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Heat-Affected Zone Cracking of Al-Mg-Si Alloys

Following a series of patch tests, Al-Mg alloys are found to be preferable to Al-Mg-Si alloys for welded fabrications, and maximum resistance to HAZ cracking is obtained using Al-Si filler metal

BY N. F. GITTOS AND M. H. SCOTT

SYNOPSIS. The heat-affected zone (HAZ) liquation cracking of the Al-Mg-Si alloy H30 (6082) has been investigated using a patch welding test. Cracking was produced around the outside of the patch when using Al-Mg filler metal but not with Al-Si filler metal.

After metallographic examination of the HAZ and the cracks, it was concluded that the observed behavior could be explained in the following way:

1. Grain boundary liquation occurred in the HAZ during welding.
2. With certain weld metal compositions, the weld metal solidus was higher than that of the base metal.
3. In these cases, the strains due to solidification of the weld metal were therefore imposed on the HAZ, while there were still liquated boundaries present, which then cracked.

Recommendations for avoiding or minimizing HAZ cracking are given.

Introduction

There are two categories of Al alloy used in structural and plant applications—non-heat-treatable and heat-treatable. The former is mainly alloyed with Mg; one of the strongest, containing 4.5% Mg and 0.75% Mn, is designated N8 in British standards,

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5083 in American, French and Japanese and AlMg4.5Mn in German standards. In the case of annealed N8 sheet and plate, the minimum 0.2% proof stress is 125 N/mm² and the minimum tensile strength is 275 N/mm²; these values also apply to welded N8.*

The commonest welded heat-treatable alloys are those based on Mg and Si. Such alloys are exemplified by the British H30 (6082), the American, French and Japanese 6061 and the German AlMgSil and contain Mg and Si in amounts up to 1% nominal, often with additions of Mn and/or Cu. Taking H30 plate in the fully heat-treated condition (i.e., solution-treated and artificially aged), the minimum 0.2% proof stress is 240 N/mm² and the minimum tensile strength 295N/mm², the former in particular being much higher than for N8.

**Editor's Note: N/mm² is equivalent to MPa and may be converted to ksi by dividing N/mm² by 6.895.*

The extensive use of these alloys is to some extent surprising for two reasons since the weldability is only mediocre:

1. The composition is such that the alloys are very susceptible to solidification cracking; to counter this, a filler metal of much higher alloy content must normally be used, the choice resting between Al-Mg or Al-Si alloys. (The exception is where thin sections are to be welded and the work is very carefully jigged, frequently to put the weld in compression, when autogenous welds are possible.)
2. The strength of the as-welded joint is well below that of the fully heat-treated base metal.

In the HAZ, this is because the major effect of the weld thermal cycle is to coarsen the hardening Mg_2Si precipitate, overaging the alloy. The composition of the weld metal is basically binary, with possibly only a little of the third element (from dilution by the base metal) to form Mg_2Si. In any case, this will not form in a hardening mode under the weld thermal cycle. As a consequence, the tensile properties of the weld are usually well below not only those of the base metal, but also those of N8.

In general terms, the use of an Al-Mg filler metal gives somewhat higher

Table 1—Composition of Two Sheets Used for the Test and the Specification of HS30, Wt-%

	Al	Mg	Si	Mn	Fe	Cr	Ti	Cu	Zn
BS 1470:1972	Remainder	0.5	0.7	0.4	0.5	0.25	0.2	0.10	0.2
HS30		−1.2	−1.3	−1.0	Max	Max	Max	Max	Max
Sheet 1	Remainder	0.63	0.91	0.53	0.32	≃0.008	0.03	0.04	0.02
Sheet 2	Remainder	0.72	0.94	0.51	0.34	< 0.005	0.04	0.03	0.02

Table 2—Analyses of the Filler Metal Wires Used in the Tests, Wt-%

	Al	Mg	Si	Mn	Fe	Cr	Ti	Cu	Zn	Si + Fe
NG21 wire	Remainder	0.13	5.00	0.02	0.37	–	–	0.02	0.01	–
BS 2901:Part 4:1970 NG21	Remainder	0.2 max	4.5-6.0	0.5 max	0.6 max	–	–	0.10 max	0.2 max	–
NG61 wire	Remainder	5.46	0.09	0.64	0.23	0.14	0.10	0.07	0.1	0.32
BS 2901:Part 4:1970 NG61	Remainder	5.0-5.5	–	0.6-1.0	–	0.05-0.20	0.05-0.20	0.10 max	0.2 max	0.40 max

strength and ductility than does an Al-Si filler metal. However, the latter has better resistance to solidification cracking for the same dilution.

To illustrate the extent of the loss in properties, ASME VIII Division 1[1] allows a design stress for unwelded 6061 of 72N/mm², but for welded 6061 this is reduced to 41N/mm² (based on a welded tensile strength of 165N/mm²) whereas 5083 is allowed 69N/mm². An even more marked reduction is given in the British code of practice CP118,[2] where the permissible stress is based primarily on the proof, not the tensile. For unwelded H30 plate, the permissible axial tensile stress is 132N/mm²; for welded H30 plate, 51N/mm². (For comparison, for N8 plate the values are 80 and 82N/mm², respectively.) An improvement in strength can be obtained by full postweld heat treatment. Since this involves rapid quenching, however, it is only possible on relatively small items of simple shape, e.g., beer barrels and LPG bottles.

Problem of HAZ Cracking

Despite the shortcomings, Al-Mg-Si alloys are frequently welded. One reason for this is that these alloys are much easier to form and to produce as sections than the high Mg content Al-Mg alloys. Therefore, it is often easier to obtain Al-Mg-Si alloys, particularly if complex shapes are wanted, than it is Al-Mg alloys. Provided that the low strength of the welds is allowed for and the welding preparations allow a sufficient amount of filler metal to be used, this may seem sufficient reason for using these alloys. Yet there is another potential problem and one that has received remarkably little attention in the literature—that of HAZ cracking.

