PHILOSOPHICAL TRANSACTIONS OF THE ROYAL SOCIETY A MATHEMATICAL, PHYSICAL & ENGINEERING SCIENCES

The Effects of Residual Elements and Deoxidation Practice on the Mechanical Properties and Stress Relief Cracking Susceptibility of <latex>$\frac{1}{2}$</latex>%CrMoV Turbine Castings

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The effects of residual elements and deoxidation practice on the mechanical properties and stress relief cracking susceptibility of $\frac{1}{2}$%CrMoV turbine castings

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The effects of residual element content on the mechanical properties and stress relief cracking susceptibility have been investigated for basic electric arc melted $\frac{1}{2}$%CrMoV steel castings deoxidized by using either aluminium or titanium practices. The residual element contents ranged from the lowest readily available in commercial practice to levels substantially higher than those common at present.

In normalized and tempered material, deoxidation by the use of aluminium resulted in lower creep rupture ductility than deoxidation by using titanium, regardless of residual element content. Only in one cast did low purity appear to correlate with low ductility.

Similarly, in simulated heat-affected zone material, the susceptibility to stress relief cracking was less in casts deoxidized with titanium. Increasing the residual element content had a slightly deleterious effect on stress relief cracking susceptibility, but austenite grain size refinement gave significant improvement.

It is concluded that adoption of the titanium deoxidation practice used in the present work would improve the creep rupture ductility and stress relief cracking resistance of $\frac{1}{2}$%CrMoV steel, but that at present there is no need for very low residual element contents. Regardless of deoxidation practice, sound welds should be obtainable when adequate grain refinement can be produced by control of the welding process.

Introduction

Stress relief cracking may occur in low alloy creep resistant steels in the coarse grained, heat-affected zones (h.a.zs) of weldments during post-weld heat treatment. It has been suggested that the problem could be related to creep rupture ductility and that it might be reduced by using high purity materials or by changing the conventional deoxidation route from an aluminium to a titanium based practice (Benes & Skvor 1972; Harris & Jones 1972; Hopkins *et al.* 1971; King 1976; Myers 1972*b*; Ratliff & Brown 1967; Stone & Murray 1965; Tipler 1972; Tipler & Hopkins 1976; Townsend 1975; Viswanathan & Beck 1975; Batte & Murphy 1973). Since residual element contents may rise in future years in normal sources of scrap, it was considered necessary to establish the levels that can be tolerated without deleteriously affecting the resistance to stress relief cracking or other high temperature properties. It was also considered desirable to verify the reported increase in stress relief cracking resistance resulting from titanium deoxidation, and further to demonstrate that this deoxidation route did not reduce the creep rupture ductility.

A collaborative programme was, therefore, designed by B.S.C., G.E.C. Turbine Generators, C. A. Parsons Turbine Generators and C.E.G.B. to study the effects of changes in both residual levels and deoxidation practice achievable in commercial practice for large $\frac{1}{2}$%CrMoV steel castings. The tensile, impact and creep rupture properties of normalized and tempered material were measured, together with the susceptibility to stress relief cracking of simulated heat affected zone material.

Test material

By using three melts from a 2.5 t basic electric arc furnace, experimental castings were produced with residual contents ranging from those associated with basic oxygen steel scrap to those substantially higher than is common at present (table 1). Each melt was split into two ladles and deoxidized in one with aluminium and zirconium (aluminium practice) and in the other with titanium and calcium silico-manganese (titanium practice). A test block, 790 mm × 250 mm square, together with a feeder head of similar volume, was cast from each of the six melt–deoxidation route combinations. While the compositions obtained (table 1) were essentially as intended, the tin content in cast 3 was somewhat low. The difference in molybdenum content between casts 2A (0.45 %) and 2B (0.56 %) was unintentional, but nevertheless within the normal range specified for steel of this type. The deoxidation additions were constant, but the final deoxidation element contents varied considerably from cast to cast.

After removing the feeder heads, the blocks were annealed at 970 °C for 12 h, furnace cooled, then reaustenitized at 960 °C for 8 h and cooled at a rate to simulate the conditions in the flange of a large turbine casing casting (Errington & Murphy 1973). Finally, the blocks were tempered at 700 °C for 12 h and furnace cooled. A slice was then cut from each block and examined by sulphur printing, macro-etching and hardness testing. The material was found to be generally uniform across the section with little segregation.

