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## Porosity and Hydrogen Absorption in Aluminum Welds

*During aluminum arc welding, hydrogen absorption rates are not constant from one alloy to another, resulting in variable weld porosity levels*

BY R. A. WOODS

**ABSTRACT.** The hydrogen absorption and porosity formation characteristics of aluminum and aluminum alloy welds have been studied by determining the density of aluminum weld metal produced with known levels of hydrogen contamination in the arc.

The mechanism of hydrogen absorption in aluminum welding agrees well with the concept of the existence of an active absorbing area beneath the arc which becomes saturated with a large volume of hydrogen. Stirring forces within the pool then distribute this material and the dissolved hydrogen throughout the weld. Changes in welding parameters affect the hydrogen content of the weld metal. The overall effects can be predicted qualitatively by considering the influence of each parameter upon the

geometry of the absorbing zone. However, the effects of any accompanying changes in physical dimensions of the weld pool must also be considered.

Weld porosity levels are controlled by four important factors; the initial hydrogen content of the materials, the rate at which hydrogen is absorbed, the volume of gas which can be held in solid solution in the weld metal, and the rate at which gas bubbles can escape. Weld metal composition was found to be of primary importance in determining both the rate of hydrogen absorption and the hydrogen solid solubility. Hydrogen absorption rates were also influenced strongly by the welding parameters.

Commercial purity aluminum shows a low solid solubility for hydrogen and a high absorption rate. This material is therefore prone to weld porosity formation. Aluminum alloying elements, such as magnesium, can significantly increase the effective solid solubility of hydrogen in the weld metal and lower the rate of hydrogen absorption from the arc. Consequently, Al-Mg alloys are much

less likely to develop weld porosity.

The behavior patterns of each alloy system were similar in both GTA and GMA welding. In general, however, reaction rates were faster during the GMA process. This was partly due to reaction of the molten droplets with hydrogen in the arc and partly to the increased size of the weld pool. In GTA welding, differences in the tungsten electrode changed the level of weld porosity. These electrode differences may account for the sudden unexplained changes in porosity level which sometimes occur in production operations.

### Introduction

Since the early days of aluminum welding, weld porosity has been a recurrent cause for concern. It is generally agreed that hydrogen is responsible for the porosity and that the gas may come from several sources in the form of contamination of various types (Refs. 1-5). By the development of good welding practices and by paying careful attention to

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R. A. WOODS is a Staff Research Metallurgist with Kaiser Aluminum and Chemical Corporation at The Center for Technology, Pleasanton, California.

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such factors as gas quality and proper storage, handling and cleaning procedures for the filler metal and plate, porosity has been minimized. However, it has not been eliminated.

During the last decade, some basic investigations have clarified the mechanisms of gas absorption during the arc melting of various metals (Refs. 6,7). These studies have shown that temperature distributions which are rather unique to arc melting processes can lead to the very rapid pickup of considerable volumes of gas. Porosity in the solidified metal can then result in two ways. Firstly by gas rejection at the advancing solid/liquid interface or, secondly, in cases of much higher contamination, by the nucleation of bubbles ahead of this interface with these bubbles subsequently becoming frozen in the weld (Ref. 8).

The absorption model developed by the earlier authors is one in which gas pickup takes place over an annular area of the pool surface. In the center of this annulus, the pool surface temperature is at or near the boiling point of the metal. Consequently, in this area, gas absorption is low. Outside this very hot central area, gas absorption increases, because the pool surface is cooler and the metal vapor pressure is lower. At increasing distances from the arc center, the pool surface temperatures decrease further. As they do, the gas solubility and therefore gas absorption reach a maximum and then progressively decrease, becoming negligible in regions not directly exposed to the arc.

When the metal in the hot zone absorbs more gas than can be retained in the cooler parts of the molten pool, gas is rejected around the perimeter of the fusion zone. Static arc-melted pools eventually reach a condition in which the rate of gas rejection due to bubbles rising to the surface becomes equal to the rate of gas absorption. The earlier authors were able to show that, under these conditions, the volume of dissociable gas dissolved in the arc-melted pool was directly proportional to the square root of the partial pressure of the gas in the arc. Also, Salter (Ref. 9) showed that the initial rate of hydrogen absorption in steel was a function of the square root of the partial pressure. This is the pattern of behavior which would be expected for a dissociable gas if the system behaved according to Sievert's principle.

However, a most important finding in the earlier work was that the equilibrium gas contents were much higher than were predicted by Sievert's Law for temperatures corresponding to an estimated average temperature throughout the weld pool. This difference was especially

pronounced in the case of aluminum where the metal absorbed large quantities of gas very rapidly. This is consistent with the observation of Lakomskii who, in levitation melting experiments, found that the maximum standard solubility of hydrogen was about 20 ml/100 g of aluminum at 2050 C (Ref. 10). Since pool temperatures of this order are believed to exist in both arc melting and welding, it can be seen how very high weld metal hydrogen contents could arise even if contamination levels in the welding system are quite low.

For an equilibrium condition, Sievert's Law states that:

$$[H] = K\sqrt{P_{H_2}}$$

where:

[H] is the total volume of hydrogen in the liquid metal

K is a constant of proportionality

$P_{H_2}$  is the partial pressure of hydrogen in the arc expressed in atmospheres

Since welding is a dynamic process, equilibrium conditions are unlikely to be established. However, the weld metal hydrogen content and the rate of gas absorption should still be a function of the hydrogen content of the arc. The total volume picked up depends upon the kinetics of the gas absorption process. Although basic gas absorption mechanisms have been proposed for arc melting situations, little information is available to indicate how well these mechanisms apply to arc welding. In particular, very little has been published concerning the role of absorption kinetics in weld porosity. The present study should help to bridge this gap by obtaining porosity and hydrogen absorption data for aluminum and its alloys under arc welding conditions.

In the experimental work, various levels of porosity were produced in aluminum by welding with shielding gases containing known quantities of hydrogen. A major problem which always exists in this type of research is that of retaining as much as possible of the absorbed gas in the weld metal. In the present work, all welding was conducted in the overhead position, thereby eliminating losses due to porosity flotation. Some gas

could have been lost by diffusion from the pool surface and by the breaking of a small number of bubbles forming near the surface. However, in general, these quantities were considered to be minimal.

The final gas content of the weld metal was determined from weld bead density measurements. The total volume of hydrogen in the weld metal consisted of:

1. that present in the plate before the weld was made
2. that absorbed by the weld metal from the arc during welding

Only the latter quantity could be determined from the density measurements. This absorbed gas was present in two forms. A certain quantity was in solution, while the remainder was present as porosity. Although only the porosity volume was measured directly, the volume absorbed and held in solid solution by the weld metal could be obtained by indirect means.

The initial phase of the work consisted of establishing behavior patterns for commercially pure 1100 aluminum. Once this was accomplished, later phases of the investigation were concerned with the effects of welding variables, alloy composition and welding methods.

## Experimental Procedures

The general approach in the present work was to produce weld beads under conditions which (apart from the overhead positioning) closely approximated those that might be used in everyday aluminum welding.

### Gas Tungsten-Arc (GTA) Welds

All welding was automatic overhead dcsp with a 2% thoriated tungsten electrode (EWTh-2) ground to a 60 deg included angle. The standard welding conditions are shown in Table 1. All runs were of the simple autogenous bead-on-plate type. The standard test plate dimensions were 12 x 3 x 0.25 in., so that each weld was about 10 in. long. The plate was thick enough to prevent penetration of the weld bead, so that no gas was lost from the back side of the weld. The arc voltage was not monitored but arc length was accurately controlled and was set with a feeler gauge between the electrode tip and the plate surface prior to each run.

All the shielding gases were dried by a molecular sieve in the gas line, and the gas system was purged before each run to remove any traces of other shielding gases. A control weld was made using 75% helium-25% argon before and after each series of welds to check that contamination

Table 1—Standard GTA Welding Conditions

Current, A	200
Arc length, in.	0.066
Travel speed, ipm	8
Gas flow, cfm	75
Electrode	EWTh-2
Electrode ext., in.	1.1
Cup height, in.	0.5
Cup diam, in.	0.75

within the system was not contributing to porosity. Furthermore, when alloys other than 1100 aluminum were being investigated, a control 1100 plate was run at one gas level to ensure that consistent results were being obtained between one set of data and another.

