onslow Bay, North Carolina Continental Shelf.
● Exploration for Phosphate in the Offshore Territory of the Congo.
● Marine Mining — the View Ahead.

Dr Dallas Peck, Director of the U.S. Geological Survey, will present the keynote address: Marine Minerals Research in the Exclusive Economic Zone.

For technical programme information, contact J. Robert Moore: 512/471-4816. For registration information, contact Gene A. Woock: 608/262-0644.