An Investigation on the Effects of Gases in GTA Welding of a Wrought AZ80 Magnesium Alloy

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ABSTRACT. Magnesium alloy components are frequently gas tungsten arc welded despite magnesium's high thermal diffusivity. Gases such as argon, helium, and nitrogen—enriched or not with hydrogen—have been investigated to determine if melting, and in particular weld penetration, can be increased. Images of the arcs, voltage readings, dimensions, defects, and microstructure of weld fusion zones have been examined. Due to a greater first ionization potential, helium increased the constant-current voltage and created more melting than argon. With diatomic gases such as nitrogen and hydrogen, voltage and weld dimensions were even further increased. However, hydrogen caused porosity, and nitrogen interacted with magnesium by leaving a nitride deposit at weld surfaces. While consequences of alloying with nitrogen were probably not disadvantageous, hydrogen pores were of greater concern. Both welding parameters and hydrogen concentration in the arc atmosphere were important in controlling porosity. The two-dimensional heat-flow conditions of the weld pool surface. Knowing that surface tension of the molten metal depends upon its partial pressure in the arc atmosphere and the temperatures of GTA weld pools must approach the material's boiling temperature (1090 °C for magnesium and 900 °C for magnesium-hydrogen phase diagram), the cathodic clean-up of the magnesium alloy melt is essential. However, compared to direct current where the electrode is negative and operates as a cathode, alternating current lowers the heat input to the base material and produces shallower welds, especially when argon is selected over helium (Refs. 6–8). Here, the addition of a diatomic gas, like hydrogen (H₂) or nitrogen (N₂), to a monatomic gas, like argon (Ar) or helium (He), is proposed to increase the melting of magnesium alloys during GTA welding.

While hydrogen and nitrogen have been safely introduced to argon or helium for increasing weld penetration in transition-metals alloys, their use has not been reported for joining magnesium. The magnesium-hydrogen phase diagram shows that hydrogen solubility in magnesium decreases by about 25 wt.% when magnesium solidifies at 650°C. This characteristic alone indicates that hydrogen accumulates rapidly at the solid/liquid interface, and if hydrogen partial pressure is sufficient, gaseous bubbles will form. In general, porosity appears in welds when the ratio of solubility at the average pool temperature and at the melting temperature (either in the liquid phase or solid phase) is high. The concentration of a gas absorbed by the weld pool depends upon its partial pressure in the arc atmosphere and the temperatures at the weld pool surface. Knowing that surface temperatures of GTA weld pools must approach the material's boiling temperature (1090 °C for magnesium and 900 °C for magnesium-hydrogen phase diagram), Sievert's law can be applied to estimate the propensity of hydrogen to generate porosity. For liquid magnesium (Ref. 10), Sievert's law can be represented as:

\[
[H] = 608 \cdot p^{1/2} \left( \frac{24,400}{RT} \right)
\]

where \([H]\) is the total volume of hydrogen in the liquid metal in mL/100 g, \(p\) is the partial pressure of hydrogen in the arc expressed in atmospheres, \(R\) is the universal gas constant (8.31 J mol⁻¹ K⁻¹), and \(T\) is molten metal temperature (in kelvins). Equation 1 predicts that the equilibrium solubility at 1-atm hydrogen pressure for 1090 °C is 30.4 and 35.3 mL/100 g, respectively, as validated by independent measurements by Fromageau et al.
values of solubility will be affected by weld vicinity of 19 mL/100 g. Although all these
temperature, being about 25% less, is in the
ity in solid magnesium at the melting tem-
Equation 1 shows that the pressure re-
drogen potency in porosity formation.
rium conditions are valuable to assess hy-
bubbles to form, which requires that the
pool cools from 1090 °C to 650°C. For gas
sion is (70.6/25.3)^2 = 7.8 atm when the
amount of dissolved hydrogen exceeds the
calculation shows that the minimum concen-
tration of hydrogen to induce porosity is
comparing data from various investigators
(Refs. 12, 13). However, no research to
hydrogen partial pressure is smaller by
This prediction that hydrogen potency to
produce pores in magnesium is far less
than in aluminum has been verified by
physiological unbound atoms (Refs. 7, 18-20).
precedes the thermal ionization of the
gases. The graph at the top-left corner is a
semilog representation of the electrical
conductivity of several pure ionized gases as a function of temperature (pressure is
one atmosphere) (Refs. 7, 19).