#### Gas Metal-Arc (GMA) Welds

GMA welds were also produced in the overhead position, but using reverse polarity. The filler metal was taken from (1/16 in. diam) commercial lots. Since the weld beads were somewhat larger than the GTA welds, the plate dimensions were increased to 12 × 4 × 0.5 in. Just prior to cleaning, a groove 3/16 in. deep was machined down the center of each plate and the weld bead was run along this. Welding current was 200 A at a travel speed of 20 ipm. Arc length was held at about 3/16 in. which gave the "transition" arc sound, as judged by an experienced aluminum welding operator. Normally, the welding voltage under these conditions was about 24 V, but again arc length rather than voltage was kept constant.

#### Shielding Gases

Contamination was introduced into the welding system through the use of shielding gases containing known amounts of hydrogen. The gases used were from cylinders of 75% helium-25% argon, to which varying small quantities of hydrogen had been added. Mass spectrometry showed hydrogen contents (% H<sub>2</sub>) that varied, as follows: 0.019, 0.066, 0.096, 0.21, 0.36 and 0.57.

#### Materials

Analyses of plate materials are listed in Table 2. Alloys with commercial designations were taken from production lots. Experimental alloys were produced in the laboratory from small ingots using high purity components.

#### Weld Metal Gas Determinations

Hydrogen contents were determined by measuring weld metal density. The measurements were made by the simple water displacement technique.

A sample approximately 1.5 in. square was cut from a welded plate (Fig. 1) and weighed in air, then in water. Following this, the weld metal was cut away using a small electric cutting tool with a range of small diameter cutters. Care was taken to remove a minimum of base plate and the area was lightly etched during cutting to distinguish between weld and

Table 2 — Composition of Base Plate Materials

Alloy	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti
1100	0.06	0.56	0.13	0.01	0.00	0.001	0.02	0.02
1% Mg	0.05	0.02	0.01	0.00	1.20	0.001	0.02	0.00
6.5% Mg	0.03	0.02	0.00	0.00	6.50	0.00	0.00	0.01
1% Zn	0.01	0.01	0.00	0.00	0.00	0.00	0.95	0.02
6.5% Zn	0.01	0.06	0.00	0.00	0.00	0.00	6.60	0.01
3003	0.29	0.67	0.08	1.13	0.01	0.005	0.02	0.01
5083	0.12	0.21	0.05	0.58	4.58	0.09	0.08	0.02
7039	0.08	0.17	0.02	0.25	2.50	0.19	3.75	0.01
2219	0.09	0.24	5.4	0.31	0.01	0.02	0.04	0.04

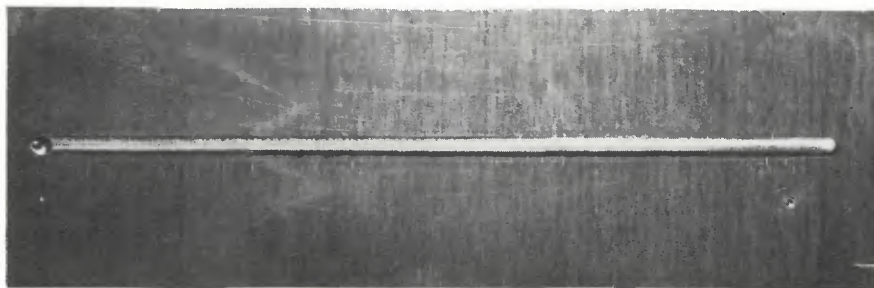


Fig. 1 — Weld bead from which density specimens were taken (X<sup>1/2</sup>)

base metal. When all traces of weld metal had been removed, the remaining plate was reweighed in air and in water. From the difference between the two sets of weighings, the density of the cut-away weld metal could be calculated. Knowing this and the density of sound weld metal (which was established by control runs), the volume of the porosity could be obtained. After correction to N.T.P., this volume was assumed to be that of the measurable hydrogen in the weld.

## Results

#### Hydrogen Absorption in GTA Welded 1100 Aluminum

*Welds at a Constant Arc Hydrogen Level.* The first welds were produced primarily to establish the techniques and reproducibility of the methods. To do this, three identical GTA welds were run on 1100 aluminum plate using a gas contaminated with 0.57% hydrogen. After discarding the bead start and stop areas, the remainder of each plate was cut into samples 1.5 in. long for density analyses. The results showed that the measured gas contents changed markedly as the weight or volume of melted base plate changed along the length of each weld (Fig. 2).

However, an approximately linear relationship was apparent between the gas content and the bead weight (Fig. 3). When plotted on this basis, the results obtained from the three plates were indistinguishable. The trend was for the gas content of the weld metal to decrease as the weight of fused metal increased. This

showed that an equilibrium gas content was not established and also meant that the data were comparable only if taken for the same weight of metal melted or if allowance were made for variations in the size of the fused zone.

*Welds with Varying Levels of Arc Hydrogen.* These welds were produced using the standard operating conditions given in Table 1 and shielding gases from the mixtures detailed earlier. The previous results had shown that comparisons of the porosity produced at various hydrogen contamination levels required weld beads of constant weight per unit length. Since this would have necessitated extremely accurate control of the welding and heat sink parameters, another approach was used. The welds were purposely produced with a wide variation in fused zone size along the length of the bead. This was done by varying the clamping pressure between each weld plate and the heavy backing plates. Then, by taking samples from different locations along each bead, a range of bead weights per unit length of weld was obtained from a single test plate. From this range, the hydrogen content at any intermediate weld bead size could be determined by interpolation.

Several welds were made at each level of hydrogen contamination. The results showed that there was a gradual increase in porosity level with increasing arc hydrogen as can be seen in the macrosections in Fig. 4. The hydrogen contents calculated from the density measurements are shown plotted in Fig. 5 as a function of the

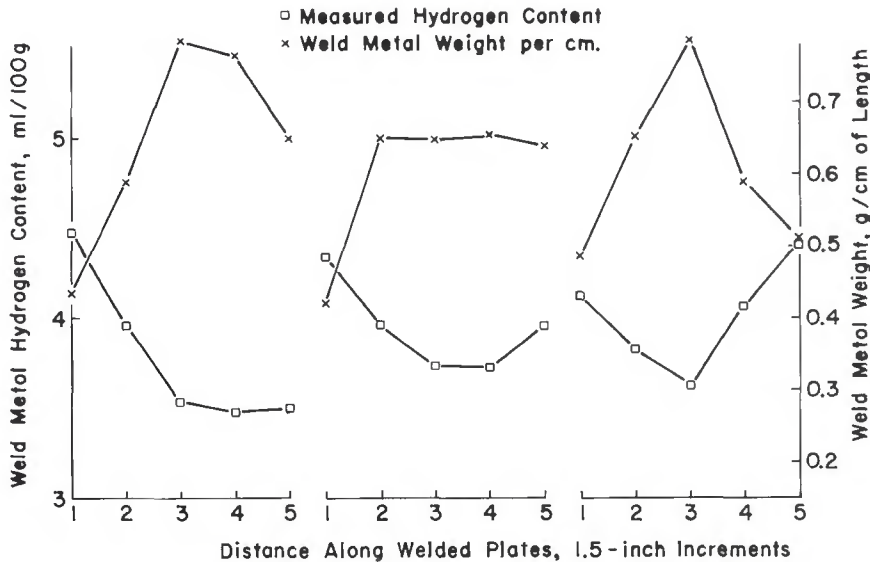


Fig. 2 — Variations in hydrogen content and weight of metal melted along three GTA welded plates

