Discussion on “The Relationship of Weld Penetration to the Welding Flux”

Introduction


The publication of the paper has since raised questions on the part of B. M. Patchett and E. A. Dancy whose comments appear below under Discussion. Corresponding response on the part of D. D. Schwemmer, D. L. Olson and D. L. Williamson follows under Authors’ Reply.

Discussion by B. M. Patchett and E. A. Dancy

The recent paper by Schwemmer et al. on the effects of flux properties on submerged arc weld bead penetration endeavored to establish connections between observed weld bead dimensions and the viscosity, surface tension and arc stabilizing properties of silicate-based welding fluxes. This paper is a valuable contribution to the analysis of the influence of slags in welding. However, we would like to suggest an alternative interpretation using a thermochronological approach in place of the rather tenuous correlations attempted in the paper.

There is no doubt that arc stability has a profound effect on weld bead shape and penetration, but it is essential to define the type of arc stability being discussed and the sources of each type in a given welding situation. There are two fundamental types of arc stability which are required in welding arcs:

1. Stable emission and collection of electrons at the cathode and anode respectively.
2. An easily ionized arc plasma between the cathode and anode to facilitate conduction of electrons and ions from one to the other.

The first requirement must be met for all welding arcs, while the second is of more importance in arc welding with alternating current when the arc is extinguished, cooled, and reignited repeatedly. With direct current the first requirement is of overwhelming importance.

One of the more enduring myths in welding research is fusion of both requirements as one problem of “arc stability” when in fact they are distinct, albeit related, difficulties. The atoms of low ionization potential (such as sodium and potassium) which improve arc plasma conduction and reignition characteristics are often also accredited with overall improvement in arc stability. However, these atoms are always present in welding situations as compounds of some kind, usually with oxygen or other oxidizing atoms.

It is well known that oxygen and oxygen-bearing gases improve arc stability in the welding of ferrous metals, notably stainless steels, although the resulting improvements in bead shape and penetration are often attributed to charges in interfacial tension between the molten and solid metal caused by the oxides produced during welding. One of the present authors has shown that oxidizing atoms alone can improve cathode emission characteristics dramatically in other cold cathode metals; this indicates that metallic ions of any type, alkaline earth or other, probably have a minor effect at the most on the thermal emission characteristics which are a necessity for direct current arc stability. They are probably, however, convenient carriers of arc stabilizing ions such as oxygen via their relatively unstable compounds incorporated in fluxes, e.g., silicates, carbonates or oxides.

The paper by Schwemmer et al. states that the improvement observed in arc stability with increases in both FeO and CaO in otherwise similar Mn silicate fluxes was caused by the presence of more easily ionized atoms. However, they also point out that FeO substituted for MnO greatly improved arc stability when welding a 4340 steel despite the fact that Fe and Mn have very similar ionization potentials. The improvement in arc stability with a 1020 steel was less, but large enough to be observed. We feel that these improvements are more likely due to an increased oxygen potential in the arc cavity as a result of changes in oxide activity in the slag.

Using a combination of published data and a qualitative understanding of slag structure based on Masson’s polymerization model, it is possible to predict trends in the partial pressure of oxygen (P_{O_2}) over silicate melts as a function of composition. In their section on viscosity, Schwemmer et al. describe the structure of silicate melts according to generally accepted ideas. Polymerization theory developed over the past fifteen years makes the picture more precise and suggests how different metal ions affect the silicate network differently. It tells us that silicate melts at compositions between the ortho- and meta-silicate compositions consist of a large proportion of polymeric ions (SiO_{4}^{4-}, SiO_{3}^{3-}, SiO_{2}^{2-}, etc.) in addition to SiO_{2}^{2-} and O^{2-} ions and the metal ions Mn^{2+}, Fe^{2+}, Ca^{2+}, etc.

