Effect of Alloying Additions on the Weldability of 70 Cu-30 Ni

The effects of four elements—P, Sb, Fe and Mn—on hot-cracking sensitivity are analyzed statistically, and Sb and the combination of Sb-Mn are found to be beneficial

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ABSTRACT. The effects of minor alloying additions on the hot-cracking sensitivity of 70 Cu-30 Ni alloy were evaluated in a full-factorial experiment. Composition modification of the as-received base material was accomplished by means of a novel remelt operation, and the weldability of the alloy was evaluated by the employment of the Tigamajig test.

By applying the method of Yates, the resulting treatment effects of four elements—P, Sb, Fe, and Mn—were tabulated for the statistically significant, as-welded cracking parameters. It was proposed that the most harmful combinatorial treatments were those which most severely lowered the effective solidus temperature of the alloy and, thus, most drastically impaired, the hot ductility of the material.

By means of a Gleeble-simulation technique, considerable improvement in the hot ductility of a P-modified lot of material was observed following a cold reduction between welding and testing events. It is suggested that improved hot-cracking resistance could be expected in subsequent passes in multipass welding if an interpass cold reduction of the weldment were employed.

Because hot cracks were backfilled, the effective distribution coefficients for the elements Fe, Mn, and Ni could be calculated from electron beam microprobe data.

Introduction

Hot cracking in a single-phase alloy generally occurs above the solidus temperature of the lowest-melting composition present. The degree and extent of solute microsegregation in the weld microstructure determines the magnitude of the lowest effective solidus temperature present. Thus, factors which modify the prevailing growth mode in the solidified weld will greatly influence hot-cracking sensitivity. A certain degree of control over both the temperature gradient existing in the liquid and the growth rate of the solid is possible by selection of the welding parameters.

Control of the shape of the weld pool by careful selection of welding conditions can drastically modify the pattern of microsegregation existing in the solidified weld.

Certain minor alloying elements are believed to exert considerable influence on weldability. The impact of the composition factor is compounded by the heat-to-heat variations in compositions of both the major and minor alloying elements permitted for a given commercial material.

As a result of input from industry (via INCRA), four elements and appropriate levels for each were selected for study, based on control difficulties in scrap processing and/or alloying content required for physical properties, e.g., corrosion resistance. The pertinent reported effects of P, Sb, Fe, and Mn on the mechanical properties and weldability of Cu, Ni, and Cu-Ni alloys may be summarized as follows:

1. Small additions of Fe, Mn, P, and Sb solid-solution strengthen Cu, Ni, and Cu-Ni alloys at the levels of concentration employed in this investigation. Sb is a much more effective hardener of Cu than is Mn or P.

2. P is universally identified as an embrittling element in Cu-Ni and Ni alloys subjected to hot work. It has been reported, furthermore, that P is also injurious to the weldability of the cupronickels. When present in oxygen-free Cu, P also results in a loss of high-temperature ductility.

3. Sb appears to have no detrimental effect on either the hot workability of Cu or Ni, or on the weldability of Cu-Ni alloys. Sb in oxygen-free Cu appears to increase the room-temperature toughness slightly.

4. Fe and Mn are generally considered innocuous alloying additions. Neither element seems to have a great effect on either the mechanical properties or weldability at low concentrations in Cu, Ni, or Cu-Ni alloys.

Object

The objectives of this experiment were to:

1. Investigate the effect of four minor alloying elements, P, Sb, Fe, and Mn, often present in commercial cupronickel, either as intentional additions or as “tramp” elements, on the hot-cracking propensity.

2. Determine whether an improvement in hot-cracking resistance could be effected by a thermo-mechanical treatment involving cold work and recrystallization.

Materials and Procedure

Hot-Cracking Evaluation

Material. The 70 Cu-30 Ni base alloy was used in this study.
The material used in this investigation was arc-cast at Inco's Sterling Forest facility and was hot rolled to ⅛ in. (3.2 mm) plate by the Revere Copper and Brass Company. Table 1 summarizes the chemical composition of the unmodified as-received, basic binary alloy.

**Experimental Design.** The effects of four minor alloying elements—P, Sb, Fe, and Mn—on the hot-cracking sensitivity were investigated by Tigamajig tests™ performed on 16 lots of modified composition produced in as-received plate by means of the remelt operation described below. The effects of these elements were investigated at two levels of concentration:

1. The nominal level of the high-purity, as-received material.
2. All of the possible 16 combinations in the full factorial experimental design of 0.01 w/o P, 0.003 w/o Sb, 0.25 w/o Fe, and 0.3 w/o Mn.