The Welding Institute first became aware that such cracking could be a serious problem in 1971, and several examples have been seen since. In each case, this was during investigations carried out for member companies of the Institute and, therefore, it is not possible to give full details; indeed, in some cases it was not possible to obtain data on factors such as the welding conditions. From the various investigations, however, the following features emerged:

1. Cracking was in the HAZ at or near the fusion boundary, but more usually the latter and normally parallel to it; it was intergranular.
2. It could break the surface or be entirely sub-surface. In the former case, it often initiated at the weld toe, but propagated away from the fusion line, even though following the same general direction.
3. The crack length could vary from a fraction of a millimeter up to several millimeters in depth.
4. Cracks were observed in plate, sheet, tube and extruded sections, in thicknesses from 1 to 13 mm (0.04 to 0.5 in).
5. The severity of cracking varied, but in most cases it was not found regardless of the postweld inspection method that was applied. Where the cracking was minor and sub-surface, it was only found by sectioning, either while examining some other defect or specifically to see if cracking had occurred. In one case of surface cracking, it was only observed after the components had been painted, some of the cracks being too wide for the paint to bridge. The most extreme case was of pipe/flange joints on a chemical plant, some of which were found to leak when the plant was filled with water prior to a pressure test. Subsequent examination showed that some 20% of the joints had surface-breaking cracks and sectioning showed subsurface cracks in others. In another instance, the closing bungs on small pressure vessels blew out during pressure testing because of HAZ cracks.
6. Cracking occurred with both GTA and GMA welds, butt and fillet joints and with and without preheat.
7. Cracking was usually associated with the use of Al-Mg filler metal rather than Al-Si filler metal.
8. In the specific case of butt welding a patch into a larger sheet, cracking was always on the outer edge, not the inner.
9. The morphology of the cracking suggested a liquation mechanism, and this was subsequently confirmed by scanning electron microscopy.
10. All the known incidences of cracking were in the H30 and 6061 alloys.

The first reference to HAZ cracking of an Al-Mg-Si alloy was in 1966,[3] wherein the occurrence of small surface cracks was described as a chronic problem. However, the cracks were markedly smaller than those observed by the Institute, being no more than 0.4 mm (0.016 in.) deep and occurring on the surface where local melting took place, due to the high frequency superimposed on the ac GTA arc.

The cracks were solidification rather than liquation cracks. Although small, they were found to have a significantly deleterious effect on the tensile strength and fatigue behavior. Such cracks have also been observed by The Welding Institute but were not correlated with the presence of larger cracks. (They are also observed where there have been arc strikes on the surface. Therefore, every effort should be made to avoid arc strikes).

HAZ cracking more like that observed by the Institute is mentioned briefly in Kaiser Aluminium's book on welding Al alloys,[4] but major cracking is stated to be rare and implicitly due to faulty welding conditions.

Metzger[5] reported HAZ cracks near the fusion line in GTA welds made

with Al-Mg filler metals, but not with Al-Si filler metals. These cracks reduced the tensile strength of specimens tested transverse to the weld. The optical illustration of such a crack was consistent with a liquation mechanism. Other workers[6] have shown with a weld thermal cycle simulator that liquation cracking can occur at grain boundaries, but considerable strain was needed.

From the Institute's own observations and the limited information in the literature, it seemed reasonable to conclude that the cracking was caused by liquation of some phase in the alloy which remained liquid while the weld metal was solidifying and opened up under the strain imposed. This mechanism would, for instance, explain why the incidence of cracking seemed higher with Al-Mg filler metals than with Al-Si, the former giving a somewhat higher melting range to the weld metal.

In view of the increasing number of cases and the seriousness of these, it was considered that the problem warranted some systematic investigation, both to try to study more accurately the cause and to be able to recommend the remedy.

Method of Investigation

From the examples of HAZ cracking studied, it was apparent that the choice of filler metal had an effect and it seemed possible that not all batches of Al-Mg-Si alloy were susceptible. Therefore, tests to determine the susceptibility to HAZ cracking were made with Al-Mg and Al-Si filler metals on more than one batch of Al-Mg-Si alloy.

It was considered essential to use a welding test rather than a thermal simulation test and one that imposed a high degree of restraint. Patch tests were, therefore, made by mechanized GTAW to ensure reproducibility.

Base Metal

For the tests BS 1470*: 1972, HS30 was chosen as being typical of the Al-Mg-Si alloys. The nominal composition is Al-1%Si-0.8%Mg-0.7%Mn and the compositional requirements of the specification are given in Table 1. Since cracking had been observed in various product forms, 3 mm (0.12 in.) thick sheet was selected as being experimentally convenient and two such sheets were obtained, care being taken to ensure that they came from different batches; analyses of these sheets are given in Table 1.

