Nitrogen Porosity in Gas Shielded Arc Welding of Copper

Low levels of nitrogen can produce porosity but this can be reduced by increased welding speed and eliminated by denitriders in the filler metal

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ABSTRACT. The effects on porosity of base metal composition, current, speed, water vapor content of the argon shielding gas, flow rate and weld preparation were examined for gas tungsten-arc (GTA) welding of copper. Use of butt welds instead of melt runs was found to increase porosity apparently by entrainment of the atmosphere into the arc. A quantitative study was then made of the effect of nitrogen content of the shielding gas, welding speed and filler metal composition on weld metal porosity for both the GTA and gas metal-arc (GMA) welding processes.

It was confirmed that nitrogen porosity can occur in copper weld metal, the problem being more severe in the GTA process than in GMA welding. The threshold levels of nitrogen for porosity formation were sufficiently low in both processes to occur by atmosphere entrainment. The amount of porosity at a given nitrogen level can be reduced by increased welding speed, and can be eliminated by use of filler metals containing small amounts of strong denitriding elements.

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Paper was presented at the 55th AWS Annual meeting held at Houston, Texas, during May 6-10, 1974.

Introduction

Although the widespread use of the inert gas shielded arc processes has eliminated many of the difficulties encountered in the welding of copper, the incidence of porosity in the weld metal has remained a problem and can result in a marked decrease in joint strength (Ref. 1). The porosity invariably takes the form of randomly distributed pores and appears to be most severe in the absence of deoxidants. For this reason the general assumption has been that it was due to the steam reaction, and consequently special filler metals containing small amounts of powerful deoxidants were developed (Refs. 2,3,4). Although use of these filler metals greatly reduced the porosity it did not eliminate the defect, so that the radiographic standards achieved remained inferior to those readily obtained in steel. This failure to eradicate the weld metal porosity suggested that it was not entirely due to the steam reaction.

In a study of argon-nitrogen gas mixtures (Ref. 5) in the metal arc welding of nonferrous metals, Kobayashi showed that extensive porosity formed in copper weld metal. At first sight this seems very surprising due to the fact that nitrogen is reported to be insoluble in copper up to 1400 C (Ref. 6). In explanation Kobayashi postulated that nitrogen was soluble in the molten copper at the higher temperatures encountered in arc welding and subsequently precipitated as the molecular gas during solidification. These results appeared to be in conflict with those investigations which have shown that porosity-free weld metal can be obtained using a nitrogen shielded arc (Refs. 3,4). However, the latter investigations employed filler metals containing aluminium and titanium which happened to be powerful denitriders as well as deoxidants so that there is no contradiction. In view of these results it appeared possible that nitrogen could be responsible for the porosity occurring when welding with an inert gas shield since the gas could be entrained into the arc from the atmosphere.

The present work investigates nitrogen porosity formation in copper weld metal with a view to establishing its role in the inert gas shielded arc processes. In the first instance a survey is made of the factors effecting porosity in general in gas tungsten-arc (GTA) welding. Subsequently, an examination is made of the factors affecting porosity produced after deliberate additions of nitrogen are made to the shielding gas in both the GTA and gas metal-arc (GMA) welding processes.

Experimental Procedure and Results

Factors Affecting Weld Metal Porosity

In order to identify the basic practical factors responsible for porosity formation, a quantitative assessment was made of the influence of the main welding and material variables on weld metal density. This was carried

Table 1 — The Independent Variables in the Survey of Factors Affecting Weld Metal Porosity in GMA Welding

Factors	Independent variable range Low level	High level
Copper base metal	Deoxidized High Phosphorous	Electrolytic Tough Pitch
Water vapor in argon	0	0.005 g/dm^3
Argon flow rate	7.5 dm^3/min	22.0 dm^3/min
Weld preparation	Butt welds	Melt runs
Welding current[a]	350 A and 225 A	380 A and 275 A
Welding voltage	11.5 V	11.5 V
Welding speed	1 mm/s	8 mm/s

Tungsten electrode size 6 mm diam, tip angle 60 deg
(a) Two levels of current to giva full penetration at high and low levels of welding speed.

