

Review: Low transformation temperature weld filler for tensile residual stresses reduction

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Abstract

An attractive, alternative approach for the reduction of harmful residual stresses in weld zones is reviewed, which utilises low temperature, solid-state, displacive phase transformations in steel. The theory, latest concepts and practice for the design of such low transformation temperature (LTT) filler alloys are considered. By engineering the phase transformation temperature of the weld metal so as to take advantage of transformation expansion, the residual stress state within the weld zone can be significantly altered, most particularly where the weld thermally contracts with any movement of base parts constrained. To date, the technique has been shown to increase fatigue strength for some common weld geometries, which may enable engineering design codes to be favourably re-drafted where such LTT filler alloys are used.

Keywords

Welding, Residual stress, Martensite, Low transformation temperature weld metal.

1. Introduction

Welding integrity is crucial to many important industrial processes; welded structures in building construction, mining equipment, ships, agricultural machinery, bridges and off-shore platforms are just a few examples. Besides the efficiency of the process, equipment and operational costs are low, making the process favourable compared to other methods of joining. On the completion of welding, due to thermal contraction and geometric construction, a pattern of compressive and tensile residual stresses will be present in and adjacent to the weld zone on cooling to room temperature. The presence of tensile residual stress is said to be the main reason why the fatigue strength of a welded joint does not increase by strengthening the base steel [1]. In this review, the theory, latest concepts and practice for the design of welding alloys capable of offsetting such tensile, residual stresses in weld zones, generated from thermal contraction are presented. Such 'smart' alloys are particularly beneficial in mitigating tensile residual stress for constrained welds.

2. Theoretical and experimental basis of LTT weld filler alloys

2.1 Transformation under constraint

The mechanism by which the austenite transforms on cooling can be described as being either reconstructive or displacive movement of iron atoms to another lattice configuration. With displacive transformation, there is a homogeneous deformation of the original crystal arrangement into a new structure, which does not involve the bonds being broken and subsequent re-arrangement of atoms as seen in reconstructive transformation to give allotriomorphic ferrite and/or pearlite. When cooling rate is rapid (e.g. most weld scenarios) or if the weld metal is heavily alloyed, the displacive transformation becomes prevalent where the resultant microstructure can be acicular ferrite, bainite or martensite. The microstructural characteristics of such phase transformations from the austenite phase are extensively reviewed [1-5], but most importantly, such transformation is always couple with tremendous shear and volume expansion.

For relatively large and heavy parts being welded, most base material flanking the weld will be constrained, thus thermal contraction on cooling and inhomogeneity of temperature distribution will induce residual stresses in the weld 'zone' (i.e. the weld and flanking heat-affected areas). In the usual

case of long, thin weld run(s) between two parts, the highest tensile residual stress magnitude will be unidirectional. In order to counter-act the thermal contraction during cooling from the austenite with volume expansion, the transformation temperature can be engineered. Such residual stress / temperature relationships is shown schematically in Figure 1a where the transformation start temperature is observed by the sudden decrease in stress, which can be attributed to the dilatation and shear strains that occur and compensate the accumulated thermal contraction.

Experimentally, this effect during welding can be clearly shown with constrained cooling tests, such as the Satoh test. The test involves a tensile test specimen, which heated so as to be fully austenitic which is then control cooled under unidirectional restraint. Some results [6-8] are shown in Figure 1b for different types of steel (martensitic, bainitic and austenitic, respectively). It can be seen that the fully austenitic steel has a near-linear thermal contraction slope, while those of the bainite and martensite transformation reflect the reduction in residual stress; however the beneficial offset in contraction strain is negated by the continued cooling to ambient after the transformation product has been exhausted. These observations led to the conclusion that the final stress state of a welded component is not only affected by transformation temperature, but it has the potential to be reduced by lowering the transformation temperature. Figure 1c shows the effect on stress in such a reduction in transformation temperature with two alloys (LTTE and Series B). It can be seen that the stress is being driven into compression or near zero stress at room temperature, compared to that of a conventional filler alloy (OK75.78) which has a tensile stress at room temperature after transformation at ~ 450 °C.

The effect of the phase transformation temperature of various weld filler alloys, on the residual stress distribution within the weld zone, was investigated by both Wang et al.[9] and Murata et al.[10]. Murata et al.'s results (Figure 2) showed that once the transformation temperatures get sufficiently low, not only are tensile residual stresses reduced to zero, but compressive residual stresses can be generated with cooling to at room temperature. These reach a maximum when the transformation of the filler alloy occurs around 200 °C. Below that temperature, the transformation and associate volume expansion are not complete on cooling to ambient temperature thus not fully cancelling the tensile residual stresses from restrained contractive cooling. If the phase transformation is absence (e.g. if an austenitic, weld filler alloy is used), then the tensile stress is expected to increase progressively due to uninterrupted thermal contraction.

Although significant research in this area only started two decades ago, it has been known for a longer period that tensile residual stresses in weld zones can be reduced by the use of 9% Ni filler alloy with a M_s of 350 °C [11] and compressive residual stresses can be obtained when the transformation temperatures are even lower (e.g. $M_s = 250$ °C), as observed in filler alloy for maraging steels [12, 13].

2.3 Unconstrained tests

For a linear weld run, the magnitude of the residual stresses generated during cooling in longitudinal and transverse direction will differ (Figure 2) with transformation start temperature. This figure illustrates that difference, plus the resultant angular distortion, from flat weld test sample [10]. This was shown clearly where unconstrained plates were V-notched and welded along the complete plate lengths [14]; angular distortion was reduced 45% by using an alloy with a transformation temperature range of 350-422°C compared with one with a range of 400-802°C. Further, distortion was minimal when using a filler alloy with a transformation start temperature of between 250 and 300°C. Such differences were also found valid for multi-pass welding [15]. The simple experimentation clearly shows that the use of such novel, LTT weld filler alloys will not only reduce harmful tensile residual stresses in constrained weld zones, but also minimise the weld zone distortion.

2.4 Residual stress surveys of welded test samples.

Techniques for measuring residual stress can be classified as destructive or non-destructive. Destructive techniques involve disruption of the residual stress from its state of equilibrium, generally by material removal to create a free surface, whilst local changes in strain or displacement are measured (e.g. hole drilling and the 'contour method') [16-19].