Physical Properties of Gases

In GTA welding, the voltage estab-
lished between the nonconsumable tung-
sten electrode and the base material is
normally self-regulated to deliver a con-
stant current. With the current being ap-
proximately invariant, Ohm’s law indi-
cates that voltage and electrical conduc-
tance of the arc are inversely pro-
portional. The conductance of the arc is
a function of its dimensions, particularly the
distance between the electrode and the
base material (i.e., the arc length), as well as the
local values of its electrical conduc-
tivity. As a simple definition, the electric
arc is a sustained discharged plasma with
physical properties that relate mainly to
the ionization of its gaseous species. In the
arc, electrical conductivity is greatest
where electrons are generated at low tem-
peratures. A monatomic gas, thus a gas
only consisting of unbound atoms in their
gaseous state, has a high electrical con-
ductance when its first ionization poten-
tial (i.e., the energy barrier to release a
first electron) is small. The lower the gas
first ionization potential, the smaller is the
voltage necessary to produce a given cur-
rent, and the smaller are the energy and
thus the average arc temperature. Figure
1 (Refs. 7, 16) demonstrates that voltage
across a GTA arc decreases noticeably
when gases with a small first ionization po-
tential, as found going down the periodic
table, are selected. Helium, because of its
high first ionization potential (24.6 eV),
conducts the current least and conse-
quently requires the greatest voltage of all
monatomic gases to carry a given current.
Comparatively, arc voltages with argon
and particularly xenon are smaller as ex-
plained by the smaller first ionization po-
tential of these two monatomic gases (15.6
eV for argon and 12.1 eV for xenon). Since
the current carrying capability of the arc
depends upon its electron population, it-
selves dependent upon the temperatures es-
ablished within the arc, any properties
that would affect arc temperatures are rel-
vant to GTA welding. Thermal conduc-
tivity and specific heat are therefore two
important properties that must be consid-
ered to understand the effects of the vari-
ous gases selected for this study.

For monatomic gases, electrical conduc-
tivity, heat capacity, and thermal conduc-
tivity all increase when first ionization
tential decreases (Refs. 7, 16). For di-
omatomic gases, the physical properties are
strongly affected by the dissociation that
precedes the thermal ionization of the
gaseous unbound atoms (Refs. 7, 18-20).
The dissociation energy of diatomic nitro-
gen is greatest (9.8 eV), followed by that
of diatomic oxygen (5.1 eV) and diatomic
hydrogen (4.5 eV) (Refs. 7, 19). Figure 2
(Refs. 7, 20) compares physical properties
of argon, helium, and several diatomic
gases. The graph at the top-left corner is a
semilog representation of the electrical
conductivity as a function of temperature.
It is seen that electrical conductivity of
argon is considerably greater than that of helium at temperatures less than about 3000 K. This is a direct consequence of argon's smaller first ionization potential. However, at higher temperatures, the population of charged particles (i.e., positive ions and electrons) rapidly levels off in both gases (Refs. 7, 16) and the electrical conductivities become comparable, at least until secondary ionization takes place. Compared to argon, diatomic nitrogen and diatomic hydrogen exhibit smaller electrical conductivities. Beyond 2000 K, their conductivities are also less than that of helium. Of the properties seen in Fig. 2, enthalpy and thermal conductivity of diatomic gases differ most significantly from those of argon and helium. The top-right corner graph reveals that enthalpies of the diatomic gases are considerably greater than those of argon or helium. As a direct consequence, raising the temperature of these gases will require more energy than with either argon or helium, regardless of the extents of the gas ionization. Diatomic gases also extract thermal energy more efficiently, as well-depicted by the lower graph. The fact that thermal conductivity of hydrogen and oxygen rapidly decrease after 4000 K indicates that a large fraction of the diatomic molecules have already split into single atoms. Due to a substantially greater dissociation energy (9.8 eV), electrical conductivity of nitrogen does not drop before 7000 K. All the property differences depicted in Fig. 2 between argon, helium, nitrogen, and hydrogen will influence GTA welding, particularly the voltage necessary to stabilize the arc, the heat that this arc generates, and thus the heat input to the base material.

