"In Situ" Determination of Transformation Temperatures in the Weld Heat-Affected Zone

A novel, simply applied method of thermal analysis proves to be very sensitive for detecting the phase transformation occurring in the HAZ of a C-Mn steel

BY R. H. PHILLIP

ABSTRACT. The development of a direct procedure for determining transformation temperatures in the heat-affected zone of an actual weld, using a new thermal analysis technique, is outlined.

A continuous cooling transformation (CCT) diagram for the carbon-manganese (C-Mn) steel examined has been constructed to illustrate the potential of the method. The problems and limitations of the thermal analysis technique are also discussed.

Finally, the transformation behavior of the C-Mn steel determined using the "in situ" method is compared with the transformation behavior of the same steel under simulated welding conditions.

Introduction

The phase transformations that occur in ferritic steels during welding are of major importance in determining their weldability. Accordingly, numerous laboratory techniques have been developed to study these phase transformations, providing valuable information about the welding properties of steels. The methods followed fall into two categories:

1. Analytical methods.
2. Direct "in situ" methods.

In the analytical methods, small test specimens are subjected to thermal cycles similar to those experienced by a particular point in the heat-affected zone (HAZ) during welding. The progress of decomposition of the austenite during cooling is recorded, either by dilatometry (Ref. 1-6), thermal analysis (Ref. 5-9), or magnetic analysis (Ref. 9). There may, however, be considerable differences between the transformation behavior as determined under simulated welding conditions, and that occurring in an actual HAZ under nominally the same thermal conditions; this is because the latter experiences severe mechanical restraints and steep temperature gradients. In addition, there may be differences in the thermal cycles, particularly on heating, due to inadequacies associated with some of the analytical methods.

The main problem area is the replication of the very rapid rates of heating associated with certain welding processes. For instance, in manual metal arc (MMA) welding, the time to achieve a 1300°C (2372°F) peak temperature in the HAZ is usually in the order of two to four seconds(s). On the other hand, the maximum heating rate achievable by some of the analytical methods can be significantly slower, with values of 7 to 12 s to attain a similar peak temperature being reported (Ref. 1, 4, 6, 7, 8).

The direct methods study "in situ" the phase transformations that actually occur during welding. Because reliable results are inherently much more difficult to obtain by these methods, considerably less attention has been paid to them.

To date, two different approaches have been attempted as discussed below under separate headings, these are dilatometry and thermal analysis.

Dilatometry

A special dilatometer was used by Hofman and Burat (Ref. 10) to measure transformation temperatures in the HAZ's of actual welds. In these experiments, the feelers of the dilatometer were inserted into two holes drilled near the surface of the V-groove of a plate to be butt welded. The relative movement of the two feelers was translated into an electrical signal and fed into an X-Y recorder. The feelers were at the same time wires of a thermocouple, the thermocouple circuit being closed by the steel within the gauge length.

Unfortunately this technique has certain experimental drawbacks. For instance, the dilatometer was a complicated apparatus and required considerable development to achieve operational status. Furthermore, it was reported that it was difficult to explore those parts of the HAZ immediately adjacent to the fusion boundary because of possible damage to the thermocouples. It is just these regions which are of greatest interest in studies of transformation temperatures.

Thermal Analysis

The thermal analysis technique was originally developed by Granjon and Gaillard (Ref. 11-13). In this method, a cylindrical implant of test material was inserted into a plate of the thickness to be welded. A thermocouple was then inserted into a blind hole in the implant in such a position as to become part of the HAZ during welding.

The decomposition of austenite is always exothermic (heat liberated), and the transformation temperatures were determined by thermal analysis of the resulting cooling curve using an electronic differentiator, and were read off directly from the discontinuities in the thermal analysis curve:

$$\frac{dT}{dt} \text{ vs. } t$$

R. H. PHILLIPS is with the Welding and Casting Section of the Materials Research Laboratories, Department of Defence Support, Ascot Vale, Victoria, Australia.

where T is temperature and t is time.

This technique has since been used by other workers (Ref. 15), and has also been modified for the determination of transformation temperatures occurring in weld metal (Ref. 16).

Problems and a Solution

The main experimental problem with the electronic differentiation method of thermal analysis lies in the elimination of "noise" from the

$$\frac{dT}{dt}$$

axis (Ref. 8). This "noise" is electronic and arises from low frequency electric currents induced into the thermocouple leads by magnetic effects associated with the welding arc. In addition, there also appeared to be some difficulty in accurately determining transformation finish temperatures.

Although some sample curves obtained from the thermal analysis technique have been published (Ref. 12-15), a rigorous assessment of its accuracy and sensitivity over a wide range of welding conditions does not appear to have been made. In particular, the sensitivity of the technique in detecting small amounts of a particular transformation in a multiple transformation event, commonly found in the weld HAZ, has not been established.

A novel technique of thermal analysis which does not require an electronic differentiator has been developed by the present author at Materials Research Laboratories, for use in determining transformation temperatures under simulated welding conditions (Ref. 7). In this method, the temperature difference between two ends of a gas-quenched specimen is plotted against the temperature of the cooler point. The moment that the cooler end begins an exothermic phase transformation, its cooling rate is retarded and the hotter end begins to catch up. The point at which the temperature difference starts to diminish rapidly therefore marks the transformation starting temperature.

This technique was found to be extremely sensitive and accurate in determining transformation temperatures; it had the added advantage of requiring a minimum of experimental equipment (only an X-Y recorder and a temperature-time recorder being required for the production of a CCT diagram).