The mixing of FeO-SiO_{2} and MnO-SiO_{2} at a given constant mole fraction of SiO_{2} would be expected to be almost ideal since the two oxides affect the polymerization similarly. (Note: because of the closely similar molecular weights of FeO and MnO, constant wt-% composition means almost constant mol-% composition.) This has been demonstrated experimentally at the orthosilicate compositions, 33.3 mol-% SiO_{2}. At 44 mol-%, the composition of these melts, the phase diagram summarizing the stability ranges of solid phases in the ternary FeO-MnO-SiO_{2} system suggests that there could be some deviation from ideality in the melt at the higher FeO contents. This is discussed in the Appendix.

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In the MnO-rich part of the system, fairly close adherence to ideality would be expected. Knowing this and the activities of the constituents of the two binary systems (at 1650°C (3002°F)), in "FeO"-SiO₂, \( a_{\text{FeO}} = 0.45 \) and \( a_{\text{SiO₂}} = 0.90 \); in MnO-SiO₂, \( a_{\text{MnO}} = 0.24 \) and \( a_{\text{SiO₂}} = 0.40 \), and knowing that FeO is overwhelmingly the least stable of these two oxides (at 1650°C, \( P_{\text{FeO}} \approx 10^{-8}, \) 10⁻¹⁵ and 10⁻¹⁷ atm). Over pure FeO, MnO and SiO₂, respectively), it can be predicted that, beginning at some quite low concentration of FeO, the \( P_{\text{FeO}} \) over the molten ternary system at 44 mol-% SiO₂ will increase with increasing FeO contents.

This prediction is supported by experimental data for the system in the solid state at 1150°C (2102°F). Thus, referring to Fig. 5 of the Schwemmer et al. paper, arc instabilities which is shown as decreasing as a function of % FeO also decreasing as a function of increasing \( P_{\text{FeO}} \).

The effect of substituting CaO for some of the MnO in MnO-SiO₂ is less easily predicted. Mixing is not ideal because of a significantly greater depolymerization effect of CaO. In addition, there is a strong tendency for Ca²⁺ ions to be closely associated with silicate ions. Finally, the mole fraction of SiO₂ is diminishing as more CaO is added.

Fortunately, experimental data exist for \( a_{\text{CaO}} \) in the ternary CaO-MnO-SiO₂ system, indicating that—more than half the composition range between the MnO-SiO₂ and CaO-SiO₂, \( a_{\text{CaO}} \) remains very nearly constant at 0.25-30 at 40 wt-% SiO₂ and 1650°C (3002°F), perhaps rising slightly as CaO is first added. This would be expected from the association between Ca²⁺ ions and silicate ions, which would also lead to a fairly steep decrease in \( a_{\text{MnO}} \) from its value of 0.40 in MnO-SiO₂.

This leaves MnO, with the highest activity and the highest \( P_{\text{MnO}} \) over the pure oxide, as the controlling factor defining \( P_{\text{MnO}} \) over the melt. Thus \( P_{\text{MnO}} \) will remain roughly constant or will increase slightly as CaO is added to the MnO-SiO₂ system at 40 wt-% SiO₂. Referring again to Schwemmer et al.'s Fig. 5, we can therefore correlate a roughly constant (slightly decreasing) arc stability with a roughly constant (slightly increasing) \( P_{\text{MnO}} \).

It should be noted that in the above discussion, all values for activities and \( P_{\text{MnO}} \) are for 1650°C (3002°F). The trends predicted would be expected to remain valid at higher temperatures (e.g., 1900°C or 3452°F), but oxygen partial pressures would be substantially higher.

Figure 4 in the Schwemmer et al. paper shows that FeO additions to the flux have a more marked influence on arc stability for AISI 4340 steel than for 1020 steel. This means that the cathode region on the plate surface is affected by the flux, since the electrode wire (anode) is presumably the same in both cases.