Experimental Procedure

To study the effects of nonatomic and diatomic gases using a limited number of experiments, only five gases were selected. They were commercially pure argon, argon with 1% and 6% hydrogen, helium with 1% hydrogen, and nitrogen with 1% hydrogen. With the first three gases, the influences of hydrogen additions could be investigated. Further, considering 1% hydrogen, the effects of argon, helium, and nitrogen could be compared. However, a direct comparison of the two diatomic gases, hydrogen and nitrogen, could not be made. In addition to five gases, three currents (30, 40, and 50 A) and four arc lengths (0.5, 1.0, 2.0, and 4.0 mm) were used, leaving a total of 60 bead-on-plate welding experiments and, with two cross sections per weld bead, 120 cross sections to examine.

Bead-on-plate gas tungsten arc (GTA) welding was completed using constant current conditions, a travel speed of 100 mm/min, and a gas flow rate of 40 L/min. A set of 2-mm-diameter tungsten electrodes was prepared with a typical tip angle of 60 deg. When tip wear was observed, the tungsten cathodes were replaced to guarantee consistent and comparable experimental results. A digital multimeter with a precision of 0.01 V and a digital camera with a dark filter were utilized to investigate contributions of the various gases on the arcs. Visual characteristics of the arc as well as voltages, measured at the welding power supply, were captured. This voltage, although not measured directly across the arc, is closely related to the arc voltage, and was therefore sufficient to compare the various gases quantitatively. Later in the paper, this voltage is referred to as welding voltage.

All the welding specimens were fabricated from a single 180-mm-diameter extruded AZ80 magnesium alloy plain cylinder. The selection of a wrought alloy as opposed to a cast alloy guaranteed that the initial material did not contain any pores, in particular hydrogen pores. The AZ80 alloy of this study had approximately 8.50 wt-% aluminum and 0.60 wt-% zinc, as estimated using glow discharge spectroscopy. In this investigation, the extruded cylinder was sliced into 3.8-mm-thick circular plates, which were then ground to exhibit smooth and shiny surfaces. Depending upon welding parameters and gases, partial and fully penetrating welds were obtained. On each plate, six bead-on-plate welds were deposited parallel to each other. No fixtures were applied during welding, as the plates were simply positioned on top of a flat and bulky aluminum support, which also eliminated the need for a secondary root shielding.

After bead-on-plate welding, the specimens were cross-sectioned, ground, and polished with silicon-carbide papers and diamond pastes (6 and 1 μm), etched with 2% nital, and examined with a stereomicroscope at magnifications from 10 to 40 x. Dimensions of fusion zones (penetration and width) were measured and porosity (if encountered) was quantified using conventional image analyses based on area measurements. Optical microscopy, scanning electron microscopy, electron dispersive spectroscopy (EDS), X-ray diffraction (XRD), and thermodynamic calculations were also used to complement the analyses of the weld fusion zones.

Results and Discussion

Arc Morphology

Figure 3 displays digital images of the arc produced using various combinations of gases and arc lengths, as the arcs were viewed from the side. For the purpose of photographing the arc, the welding current of 30 A was preferred. Of the three selected currents, this current generated
the least heat, and therefore minimized the brightness of the arc on the captured images. With a few exceptions, the images exhibited good contrast and they were clear enough for analyses.

By first examining the images of Fig. 3 from top to bottom, the influence of each gas can be examined. The color of the arc was found to vary with the composition of the gases. Since temperatures, radiation wavelength, and color are related, as proven by emission spectroscopy, the colors captured on the images were important features to examine. Behind the dark filter, the arcs with pure argon were yellow. The additions of hydrogen caused the yellow arcs of argon to become increasingly green. The substitution of argon by helium and nitrogen produced entirely green arcs. The arcs were also considerably larger. A comparison of the various argon arcs reveals that hydrogen did not broaden the arc, as could have been expected from the arc colors observed with helium and nitrogen. On the contrary, the argon-hydrogen arcs appeared as if they were constricted, which was particularly apparent by examining the brightest region of the 2.0- and 4.0-mm arcs. Based upon this observation, the possibility was raised that fusion zone penetration might be improved by the use of hydrogen additions, likely through an increase in current density and thus Lorenz electromagnetic force.