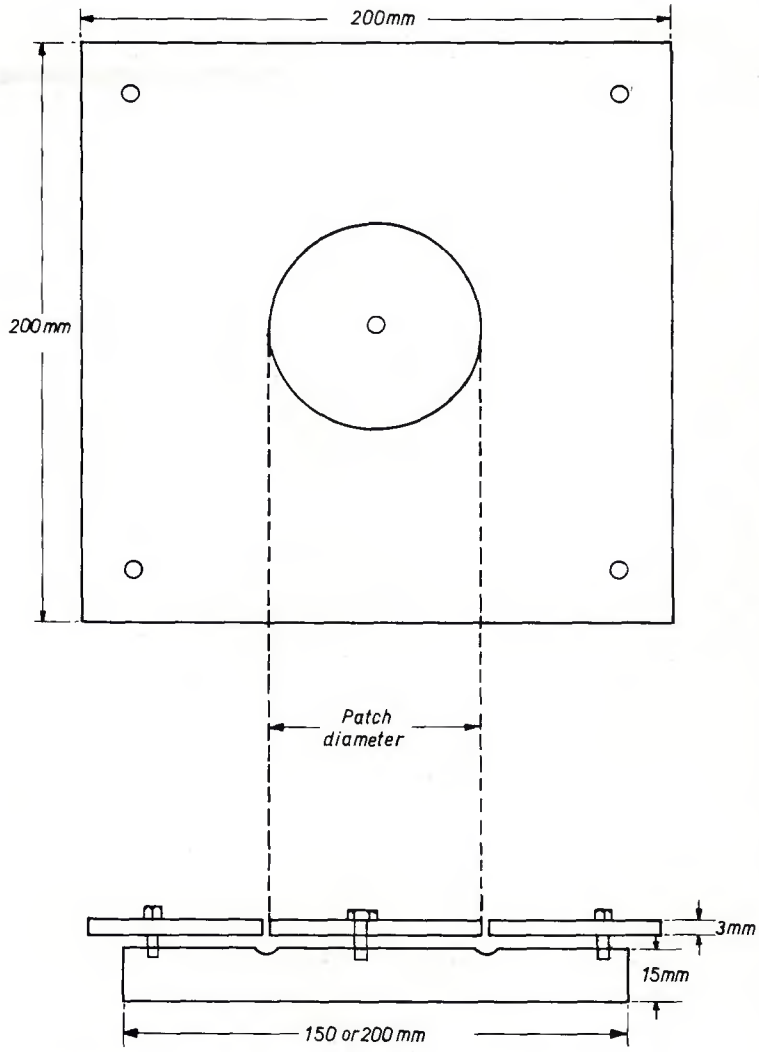

Fig. 1—Plan and sectional arrangement of the patch tests

Filler Metal

As mentioned in the Introduction, a filler metal of higher alloy content must usually be used when welding Al-Mg-Si alloys, the choice resting between Al-Mg or Al-Si, both nominally containing 5% of the alloying addition. (Very occasionally, Al-10%Si is used, but this would not be expected to behave very differently from Al-5%Si because their solidus temperatures are identical.)

The filler metals chosen were BS 2901.† Part 4: 1970, NG21 (Al-5%Si) and NG61 (Al-5.2%Mg-0.8%Mn), both as 1.6 mm (0.06 in.) diameter wire. The requirements of BS 2901 and analyses of the wires are given in Table 2.

Welding Test

The patch test was selected for this investigation. This test has frequently been used for studying weld metal cracking, because it imposes a relatively high restraint in the weld zone transverse to the weld. Thus, in this sense, it seemed well suited to the present requirement. Moreover, during one of the service investigations it had been noted that HAZ cracking had occurred on what was effectively a small patch, but had not occurred when making a larger diameter weld. Therefore, there was some reason to suppose that changing the diameter of the patch would alter the restraint and possibly the degree of cracking.

The experimental arrangement is shown in Fig. 1. It consisted of a 200 mm (7⅞ in.) square of HS30 sheet with a central hole being bolted to a mild steel backing plate, with a disc filling the hole being bolted to the backing plate, the diameter of the hole and disc ranging from 50 to 150 mm (2 to 5.9 in.). The machining tolerances

British Standard 1470: 1972—Wrought Aluminium for general engineering purposes-plate, sheet and strip.

†*British Standard 2901: Part 4:1970 Filler rods and wires for gas shielded arc welding—aluminium and aluminium alloys and magnesium alloys.*

Table 3—Welding Conditions for All Welds[a]

Process	AC mechanized GTA
Shielding gas: Type	Argon
Flow rate	12 liters/min
Nozzle diameter	16 mm (0.63 in.)
Tungsten electrode diameter	4 mm (0.16 in.)
Arc voltage	20 V
Current	250 A
Electrode–work gap	2.4 mm (0.094 in.)
Wire filler metal diameter	1.6 mm (0.063 in.)
Wire filler metal feed speed: Welds 1-10	920 mm/min (36 ipm)
Weld 11	480 mm/min (19 ipm)

[a]The welding speed was varied and is given in Table 4.

were +0.25 − 0.00 and +0.00 − 0.25 mm for the hole and disc, respectively. Backing grooves of appropriate diameter were machined in the backing plate.

Welding Conditions

The AC GTA process was used and, in order to maintain reproducibility, a mechanized system was adopted. The test pieces were degreased, wire brushed and degreased, then bolted on to the backing plate. The latter was mounted on a turntable, so that the work was rotated under a stationary torch. It was then possible to vary the linear arc energy of the welds (in the range 0.7–1.2 kJ/mm) by altering the rotational speed of the turntable.

The welding conditions are given in Table 3.

Metallography

After removal from the backing plate, the welded test piece was examined visually and, if necessary, under a binocular microscope. This was followed by metallographic sectioning and examination of the crack face in a scanning electron microscope fitted with an energy dispersive analyzer. Thin foils from the base material and HAZ were examined by transmission electron microscopy; in addition, some microanalysis was performed using a Philips 400 scanning transmission electron microscope (STEM) fitted with an X-ray energy dispersive analyser.

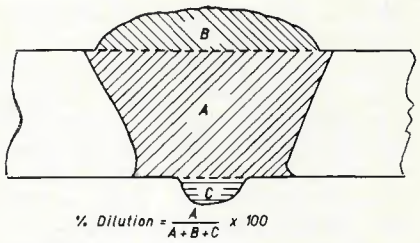

Fig. 2—Schematic weld cross section showing the method for estimating % dilution of the weld by the base metal

The transverse metallographic sections of the patch tests were macroetched in Keller's reagent so that the position of the fusion boundary was clearly visible when observed on a projection microscope at a magnification of X7.

Tracings of the weld beads were taken and the areas and dilutions obtained graphically as indicated in Fig. 2. The calculation assumed no joint gap. In the worst case the gap could have been 0.5 mm (0.02 in.). This would have led to an over-estimation of the dilution by less than 5%.

Results

The results of the patch tests are given in Table 4 and the dilution measurements in Table 5. Table 4 shows that every weld made with NG61 showed HAZ cracking, irrespective of patch size of linear arc energy, while no weld made with NG21 cracked at all. As with the previous service investigations, the cracking was at the outer edge of the weld, not the inner, and in most cases extended through 360 deg. Figure 3 illustrates the marked difference in behavior between the two filler metals for 100 mm (3.94 in.) patches; with the NG61 filler metal, the patch could be pulled out with no difficulty.

The dilutions on all welds were high, but in the range that is expected for close square butt GTA welds. Based on the dilution measurements, the weld metal Mg and Si contents were calculated and are also given in Table 5. The

Table 4—Details of Patch Tests[a]

Weld no.	Sheet no.	Ring diameter, mm	Filler metal	Travel speed, mm/min	Linear arc energy, kJ/mm	Comments
1	1	100	NG21	288	1.0	–
2	1	75	NG21	288	1.0	–
3	1	50	NG21	288	1.0	–
4	1	50	NG61	257	1.2	HAZ cracking
5	1	75	NG61	257	1.2	HAZ cracking
6	1	100	NG61	257	1.2	HAZ cracking
7	2	100	NG61	290	1.0	HAZ cracking
8	2	100	NG21	290	1.0	
9	2	150	NG61	257	1.2	HAZ cracking
10	2	150	NG61	404	0.7	HAZ cracking
11	2	150	NG61	257	1.2	Weld metal cracking, fine HAZ cracks

[a]mm ÷ 25.4 = in.