Advances in non-destructive, diffraction based techniques such as in synchrotron X-Ray diffraction and neutron diffraction in recent years has facilitated the study of the effect of phase transformation and residual stress in 3-dimensional detail without destroying the sample [8, 20, 21]. Both techniques are essentially using the lattice parameter as a strain gauge [22]. Figure 3 shows neutron diffraction surveys across test welds for the three filler alloys; one standard, two LTT. Their chemical compositions and some respective mechanical properties of the alloys are given in Tables 1 and 2, respectively. The works have provided an insight into stress generation and has further shown that the use of LTT alloys

would result in near-zero or compressive residual stresses over the cross section of the weld metal, whilst tensile residual stresses remain in the non-transformable HAZ (Heat Affected Zone). The measured tensile residual stresses in the HAZ are lower compared to those seen in welds where conventional, high transformation temperature, filler alloys have been used [8, 21]. This is due to the expansion force of the weld metal phase transformation that counteracts the tensile residual stress generated due to the contraction of non-transformable HAZ.

3.0 Design of LTT 'smart' weld filler alloys

The previous sections have shown the importance in lowering the phase transformation temperature in these LTT alloys to ensure maximal expansion to relieve tensile, residual stresses generated during thermal contraction. It is also importance to consider other design requirements during alloy design. The LTT alloy should have higher or equal strength compared to the base metal as well as appropriate toughness. Whilst design factors such as 'weldability' and susceptibility to hot and cold cracking should be considered when new weld alloys are designed, factors such as the corrosion resistance would depend upon the application of the alloy.

The temperature of phase transformation of the LTT alloy is strongly dependent on the content of the austenite (γ)-stabilizing alloying elements. The known and most commonly used γ -stabilizing alloying elements are carbon, nickel, manganese and chromium; these are discussed in detail in below. Various LTT alloys, which have been investigated by various researchers, are listed in Table 1 with respect to chemical composition, M_s temperature and expansion strain where given; in the following Table 2, mechanical properties are listed where given. All data are cross-referenced to respective sources within the tables.

Low carbon and nitrogen contents are very desirable for a filler weld alloy to prevent the formation of hard and brittle martensite in the HAZ, which is susceptible to hydrogen-assisted cold cracking. The lower limit of carbon content in weld alloy is normally reported being about 0.03 wt%; however if this value could be further lowered, to say 0.003 wt% (as obtained in an interstitial free steel), significant improvement of toughness would be expected. As shown in Table 2, the toughness of LTT weld metal 'B206' was significantly improved when the carbon content was reduced to 0.01 wt%.

With reference to Table 1, low carbon and nitrogen plus the addition of certain alloying elements are required lower the transformation temperature. Nickel can be a popular choice of alloying as it is known to improve toughness, when the cost of nickel is not too expensive. Francis et al.[21] and Darcis et al.[23] have designed a LTT alloy with nickel as a major alloying element; however its toughness was not significantly better than alloys with a chromium and nickel combination. A more cost effective alloying element is manganese and attempts has been made to design manganese based, weld filler alloys in which silicon has been added to improve the bead morphology; however their toughness has been low [24] Martensitic Fe-Mn alloys can be very brittle [25-29] probably as a result of a co-segregation of manganese and phosphorus to prior austenite grain boundaries [25, 27, 30].

Although chromium is a ferrite stabiliser, about 10-14 wt% can be added to decrease the transformation temperature. In this instance, the alloy can be easily designed to have a minimum of 12 wt% so that the alloy can be 'stainless'. Low carbon content is also required to prevent the formation of chromium carbides during the reheating process. Formation of such precipitates would result in inter-granular attack due to chromium depletion. Chromium additions also improve the oxidation resistance.

Most of the LTT alloys designed in the past two decades to date have had a combination of Cr and Ni with most also having a low carbon content [31-33]. The content of nickel has been adjusted to between 5 and 11 wt% to control the phase transformation starting temperature. This is considered to be the most economic option where reasonable toughness can be obtained. There was a concern of galvanic corrosion when these types of LLT alloy were used to weld high strength low alloy steel due to the compositional mismatch. However, work in Japan had shown no trace of selective/galvanic corrosion in synthetic sea water when LLT filler alloys were use to weld high strength low alloy steel [34]. Similar results were also obtained by Zenitani et al.[35].

Impurity elements like sulphur and phosphorus should be kept as low as possible whilst elements like molybdenum and tungsten can be added dependent upon whether the application, for example if pitting corrosion is to be prevented. Micro-alloying elements such as titanium and niobium can be used to control nitrogen content.

Considerable experience in stainless steel welding research has shown that alloys that solidify as primary ferrite are resistant to hot cracking compared with alloys that solidify entirely as austenite.

Therefore, it is also preferable to design 'smart' LTT alloys with primary ferrite solidification. Small amount of discontinuous δ ferrite can be left in the weld with ferritic solidification as this is believed to be beneficial to toughness [36]. However, if there is a significant amount of δ ferrite such that a continuous network forms, this has been found to be detrimental to impact toughness with a high ductile to brittle transition temperature [37]. The austenite grain size will also be finer where it is formed from the decomposition of δ -ferrite compared with the grain size of austenite formed directly from the molten weld metal [38]. A fine grain size will also contribute to higher toughness.

It should be noted that the phase transformation temperature of the LTT filler alloy in actual welds, as against Satoh type test rigs, will be affected by other variables, such as dilution by the base metal, the high temperature cooling rate, which affects the amount of δ ferrite formed, which in turn will affect the carbon content of the austenite.

4.0 Calculation of the M_s and martensite transformation

It has been shown that the phase transformation temperature of the filler metal can significantly change the distribution of residual stress; therefore it can be very beneficial to determine an alloy composition which will give an advantageous transformation temperature together with a suitable transformed microstructure. When the phase transformation is depressed to very low temperature or when significant alloying elements are added, it is often the case that the martensite will form. The temperature of the martensite start temperature (M_s) of the filler weld metal is strongly dependent on the content of austenite stabilizing alloying elements. The effect of the chemical composition on M_s temperature has been extensively researched over the years. Several linear regression empirical equations have been proposed [39-43] whilst thermodynamic [44-47] and artificial neural network models [48-50] were also developed to predict M_s temperatures. All such methods provide a reasonable guide during the design stage as long as they are used within the data boundaries stipulated by the respective model. This includes a thermodynamic model where its accuracy is limited by the underlying thermodynamic, empirically obtained, database [49].