Table 2 — Significance Levels of the Effect of Single Factors and Interactions on Weld Metal Density

Factor or interaction	Signif-icant level	Factor or interaction	Signif-icant level[a]
Water vapor	99%	Vapor-flow	90.0%
Flow rate	N.S.	Vapor-speed	99%
Welding speed	99%	Vapor-preparation	N.S.
Weld preparation	99%	Base metal-vapor	99%
Base metal	99%	Flow-speed	90%
Current	N.S.	Flow-preparation	99%
Current-vapor	N.S.	Flow-base metal	97.5%
Current-flow	99%	Speed-preparation	99%
Current-speed	95%	Speed-base metal	99%
Current-preparation	N.S.	Preparation-base metal	95%
Current-base metal	97.5%		

(a) N.S. = Not Significant

Table 3 — The effects of Single Factors on Weld Metal Density

Variable	Difference of means low to high kg/dm^3
Water vapor	−1.65
Base-metal	−1.31
Weld preparation	+0.57
Welding speed	+0.23
Welding current	+0.03
Flow rate	+0.03

out with high purity copper plate and autogenous welding using the GTA process. From a review of the likely sources of gas porosity, particularly those likely to cause the steam reaction, the following six variables were selected for study.

1. Deoxidized state of the base metal
2. Composition of the shielding gas
3. Efficiency of shielding gas coverage
4. Weld preparation and fit up
5. Welding current
6. Welding speed

In view of the large number of variables and their possible interaction the experiments were arranged according to a 2^6 factorial design. Thirty-two experiments were conducted with the six variables at low and high levels using a standard half replicated design.

In principle the experiments involved making full penetration weld runs in 3 mm thick copper plates using the GTA welding process at two levels of base metal composition, water vapor content of the argon, argon flow rate, weld preparation, speed and direct current. From previous work (Ref. 7) a specimen size of 300 × 300 mm was taken as representing an acceptable approximation to semi-infinite heat flow conditions. Welding was done by traversing the torch along the centerline of the plates, these being held rigidly in a jig with a thin copper backing strip. High purity argon (99.999%) was used as shielding gas and water vapor was introduced by bubbling through water at room temperature. A level of 1/40th saturation was selected as the upper limit, this giving no condensation problems in the equipment, and it was obtained by mixing streams of "wet" and dry argon in suitable proportions. Details of the levels of the independent variables are given systematically in Table 1. The actual order of execution of the test runs was deliberately random and in each case the weld metal was examined by x-ray radiography and density determination. Invariably the porosity when found proved to be in the form of randomly scattered spherical pores.

An analysis of variance was conducted on all the density results to determine the effect of the six single factors and their fifteen first order interactions. A summary of the calculated significance levels is given in Table 2 where it can be seen that four of the six single factors and six of the interactions achieved the 99% significance level. An estimate of the size and magnitude of the effect of the single factors is given in Table 3.

It is apparent that porosity is increased (density decreased) by increasing water vapor content of the shielding argon, by changing from Deoxidized High Phosphorous Copper to Electrolytic Tough Pitch Copper, by reducing welding speed and by changing the weld preparation from melt runs to butt welds. Over the range examined neither current nor argon flow rate had significant effects on porosity. In general these effects can be explained in terms of the occurrence of the steam reaction in the weld pool and the subsequent formation of porosity (Ref. 8). The major exception is the increase of porosity brought about by the change from melt runs to butt welds.

In order to explore the effect of weld preparation more fully, its interaction with speed was investigated by making melt runs and butt welds at 2, 4, 6 and 8 mm/s using argon shielding. Electrolytic Tough Pitch Copper was selected as the base metal because of its sensitivity to steam porosity. The argon flow rate was increased to 22 dm^3/min to increase coverage of the plate surface and the current varied from 265-365 A in keeping with producing a full penetration run at each speed.