**GTA-Remelt Operation.** The remelt operation used to modify the compositions of a relatively small volume of the plates is shown in Fig. 1 and is summarized below:

1. An alloy insert, prepared by cold compacting an appropriate mixture of high-purity metal powders for each combinational treatment, was inserted into a pre-machined ⅛ in. (8.7 mm) wide groove along the 5 in. (127 mm) center line of each of the sixteen ⅛ X 5 X 6 in. (3.2 X 127 X 152 mm) specimen plates.
2. A series of three GTA passes were made using a sinusoidal wave pattern to remelt and mix the insert with a volume of the basic binary plate. The plates were mechanically restrained during this operation on an Al heat-sink to maintain constant heat-flow characteristics from plate to plate. By standardization of the remelt-welding procedure, the extent, shape, and prevailing growth mode of the composition-modified remelted region were kept as similar as possible. The welding conditions summarized in Table 2 were chosen to maintain a full-penetration weave bead.
3. Atmospheric contamination was minimized by performing the GTA-remelt operation in high-purity Ar within a dry box. Starting and run-off tabs were applied to the plates to minimize end effects.

The compacts were made from high-purity Cu, Ni, Sb, Fe, Mn, and Cu,P powders. Following blending, the powders were cold-compacted under 30 tons/in.² (4220 kg/cm²) using a minimal film of zinc stearate-acetone lubricant applied to the sides of the die plunger. The dimensions of each compact were ⅛ X ⅛ X 3 in. (1.6 X 8.7 X 76 mm); two were required for each plate.

Following the GTA-remelt operation, reinforcements on the bottom of the plates were milled flat, and the plates were cut, drilled, and degreased to provide test specimens, as illustrated in Fig. 2. Microscopic observation at X40 failed to indicate any cracking in either the remelted region or in the base material in any of the GTA-remelted specimens.

**Tigamajig Testing.** With the welding conditions shown in Table 3, a stationary GTA arc was struck at midspan of each sample and maintained for 9 seconds (s). At the end of that time, a nominal augmented strain of 0.785% was suddenly applied to the specimen through the action of the die block, with arc cut-off occurring simultaneously. The arc time was selected to provide a weld pool showing incomplete penetration prior to the establishment of steady-state heat-flow conditions.

As a result of these events, hot

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**Table 1—Chemical Analyses of As-Received Material and GTA-Remelted Samples**

<table>
<thead>
<tr>
<th>Element</th>
<th>Unmodified (as-rec'd material)</th>
<th>Modified (GTA-remelted)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cu</td>
<td>69.96</td>
<td>-</td>
</tr>
<tr>
<td>Ni</td>
<td>29.99</td>
<td>-</td>
</tr>
<tr>
<td>P</td>
<td>0.0068</td>
<td>0.006</td>
</tr>
<tr>
<td>Sb</td>
<td>0.002</td>
<td>0.0025</td>
</tr>
<tr>
<td>Fe</td>
<td>0.03</td>
<td>0.18</td>
</tr>
<tr>
<td>Mn</td>
<td>0.0013</td>
<td>0.23</td>
</tr>
<tr>
<td>Pb</td>
<td>0.0001</td>
<td>-</td>
</tr>
<tr>
<td>S</td>
<td>0.0017</td>
<td>-</td>
</tr>
</tbody>
</table>

**Table 2—Welding Conditions Used During Remelt Operation**

- Welding current: 144 A dcsp
- Welding voltage: 9-10 V
- Arc gap: ½ in. (2.4 mm) measured cold
- Electrode: ½ in. (3.2 mm) EWT50
- Oscillation: 0.9 ipm (22.9 mm/min)
- Oscillation frequency: 26 oscillations/min; sinusoidal pattern
- Fusion travel speed: 90 deg incl. angle
- Welding current: 9-10 V
- Welding voltage: 7-8 V
- Arc gap: ½ in. (2.4 mm) measured cold
- Electrode: ½ in. (3.2 mm) EWT50
- Oscillation: 0.9 ipm (22.9 mm/min)
- Oscillation frequency: 26 oscillations/min; sinusoidal pattern
- Fusion travel speed: 90 deg incl. angle
- Welding current: 9-10 V
- Welding voltage: 7-8 V
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Fig. 1—Schematic illustration of remelt operation: (A) machined plate; (B) melting and mixing of powder metal compact (plate is mechanically restrained along AB and CD); (C) preparation of Tigamajig specimens.
cracks were produced within the modified region adjacent to the GTA spot weld in all specimens. This situation, therefore, resembles hot cracking that occurs in a previous pass during a multiple-pass weld made under restraint. The appearance of the hot cracks produced is illustrated in Fig. 3.

