

CRACKS – RUNNING HOT AND COLD A TALE OF WELD METAL INTEGRITY

*DUNNE D. P., **ALAM N., ***LI H., AND ****CHEN L.

* Professor of Materials Engineering at the University of Wollongong, NSW, Australia.

** CRC postdoctoral fellow at the University of Wollongong and is currently a Research Scientist, CSIRO Manufacturing Science and Technology, Adelaide.

*** CRC postdoctoral fellow at the University of Wollongong and is currently a Research Scientist, ANSTO, Sydney.

**** Postgraduate research student at the University of Wollongong.

Paper taken from the proceedings of the "International Institute of Welding – Asia Pacific International Congress" Melbourne 29th Oct – 2nd Nov., 2000.

ABSTRACT

The welding of steel is a casting process in which molten steel solidifies within the confines of a steel mould formed by the plates being joined. Without counteractive measures it is therefore subject to common casting defects such as segregation, cold shuts, contamination, oxidation, gas and shrinkage porosity and hot tearing. However, welding differs in several major respects from batch or continuous casting processes. One obvious difference is that welding is a joining process in which the cast metal unifies with the metal mould and, as a consequence, stresses develop from the restraint to shrinkage imposed by the mould. In addition, the mould material melts and alloys with the weld filler metal to form a "diluted" slug of weld metal. Welding can also be regarded as a mini-electric-arc steelmaking process in which molten metal interacts with elements in the arc plasma and with fluxing materials to produce a weld bead with a high volume fraction of non-metallic inclusions, compositional gradients and segregation, as well as relatively high concentrations of the gaseous elements oxygen, nitrogen and hydrogen.

This gassiness, the effect of restraint and compositional segregation increase the susceptibility to "hot" cracking at temperatures near the primary solidus and to "cold" cracking at temperatures well below the primary A_{r3} temperature of the weld metal. This paper reviews hot cracking of carbon steel weld metals and discusses recent research results on hot and cold cracking of pipeline root pass girth welds, the interactivity between the two types of cracking and welding conditions which reduce the likelihood of cracking.

Keywords :

Hot cracking, solidification cracking, weld metal, restraint force, transformational strains, shrinkage strains, peritectic reaction, effect of sulphur and phosphorus, pipeline girth welds.

INTRODUCTION

Both "hot" and "cold" cracks can occur in steel weldments. Cold cracks form at temperatures below about 200°C, initiating and propagating in the ferritic structure of carbon steels. Cold cracking is usually assisted by hydrogen acting at stress concentrating macroscopic or microstructural features, in the presence of residual stresses arising from restraint of the thermal and transformational strains generated on cooling of the weldment. The nature and mechanisms of hydrogen assisted cold cracking (HACC) are covered in several recent reviews [1-3] and the main thrust of the present paper is an analysis of hot cracking, with a brief reference to its interaction with cold cracking.

Hot cracking occurs at high temperatures shortly after weld deposition, during or soon after primary solidification to either delta ferrite or austenite. Hot cracking can also occur in the solid state at high temperatures due to poor hot ductility in the presence of tensile restraint stresses; and at lower (but still elevated) temperatures during postweld heat treatment (re-heat cracking). Hot cracks are frequently intergranular, initiating and growing along grain boundaries.

The term 'solidification cracking' is used to distinguish hot cracking that occurs in weld metal at temperatures below the liquidus temperature but within the solidification range of the

steel. These cracks occur predominantly at the weld centre-line or between columnar grains ('dovetail' or 'flare' cracking [4]). The crack surface usually shows a cellular-dendritic solidification morphology. Solidification cracking is also common in the weld crater at the end of a weld run.

Weld metals solidifying as ferrite can be susceptible to solidification cracking, but solidification to austenite can be even more problematic, because of the lower solubility of S and P in austenite, resulting in their rejection and concentration in the interdendritic liquid.

MECHANISMS OF SOLIDIFICATION CRACKING

It is well known that one of the major causes of solidification cracking is partitioning or rejection of particular alloying elements at columnar grain boundaries, ahead of the advancing solid-liquid interface [5-7]. Solidification cracking can be produced by the presence of low melting compounds forming as a result of segregation during the liquid to solid phase transformation, and/or by the presence of welding stresses. As the solidifying weld metal contracts, regions which remain liquid to relatively low temperature may be ruptured, because of contraction strains transmitted through the surrounding continuous network of solidified metal. Cracking may then progress in the solid state

because of low intergranular ductility and the increasing restraint stress. Even in cases where no appreciable segregation occurs, solidification 'cracking' can still develop due to failure of liquid feeding to compensate for solidification shrinkage, i.e. if the geometry of the fusion zone is unfavourable.

During the weld solidification process, columnar grains grow epitaxially from the fusion boundaries and as a result of the relatively wide range of solid compositions developed over the freezing range (low partition coefficients), some alloying elements are rejected into remnant liquid between the cells/dendrites and ahead of the advancing solid-liquid interface, giving rise to marked segregation. The liquid containing the partitioned elements can freeze to a compound at a temperature well below the primary solidus. Alternatively, the liquid can be a low freezing eutectic liquid or form a highly wetting liquid film along the boundaries. During the solidification process weld metal contracts, restraint stresses develop and under the influence of tensile forces, rupture of liquid/semi-liquid films, or cracking of solidified low ductility zones, can occur in the interdendritic or intercellular regions [8], see Fig. 1.

