Anisotropy of mechanical properties in quaternary Al–Li–Cu–Mg alloys

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MS received 29 January 1992; revised 27 March 1992

Abstract. Considerable anisotropy in the mechanical properties of quaternary Al–Li–Cu–Mg alloys was observed in both plate and sheet products. These alloys showed more than 100% increase in tensile ductility in the test direction oriented at 45–60° to the rolling direction as compared to that in the rolling direction (longitudinal, L). A concomitant decrease in strength was also found. These alloys exhibit superior low cycle fatigue resistance in the long-transverse (LT) direction as compared to the longitudinal (L) direction. Another observation is the occurrence of strength differential (S–D), which is seen to be directional. The trends in S–D are similar under monotonic as well as cyclic loading conditions. The fracture resistance is also highly anisotropic. An attempt is made here to correlate the observed anisotropy in the mechanical behaviour of these alloys with the microstructure and crystallographic texture.

Keywords. Quaternary Al–Li–Cu–Mg alloys; microstructure; texture; anisotropy; tensile; low cycle fatigue; strength differential; fracture resistance.

1. Introduction

Aluminium-lithium alloys are candidate materials for aircraft structural applications. These alloys possess superior specific modulus and strength as well as fatigue crack growth resistance as compared to the conventional 2XXX and 7XXX aircraft structural aluminium alloys (Baker *et al* 1986). On the other hand, significant improvement in ductility and fracture resistance needs to be achieved if these alloys were to replace the conventional aircraft structural aluminium alloys. Numerous attempts have been made in recent years to modify the alloy chemistry, processing schedules and thermomechanical treatments to achieve improved strength-toughness combination (Gregson and Flower 1985; Jata and Starke 1986). However, studies on the anisotropy in mechanical properties of these alloys are mostly confined to the tensile properties (Doorbar *et al* 1986; Peel *et al* 1986, 1987; Takahashi *et al* 1987). The present paper reports the results of a comprehensive study on the anisotropy in the mechanical properties of 8090 Al–Li–Cu–Mg alloys.

2. Experimental

The quaternary Al–Li 8090 alloys have the nominal composition, by wt%, 2.2–2.51 Li, 1–1.4 Cu, 0.7–1 Mg, 0.02–0.16 Zr and the balance aluminium. Lital C contains higher Cu and Mg and lower Zr as compared to Lital A. Various product forms, namely the Lital A plates (8 and 12.5 mm thickness) in T8E51 condition, Lital A sheets (2 and 3 mm thickness) in T6 condition, Lital C sheets (1.6 and 3 mm thickness) in T81 condition and Lital C sheet (1.6 mm thickness) in T3 condition were tested in the as-received condition.

Tensile tests were conducted on specimens oriented differently with respect to the rolling direction. In the case of 12.5 mm thick plate, specimens with 25 mm gauge length and 6.5 mm gauge diameter were used. On the other hand, flat tensile specimens with 25 mm gauge length and 8 mm width were employed to determine the tensile properties of the sheet products.

Fully-reversed, total-strain controlled low cycle fatigue (LCF) tests were conducted on MTS 880 servohydraulic test system using specimens of 15 mm gauge length and 6.5 mm gauge diameter. The strain was monitored using a 10 mm axial extensometer.

Plane-strain fracture toughness (K_{Ic}) tests were conducted on 12.5 mm thick plate in L–T (the notch being perpendicular to the rolling direction) and T–L (the notch being parallel to the rolling direction) orientations. Standard 12.5 mm thick compact tension specimens were used. For sheet products, the plane-stress fracture toughness (K_c) was evaluated in both L–T and T–L orientations using specimens of 101.6 mm width and half-height to width ratio of 0.6. Fracture toughness tests were conducted following ASTM standards E-399 (K_{Ic}) and E-516 (K_c) on computer-controlled servohydraulic MTS 880 and Instron 8500 test systems.