The weld metal density results for both types of preparation are plotted against speed in Fig. 1. The melt runs remained sound whereas the butt welds showed increasing porosity as the speed rose. In arc welding with a constant partial pressure of reacting gas, the weld metal gas absorption and porosity are known to decrease with increasing welding speed (Refs. 8,9). This suggests that, in the present case, the partial pressure of reacting gas in the arc atmosphere must have been increasing as the welding speed increased. Such an effect could occur as the result of atmosphere entrainment since this would increase with welding speed. The likelihood of entrainment occurring is much greater with a butt weld preparation than with melt runs.

If the porosity in butt welds is due to atmosphere entrainment it is evident that it could be associated with either water vapor or, in view of Kobayashi's (Ref. 5) findings, nitrogen. In other work with Electrolytic Tough Pitch Copper plate (Ref. 8) it was found that, at 4 mm/s, porosity equivalent to that present in the butt welds occurred at a water vapor level of about 150 vpm. The average water vapor pressure in England on a summer afternoon amounts to about 15 millibars (Ref. 10), which represents about 15,000 vpm. Consequently, the level of entrainment must reach about 1% in the arc shroud if water vapor were to be responsible for the porosity. Detailed information about the minimum levels of nitrogen in the shielding gas giving porosity is not available. However, it is significant that in the case of nickel which could be analagous to copper, Pease et al (Ref. 11) observed porosity at nitrogen levels in the arc shroud of 0.1-0.2%.

Nitrogen Addition, Speed and Filler Metal Composition in GTAW

A series of preliminary experiments were conducted with nitrogen shielded arc welding using Oxygen Free High Conductivity (OFHC) Copper in order to confirm the occurrence of porosity. Melt runs were made in 3 mm plate (150 mm × 114 mm) at speeds of 2-8 mm/s, 265-365 A, 11 V and a nitrogen flow rate of 14.25 dm^3/min. Appreciable weld porosity was obtained at all speeds, the extent decreasing (density increasing) with increasing speed, Fig. 2. This would appear to confirm that nitrogen porosity does occur and that it is a typical gas-metal reaction. The morphology of the porosity differed from the random spherical type due to water vapor. There was a clearly defined region of large pores at the surface with fine pores in the weld metal interior, Fig. 3.

In order to examine the small amounts associated with the occurrence of entrainment, the effect of nitrogen additions to the shroud in the range 0-1.0% was determined for melt runs in OFHC copper plate. These tests were made at speeds of 2 and 6 mm/s, 250 and 315 A, 11 V and an argon flow rate of 15 dm^3/min. Accurately calibrated flow meters were used to make up the argon-nitrogen mixtures. The resulting weld metal densities are plotted against nitrogen content of the shielding gas for both speeds in Fig. 4. It can be seen that as little as 0.1% nitrogen added to the shielding gas is enough to produce appreciable porosity at both speeds.

Further addition of nitrogen produced a progressive increase of porosity (density decrease) until a steady level was reached at about 0.6%. For any given level of nitrogen in the argon, increasing the welding speed reduces the amount of porosity. In order to identify the gas responsible for the porosity, weld metal samples were examined at the Welding Institute using a special technique for direct analysis of the gases present in pores. The technique which is described by Jenkins and Coe (Ref. 12) involved fracture of selected pores under high vacuum and analysis of the evolved gas by a small mass spectrometer. Two such analyses carried out on a weld metal sample made using a shielding gas of 0.4%N_2-Argon mixture gave results of 91 and 94% nitrogen. Clearly, such pores could not have been formed by entrapment of the shielding gas but must have been formed by absorption and subsequent rejection of nitrogen from solution in the molten weld pool.

