Subzero tensile properties of vapour quenched aluminium alloys

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Supersaturated and metastable aluminium alloy solid solutions containing a dispersed phase have been produced by a vapour quenching technique. Binary alloys contained 3.5% Fe and 5.5% Mn; ternary alloys contained 6 to 9% chromium and 0.5 to 1.2% iron. After rolling into sheet the tensile properties were determined in the temperature range 293 to 77 K. At 77 K tensile strengths of 1115 and 1036 MPa were obtained for two Al–Cr–Fe alloys, equivalent to E/82 and E/83, respectively. These are the highest strengths ever reported for an aluminium alloy. The deformation behaviour at subzero temperatures has indicated the potential for further strengthening of metastable rapidly solidified aluminium alloys by dislocations alone.

1. Introduction

High strength Al–Cr–Fe alloys have been produced by an electron beam evaporation and vapour quenching technique [1, 2]. The microstructure of these alloys is unique, consisting of a supersaturated aluminium–chromium solid solution in which is dispersed a high volume fraction of small particles of an Fe–Cr phase. The tensile properties of the alloys at room temperature and at temperatures up to 300° C have been reported [2]. The alloys have been considered for airframe structures and since the latter are required to have adequate properties at temperatures down to − 50° C [3], some subzero tests have been carried out.

The deformation behaviour at subzero temperatures is also of considerable theoretical interest. Since thermally activated deformation processes become increasingly difficult with decreasing temperatures [4, 5], very high shear strengths and high strain hardening rates have been obtained in alloys tested at subzero temperatures [6]. In a review of the deformation mechanisms in metallic systems, Conrad [4] concluded that ultra high strengths might be achieved by deformation at very low temperatures and that a large fraction of this strength should be retained at temperatures up to 0.5 T_M, where T_M is the metal matrix melting point in degrees Kelvin. Some evidence for this increase in strength and its retention at room temperature has been reported in commercial purity titanium [7]. The exceptionally high strengths obtained in two phase alloys has also been attributed [8] to the additional geometrically necessary dislocations required in such materials. Since the vapour quenched alloys in the form of rolled sheet already contain a high density of dislocations and exhibit very high rates of strain hardening and high strength at room temperature, the deformation and strength of these alloys at subzero temperatures is of special interest and the results could have inplications for the production of very high strength materials.

This paper describes the sheet tensile properties of two binary and five ternary vapour quenched aluminium alloys in the temperature range 77 to 293 K. The properties are compared with those obtained for 1.78 mm thick sheet in alloys Al–6Zn–2.3Mg–1.7Cu (7010-T7651) Ti–6Al–4V and Ti–4Al–4Mo–2Sn–0.5Si (IMI 550) under the same conditions. A brief reference is made to the effect of rolling at low temperatures.

2. Experimental techniques

Aluminium alloys were electron beam evaporated and quenched from the vapour phase on to a substrate held at temperatures in the range 230 to 340° C. Two binary alloys (Al–5.5% Mn and Al–3.5% Fe) and one ternary alloy (HS23, Al–7.1% Cr–0.92% Fe) were quenched onto a stationary substrate [2] and after removing from the substrate were machined into small slabs and pressed and rolled to sheet 1.6 to 1.8 mm thick. The deposition and rolling conditions for these alloys are given in Table I.

Reciprocating substrate and scanning electron beam techniques were used to produce four ternary Al–Cr–Fe alloys deposits with reduced porosity, higher alloy content and higher strengths [2]. These alloys, designated SB24, SB43, SB45 and SB46, contained 6 to 9% chromium and 0.5 to 1.2% iron and had a layered microstructure, each layer being 70 to 80 μm thick. The deposits were removed from the substrate, machined into slabs and rolled. The substrate temperatures for these alloys are given in Table II. Since the composition varied slightly throughout a deposit, the composition of each test piece, determined using an electron microprobe analyser, is given in Table II.

At the deposition temperatures used for the Al–Cr–Fe SB alloys (270° C and below, Table II) the aluminium lattice parameter indicated [2] that almost all the chromium was retained in supersaturated solid solution whilst the iron was combined with chromium to form a uniform dispersion of very small precipitates

TABLE I Processing conditions for alloys produced on a stationary substrate

Deposit composition (wt %)	Substrate temperature (°C)	Working temperatures and reductions
Binary Al–5.5%Mn	305 to 340	Pressed 300° C, 28% reduction Rolled 300° C, 70% reduction
Binary Al–3.5 % Fe	310 to 320	Pressed 250° C, 64% reduction Rolled 200° C, 78% reduction
Ternary Al–7.1%Cr–0.92%Fe alloy (HS23)	320	Pressed 245° C, 72% reduction Rolled 230° C, 69 to 81% reduction

or zones 3 to 5 nm diameter (Fig. 1a). The Al–Cr–Fe alloy HS23 deposited at ~320° C still contained most of the chromium in solid solution and the fine dispersion, but the grain boundaries contained occasional coarser (200 nm diameter) $CrAl_7$ intermetallic particles associated with a solute depleted matrix (Fig 1b). The binary Al–Fe alloy HS18 also deposited at ~320° C, had a lattice parameter close to that of aluminium and contained a uniform distribution of 3 nm diameter particles within the grains and some 100 to 200 nm particles of $FeAl_6$ in the grain boundaries associated with denuded zones. The binary Al–Mn alloy HS7 deposited at a slightly higher maximum temperature of 340° C contained larger grains and coarser grain boundary precipitates than in the binary Al–Fe alloy. Working had little effect on the particle sizes but produced very small pancake shaped grains in all the alloys, with typical planar dimensions 1 to 4 μm and thickness 0.1 μm. The grains contained a very high density of dislocations.

The test piece gauge length was 30 mm long and 6 mm wide. Notched tensile test pieces were made by machining 60° edge notches to reduce the cross-sectional area by 30%; the calculated K_t was 9.2. The testing technique was identical to that described previously [3]. Tests were carried out at 77, 173, 193, 223 and 293 K.

Small Al–Cr–Fe (HS23) alloy specimens were also rolled at 293 K, cooled to 77 K in liquid nitrogen and rolled at a temperature close to 77 K. The latter involved rapid transfer (~2 sec) to a cold rolling mill which produced reductions per pass of about 1.5%. The specimens were re-cooled after each pass.