It was therefore decided to modify this technique for the "in situ" determination of HAZ transformation temperatures under actual welding conditions, and to assess its sensitivity in detecting transformations over a wide range of welding conditions. A C-Mn steel was used for these experiments. In addition, the transformation behavior of the C-Mn steel determined under actual welding conditions was compared with its transformation behavior under simulated welding conditions.

Experimental Procedures and Results

Experimental Setup

The experimental technique used to determine HAZ transformation temperatures under actual welding conditions combined the technique of implant as proposed by Granjon (Ref. 11), in which a plug of test material is inserted into a mild steel host plate, with the above mentioned method of thermal analysis developed at the Materials Research Laboratories (Ref. 7). The "implant" technique was used, because it greatly reduced the amount of test material required. This is clearly a significant advantage where only limited amounts of material are available or if the material is not in plate form of the appropriate thickness.

All plugs of test material were machined from a hot rolled carbon-manganese steel plate, code XK 1320, whose composition is shown in Table 1. The dimensions of the mild steel host plate were 240 mm (9.45 in.) long and 115 mm (4.5 in.) wide. In order to cover a wide range of cooling rates, plate thicknesses of 6, 12.5 and 25 mm (¼, ½ and 1 in.) were used. The cylindrical "implant" of test material was 10 mm (0.39 in.) diameter and the same length as the host plate

Table 1—Composition of Experimental Materials, %

	XK 1320 C-Mn steel	Undiluted weld metal
C	0.24	0.09
Mn	1.59	1.07
Si	0.23	0.32
S	0.019	0.016
P	0.024	0.014

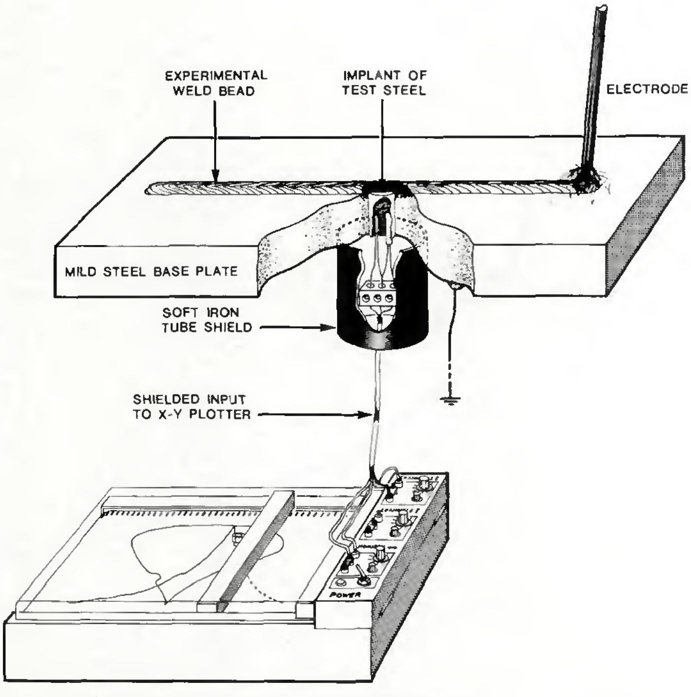

Fig. 1—Schematic overview of experimental setup

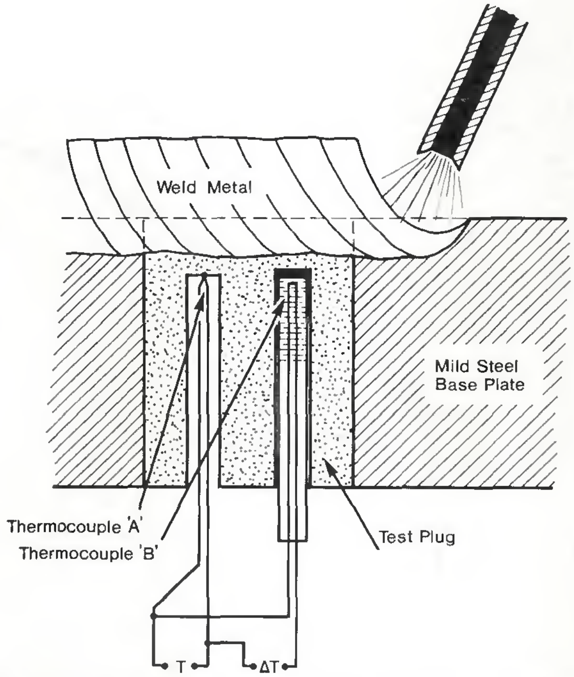

Fig. 2—Detail of thermal analysis setup for the measurement of "in situ" HAZ transformation temperatures

thickness. Plugs were a drive fit in the host plate. The experimental weld beads were deposited using an automated metal arc process, with a low hydrogen, carbon-manganese steel electrode. Both travel speed and electrode feed rate were automated. The composition of the undiluted weld metal is shown in Table 1. The overall experimental set-up is shown schematically in Fig. 1.

The method of thermal analysis used to determine the transformation temperatures involved inserting chromel-alumel thermocouples into two blind holes drilled in the implant in such a position as to become situated in the coarse grained HAZ adjacent to the fusion boundary during welding. This is shown in detail in Fig 2. Thermocouple A, the temperature measuring couple, was sheathed in a two-hole alumina insulator and welded to the top of the 1.35 mm (0.05 in.) diameter blind hole using a capacitor discharge apparatus (see Fig 2). Thermocouple B, the reference couple, was a mineral insulated, isolated junction thermocouple encased in stainless steel sheath of 1.0 mm (0.04 in.) diameter.