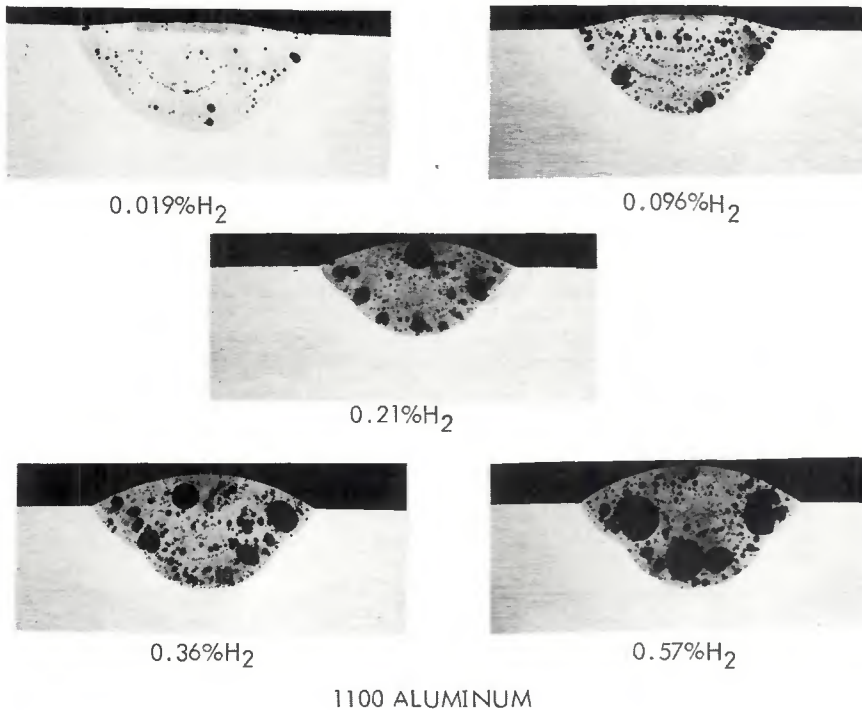


Fig. 4 — Macrographs of GTA welds made in 1100 plate showing increasing porosity with increasing hydrogen contamination (X5, reduced 19%) (½% HF etchant)

weight of metal melted per cm of weld length. From this, an arbitrary weight of 0.6 g/cm of bead length was chosen to compare the hydrogen contents obtained with each gas.

Plotting the values at 0.6 g/cm as a function of the square root of the hydrogen partial pressure in the arc yielded a linear relationship (Fig. 6) as would be expected if the system behaved according to Sievert's principle. The slope of the line gives a value for the constant of proportionality (K) in Sievert's relationship. This, in effect, is a measure of the suscept-

ibility of the alloy to hydrogen porosity. The higher the value of K, the greater will be the rate of increase in weld porosity for a given increase in hydrogen contamination of the arc. However, it should be emphasized that the value of K = 64, determined here, is valid only for the particular alloy, welding conditions and bead weight chosen. Changes in any of the welding parameters, or in the size of weld bead, would change the value of K.

From Fig. 6 it is also possible to determine (a) the threshold arc con-

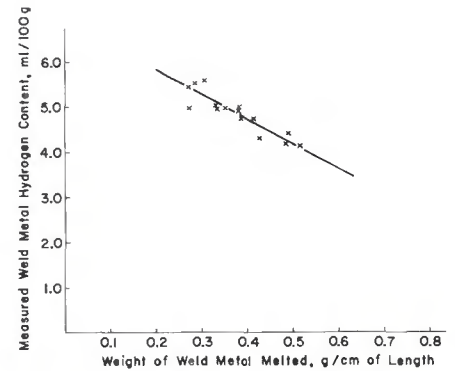


Fig. 3 — The measured weld metal hydrogen content as a function of the bead size

tamination level or hydrogen level at which porosity first appeared and (b) the volume of hydrogen absorbed from the arc and retained effectively in solid solution by the weld metal.

In this instance, the threshold arc hydrogen level was about 0.012% H or 120 parts per million of hydrogen. The volume of hydrogen absorbed from the arc and held in solid solution can be estimated by the extrapolation of the hydrogen content measurements backwards to the zero arc contamination level. The negative intercept of the line upon the hydrogen content axis then indicates the volume of gas absorbed from the arc and held in solid or supersaturated solid solution. The value obtained in this instance was 0.7 ml/100 g. Knowing this value, as well as the amount of gas present as porosity and the welding speed, it was possible to calculate the rate of gas absorption at any of the gas contamination levels used. This calculation for the 0.57% H<sub>2</sub> shielding gas showed that the gas absorption rate first increased as the pool size increased, then reached a constant value of about  $10.7 \times 10^{-3}$  ml/sec as the bead size approached 1.0 g/cm of length (Fig. 7). This may mean that, once the weld pool reaches a certain size, any further size increases do not significantly influence the area over which absorption takes place. The variation in absorption rates again emphasizes the need to consider all results at a constant weld bead size.

#### Hydrogen Absorption in GTA Welded Aluminum Alloys

The next stage in the investigation was to determine how changes in weld metal composition affected hydrogen absorption. By holding all the welding parameters constant, the porosity formation characteristics of various aluminum alloys could be determined and compared to those of 1100 aluminum. Specimens and techniques identical to those used earlier for 1100 aluminum yielded data and

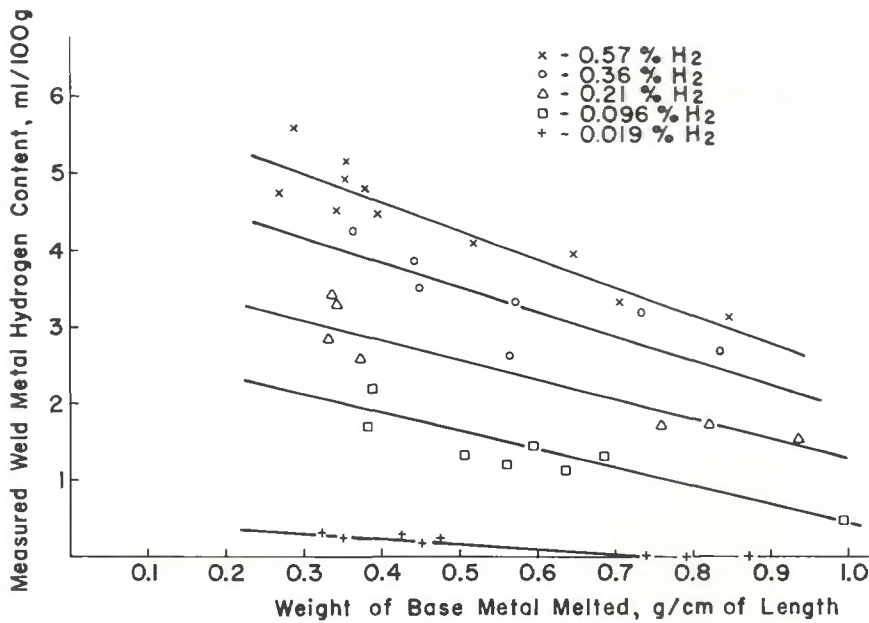


Fig. 5 — Measured hydrogen content of 1100 aluminum GTA weld metal as a function of the bead size and shielding gas composition

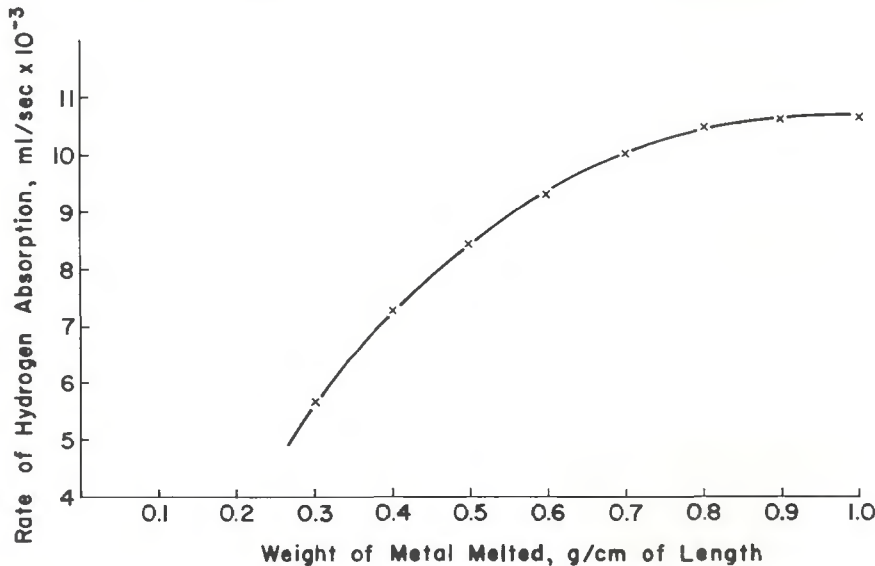


Fig. 7 — The rate of hydrogen absorption as a function of pool size for 1100 aluminum welded with 0.57% arc hydrogen

relationships similar to those shown in Fig. 5. From these, the gas contents of each alloy, at a constant bead weight of 0.6 g/cm, were again expressed in terms of the square root of the partial pressure of hydrogen in the arc.