The martensite reaction occurs athermally and is effectively not time dependent; the amount transformed is dependent on the degree of undercooling beneath the M_s temperature. This is expressed by the equation of Koistinen and Marburger [51]:

$$1 - V_{\alpha'} = V_{\gamma} = \exp\{-b(M_s - T_q)\} \text{ where } b \approx 1.10 \times 10^{-2} \text{K}^{-1} \quad [1]$$

where $V_{\alpha'}$ is the fraction of martensite, V_{γ} the volume fraction of retained austenite and T_q the 'quenching' temperature below M_s to which the sample is cooled. Whilst a value of material constant $b = 1.1 \times 10^{-2} \text{K}^{-1}$ is normally quoted for ferrous martensite [4], it has been determined that 'b' lies within the range, $1.75\text{-}1.90 \times 10^{-2} \text{K}^{-1}$, in low carbon, chromium and nickel rich welding alloys [52].

5.0 Applications of LTT 'smart' alloys

5.1 Application of LTT 'smart' alloys to improve fatigue strength

In comparison with its static strength, the fatigue strength of a joint welded using a conventional filler alloy is much weaker. As shown in Figure 4, welded components have a comparatively low fatigue resistance. Although the fatigue strength of the base metal will increase with increasing base metal yield strength, the fatigue strength for a particular welded component remains more or less constant, despite weld dilution. This is why the fatigue tolerances for welded structures in the design code are classified into joint type according to the type of weld and its orientation with respect to the applied fluctuating loads rather than being based on the yield strength of the base materials [53].

The low fatigue resistance of the welded joint is mainly due to a short crack initiation period. This is significantly influenced by the joint geometry and also the tensile residual stress distribution in the joint weld zone. The former can include the abrupt shape change and/or a surface defect that induces stress concentration.[54] The sharper the transition between the weld bead and the parent material, the higher the stress concentration will be and thus the greater effect on fatigue life. The introduction of tensile residual stresses not only reduces the fatigue crack initiation resistance and the fatigue limit of the welded components,[55] but also increases the fatigue crack growth rate [9, 56].

It is possible to apply post weld treatments to improve the fatigue performance and significant work has been done to extend the crack initiation life of welded joints. In general, the post weld improvement methods can be classified into weld geometry improvement and residual stress alteration. The weld geometry can be improved by, for example, TIG-dressing [57, 58] or grinding to reduce the

stress concentration factor. However, in fillet welds, it is not always possible to eliminate the geometrical effect. The second approach is to alter the residual stress present, either by mechanical or thermal means. It can be the introduction of surface compressive stresses by locally deforming the surface e.g. shot-peening or hammering etc. [59, 60] or by post-weld heat treatment [61] to relieve the residual stresses. While all of these techniques are efficient in increasing fatigue life, these processes also required additional work after welding, which are both time consuming and an additional cost [62, 63].

LTT welding facilitates not only the reduction of tensile residual stresses, particularly transversely across the weld zone, but also the generation of compressive residual stresses in the weld zone, thus increasing fatigue strength. Such improvements have been found for all types of welded joint [31, 33, 64-70]. This achievement is based entirely on the fact that the reduction of the transformation temperature allows the volume expansion originated from martensite transformation to compensate for the accumulated thermal contraction strains, with that expansion variant directional when the weld cools within a directional, non-hydrostatic, stress field. Early work has also shown that the fatigue strength of a non-load-carrying cruciform welded joint increases in proportion to the increased base metal strength when LTT welding wire alloy is used [71, 72]. A further benefit in using LTT filler alloys has been found where there are weld surface defects, such as blowholes. It has been claimed that blowholes of about 4.5 mm diameter in a LTT weld were found to have not reduced fatigue strength, as the compressive residual stress around them prevented the fatigue crack initiation by the stress ratio effect [65].

Cruciform joints [23] and out-of-plane gusset (box) welded joints [73, 74] have, to date, been the most common test geometry in fatigue strength investigations of the LTT weld filler alloy effect. As is shown in Figure 5, that these types of joint generally a comparatively low fatigue strength and there is an obvious cost and design benefit if their fatigue strengths can be improved by using such filler alloys.

The LTT welds have shown a considerable improvement of fatigue resistance compared to that of conventional joints in constant amplitude fatigue testing, with some reports showing about 40% increase in mean fatigue strength [23, 31, 64, 66, 68, 75]. However with variable amplitude fatigue testing, the improvement observed was not as significant (12%) compared with that found during

constant amplitude loading [75]. This was mainly due to the relaxation of the compressive residual stresses at high amplitude loading during variable amplitude fatigue.

5.3 Welding processes suitable for use of LTT 'smart' alloys

As discussed in the previous section, very low carbon and nitrogen are paramount to obtain good toughness and this can be obtained readily with metal-cored wires [37]. The wires are suitable for submerged arc (SAW), tungsten inert gas (TIG) and metal inert/active gas (MIG/MAG) welding, offering the requisite versatility and high output. When MIG/MAG is chosen, a gas mixture containing a high proportion of carbon dioxide must be prohibited since this will lead to an increase in carbon content of the deposited metal. Another significant effect on toughness is oxygen content, which varies with respective welding processes. As shown in Figure 6, the toughness of martensitic steel welds decrease with increasing oxygen contamination [37, 76-78]. It also shows that TIG process produce the lowest oxygen content compare to other process. Excessive inclusion formation and significant loss of alloying elements to the slag are the main reason why the mechanical properties of weld metal are diminished by high oxygen content.

5.4 Effect of inter-pass temperature on multi-pass LTT alloy welds

The interpass temperature was found significant, as distinctly different residual stress fields were obtained with different interpass temperatures. If interpass temperatures are higher than the martensite start temperature (M_s) thus giving an austenitic matrix, all passes will only transform into martensite after respective welding runs are completed. Therefore, much of the fusion zone will be left under compressive stress. Where interpass temperatures are lower, with underlying pass temperatures not exceeding the M_s temperature, the underlying weld runs zones will be tempered rather than re-austenitized. As a consequence, only a small zone around or within the final weld run will be in compressive stress, while other regions will have tensile residual stress zones [79].

5.5 Effect of post-weld treatment where LTT weld filler alloy has been used

Thibault et al.[79] have characterized residual stress patterns on LTT weld that had undergone a standardized post-weld heat treatment. They showed that post weld heat treatment not only reduced the

harmful tensile residual generated in non-transformable HAZ, but also generated beneficial compressive stresses.