It was found necessary to use an isolated junction for thermocouple B, instead of the standard type, in order to reduce "noise" on the ΔT axis. This noise was caused by low frequency electric currents in the thermocouple leads induced by the magnetic effects associated with the welding arc. Although the use of pyrotenax thermocouples reduced the noise level very substantially, further modifications to the system were necessary to eliminate it completely. These modifications included shielding the X-Y recorder input leads and encasing the thermocouple wires and cold junction box in a soft iron tube.

The thermocouples were connected so as to measure the temperature (T) at the cooler point during the cooling cycle (thermocouple A), and the temperature difference (ΔT) between the warmer and the cooler point, (B-A). The resulting curve of ΔT v T was plotted on an X-Y recorder. (Examples of such curves are shown in Figs. 3 and 4). Simultaneously, the temperature-time curve was recorded.

In determining the transformation temperatures from the thermal analysis graphs (see Figs. 3 and 4), the onset of transformation was taken as that point where the temperature difference during cooling starts to diminish rapidly. Note that thermocouple A is on the cooler side and first registers any transformation which in turn causes the temperature difference between B and A to decrease. Because the amount of heat evolved at the thermocouple junction is relatively small, it is important that the temperature difference between A and B also be relatively small so that the most sensitive recorder scale can be used. This provides a high degree of magnification of the heat sensing effect. The completion of transformation was taken as the point at which the deviation from a smooth curve in the transformation discontinuity appeared to end.

In many cases it was difficult to accurately determine this point due to the relatively slow decay in the transformation rate as the transformation finish temperature was approached.

Location of Thermocouples

Preliminary bead-on-plate tests were conducted to determine the appropriate thermocouple positions which would be necessary to record both a peak temperature of 1300°C (2372°F) and also produce a thermal analysis curve of the required sensitivity. For weld beads deposited at a heat input of 1.4 kJ/mm (35.6 kJ/in.), the grain coarsened region of the weld was observed to be 1.10 mm (0.04 in.) wide and a peak temperature of 1300°C (2372°F) was attained when the thermocouple was located 0.38 mm (0.015 in.) from the fusion boundary.

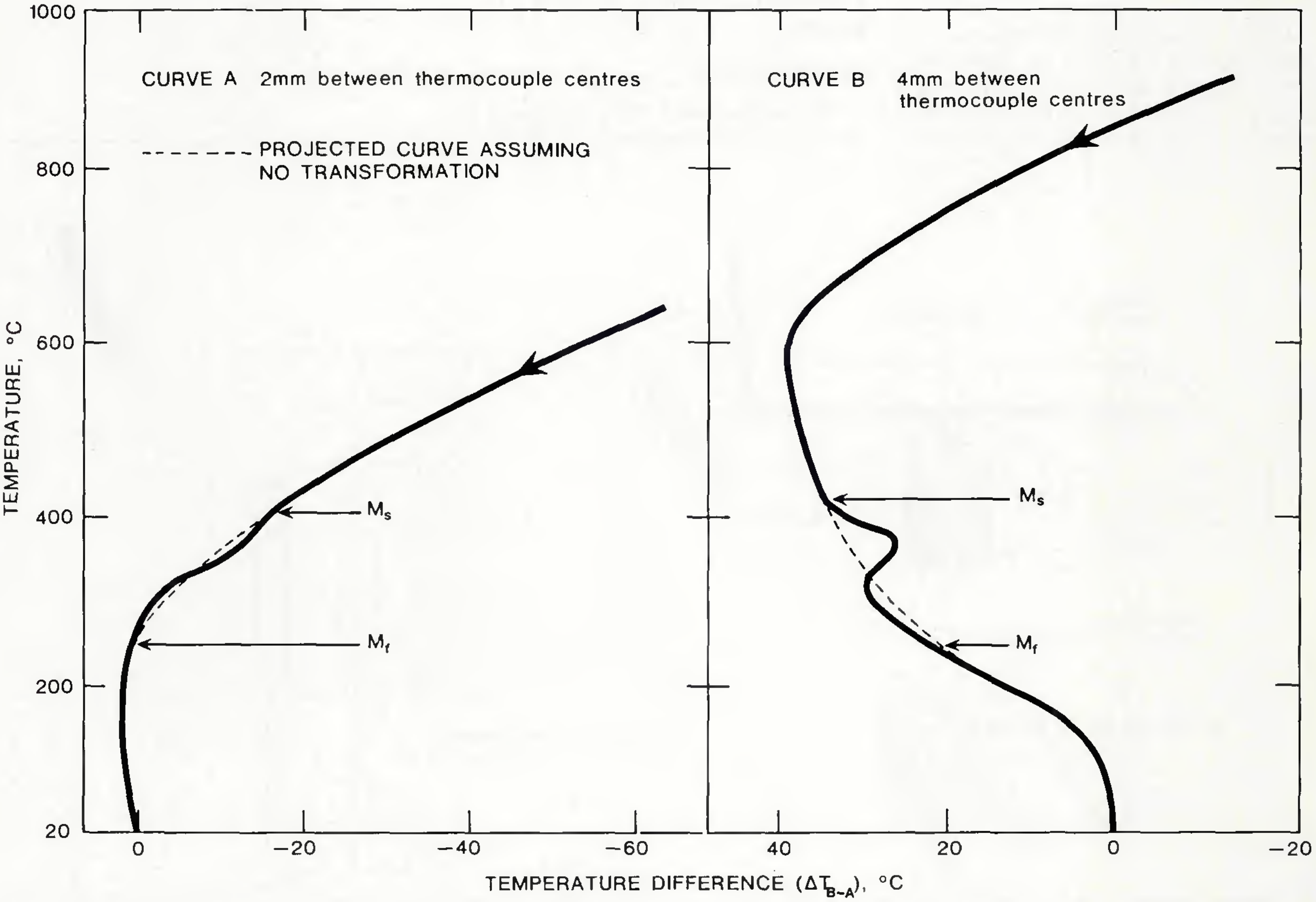

Fig. 3 — Effect of distance between thermocouples on form and sensitivity of thermal analysis curve. Cooling rate 1