**Alloys Containing Magnesium.** Commercial 5083 plate and two simple binary magnesium alloys containing 1% and 6.5% Mg were investigated (Table 2). Results with all three alloys conformed well to Sievert's principle. The volume of porosity, however, was considerably lower in each alloy at a given arc contamination level than it had been in 1100 aluminum (Figs. 8, 9) and, in general, as the magnesium content of the alloys increased, the

weld porosity decreased. In all alloys, however, once porosity appeared, its volume increased at about the same rate with a given increase in arc contamination.

The threshold arc contamination level became higher as the magnesium content of the alloys increased. For example, in the 6.5% Mg alloy about 1090 ppm of hydrogen in the arc was required before porosity appeared. In the 1.0% Mg alloy, 625 ppm was sufficient, and in 1100 aluminum it was previously determined that an arc level of only 120 ppm of hydrogen would cause porosity.

Two factors contributed to these differences. Firstly, the absorption co-

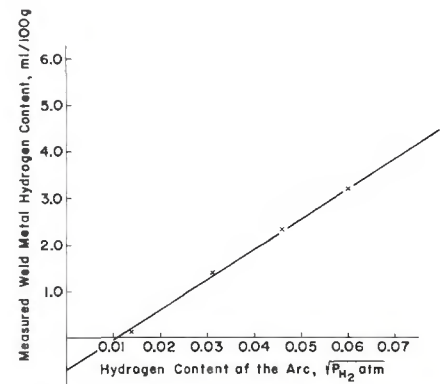


Fig. 6 — The relationship between the weld metal hydrogen content and the square root of the partial pressure of hydrogen in the shielding gas at a bead weight of 0.6 g/cm

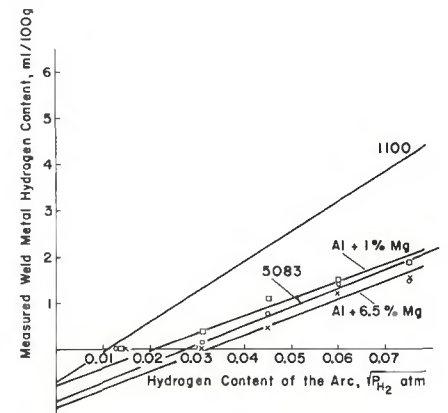


Fig. 8 — The hydrogen content of aluminum-magnesium alloy welds as a function of the square root of the arc hydrogen content

efficient (K) of the Al-Mg alloys was lower (approximately 40 as compared to 64 for 1100 aluminum). Secondly, the magnesium alloys could absorb more hydrogen before porosity appeared. For example, the 6.5% Mg alloy absorbed about 1.25 ml H<sub>2</sub>/100 g and the 1% Mg about 0.75 ml H<sub>2</sub>/100 g before porosity began to form. The 1100 aluminum could only absorb 0.7 ml H<sub>2</sub>/100 g. Therefore, even if the gas absorption rates had been the same, porosity would have appeared at lower arc hydrogen levels in 1100 aluminum than in the Al-Mg alloys.

**Alloys Containing Zinc.** This study included alloy 7039 and two binary zinc alloys containing 1% and 6.5% Zn, respectively. As shown in Fig. 10, the 1% Zn alloy was slightly more prone to porosity formation than 1100 aluminum. The 6.5% zinc alloy, on the other hand, was considerably less so. The marked difference in porosity levels between this alloy and 1100 aluminum can be seen by

comparing the macrostructures in Figs. 11 and 4. The low rate of increase of porosity in the 6.5% zinc alloy corresponded to a low absorption coefficient ( $K = 25$ ) when compared to 1100 aluminum ( $K = 64$ ). Alloy 7039 also showed lower porosity levels than 1100 aluminum, largely because it had very high solid solution capacity (approximately 2.4 ml  $H_2/100$  g). However, once porosity did occur in this alloy, the rate of increase was faster than that in 1100 aluminum, indicating a slightly higher absorption coefficient.

Summarizing, in the binary Al-Zn alloys, the porosity formation tendency increased slightly in the 1% Zn alloy but, following the same trend as in the magnesium alloys, decreased in the more highly alloyed (6.5% Zn) material. However, 7039, which is also highly alloyed (3.75% Zn, 2.5% Mg), showed a surprisingly high rate of porosity increase, although this was offset by a very high solid solution capacity.

**Aluminum Alloys 3003 and 2219.** Alloys 3003 and 2219 were investigated, since both are commonly welded in industry. The results are shown in Fig. 12. The behavior of 3003 was very close to that of 1100 aluminum, although there was a slight increase in both the quantity of arc hydrogen held in solid solution, and the rate of porosity increase.

The copper-containing 2219 alloy showed a low absorption coefficient ( $K = 29$ ) and a low solid solubility limit (0.45 ml  $H_2/100$  g). As a consequence, the porosity formation tendency of this alloy was very low.

#### Influence of Welding Variables

Although chemical composition is a major factor affecting hydrogen absorption and porosity in the weld metal, changes in the welding variables would also be expected to produce an effect. The next step in the investigation dealt with this aspect. A major difficulty was encountered here in that changing any welding parameter almost inevitably led to a change in the size of the fused zone. Since it had been shown earlier that fused zone size changes affected the measured hydrogen contents, a determination of the trends resulting from each parameter change was usually the best that could be achieved. The welding parameters investigated were varied over as wide a range as practicable. During the investigation of each variable, all other conditions were maintained as given in Table 1. Again, the welding was GTA overhead on 1100 aluminum.

**Welding Current.** The influence of current was investigated over a range of 100 to 300 A. The shielding gas contained 0.066% hydrogen. Variations

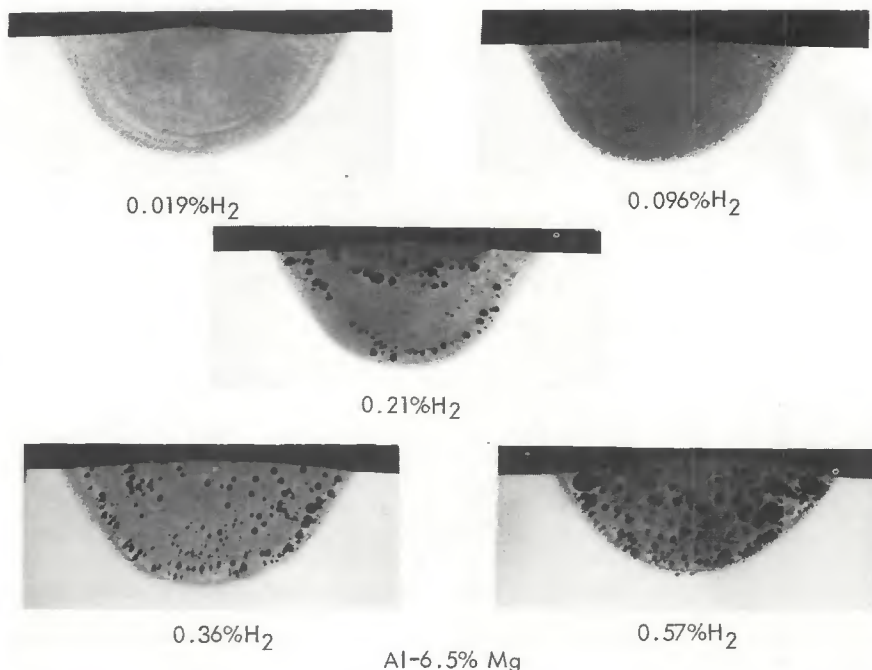


Fig. 9 — GTA welds made in Al-6.5% Mg alloy show increasing porosity levels with increasing arc hydrogen, X5, reduced 22% (1/2% HF etchant)

in the size of the fused zone were minimized by using the combination of plate thicknesses and preheat treatments outlined in Table 3. Since the production of a range of hydrogen content/bead weight specimens for each amperage level was a time-consuming task, the results at each current level were adjusted to a bead weight of 0.6 g/cm. To apply this adjustment, it was assumed that (a) the true 0.066% hydrogen content/bead weight relationship was a straight line falling between the 0.096%  $H_2$  and 0.019%  $H_2$  results shown in Fig. 5, and (b) that the slope of this line would be parallel to the 0.096%  $H_2$  results. Knowing the two quantities  $H_x$  and  $x$ , which were respectively the individual experimental determinations of the hydrogen contents and corresponding bead weights using the 0.066%  $H_2$  shielding gas at each current level, the above assumptions enabled the equation:

$$H_{0.6} = H_x + 2.38x - 1.43$$

to be used in calculating the adjusted gas content  $H_{0.6}$  at a bead wt of 0.6 g/cm for each determination.