Cracking Data. After testing, the total crack length, the maximum crack length, and the average crack length were measured at X 40 in each specimen in the as-welded condition. Similar data were accumulated for each specimen following metallographic polishing to a horizontal plane located approximately 0.010 in. (0.254 mm) below the original surface.

Chemical Analyses. Chemical analyses of the remelted regions were made for each of the experimental specimens. From the original specimens, groups were selected on the basis of their chemical composition. The distributions of compositions and important statistical quantities of these groups are illustrated in Figs. 4-7. Only the cracking data from these specimens were considered in this investigation.

The average compositions of the alloy modifications, shown in Figs. 4-7, are also presented in Table 1 for comparison with the analysis of the unmodified base metal.

Electron-Beam Microprobe Analysis. Most of the hot cracks were backfilled, as shown in Fig. 8. The liquid in the diffusion gradient at the solid-liquid interface tends to penetrate, or backfill, the newly opened crack. Thus a good estimate of the distribution coefficients of the various solutes can be obtained by analysis of the material in the backfilled crack.

This segment of the investigation was carried out on an electron probe microanalyzer by the point-count technique. A typical traverse is indicated by line AB in Fig. 8. The graphical results of typical analyses are shown in Figs. 9, 10, and 11, for Ni, Fe, and Mn, respectively.

Weldability Improvement

This portion of the investigation, concerned with improving the hot-cracking sensitivity of a particularly susceptible, P-modified cupronickel, involved hot-ductility measurements carried out on the Gleeble. Two types of specimens were prepared:

1. Plates initially 0.125 in. (3.2 mm) thick and modified with a P-addition

![Diagram](image-url)
Fig. 6—Distribution of compositions for specimens containing Fe: (A) unmodified, (B) modified

Fig. 7—Distribution of compositions for specimens containing Mn: (A) unmodified, (B) modified

Fig. 8—Backfilling in a Tigamajig hot crack. Line AB denotes representative electron-beam microprobe traverse. Modified Marbles reagent etch, section parallel to plate surface. ×225 (reduced 27% on reproduction)

Fig. 9—Electron-beam microprobe analysis results for Ni

Fig. 10—Electron-beam microprobe analysis results for Fe

Fig. 11—Electron-beam microprobe analysis results for Mn

Gleeble specimens, 0.1 in. (2.5 mm) thick by ¾ in. (7.9 mm) wide by 3½ in. (82.6 mm) long, were machined from these plates with the composition-modified region centered longitudinally. These specimens were subjected to either of two thermo-mechanical cycles (Program A or B) in the Gleeble, with all testing being done in an argon (Ar) atmosphere chamber.

Program A. Specimens were heated to a peak temperature, $T_p$, and held for times ranging between 0 and 1500 s before rapid strain application. The peak temperature in every case was reached in 9 s, the same time interval used prior to loading during Tigamajig testing.

The magnitude of $T_p$ was chosen to lie between the nominal solidus (2138 F, 1170 C) and the instantaneous temperature of the region to which cracking was observed to extend. The approximate extent of this hot-cracking range was measured by thermocouples attached to Tigamajig specimens during testing. Final selection of $T_p$ was based on tests made on Gleeble
specimens heated to temperatures within the hot-cracking range measured above.

Specimens heated to 2113 F (1156 C) and 2097 F (1147 C) exhibited gross melting; two specimens heated to 2081 F (1138 C) exhibited approximately zero ductility, with no extensive bulk melting. On this basis, 2081 F (1138 C) was chosen as the peak temperature. The straining rate of the Gleeble was made equal to the straining rate of the Tigamajig.

Program B. Program B consisted of two separate operations. The first consisted of subjecting Gleeble specimens to a typical annealing cycle by heating to 1450 F (788 C) at the Tigamajig heating rate and holding at that temperature for various lengths of time (from 0 to 720 s), followed by free cooling to room temperature.