A higher welding speed induces the dendrites to grow from the two sides of the weld bead towards the central plane and therefore, strong

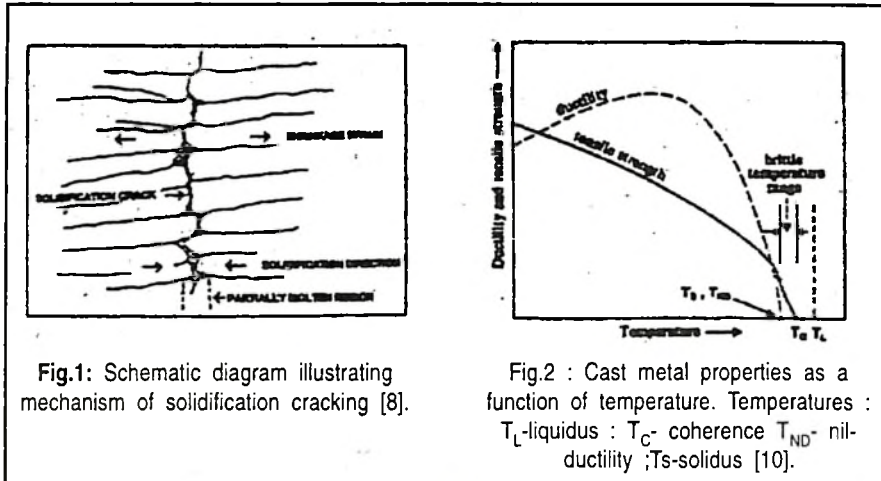


Fig.1: Schematic diagram illustrating mechanism of solidification cracking [8].

Fig.2 : Cast metal properties as a function of temperature. Temperatures : T_L -liquidus ; T_C - coherence ; T_{ND} - nil-ductility ; T_S -solidus [10].

segregation of impurities can occur at the weld central plane, increasing the propensity for solidification cracking. Thermal stresses play an important role in solidification cracking. Generally, end crater cracking is associated with high transverse stresses, and centre-line cracking may also be promoted by high longitudinal stresses [9]. Dovetail cracking can be induced by high residual stresses associated with relatively coarse columnar grains. This kind of cracking generally occurs in concave regions of the weld bead [4].

In general, solidification cracking will develop if the tensile stress arising from shrinkage exceeds the local fracture strength of the weld bead at any temperature in the solidification range [10]. Low fracture strength may be associated with low ductility resulting from the semi-solid nature of the weld bead. Once a solidifying weld bead forms a continuous ('coherent') network (at the coherence temperature, T_C) it

acquires mechanical strength which rises steeply as the temperature falls and solidification progresses. However, the semi-solid weld bead has little tensile ductility because of easy rupture of liquid regions at low strains. This condition applies above the nil-ductility temperature, T_{ND} which is close to the solidus temperature, T_S , [10]. The temperature range between T_C and T_{ND} is called the brittle range and weld metals with a wide brittle range have increased susceptibility to solidification cracking (Fig. 2). If the tensile restraint stress rises more steeply than, and exceeds, the tensile strength of the weld bead, then cracking will occur. The tensile restraint stress experienced by the solidified network is given by $xE\alpha$ where E and α are the Young's Modulus and thermal contraction coefficient at the temperature of the weld bead. For steels, phase transformation imposes extra restraint as discussed in Section 3.2 and 7. The factor x is the restraint factor which is a function of the type

of joint, the weld gap, plate thickness, plate strength and the relative thickness of the weld bead and the plate. The restraint increases for small bead/plate thickness ratio and a rigid joint set-up.

EFFECTS OF ALLOYING AND IMPURITY ELEMENTS

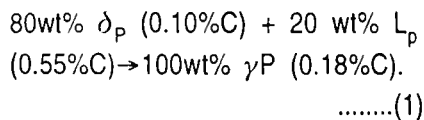
Alloying elements have a strong influence on solidification structure and hot cracking behaviour. Basically, any alloying element that lowers the melting temperature and increases the freezing range will increase the risk of cracking. Elements which segregate strongly between the solidifying metal and the liquid (ie. those with low partition coefficients) increase the cracking tendency. In general, elements causing solidification cracking in steels

- have a low partition coefficient between liquid and solid (γ or δ - ferrite, depending on how the steel solidifies);
- readily form low melting point compounds or eutectics with the base metal; and
- promote a low wetting angle between solute enriched liquid and base metal, resulting in liquid films around the solidifying metal surface.

In addition, variations in solubility of alloying elements in the two solid phases of iron can increase the risk of solidification cracking.

The Peritectic Reaction

The presence of a peritectic reaction at high temperature in the Fe-C binary system (Fig. 3) makes steel vulnerable to solidification cracking. At high temperature, steels with carbon contents between 0.10-0.55wt% (the peritectic composition range) start to solidify as δ -ferrite, then form γ phase. Generally, low alloy Fe-C weld metals with $< 0.10\%C$ solidify as δ . For the peritectic alloy (0.18%C) the primary crystals are δ phase but a peritectic reaction occurs at $1492^{\circ}C$ which produces γ by reaction between δ_p and peritectic liquid, L_p :



For $0.10\% < C < 0.18\%$, the primary delta phase δ_p (0.10%C) reacts with peritectic liquid L_p (0.55%C) at $1492^{\circ}C$ to produce peritectic austenite γ_p and excess δ_p . For hyper-peritectic alloys (0.18-0.55%C) excess liquid remains on cooling below $1492^{\circ}C$, and solidification to γ then occurs. Hence, the freezing range is increased. These changes increase the propensity for solidification cracking. Although the foregoing discussion applies to equilibrium freezing the actual solidification process is highly dynamic and the high temperature range of solidification should ensure a rapid approach to equilibrium, particularly for the peritectic reaction and the