The results are shown in Fig. 13. The trend established was one of increasing porosity as the welding current was increased. Since it was determined earlier that welds in 1100 aluminum can retain about 0.7 ml  $H_2/100$  g of weld metal in solid

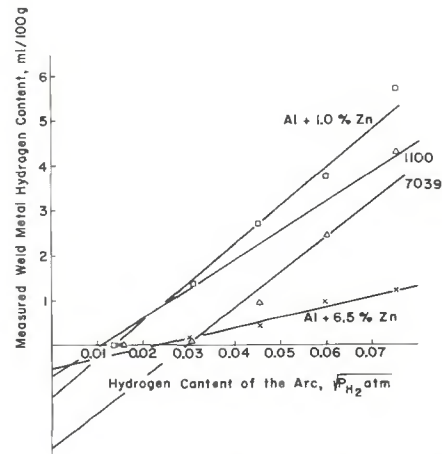


Fig. 10 — Weld metal hydrogen contents in aluminum-zinc alloys as a function of the square root of arc hydrogen

solution, the hydrogen absorption rates could also be calculated and, as would be expected, these followed the same trend. Following the absorption model outlined earlier, it is proposed that increasing the current raises the average surface temperature of the weld pool, enlarging the area of the hot annular zone over which the hydrogen absorption takes place. This results in an overall increase in the rate of gas pickup as current increases.

**Arc Length.** In this study the arc length was varied between 0.020 and 0.070 in. As mentioned earlier, arc length was set with a feeler gauge be-

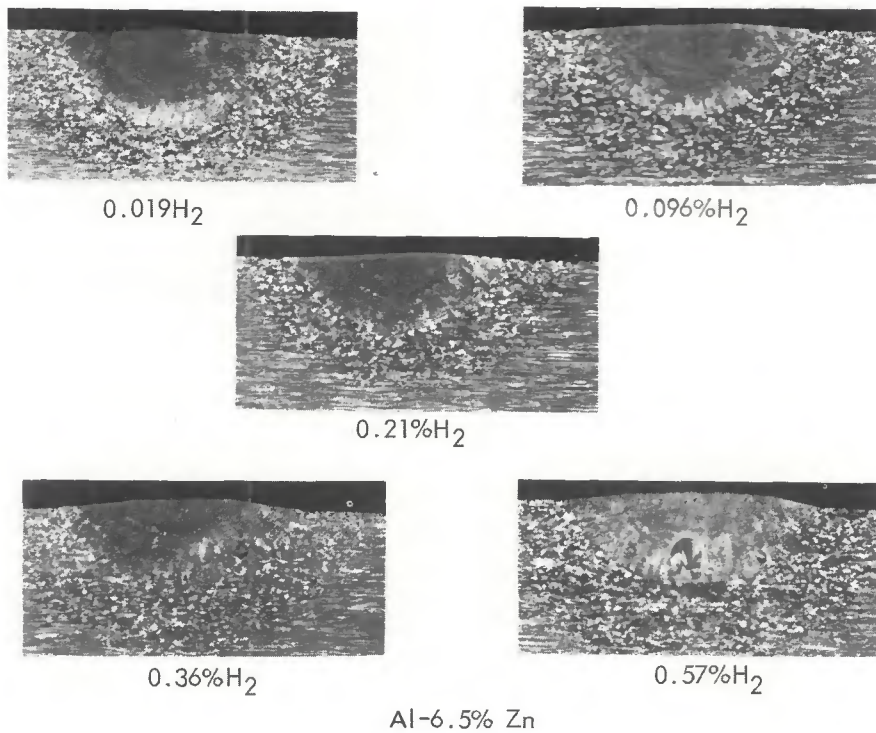


Fig. 11 — Increasing porosity in Al-7.5% Zn alloy welds made with hydrogen contaminated shielding gases. X5, reduced 15% (1/2% HF etchant)

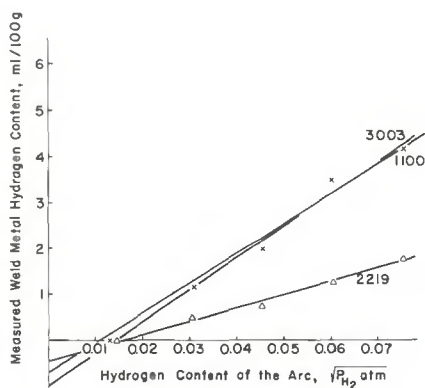


Fig. 12 — Hydrogen contents in 3003 alloy and 2219 alloy as a function of the square root of the arc hydrogen

fore each run was started and the electrode-to-plate distance was not adjusted during the run. The true arc lengths, therefore, would have been somewhat longer than those measured, due to the depressing action of the arc forces on the pool surface.

The weld bead weight per unit length was about 0.4 g/cm over the range of arc lengths investigated, so no correction was made for variations in pool size (Fig. 14).

The measured hydrogen contents showed a general trend of increasing gas pickup as the arc lengthened. This trend was probably due to the modification of the surface temperature profile across the pool. As the arc length increases, a greater area of the pool surface is exposed to the arc, and the area of the annular absorbing zone becomes greater. In

general, therefore, longer arcs would be expected to cause more weld porosity than short arcs.

**Travel Speed.** Up to this point, the standard travel speed throughout the experimental work had been 8 ipm. To examine the effect of this as a variable, additional tests were run at 12 and 20 ipm holding all other parameters constant. Test plates used for the 20 ipm set of results were preheated to 520 F before welding, in order to produce weld beads of comparable size to those produced at the slower welding speeds.

The data in Fig. 15 show that at each travel speed, the gas contents increased linearly in accordance with the square root of the partial pressure law. However, as the travel speed increased, the weld porosity at any given arc contamination level decreased. This latter effect may be interpreted as a progressive lowering in the value of the absorption coefficient, K. The volume of gas held in solid solution by the 1100 aluminum remained constant but the threshold value for the onset of porosity was lowered. This means that more highly contaminated welding systems can be tolerated at higher travel speeds. Calculations of the rate of hydrogen absorption, again making allowances for the volume of gas held in solid solution, demonstrated that the rate of gas absorption was unaffected by a change in travel speed (Fig. 15b). Thus, the reduction in porosity caused by going faster was simply a consequence of the reduced time available for gas pickup.

Table 3 — Preheat and Plate Thickness Combinations Used to Investigate the Effect of Current on Absorption

Current, A	Plate thickness, in.	Preheat, F
100	3/16	500
150	1/4	500
200	1/4	None
250	1/4	None
275	1/2	None

**Shielding Gas Flow Rate.** Another variable investigated was the shielding gas flow rate. For these tests, the shielding gas used contained 0.066% hydrogen while welding conditions other than the gas flow rate were as shown in Table 1. Two sets of results produced under identical conditions over the range 30 to 105 cubic feet per hour are shown in Fig. 16. The difference in porosity levels between one set and another was caused by a small change in the volume of the weld beads. Indications were that weld porosity levels and gas absorption rates were slightly reduced at high gas flow rates, possibly because of a gas cooling effect on either the arc or the pool surface.

**The Effect of Electrode Changing and Regrinding.** This phase of the work was undertaken to investigate whether changes at the electrode tip could affect porosity levels. Such changes in the tip might be brought about either by arc wear, by mechanically redressing the tip or by changing from one electrode to another. All the electrodes used were taken from a batch of 2% thoriated tungsten rods which were ground to an included tip angle of 60 deg  $\pm$  2 deg. Welds were produced using 0.57% H<sub>2</sub> gas and standard operating conditions. The results are shown in Fig. 17. A series of five consecutive welds and another of ten welds were made with a new electrode at the beginning of each series. Two more series of five welds each were made using the same electrode throughout each series but regrinding the tip between each run.