The second part consisted of ductility evaluation by heating of the specimens, at the Tigamajig heating rate, to Tp = 2081 F (1138 C), followed by rapid strain application.

Transmission Electron Microscopy. TEM was employed to determine, qualitatively, the effect of the various thermo-mechanical histories on the distribution of dislocations about grain and/or substructure boundaries in the material. Three Gleeble specimens were studied:

- Specimen 1, in the as-welded condition, was representative of the composition-modified region of the Tigamajig specimens.
- Specimen 2, in the cold-worked condition, corresponded to approximately 20% reduction in sheet thickness.
- Specimen 3, originally cold worked, was recrystallized by a thermal excursion to 2081 F (1138 C) for 300 s, followed by free cooling.

The resulting electron micrographs are reproduced in Figs. 12, 13, and 14, respectively.

Results

Hot-Cracking Results

The method of analysis proposed by Yates2 was applied to the cracking data. Statistical estimates of the effects of each treatment and estimates of error were calculated for the following three cracking parameters: total, maximum, and average crack length.

The results of the calculations for the as-welded cracking parameters are summarized in Table 4. The values recorded in this table may be interpreted as follows. In the case of the Sb-Mn experimental treatment (e.g., the 23 total-crack-length value at the 5% level of significance), there is a 95% chance that the total crack length will decrease by 0.023 in. (0.58 mm) relative to the unmodified alloy. Again considering the Sb-Mn treatment (e.g., the 1.2 average-crack-length value at 5% level), there is a 95% chance that the average crack length will decrease by 0.0012 in. (0.03 mm) relative to the unmodified alloy.

For comparison purposes, the average values of the cracking parameters for the unmodified base alloy were: total crack length, 0.199 in. (5.1 mm); maximum crack length, 0.028 (0.7 mm); and average crack length, 0.016 in. (0.4 mm).

Analyses of data taken following metallographic polishing of the specimens yielded no significant effects at the 1, 5, or 10% levels and, therefore, are not considered further in this discussion. The absence of statistically significant effects was related to the backfilling process which obscured the hot crack.

For the as-welded data, effects of the factorial treatments on the total-crack-length parameter were significant at the 1% level in the case of six treatments. In other words, in those six cases, there is no greater a probability than 0.01 that the effects measured were due to chance fluctuations in the data. The following summarizes the analysis of total-crack-length data at the 1% level:

1. P was detrimental to the weldability of this alloy, especially in the presence of Sb; although Sb alone appears to be a beneficial addition.
2. Fe and Sb when present together are somewhat detrimental, while Sb alone appears to be beneficial and Fe alone does not have a significant effect.
3. In the presence of Mn, which alone is detrimental, the combination of Fe and Sb represents the worst effect noted.

A similar analysis of data for the maximum-crack-length parameter yielded the following:

1. In the presence of Mn, the combination of Fe and Sb was detrimental.
2. The elements P and Sb when each was present with Fe, were injurious to this cracking parameter, although none of these elements had a significant effect when present alone.

Finally, the following results were noted for the average-crack-length parameter:

1. Sb decreased the average crack length.
2. The presence of P with either Sb, Fe, or Mn increased the average crack length.

Electron-Beam Microprobe Analysis Results

Microprobe data gave the following results:

1. Mn segregated to the last liquid to solidify. The average composition within the backfilled hot cracks was about 0.57 wt-%, while the average composition of the uncracked, remelted region was about 0.38 wt-%.
2. Fe segregated away from the last solidifying liquid. While the average composition of the unaffected remelt region was about 0.32 wt-%, the average composition within the backfilled hot cracks was about 0.16 wt-%.
3. The composition of Ni within the backfilled hot cracks averaged about 18 wt-%, while the composition of the remelted region averaged about 30 wt-%.
4. The levels of Sb and P did not differ significantly from the unaffected to the cracked regions. The levels at which these elements were present in the composition-modified region were too low for accurate measurements.

5. When Fe and Mn were present together at their high levels, no difference in the type of segregation each exhibited individually was noticed.