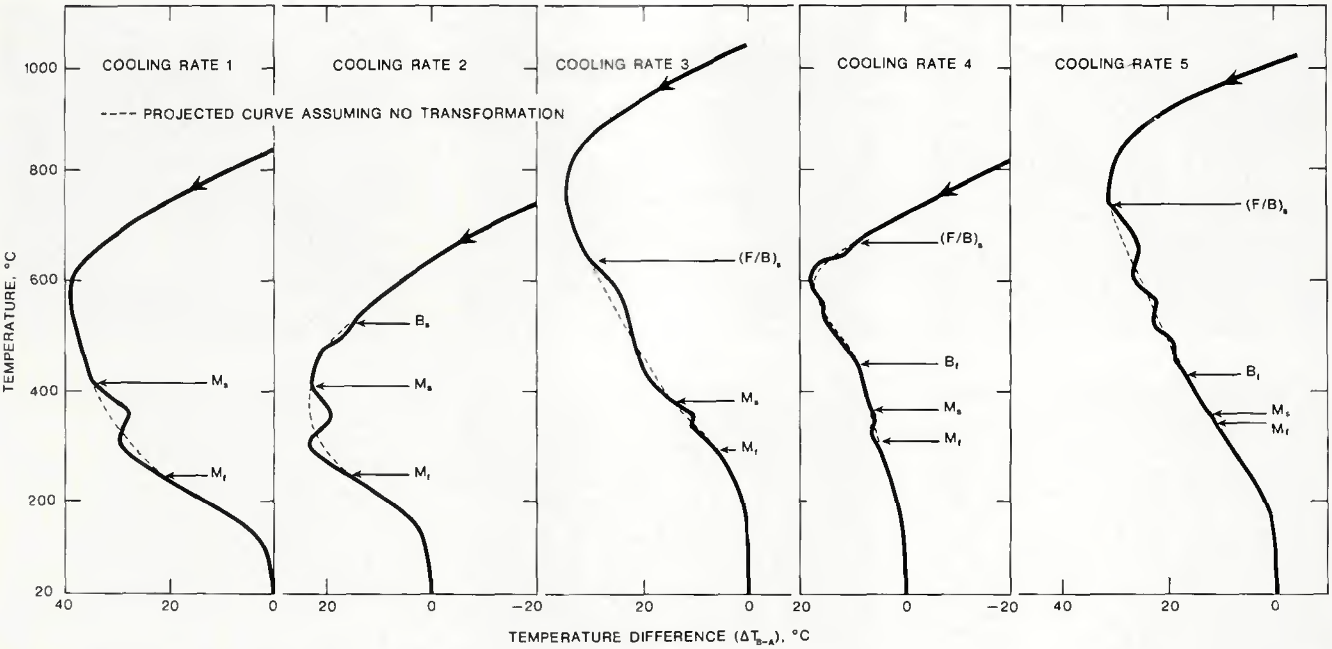

Fig. 4 — Thermal analysis curves for cooling rates 1 to 5

The optimum sensitivity for detecting transformation temperatures from the thermal analysis curve was achieved when $\Delta T_{B\text{-}A}$ was slightly positive (between 0 and 40°C, i.e., 32 and 104°F) over most of the cooling cycle below 1000°C (1832°F). This was achieved by placing the thermocouples with their centers 4 mm (0.16 in.) apart. If the thermocouples were separated by significantly greater distances — say, 6 mm (¼ in.) — then $\Delta T_{B\text{-}A}$ became too large and overshot the optimum recorder scale. On the other hand, if the thermocouples were placed too close together — say, 2 mm (0.08 in.) apart — the value of $\Delta T_{B\text{-}A}$ remained negative over almost the entire cooling and the sensitivity in detecting transformation temperatures was greatly reduced.

An example of the effect of which the distance between the thermocouples has on the form of the thermal analysis curve, and hence its sensitivity in detecting transformation temperatures for the fastest cooling rate (Ref. 1) used, is shown in Fig 3.

Determination of the "In Situ" CCT Diagram for XK 1320 Steel

In order to more fully evaluate the "in situ" method of determining transformation temperatures outlined previously, thermal analysis curves were produced for a wide range of cooling rates. These cooling rates covered the range which would normally be encountered in the HAZ during metal arc welding. This was achieved by using appropriate combinations of plate thickness (6 to 25 mm, i.e., ¼ to 1 in.) and heat input (1.0 to 1.8 kJ/mm or 25.4 to 45.7 kJ/in.).

Five typical thermal analysis curves covering the range of cooling rates used are shown in Fig. 4; the transformation temperatures are marked on these curves. An "in situ" CCT diagram for the experimental C-Mn steel was then derived from all thermal analysis experiments conducted — Fig. 5.