Normalized and tempered material

(a) *Microstructure*

After normalizing and tempering, each cast had a hardness of 135 to 165 h.v. and a predominantly ferritic microstructure containing 7–11 % pearlite. The ferrite had in each case a grain size within the range ASTM 6–7 and contained a heterogenous distribution of vanadium carbide, characteristic of this type of material (Murphy & Branch 1969; Branch *et al.* 1974). The volume fraction of manganese sulphides was low, compatible with the low sulphur levels obtained, and the particles were associated mainly with other types of inclusion. Isolated sulphide trails were, however, noted in casts 2A and 2B, which had approximately twice the sulphur content of the other casts.

(b) *Tensile and impact properties*

Tensile tests were carried out on material from each test block at temperatures up to 725 °C. The tensile properties (figure 1) fell generally within a scatter band determined for thick section ½%CrMoV castings and closed die forgings tested in the authors' laboratories and at E.R.A. (Branch *et al.* 1974). At room temperature there was a small increase in strength level with increasing residual content, but the maximum difference in tensile strength amounted to only 50 MPa. At temperatures above 500 °C there was even less difference. Across the whole temperature range the variation in ductility between the casts was insignificant, all being generally within the scatter band for normal production material.

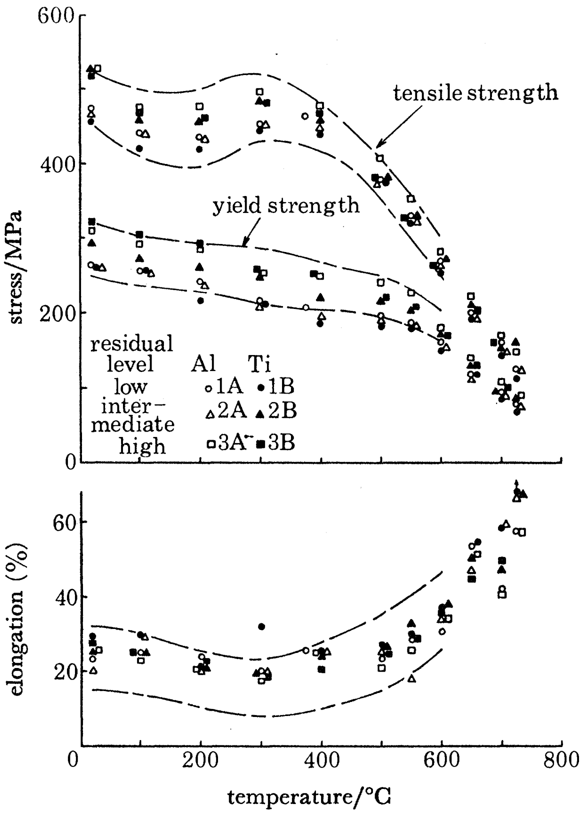

FIGURE 1. Effect of test temperature on tensile properties of normalized and tempered material, including scatter bands for thick section ½%CrMoV components.

Charpy V-notch impact transition data were determined for each test block. There were no significant trends within the data (f.a.t.t. 38–80 °C), which lay essentially within the range established within the authors' laboratories for production aluminium deoxidized material (f.a.t.t. 35–90 °C).

(*c*) *Creep rupture properties*

The results of the creep rupture tests at 550 °C (figure 2) also mainly lay within the scatter bands for production ½%CrMoV castings and forgings tested in the authors' laboratories and at E.R.A. (Branch *et al.* 1974). There was no consistent effect of residual element content or deoxidation practice on the rupture strength. However, the rupture ductility of the aluminium deoxidized casts fell earlier than that of the titanium treated material, such that after 1000 h testing at 550 °C, the aluminium deoxidized material was markedly less ductile. Increasing the