The observation that arcs were narrower in the presence of hydrogen is consistent with Fig. 2, where both thermal conductivity and specific heat are significantly increased by hydrogen, especially at the low temperatures where hydrogen is still in its diatomic form. In the arc periphery, where temperatures are lower than in the arc central region, hydrogen is therefore most effective in restraining the arc from expanding. In the case of the helium enriched with 1% hydrogen, the arcs did not appear to be as bright as the argon arcs. The arcs were also distinctively deflected in the direction opposite to the displacement of the tungsten electrode. This arc deflection was most evident when the arcs were longer, as seen by comparing the pictures of Fig. 3 from left to right.

Compared to the other gases, the arcs produced with the nitrogen-rich gas were noticeably different. Figure 3 does not indicate that these arcs were deflected. On the contrary, they were quite symmetrical, and therefore the direction of electrode displacement, thus arc deflection, could not be determined, as was found with the other gases. As shown in Fig. 3, the digital images of the arcs with the nitrogen-rich gas were also all blurry irrespective of the welding parameters. These hazy images were attributed to the heavy fumes encountered when welding was conducted with this diatomic gas. Figure 3 shows that the images corresponding to arc lengths of 0.5 and 1.0 mm exhibited the brightest features. This observation could be indicative of exceptionally high arc temperatures, and most probably high welding voltages, since these two characteristics are largely related, as explained previously.

Voltage

Figure 4 shows five voltage-current lines, all constructed from average voltage readings with the different gases and all accompanied with a least-square root equation. The error bars indicate that voltage readings were dispersed within a range that represented about 10% of their values. Due to scatter, the voltage measurements made with the different gases occasionally overlapped. However, the data points could still be properly fit using five distinct straight lines. These five separate lines are proof that electrical conductance (defined as the ratio of current over voltage) for the various arcs differed noticeably. Also, the fact that the five lines were parallel (with a constant slope of 0.04 V/A) indicates that voltage and current varied at a constant rate, and this rate did not depend on the gas.

Regardless of the current, the arcs with pure argon exhibited the smallest voltages, followed by the arcs with argon containing 1% and 6% hydrogen, respectively. On average, the addition of 1% hydrogen increased voltage by about 15% over that of argon, and the addition of 6% hydrogen doubled this voltage increase. The helium enriched with 1% hydrogen increased the voltage even further, leading to an average 25% increase in voltage over the 1% hydrogen-enriched argon gas, and about 40% over the pure argon gas. With nitrogen as a substitute to argon and hydrogen concentration still at 1%, the voltage was increased by 45% beyond that of

![Fig. 4 — Voltage vs. current characteristics for the five gases (arc length: 0.5 mm).](image1)

![Fig. 5 — Welding voltage vs. arc length for the five selected gases (current: 30 A).](image2)
the 1% hydrogen-enriched argon and about 65% over that of the pure argon gas. The dramatic increase in voltage observed with nitrogen compared to argon was well substantiated by Fig. 2, where enthalpy and thermal conductivity of the nitrogen are considerably greater than those of argon and helium. These characteristics suggest that the elevated temperatures needed to ionize and thus create an electrically conductive arc gap require more heat generation with nitrogen. Such a heat generation, and thus heat input to the material, was confirmed by the voltage readings, the fumes observed during welding (which explained the hazy images seen in Fig. 3 with nitrogen), and the bright spots revealed in the same images.

Figure 5 directly complements the set of photographs of Fig. 3 by also describing the effects of both arc length and gas. In agreement with Fig. 3, the voltage (thus the heat generated by the arc) increased with the arc length. As in Fig. 4, the voltage was minimum with argon and increased with the other gases, as given by the sequence shown from top to bottom in Fig. 3. Figure 5 also demonstrates that voltage reached about 21 V when electrode and base material were 4 mm apart, and when 1% hydrogen was added to nitrogen. When examining the effects of arc length, note that voltage appeared to increase at a greater rate in the presence of short arcs.

With both Figs. 4 and 5, the effects of hydrogen could be further analyzed. In both figures, the argon gas containing 1% hydrogen is located about halfway between the pure argon and the argon with 6% hydrogen. This result indicates that the first 1% hydrogen was about as effective in increasing the voltage as the 6% hydrogen that was subsequently added. Consequently, the arcs produced with hydrogen were conceivably sustained with both a limited dissociation of the diatomic molecules and a slight ionization of the hydrogen atoms. If that is the case, the major contribution of hydrogen would be to cool the arc central region, which would be made possible by increasing thermal conductivity and specific heat of the arc atmosphere.