Table 5—Dilution Measurements on Welds 6-11 and Computed Weld Metal Compositions

Weld no.	Filler metal	Dilution of weld metal by base metal, %	Calculated weld metal composition, wt-% Mg	Calculated weld metal composition, wt-% Si
6	NG61	77	1.73	0.72
7	NG61	77	1.81	0.74
8	NG21	78	0.59	1.83
9	NG61	77	1.81	0.74
10	NG61	82	1.57	0.79
11	NG61	88	1.29	0.84

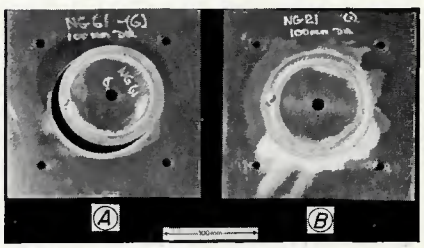

Fig. 3—Results of welding 100 mm (3.9 in.) diameter patches using indicated filler metals: A—NG61; B—NG21

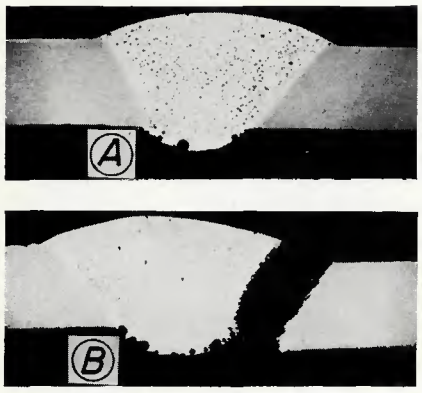

Fig. 4—Transverse sections of welds 3 and 4 welded with indicated filler metals: A—NG21; B—NG61 etched in Keller's reagent. ×6 (reduced 58% on reproduction)

levels are borderline for good resistance to weld metal cracking[7] but this only occurred on one specimen.

Figure 4 shows transverse sections of the two 50 mm (1.97 in.) diameter patch test welds 3 and 4 made using NG21 and NG61 filler metals. Weld 3 is typical of the welds made with NG21 filler metal, showing no cracking; weld 4 illustrates the typical HAZ cracking associated with the welds made using NG61 filler.

Figure 5 shows sections of weld 11 in which there was a marked reduction in HAZ cracking. This was achieved by using the welding conditions which gave rise to extensive HAZ cracking in weld 9, except that the filler metal wire feed speed was reduced. This produced the slightly concave weld profile shown in Fig 5A. Although no cracking occurred which broke the surface, fine sub-surface HAZ cracks were observed on another section, Fig. 5B. Reference to Table 5 shows that the dilution of this weld was greater than those of the previous tests. The resultant weld metal composition and also possibly weld shape produced limited centerline solidification cracking.

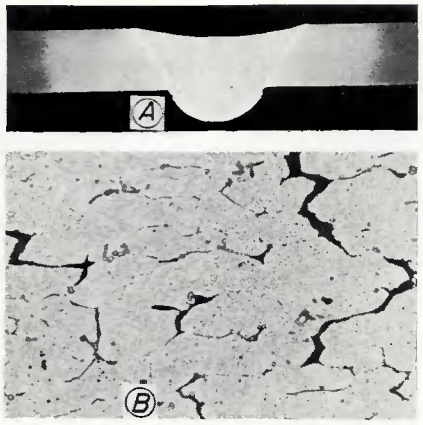

Fig. 5—Transverse sections of weld sections of weld 11: A—uncracked, ×12; B—subsurface HAZ cracks etched in Keller's reagent, ×500 (A and B reduced 58% on reproduction)

Figure 6 shows the microstructure of the base metal and HAZ. In the base material, which was in the fully heat-treated condition, two types of Al-Fe-Mn-Si intermetallic compounds are visible, one relatively coarse (1-10 μm in their largest dimension) and the other as a high density of dark speckles barely resolved in the optical microscope. These fine particles were revealed in the transmission electron microscope (Fig. 7) as rounded particles ~ 0.1 μm in diameter.

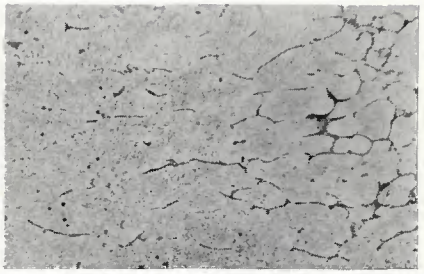

Fig. 6—HAZ microstructure of weld 6 on the inner uncracked side. Etched in Keller's reagent; ×500 (reduced 50% on reproduction)

Transmission electron microscopy also revealed the fine Mg_2Si precipitates which are responsible for hardening in this alloy—Fig. 7B. Fine precipitates were present up to within approximately 0.1 μm of the grain boundaries, suggesting that there were no marked Mg and Si composition gradients at the boundaries before aging.

In the HAZ the density of fine particles was reduced and some of the grain boundaries appeared outlined by films or particles—Fig. 6. On the right-hand edge of the micrograph the typical cellular weld metal structure is visible. Transmission electron microscopy of the HAZ showed it to be in the solution-treated condition, the fine Mg_2Si precipitates having been taken into solution. In agreement with the optical observations it can be seen that the density of fine particles was lower and that they were also coarser than in the base metal—Fig. 8.