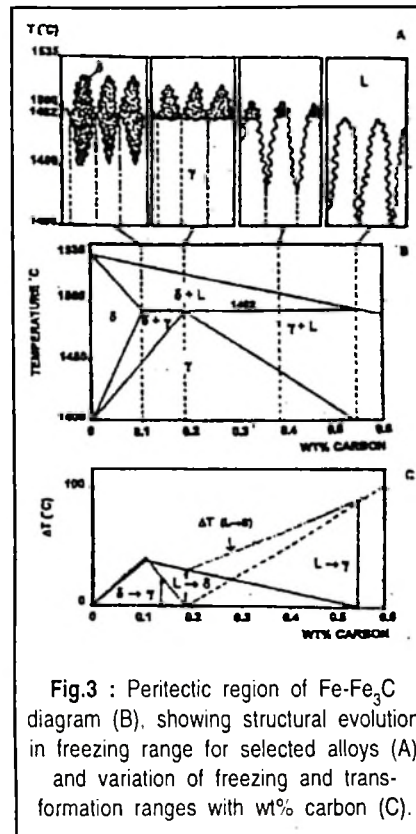


Fig.3 : Peritectic region of Fe-Fe₃C diagram (B), showing structural evolution in freezing range for selected alloys (A) and variation of freezing and transformation ranges with wt% carbon (C).

redistribution of C between the liquid, δ ferrite and γ phases. The solidification process in multi-component steel weld metals will be much more complex, but consideration of the Fe-C binary system serves to clarify the evolution of the major phases and their sequence of formation.

The Effect of Carbon

Because of the peritectic reaction and the different solubilities of S in δ and γ , the effect of C on solidification cracking is rather complex. Although C is generally regarded as a promoter of solidification cracking, cracking can also occur for low carbon steels ($< 0.10\%C$) in which solidification

occurs completely to δ -ferrite. Furthermore, recent work indicates that the propensity for solidification cracking can increase with decreasing C content from 0.065%C down to 0.04%C [12]. The transformation of δ to γ introduces a contraction strain which adds to the solidification and thermal contraction strains, leading to increased risk of cracking at elevated temperatures. It is uncertain whether the cracking in low carbon steels is solidification cracking due to residual liquid phase segregates or hot cracking of low ductility solid regions. However, Bailey [11] has proposed, without citing evidence, that on approaching zero carbon content the resistance to hot cracking increases. The temperature of transformation of the first-formed δ phase to γ and the temperature sensitivity of the tensile strengths of δ and γ may be critical in determining the cracking susceptibility of low carbon weld metals, particularly for $C < 0.10\%$. For C close to zero wt%, the δ to γ transformation does not occur until the temperature falls to about $1400^{\circ}C$, whereas for a C content approaching 0.10%, transformation starts at a temperature close to $1490^{\circ}C$. This difference and the dependence of strength on carbon level may be relevant to the observed decrease in cracking resistance from 0.065-0.04wt%C and the proposed increase in cracking resistance as C decreases from 0.04% to 0%.

Yurioka et al. [12] and Oshita et al. (13) carried out hot crack testing of welds with a range of carbon contents (0.04-0.33 %C). Contrary to the conventional wisdom, it was found that the risk of solidification cracking in the root pass welds of butt joints is increased in low carbon diluted weld metals (Fig. 4). For cellulosic SMAW root pass butt welds of 16-25 mm thick X60 or X70 plate/pipe steel, Yurioka et al. (12) found no cracking for weld metals with C in the range 0.04-0.20 for welding speeds below 330 mm/min. However, for higher speeds (up to about 550 mm/min), cracking was observed only for welds with C contents at the lower end of the range (0.065-0.04 wt%). Solidification cracking was prevented when the C content of the weld metal exceeded a critical value of about 0.07 wt%, close to the threshold of the peritectic reaction. Although nickel is generally thought to promote solidification cracking by enhancing austenite solidification, it was found to partially counteract the adverse effect of reduced C content on weld metal cracking [12], Figure 4, by displacing the susceptible range to lower carbon levels. Nickel, like manganese, is an austenite stabiliser and they both raise the peritectic temperature and reduce the C content of the peritectic phases [15]. This effect is only minor for additions < about 2 wt%.

GMA welds studied by Yurioka et al. (12) showed similar hot cracking

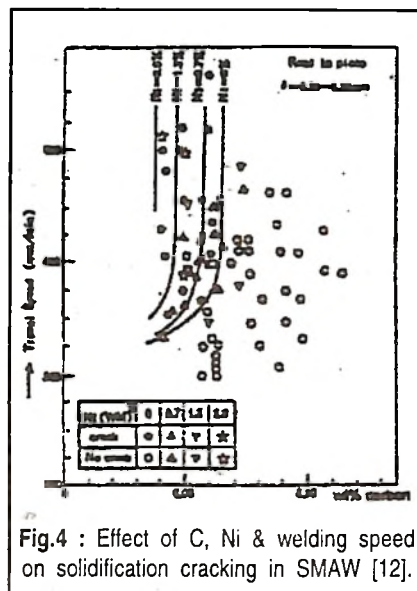


Fig.4 : Effect of C, Ni & welding speed on solidification cracking in SMAW [12].

behaviour to the SMA welds, but for welding speeds above about 500 mm/min.