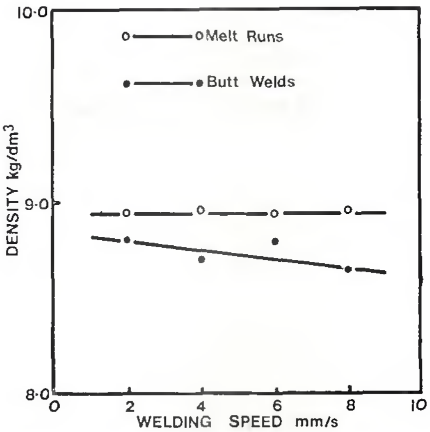

Fig. 1 — The effect of weld preparation and speed on the density of Electrolytic Tough Pitch Copper weld metal

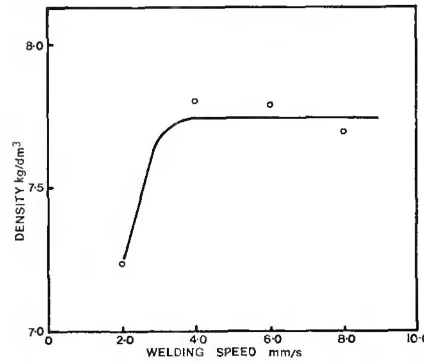

Fig. 2 — The effect of welding speed on the density of Oxygen Free High Conductivity Copper weld metal using commercial purity nitrogen shielding gas

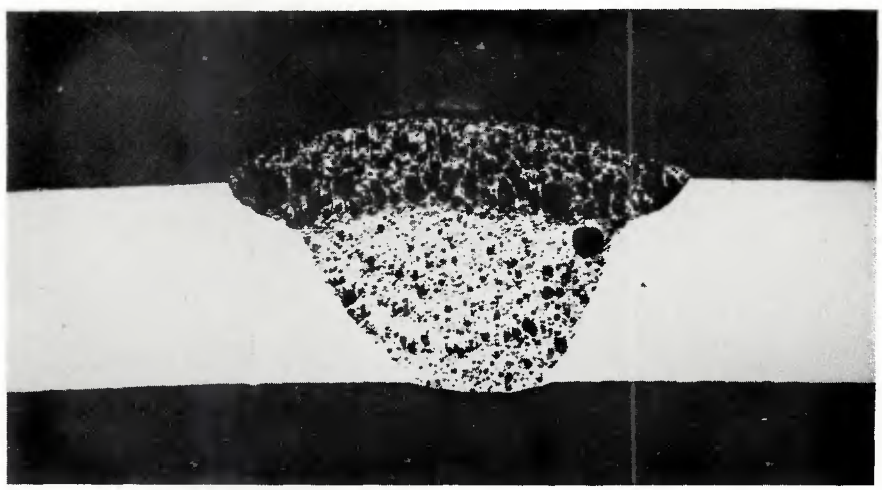

Fig. 3 — Macrograph of OFHC Copper weld metal produced with nitrogen shielding gas showing coarse porosity at surface and fine porosity in the interior. X10, reduced 11%

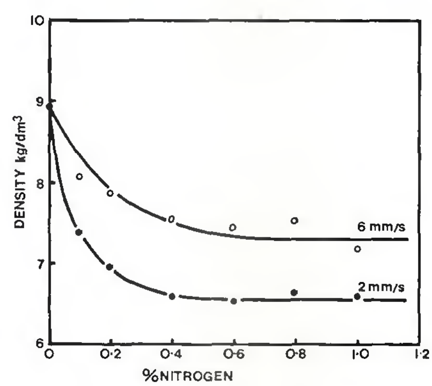

Fig. 4 — The effect on weld metal density of nitrogen content of shielding gas and welding speed in GTA melt runs in Oxygen Free High Conductivity Copper plate

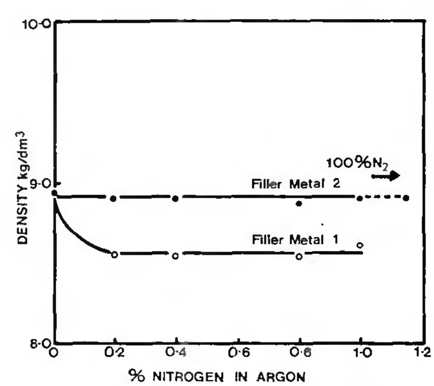

Fig. 5 — The effect on weld metal density of the nitrogen content of the shielding gas in GTA bead-on-plate runs on Oxygen Free High Conductivity Copper plate using filler metals containing (1) 0.18% wt Si, 0.23% wt Mn; (2) 0.20% wt Al, 0.10% wt Ti