Cathode electron emissivity in non-thermionic metals such as iron is quite sensitive to oxygen activity simply because of the iron oxide FeO. The available level of active oxygen will be jointly determined by the amount of oxygen supplied to the cathode surface (from dissolved gases in the steel, oxide inclusions in the steel, atmospheric gases, and flux decomposition), and the degree of removal of such oxygen by deoxidants (Si, Mn, Cr, C, etc.). AISI 4340 steel has a higher level of deoxidants than 1020 as shown in Table 1, and with a lower availability of active oxygen from the plate the 4340 steel would be expected to be more responsive to oxygen supplied to the cathode via FeO additions to the flux.

Figures 8 and 9 add substance to this interpretation, since penetration increases more rapidly as a function of wt-% FeO for the 4340 steel than it does for 1020.

The improved arc stability resulting from the higher oxygen potential of the gas over the flux will automatically lead to deeper penetration simply because the arc wander is minimized allowing the more efficient transfer of heat. Thus, it is hazardous to attempt to link increased penetration with increased viscosity when the same composition change, in this case increased FeO, leads to a decrease in arc instability. The effect of viscosity on penetration is likely to be small compared with the effect of improved arc stability, especially as the variations in viscosity are very limited in practical terms.

For example, the substitution of half the manganese by iron increases the viscosity from 4.5 to 5.7 poise at 1500°C (2732°F), a fairly insignificant change in practical terms, since liquids with viscosities in this range flow easily. In fact, the flux immediately surrounding the arc undoubtedly has an even lower viscosity (of the order of 2 at 1900°C (3452°F), calculated using an activation energy of 19.8 kcal derived from published data for MnO-SiO₂.)

On the other hand, within millimeters, the flux is solid; there has not been time for heat from the plasma to be transferred further, and there is thus a very steep gradient in viscosity values from the arc cavity wall to the unmelted flux burden.

Which value of the viscosity is supposed to be relevant? There is obviously a requirement that the flux not be too fluid, i.e. like water, but then we would be talking of viscosities two orders of magnitude lower, not of differences of a fraction of one order. The amount of scatter to be expected in viscosity measurements and in the measurements of penetration (which vary by only a total of 2 mm or 0.079 in.) would also, in our view, cast some doubts on the validity of inferring a causal relationship between viscosity and penetration.

Before leaving the discussion on viscosity, it should be mentioned that an interpretation of the structural implications of increasing viscosity and decreasing Moessbauer linewidth as a function of composition is presented in the Appendix to these comments. It is not included here, since, although interesting, it is not an essential part of the main discussion on weld penetration.

Surface tension, or interfacial energy, is also unlikely to have a primary effect on penetration. Measurements of contact angles as done by Schwemmer et al. certainly show that interfacial energy differences exist as flux composition changes, but these measurements were done on the as-solidified weld bead shape.

Weld metal solidification takes place well behind the area of arc impingement on the base metal surface, in an area unaffected by the arc plasma. The thermal energy between the molten metal and the base metal surface will be primarily influenced by thermal energy from the arc plasma rather than by interfacial energy.

In relation to this, there is another phenomenon that should be mentioned. Renwick and Patchett found that there is a transition current phenomenon in submerged arc welding similar to that in GMAW. Above the transition current, which is primarily a function of current level and electrode diameter, penetration increases rapidly, and depth-to-width ratios even more rapidly. This occurs because a component of the arc energy is concentrated in the central region, producing physical pressure on the weld pool surface, which alters the fusion depth considerably.

Important side effects that occur are a decrease in the weld bead width and a resulting increase in reinforcing necessary to accommodate the volume of deposited metal. Both of these dimensional changes will influence the final weld bead shape above the plate surface. The shape and penetration below the plate surface will be
determined by the balance between heat transfer from the arc and the pressure exerted on the molten pool surface by the arc plasma. Therefore, even with one flux, one base metal plate and the same welding conditions, a change in electrode diameter alone can greatly alter the weld bead shape and penetration values, as shown in Fig. 13 of Masson.18

The current of 500 amperes at 30 volts used by Schwemmer et al. is close to the predicted transition current of 450 A for a 2.4 mm (%n) diameter wire. Further work well above the transition level may produce different results, as would change in electrode diameter, stickout (i.e., electrode extension), etc. The net result would be a wide variation in the constant “k” in equation (4) of the paper, i.e.,

\[ P = k[y^2 - \gamma^2] \times n [V - \Delta V] + C. \]