TABLE II Alloy compositions for Al–Cr–Fe alloys produced on a reciprocating substrate

Deposit	Test piece number	Composition	
		Chromium (wt %)	Iron (wt %)
SB24	24/3/15	6.04	0.53
Substrate	24/3/6	5.74	0.53
temperature	24/6D/2/2	7.10	0.76
270° C	24/6D/2/3	6.73	0.72
	24/5A/2		
	24/6D/2/4	7.23	0.86
	24/6D/2/6	6.98	0.71
	24/6B	7.11	0.77
	24/5B/3	6.60	0.70
	24/5B/1	7.04	0.77
	24/5B/2	6.97	0.78
	24/5A/3/1	6.33	0.63
	24/5A/1/1	6.73	0.71
	24/5A/1/2	6.41	0.63
SB43	43/4	7.62	1.0
Substrate	43/1	7.04	0.82
temperature	43/3	6.01	0.76
230 to 240° C	43/2	7.54	1.00
SB45		8.0	1.0
Substrate			
temperature			
240° C			
SB46	46/1	9.15	1.15
Substrate	46/2	9.02	1.08
temperature	46.4	8.53	1.01
240 to 260° C	46/3	8.48	0.98

3. Results

3.1. Tensile properties

The tensile properties of Al–Mn, Al–Fe and Al–Cr–Fe alloys at test tempertures in the range 77 to 293 K are given in Tables III to V. The tensile strength increased with decreasing test temperature (Fig. 2).

At the lower test temperature of 77 K, tensile strengths greater than 1000 MPa were obtained for Al–Cr–Fe alloys (Tables III to V) with total elongations of 2.5 to 9.0%. At 77 K the high strength alloys SB43, SB45 and SB46 fractured below the 0.1% proof stresses with zero elongation. Several test pieces of these alloys fractured in the loading holes. One SB45 test piece was machined with an increased cross-section in the head and a decreased cross-section in the gauge length. This test piece fractured simultaneously in the loading-hole and the gauge length when the stress in the gauge length was 1003 MPa.

The percentage increase in tensile strength with decrease in test temperature is shown in Table VI. Decreasing the test temperature from 293 to 173 K increased the tensile strength of the Al–Cr–Fe alloys by 20 to 25% and decreasing the test temperature to 77 K increased the tensile strengths of the binary Al–Mn alloy and the ternary Al–Cr–Fe alloys by 65 to 70%. These increases were about twice the increase for the 7010 aluminium alloy (28%) and greater than the increase for the Ti–6Al–4V alloy (60%) [3]. The smaller increase for the binary Al–Fe alloy was probably due to the lower dispersed phase volume fraction at the lower alloy content. However, the tensile strength of this alloy at 293 K (636 MPa) and 77 K (848 MPa) (Table III) was impressive for such a dilute alloy, and comparable with the strength obtained for the more highly alloyed ternary alloy (Table IV).

Notched tensile data were only obtained for the SB24 alloy and were limited to two test temperatures of 293 and 223 K. The results are given in Table VII. At both temperatures the tensile strengths were

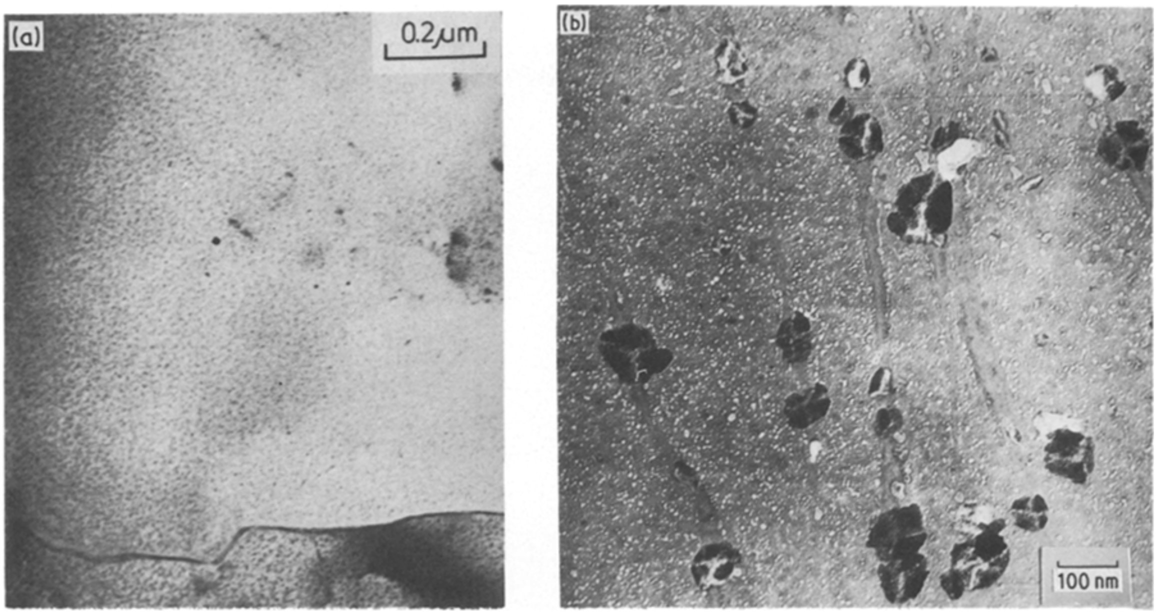

Figure 1 (a) Transmission electron micrograph showing fine Fe–Cr precipitates in Al–Cr–Fe alloys, (b) replica showing denuded zones around $CrAl_7$ precipitates in Al–Cr–Fe alloys.

greater for the notched (NTS) than the unnotched (TS) test pieces, with ratios NTS/TS of 1.12 to 1.13. The ratio NTS/PS (PS = 0.2% proof stress) was slightly greater than the corresponding ratios for 7010 and 7075 aluminium alloys, although the tensile strength was about 8% higher for the SB24 alloy. These results show that the SB 24 ternary alloy was not notch sensitive in the temperature range 293 to 223 K.

The ductility parameters are compared over the temperature range 293 to 77 K in Fig. 3. There was a trend towards increasing uniform elongation with decreasing test temperature for some SB24 alloy test pieces (Fig. 3a) but there was little change in the total elongation. In the high strength ternary alloys (SB43, SB45, SB46) the uniform and total elongation were not very dependent on test temperature down to 173 K. Both were zero at 77 K when the deformation to maximum load appeared to be completely elastic.