5.6 Effect of welded geometry and dilution

These have also been found to affect significantly the M_s temperature [80, 81]. The optimum M_s temperature of the weld filler metal required for the introduction of residual compressive stress and minimum distortion in a butt weld is $\sim 200^\circ\text{C}$, whereas for a fillet welded T joint, optimum M_s is $\sim 400^\circ\text{C}$, where the transient angular distortion is minimised. Recent work by authors have shown that dilution must be considered when using LTT filler alloys for welding austenitic stainless steels, which may have higher alloying concentrations than the welding alloy. Such concentrations may lower the M_s temperature such that significant transformation may not occur. It has been shown that after a first pass the microstructure can remain austenitic, whereas with subsequent passes there was transformation to martensitic. This shows that different configuration shape of the welded joint, or structural restraint, as well as welding conditions all influence the residual stress pattern generated.

Heavily alloy weld metal has been produced to counter the dilution effect [70]. It is found that the element concentration of welds was significantly lower than the initial pure alloy material with dilution levels varying between $\sim 25\text{--}35\%$ for single-pass and $\sim 10\%$ for two-pass welds [70]. The work highlights that single pass dilution effects and the improvements in fatigue performance that can be achieved. The dilution effect on fatigue was considered to be a result of the M_s change influencing residual stress rather than microstructure.

5.7 Cold cracking resistance in LTT 'smart' alloys

Whilst it has been shown that the use of LTT weld alloys is effective in the mitigation of harmful residual stresses, the martensitic microstructure of some LTT alloys can be susceptible to cold cracking. Zenitani et al.[82, 83] and Kromm et al. [84] have investigated the cold cracking behaviour of LTT weld alloys using the Tekken test. They found that LTT weld metals with a two phase microstructure (martensite and retained austenite) had excellent cold cracking resistance where the M_s temperature is even lower than normal for LTT alloys and where some retained austenite was still

present at room temperature. Such retained austenite facilitated hydrogen entrapment hence reducing cold cracking susceptibility.

6.0 Finite Element Modelling

Whilst the residual stresses would be generated from uneven heat dissipation and any restrictions on expansion and contraction within localised areas, their magnitude would vary dependent upon the joint restraint, its geometry and the specific welding process employed. If the parts to be joined were free to expand and contract then the residual stresses would be minimal. Where the heat input range varies with different welding processes, the residual stresses generated will differ [77]. The material's physical and mechanical properties, that affect the type and magnitude of residual stress will include the heat capacity, thermal expansion coefficient, density and strength. Poor thermal conductivity and high thermal expansion coefficient would increase significantly the residual stresses in a weldment, for example, austenitic stainless steel.

Given the complexity of the physical processes that are taking place, it would be of great benefit to know whether a weld zone residual stress distribution could be predicted on the basis of the weld filler alloy transformation behaviour. To this end, numerous finite element software packages [85, 86] were created to account all aspects of the respective welding processes; these included heat flow, material behaviour and evolution of stress and strain. These newly developed models also accounted for transformation plasticity [87, 88]. While the welding condition and material properties inputs are straight forward, it has been reported that finite element simulation has shown that the transformation temperature of the weld metal does not have a significant effect on the magnitude of the peak stresses beyond the HAZ and that this is due to the unaccounted variant selection during simulation [20]. In this respect, it was suggested that variant selection must be incorporated in the simulation in order to more accurately predict the weld zone residual stress distributions where LTT weld filler alloys have been used.

6. Conclusions

The steady cooling curve for the residual stress - temperature relationship for the contraction of steel weld zone on cooling is interrupted by the period of expansion due to crystallographic phase

transformation on cooling from austenite. Where transformation is completed well above room temperature, the transformed product will then further contract introducing the tensile residual stress. The lower the transformation temperature, the less the effect of this. If full transformation occurs near room temperature, for some situations around 200 °C, then the benefit of transformation expansion can be exploited to greatly reduce critical concentrations of any harmful tensile residual stresses and, with some alloys, even generate beneficial, compressive residual stresses within the weld zone. Such improvements apply to multi-pass welds as well as single pass welds, although the magnitude of the effect is dependent upon re-heat temperatures of previous passes.

Reducing harmful tensile residual stress concentrations or even generating compressive residual stresses in weld zones will give increased resistance to fatigue. This in turn, may require design codes to be reviewed where LTT filler alloys are used. Additionally, proscribed post weld heat treatments may not be required.

Further research and development into suitable LTT alloys should continue and be expanded to quantify their beneficial affects with different weld configurations, particularly on-site, with different weld environments and base steels.

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List of Tables

Table 1: Chemical composition (wt.%), M_s temperature and expansion strain of LTT weld metals from various sources [8-10, 23, 31, 33, 64, 79, 82-84, 89-95]. Also included in the table are the chemical composition of the standard weld metals and base metal used in this review.

Table 2: Mechanical properties of LTT weld metals from various sources [21, 23, 64, 66, 92, 96].

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Fig. 1. Effect on residual stress with cooling; the balance of phase transformation expansion and thermal contraction.

(a) Stress temperature schematic (after [6]).

(Note - Thermal expansion coefficients for ferrite, $13 \times 10^{-6} \text{K}^{-1}$; for austenite $21 \times 10^{-6} \text{K}^{-1}$ [97]).

(b) Satoh test results for final residual stress of different steel phase types on cooling (after [6]).

(c) Satoh test results for conventional (OK75.78) and LTT weld filler alloys (LTTE & Series B). Samples had cooled from 850°C at 10°C s^{-1} (after [8]). Respective chemical composition data estimated transformation start temperatures and properties are shown in Tables 1 and 2.

Fig. 2. Directional differences in residual stress and angular distortion on cooling a test weld down to room temperature. (a) Relation between residual stress (σ_x , σ_y) of weld metal and starting temperature of transformation [10]. Similar results were also obtained by [9]. (b) Relation between total angular distortion after the final pass and the measured transformation starting temperature [10].

Fig. 3. Axial residual stress map (MPa) across HAZ's of welds using conventional filler alloy ('OK 75.78') and two LTT alloys ('LTTE', 'Series B'). (Their compositions, physical and cooling-residual stress properties are shown in Tables 1 & 2 and Figure 2c). Stresses were determined by neutron diffraction and superimposed on macrographs; stresses were averaged around the centre lines; i.e. "mirrored" (taken from reference [8]).

Fig. 4. Fatigue results [70] for single-pass fillet welds using a conventional filler alloy ('OK Autorod 89') and two LTT alloys ('LTT- C', 'LTT- S'). LTT-S provides evidence for improved fatigue

performance through designing an alloy to account for dilution. Open symbols represent specimens that did not fracture. Chemical compositions and Ms temperatures are given in Table 1.

Fig. 5. Effect of base metal test piece configuration on welded component fatigue life [67, 98].