Weld Fusion Zone Morphology

Figure 6 reproduces optical micrographs of five bead-on-plate welds for which the arcs were presented in the last column of Fig. 3. The welds gathered in Fig. 6 have striking differences. First, starting with the weld made with argon, the presence of transverse striations at its surface as well as frequent lateral deviations suggest that the heat input to the base material was unstable and insufficient to form a stable and penetrating weld pool. In contrast, the introduction of 1% hydrogen to argon clearly improved arc stability, as the welds were straight and rather indistinguishable from beginning to end. Since the striations at the weld surface were still present, melting also probably occurred within a shallow depth under the surface. However, with an addition of 6% hydrogen, the welds were distinctly wider, as expected from the voltage measurements presented previously. With 6% hydrogen, spherical-like protrusions, perhaps representing subsurface pores caused by the rejection of hydrogen during solidification, also became visible. When helium was replaced by argon, the widths of the welds were further increased. The surface of the welds was still smooth, a condition that signaled the presence of lesser defects. If that was indeed the case, the 1% hydrogen added to helium could be advantageous. When nitrogen was selected as replacement to argon or helium, weld surface looked noticeably different. As seen in Fig. 6, the coloration of the nitrogen-rich welds was darker. Closer examination by scanning electron microscopy revealed that an agglomerate of fine particles, as shown by the inset image, covered the entire weld surface. Such proof of chemical interactions between the weld fusion zone and nitrogen will be further developed later where the bulk of the fusion zone will be examined in great detail.

Relationships between penetration and width of fusion zones were searched to further characterize the morphology of the produced bead-on-plate welds. As shown in Fig. 7, widths, as seen in Fig. 6, and penetration of the fusion zones were mathematically related irrespective of the gases. Since dimensions of fusion zones were at a first estimate interdependent, the results of Figs. 6 and 7 infer that fusion zone penetration with the hydrogen additions was greater than those made with pure argon, and the same ranking of gases seen for the
Figures 8 and 9 specifically show how fusion zone penetration with the different gases varied as a function of arc length and current. Despite scatter caused by measurements on different cross sections, the scatter was small enough to guarantee excellent comparisons. Both Figs. 8 and 9 appear to substantiate the ranking earlier established with the voltage readings, except that the helium with 1% hydrogen gas produced slightly greater voltages than the 1% enriched nitrogen gas. As found for the voltage, fusion zone penetration increased with both arc length and current. In contrast with voltage readings, however, the slopes for each line seen in Figs. 8 and 9 varied differently from one gas to the other. The argon line exhibited the smallest slope, followed by that of the argon with 1% hydrogen and that of the argon with 6% hydrogen. The lines for the helium and nitrogen gases were noticeably steeper, although precise values for their slopes could not be determined from the two data points only available. Figures 4, 7, and 8, among others, clearly reveal that enlargements of the fusion zones are to some extents related to the voltage readings. Despite significant scatter, Fig. 10 confirms that such a correlation between fusion zone penetration and voltage existed. Figure 10 not only describes the effects of both voltage and current on fusion zone penetration, it also clearly shows that differentiating between gases becomes practically irrelevant when the voltage is considered.

Figures 8-10 also reveal that the fusion zones left by the nitrogen enriched with 1% hydrogen were slightly smaller than those anticipated from the voltage readings — Figs. 4, 5. An explanation can perhaps be found by considering the distinct arc profiles seen in Fig. 3 for the nitrogen-rich gas. With nitrogen as a substitute for argon or helium, both the arcs and the fusion zone penetrations indicate that more heat must have been dissipated into the surrounding of the arc. In other words, nitrogen likely created less energy density than the other gases, although overall heat generated within the arc was greater, as indicated by voltage. This possibility is also strongly suggested by Fig. 2, which indicates that nitrogen had the greatest electrical conductivity of all the gases, in addition to a particularly high enthalpy.