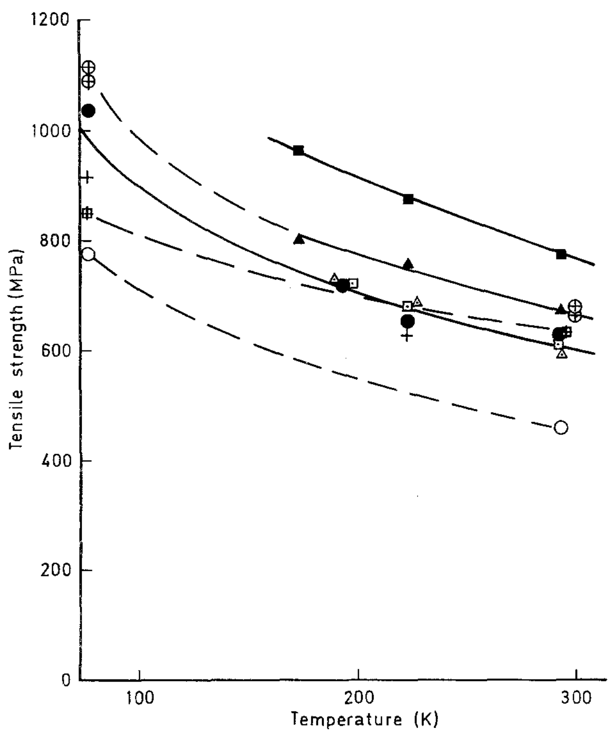

Figure 2 Tensile strength against test temperature. (○) Al–Mn; (⊞) Al–Fe; (⊕) HS23, Al–Cr–Fe; (+, □, △, ●) SB24, Al–Cr–Fe; (▲) SB43, Al–Cr–Fe; (■) SB46, Al–Cr–Fe.

The reduction of area for the SB24 alloy decreased with decreasing test temperature, the decrease being particularly marked in the cold rolled material (Fig 3c). The reduction of area values for the high strength alloys were not very dependent on temperature down to 173 K but at 77 K very low values were obtained (Fig. 3c).

The ratios of uniform/total elongation are compared in Table VIII. In the temperature range 293 to 173 K ratios were in the range 0.07 to 0.38 for the alloys SB24, SB43 and SB46 compared with ratios of 0.72 to 0.83 for the 7010 alloy and 0.54 to 0.67 for Ti–6Al–4V alloy. For the lower strength SB24 alloy at 77 K, the ratio increased to about 0.8. This value was greater than the ratio obtained for Ti–6Al–4V

TABLE III Tensile properties of binary and ternary alloy sheet

Alloy	Test temperature (K)	Tensile strength (MPa)	Elongation (%)
Binary Al–5.5%Mn	293	455	4.2
	77	775	6.0
Binary Al–3.5%Fe	293	636	10
	77	848	5
Ternary Al–7.1%Cr–0.92%Fe (HS23)	293	662	6
		667	10
	77	1090	2.5
		1115	8

TABLE IV Tensile properties of Al–Cr–Fe alloy (SB24) sheet

	Test piece number												
	6D/2/2	5B/3	6B	5A/3/1	3/15	6D/2/3	5B/1	5A/1	2/4	5B/2	5A/1/2	3/16	2/5
Symbol for working conditions	●	△	△	□	+	●	△	□	●	△	□	+	●
Rolling temperature (° C) and rolling reduction	350/17% 320/83%	350 to 370/23% 320/84%		350/21% 320/66% 20/55%	320/79%								
Test temperature (K)	293				223				193			77	
Tensile strength (MPa)	627	598	586	603	616	653	682	676	718	718	712	915	1036
0.5% proof stress (MPa)	627	596	577	600	615	649	677	674	711	702	709	887	972
0.2% proof stress (MPa)	596	551	532	564	598	609	647	632	680	674	678	844	930
0.1% proof stress (MPa)	548	481	466	515	563	553	600	574	634	631	629	794	882
Young's modulus (GPa)	←	83		→	87.5	←	86.7	→	←	88.3	→	91.5	92.3
Uniform elongation (%)	0.7	0.9	1.2	1.0	3.8	1.1	2.9	1.1	1.5	3.9	1.5	3.3	7.6
Total elongation (%)	4	8	10	10	10	3.0	9.0	5.0	7.5	10.5	8.0	4.5	9.0
Reduction of area (%)	42	43	52	43	34.4	23.4	27.9	15	31	32	14.8	10	14

TABLE V Tensile properties of Al–Cr–Fe alloy (SB43, SB46, SB45) sheet (rolling temperature 290° C, rolling reduction 78%)

	Alloy							
	SB43			SB46			SB45	
Test piece number	4	1	3	1	2	4	1	2
Test temperature (K)	293	223	173	293	223	173	293	77
Tensile strength (MPa)	663	755	799	769	869	958	695	1003
0.5% proof stress (MPa)	–	–	793	–	866	954	689	–
0.2% proof stress (MPa)	628	729	779	723	823	902	652	–
0.1% proof stress (MPa)	569	687	737	654	750	820	613	–
Young's modulus* (GPa)	85.5	88.7	91.0	88.4	93.5	95	88	97
Uniform elongation (%)	0.39	0.44	0.75	0.42	0.71	1.04	1.1	1.0
Total elongation (%)	6	6	7	5	5	7	5	1.0
Reduction of area (%)	27	27	27	28	35	18	29	–

*Young's modulus at 77 K was 96 GPa and 98 GPa for alloys SB43 and SB46, respectively.

TABLE VI Increase in tensile strength with decrease in test temperature to 173 K (σ_{173}) and 77 K (σ_{77})

Alloy	Strength increase at 173 K		Strength increase at 77 K	
	($\sigma_{173} - \sigma_{293}$) (MPa)	(%)	($\sigma_{77} - \sigma_{293}$) (MPa)	(%)
Vapour quenched alloys				
Al–Mn			320	70
Al–Fe			212	33
Al–Cr–Fe (HS23)	140*	21*	448	67
Al–Cr–Fe (SB24)	140*	23*	409	65
Al–Cr–Fe (SB43)	136	20		
Al–Cr–Fe (SB46)	189	25		
Conventional alloys				
7010	69	12	163	28
Ti–6Al–4V	241	23	618	60

*Interpolated from curves in Fig. 2.

and slightly smaller than the ratio for the 7010 alloy at this temperature.

In the temperature range 173 to 293 K, Young's modulus values (E) for the alloys SB24, SB43 and SB46 showed a linear dependence on test temperature (T) (Fig. 4); the dependence was described by the expression

$$E = M_1 T + C_1 \quad (1)$$

where M_1, C_1 were constants independent of temperature but dependent on alloy content. Values for the constants M_1 and C_1 are given in Table IX; M_1 values were between -0.037 and -0.051 GPa °C^{-1}. These values were greater than the value obtained for the 7010 alloy (-0.032 GPa °C^{-1}) but less than the value obtained for Ti–6Al–4V (-0.061 GPa °C^{-1}).

For many metals the yield stress is linearly dependent on test temperature at subzero temperatures. A good approximation to the yield stress is the 0.1% proof stress ($\sigma_{0.1}$). Values for $\sigma_{0.1}$ are plotted against test temperature in Fig. 5a. A linear regression analysis of the data showed that in the temperature range 175 to 293 K, $\sigma_{0.1}$ was related to temperature by the expression

$$\sigma_{0.1} = MT + C \quad (2)$$

where M and C were constants dependent on the alloy Table X). The temperature dependence (M in Equation 2) for the vapour quenched alloys was about twice that found for the 7010 alloy and about half that found for the Ti–6Al–4V alloy.