Fig 6. Influence of oxygen content on Charpy-V toughness at -20°C and -40°C in Mo-alloyed supermartensitic weld metal [37].

Table 1: Chemical composition (wt.%), M_s temperature and expansion strain of LTT weld metals from various sources [8-10, 23, 31, 33, 64, 79, 82-84, 89-95]. Also included in the table are the chemical composition of the standard weld metals and base metal used in this review.

LTT alloy identification	C	Si	Mn	Cr	Ni	Mo	Other	M_s (°C)	Expansion Strain (%)	Ref.
10Cr-10Ni	0.025	0.32	0.7	10	10	0.13		180	0.55	[31, 74]
L2	0.04	0.17	0.27	11.69	10.01	0.04		80	0.28	[9]
L3	0.04	0.17	0.3	10.51	9.62	0.05		138	0.52	
L4	0.07	0.230.23	1.25	9.1	8.46	0.05		191	0.63	
L5	0.08	0.230.17	1.35	7.78	6.88	0.06		242	0.54	
L6	0.07	0.23	1.3	6.3	5.29	0.04		287	0.48	
L7	0.09	0.47	1.51	5.07	4.43	0.06		325	0.44	
L8	0.08	0.45	1.5	3.91	3.23	0.04		408	0.36	
LTTE	0.07	0.2	1.3	9.1	8.5			200		
Series B	0.03	0.65	0.5	1	12	0.5		275		
B206/ OK Tubrod 15.55	0.01	0.4	1.1	12.5	6.7	2.5	0.5Cu			[64, 89]
13Cr/LC35	0.04	0.5	0.8	12.3	7.3	2.2				[64]
Optimised 'D'	0.047	0.4	1.5	11.1	8.6	0.3		94		[83]
A6	0.08	0.19	0.89	14.7	0.27	0.04		360	0.244	[23, 90]
B5	0.04	0.19	0.86	13	1.7	0.04		300	0.578	
C5	0.05	0.22	0.41	3	13.2	0.35		270	0.57	
Camalloy 4/2b	0.014	0.76	1.36	12.66	5.24	0.1	0.053Ti	216	0.47	[36, 99]
LTTW	0.04	0.32	0.36		11.9					[91]
A	0.042	0.22	0.65	10.91	9.37	0.26		194		[82]
B	0.035	0.25	0.43	12	5.26	0.46		320		
A	0.046	0.22	0.69	11.05	9.51	0.296		122		[95, 100]
B	0.048	0.23	0.67	11.03	6.49	0.307		220		
R	0.046	0.22	0.7	12.54	9.48	0.298		60		
O	0.025	0.32	0.7	10	10	0.13		210		
C13N	0.029	0.15	0.19	15.66	7.07	-		250		[92]
C15N	0.024	0.15	0.19	12.98	8.9	-		250		
XTT8	0.04	0.26	0.56	7.6	10	0.47		209		[10]
XTT10	0.05	0.21	0.61	5.9	7.8	0.5		295		

LTTE1	0.04	0.17	0.27	10.69	10.01	0.04		79		[33]
LTTE2	0.04	0.17	0.3	10.51	9.62	0.05		144		
LTTE3	0.07	0.23	1.25	9.1	8.46	0.05		191		[33, 101]
'410NiMo' LTT filler alloy (bead centre measurement)	0.020	0.34	0.48	12.5	3.8	0.47	0.06 Cu			[79]
LTT 'weld 1' Nom. 8% Ni alloy	0.07	0.45	1.05	9.52	7.33	< 0.02				[84, 93, 94]
LTT 'weld 1' Nom. 10% Ni alloy	0.06	0.45	1.05	9.57	9.44	< 0.02				
LTT 'weld 1' Nom. 12% Ni alloy	0.08	0.40	1.00	9.19	10.81	< 0.02				
LTT-C	0.014	0.7	1.27	13.4	6.1	0.07		281		[70]
LTT-S	<0.020	<1.0	<2.0	15 - 18	6 - 8	<2.0		221		
Standard weld										
OK 75/78	0.05	0.19	2.0	0.4	3.1	0.6				[8, 102]
OK Autorod 89	0.09	0.8	1.9	0.3	2.2	0.6				[70]
Base Metal										
Weldox 960	0.20	0.50	1.6	0.7	2.0	0.7	0.3Cu			[8, 102]

Table 2

Table 2: Mechanical properties of LTT weld metals from various sources [21, 23, 64, 66, 92, 96].

Weld Metal	0.2% PS (MPa)	UTS (MPa)	Elong. (%)	Impact toughness (J)					Ref.
				-40°C	-30°C	-20°C	0	20°C	
10Cr-10Ni	822	1192	36			48			[66]
10Cr-10Ni	708	1021	7						[96]
B206/ OK Tubrod 15.55	700-850	950-1050	>15	>100				>110	
13Cr/LC35	680	1050	15	30				36	[64]
A6	500	624					8	10	[96] [23]
B5	886	1002					12	14	
C5	702	1155					14	14	
Camalloy 4/2b	838	1069	17.6			53		72	[36]
LTTE(3)	1135	1287	6		17	15		20	[21]
Series B					22	27		28	
C13N	495	1051	7						[92]
C15N	671	1117	7.7						
Standard Weld									
OK 75/78	~ 1100			79		90		96	[21]
Base Metal									
Weldox 960	> 960	980 - 1150	> 12		27	27-30			[21]