Similar to the effects of nitrogen, the effects of the hydrogen concentration added to argon could also be further examined. As mentioned earlier for the voltage, the line for the 1% hydrogen argon gas was located approximately halfway between the lines of the pure argon and the argon with 6% hydrogen. Figures 8 and 9 indicate that the contribution of 1% hydrogen on the fusion zone dimensions was closer to that of pure argon than to that of the argon with 6% hydrogen. This result, which might appear counterintuitive at first, could also simply suggest that the heat flow had changed, as voltage started to exceed some critical values from which heat flow quickly changed from three-dimensional to two-dimensional (Refs. 6, 22, 23). Figure 10 confirms that fusion zone penetration and voltage were not proportional to each other, but that fusion zone penetration appeared to increase at a greater rate with the voltage.

Weld-Gas Interactions

The cross-sectional views of Fig. 11 demonstrate that hydrogen is a source of porosity, as theorized previously, whereas nitrogen created weld fusion zones with two distinct regions, which for convenience were designated as FZ1 and FZ2. The first region, FZ1, possessed the normal composition and microstructure of welds made with the monatomic gases, and therefore will not be described in this article. However, the second region found in the upper part of the weld fusion zone, FZ2, revealed an unusual microstructure, as already detected in Fig. 6. In this section, interactions between the AZ80 magnesium alloy and gases like hydrogen and nitrogen are discussed.

Effects of Hydrogen

As suggested in Fig. 6 with the weld produced with 6% hydrogen, Fig. 11 confirms that large spherical pores were encountered when hydrogen was present in the arc atmosphere. However, based upon Figs. 6 and 11, the size and perhaps amount and concentration of pores appeared to be influenced by the concentration of hydrogen, the gas that is mixed with hydrogen, as well as the current that is selected. In this section, the influence of welding parameters is therefore pri-
mary to discuss. Since porosity is the result of a nucleation and growth process depending upon mass transport and solubility of hydrogen (Ref. 21), the effects of process parameters that affect hydrogen transport, particularly the arc length and the current, were studied. Further, relations with weld fusion zone morphology, already discussed, were searched.

In Fig. 12, the concentration of hydrogen pores (defined in percent of fusion zone area) at currents of 30 and 40 A is graphically represented as a function of the arc length for each of the hydrogen-enriched gases. Due to variations between weld cross sections, the data were considerably dispersed. While the contribution of current to porosity could not be established using the two currents shown in Fig. 12, a clear correlation between pore concentration, arc length, and hydrogen concentration was found. Figure 12 shows that porosity decreased when the arc length was increased and when hydrogen concentration was reduced.

To understand this last result, some of the results already presented must be reviewed. First, recall that voltage stepped (Ref. 22, 23), the heat flow could be categorized as two-dimensional. A consequence of two-dimensional heat flow is that cooling rates are typically one to two orders of magnitude smaller (Refs. 4, 6, 8, 22, 23). Based upon Fig. 12, in welds associated to two-dimensional heat flows, the hydrogen dissolved in the liquid magnesium had sufficient time to diffuse out of the pool and effectively reduce porosity. The concentration of pores was considerably different when the arc length was reduced. The fusion zone then partially penetrated into the base material (Fig. 9) and established three-dimensional heat flow conditions, where cooling is comparatively faster.

To address the mechanisms of pore formation, pore diameters were measured, and pore size distributions were established for various welding parameters and the 6% hydrogen gas. Figure 13 shows that the nucleation and growth of pores within the liquid pool did not appear to be measurably restricted by the faster cooling of partially penetrating welds. In fact, Fig. 13 shows that the shallow fusion zones produced with the 0.5-mm arcs and the current of 30 A, where heat flow was three-dimensional, included more pores of any size than any other welds considered in Fig. 13. In contrast, population and size of pores were considerably less when the heat flow was two-dimensional, as promoted by the 4-mm arc length. Regardless of welding parameters, Fig. 13 reveals that the population in any given pores gradually decreases as their average diameter increases. This trend demonstrates that there was always sufficient hydrogen to nucleate small hydrogen pores regardless of the welding parameters. With both large currents and extended arc lengths (i.e., two-dimensional heat flow conditions), the larger pores were eliminated. Porosity was practically eliminated, not because nucleation and growth of pores were restricted, but because hydrogen could leave the weld pool.