The Effect of Phosphorus and Sulphur

Other work on the effect of C on hot cracking has been reported by Tamaki et al. [14], based on a circular groove welding test using GMAW. Weld metals containing 0.07-0.95% C were considered, together with the effect of S and P. This work indicated that the critical S and P contents required to produce hot cracking varied with carbon content, showing a sharp decrease close to the peritectic composition and a higher rate of decrease with increasing C for hyper-peritectic than for hypo-peritectic weld metals, Fig. 5. The hot cracking propensity (Fig. 6) was found to be uniformly high for hyper-peritectic compositions (0.2 - 0.6 wt% C) and was correlated with a

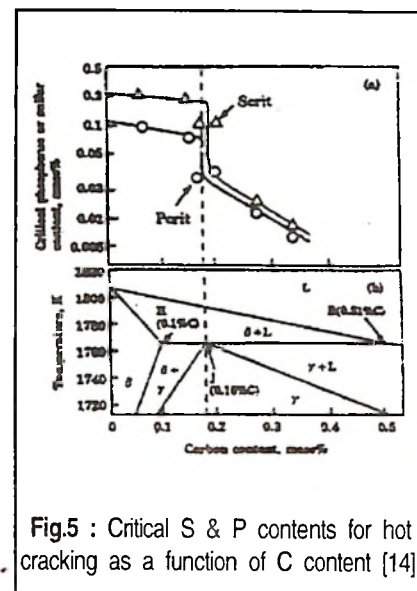


Fig.5 : Critical S & P contents for hot cracking as a function of C content [14].

high concentration of phosphide particles at interdendritic boundaries as a result of the peritectic reaction. Although sulphides were also detected and they also exert a strong deleterious effect, the concentration was relatively insensitive to carbon content and thus the solidification sequence. For weld metals that solidified to γ alone ($C > 0.6\%$) the phosphide concentration increased sharply but then became, like S, relatively insensitive to C content. The authors concluded that the observed increase in hot cracking sensitivity with increasing C above 0.6% is due either directly or indirectly to C, rather than S and P content. For example, hot cracking could be enhanced by the increase in freezing range, thus allowing strong segregation of S and P even at very low concentrations.

Sulphur and phosphorus have a strong influence on the peritectic reaction particularly with respect to

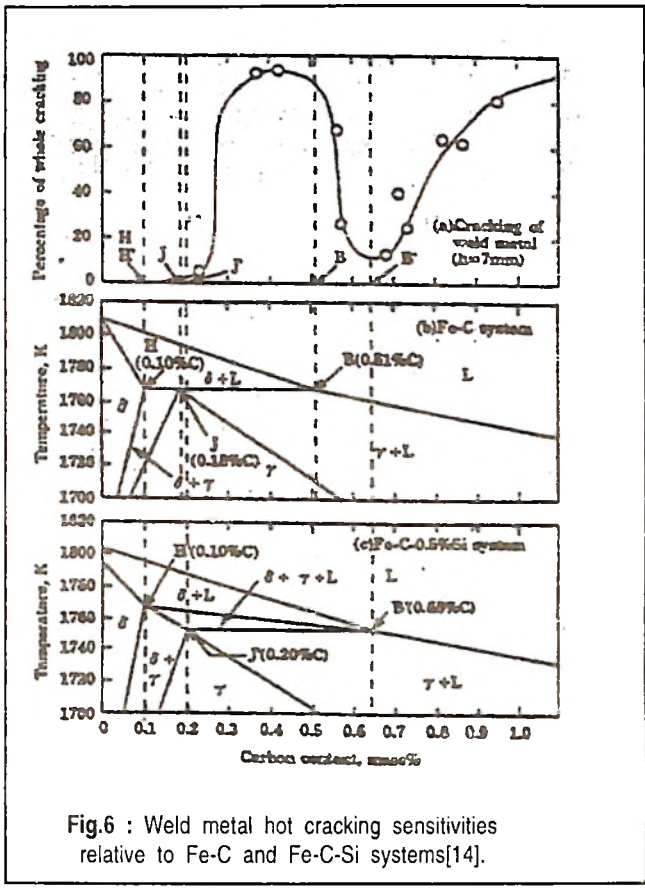


Fig.6 : Weld metal hot cracking sensitivities relative to Fe-C and Fe-C-Si systems[14].

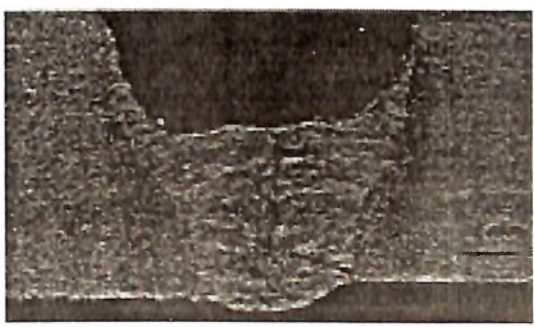


Fig.8 : Diagonal crack path of root pass E9010 weld in WIC test.

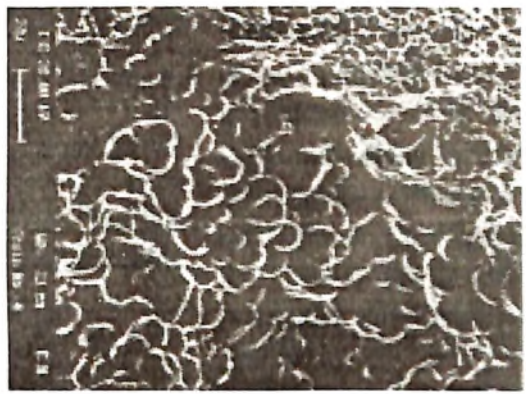


Fig.9 : Cellular-dendritic crack surface in central region of root pass weld in WIC test using SMAW with E9010 electrode.

the temperature range of the reaction. Segregate-rich liquid can persist at temperatures well below the primary solidus temperature [15].