Results of Gleeble Investigation

The data resulting from the Gleeble studies are plotted in Fig. 15. The following points are evident:

1. The recovery of ductility measured at 2081°F (1138°C) was most pronounced in the case of material cold rolled 20% between welding and testing operations and least noticeable in the material tested in the as-welded condition.
2. The hot ductility of a cold-rolled specimen tested immediately on reaching 2081°F (1138°C) is 8%, while the ductility of the as-welded specimens tested at the same point is essentially zero.
3. Annealing of cold-rolled specimens at 1450°F (788°C) following welding also leads to an improvement in ductility with time at annealing temperature, but the rate of ductility recovery is slower than in the case of specimens exposed to higher temperatures for the same times.

Discussion of Results

Hot-Cracking Studies

Following establishment of the arc spot during Tigamajig testing, the resulting instantaneous temperature gradient encompasses a portion of the remelted region where the actual
temperatures lie between the effective liquidus and solidus. Because of the point-to-point variation in the effective solidus resulting from microsegregation, those portions solidifying last during the remelt operation (i.e., interdendritic regions) experience actual temperatures above their effective melting points. This region of partial melting, which is characterized by liquid-covered interdendritic surfaces, exists at the instant of augmented strain application.

A large majority of the cracks observed in this investigation had the appearance illustrated in Fig. 8. It was noted that:

1. Initiation of cracking occurred in the partially melted region.
2. The cracking was intimately related to the solidification substructure boundaries.
3. The cracks, most of which were backfilled, seemed to have extended well beyond the partially melted region.

It is proposed that the hot cracking observed depended on three contributing factors, which occurred on sudden loading during Tigamajig testing:

1. Sliding of liquid-covered interdendritic boundaries under shear stresses.
2. Separation of liquated boundaries under the influence of normal stresses.
3. Crack extension through liquated boundaries in the solidification substructure.

The first two events may be identified with the crack-initiation event, while the last factor is related to crack propagation.

This crack-extension process may be expected to continue until the crack is effectively stopped through the change in crack-tip geometry and/or the inability of the crack to extend through an unliquated boundary by conventional fracture mechanisms. The degree of crack extension tended to increase in those regions where a solute element solution strengthened the alloy, yet at the same time impaired the ductility and toughness. For example, the expected segregation of P to the last-solidifying liquid in the remelt region should be responsible, based on the literature, for the solid-solution strengthening of that region, with a concurrent notable loss in ductility, relative to regions of lower P concentration. This combination of properties may reasonably be expected, therefore, to result in an increase in the degree of crack extension as noted in this experiment.

**Hot-Ductility Studies**

Figure 12A, an electron micrograph, illustrates the distribution of dislocations in the vicinity of a grain boundary in the P-modified region of an as-welded Tigamajig specimen. Figure 12B represents the situation in regions removed from such boundaries. Comparison of these figures leads to the identification of the following features:

1. Dislocations are present in greater density in the vicinity of the boundary.
2. In several locations, dislocations appear to be anchored at various points along the dislocation length. While this pinning may be attributed to interaction with "tree" dislocations located on intersecting slip planes, this feature probably also reflects the effect of solution strengthening.
3. The above hypothesis is strengthened by the zig-zag dislocation configuration which is thought to reflect the interaction of an otherwise straight dislocation line with solute-atmospheres.

The above-mentioned features tend to confirm the proposal that relatively low ductility may be expected in the vicinity of solidification-substructure boundaries in the remelted region of Tigamajig specimens. When an as-welded specimen was rapidly strained in the Gleeble upon reaching a temperature representative of the cracking range expected in the Tigamajig specimen, essentially zero reduction in area was noted.

Following a mild cold working of a specimen having the prior history described above, the density of dislocations, as illustrated in Fig. 13, increased significantly, as would be expected, relative to Figs. 12A and 12B. More importantly, no features identified with solution strengthening were apparent in the vicinity of the boundaries. A Gleeble specimen of this description, when heated to 2081 F (1138 C) and strained immediately, showed a reduction in area of approximately 6% (Fig. 14). This recovery in ductility is attributed both to the disruption of the solute atmospheres and to recrystallization.

The microstructure in Fig. 15 represents a cold-worked specimen which was heated to 2081 F (1138 C), maintained for 300 s and allowed to free cool in the Gleeble. The transmission electron micrograph shows a marked decrease in dislocation density relative to the cold-worked structure (Fig. 13). In addition, there is again a general absence of those features associated with solution-hardening in the vicinity of grain boundaries, as shown previously in Fig. 12A.