Scanning transmission electron microscope microanalysis did not reveal any segregation at grain boundaries. Also, as shown in the micrographs,

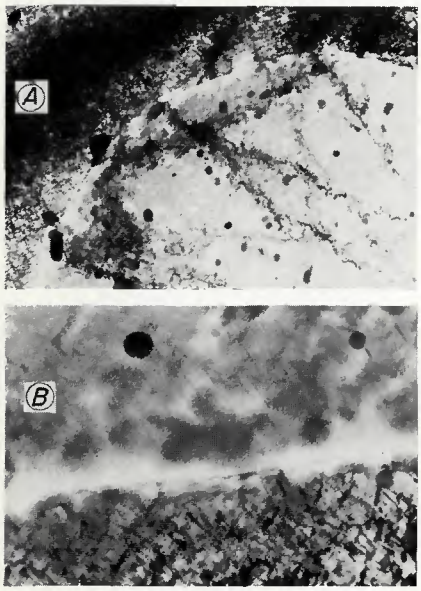

Fig. 7—Thin foil electron micrographs of base metal in the as-received (solution treated and aged) condition. Rounded particles (~ 0.1 μm in diameter) are Al-Mn-Fe-Si intermetallic, and fine needles or rods are Mg_2Si precipitates. A—×32,000; B—×110,000 (A and B reduced 61% on reproduction)

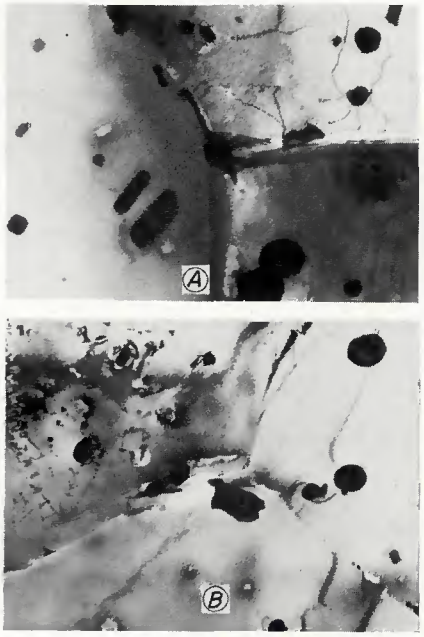

Fig. 8—Thin foil electron micrographs of grain boundaries in the HAZ. Note the absence of fine Mg_2Si precipitates and the decreased density and increased size of Al-Mn-Fe-Si intermetallic participates. ×32,000 (reduced 63% on reproduction)

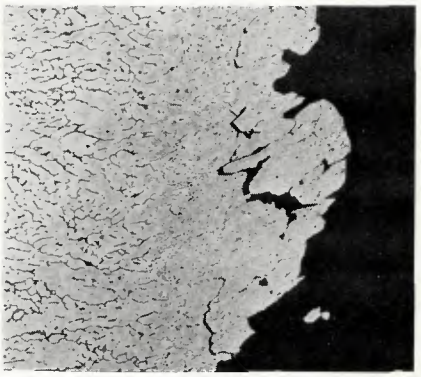

Fig. 9—HAZ cracking in weld 6. The cracking in intergranular and close to the fusion boundary. Etched in Keller's reagent; ×200 (reduced 50% on reproduction)

there were no continuous grain boundary films. One small region at a grain boundary apparently rich in silicon was observed. This is in line with the findings of Thomas,[8] who observed that Si particles can form at grain boundaries on quenching.

The location of the HAZ cracking is shown in Fig. 9. The fracture path runs intergranularly within a few grains of the fusion boundary. The intergranular nature of the cracking is clearly shown in the scanning electron micrographs of a crack surface—Fig. 10. The crack surface had the smooth and rounded features characteristic of intergranular liquation. There were isolated particles enriched in Fe and Mn distributed over the fracture—presumably, Al-Mn-Fe-Si intermetallic compounds.

Qualitative X-ray analysis of the fracture surface in the SEM using the normal operating voltage of 25 kV indicated no apparent differences from the bulk composition. However, when the accelerating voltage was reduced to 7.5 kV, the relative counts for Mg and Si increased compared to those for Al—Fig. 11. One of the effects of reducing the accelerating voltage is to reduce the depth from which X-rays are emitted from the sample; therefore, this result suggests that the composition of the surface layer differs from material lying below it, being relatively richer in Mg and Si.

Such differences in composition would be expected to modify the distribution of Mg_2Si precipitates obtained on subsequent aging. A section of one of the ring test was therefore given an aging heat treatment at 175°C (347°F) for 8 h. Thin foil electron microscopy (Fig 12) showed a similar distribution of Mg_2Si precipitates to that observed in base metal, given the conventional solution treatment and aging.

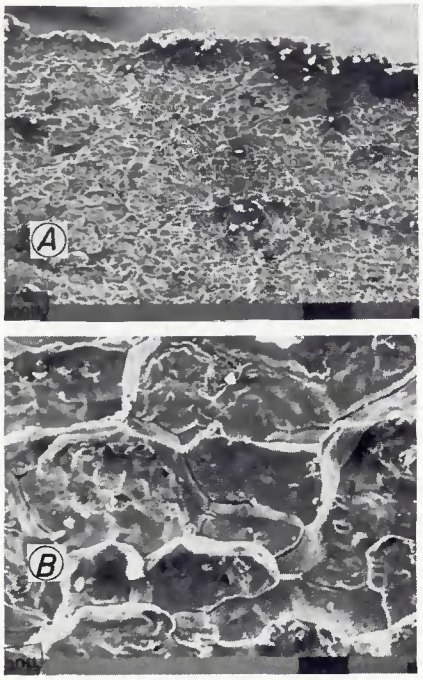

Fig. 10—Scanning electron micrographs of a HAZ crack surface: A—×130; B—×650 (A and B reduced 50% on reproduction)

Discussion

In many alloy systems in which liquation occurs (e.g., Ni-Cr-Fe alloys), there is some evidence that the liquation is caused by minor elements in the alloy and increased concentrations of such elements can sometimes be detected on fracture faces. In some of the service failures in Al-Mg-Si alloys, it had been noted that there was a tendency for a second phase to be present at grain boundaries near cracks. However, the investigation of HAZ microstructure has not revealed any continuous or even semi-continuous network of low melting point constituents at grain boundaries likely to have liquated during welding. The only particles observed in the HAZ during microstructural and crack-face examination consisted of Al-Fe-Mn-Si intermetallics and possibly some Si particles. The intermetallic particles showed no evidence of melting and were not present as a grain boundary network.