Lancaster [10] has reported that C, S and Mn are the elements which have the strongest effect on solidification cracking of carbon steel weld metal. Fig. 7 shows the influence of Mn:S ratio and of C content on the susceptibility to solidification cracking in carbon steel weld metal. The cracking tendency increases with increasing C content, and above about 0.14% C a sharp rise in the crack/no-crack boundary occurs which may be associated with the peritectic composition (0.18% C),

and the sharp decrease in critical S content for cracking (see Fig. 5). Manganese tends to counteract the influence of S, and the minimum ratio of Mn to S required to avoid cracking increases with increasing C content [10].

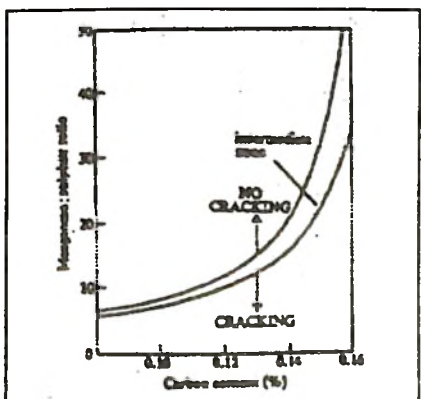


Fig.7 : Effect of C content on Mn:S ratio required to prevent hot cracking [10].

Effect of Other Elements

Yurioka et al. [12] have also investigated the effect of Si on solidification cracking. It was shown that Si has little effect on the critical conditions for solidification cracking. However, Si at a typical level of 0.5% modifies the peritectic reaction to a range of temperatures (about 1490 to 1470°C), with the critical compositions being 0.10%C (δ_p); 0.20%C (γ_p); and 0.65%C (LP).

Elements can also have adverse interactive effects. For example, oxygen interacts unfavourably with sulphur in the following way: FeS freezes at 1190°C and the FeS-FeO

eutectic freezes at 940°C [4]. At low levels of oxygen (up to 0.010%) the sulphide compound has a low surface tension and spreads easily along grain boundaries [11], thus making solidification cracking easy. The presence of Mn is considered to be beneficial since its presence prevents the formation of low melting FeS compounds by promoting a mixed sulphide (Mn, Fe)S, which has a high eutectic temperature, thus reducing the susceptibility to cracking. In addition, MnS has a higher wetting angle than FeS resulting in particles rather than grain boundary films. Further, the freezing temperatures of MnS and MnS-MnO eutectics are high (1600 and 1300°C, respectively) preventing persistence of regions of liquid to relatively low temperatures.

The strong carbide formers Cr, Mo and V reduce the risk of cracking, since these elements promote the formation of ferrite. The above effects of alloying and impurity elements are reflected, at least in part, in various empirical cracking susceptibility formulae that have been proposed for specific welding processes and composition ranges. All of these formulae indicate a strong deleterious effect of C, S and P and a favourable influence of Mn.

EFFECT OF PROCESS PARAMETERS

Welding parameters affect the shape of the weld bead as well as the

evolution of the microstructure of the weld metal by solidification and solid state transformation. Where weld metal compositions are susceptible to cracking, high welding speed and high weld bead depth/width (D/W) ratio should be avoided. The (D/W) ratio is usually controlled by keeping the current low. Deep penetration welds of higher carbon steels also enhance weld bead C enrichment because of high dilution, thus increasing the risk of cracking.

Weld Bead Shape

The weld bead shape has a significant effect on solidification cracking. If the weld pool is deep and narrow, ie. high D/W ratio, feeding of liquid metal during solidification becomes more difficult, and cracking is more likely than if the D/W ratio is near the optimum. The solidification pattern of a weld with a high D/W ratio tends to be inwards from the sides and this increases the severity of segregation to form liquid with a low freezing temperature in the centre-line. A shallower pool usually allows upward solidification, thus reducing the severity of centre-line segregation and allowing solidification contraction to be taken up by sinking of the weld surface.

Heat Input

One of the main factors which influences solidification cracking is the coarseness of solidification microstructure. The size of the epitaxial columnar grains of weld

metal depends on the austenite grain size of the HAZ at the fusion boundary. High energy welding processes such as electroslag and submerged arc welding usually give the greatest grain growth in the HAZ and hence the coarsest microstructure in the weld metal. Generally, the coarser the cellular-dendritic structure, the higher the segregation tendency and the higher the susceptibility to solidification cracking.

Effect of Stress

Weld residual stresses result from cooling and subsequent contraction of the weld metal. Under the influence of stress any weak plane in the weld becomes a potential site for cracking. However, stresses in the high temperature regime only result from the solid state transformation of δ -ferrite to austenite and from solid state thermal contraction under restraint. If residual liquid regions are isolated and cannot be fed from a liquid reservoir then shrinkage voids are likely as freezing occurs and/or 'rupture' can be induced by tensile stresses resulting from restrained contraction of surrounding solid material.

Effect of Welding Speed

Speed of welding has a significant influence on the solidification pattern of weld metal and hence on solidification cracking. Speed influences the feeding of liquid metal

to compensate for the contraction on solidification and any deficiency in liquid metal feeding is likely to lead to shrinkage or solidification cracking. At lower speed the molten pool shows little elongation and the crystals grow perpendicular to the fusion boundaries. Near the weld centre-line the solidifying grains curve progressively in the direction of the moving molten pool. As the speed of welding increases the weld pool becomes tear drop shaped and the crystals tend to grow perpendicular to the fusion boundaries, meeting in the central plane of the weld bead. The probability of segregation to form a low freezing temperature liquid along the centre-line is increased. Yurioka et al. [12] have shown that solidification cracking does not occur when the travel speed is less than a critical level, see Figure 4.