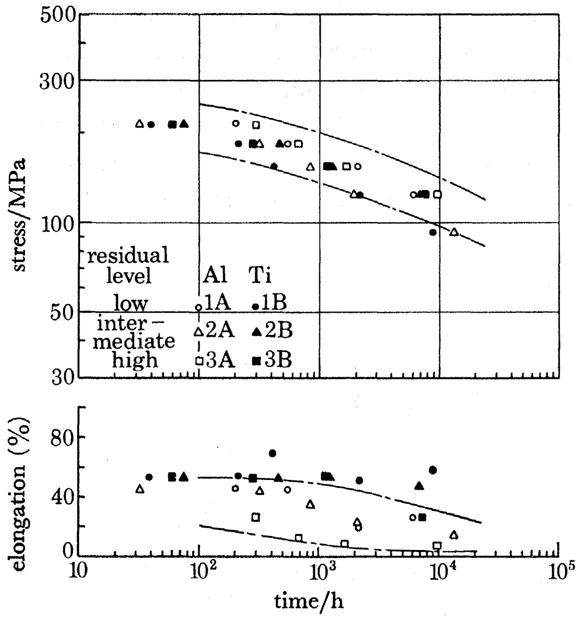

FIGURE 2. Creep rupture properties of normalized and tempered material at 550 °C, including scatter bands for $\frac{1}{2}$%CrMoV production components.

residual element content reduced the creep rupture ductility slightly. Similar, but less discernible, trends were observed in tests carried out at 500 °C.

STRESS RELIEF CRACKING TESTS ON SIMULATED HEAT-AFFECTED ZONE MATERIAL

(*a*) *General test procedure*

To simulate the effect of a welding cycle in the h.a.z. region of a thick section joint, specimens from each cast were given high temperature heat treatments of short duration (table 2). In each case the treatment resulted in a fully bainitic microstructure with a range of prior austenite grain sizes (table 3). These materials were then subjected to various tests (table 2), designed to assess their susceptibility to stress relief cracking. In all of these tests failure occurred in an intergranular manner, typical of stress relief cracking observed in power plant joints.

(*b*) *Tensile ductility at* 550 °C

In the untempered, coarse grained condition the tensile ductilities of the titanium deoxidized casts were slightly higher than those of the corresponding aluminium treated materials (figure 3*a*). The two lowest residual content casts more than doubled in ductility on tempering, whereas the other steels were essentially unaffected. The elongation after 10 h tempering at 700 °C (table 4), which has previously been taken as a criterion for susceptibility to stress relief cracking (King 1976) suggested detrimental effects of residuals and aluminium deoxidation. The stress relaxation treatment (table 2) reduced the ductility of the low residual casts (table 4), although little systematic effect of residuals or deoxidation was apparent. However, the high residual aluminium treated cast (3A) failed during relaxation.

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TABLE 1. COMPOSITIONS OF STEELS USED (PERCENTAGES BY MASS)

cast no.	C	Si	Mn	P	S	Cr	Mo	Ni	Al	As	Bi	Co	Cu	N	Pb	Sb	Sn	Ti	V	Zr	O
1A	0.13	0.40	0.64	0.006	0.005	0.36	0.50	*0.06*	*0.044*	*0.006*	0.0005	0.01	*0.06*	0.012	0.0005	*0.001*	*0.008*	*0.001*	0.25	*0.040*	0.0057
1B	0.13	0.34	0.64	0.005	0.004	0.37	0.49	*0.06*	*0.010*	*0.006*	< 0.0005	0.01	*0.07*	0.012	0.001	*< 0.001*	*0.007*	*0.036*	0.26	*0.001*	0.0065
2A	0.13	0.36	0.60	0.007	0.012	0.40	*0.45*	*0.11*	*0.022*	*0.006*	< 0.0005	< 0.01	*0.20*	0.015	0.001	*0.002*	*0.012*	*0.001*	0.26	*0.043*	0.0078
2B	0.12	0.37	0.58	0.005	0.009	0.38	*0.56*	*0.11*	*0.011*	*0.006*	< 0.0005	< 0.01	*0.20*	0.013	0.001	*0.002*	*0.014*	*0.028*	0.25	*0.002*	0.0064
3A	0.14	0.39	0.62	0.006	0.004	0.44	0.49	*0.25*	*0.035*	*0.021*	< 0.0005	0.01	*0.41*	0.012	0.0005	*0.003*	*0.018*	*0.001*	0.26	*0.045*	0.0082
3B	0.12	0.32	0.60	0.006	0.004	0.43	0.52	*0.25*	*0.005*	*0.020*	< 0.0005	0.01	*0.47*	0.011	0.0005	*0.004*	*0.018*	*0.024*	0.23	*0.001*	n.a.
normal range	0.10–0.15	0.45 max.	0.40–0.70	0.03 max.	0.03 max.	0.30–0.50	0.40–0.60	0.30 max.					0.30 max.				0.05 max.		0.22–0.30		

n.a., not analysed.