Effects of Nitrogen

Figure 14A-D depicts optical and secondary electron images with EDS and XRD results for the region in Fig. 11 that has been designated as FZ2. This new region demonstrates that a measurable amount of nitrogen had permanently entered the weld, most particularly its upper part, FZ2. In one of two reasonable explanations, nitrogen interacted with the weld pool well before its dimensions reached that of the fusion zone. The presence of nitrogen at the weld pool surface, either in solution or as a nitride layer covering the pool, could explain why hydrogen did not induce porosity. In this situation, the contribution of nitrogen would have been to prevent hydrogen to enter the weld pool. Alternatively, nitrogen could have been dissolved in the entire fusion zone, implying that during solidification, hydrogen would have been gradually rejected by the fast-growing magnesium solid phase. In this second explanation, FZ2 would have formed the last, after the material in FZ1 would have solidified without being effectively influenced by hydrogen or nitrogen. To vali-
date this second explanation, hydrogen would have been dissolved in restricted amounts in the magnesium liquid phase. Reduced hydrogen solubility, thus low hydrogen partial pressure for pore nucleation, could well explain the absence of pores in all the fusion zones made with nitrogen. Although a complete understanding of the interaction between the magnesium weld pool, nitrogen, and hydrogen might be out of reach given the initial purposes of this study, the data of Fig. 14 are worth examining to better understand weld formation with the nitrogen-rich gas.

Figure 14A and B clearly shows that FZ2 exhibited an uncharacteristic banded microstructure. Such microstructures are as intriguing and puzzling as the presence of the two regions, FZ1 and FZ2. The banded microstructure of FZ2 indicates that nitrogen had entered the weld pool, and had possibly partitioned from a magnesium-rich phase to periodically reach concentrations large enough to stabilize at least one new nitrogen-rich phase. If that is the case, the rejection of nitrogen from a magnesium-rich phase would demonstrate that this magnesium alloy has no solubility for nitrogen. Thermodynamically, this would mean that enthalpy of mixing for a solution comprising magnesium, its alloying elements, and nitrogen is strongly positive, at least when the solute composition is far from that of the new nitrogen-rich phase. If the microstructures of Fig. 14A and B are the result of normal partitioning, the new nitrogen-rich phase must have formed after the magnesium-rich phase, as would be seen in a eutectic transformation. In the presence of a eutectic transformation, the temperatures where the alloyed region, FZ2, would be liquid would be lower than that of the AZ80 alloy (i.e., the material of FZ1). The presence of a lower melting point eutectic mixture, although not validated by other measurements or a complete binary phase diagram (unavailable), is in agreement with the observation that FZ2 is located near the weld upper surface, where it would be expected if it had solidified at last.

However, the fact that nitrogen accumulated along concentric bands, well following the solid/liquid interface, also appears to be atypical of eutectic decompositions, and this last explanation may therefore be invalid. Although no data are provided to support this discussion, the unusual phase morphology seen in Fig. 14A and B may be better explained if capillarity is considered. Rejection of nitrogen from magnesium-rich growing dendrites and transport of nitrogen along the dendrite boundary toward the dendrite tips (driven by capillarity) would be a satisfactory explanation for the banded microstructures of Fig. 14A and B, if supporting data were available. In that particular situation, nitrogen would constantly advance at the same time as the solid/liquid interface until the nitrogen buildup would become enough for a new phase to form and magnesium to continue solidifying beyond the magnesium nitride phase.

Figure 14C presents a typical energy dispersive spectrum for the darker phase seen in the banded microstructure of Fig. 14A and B. The inset secondary electron image, accompanying the energy spectrum, reveals that this new phase is less conductive than magnesium (because brighter in the SEM), and also hard and brittle, as revealed by its granular appearance. The EDS measurements revealed that nitrogen constituted 40 at.-% of this brittle-looking phase, whereas nitrogen was not present in the magnesium-rich phase, as was suspected from previous analysis. These measurements not only confirm that magnesium and nitrogen did not mix, but that nitrogen and magnesium formed a stoichiometric compound with 40 at.-% nitrogen. Measurements by XRD, presented in Fig. 14D, identified this phase as the normal magnesium nitride phase, Mg3N2. Compared to other nitrides, magnesium nitride could have been expected. Of the elements present in an AZ80 alloy (i.e., magnesium, aluminum, and zinc), magnesium's interaction with nitrogen is the strongest, as indicated by comparing Pauling's electronegativity differences with nitrogen, or comparing free energies of magnesium nitride, aluminum nitride, and zinc nitride.