Therefore, an alternative explanation must be sought. Such an explanation can be found in the solidus figures of the Al-Mg-Si ternary system as given by Phillips.[9] Figure 13 shows the solidus temperature in the Al-rich corner of the ternary diagram. It will be noted that quite small changes in composition alter the solidus temperature by several degrees.

The compositions and temperatures of the solidi given by Phillips apply only to the pure ternary system and, in commercial alloys, the exact form of the diagram may be modified by the presence of additional elements and trace impurities. The presence of Mn will probably not alter the temperature greatly. Since Mn in the amounts concerned has little effect in the Al-Mn binary alloy, it seems probable that the solidus temperature will be lower than for the Al-Mg-Si ternary. Thus, if the concentration of Mg and Si was higher at grain boundaries, this could lower the solidus temperature locally and make the grain boundaries more susceptible to liquation during welding.

The aging experiment showed no evidence that the composition changed with respect to Mg and Si as the grain boundaries were approached. However, the X-ray analysis carried out in the SEM suggested that there was some increase. Application of the ternary diagram to weld metal compositions show that there is a significant difference between the Al-Si

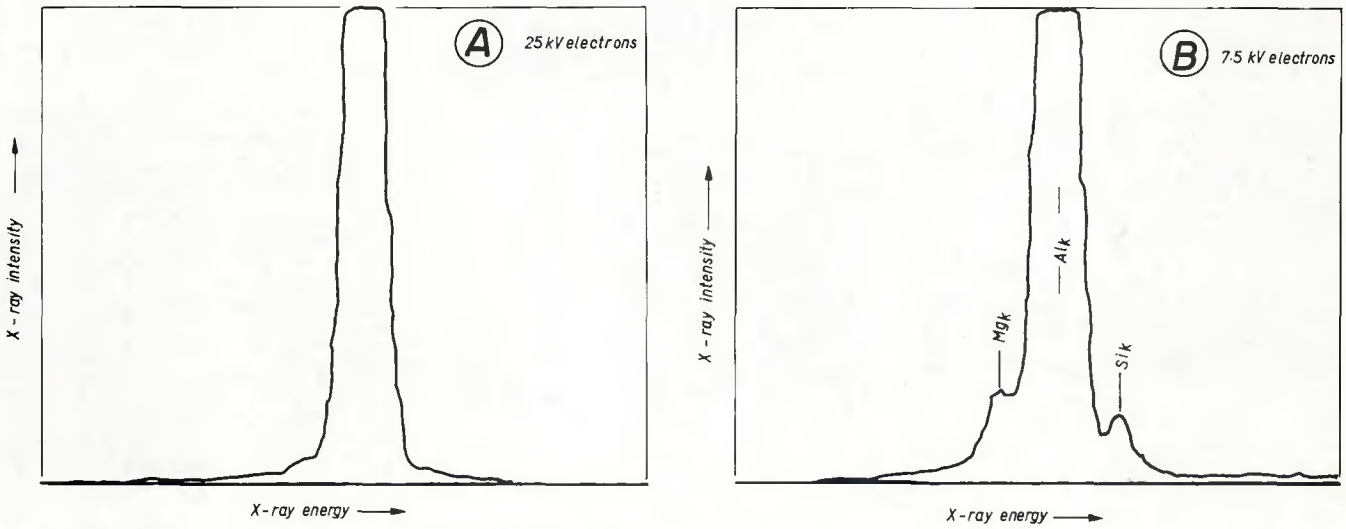

Fig. 11—Qualitative X-ray energy dispersive analysis of the crack surface of one of the HAZ cracks using different accelerating voltages: A—25 kV; B—7.5 kV

and Al-Mg filler metals.

The composition of sheet 2 and the two filler metals used for the tests have been superimposed on Fig. 13. By drawing straight lines between the sheet compositions and the filler metal compositions, it is possible by simple proportion to establish the weld metal composition for a known dilution and interpolate the solidus temperature of that weld metal.

The solidus temperature of sheet 2 is 580°C (1076°F). Welds in this sheet of less than 95% dilution made with NG21 filler metal of the composition used for the tests will have solidus temperature below 568°C (1054°F), as shown in Fig. 14. However, Fig. 14 shows that it is possible for welds made with the NG61 filler metal to have solidus temperatures in excess of that of the base material, if the dilution is in the approximate range 50-90%.

If an alloy is heated to above its solidus, it is quite common for melting to start at grain boundaries; this will happen in the HAZ near the fusion boundary. With certain dilutions of NG61 (5% Mg) filler metal, it is apparent that the weld metal will solidify before any such constitutionally liquated boundaries. The shrinkage strains produced by solidification of the weld metal could then cause parting of the liquated boundaries in the HAZ.

Fig. 12—Thin foil electron micrograph of HAZ after aging for 8 hours at 175°C (347°F). ×40,000 (reduced 50% on reproduction)

Severe cracking was observed with all the welds made with NG61 except weld 11. The higher dilution of the weld reduced the solidus temperature close to that of the base metal; consequently, conditions for cracking were borderline. Both the sheet compositions used for the patch tests contain similar levels of the important alloying elements Mg and Si. Thus, the solidus temperatures and hence cracking susceptibility were very similar.

According to the above model, no cracking would be predicted (or was observed) using NG21 (5% Si) and HS30 of the compositions used in the tests. Nonetheless, the range of composition allowed in the specification of HS30 includes alloys which could have a solidus temperature of 555°C or 1031°F (Mg ≈ 0.85% and Si ≥ 1.1%). Any dilution of NG21 filler metal by such an alloy will produce a weld metal of higher solidus temperature. An alloy of this composition would also produce weld metal of higher solidus temperature when welded