TABLE 2. WELD SIMULATION TREATMENTS AND TEST PROCEDURES FOR SIMULATED H.A.Z. MATERIAL

test	austenitizing conditions	mean cooling rate in the range 700–500 °C K/s	testing procedure	references
tensile tests at 550 °C	*coarse grained material* austenitized at 1300 °C for 10 s; *fine grained material* as above, reaustenitized at 950 °C for 30 min	1.4	tensile properties determined at 550 °C ($\dot{\epsilon} = 10^{-5}$/s) (*a*) in untempered condition (*b*) after tempering at 700 °C for 3 or 10 h (*c*) after loading to 100 MPa and stress relaxation allowed to occur for 3 or 10 h at 700 °C	King (1976) Townsend (1975)
creep rupture tests at 690 °C	heated to 1300 °C in 20 s	17	specimens of waisted geometry loaded at room temperature and heated at 85 K/h to 690 °C and rupture testing carried out	Myers (1972*a*, *b*) Myers & Price (1977)
hot c.o.d. tests	heated to 1270 °C in 7 s	8	fatigue pre-cracked specimen loaded in 3 point bending, heated at 50 K/h to 700 °C and held 12 h or until failure occurred (failure defined as arbitrary loading point displacement of 5 mm)	Harris & Jones (1972)
notched bend relaxation tests	heated to 1400 or 1250 °C in 5 s and held 2 s	14	Charpy type specimens loaded in 3 point bending in jig, heated at 50 K/h to 700 °C and immediately cooled at same rate; samples broken open at − 196 °C and area of oxidized crack surface measured.	Miller & Batte (1975)
Murray tests	heated to 1300 °C over 55 s and air cooled	38	notched tensile specimens heated to temperatures in the range 575–725 °C at 25 °C intervals and after 20 min at temperature loaded to 120 % of 0.2 % proof stress of the normalized and tempered material at that temperature; time to failure recorded.	Murray (1967)

Table 3. Grain sizes of simulated h.a.z. material

(Mean linear intercept grain sizes are given in micrometres and, in parentheses, as ASTM numbers.)

cast no.	tensile tests at 550 °C		creep rupture tests at 690 °C	hot c.o.d. tests	notch bend relaxation tests		Murray tests
	coarse	fine			coarse	fine	
1A	130–180 (1.5–2.5)	22–32 (6.5–8)	55 (5.0)	34 (6.5)	46 (5.5)	22 (7.5)	79 (4.0)
1B	130–180 (1.5–2.5)	16–32 (6.5–8.5)	51 (5.5)	19 (8.0)	44 (5.5)	15 (9.0)	64 (4.5)
2A	130–180 (1.5–2.5)	22–32 (6.5–8)	—	29 (7.0)	44 (5.5)	20 (8.0)	84 (4.0)
2B	130 (2.5)	22–32 (6.5–8)	—	18 (8.0)	31 (7.0)	13 (9.5)	100 (3.5)
3A	180 (1.5)	22 (8)	51 (5.5)	37 (6.0)	48 (5.5)	22 (7.5)	81 (4.0)
3B	180 (1.5)	16–22 (8–8.5)	55 (5.0)	27.5 (7.0)	22 (7.5)	14 (10.0)	107 (3.0)

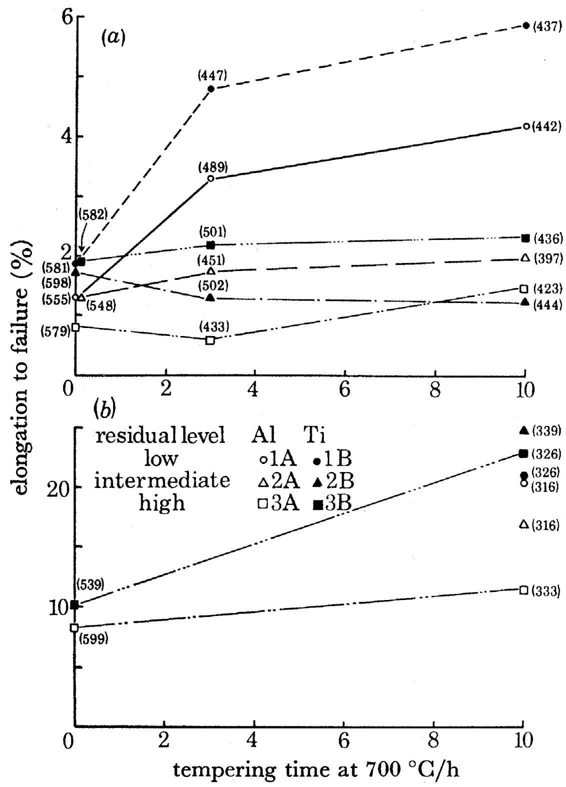

Figure 3. Effect of tempering time at 700 °C on the subsequent ductility in tensile tests at 550 °C on (*a*) coarse grained and (*b*) fine grained simulated h.a.z. material. Figures in parentheses are tensile strengths in megapascals.