As supported by optical metallography, the combination of time and temperature employed was sufficient to cause recrystallization and the separation of grain and subgrain boundaries from the regions of microsegregation. As might be expected, the ductility of this specimen measured following rapid loading after 300 s at 2081 F (1138 C), approximately 46% reduction in area as shown in Fig. 14, was considerably improved over that of the previous two specimens.

The results of this section suggest a practical means by which this improvement in ductility might be applied to decrease hot-cracking sensitivity during multipass welding. It has been shown that a relatively mild cold working of the previous pass was responsible for the ductility recovery. This mechanical deformation could be effected by shot peening, hammer peening, or roll planishing between welding passes. Selection of welding conditions which could effectively increase the time spent at elevated temperatures (e.g., slower travel speed) could be expected to decrease, further, the hot-cracking susceptibility during multipass welding.

**Table 5—Distribution Coefficients of Indicated Elements**

<table>
<thead>
<tr>
<th>Element</th>
<th>Equilibrium value from diagram</th>
<th>Calculated (microprobe data)</th>
<th>Reported in literature</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe</td>
<td>1.4</td>
<td>2.0</td>
<td>3.8-5.2</td>
</tr>
<tr>
<td>Mn</td>
<td>0.75</td>
<td>0.67</td>
<td>0.45</td>
</tr>
<tr>
<td>Ni</td>
<td>1.2</td>
<td>1.66</td>
<td>1.9-2.4</td>
</tr>
</tbody>
</table>

Fig. 15—Transmission electron micrograph of a Gleeble specimen cold worked 20% following welding and subjected to 2081 F for 300 s followed by a free cool to ambient temperature X 90,000 (reduced 55% on reproduction)
Calculation of Effective Distribution Coefficients

Cs, the concentration of a solute in the last regions of a weld to solidity, is equal to the expression Cc/kc, where kc is an effective distribution coefficient for the particular solute, and Cc is the nominal composition of the alloy. A measure of kc may be calculated by dividing the average concentration of the solute in the region exterior to the crack, Cc, by the average concentration of the solute within the confines of the crack, Cc.

The microprobe data generated in this investigation allowed calculation of effective distribution coefficients for Fe, Mn, and Ni, as shown in Column 2 of Table 5. Values given in Columns 1 and 3 indicate, respectively, the equilibrium value of kc measured from phase diagrams and the value of kc as measured by other investigators. In the case of Fe and Ni, the deviation of the reported values from the equilibrium values is positive; in the case of Mn the deviation is negative. One would expect these deviations because of the suppression of the solute by Fe had no effect, and Sb was beneficial. Secondary-order combinations of P-Fe and Sb-Fe were detrimental, and the combination P-Sb was the most detrimental of all. On the other hand, the combination Sb-Mn was beneficial.

6. A Gleeble-simulation technique showed that, following an interpass cold reduction of the weldment, a significant improvement could be expected in the hot ductility of a region in which Tigamajig cracking was observed. In this case, a P-modified specimen exhibited zero ductility when strained immediately upon reaching a temperature typical of the partially-melted region. A similar specimen, reduced 20% in thickness before being tested under the same conditions, showed a reduction in area of 8%. Longer times prior to the application of load resulted in larger recoveries in hot ductility.

7. Effective distribution coefficients in the 70 Cu-30 Ni alloy were measured by electron-beam microprobe traverses across back-filled hot cracks for the elements Fe, Mn, and Ni.

Acknowledgments

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References


Conclusions

1. The remelt operation, used to modify the minor-element composition of the as-received 70 Cu-30 Ni base material, provided a reasonably reproducible solidification substructure upon which weldability evaluations could be based.
2. A full-factorial experimental design allowed calculation of compositional effects on the hot-cracking propensity. Statistical levels of confidence for those effects were also determined.
3. The hot cracking observed in the Tigamajig test specimens appeared to have originated in liquated boundaries adjacent to the GTA spot. However, the cracks extended beyond the apparent limits of the partially-melted region.
4. The weldability of the composition-modified region was impaired by the presence of minor alloying elements which lowered the effective solidus temperature and had a disproportionately adverse effect on the ductility.
5. Specific cracking tendencies in the 70 Cu-30 Ni alloy were noted from a Yates analysis of as-welded cracking data. As examples, P and Mn were detrimental to the weldability, whereas