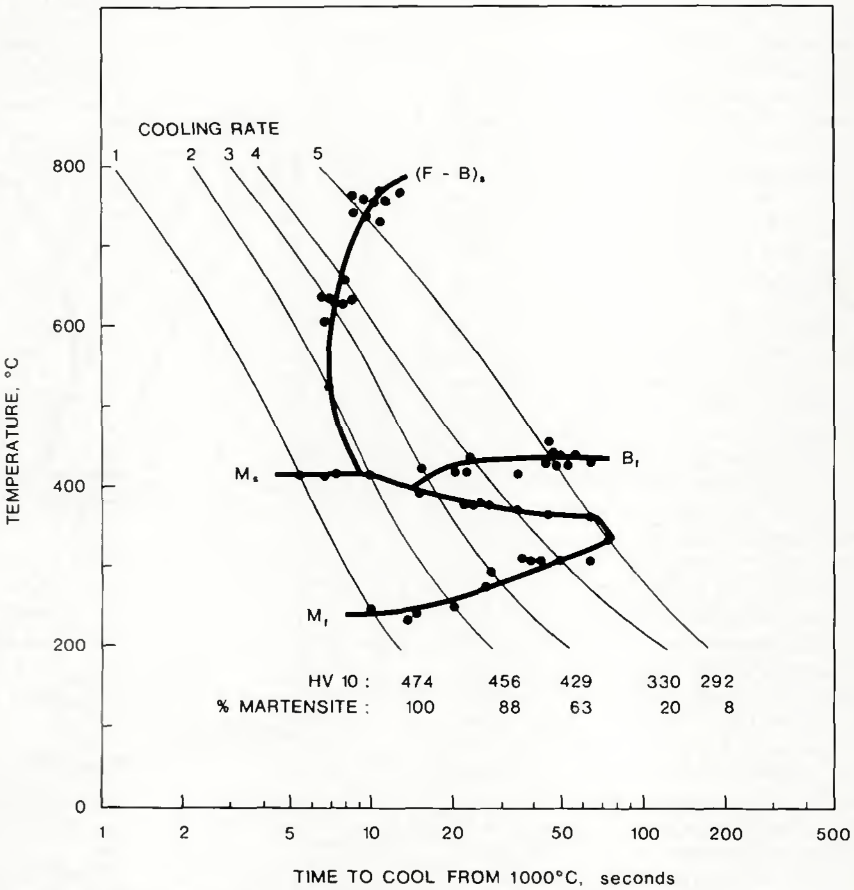

Fig. 5 — "In situ" CCT diagram for XK 1320 steel

All implants of the XK 1320 steel used in the thermal analysis experiments were sectioned parallel to the welding direction to include the thermocouple locations and examined metallographically. The microstructures and hardness values

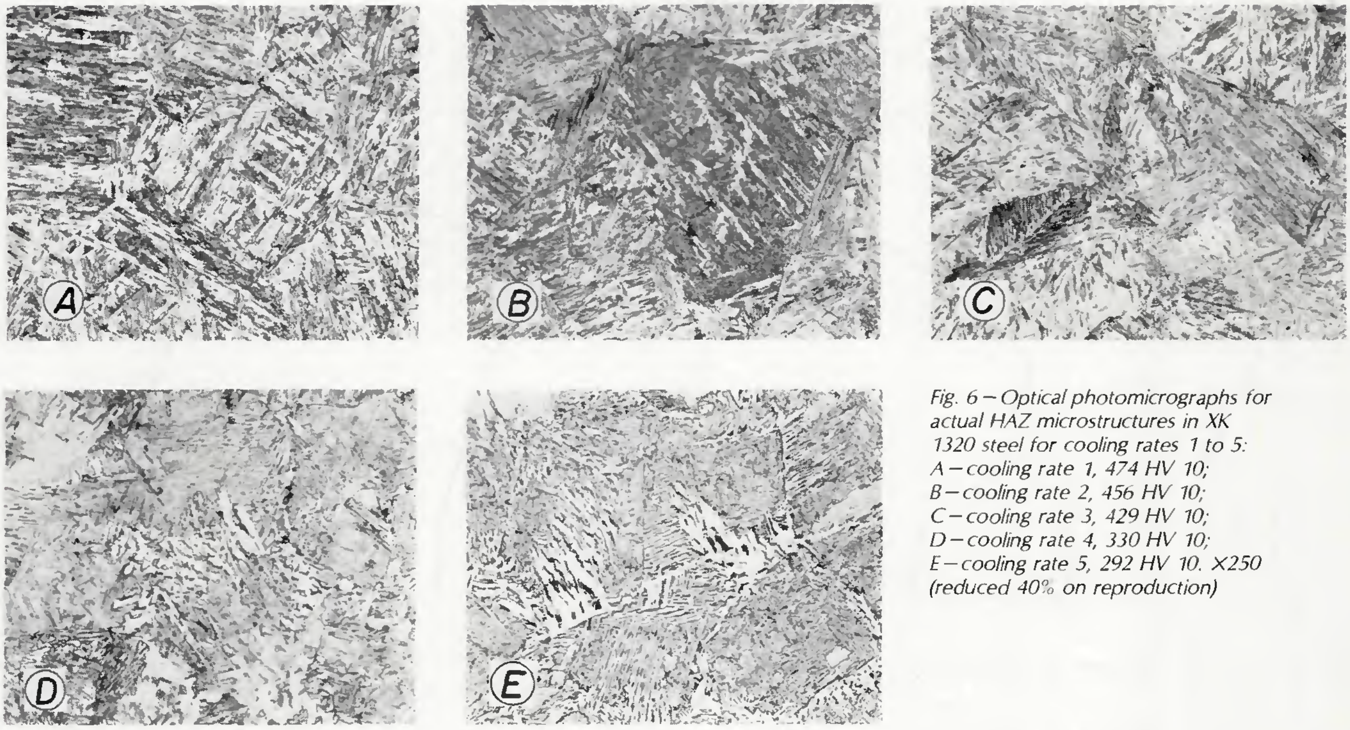

Fig. 6 — Optical photomicrographs for actual HAZ microstructures in XK 1320 steel for cooling rates 1 to 5: A — cooling rate 1, 474 HV 10; B — cooling rate 2, 456 HV 10; C — cooling rate 3, 429 HV 10; D — cooling rate 4, 330 HV 10; E — cooling rate 5, 292 HV 10. X250 (reduced 40% on reproduction)

corresponding to the five cooling rates mentioned above are shown in Fig. 6. The micrographs were taken with centers 0.38 mm (0.015 in.) from the fusion boundary. This corresponds to a peak temperature isotherm of 1300°C (2372°F).