The evidence obtained by Kobayashi (Ref. 5) and Moore and Taylor (Ref. 4) showed that porosity was reduced in the presence of elements which were denitriders, although latter authors treated them as deoxidants. In order to investigate this effect in GTA welding, bead-on-plate runs were made on 3 mm OFHC copper plate using two commercially available filler metals (BS 2901 Part 3, C7 and C8) which contained respectively 0.18% wt Si, 0.23% wt Mn and 0.20% wt Al, 0.10% wt Ti.

The first filler metal contained elements which were strong deoxidizers but only mind denitriders, while the second contained elements which were both strong deoxidizers and strong denitriders. Nitrogen was added to the argon over the range 0-1.0% for a total gas flow rate of 15 dm^3/min. The welding conditions were 12 V, 240-273 A and 2 mm/sec. The densities of the resulting weld metal are plotted against nitrogen content of the shielding gas in Fig. 5. It is evident on comparing these results with those in Fig. 4 that the presence of the strong denitriding elements, titanium and aluminium, has entirely eliminated the porosity over the nitrogen addition range 0-1.0%. The second filler metal containing the mild denitrider, silicon, has reduced the porosity but not eliminated it.

Nitrogen Addition, Speed and Filler Metal Composition in GMAW

Bead-on-plate test runs were made on Deoxidized High Phosphorous (DHP) copper plate using the GMA process, electrode positive, for argon-nitrogen shielding gas mixtures containing 0-100% nitrogen. A stationary water cooled torch was mounted over a variable speed carriage on which a test plate 300 mm × 300 mm × 5 mm was placed. The required gas mixtures were obtained by metering appropriate amounts of high purity Argon (99.999%) and high purity nitrogen (99.99%) through accurately calibrated flow meters. The filler metals (1.6 mm diam) were to the same specifications as those used in the GTA experiments (i.e. BS2901, Part 3. C7 and C8) containing respectively 0.18% wt Si, 0.23% wt Mn and 0.20% wt Al, 0.1% wt Ti.

Other welding conditions were 4 mm/s and 8 mm/s, 15 dm^3/min gas flow, and electrode stickout 12.7 mm using 27 V, 330 A for the Si-Mn containing filler metal and 25 V, 270 A for the Ti-Al containing filler metal. The welding conditions gave spray transfer for the argon shielding conditions with both filler metals. However, the spray transfer changed to a coarse "globular" type transfer when the nitrogen content reached 20% for the Si-Mn filler metal and 40% for the Ti-Al filler metal. It was noticeable that the surface appearance of the weld beads of the latter was much rougher than the former.

Experiments were made at 0, 1, 2, 3, 4, 5, 8, 20, 40 and 100% nitrogen additions to the shielding gas with extra runs at 13% and 17% for the Si-Mn filler metal. The density values obtained from the deposits with the Si-Mn containing filler metal are plotted against nitrogen addition in Fig. 6. At 4 mm/s there is an immediate increase in porosity (fall in density) on adding 1% nitrogen and the porosity then increases to a maximum at 8% nitrogen before reaching an intermediate level at 100% nitrogen. Increase of welding speed to 8 mm/s delays the onset of porosity up to the 8% nitrogen level, while at higher nitrogen levels the behavior is similar to that at 4 mm/s.

Radiographic and metallographic examination of the weld metal showed spherical porosity throughout the fused zone with larger pores concentrated at the fusion line, Fig. 7. By contrast the weld beads deposited by the Ti-Al-containing filler metal showed virtually no change in density over the whole range 0-100% nitrogen at both speeds, i.e., zero porosity. Examination of these weld beads by electron probe microanalysis indicated the presence of titanium and aluminium rich areas presumably associated with nitrides. Vacuum fusion gas analysis was carried out on samples of the weld beads with both filler metals. The nitrogen contents were very low, ~0.0002% wt; a level too low for accurate measurement by this technique.