All of the parameters used in the equation (viscosity surface tension and voltage instability) have been shown to be dependent on FeO levels in the flux, but this does not necessarily mean that a causal relationship exists between each of them and the weld bead penetration. Schwemmer et al. agree that a decrease in arc instability will lead to increased penetration via more consistent heat transfer conditions, and we feel that this is the dominant influence on penetration and, when ferrous metals are involved, primarily a function of \( P_n \). Thus, we suggest viscosity and surface tension are made to appear to have an influence on penetration simply because they change with slag composition in a manner similar to the change in instability for the particular slag system used, and not because a direct causal relationship has been established.

Since each of the three parameters can be graphically plotted to produce an apparently positive relationship with penetration, a dependence of penetration on the three multiplied together is automatic but not necessarily significant. Much depends on the type of relationship assumed. For example, Fig. 12 of the paper shows a linear dependence of penetration on viscosity for both the 020 and 4340 steels, although the relationship for 4340 appears to be better suited to a non-linear curve. This would bring the data more in keeping with the non-linear relationships for 4340 in Fig. 11 (width/depth ratio vs. wt-% FeO), Fig. 9 (penetration vs. wt-% FeO) and Fig. 4 (arc instability vs. wt-% FeO).

Similar results for 1020 steel are more linear—see Figs. 10, 8 and 4 respectively. However, despite the non-linear relationship found for the 4340 steel, the authors propose that a linear relationship be assumed to exist for both steels in order to suggest the equation relating surface tension, viscosity and arc instability to penetration in the form:

\[ P = (f(n,\Delta V)) + C. \]

We submit that the apparently linear results thus obtained, are fortuitous despite the high correlation coefficients, and suggest that more discretion be exercised when mathematically analyzing or graphically plotting data of one type against another without first establishing reasonable grounds for a causal relationship.

The equations derived in the paper indicate that very large variations in the surface tension-viscosity-voltage instability relationship are needed to yield very small changes in penetration. Penetration is approximately (in mm) 9.0 + 0.015 (\( \Delta y \), n, 30 - \( \Delta V \)). The value of (\( \Delta y \),n,30-\( \Delta V \)) varies from about −50 to +100 when slag FeO contents range between 0 and 30%. Calculated penetration, therefore, varies between 9.0 − 0.75 = 9.25 mm (0.36 in.) and 9.0 + 1.5 = 10.5 mm (0.41 in.).

This is a total range of only 1.75 mm (0.069 in.), and scatter alone can account for a good portion of that in a series of experiments using identical welding conditions. Results obtained by Renwick10 show that the range of penetration for a given flux can be up to 22 mm in many cases, and occasionally more. Although the correlation coefficients obtained in the Schwemmer et al. paper are high, we would suggest that the mathematical relationships may become much less precise if the number of data points is increased, possibly making accurate predictions of penetration very difficult if not impossible.

There is a need to assess other silicate slag systems with similar \( P_n \) variations but different viscosity and surface tension effects in order to discriminate between the points of view expressed in the paper and this critique. We are in the process of starting such an assessment, and hope to report the results in due course.

References


Appendix

The authors’ interpretation of the observed Mössbauer linewidth measurements runs contrary to their own viscosity measurements, to their own description of silicate melts, and to polymerization theory. High viscosities, arising because of the directional nature of the Si-O bonds and the polymeric silicate ions, impart a tendency for a melt to form a glass (amorphous) rather than crystalline solid on cooling.

Put simply, when the melt is cooled rapidly, the large silicate units do not have time to organize themselves into the positions required to give a crystalline solid. The short-range order in such liquids and solids is very nearly that of the crystalline material, but the long-range order is not. In other words, the nearest neighbors of any given atom are almost the same in number and position as they would be in the crystalline material, but next-nearest and more distant neighbors are progressively less likely to be found in the positions they would occur in the crystalline array.