HOT CRACKING TESTS

Weld solidification cracking has been studied extensively over the past 40 years using a variety of weldability tests. These tests can be classified on the basis of two concepts: (1) using the selfrestraint of a laboratory scale specimen to induce cracking, and (2) simulating the restraint force on a small specimen to approximate a production situation.

The circular groove test used by Tamaki et al. [14] and the butt weld test procedure described by Yurioka et al. (12) are examples of self-restraint tests, whereas the

Transvarestraint test (16-18) is based on the application of an external force. Longitudinal solidification cracking can be produced when a predetermined strain is applied across a solidifying weld pool.

Many of the tests developed are only capable of assessing the relative cracking tendencies of different weld metals and they may not be good predictors of susceptibility in an actual production weld. Yurioka [19] holds the view that hot crack testing should be conducted with the same groove and under the same restraint conditions present in actual welding practice in order to properly assess the susceptibility to hot cracking.

HOT AND COLD CRACKING OF CELLULOSIC ROOT PASS WELDS IN X80

An extensive investigation of cracking in high strength cellulosic root pass pipeline girth welds has been conducted by the authors over recent years [20]. The focus of this work was initially, and predominantly, hydrogen assisted cold cracking (HACC). A WIC testing procedure was developed for 8.6 mm X80 strip to determine the effect of welding conditions, specifically the type of E9010 consumable, pre-heat and restraint stress, on propensity for HACC. Although cracking could be avoided by modest preheat and/or reducing the restraint force, cold cracks were pervasive for all consumable types for room

temperature welding. The crack path extended from one of the characteristic wagon tracks on the weld face directly through to the root or, just as frequently, followed a macroscopically diagonal path through the weld centre-line (Fig. 8). On a fine scale the diagonal cracks were stepped with intergranular segments along ferrite veins and transgranular segments across the grains. Taking the long path in this way at a large angle to the direction of cross-weld tensile restraint force was a puzzling observation, until cracks were found that consisted of isolated segments of centre-line cracking. Careful examination of the surfaces of these cracks by fracturing open the weld revealed segments of cellular-dendritic structure (Fig. 9) which clearly show that solidification cracking had occurred along the centre-line at high temperatures. These hot cracks provide effective stress concentrations for hydrogen assisted cold cracking at lower temperatures, extending the cracked surface and directing stepwise crack growth from the wagon track towards the upper tip of the hot crack and between the weld root and lower tip. This built-in, internal stress concentrator facilitates subsequent cold cracking when the welding conditions do not allow effusion of hydrogen to below the critical level for HACC.

Even if extension of the hot cracking by cold cracking does not occur, the potential remains for a buried hot

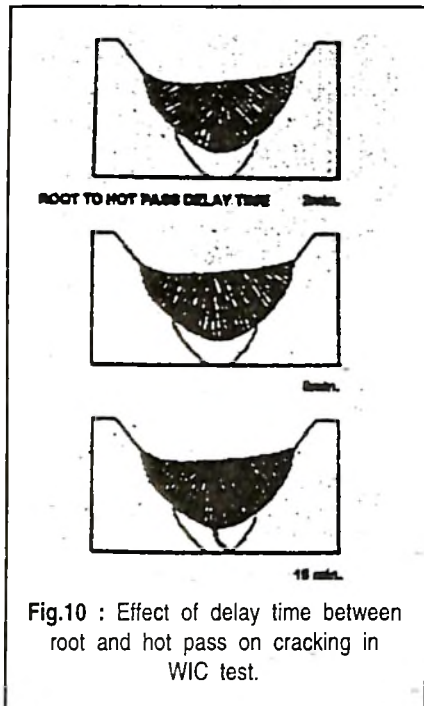


Fig.10 : Effect of delay time between root and hot pass on cracking in WIC test.

crack in the root pass weld bead. Perhaps fortuitously, the welding procedures that have been adopted by the Australian pipeline welding industry for SMAW with cellulosic electrodes involve deposition of the hot pass before the temperature falls to a level at which HACC becomes a threat, as well as remelting of a significant part of the root pass weld, including regions likely to contain a hot crack. Fig. 10 shows schematically the results of hot pass welds made after increasing interpass time intervals and illustrates the danger of a significant delay before deposition of the hot pass.

The observation of hot cracking in cellulosic root pass weld beads is not novel. Yurioka et al. [12] showed a macrograph of centre-line cracking in SMA welds of 16-25 mm thick X60

and X70 plate (Figure 6 in ref. 12). However, the restraint stress is likely to have been much higher in this case than in the WIC testing program because of restraint welds at either end of the weld groove, much thicker plate and smaller bead to plate thickness ratio. Therefore the probability of solidification cracking would be expected to be much higher. Yurioka et al. also showed the effect of the hot pass in removing the root pass solidification crack (Fig. 8 in ref. 12).

SIMULATION OF SOLIDIFICATION CRACKING IN CELLULOSIC WELD METALS

An investigation of the relative susceptibilities of commercially available E9010 cellulosic electrodes to hot cracking was undertaken by the authors using a Gleeble thermomechanical simulator. Welds were deposited on X80 strip and 5.8 mm diameter rods were extracted from the strip with the diluted weld bead being located in the middle of the rod. Using a slotted quartz cylinder as a crucible, the central zone was remelted by resistance heating and then cooled under conditions simulating a typical root pass weld thermal cycle. The restraint force developed on cooling was monitored and samples were sectioned for evidence of solidification cracking. Limitations to this technique are: (i) a volume greater than the original weld bead was remelted and therefore more

extensive dilution with X80 occurred; (ii) the shape of the molten slug and the heat transfer paths do not allow efficient feeding of the solidifying weld and the weld bead is therefore prone to shrinkage cavities; (iii) the remelted weld bead will no longer be hydrogen saturated, reducing the effect, if any, of hydrogen on solidification cracking; and (iv) control of the uniformity of the temperature during melting and freezing is difficult.