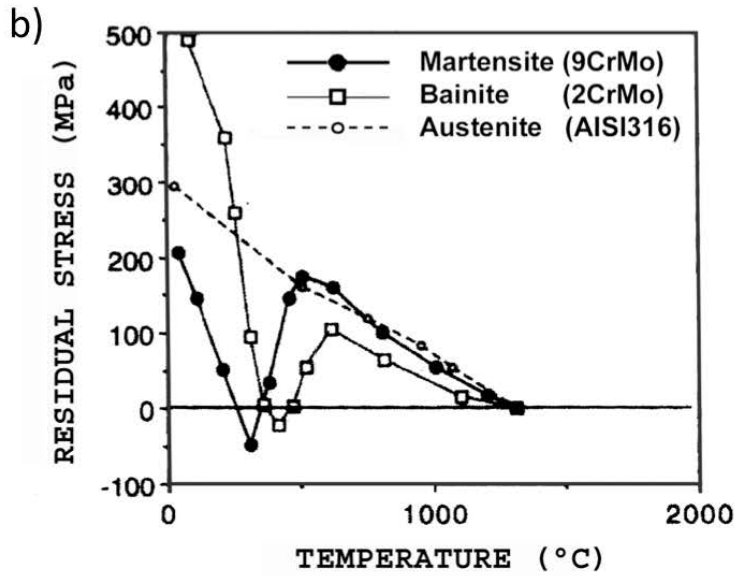
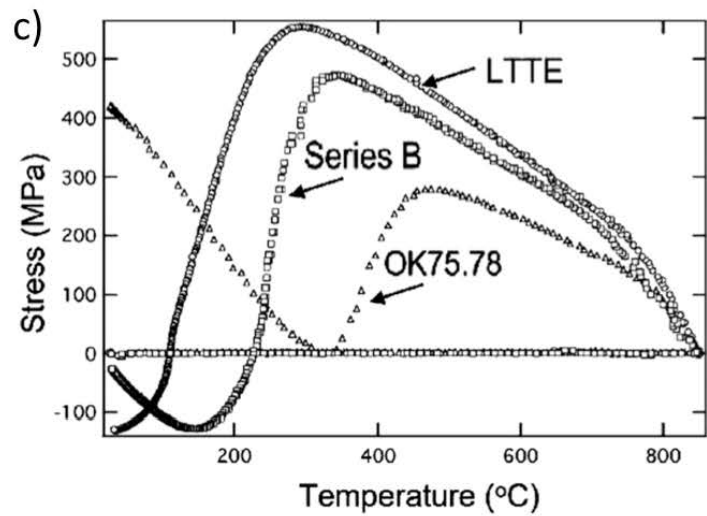
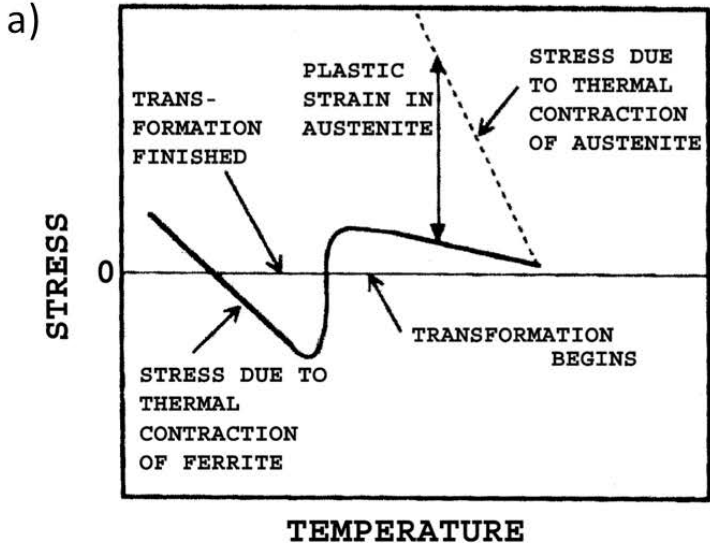
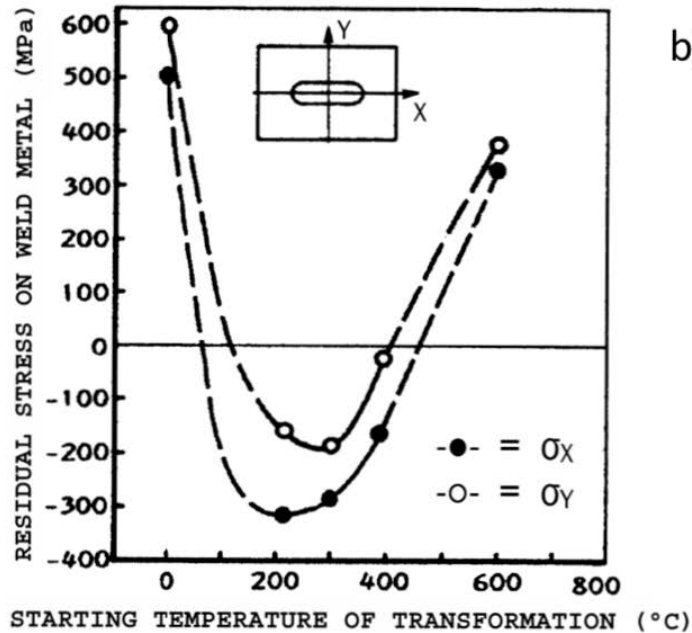
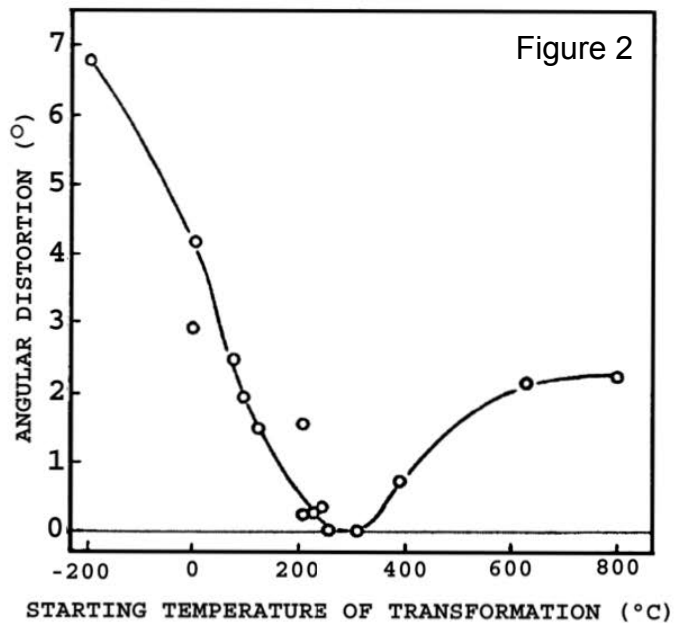


Figure 1

a)



b)