To determine when during welding nitrogen interacted with magnesium, thermodynamic calculations of the equilibrium state of an equi-molar mixture of magnesium, nitrogen, and hydrogen were conducted. Although the proposed stoichiometry is likely different from those found in the weld, results of this calculation were sufficient to incorporate temperature in the analysis. Results of the thermodynamic calculation are graphi-
Fig. 14 — Microstructures of the FZ2 region shown in Fig. 11. A and B — Optical micrograph; C — secondary electron image with an EDS spectrum; D — X-ray diffracted intensities of the surface of a weld made with nitrogen enriched with 1% diatomic hydrogen gas.

Fig. 15 — Equilibrium products of the equi-molar reaction between magnesium, hydrogen, and nitrogen as a function of temperature.

cally represented as a function of temperature in Fig. 15. In this figure, it is confirmed that the magnesium nitride phase, Mg$_3$N$_2$, was the most stable equilibrium product. However, it is also shown that the Mg$_3$N$_2$ phase formed at much higher temperatures than anticipated (about 1500 K), and certainly at temperatures where magnesium was still in a gaseous state. This new result therefore strongly indicates that some unknown fraction of the magnesium nitride seen in Fig. 14 had formed in the arc atmosphere before condensing on top of the weld. Consequently, the first analysis, suggesting that hydrogen was prevented from entering the weld pool because of nitrogen at the weld surface, is validated. This new result still does not mean that the second explanation, involving the nitrogen partitioning, is invalid. It simply means that understanding weld pool formation in the presence of nitrogen is more complicated than initially thought and requires further research.

Conclusions

The effects of gases, in particular hydrogen and nitrogen, have been investigated with the general objective of increasing melting during GTA welding of magnesium alloys. The following conclusions were reached:

- The weld fusion zone dimensions, particularly the penetration (Fig. 7), were increased with gases having a high first ionization potential, because voltage and thus heat input (under a constant current condition) were increased — Figs. 4, 5.

- Helium, due to its high first ionization potential (24.5 eV), produced deeper fusion zones than argon (15.9 eV).

- When 1% hydrogen was added to either argon or helium, the fusion zones were more penetrating than with only the monatomic gases. Also, nitrogen substituted for either argon or helium, generated even more melting, although its first ionization potential is less than that of either argon or helium. The properties of diatomic gases (Fig. 2), particularly enthalpy and thermal conductivity, could well explain the increased voltage and penetration observed with these gases — Figs. 4, 5.

- Despite desirable increase in melting, hydrogen was also a powerful source of porosity — Fig. 13. However, porosity could be prevented by establishing fully penetrating welds, where heat flow is twodimensional. Due to a slower cooling than in partial-penetration welds, pores could nucleate, grow, and leave the weld pool.

- Regardless of the welding parameters, nitrogen stabilized a second phase near the weld surface — Fig. 14. Although examined in this paper, the interaction between magnesium and nitrogen is not yet well understood.

- Minor additions of hydrogen to argon or helium could be recommended to increase melting. With nitrogen, the properties of the nitride layer must be further investigated.

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References


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**Call for Papers**

The American Welding Society announces a Call for Papers for the 2005 Professional Program to be held as part of Welding Show 2005 on April 26-28, 2005, in Dallas, Texas.

Submissions should fall in one of the following three categories and will be accepted only in a specific format. Individuals interested in participating should contact Dorcas Troche, Manager, Conferences & Seminars via email at dorcas@aws.org for specific details. Deadline for submission of paper is Friday, July 30, 2004.

**Technical/Research Oriented**
- New science or research
- Selection based on technical merit
- Emphasis is on previously unpublished work in science or engineering relevant to welding, joining and allied processes.
- Preference will be given to submittals with clearly communicated benefit to the welding industry

**Applied Technology**
- New or unique applications
- Selection based on technical merit
- Emphasis is on previously unpublished work that applies known principles of joining science or engineering in unique ways
- Preference will be given to submittals with clearly communicated benefit to the welding industry.

**Education**
- Welding education at all levels
- Emphasis is on education/training methods and their successes
- Papers should address overall relevance to the welding industry