As already mentioned, the FeO and MnO containing fluxes have similar degrees of polymerization in the 44 mol-% SiO₂ composition range. However, the MnO flux probably has a somewhat higher proportion of smaller silicate units. Thus, substitution of MnO by FeO would be expected to increase the viscosity slightly, as Schwemmer et al. found experimentally, the increase being due not to increased “crystallinity,” this term pre-
sumably implying increased long-range order, but to a small increase in the polymerization of the silicate units. This can be worded more traditionally by saying that iron is not as good a network modifier as manganese, i.e., it is more covalent in nature than manganese and, so, gives up its oxygen less easily to the silicate network, resulting in less breaking up of the complex silicate units.

If the increased viscosity does not result from an increase in the “crystallinity,” it remains to offer an alternative explanation for the change in the Mössbauer linewidth. In the main part of the text of these “comments,” possible deviation from ideality at higher FeO contents was mentioned. This remark was prompted by the fact that in the solid state at 1150°C (2102°F), there is a change in the phases present in passing across the MnSiO₃-FesiO₃ join.

At the orthosilicate composition (33.3 mol-% SiO₂), a continuous solid solution exists between FeSiO₃ and Mn₂SiO₄. The metasilicate solid solution (50 mol-% SiO₂), on the other hand, extends only part way from the MnSiO₃ end member towards FeSiO₃, because the latter is unstable relative to the phase assemblage Fe₂SiO₄ plus SiO₂. Thus, substitution of FeO for MnO in the 44 mol-% SiO₂ system in the solid state at 1150°C (2102°F) will, at first, result in progressive enrichment in FeO of the two coexisting solid solution phases, (Mn,Fe)SiO₃ and (Mn,Fe)SiO₄.

Slightly on the FeO-rich side, a composition is eventually reached beyond which further FeO enrichment results in disproportionation of the (Mn,Fe)SiO₃ phase into (Mn,Fe)₂SiO₄ and SiO₂. This means that on the MnO-rich side, the metal ions will be distributed between two different types of local environment while on the FeO-rich side they will all be in identical environments.

In the liquid state, as we have already discussed, order exists at the nearest neighbor level. Thus, in MnO-SiO₂, Mn⁺⁺ ions will be found distributed between “sites” approximating those in crystalline Mn₂SiO₄ and MnSiO₃. At low FeO concentrations, Fe⁺⁺ ions will be similarly distributed. Because of the lack of structural constraints in the liquid state compared with the solid crystalline state, the changes occurring as a function of FeO content will be gradual rather than abrupt. Thus, as the FeO content increases, the short range order in the vicinity of more of the metal ions will become closer to that of the orthosilicate, with neighboring regions becoming closer in structure to pure silica.

The result of such a change as a function of composition would be a deviation from ideality at higher FeO contents. It would also mean that Fe⁺⁺ ions would increasingly be found in orthosilicate “sites” rather than distributed between two types of “sites”. Put differently, this means that as the FeO content increases, the Fe⁺⁺ ions increasingly occupy “sites” with the same local environment, accounting for the observed Mössbauer linewidth changes for the rapidly quenched material.

Authors’ Reply

Welding flux research has produced data based on a variety of process parameters, using commercial grade fluxes and usually too many variables for effective analysis. It was the intent of the work of Schwemmer et al.¹ to systematically investigate the physical behavior of the MnO-SiO₂-FeO welding flux system using reagent grade flux materials and constant welding parameters and report their results in a manner in which other investigators can make use of them, as did Patchett and Dancy.

The work reported by Schwemmer et al.¹ is only a very small part of the work performed at the Colorado School of Mines on the manganese silicate welding flux systems².¹ Fourteen different flux systems have been investigated including additions of alkali, alkaline earth, and first row transition metal ions. These results indicate good agreement in correlating the data with the penetration expression (equation 4 in Schwemmer et al.), especially for 1020 steel. In fact, the penetration expression was reported in the Welding Journal only after it was known that it was successful in correlating data in other flux systems as shown in Fig. 1.