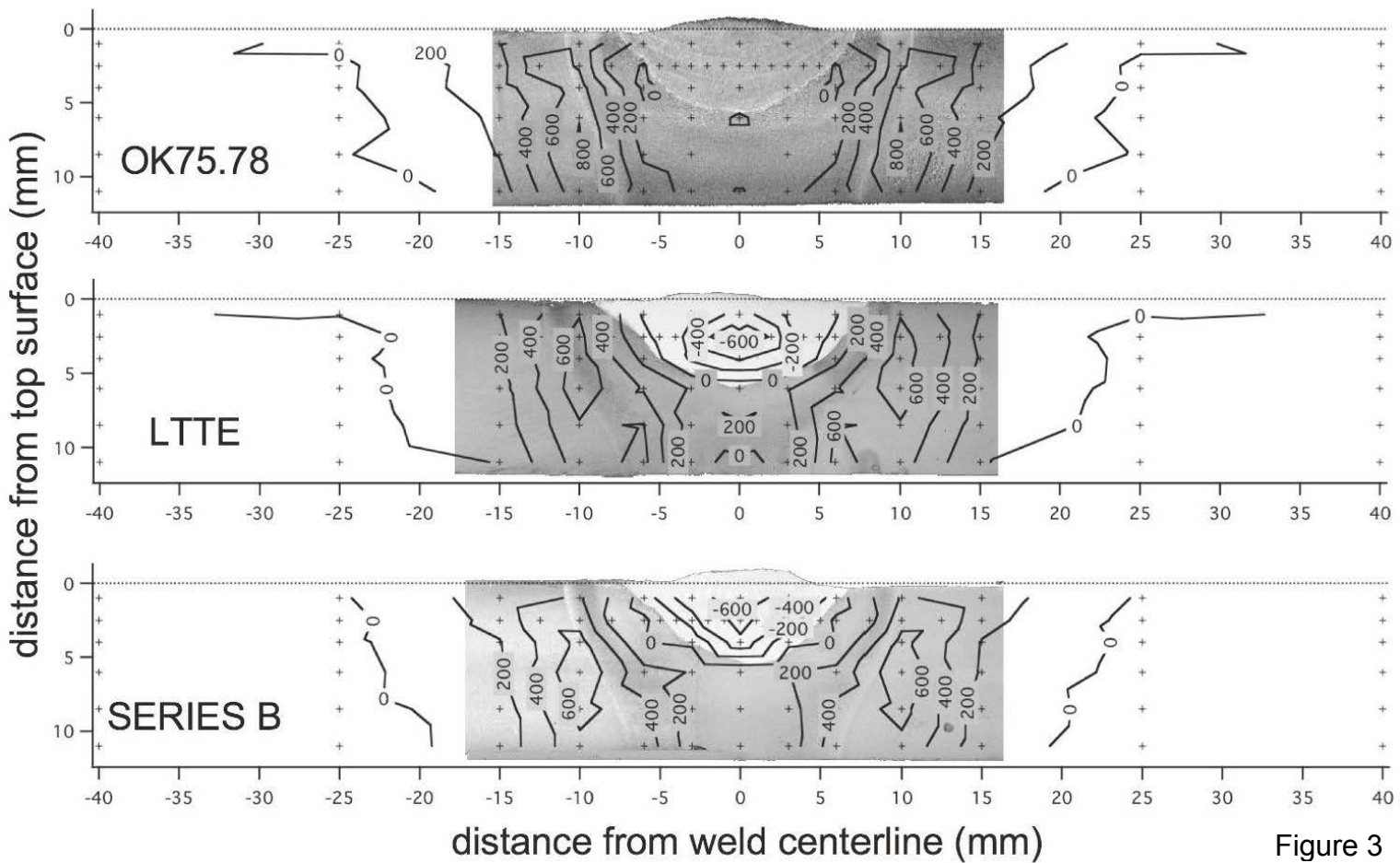


Figure 3

Stress Range (MPa)

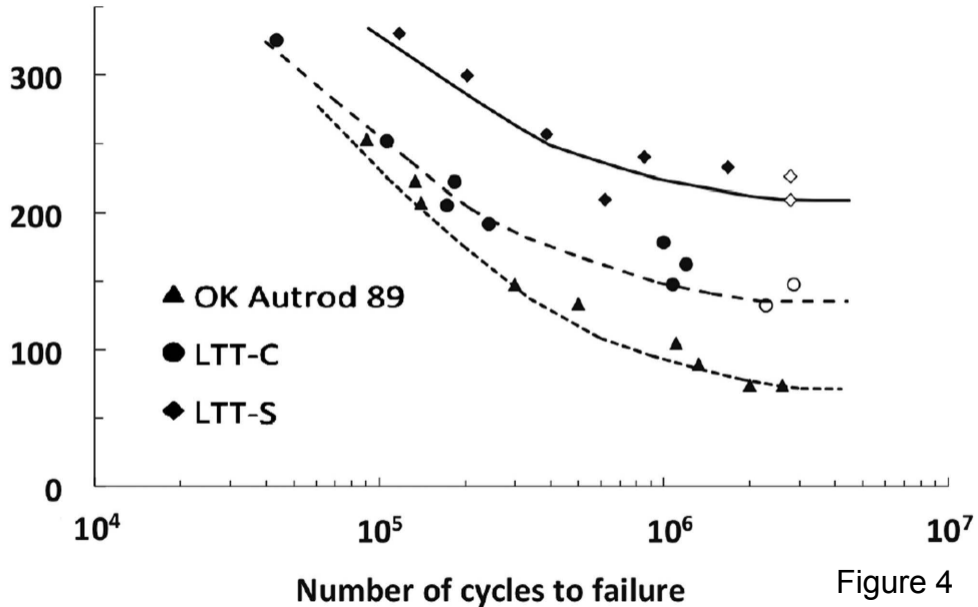


Figure 4

Fatigue strength at 2×10^6 cycles,

$\Delta \sigma$ (MPa)