The flow stress of an alloy depends on Young's modulus which is itself dependent on temperature; the modulus corrected yield stress ($\sigma_{0.1}/E$) is plotted in Fig. 5b. A significant temperature dependence was still apparent in the vapour quenched aluminium alloys, but was less in the 7010 alloy, suggesting that a greater part of the temperature dependence of $\sigma_{0.1}$ in the vapour quenched alloys was caused by factors other than modulus. The temperature dependence of $\sigma_{0.1}/E$ was greater for the Ti–6Al–4V alloy than for the aluminium alloys.

3.2. Strain hardening behaviour

A measure of the mean strain hardening rate during 0.1% plastic strain in the early stages of plastic deformation can be obtained from the difference between the 0.2% proof stress ($\sigma_{0.2}$) and the 0.1% proof stress ($\sigma_{0.1}$), i.e.

$$\frac{d\sigma}{d\varepsilon_{0.1}} = (\sigma_{0.2} - \sigma_{0.1}) \times 10^3 \quad (3)$$

The maximum strain hardening under uniform strain conditions in a tensile test (see Appendix) is given by

TABLE VII Notched tensile strength (NTS) of SB24 alloy and 7010 and 7075 alloys at 223 K (– 50° C)

	Alloy				
	Al–Cr–Fe alloy SB24		7010–T7651	7075–T6	
Test temperature (K)	293	223	293	293	223
Test piece number	24/3/13	24/6D/2/3			
Tensile strength, TS (MPa)	582	653	537	540	
0.2% proof stress, PS (MPa)	570	609	472	485	
NTS (MPa)	653	739 (24/5A/2)	529	507	
NTS/TS	1.12	1.13	0.99	0.94	
NTS/PS	1.14	1.21	1.12	1.05	1.01

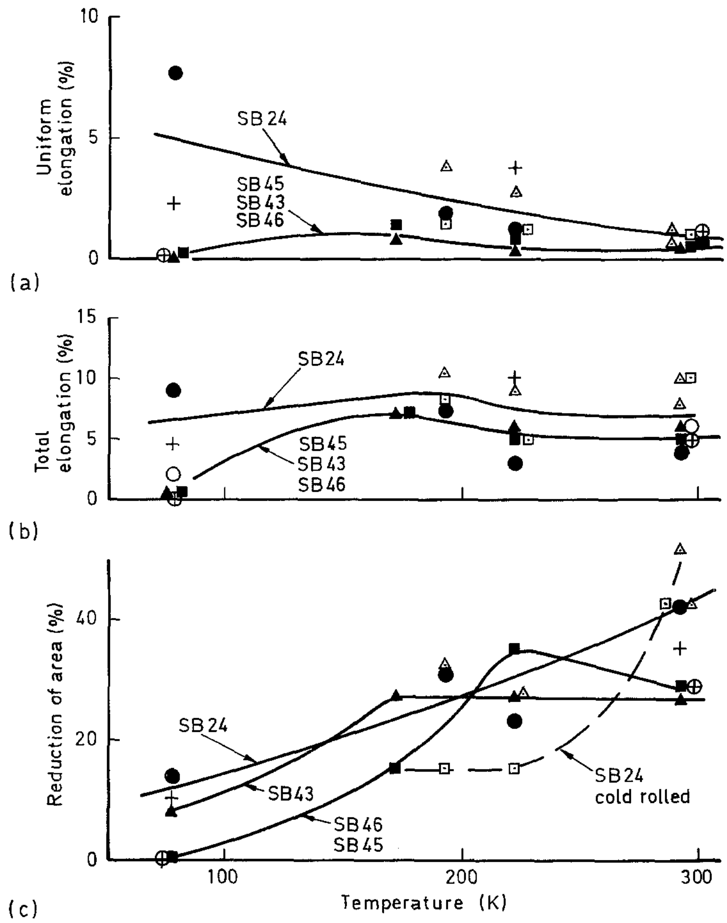

Figure 3 Ductility parameters against test temperature for Al–Cr–Fe alloys. (○) HS23; (⊕) SB45; (▲) SB43; (■) SB46; (+, ●, ⊡, △) SB24.

$$\Delta\sigma = (\bar{\sigma}_{TS} - \sigma_{0.1}) \tag{4}$$

where $\bar{\sigma}_{TS} = \sigma(1 + \varepsilon_u)$ and $\bar{\sigma}_{TS}$ is the true stress at maximum load, σ is the nominal stress at maximum load, ε_u is the uniform elongation.

The parameters $d\sigma/d\varepsilon_{0.1}$ and $\Delta\sigma$ are plotted against test temperature in Fig. 6. In the temperature range 77 to 293 K strain hardening rates of 40 to 60 GPa were obtained for the SB24 and SB43 alloys, but very high values of 68 to 85 GPa were obtained in the range 173 to 293 K for the SB46 alloy, and these high values were greater than the values for 7010 and for Ti–6Al–4V (Fig. 6b). However, at 77 K values of 130 GPa were obtained for Ti–6Al–4V; this high rate of strain hardening is well documented for titanium alloys at cryogenic temperatures and has been shown to be associated with multiple slip systems and twinning deformation [6].

The maximum strain hardening $\Delta\sigma$ for the vapour quenched aluminium alloys was not very sensitive to temperature in the range 223 to 293 K, but increased rapidly in alloy SB46 below 233 K and in alloy SB24 below 173 K (Fig. 6c). In the temperature range 223 to 293 K the $\Delta\sigma$ values were less in the vapour quenched alloys than in the 7010 alloy (70 to 135 GPa and 175 GPa, respectively). The $\Delta\sigma$ values were greater in the titanium alloy and increased rapidly below 173 K.

The ratio 0.2% proof stress/tensile strength (PS/TS ratio) is often used to compare the strain hardening of metals [11] (see Appendix). Values for this ratio are given in Table XI for the three alloy types. The ratio for the vapour quenched alloys (0.90 to 0.97) was greater than for 7010 (0.84 to 0.86) and similar to Ti–6Al–4V (0.92 to 0.95). However when corrections were made for true tensile stress (see Appendix) the ratios were reduced for the 7010 and Ti–6Al–4V

TABLE VIII Effect of test temperature on the ratio uniform elongation/total elongation

Test temperature (K)	Ratio uniform elongation/total elongation				
	Evaporated aluminium alloys			7010	Ti–6Al–4V
	SB24	SB43	SB46		
293	0.10 to 0.18	0.07	0.08	0.72	0.54
223	0.22 to 0.38	0.07	0.14	0.82	0.67
193	0.19 to 0.33	–	–		
173	–	0.11	0.15	0.83	0.67
77	0.73 to 0.84	–	–	0.95	0.55

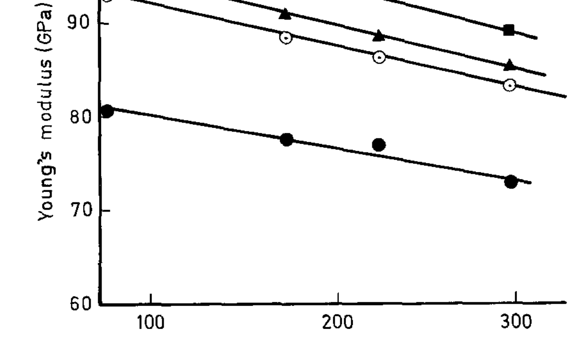

Figure 4 Young's modulus against test temperature for Al–Cr–Fe alloys and 7010. (■) SB46; (▲) SB43; (⊙) SB24; (●) 7010.