Evaluation of the thermal analysis curves in Figs. 4 and 5 showed that the transformation product was fully martensitic only at the fastest cooling rates. For most cooling rates investigated, a mixed ferrite/bainite/martensite transformation product was indicated, whilst for the slowest cooling rates (i.e. slower than rate 5) the martensite transformation could not be detected. It can be seen from Fig. 4 that the size of the martensite inflection decreases rapidly with cooling rate.

Reference to the CCT diagram (Fig. 5) shows the reproducibility achieved in the determination of transformation temperatures using the "in situ" thermal analysis method. For M_s temperatures a reproducibility of ± 7°C (± 12°F) was achieved, while for M_f this fell to ± 20°C (± 36°F). For the higher temperature ferrite-bainite transformation reproducibility was somewhat poorer being ± 22°C (± 40°F) for $(F\text{-}B)_s$ and B_f.

Metallographic examination of the test implants showed that the microstructures adjacent to the fusion boundary varied as follows: from fully martensitic at the fastest cooling rates, through mixed martensite-bainite to a mixture of pro-eutectoid and Widmanstatten ferrite and bainite and a small amount of martensite at the slowest cooling rates. At cooling rate 5, 8% martensite was present while, at the slowest cooling rate achieved, only 3% martensite was detected. The concentration of the respective phases was determined using a linear intercept method.

At the slow cooling rates it was noted that the thermal analysis method could not detect martensite when present in concentrations less than 8%. At the faster cooling rates, the bainite transformation was detected clearly when the final proportion of bainite was 12% or greater (see cooling rate 2, Fig. 4).

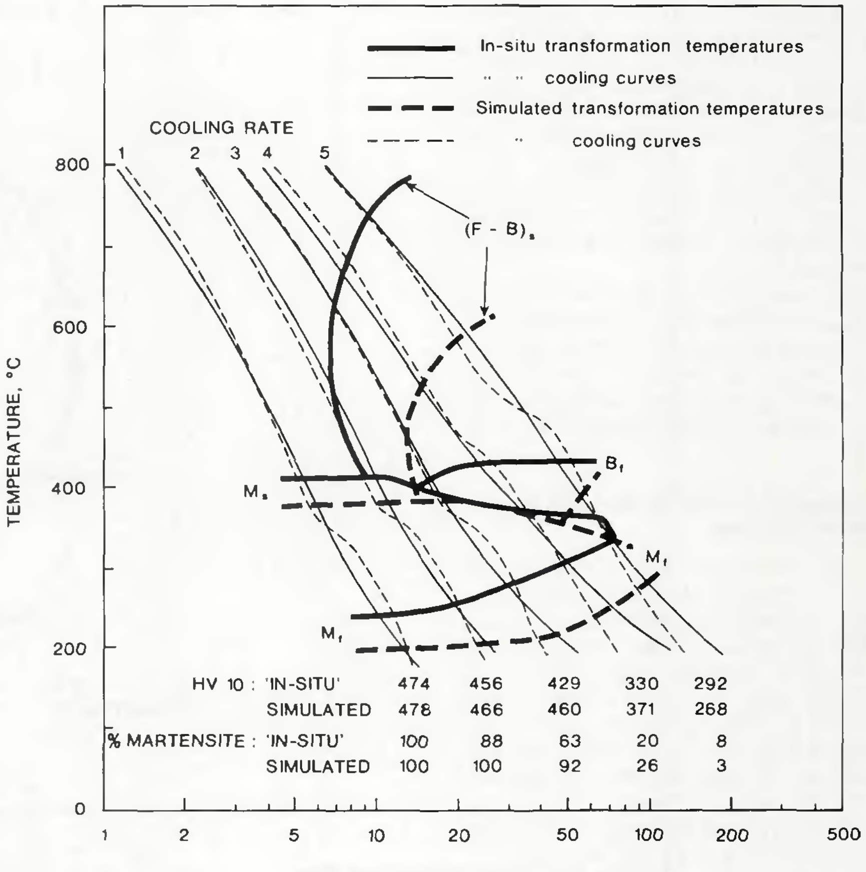

Fig. 7 — Comparison between "in situ" and simulated CCT diagrams for XK 1320 steel