In the fine grained condition all casts had a much higher ductility after 10 h tempering than their coarse grained counterparts (figure 3*b*, table 4). In each case the titanium deoxidized material was more ductile than the corresponding aluminium treated cast, this effect being greatest at the high residual level. In this fine grained condition casts 3A and 3B showed no impairment of properties after stress relaxation (table 4).

TABLE 4. SUMMARY OF RESULTS OF TESTS ON SIMULATED H.A.Z. MATERIAL

test	tensile tests at 550 °C: elongation to failure after 10 h at 700 °C for tempered (T) and stress relaxed (R) material (%)				creep rupture tests at 690 °C	hot c.o.d. tests	notched bend relaxation tests: notch opening displacement to give 10% cracking/μm		Murray tests
	coarse grained		fine grained						
cast no.	T	R	T	R	1 h rupture strength/MPa	initial K to give failure in 12 h / $\mathrm{MN\ m^{-\frac{3}{2}}}$	coarse grained	fine grained	time to failure at 675 °C / min
1A	4.1	2.3	20.4	n.t.	174	5.2	90	250	12
1B	5.9	1.4	21.0	n.t.	212	8.3	150	> 350	†
2A	2.0	2.5	16.9	n.t.	158	4.9	90	310	†
2B	1.2	0.9	24.8	n.t.	150	7.6	75	> 350	†
3A	1.5	failed	11.5	10.7	127	4.6	75	155	*ca.* 2
3B	2.3	1.7	23.0	19.7	150	7.3	110	> 350	†

n.t. Not tested.

† Indeterminable owing to scatter.

(*c*) *Creep rupture strength at* 690 °C

The creep rupture strength at 690 °C of the high purity titanium deoxidized cast was significantly higher than that of all the other casts, the low purity aluminium deoxidized cast having the lowest strength of all (figure 4, table 4). It has been shown previously (Myers 1972*b*) that the h.a.z. creep rupture strength correlates with the resistance to stress relief cracking and, hence, the trend of these results indicates increased resistance to stress relief cracking with low residual levels and titanium deoxidation. Previous data on wrought commercial steels has indicated that only steels with 1 h rupture strengths below 150 MPa exhibited stress relief cracking in practice. On this basis only cast 3A with both high residual element levels and aluminium deoxidation would be susceptible to cracking.

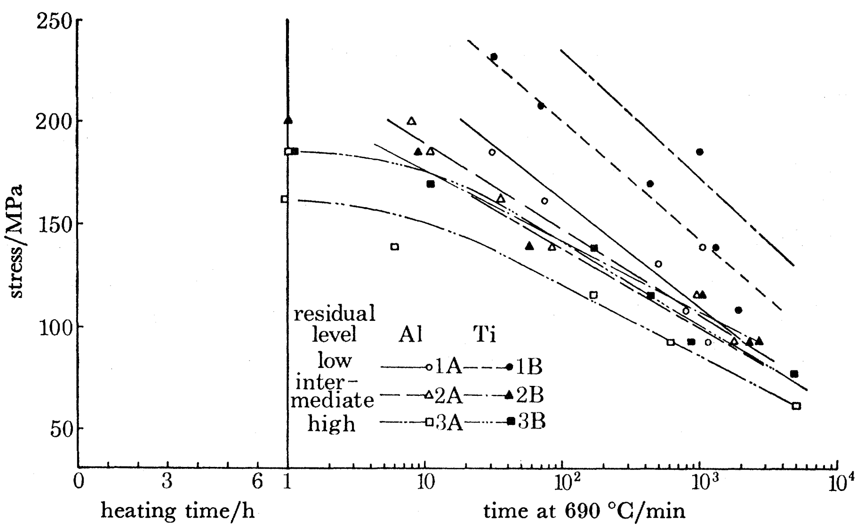

FIGURE 4. Creep rupture properties of simulated h.a.z. material at 690 °C, including scatter band for non-susceptible wrought ½%CrMoV steels.