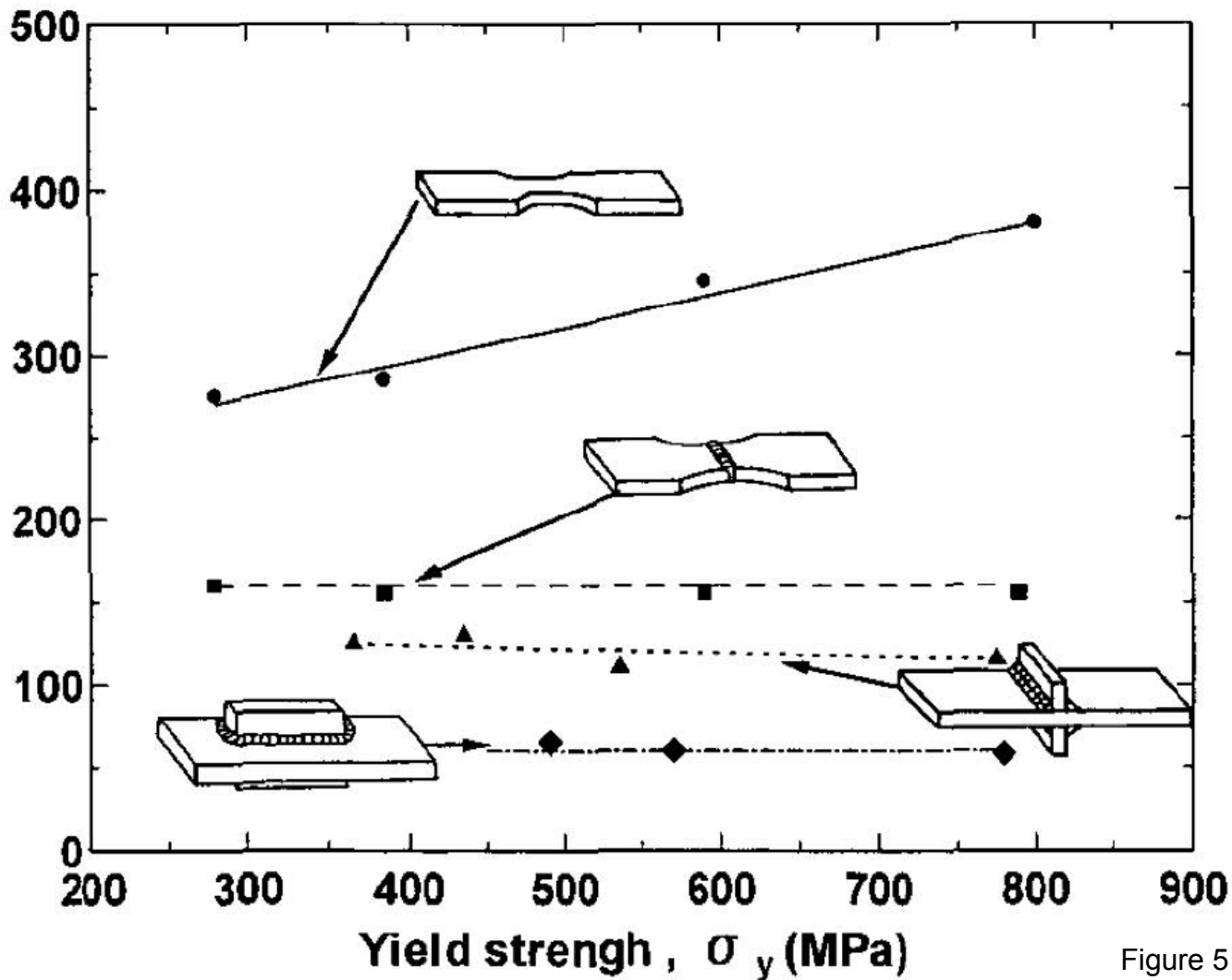


Figure 5

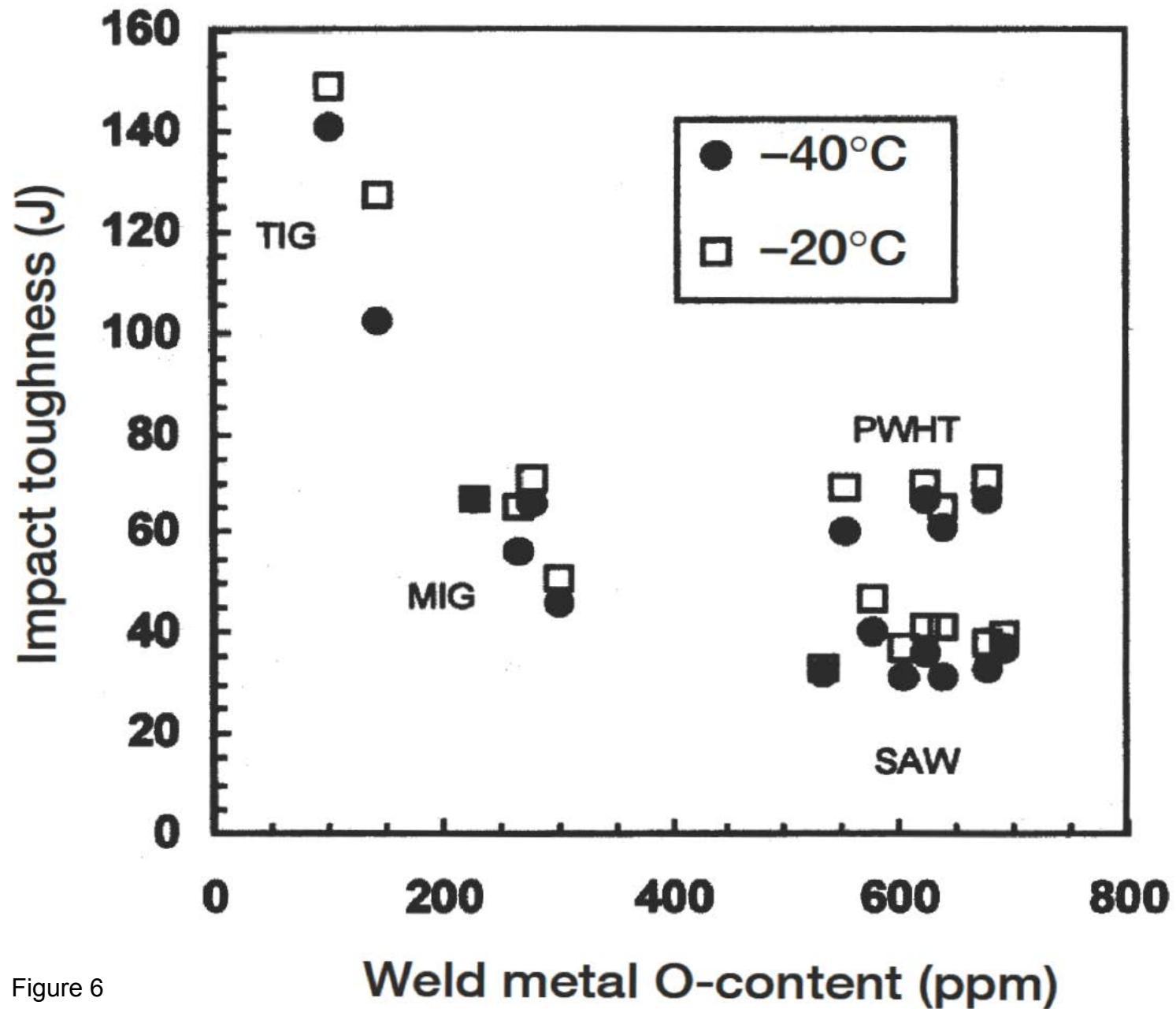


Figure 6