TABLE IX Values for the constants M_1 and C_1 in equation $E = M_1T + C_1$, Equation 1

Alloy	Constants	
	M_1 (10^{-2} GPa °C^{-1})	C_1
SB24*	−3.69	95.1
SB43*	−4.86	99.6
SB46‡	−5.09	104.2
7010*	−3.2	83.2
Ti–6Al–4V‡	−6.1	128

*Applicable in the temperature range 293–77 K.
‡Applicable in the temperature range 293–173 K.

TABLE X Values for the constants M and C in the equation $\sigma_{0.1} = MT + C$, Equation 2

Alloy	M (MPa °C^{-1})	C	Temperature range (K)
Vapour quenched alloys			
SB24	−1.24	861	293 to 173
SB43	−1.42	990	293 to 173
SB46	−1.38	1059	293 to 173
Conventional alloys			
7010	−0.68	639	293 to 77
Ti–6Al–4V	−2.55	1605	293 to 77

alloys, indicating greater hardenability in these alloys. Since $\Delta\sigma$ is based upon yielding at $\sigma_{0.1}$ and the true tensile strength, the $\Delta\sigma$ parameter is believed to be a better measure of strain hardening than the PS/TS ratio, especially when comparing materials with a high rate of strain hardening, as discussed in the Appendix.

3.3. Elastic strain energy

The elastic strain energy associated with the plastic deformation of the vapour quenched alloys was much greater than that found in conventional aluminium alloys [3, 12]. The elastic strain energy/unit volume at the 0.1% proof stress is equal to $\frac{1}{2}\sigma_{0.1}^2/E$; the alloys therefore lie between the conventional 7010 alloy and Ti–6Al–4V alloy (Fig. 7a). However at the maximum load in a tensile test the ratio

$$\frac{\text{elastic strain energy/unit volume}}{\text{total (elastic + plastic) strain energy/unit volume}}$$

was 3 to 6 times greater for vapour quenched alloys SB43 and SB46 than for alloys 7010 and Ti–6Al–4V (Fig. 7b). This reflected the combination of small uniform strain and high tensile strength in the vapour quenched alloys.

The strain energy/unit weight in the higher strength vapour quenched alloy SB46 was similar to that for Ti–6Al–4V and much greater than the values for 7010 and a typical steel (EN42) used for springs. Since vapour quenched alloys have good fatigue and corrosion resistance [2], the high strength alloys could be considered for lightweight springs.

3.4. Effect of rolling at low temperatures on the tensile strength

The room temperature tensile properties of the Al–Cr–Fe alloy HS23 given a final 34% rolling reduction at either 293 K or approximately 77 K are given in Table XII. The specimen rolled at 77 K had a significantly higher tensile stength, with little change in tensile ductility, compared with the test piece rolled at 293 K. The strength increase was due to an increase in the yield stress ($\sigma_{0.1}$) of 147 MPa; the initial strain hardening rate ($\sigma_{0.2} - \sigma_{0.1}$) and the maximum strain

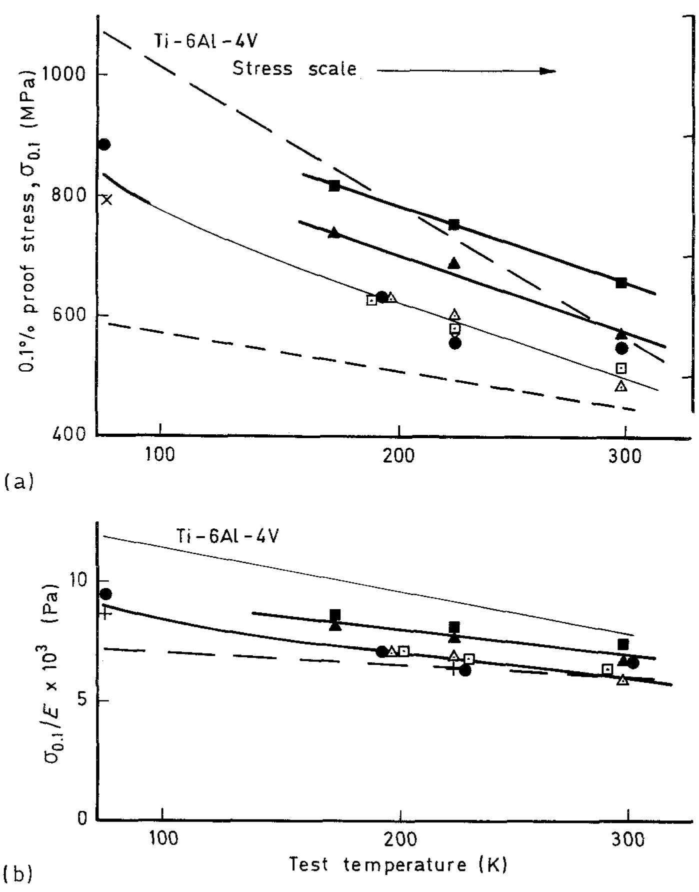

Figure 5 Temperature dependence of the 0.1% proof stress ($\sigma_{0.1}$) for Al–Cr–Fe alloys, 7010 and Ti–6Al–4V alloys (a) $\sigma_{0.1}$ against temperature (b) $\sigma_{0.1}/E$ against temperature. (○) SB45; (■) SB46; (▲) SB43; (●, +, △, ⊡) SB24; (– – –) 7010.