(*d*) *Hot crack opening displacement tests*

The time to failure (or temperature of failure, when failure occurred during heating), increased with decreasing initial stress intensity, *K*, until a limiting value was reached, below which failure did not occur during the cycle (figure 5). These limiting values (table 4) were consistently higher for the titanium treated casts than for aluminium deoxidized casts. Also, within each deoxidation practice the limiting *K* value decreased with increasing residual element content; however, this effect was slight. Although the present results lie within the range of values observed previously for commercial $\frac{1}{2}$%CrMoV steels (Harris & Jones 1972), the present grain sizes (table 3) were significantly finer than those obtained under similar conditions in that previous investigation.

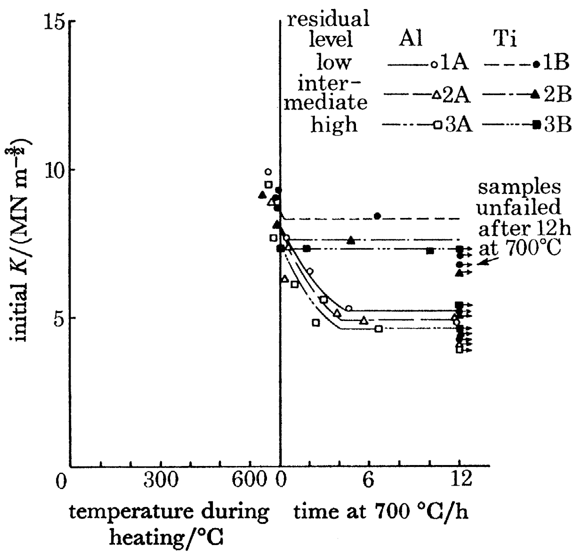

FIGURE 5. Effect of initial *K* on time to failure in hot crack opening displacement tests on simulated h.a.z. material. Heating was at 50 K/h.

(*e*) *Notched bend relaxation tests*

The notch opening displacement (n.o.d.) to give 10% cracking in the notched bend relaxation test (Batte *et al.* 1976) decreases with increasing susceptibility to stress relief cracking. In both coarse and fine grained materials, the titanium deoxidized casts were generally less susceptible to cracking than the aluminium deoxidized casts (table 4), except for the coarse grained condition for cast 2. However, there was no clear effect of residual element content. In the fine grained condition the titanium treated materials exhibited no cracking up to the highest n.o.d. (350 μm) applied (table 4), but nevertheless the present results lay within the range of values expected from results obtained in the past for commercial $\frac{1}{2}$%CrMoV steels.

(*f*) *Results of notched tensile relaxation tests* (*Murray tests*)

There was a large degree of scatter in the times to failure in the Murray tests and it was possible to construct C-curves (Murray 1967), only through the points for casts 1A and 3A (figure 6). However, the titanium deoxidized cast with low residual element content exhibited the greatest resistance to fracture with many samples not having fractured after 20 h, while

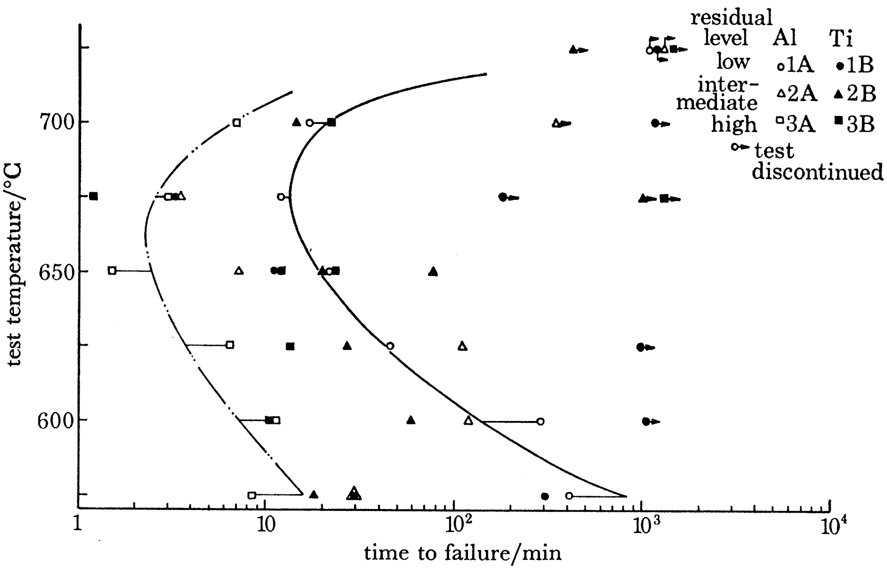

FIGURE 6. Murray notched relaxation test results for simulated h.a.z. material.

the high residual element content, aluminium deoxidized cast had, overall, the shortest times to failure with the 'nose' of the C-curve at 2–3 min.