Figure 2 illustrates the correlation for the manganese silicate welding fluxes with alkali ion additions on 1020 steel. Recent work with alkali additions on 4340 steel did indicate problems in correlating penetration to the relative interfacial tensions.

The concept of Patchett and Dancy to relate penetration directly to oxygen activity is interesting and worthy of further investigation. It must be realized that the partial pressure of oxygen on which their concept is based is orders of magnitude lower than the residual oxygen contamination level in inert gas used in the GTA and GMA welding processes. The very similar arc instability behavior for the
two different flux systems shown in Fig. 3 suggests that arc stability under a flux may not be simply controlled by just the oxygen activity since it is known that CaF₂ additions cause major alterations in oxygen activity. Also notice in Fig. 4 that 4340 steel has a lower arc instability compared to 1020 steel for the same flux, indicating that the interpretation of Patchett and Dancy that alloy additions of 4340 steel effectively getter oxygen and thus promote arc instability is a tenuous concept.

In a more recent investigation the arc instability was found not to be simply explained by ionization potential concepts, for the arc instability actually decreased with ionic additions of increasing ionization potential as illustrated in Fig. 5. This does suggest that the cathode spot interactions are probably primary to arc stability.
Based on the Mössbauer resonance linewidth data (Fig. 3 in Schwemmer et al.), Patchett and Dancy have offered a plausible interpretation based on the concept of short range order. However, the Mössbauer spectra from the fluxes indicate the situation to be more complex than their suggested two-site versus one-site Fe$^{3+}$ environment expected for (Mn,Fe)$_2$SiO$_4$ and (Mn,Fe)SiO$_3$ for the MnO-rich side vs. only (Mn,Fe)$_2$SiO$_4$ for the FeO-rich side.

Our analysis has accounted for up to four types of sites in a given sample (as evidenced by structure in the resonance spectrum) such that the linewidth presented represents that for a single type of site. Thus the broad lines observed for low FeO concentrations are interpreted by us to be a measure of the lack of long-range order.

An important point to be remembered in considering the Mössbauer data is that all measurements were made at room temperature on fluxes quenched from the liquid state. Therefore, the decrease in linewidth with increasing FeO concentration may indeed correspond to increasing crystallinity (long range order) due to a quench rate insufficient to preserve the amorphous state as the FeO content increases. This means that this research effort was looking into functionality of $k'$ in terms of physical properties of the welding flux. The coefficient $k$ in the penetration expression of Schwemmer et al. is definitely a function of the welding parameters. Also comparing Figs. 1 and 2 it can be seen that the value of $k$ will vary from one major flux system to another. Thus a specific penetration equation is probably only accurate for a given system. The functional form of this expression may be useful to the welding flux designer. Researchers at the Colorado School of Mines will make an effort to test the concepts of Patchett and Dancy as well as others as we systematically analyze and report our data.

$P = k'^2 \sqrt{\frac{P}{SE^2}}$

Their coefficient $k'$ is a function of the specific flux. In the work of Schwemmer et al., the welding parameters were intentionally held constant. This means that this research effort was looking into functionality of $k'$ in terms of physical properties of the welding flux. The coefficient $k$ in the penetration expression of Schwemmer et al. is definitely a function of the welding parameters. Also comparing Figs. 1 and 2 it can be seen that the value of $k$ will vary from one major flux system to another. Thus a specific penetration equation is probably only accurate for a given system. But the functional form of this expression may be useful to the welding flux designer. Researchers at the Colorado School of Mines will make an effort to test the concepts of Patchett and Dancy as well as others as we systematically analyze and report our data.

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References


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(1) A Critical Evaluation of Plastic Behavior Data and a Unified Definition of Plastic Loads for Pressure Components
by J. C. Gerdeen

(2) Interpretive Report on Limit Analysis and Plastic Behavior of Piping Products
by E. C. Rodabaugh

(3) Interpretive Report on Limit Analysis of Flat Circular Plates
by W. J. O'Donnell

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