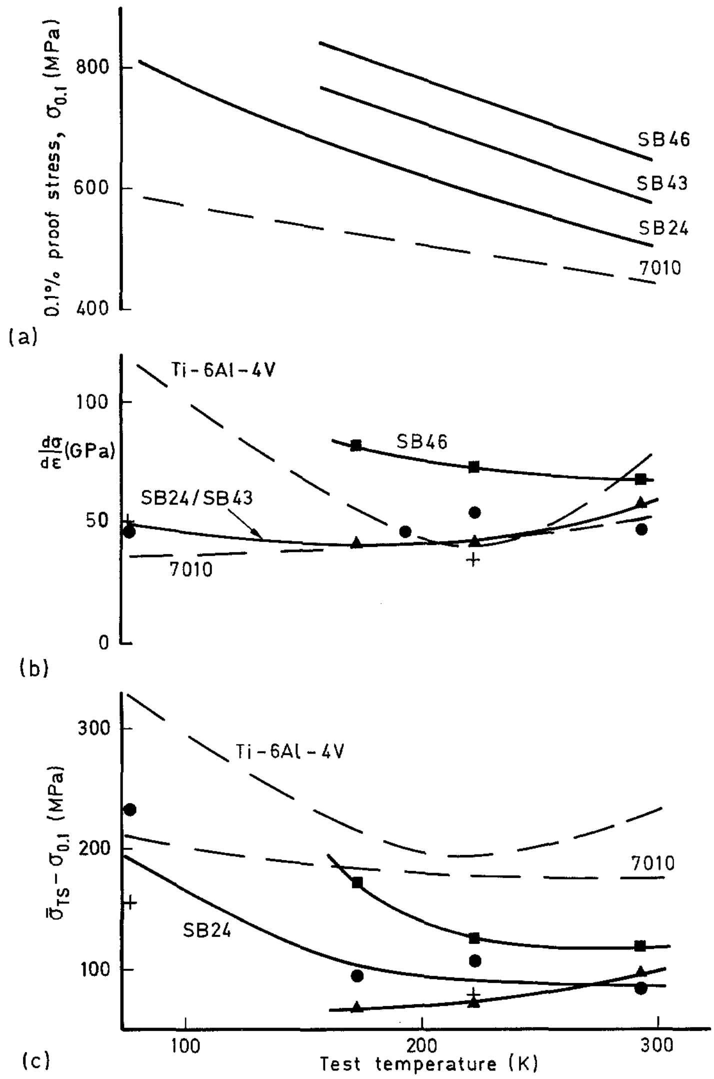

Figure 6 Temperature dependence of (a) $\sigma_{0.1}$, (b) strain hardening rate $d\sigma/d\varepsilon$ (■) SB46; (▲) SB43; (●) SB24, (c) strain hardening ($\bar{\sigma}_{TS} - \sigma_{0.1}$).

hardening ($\Delta\sigma = \bar{\sigma}_{TS} - \sigma_{0.1}$) were virtually the same as for the test piece rolled at 293 K. The slightly higher alloy content made some contribution to the strength of the 77 K rolled material, but could not acount for the magnitude of the strength increase. This is apparent in the summary Table XIII, in which the tensile and 0.1% proof stresses for several different Al–Cr–Fe alloy compositions are compared. The strengths for the 77 K rolled material (test piece 23/15/1B) were greater than those for the higher chromium test piece (SB24/6D/2/2) or the higher iron test piece (61/4/1).

After numerous tests on the Al–Cr–Fe alloy and when differences in alloy composition were taken into account, no significant increase in tensile strength was obtained by rolling at room temperature instead of 200 to 320° C; an example for the SB24 alloy is provided by the test pieces SB24/5A/3/1 and SB24/6D/2/2 in Tables II and IV and compared in Table XIII. However there was evidence for a reduction in the non-uniform elongation of cold rolled material at subzero test temperatures (Fig. 3c). An indication of the effect of rolling alloy HS23 at 293 and 503 K is provided by test pieces 23/15/1A or 23/19/1 in Table XIII; the tensile strengths were similar for the two conditions.

These results show that the tensile properties were not dependent on rolling temperatures in the range 293 to 593 K. But rolling at a very low temperature (77 K) can introduce a still higher dislocation density into the metastable alloy microstructure compared with that produced by rolling at 293 K or above. The

TABLE XI Ratio 0.2% proof stress/tensile strength in the temperature range 293 to 77 K

Test temperature (K)	Ratio = 0.2% proof stress/tensile strength				
	Vapour quenched alloys			Conventional alloys	
	SB24	SB43	SB46	7010	Ti–6Al–4V
293	0.91 to 0.95	0.95 (0.94)	0.94 (0.94)*	0.86 (0.80)	0.92 (0.86)
223	0.93 to 0.97	0.97 (0.96)	0.95 (0.94)	0.86 (0.80)	0.94 (0.88)
193	0.94 to 0.95				
173		0.97 (0.97)	0.94 (0.91)	0.86 (0.79)	0.95 (0.89)
77	0.90 to 0.92			0.84 (0.76)	0.93 (0.88)

*Ratios in brackets based upon true tensile strength, i.e. $R' = \sigma_{0.2}/\bar{\sigma}_{TS}$.

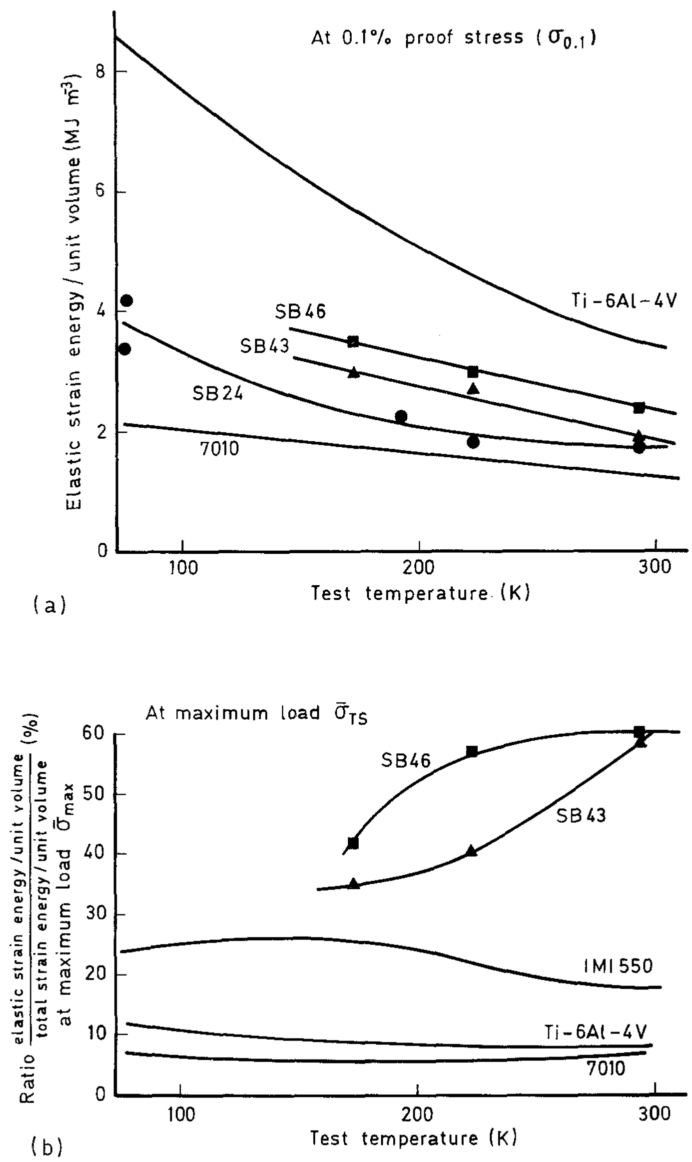

Figure 7 Elastic strain energy/unit volume at $\sigma_{0.1}$ and $\bar{\sigma}_{TS}$.

higher dislocation density is retained at room temperature and leads to an increase in the yield strength and hence the tensile strength.