Each set of welds showed a wide range of inter-run absorption rates. This occurred whether the electrode was reground or whether the same tip was used again. There appeared to be a general trend toward higher absorption after prolonged use of the same electrode. The effect, however, is difficult to explain on the basis of tip aging, since two sets of data where the electrodes were reground between each run also showed the same trends. Presumably, the differences between runs are due to slight changes in arc structure each

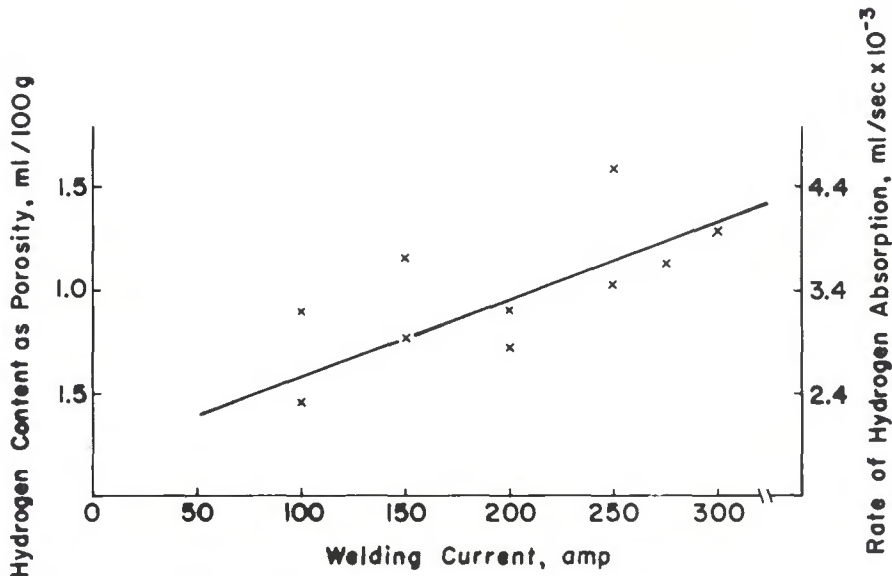


Fig. 13 — The effect of welding current on weld porosity and the rate of hydrogen absorption in 1100 aluminum

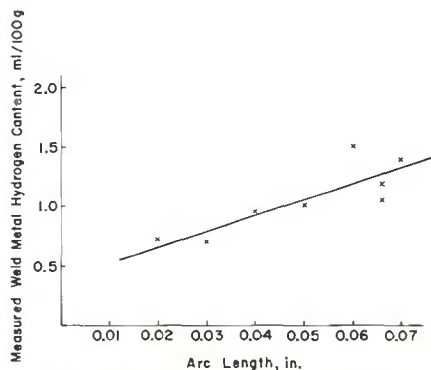


Fig. 14 — The effect of arc length upon the volume of hydrogen porosity in the weld

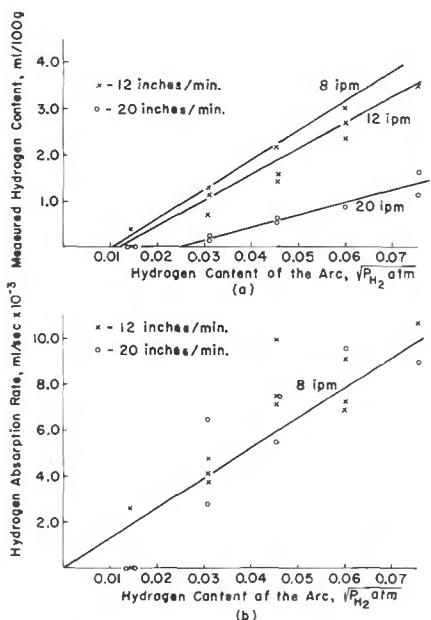


Fig. 15 — Weld metal hydrogen content and hydrogen absorption rate as a function of travel speed

time the arc is initiated. Overall, the individual run variations could be quite marked. These variations undoubtedly contributed some scatter to the results reported in this work, and they could also lead to variable porosity levels in commercial weldments.

#### Analysis of Gas Metal-Arc Welds

In the next phase of the study, the primary objective was to determine whether the trends established for GTA welds also held true for GMA welds. Trials were made with various filler metal/base metal combinations and hydrogen contaminated gases in the consumable metal-arc system. These results were somewhat more difficult to compare on an absolute basis, since it is well known that filler metal surface contamination can contribute significantly to the weld metal hydrogen content in the consumable metal-arc system. No electrode cleaning procedures were employed, but each alloy combination was checked by a control run prior to each welding series to insure that there was no porosity.

Bead sizes were in the region of 1.5 g/cm in all alloy combinations used in this phase of the work. As a consequence, when plotting the results, no corrections were made for size variations. The results are shown in Fig. 18.

As in GTA welding, 1100 aluminum was the alloy most sensitive to porosity formation. The volume of gas absorbed by the metal before porosity appeared was approximately the same as in the GTA work (0.7 ml H<sub>2</sub>/100 g). However, the rate of increase in porosity with increasing gas

contamination levels was much higher than in the GTA welds produced at the same welding speed.

One possible factor contributing to this was reaction between the molten electrode droplet and the shielding gas. The contribution of the droplet to the final weld metal gas content was estimated by comparing the hydrogen content of droplets obtained from arcs struck in pure 75% helium-25% argon shielding gas to that of droplets from an arc contaminated with 0.57% hydrogen.

Samples were obtained by shutting off an established arc and collecting the drops which were frozen on the wire tip. A metallographic examination showed that there was more porosity in those drops obtained using the hydrogen-contaminated gas. Hydrogen analyses by sub-fusion vacuum extraction indicated that the drops produced in the pure shielding gas contained about 1.1 ml H<sub>2</sub>/100 g. Those produced in the contaminated gas contained 2.3 ml H<sub>2</sub>/100 g. The difference of 1.2 ml H<sub>2</sub>/100 g was attributed to absorption from the shielding gas at the electrode tip.

The average weight of a drop was  $7.8 \times 10^{-2}$  g and from high speed movies it was estimated that drops existed for  $4 \times 10^{-3}$  sec on the electrode and  $2 \times 10^{-3}$  sec in the arc. If the same reaction rate is assumed while the drop is in flight across the arc as at the wire tip, the drop entering the pool might be expected to contain about 1.8 ml H<sub>2</sub>/100 g of metal. Measurements showed that the electrode contributed about 75% towards the final bead weight so that the filler metal contribution to the pool hydrogen would have been about 1.4 ml H<sub>2</sub>/100 g of weld bead.

Earlier gas content measurements showed that at the 0.57% arc hydrogen level, the weld metal absorbs about 4.4 ml H<sub>2</sub>/100 g. Therefore, the molten droplet reactions in the 1100 filler metal/1100 base plate system contributed approximately one-third of the final gas content of the pool. Subtracting the droplet contribution leaves a figure of 3.0 ml H<sub>2</sub>/100 g for the weld pool/arc reaction which is still 50% higher than for GTA welding at 20 ipm. This leads to the conclusion that hydrogen pickup is more rapid in GMA weld pools than in GTA pools. Factors contributing to this may have been the longer arcs used in the GMA work, the effect of the polarity reversal, or the increased surface area of the weld pools.

Considering the other filler metal/base metal systems, all showed lower porosity levels than the 1100/1100 system. This resulted from a combination of lower absorption coefficients and the absorption of larger volumes of gas before porosity



appeared. The least improvement was shown by 4043 and 2319 alloy electrodes on 1100 base plates. This latter result was unexpected since the GTA welds on 2219 plates had shown a very low hydrogen absorptivity. The use of magnesium-containing 5183 wire and magnesium and zinc in 5039 electrodes both resulted in a lowering of porosity levels. The same effects were found when 1100 filler metal was welded on 5083 plate. In fact, the hydrogen levels were slightly lower than those in welds made with 5183 electrodes on 1100 plate. Increasing the magnesium content of the weld metal by welding 5183 filler on 5083 plate did not reduce porosity further. These results again followed the GTA trends in that:

1. even a small percentage of magnesium in the weld markedly lowered the porosity content; and

2. no extra benefit was obtained by increasing the magnesium content to higher levels.

This behavior was confirmed in welds using 4043 filler metal on 6061 alloy plate. With 25% plate dilution, the weld metal probably did not contain much over 0.25% Mg. Yet the threshold arc hydrogen contamination level was of the same order as that of other magnesium-containing alloys, the solid solubility was about 1 ml H<sub>2</sub>/100 g of metal and the absorption coefficient (K) was slightly lower than for the other alloys. These properties resulted in the lowest porosity-forming characteristics of all the GMA welding combinations tested.

### Discussion

The behavior of the aluminum/hydrogen system during welding was

generally consistent with the pattern expected from the model proposed for arc melting by previous investigators. The present results show how the general principles of this type of gas/metal reaction may be applied to the problem of porosity formation during arc welding. An important finding was that although each aluminum alloy behaved according to the principles of Sievert's Law, each reacted differently to a given level of welding arc contamination.