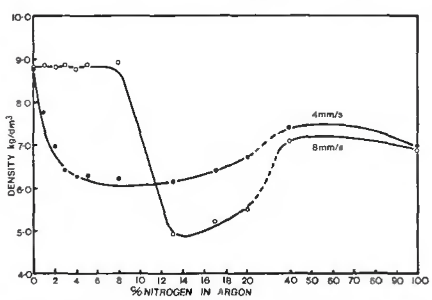

Fig. 6 — Variation of density with nitrogen addition to the shielding gas and welding speed for GMA welded bead-on-plate tests using the Si-Mn containing filler metal

Discussion

Nitrogen Porosity Formation in Copper Weld Metal

This investigation has confirmed that the introduction of nitrogen to the shielding gas in the gas shielding arc processes leads to the formation of gross porosity in the weld metal. The progressive increase in porosity with increasing nitrogen content of the shroud and decreased porosity at increased speed is typical of chemical absorption and desorption of a gas by the weld pool and not physical entrapment of shielding gas. Elimination of the porosity by the strong denitriding elements, aluminium and titanium, but not by strong deoxidants indicates that it is almost certainly due to nitrogen itself and not to trace amounts of the water vapor in the nitrogen. Conclusive evidence for nitrogen absorption and desorption is provided by the fact that pores produced by welding with an argon −0.4% nitrogen gas mixture were found to contain over 90% nitrogen.

In gas tungsten arc welding the threshold nitrogen level to give porosity is less than 0.1% so that the porosity found in butt welds made in OFHC copper plate welded in argon, Fig. 1, must have been due principally to nitrogen entrained from the atmosphere. The equivalent threshold levels for GMA welding are higher due presumably to such factors as the lower arc column temperature compared to GTA welding. In both processes the increase of speed reduced the level of porosity and in the case of GMA welding in-

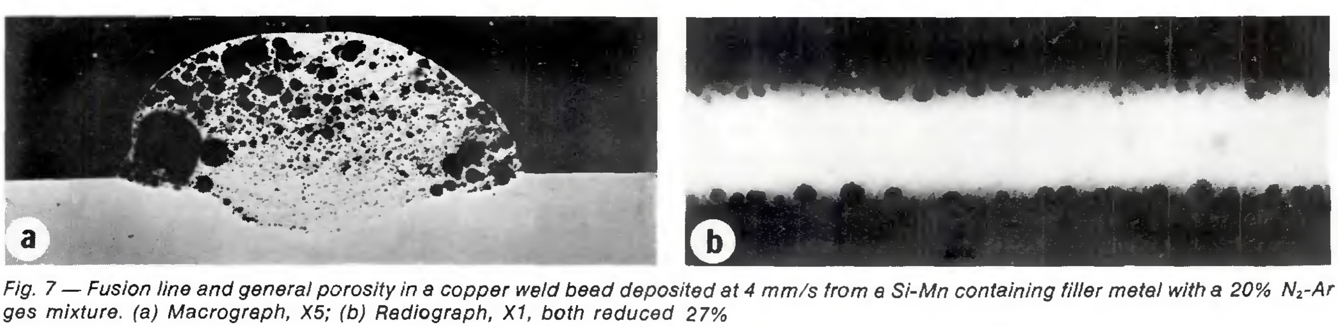

Fig. 7 — Fusion line and general porosity in a copper weld bead deposited at 4 mm/s from a Si-Mn containing filler metal with a 20% N_2-Ar gas mixture. (a) Macrograph, X5; (b) Radiograph, X1, both reduced 27%

creased the threshold level of nitrogen for porosity formation. This is an effect found in other gas-metal systems in arc welding (Refs. 8,9) and can be explained in terms of the reduced time available for gas absorption and release.