4. Discussion

Airframes are required to operate at temperatures down to -50° C (223 K). The present results show that there is no significant change in the ductility or strain hardening for the vapour quenched alloys at temperatures down to -73° C (200 K), Figs 3 and 6, although the modulus and strength parameters show a linear increase with decreasing temperature according to Equations 1 and 2. The plane and notched (SB24 data only) tensile properties of these alloys were satisfactory at temperatures down to -50° C (223 K).

For space applications structures may be required to withstand temperatures of -150° C (123 K). At this temperature the non-uniform ductility of the vapour quenched alloys was decreasing and the strength increasing more rapidly than at higher temperatures and the high strength alloys (SB43, SB45 and SB46) failed in a brittle manner at -196° C (77 K). The vapour quenched alloys tested in the present programme are therefore unsuitable for use at these very low temperatures. However, elongations of 4 to 9% were obtained at 77 K in the alloys SB24 and HS23 with a tensile strength of 1036 to 1115 MPa.

The brittle behaviour of SB43, SB45 and SB46 alloys could be associated with residual porosity or with contamination at the layer interface. In addition, the shear bands developed during rolling, which were associated with grain boundary precipitation, could lead to poor ductility in these very small areas. If the tensile strength against temperature curve for the

TABLE XII Effect of rolling at room temperature and approximately 77 K on the room temperature tensile properties of Al–Cr–Fe alloys HS23. The initial pressing temperature and reduction were 473 K and 62%, respectively. The initial rolling temperature and reduction were 503 K and 49%, respectively

	Final rolling temperature and reduction	
	293 K, 34%	77 K, 34%
Test piece number	23/15/1A*	23/15/1B†
Tensile strength (MPa)	570	710
0.2% proof stress (MPa)	556	700
0.1% proof stress (MPa)	521	668
Young's modulus (GPa)	78	81
Uniform elongation (%)	0.5	0.4
Total elongation (%)	5	4

*Composition (wt %) 5.75% Cr, 0.75% Fe
†Composition (wt %) 7.06% Cr, 0.84% Fe

TABLE XIII Tensile strength and 0.1% proof stress for Al–Cr–Fe alloys rolled at high and low temperatures

	Reciprocating substrate		Stationary substrate			
Test piece number	SB24/6D/2/2	SB24/5A/3/1	23/15/1A	23/15/1B	23/19/1	61/4/1
Composition (wt %): chromium	7.10	6.33	5.75	7.06	5.78	6.93
iron	0.76	0.63	0.75	0.84	0.66	1.07
Pressing temperature (° C) and reduction	350, 17%	350, 17%	200, 62%	200, 62%	260, 38%	270, 36%
Rolling temperature (° C) and reduction: initial	320, 83%	320, 83%	230, 49%	230, 49%	230, 86%	270, 79%
final		20, 55%	20, 34%	−196, 34%		
Tensile strength (MPa)	627	603	570	710	572	627
0.1% proof stress (MPa)	548	515	521	668	535	598

SB46 alloy (Fig. 2) is extrapolated to 77 K a value of about 1260 MPa is obtained. It is unlikely that this very high strength could be achieved with defects in the alloy. The conventional 7010 alloy also fractured in a brittle manner at 77 K [12]; fracture occurred through the grain boundary denuded zones. Hence even in defect free vapour quenched alloys, the denuded zones associated with grain boundary precipitates could give rise to lower than expected strengths.

The tensile deformation behaviour of the vapour quenched alloys at subzero temperatures had many of the characteristics of the Ti–6Al–4V alloys. The temperature dependence of the proof stress and Young's modulus for the vapour quenched alloys was intermediate between that for 7010 and that for Ti–6Al–4V (Tables IX and X). The increase in the strain hardening $\Delta\sigma$ (Fig. 6c) and the magnitude of the strength increase (Table VI) were similar to that found in Ti–6Al–4V.

The tensile strengths of 1036 and 1115 MPa obtained for SB24 and HS23 alloys, respectively, at 77 K (Tables III and IV) are the highest strengths ever reported for an aluminium alloy. The maximum tensile strengths of alloys are normally between $E/100$ and $E/1000$ compared with the theoretical value of about $E/10$ [13, 14]: when the above tensile strengths are converted to true stress, they become $E/83$ and $E/82$, respectively. The corresponding values at 293 K are less than $E/100$, as are the tensile strengths of 7010 alloy at all test temperatures. The tensile strength of Ti–6Al–4V was $E/103$ at 293 K and $E/69$ at 77 K. Thus the strengths obtained for the vapour quenched and rolled alloys at 77 K were a greater fraction of the theoretical strength than has ever hitherto been obtained for an aluminium alloy and were approaching the value for Ti–6Al–4V at the same temperature.

The remarkable strength obtained for the dilute Al–3.5% Fe alloy (636 MPa at 293 K and 848 MPa at 77 K, Table III) is an example of the efficient use of an alloying element to strengthen an alloy with a dispersed phase and a high dislocation density.

The tensile strength can be considered [3, 12] to be composed of a yield stress term ($\sigma_{0.1}$) and a strain hardening term $\Delta\sigma$, i.e.

$$\bar{\sigma}_{TS} = \sigma_{0.1} + \Delta\sigma \qquad (5)$$

where $\bar{\sigma}_{TS}$ is the true stress, $\sigma_{0.1}$ is the 0.1% proof stress and $\Delta\sigma = (\sigma_{0.2} - \sigma_{0.1})$. In the vapour quenched alloys $\Delta\sigma$ was not very dependent on test temperature down to $-75°$ C (Fig. 6c) and was about half the $\Delta\sigma$ value for 7010 alloy. The increase in strength with decreasing temperature was caused largely by the increase in $\sigma_{0.1}$ (Fig. 5a). Below $-75°$ C both $\Delta\sigma$ and $\sigma_{0.1}$ contribute to the strength. At 293 K $\Delta\sigma$ contributed about 15% to the strength of the vapour quenched alloys compared with 28% for 7010 and 21% for Ti–6Al–4V. At 77 K, however, $\Delta\sigma$ contributed 16 to 21% compared with 25% for 7010 and 14% for Ti–6Al–4V.