DISCUSSION

Examination of both the room temperature and elevated temperature properties of normalized and tempered material has indicated that the properties of the present controlled test materials lay generally within the scatter bands of data for normal production ½%CrMoV steel. Within this range there was a small general increase in room temperature strength with increasing total residual element content, although the effect was negligible at temperatures of 500–600 °C. There was no consistent effect of residual element content or deoxidation practice on the tensile ductility or impact properties.

In creep rupture tests lasting beyond 1000 h at 550 °C on the normalized and tempered material, the aluminium deoxidized casts were less ductile than the titanium deoxidized casts (typically 15% elongation as against 50% elongation) (figure 2). However, this decrease in ductility was not associated with a reduction in rupture life under the temperature–stress combinations used. Only in the highest residual material deoxidized with aluminium (cast 3A) was there any indication that increased residual element content had a detrimental effect on rupture ductility. These observations are generally in line with the detrimental effect of aluminium and the beneficial effect of titanium on rupture ductility reported for bainitic CrMoV steels with 0.20–0.45% C (Ratliff & Brown 1967; Viswanathan & Beck 1975; Stone & Murray 1965). In contrast to the present work, Tipler and coworkers (1971, 1972, 1976) using both bainitic and ferritic CrMoV steels, and Bruscato (1970) using 2¼%CrMo weld deposits, have suggested significant effects of increasing residual element contents in reducing rupture strength and ductility, although Viswanathan (1975*a*, *b*) using 1¼%CrMo steel found no effect on rupture ductility of additions of phosphorus, tin and antimony. However, in each of these investigations, different combinations and ranges of the residual elements were studied or full analyses were not published. Detailed comparison cannot therefore be made.

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The tests on simulated h.a.z. material indicated that the aluminium deoxidized steels had a greater susceptibility to stress relief cracking than those deoxidized with titanium (table 4). Increasing residual element content within the range examined, however, appeared to have only a slightly deleterious effect (figure 7), whereas previous studies with the use of similar test tech-