The mechanism of absorption is one in which hydrogen becomes dissociated in the arc atmosphere, entering the molten weld metal in atomic form. During both GTA and GMA welding, the volume of gas absorbed by the weld pool was directly proportional to the square root of the hydrogen partial pressure in the arc. However, the volume of gas absorbed was

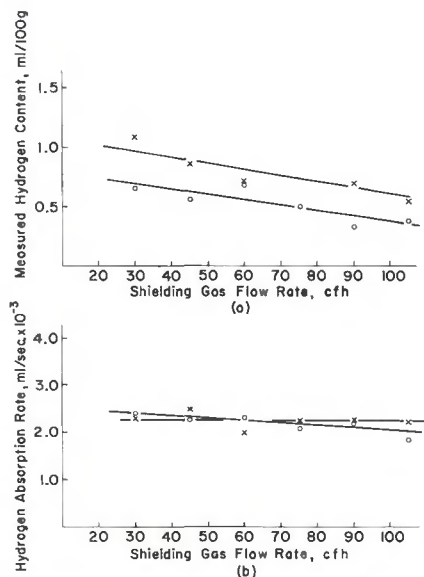


Fig. 16 — The effect of shielding gas flow rate upon weld porosity level and the rate of gas absorption

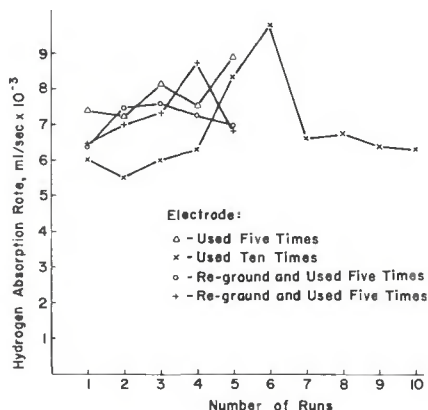


Fig. 17 — Hydrogen absorption rate variation with tungsten electrode usage

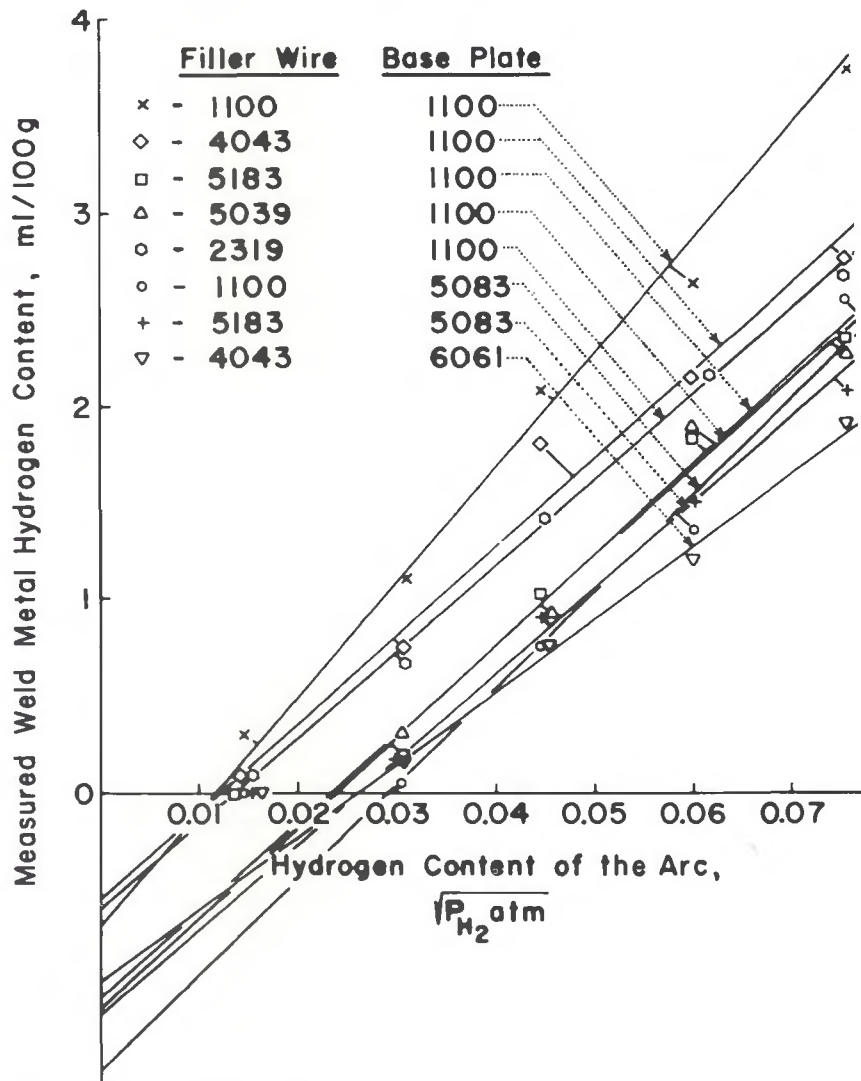


Fig. 18 — GMA weld metal hydrogen contents as a function of the square root of the arc hydrogen

also affected by the welding conditions through changes in the absorption rate.

The total hydrogen in the weld metal is composed of the hydrogen in the materials before the weld was made and the hydrogen introduced during welding. Upon solidification, a certain volume of the hydrogen is retained effectively in solid or supersaturated solid solution in the weld metal. The remainder of the gas is potentially available to form porosity in the weld bead. In the present work, a situation existed in which no gas was lost from the weld by porosity flotation. It follows from what has been established that at a specific level of contamination in a given welding system, the weld porosity in an alloy will be determined by:

1. The initial material hydrogen contents
2. The effective solid solubility of hydrogen in the metal
3. The value of the hydrogen absorption coefficient pertaining to the particular welding and arc contamination conditions

Therefore, in terms of these considerations alone, an aluminum alloy showing minimum weld porosity would be one which is well degassed, has a high hydrogen solid solubility and a low hydrogen absorption coefficient.

In the present work, two components were obtained: (a) the volume of gas absorbed from the arc which was retained in solid or supersaturated solid solution in the weld metal, and (b) the volume of gas present as porosity. Together, these two quantities represent the total volume of hydrogen absorbed from the arc. The unknown quantity was the volume of gas present in the materials before welding began. This is important since it would be expected to influence the threshold porosity level. If it is assumed that any particular welded alloy can hold only a specific volume of hydrogen in solid or supersaturated solution in the weld metal, then the higher the initial plate hydrogen concentration, the lower the level of arc contamination that can be tolerated.

Some preliminary evidence supporting this concept was obtained using an experimental cast of 5182 alloy which was degassed to two levels during pouring. Sub-fusion vacuum analyses of samples of unwelded plate fabricated from these castings yielded hydrogen contents of 0.35 ml/100 g and 0.08 ml/100 g, i.e., a difference of 0.27 ml/100 g in gas contents. During overhead GTA welding tests following the standard procedures used in most of this work, differences in gas contents between

Table 4 — Absorption Rates and Solid Solubilities of Various Alloy Systems

Alloy	Absorption rate <sup>a</sup> ml/sec × 10 <sup>-3</sup> GTA Welding	Solid solubility, ml/100 g
1100	2.63	0.7
Al + 1% Mg	2.64	0.75
Al + 6.5% Mg	1.58	1.25
5083	1.62	1.1
Al + 1% Zn	3.45	1.15
Al + 6.5% Zn	1.93	0.5
7039	3.15	2.3
3003	2.78	0.95
2219	1.16	0.45
	GMA Welding	
1100/1100	15.23	0.7
5183/5083	11.22	1.2
4043/6061	9.55	1.0

(a) The absorption rate was calculated for an arc contamination level of 400 ppm hydrogen.

welds on the two plates were determined by means of weld metal densities. These differences were found to be 0.31, 0.34 and 0.23 ml H<sub>2</sub>/100 g at three different arc contamination levels — figures which agree closely with the difference in the vacuum extraction measurements on the plate materials before welding. This illustrates the importance of using well degassed plate.

The data also indicated that the plate containing 0.08 ml H<sub>2</sub>/100 g could be welded at an arc contamination level of almost 1500 ppm hydrogen before weld porosity appeared. The badly degassed plate could only tolerate just over 1100 ppm hydrogen in the arc. The basic reason for this difference was that the well degassed plate could absorb and hold about 1.9 ml H<sub>2</sub>/100 g of arc hydrogen in solid solution, while the other plate could only absorb about 1.6 ml/100 g of hydrogen before porosity appeared. Thus, the total volume of gas held in solid solution was almost 2.0 ml H<sub>2</sub>/100 g in this particular alloy.