Notwithstanding these limitations, the simulation gave good insight into the interaction of transformational strains and thermal contraction in the evolution of the restraint force, as well as limited data on the relative extents of solidification cracking in the three consumables examined. The cooling rate (equivalent Δt_{8-5} of 7s) was similar to that expected for root pass welding of 8.6mm strip with no preheat. Two types of cavities were observed: intergranular shrinkage cavities and large centrally located gas pores. The former type of cavity was characteristic of Lincoln E9010 and E6010 electrodes and the latter of Bohler and Hobart E9010 electrodes. In both cases, liquid-solid shrinkage is compensated for by cavity formation, but the source of the gas porosity is uncertain. Pre-annealing in an attempt to de-gas the weld bead did not prevent the formation of gas pores on subsequent melting, and possible sources of the gas are deeply trapped hydrogen or gaseous

decomposition products of compounds or inclusions peculiar to the Bohler and Hobart weld metals. Although these experiments demonstrated the proneness of the weld metals to cavity formation, the work does not have strong relevance to real welds for which the preparation is designed to allow adequate liquid feeding, thus eliminating any cavities due to liquid-solid shrinkage.

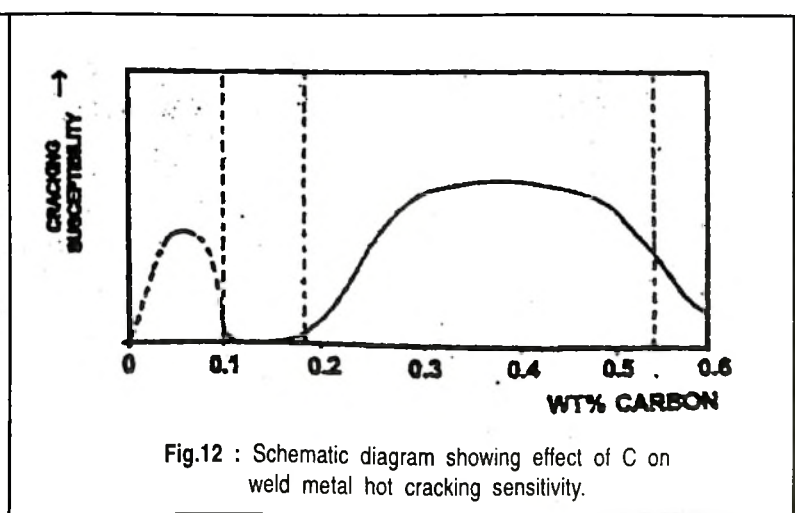
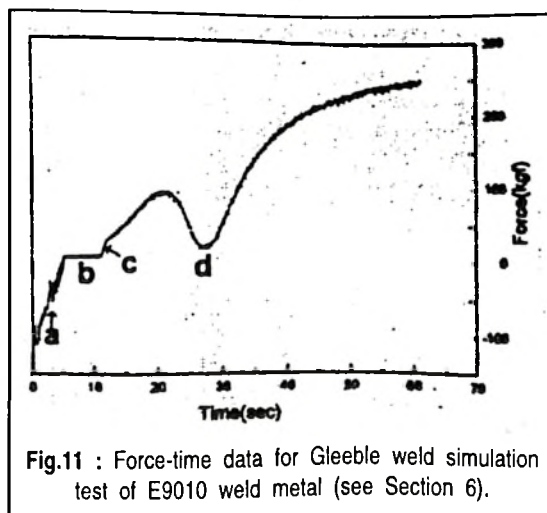
Fig. 11 shows force-time data for a cracking simulation test on X80 diluted E9010 weld metal. An initial compressive force was manually reduced during heating to the melting point. The blip at a corresponds to the effect of the volume contraction associated with α to γ transformation and the horizontal section at b shows the zero force condition when the central zone is melted. The machine was then switched to stroke control so that the specimen was rigidly restrained and the restraint force could be monitored. A rapid onset of

tensile restraint force occurs (c) when solid bridging occurs on cooling (at the coherence temperature T_c) and this force is amplified by transformation of δ to γ which produces a volumetric contraction of 0.82%. The force subsequently rises at a lower rate consistent with the thermal contraction of γ phase before the onset of γ to α transformation at d. The volume expansion causes almost complete stress relaxation before thermal contraction of the α causes a significant rise in tensile restraint force.

If the restraint stress developed during solidification exceeds the inherent strength of the weld metal at any temperature over the freezing range then solidification cracking will occur. The combination of the high tensile restraint stress developed on cooling to low temperatures with stress concentrations due to embedded hot cracks would then constitute a dangerous condition.

EFFECT OF CARBON ON WELD METAL CRACKING

The adverse effect of P and S on weld metal solidification cracking is well known. The effect of carbon per se and in concert with P, S and other elements is less clear. Work by Tamaki et al. [14], discussed in Section 3.3, indicated that cracking propensity varies in a non-monotonic way with increasing C content. They showed that the critical P and S contents required to induce hot cracking increased sharply for C contents in excess of the peritectic composition and that the incidence of hot cracking also rose sharply to a maximum level for steels with C in the hyperperitectic range (0.20-0.65 %). They associated this effect with a sudden increase in the amount of phosphide at the interdendritic boundaries. For C in excess of about 0.6%, only austenite solidification occurs and cracking susceptibility increases continuously without any apparent increase in phosphide or sulphide formation.