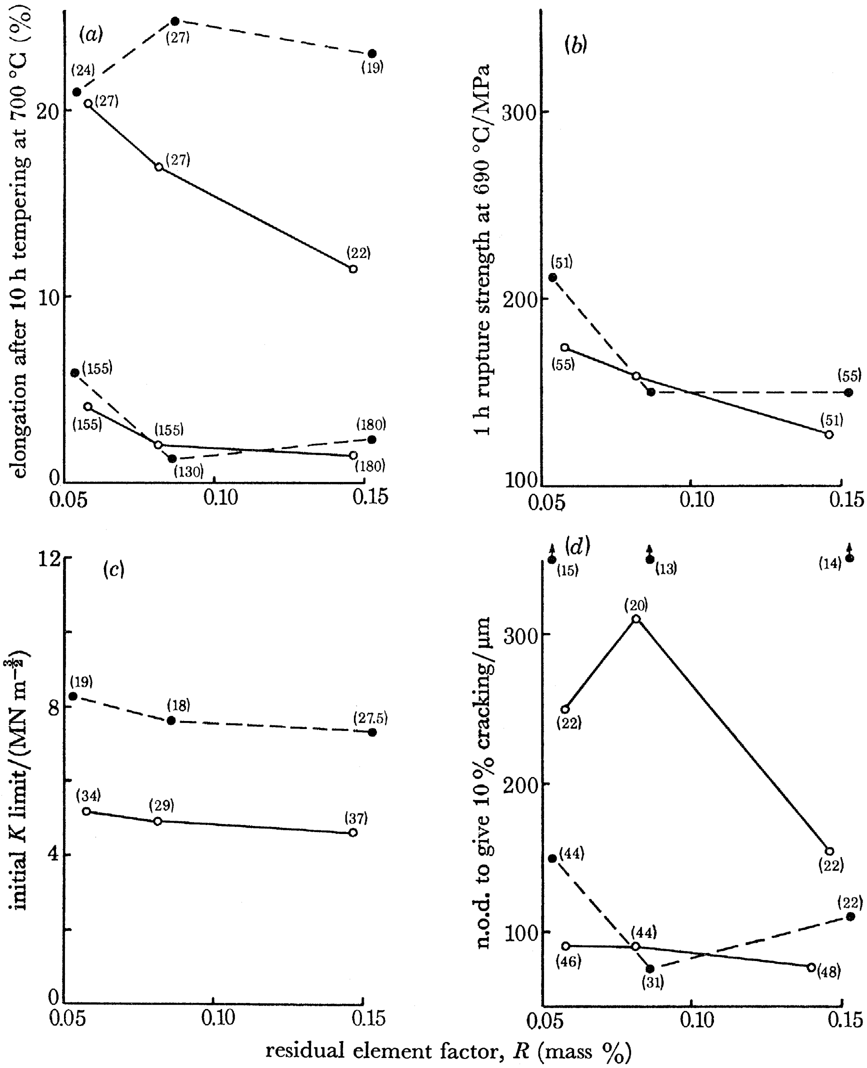

FIGURE 7. Variation in susceptibility to cracking of simulated h.a.z. material with residual element factor, R = [P] + 2.43[As] + 3.57[Sn] + 8.16[Sb] (percentages by mass), proposed by King (1976) in (*a*) tensile tests at 550 °C, (*b*) creep rupture tests at 690 °C, (*c*) hot crack opening displacement tests and (*d*) notched bend relaxation tests. Figures in parentheses are grain sizes in micrometres. ○, Al/Zr; ●, Ti/CaSiMn.

niques have indicated detrimental effects of both aluminium deoxidation and high residual element content (Harris & Jones 1972; King 1976; Myers 1972*b*; Townsend 1975). Nevertheless, in a range of production ½%CrMoV castings and forgings deoxidized with aluminium or silico-manganese there was little correlation between residual element content and stress relief cracking susceptibility (Miller & Batte 1975).

The present work has demonstrated again the importance of grain refinement in controlling stress relief cracking (Batte *et al.* 1976; King 1976). The titanium deoxidized specimens had finer austenitic grain sizes than the aluminium deoxidized specimens after the simulation heat treatments used for the hot crack opening displacement and notched bend relaxation tests (figure 7). This suggests that the relatively large effect of deoxidation practice seen in these tests could have resulted partly from grain refinement. Nevertheless, the tensile tests (figure 7) indicate that there was an effect of deoxidation practice over and above the grain refining action.

It appears from the results obtained that an improvement in the creep rupture ductility of normalized and tempered CrMoV castings would result if the present aluminium–zirconium practice were to be superseded by a titanium based practice, presuming the latter to meet all the foundry requirements. The results show that with either deoxidation practice there is no need to restrict residual element contents to limits below those currently obtained in order to control creep rupture ductility. Changing from aluminium based to titanium based deoxidation practice, however, would probably reduce the incidence of stress relief cracking, the inherent effect being amplified by the apparent grain refining effect of titanium deoxidation. The results confirm that an improvement in stress relief cracking resistance can be expected with decreasing h.a.z. grain size and, hence, that if adequate refinement of the structure can be achieved by appropriate welding procedures, sound weldments should result.

Conclusions

The effect of residual element content on the mechanical properties and susceptibility to stress relief cracking of ½%CrMoV castings deoxidized with either aluminium or titanium has been investigated. The residual element contents ranged from those associated with basic oxygen steel scrap to levels substantially higher than those common at present.

In normalized and tempered material, deoxidation with aluminium resulted in lower creep rupture ductility in times in excess of 1000 h at 550 °C than deoxidation with titanium, regardless of the residual element content. Increasing the residual element content reduced the creep rupture ductility slightly.

Various tests on simulated h.a.z. material, designed to assess the susceptibility to stress relief cracking, indicated that casts deoxidized with titanium exhibited a lower susceptibility to cracking than aluminium deoxidized material. Within the range of the results, increasing residual element content tended to increase the susceptibility to stress relief cracking, but a refinement in grain size markedly reduced the susceptibility.

It is apparent that changing from the aluminium to the titanium based deoxidation practice used in the present work could improve creep rupture ductility and stress relief cracking resistance in ½%CrMoV castings, but there does not at present appear to be a need for very low residual contents. Control of the welding process to maximize grain refinement should result in welds free from stress relief cracking.

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