At this stage it is not clear in what form this much gas is retained by the weld metal. For example, in the case of 1100 aluminum, the volume of arc hydrogen held in solid solution was about 0.7 ml/100 g of weld metal, a value which is in good agreement with that predicted by Nikiforov (Ref. 11). However, this volume is much greater than the accepted solid solubility level of hydrogen in pure aluminum of 0.04 ml/100 g as measured by Ransley (Ref. 12). Whether the values obtained represent true solid solubility levels, or whether there is considerable supersaturation, is not known.

However, it is noteworthy that for 1100 aluminum, the apparent solid solubility was independent of travel

speed which implies an independence of the rate of solidification. Solid solubility estimates for some of the other alloys investigated are summarized in Table 4. These are probably conservative figures since the initial plate hydrogen contents were not known. Solid solubility was also independent of the weld method (i.e., GTA or GMA) and the welding variables, although it appeared to be markedly dependent upon the alloy composition. Most alloys were able to hold more hydrogen in solid or supersaturated solid solution than pure aluminum. But, again, without knowing the initial plate hydrogen contents, it is not possible to state by how much the total solid solubility values of hydrogen in the alloys differed.

However, the differences encountered in the present work were large enough to make significant differences to weld porosity levels in the various alloy systems, showing that the capacity of the alloy to retain hydrogen in some form in the solid metal is an important factor in determining the final weld porosity level.

With a given solid solution value in any alloy, the level of arc contamination at which porosity first appeared is determined by the absorptivity. This parameter, which is a measure of the rate of increase in gas absorption with increasing arc hydrogen content, is difficult to evaluate fully. This is mainly because of its dependence upon such variables as the weld metal composition, welding parameters, arc geometry and physical factors, such as the size of the weld pool. In general, the effect of each parameter change upon absorption could be predicted from its expected effect upon the area of absorption.

Absorption was faster in the GMA

system than in the GTA. This was due partly to parameter and physical differences between the two welding systems and partly to the extra gas contributed by the hydrogen/molten droplet reaction. Values for the rate of gas absorption in some GMA alloy systems as well as GTA welded alloys are given in Table 4. These values were all calculated assuming an arc contamination level of 400 ppm hydrogen.

It is difficult to determine any definite trends in these results apart from the marked decrease in absorption rate in the aluminum-magnesium alloys. The reasons why magnesium acts as it does are not fully understood, although alloys exhibiting a high vapor pressure might be expected to show reduced absorptivities owing to a modification of the active absorbing area. However, this trend was not followed by the 7039 alloy or by the 1% Zn alloy, both of which showed high absorptivity. In contrast, 2219 alloy which contains low vapor pressure elements, showed a low absorptivity.

Although the present observations on the effect of welding parameters upon hydrogen absorption during the arc welding of aluminum are consistent with the concept of an annular zone over which absorption occurs, there are some basic differences which must be considered when applying this model to welding. This is especially true of overhead welding procedures such as were used in this present investigation.

As described previously, the small volume of very hot, highly reactive metal in the active surface zone of the weld pool picks up a high level of gas. Turbulence within the pool then transports metal from this zone into cooler parts of the pool. Continued stirring and diffusion within the pool tend to produce a more even distribution of gas throughout the liquid. If some metal containing more than the solid solubility limit of hydrogen should solidify, porosity will form.

In metal which does not solidify, the cooling may be sufficient to cause the metal to become saturated. Should it become supersaturated by more than a factor of 1.2 times its one atmosphere solubility, bubbles may form spontaneously ahead of the freezing interface (Ref. 13). In the overhead position, these bubbles will rise to the top of the weld bead (in this case, the root) where they will become entrapped. The net effect is that the liquid metal has lowered its dissolved hydrogen content. Continuing turbulence eventually sweeps this hydrogen-depleted metal back to the pool surface again where more hydrogen is absorbed from the arc. The cycle is then repeated and with this

continuous absorption and rejection, there is no theoretical limit to the quantity of gas that may be "pumped" into the weld metal. Practical limitations are imposed by the welding parameters and, ultimately, by the physical ability of the comparatively small volume of liquid metal in the weld pool to retain the large volume of gas which can be released.

In flat or any other position welding, there is more opportunity for bubbles to escape, although a similar absorption mechanism would be operative. Consequently, as in arc melting, a dynamic equilibrium could be established, whereby the rate of gas evolution through escaping bubbles would equal the rate of gas absorption under the arc. Thus, a limit would be set to the level of porosity which could be reached. The extent of the retained porosity depends upon welding position and welding variables.

It is common practice in flat-position aluminum welding to reduce weld porosity by increasing the welding current or reducing the travel speed — changes which, in view of the present results, would be expected to increase the volume of hydrogen picked up from the arc. For in general terms, the work reported here shows that hydrogen absorption — and therefore porosity levels in aluminum welding — should be minimized by using a low current, a short arc, a high gas flow rate and fast travel speed. However, associated with reductions in travel speed and increases in current are reductions in the rate of solidification. A change of this nature is believed to allow more time for gas bubbles to form and float out of the weld.

In addition to this effect, the above parameter changes would also create a larger volume of liquid metal in the weld pool. If the increased gas absorption, caused, for example, by the increased current, were insufficient to compensate for the increase in pool volume, then a lower average concentration of dissolved hydrogen would result. This, in turn, would lead to the formation of fewer bubbles. This reduction in numbers, together with the improved opportunity for bubble growth and escape, would be expected to lead to lower levels of hydrogen porosity in the weld. Thus, the reduction in weld porosity levels caused by changes in travel speed or current changes may be due to a combination of effects rather than to a simple pore flotation mechanism.

The effect of the tungsten electrode upon gas absorption during aluminum welding was surprisingly great. However, an electrode effect has been noted before in arc melting experiments (Ref. 7). As pointed out earlier,

the changes in absorption were attributed to changes in the arc or anode spot structures each time an arc was initiated and there was generally no way of predicting whether porosity would increase, decrease or remain unchanged. This effect may account for some of the sudden unexplained changes in GTA weld quality which are sometimes encountered in the field. Until there is a better understanding of what causes these changes, the only solution would seem to be the rather unsatisfactory one of attempting to lower the weld porosity by regrinding the electrode.

Although the basic model of hydrogen absorption processes during arc welding seems to be adequately established, this study illustrates the complexity of factors which must be considered in determining causes and solutions of porosity problems in aluminum welding. It also illustrates how welding conditions and cleanliness precautions which are satisfactory for welding one aluminum alloy may produce welds which contain unacceptable porosity levels in another.

Deterioration in weld quality could also occur after changing from one batch of plate to another if the second plate batch has been less well degassed. The additional hydrogen present in this plate might then be sufficient to bring the total weld hydrogen above the threshold level. The same effect might also be encountered in going from one batch of filler metal to another. It is perhaps fortunate that 1100 aluminum, the alloy most difficult to weld without forming porosity, is seldom used where high integrity, high strength joints are required. Welds in alloys such as 5083, which are subject to much more stringent inspection requirements, can be porosity-free even if welded under comparatively highly contaminated conditions.

## Conclusions

1. The rate and volume of hydrogen absorbed from the arc atmosphere by aluminum weld metal is in direct proportion to the square root of the hydrogen partial pressure in the arc.

2. The porosity level may vary widely in welds made under identical conditions in different aluminum alloys. The reasons for this are that:

- a. The effective solid solubility of hydrogen in each alloy is different.
- b. The hydrogen absorptivity of the weld metal varies.

There may also be differences in the volume of hydrogen contained in the unwelded material which will affect the volume of hydrogen which can

be absorbed from the arc before porosity appears.

3. In any given alloy the rate of hydrogen absorption is increased by increasing the arc current, arc length, and weld pool surface area. The rate of absorption is unaffected by changes in travel speed and essentially so by changes in shielding gas flow rate.

4. The rate of hydrogen absorption is higher during GMA welding than during GTA welding of aluminum.

5. The balance between the rate of hydrogen absorption and the volume of molten metal in the weld pool is critical in determining the level of porosity in a weld.

6. The mechanism of absorption of hydrogen by aluminum under arc welding conditions is consistent with the basic model of gas absorption over an annular zone of the weld pool surface.

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