If the results of Yurioka et al. [12] are combined with those of Tamaki et al. [14], cracking susceptibility increases again in the range 0.065-0.04 wt% C, even though no peritectic reaction is expected and tolerance to S and P should increase. These trends are presented schematically in Fig. 12, which shows that the hypo-peritectic range of C contents provides the highest resistance to weld metal solidification cracking. This result can be compared with data from continuous casting of low carbon steel plate for which C compositions in the ranges 0.10-0.12% (close to δ_p) and 0.18-0.20 (close to γ_p) are considered to be prone to solidification cracking and are therefore avoided [21]. Although the latter range is close to the border of the crack sensitive range in Figures 6 and 12, the other is within the "safe" range.

The results of a hot tear test on cast steel pins by Masumoto and Ozaki [22] showed a marked increase in hot cracking for steels with more than 0.1% C. Although this work was designed to assess weld metal cracking, the pins were cast, not produced by welding, and the oxygen contents were relatively low (about 100ppm). The results are therefore more likely to be relevant to continuous casting than welding which differ in some significant ways. One is the much higher oxygen contents of weld metal and potentially unfavourable interaction

with S (see Section 3.4). Another is the relative absence of restraint on the solidifying cast strand. In addition, the balances of alloying elements and impurities will generally differ for the same nominal C content, shifting the peritectic composition points and temperature ranges.

CONCLUSIONS

1. Fast welding promotes cracking because of a lower bead to thickness ratio and a less favourable orientation of the columnar grains associated with the tear-drop shaped weld bead.
2. Sulphur and phosphorus segregate strongly to interdendritic boundary regions, increasing the risk of solidification cracking. Welding conditions (such as high welding speeds) which promote centre-line segregation of S and P are very deleterious.
3. Carbon has a complex effect which is related to the peritectic reaction. This reaction and subsequent phase transformation affect the temperature rate of development of restraint force imposed on the weld bead. Welding conditions which determine the bead shape and size also influence the restraint force.
4. The effects of carbon, sulphur and phosphorus are strongly

interactive because of the polymorphic transformation of δ to γ .

5. Ultimately, the rate of increase in the restraint force relative to the rate of increase in weld metal strength with falling temperature is critical in determining whether or not solidification cracking will occur.
6. The optimum carbon compositional ranges for resistance to solidification cracking are controversial and differ between welding and casting of steels with similar but not equivalent compositions. However, weld metals free of solidification cracking should be possible for weld metals with very low C contents up to about 0.2% C provided the S and P contents are low, the weld design is appropriate and the welding conditions are controlled to prevent adverse bead shapes and sizes and unfavourable patterns of segregation.
7. If even minor hot cracking does occur then it provides a significant stress concentrating defect for the promotion of subsequent cold cracking.

ACKNOWLEDGEMENTS

The review presented and the experimental results discussed were conducted under the auspices of the

CRC for Welded Structures and the Australian pipeline industry . Their support is acknowledged with gratitude.

REFERENCES

1. Yurioka N and Suzuki J 1990. *Int. Mat. Reviews* 35 217-249.
2. Davidson J L, Lynch S P and Majumdar A 1996. Proc. of DSTO-WTIA Joint Seminar on Hydrogen Management of Steel Weldments," Melbourne ed. by L. Davidson and D. Olsen 1-20.
3. Dunne D P 1999. Proc. of 1st Int. Conf. on Weld Metal Hydrogen in Pipeline Girth Welds, Wollongong, ed. by N. Yurioka, 3-1.
4. Easterling K 1985. *Introduction to the Physical Metallurgy of Welding*, Butterworths.
5. Singer A R E and Jennings P H 1947. *J.I.M* 73 273.
6. Bishop H F et al 1960. *Transaction AFS* 68 518.
7. Borland J C 1960. *British Welding Journal* 7 508.
8. Baker R G 1975. *Weldability and its Implications for Materials Requirements*, Rosen Centenary Conference Proceedings, Royal Society 129.
9. Ueda Y and Yamakawa T 1971. *International Symposium on Cracking and Fracture in Welds*, Japanese Welding Society, Tokyo, IC5.I.
10. Lancaster J F 1987. *Metallurgy of Welding*, Allen and Unwin Ltd, London 4 82-84.
11. Bailey N 1995. *Weldability of Ferritic Steels*, ASM International, Abington Publishing.
12. Yurioka N, Ohshita S, Saito S and Sakurai H 1984. *Welding in Energy-Related Projects*, Welding Institute of Canada, Toronto, Pergamon Press, 53-62.
13. Ohshita O, Yurioka N, Mori N, and Kimura T 1983. *Welding Research Supplement*, 62 129s.
14. Tamaki K, Kawakami H, and Suzuki J 2000. *Effect of Carbon Content and Peritectic Reaction on Hot Cracking of Weld Metal of High Carbon Steel*, IIW-Doc. IX-1967-00.
15. Kanaga A and Okamoto T 1986. *Mat. Sci and Tech*, 2 997-1008.
16. Farrar J C M and Garland J G 1973. *Welding Institute Research Report*, M/77/73, Cambridge, UK.
17. Garland J G and Bailey N 1975. *Welding Research International*, 5 (3) 1.
18. Garland J G and Bailey N 1976. *Welding Institute Research Report*, M/28/76, Cambridge, UK.
19. Yurioka N. Private communication.
20. Alam N, Dunne D P and Barbaro F 1999. Proc. of 1st Int. Conf. on Weld Metal Hydrogen in Pipeline Girth Welds, Wollongong, ed. by N. Yurioka, 9-1.
21. Williams J. Private communication.
22. Masumoto I and Ozaki H 1973. *J. Jap. Weld. Soc.*, 45 674-684.