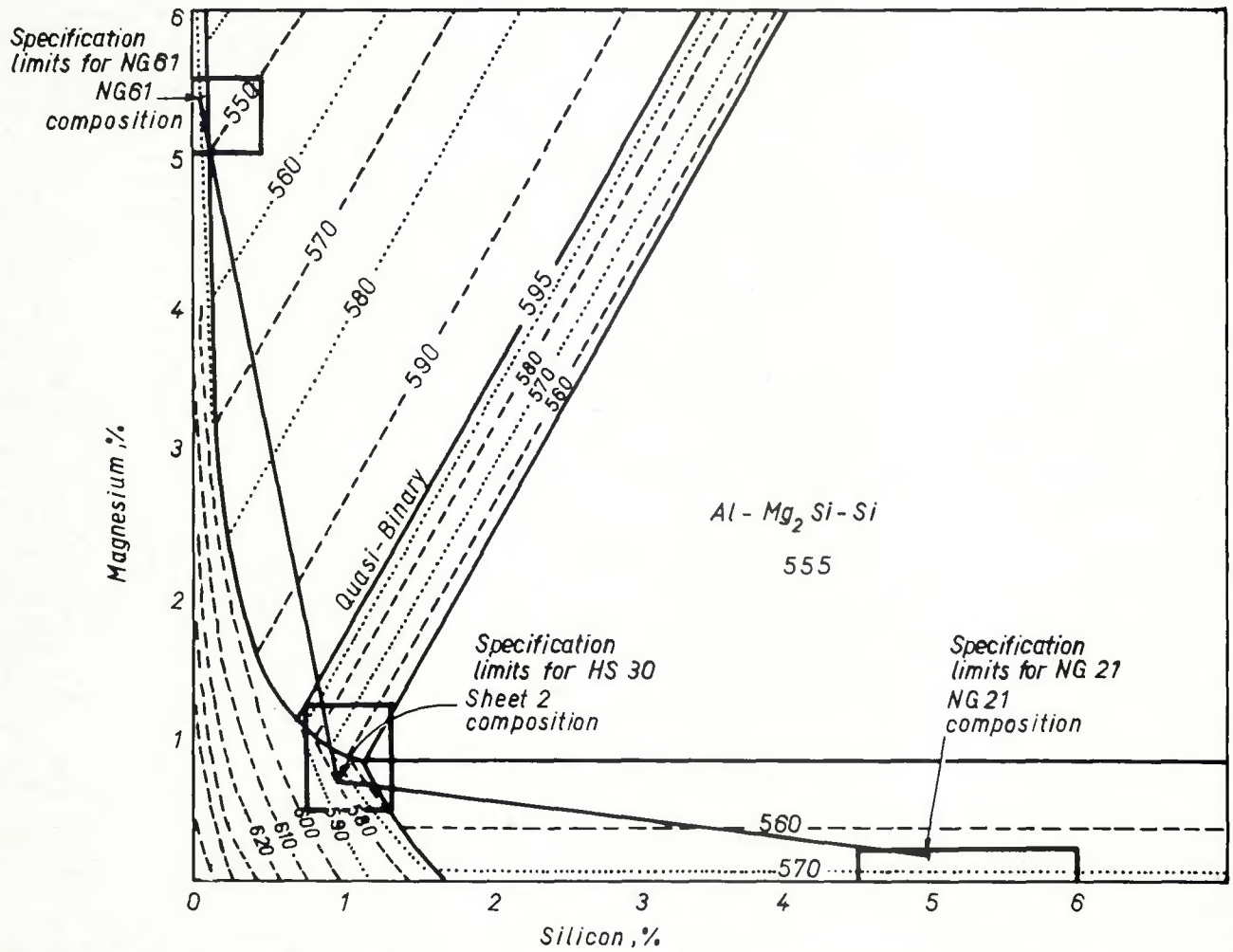

Fig. 13—Al-rich corner of the Al-Mg-Si ternary system showing solidus temperatures (°C), compositions of the experimental materials, and specification limits for the experimental materials

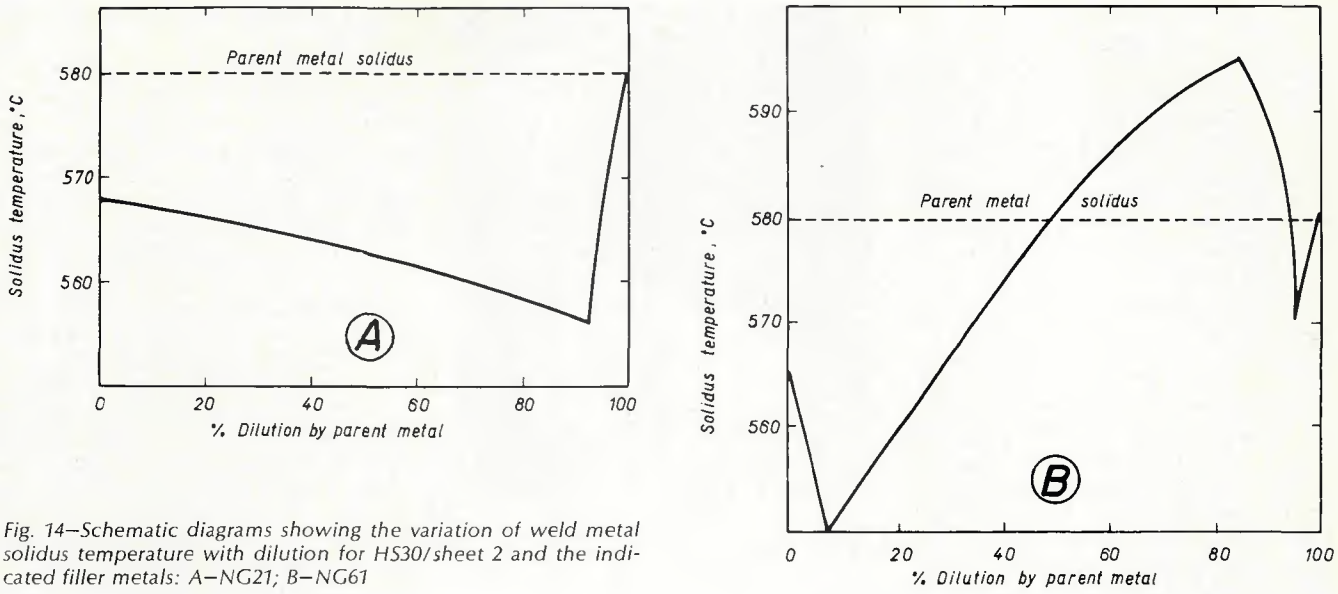

Fig. 14—Schematic diagrams showing the variation of weld metal solidus temperature with dilution for HS30/sheet 2 and the indicated filler metals: A—NG21; B—NG61

with NG61 consumables over all but a very restricted range of low dilutions. Conversely, batches of H30 of lean alloy content would have higher solidus temperatures and be less susceptible to cracking.