Since porosity was eliminated even in 100% nitrogen by small amounts of titanium and aluminium added to the weld pool from the filler metal, the total amount of nitrogen absorbed by the weld pool must have been small. Probably when nitrogen porosity is formed in copper welds the volume of pores corresponds to the total nitrogen content. Some calculations by Littleton (Ref. 13) showed that this was the case on vacuum fusion analysis of a sample GTA welded with a 0.4% nitrogen-argon gas mixture.

Nitrogen-Copper Reaction During Arc Welding

A possible explanation of the reaction of the nitrogen with the copper weld pool is that proposed by one of the authors (Ref. 14) following the publication of Kobayashi's results. This is based on the work of Howden, Salter and Milner (Refs. 9,15) with their concept of a "hot spot" in the weld pool. They maintain that the gas-metal reactions are dominated by the absorption and reaction of gas in a high temperature zone immediately under the arc. Gas absorbed in this zone is then transported by Lorentz forces into the remainder of the weld pool which is at a lower temperature.

Howden and Milner (Ref. 15) from their hydrogen absorption studies calculated a hot spot temperature of 1700 C for copper buttons arc welded at 150 A. While the solubility of nitrogen in copper is negligible at 1400 C it may be appreciable above 1700 C. In which case nitrogen could be absorbed in the reactive zone at these temperatures and then be transported into the remainder of the weld pool where the temperature is such that the solubility is negligible. Nitrogen is then rejected from solution as bubbles to become entrapped as pores in the weld metal. The spherical nature of nitrogen porosity in copper fits in with its having been formed as bubbles in the melt.

However, the explanation may not be so simple. Studies in arc melted iron and iron alloys (Refs. 16,17,18) indicate that nitrogen is being absorbed to levels higher than that predicted by normal solubility relationships involving molecular gas at any feasible "hot spot" temperature. It has been suggested in this case that the reaction involves atomic nitrogen. A similar situation could exist in the case of copper although in the absence of further work this can only remain a suggestion.

Practical Consequences for Welding Copper

Under practical welding conditions the entrainment of the atmosphere into the arc will certainly exceed that encountered in laboratory butt weld experiments and frequently produce nitrogen levels much greater than the 0.1% nitrogen threshold. This means that nitrogen porosity must often occur in GTA welding of copper. The higher threshold levels of nitrogen required for porosity formation in GMA welding suggest that with this process nitrogen porosity may not be such a problem. However, despite the absence of detailed information about entrainment levels for practical welding conditions, it is not difficult to envisage situations in GMA welding when the entrained nitrogen in the arc exceeds 1%. Consequently it seems reasonable to conclude that nitrogen porosity due to entrainment should also occur in production GMA welding of copper.

In the light of these findings it is not surprising that filler metals containing additions of silicon and manganese have not guaranteed weld metal soundness in copper fabrications. The immediate practical significance of this work would appear to be that commercial filler metals for gas shielded arc welding of copper should always contain small amounts of strong denitriding elements. Fortunately denitriders such as titanium are also powerful deoxidants so that any steam porosity will be automatically dealt with at the same time.

Conclusions

1. It has been confirmed that porosity formation can occur in copper weld metal in the gas shielded welding process due to the absorption and subsequent evolution of nitrogen.
2. The levels of nitrogen in the shielding gas to produce appreciable porosity are very low in GTA welding, 0.1%, and can be less than 1.0% in GMA welding so that nitrogen porosity can occur by entrainment of the atmosphere into the arc.
3. The amount of porosity at a particular level of nitrogen in the shielding gas is reduced as the welding speed is increased.
4. Nitrogen porosity can be eliminated by use of filler metals containing small amounts of strong denitriding elements.

Acknowledgments

The authors thank Professor R. H. Thornley, Head of the Department of Production Engineering at the University of Aston for providing laboratory facilities and acknowledge the assistance of many others of their colleagues. The whole program of work was sponsored by a "Grant in Aid" from the International Copper Research Association which the authors thank for permission to publish this paper.

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