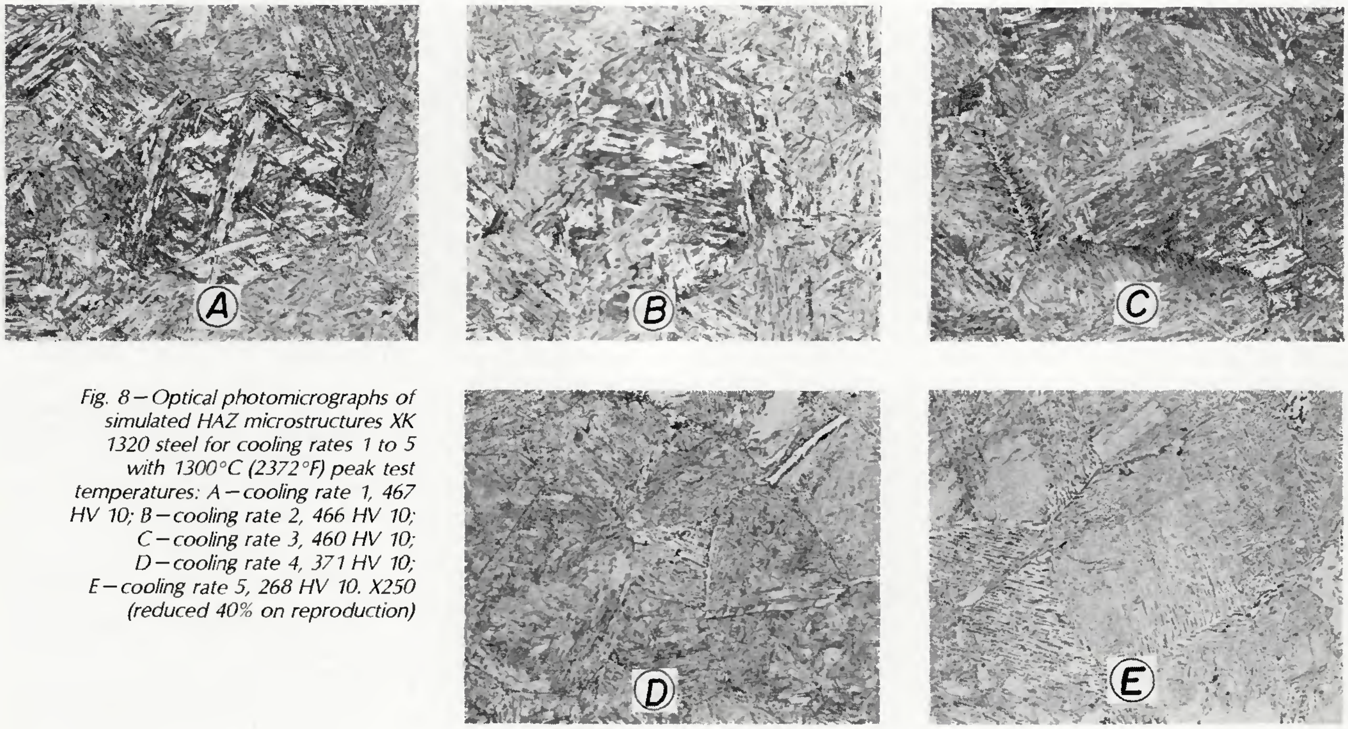

Fig. 8 — Optical photomicrographs of simulated HAZ microstructures XK 1320 steel for cooling rates 1 to 5 with 1300°C (2372°F) peak test temperatures: A — cooling rate 1, 467 HV 10; B — cooling rate 2, 466 HV 10; C — cooling rate 3, 460 HV 10; D — cooling rate 4, 371 HV 10; E — cooling rate 5, 268 HV 10. X250 (reduced 40% on reproduction)

Comparison Between CCT Diagrams Derived from Actual and Simulated Welding Conditions

For comparison with the CCT diagram produced under actual welding conditions, a second CCT diagram for the experimental C-Mn steel was produced using simulated welding conditions. In these experiments small tubular specimens (6.4 mm long, 6.4 mm OD and 3.2 mm ID, i.e., ¼ × ¼ × ⅛ in.) of the XK 1320 steel were thermally cycled to a peak temperature of 1300°C (2372°F) in a rapid response dilatometer (Ref. 4, 7) to simulate welding conditions.

The time to reach 1300°C (2372°) was 8 s. Thermal cycling was achieved by induction heating and inert gas quenching. Transformation temperatures on cooling were determined using dilatometry and confirmed using the method of thermal analysis described previously.

The CCT diagram obtained using the simulated welding conditions at five standard cooling rates is shown in Fig. 7. The microstructures obtained for the five standard cooling rates are shown in Fig. 8.

Comparison between the "in situ" and simulated methods of determining transformation temperatures in the HAZ (Fig. 7) showed that, although the respective CCT diagrams were broadly similar for the C-Mn steel investigated, some significant differences were nonetheless apparent.

For instance, it can be seen that both the martensitic and non-martensitic transformations commenced at lower temperatures under simulated welding conditions than recorded in the actual HAZ. In the case of martensitic transformations, differences up to 34°C (61°F) were noted, while for the higher temperature transformation products differences up to 140°C (252°F) were noted. In addition, bainite first appeared at somewhat faster cooling rates in the actual HAZ, than under simulated welding conditions.

Comparison of hardness values (Fig. 7) showed relatively small differences. No significant difference was observed at the fastest cooling rate. However, for cooling rates 2, 3 and 4, the simulated HAZ's were up to 41 HV, 10 points higher than actual HAZ values. This situation was reversed at cooling rate 5 where the actual HAZ was 24 HV, 10 points higher than the simulated condition.

The microstructures of the actual HAZ (Fig. 6) were broadly similar, at a given cooling rate, to those obtained under simulated welding conditions (Fig. 8). Nonetheless, some significant differences were again observed. For instance, at the faster cooling rates the proportion of bainite tended to be higher in the actual HAZ than the simulated HAZ. At cooling rate 5, however, the proportion of martensite was higher in the actual HAZ (8%) than the simulated HAZ (3%).