On the basis of its Mg and Si contents (1.2 and 0.8% max. respectively), 6061 would not be expected to be as susceptible to HAZ cracking as H30. Indeed, it may not be, but several of the service cases seen by the Institute have been in 6061. However, 6061 does contain Cu and, as discussed earlier, the presence of this element would be expected to decrease the solidus and hence increase the susceptibility.

Alloys of specified lower alloying content such as 6063 and H9 (containing approximately 0.5% of Mg and Si) would have solidus temperatures well above that of any likely weld metal and, therefore, would not be expected to crack. Indeed, no instances of cracking in H9 are known to the Institute, but too much emphasis must not be put upon this since it may be due to less H9 being welded (it being of lower strength than H30) rather than to any differences in susceptibility.

In Fig. 13 the composition ranges specified for HS30, NG61 (5% Mg) and NG21 (5% Si) have been superimposed on the diagram of solidus temperatures. It can be observed that the difference in solidus temperature between weld pool and base metal material is more sensitive to the composition of the base metal than the filler metal. Indeed, any Si content above about 2% in the filler metal will give the same relation between dilution and solidus temperature. There is rather more sensitivity to Mg content and the higher this is, the greater the tolerable range of dilution, although the effect is small.

In the present tests, compositional factors outweighed any influence of restraint, except insofar as cracking always occurred on the outside edge of the patch, never the inside. This is almost certainly because the HAZ on the inside of the weld is in compression, whereas the outside is in tension. This illustrates that there are two requirements for HAZ liquation cracking, the appropriate relation between weld metal and base metal solidus temperature and tensile strain.

As regards remedies, one absolute remedy is to use Al-Mg alloys for welded fabrications, rather than Al-Mg-Si alloys, wherever possible. This has three advantages:

1. Elimination of the risk of HAZ cracking.
2. Reduction of the risk of weld metal cracking.
3. Greater strength in the as-welded state.

However, the greater availability of Al-Mg-Si in some product forms, particularly extruded sections, will inevitably restrict the application of this remedy.

If an Al-Mg-Si alloy must be used, then lean alloys such as H9 or 6063 would be preferable, supposing that the strength is adequate. If higher strength is required and H30 or 6061-type alloys must be used, an alloy with composition at the lower end of the specification should be less susceptible. However, it will seldom be possible to exercise such a choice.

A significant influence on the risk of cracking is exerted by the filler metal. Filler metals containing Si, such as NG21, 4043 or S-AlSi5, are much less likely to cause cracking than those containing Mg, such as NG6, NG61, 5356, 5556, 5183 or S-AlMg4.5Mn. However, it will not always be possible to use a Si-bearing filler metal, e.g., where postweld anodization is to be carried out or where the Al-Mg-Si alloy is to be welded to an Al-Mg alloy. In these cases, a Mg-bearing filler metal must be used. Moreover since a higher Mg content seems likely to be beneficial, filler metals such as NG61 or 5556 may have some advantage over NG6, 5356 or 5183, since the Mg content of the former group will usually be higher.

Where the components permit, the risk of HAZ cracking may also be reduced by jigging them for welding, so that the joint is in compression. This expedient is, however, necessarily of limited application.

Conclusions

A series of patch tests were GTA welded using the Al-Mg-Si alloy HS30 and Al-Mg and Al-Si filler metals (NG61 and NG21 respectively). After metallographic examination of the test pieces, the following conclusions can be drawn:

1. Severe HAZ liquation cracking occurred when using NG61 filler metal with base metal dilutions of ~ 80%.
2. Heat-affected zone cracking did not occur when using NG21 filler metal.
3. No secondary low melting point phases were detected in the HAZ or on liquated crack surfaces.
4. No evidence was found that minor elements were associated with liquation, but there was some indication that the levels of Mg and Si were higher at crack surfaces than in the bulk material.
5. The proposed mechanism of HAZ cracking is that, during welding, grain boundary melting occurs in the HAZ and with certain base metal and

weld metal compositions, it is possible for the base metal solidus to be below the weld metal solidus. Thus, when the tensile strains arising from weld metal solidification are imposed on the HAZ, cracking occurs at such boundaries.

Recommendations for Avoiding HAZ Cracking

1. For welded fabrications Al-Mg alloys, when available, are preferable to Al-Mg-Si alloys.
2. When alloys of the Al-Mg-Si type are welded they should, if possible, be of low alloy content.
3. The maximum resistance to HAZ cracking would be expected using Al-Si filler metal. When an Al-Mg filler metal is used, the Mg content should be as high as possible.
4. In some cases, careful control of dilution can be used to avoid cracking when compositions are known. In general, low dilutions may prevent cracking with Al-Mg filler metals while, if problems are encountered with Al-Si filler metal, increasing the dilution is more likely to reduce HAZ cracking, although this will increase the risk of weld metal cracking.
5. Since tensile strain is necessary to cause cracking, it may be possible to avoid cracking with susceptible material by adapting the procedures used for the autogenous welding of these alloys, i.e., ensuring that the HAZ is in compression.

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The Influence of multiaxial Stress on Low-Cycle Fatigue of Cr-Mo-V Steel at 1000°F

by R. H. Marloff and R. L. Johnson

Low-cycle, strain-controlled fatigue tests were performed on 1Cr-1Mo-1/4V rotor steel at 1000°F (538°C) under uniaxial and biaxial states of stress corresponding to plane strain. An elastic-plastic finite element analysis of the plane strain test specimen was used to evaluate the strain at the notch tip as a function of the notch opening displacement which was controlled in the tests. The results were also extended to other biaxial stress conditions.

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WRC Bulletin 265
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Interpretive Report on Small-Scale Test Correlations with K_{ic} Data

by R. Roberts and C. Newton

Correlations of Charpy test results for the upper shelf region and three types of transition region correlations are evaluated. The efforts of plate position and scatter of the experimental results are also noted. The materials reviewed are steels with yield strengths between 250 and 760 MPa (36 and 100 ksi).

Publication of this report was sponsored by the Subcommittee on Effective Utilization of Yield Strength of the Pressure Vessel Research Committee of the Welding Research Council.

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