The deformation characteristics of the vapour quenched alloys are consistent with the high density of dislocations present in the microstructure in the as-rolled state* and the high value of Young's modulus for these alloys, since [18]

$$\Delta\sigma = \bar{\sigma}_{TS} - \sigma_{0.1} \approx \alpha\mu f^{3/2}\varepsilon \qquad (6)$$

where α is a constant, μ is the shear modulus, f is the volume fraction of particles and ε is the strain which is proportional to the dislocation density.

The ability to increase the room temperature strength by very low temperature rolling (Table XII) indicates the potential for further strengthening of the metastable alloys by dislocations alone as reported for conventional alloys [7, 15, 16, 19].

A high dislocation density in conventional alloys tends to be associated with a high yield stress; the high strength levels attained in the present work, with sometimes virtually elastic behaviour up to 0.1% strain (Section 3.1), was consistent with this tendency. A very high concentration of thermally stable dislocations may be unavoidable in high strength dispersion hardened marterials after working and for optimum mechanical properties precise control of the solute concentration and of the dispersed phase volume fraction will be necessary. It may only be possible to achieve this degree of control in microstructures produced by the vapour quenching technique.

5. Conclusions

1. The tensile properties of some vapour quenched binary and ternary aluminium alloys have been determined in the temperature range 77 to 293 K. At the

*The dislocation configuration developed by rolling differs from that produced by uniaxial tension [15, 16]. Thus some uniform elongation measured in the tension test probably arises from rearrangement of the pre-existing dislocation arrays; this would not produce significant strain hardening. Thereafter the strain would give rise to normal strain hardening according to Equation 5: the magnitude of the strain hardening in then very high in the vapour quenched alloys [17].

minimum operating temperatures for airframes, 123 K (− 50° C), all the ternary alloys had satisfactory tensile properties.

2. The strength of the alloys increased with decrease in test temperature. At 77 K the increase was 65 to 70% for the vapour quenched alloys compared with increases of 28% for 7010 and 60% for Ti–6Al–4V.

3. Tensile strengths at 77 K for two Al–Cr–Fe alloys were 1115 MPa (HS23 alloy) and 1036 MPa (SB24 alloy) and were the highest strengths ever reported for an aluminium alloy.

4. The temperature dependence of the 0.1% proof stress in the temperature range 173 to 293 K for the vapour quenched Al–Cr–Fe alloys was about twice that for 7010 and half that for Ti–6Al–4V.

5. In the temperature range 173 to 293 K vapour quenched Al–Cr–Fe alloys exhibited high elastic strain energy densities and could be suitable for springs.

6. The strength of Al–Cr–Fe alloys at 293 K was not increased by rolling at 293 K but was increased substantially by rolling at ~77 K.

7. The deformation behaviour at subzero temperatures has indicated the potential for further strengthening of metastable evaporated aluminium alloys by dislocations alone.

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Appendix
Strain hardening in the tensile test
The strain hardening $\Delta\sigma$ in a tensile test is the difference between the yield stress and the true tensile stress at maximum load ($\bar{\sigma}_{TS}$). In practice it is convenient to take the yield stress as the 0.1% proof stress ($\sigma_{0.1}$) and then

$$\Delta\sigma = (\bar{\sigma}_{TS} - \sigma_{0.1}) \tag{A1}$$

Normalizing with respect to the tensile stress gives

$$\frac{\bar{\sigma}_{TS} - \sigma_{0.1}}{\bar{\sigma}_{TS}} = \left(1 - \frac{\sigma_{0.1}}{\bar{\sigma}_{TS}}\right) \tag{A2}$$

The ratio $\sigma_{0.1}/\bar{\sigma}_{TS}$ is analogous to the proof/ultimate ratio (PS/TS) frequently reported for metals, which is based upon the 0.2% proof stress (PS) and the nominal tensile stress (TS) at maximum load. The ratio PS/TS is sometimes used to indicate the strain hardenability of an alloy; the smaller the ratio the greater the hardenability.

A single proof/ultimate ratio can indicate different strain hardening ($\Delta\sigma$) values and strain hardening rates and neither $\Delta\sigma$ nor PS/TS ratio take account of the uniform elongation, which is important when toughness is required. These features are illustrated schematically in Fig. A1, which depicts stress–strain curves with the same PS/TS ratios but different $\bar{\sigma}_{TS}$ (at A and B) or different ε_u at A and C. Since

$$\bar{\sigma}_{TS} = \sigma(1 + \varepsilon_u) \tag{A3}$$

The correction factor for true stress is small for vapour quenched aluminium alloys ($\varepsilon_u \leqslant 0.01$) but significant for conventional structural alloys ($\varepsilon_u \sim$ 0.05 to 0.10). When comparing such materials it is necessary to correct for true stress.

With increasing cold rolling reduction the uniform elongation decreases and the PS/TS ratio increases indicating reduced hardenability. The small uniform elongation and high PS/TS ratio typical of vapour quenched aluminium alloys is characteristic of cold worked materials. However, certain Ti–Mo alloys, heat-treated to a high strength, have small ε_u and $\Delta\sigma$ values and high PS/TS values, but fracture in a brittle mode [9], whilst a combination of a dispersed phase and thermomechanical working can lead to strain induced transformations in steels which produce increased strength and PS/TS ratios without sacrificing ductility [10]. Thus ideally stress–strain curves should combine a high proof stress with adequate $\Delta\sigma$ and ε_u values. Although it is difficult to give a weighting for each of these parameters, typical target values for advanced alloys might be $\sigma_{0.2} = (E/100)$, $\varepsilon_u \geqslant 0.01$ and $\Delta\sigma \simeq E/1000$, where E = Young's modulus.

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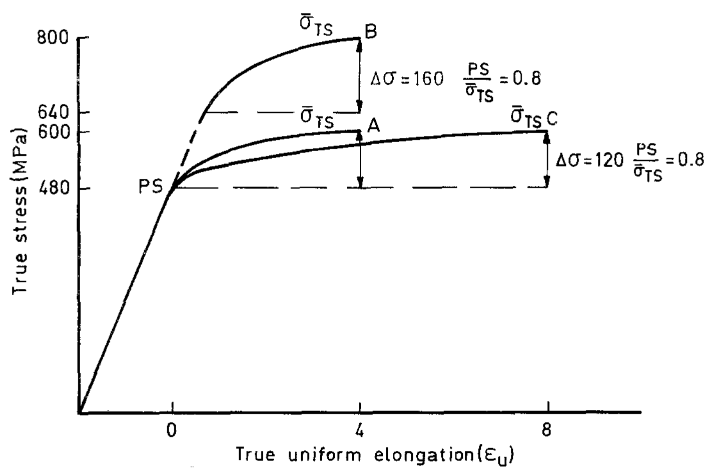

Figure A1 Schematic diagram of stress–strain curves with PS/TS ratios 0.8 but different maximum load and ε_u values.

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