At the slower cooling rates, the high temperature transformation products have a much coarser appearance in the actual HAZ than the simulated HAZ. For cooling rate 5, pro-eutectoid ferrite outlines the grain boundaries in the actual HAZ but is not so readily apparent in the simulated HAZ.

Discussion

It was apparent from the results that the method of "in situ" thermal analyses described was extremely sensitive in detecting HAZ transformations during cooling. For the C-Mn steel investigated, almost all the transformations occurring over a wide range of cooling rates were detected. It was noted, however, that the technique is not sufficiently sensitive to detect small concentrations of a particular transformation product in a mixed transformation. For instance, at slow cooling rates, the martensite transformation was not detected at concentrations less than 8%. It is possible, however, that increasing the distance between the two thermocouples would increase ΔT and thereby increase sensitivity.

No quantitative estimate was made of the lowest concentration of bainite which could be detected using the thermal analysis technique because of lack of data between cooling rates 1 and 2. Certainly the bainite transformation was very clearly indicated at cooling rate 2 where subsequent metallographic evaluation showed 12% bainite.

The reproducibility of the recorded transformation start temperatures was ± 7°C (± 13°F) for M_s, which was considered very good, and ± 22°C (± 40°F) for start of the higher temperature transformation $(B\text{-}F)_s$. The latter value was considered acceptable in the knowledge that the high temperature transformations usually commence more slowly than the martensite transformation, and it is thus more difficult to determine the initial deviation.

The reproducibility of the recorded transformation finish temperatures was ± 20°C (± 36°F) for M_f and ± 22°C (± 40°F) for B_f. Although this reproducibility was considered acceptable, it was perceived that the recorded transformation finish temperatures might be subject to a systematic error due to the slow decay in the rate of transformation and hence heat liberation as the finish temperature was approached. This would result in the transformation finish temperature recorded using "in situ" thermal analysis being somewhat higher than the actual transformation finish temperature.

An indication of the probable magnitude of this difference was obtained by comparing transformation finish temperatures determined using rapid response dilatometry and thermal analysis under simulated welding conditions (Ref. 7). These results showed that the maximum systematic error was 35°C (63°F) for M_f and 25°C (45°F) for B_f, i.e., the transformation temperature determined using thermal analysis was up to 35°C (63°F) higher than actual for M_f and 25°C (45°F) for B_f.

Apart from its sensitivity in detecting HAZ transformations, the "in situ" thermal analysis method outlined has the added advantage of being particularly easy to set up; it requires only a very modest amount of experimental equipment—namely, two sets of thermocouples, one X-Y recorder and one time-temperature recorder. The present thermal analysis method also appears to have one other inherent advantage over the method electronic differentiation of cooling curves in that it is probably less sensitive to "noise" on the ΔT axis (Ref. 8). Certainly for the present thermal analysis method the simple precautions outlined previously are sufficient to eliminate all noise, at the level of sensitivity selected.

Comparison between the CCT diagrams, and corresponding microstructures, for the XK 1320 steel determined "in situ" and under simulated welding conditions showed that, although they were broadly similar, some significant differences were apparent. For instance, the bainite transformation appeared at considerably faster cooling rates in the actual HAZ than under simulated welding conditions. Also, there was a greater proportion of bainite in the actual HAZ microstructure at the faster cooling rates. This position was reversed at the slowest cooling rate where the proportion of martensite in the actual HAZ was somewhat higher than in the simulated HAZ.

Differences were also noted between transformation temperatures recorded using the two techniques. For instance, the $(B\text{-}F)_s$ temperature recorded "in situ" was 100 to 150°C (180 to 270°F) higher than the corresponding $(B\text{-}F)_s$ temperatures recorded under simulated conditions, at similar cooling rates. This was consistent with the comparison of microstructures, which, at the slower cooling rates, were considerably coarser for the actual HAZ than for simulated HAZ's. This is well exemplified in rate 5 where the formation of grain boundary pro-eutectoid ferrite is quite pronounced in the actual HAZ but far less evident in the simulated HAZ.

It was difficult to explain these differences unambiguously; they could be due to a number of factors, such as steep temperature gradients in the actual HAZ, severe restraint in the HAZ or differences in the heating cycle. In this latter respect it took 8 s to achieve a peak temperature of 1300°C (2372°F) under simulated welding conditions, while under actual weld conditions, it took between 2 to 4 s to reach a similar peak temperature.

Conclusions

1. A novel method of thermal analysis has been used to determine the transformation temperatures occurring in the HAZ of actual welds during cooling. This method was shown to be very sensitive in detecting the phase transformation occurring in the HAZ of a C-Mn steel.

2. The method has the added advantage of being simple to apply and requires a minimum of experimental equipment.

3. Rigorous examination over a wide range of cooling rates has shown that the method can detect almost all phase transformations that occur during cooling. The notable exception was that the martensite transformation was not detected when present in concentrations less than 8%, at slow cooling rates.

4. In the determination of transformation temperatures, the reproducibility of M_s was ± 7°C (± 13°F), $(F\text{-}B)_s$ was ± 22°C ± 40°F), M_f ± 20°C (± 36°F) and B_f ± 22°C (± 40°F). Because of systematic error in the determination of transformation finish temperatures it was possible that the M_f temperature recorded using thermal analysis was a maximum of 35°C (63°F) above actual, while B_f was a maximum of 25°C (45°F) above actual.

5. Comparison between the "in situ" and simulated methods of determining HAZ transformation temperatures showed that, while the CCT diagrams were generally similar in form, under simulated welding conditions the whole diagram was depressed to lower temperatures and longer times.

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