Crystallographic texture of both plate and sheet products was evaluated in the as-received condition. Three incomplete $\{111\}$, $\{100\}$ and $\{220\}$ pole figures up to a maximum tilting angle of 80–85° were obtained by X-ray diffraction technique using Siemens AG texture goniometer fitted with a D-500 diffractometer with Fe radiation.

3. Results and discussion

3.1 Microstructure and texture

The alloys in the as-received condition possess elongated, pancake structure with partial recrystallization. The optical micrograph of 12.5 mm thick plate, shown in figure 1a, reveals a wide variation in grain width and aspect ratio in three orthogonal directions. Such an anisotropy in grain structure was also seen to be present in the other plate and sheet products under consideration. As the grain width decreases, the grain aspect ratio increases. Among the products under consideration, Lital C sheets were characterized by predominantly recrystallized grain structure (figure 1b). At higher resolutions, the microstructure consists of spherical, dense and uniformly distributed Al_3Li (δ' , the strengthening phase) and lenticular Al_2CuMg (S, the phase leading to slip homogenization) precipitates. The S phase is nearly absent in Lital A sheet products of 2 and 3 mm thickness in T6 condition and Lital C sheet of 3 mm in T81 condition. A strong cross-rolled texture was observed in the 12.5 mm thickness plate, while the 8 mm thick plate and other sheet products exhibited predominant straight-rolled texture. The microstructure and texture characteristics of the alloys are summarized in table 1.

3.2 Orientation dependence of tensile properties

The tensile properties of 8090 Al–Li alloys evaluated with specimens oriented differently with respect to the rolling direction are given in table 2. Both the plate and sheet products exhibit significant anisotropy. As the orientation changes from 0

Figure 1. (a) Triplanar optical micrograph of Alcan 8090 plate of 12.5 mm thickness in T8E51 condition; (b) optical micrograph of Lital C-T81 in L direction showing recrystallized grain structure.

to 60°, the strength decreases and the ductility increases. In the case of 12.5 mm thick plate, the maximum decrease in yield strength is 21% (occurring at 60°), while the UTS decreased by 14% (occurring at 45°). The corresponding increase in ductility is around 110%. Both work-hardening coefficient (n) and uniform elongation (ϵ_u) attain maximum values at 45–60° to the rolling direction. A similar trend in the orientation dependence of tensile properties, though to a varying degree, has been observed with other 8090 Al–Li alloy products (table 2).

The Al–Li alloys under consideration, containing Li, Cu and Mg as the major alloying additions, exhibit varying degree of serrated plastic flow depending upon the aging treatment and thereby the concentration of Li, Cu and Mg in the solid solution and the

Table 1. Microstructure and texture characteristics of Al-Li-Cu-Mg alloy products.

Description	Plate products			Sheet products		
	Lital A-T8E51		Lital A-T6	Lital A-T6		Lital C-T81 (3 mm thick)
	(12.5 mm thick)	(8 mm thick)		3 mm thick	2 mm thick	
1. Degree of recrystallization	PR	PR		PR	PR	R
2. Grain width (μm)						
(a) Longitudinal	5-9 [†]	8-15 [†]		4-6	3-4	40-60
(b) Long-transverse	1-2	1-2		0.5-1	0.4-0.8	30-40
(c) Short-transverse	2-3	2-4		1-2	1-1.5	25-30
3. Grain aspect ratio						
(a) Long-transverse	40-45	30-40		40-60	50-70	3-6
(b) Short-transverse	20-25	20-25		20-40	30-40	2-4
4. Intragranular precipitates						
(a) $\delta'(\text{Al}_3\text{Li})$	High density, 30 nm dia.	High density, 25 nm dia.		High density, 10-20 nm dia.	—	High density, 5-10 nm dia
(b) $\text{S}(\text{Al}_2\text{CuMg})$	High density	High density		Sparse	—	Sparse
5. Width of precipitate free zone(μm)	0.1-0.12	0.07-0.1		≈ 0.01	—	0.005
6. Texture	Strong, cross- rolled	Strong, straight-rolled		Strong, straight- rolled	—	—

PR—Partially recrystallized (10-30% recrystallization);

R—Recrystallized (80-90% recrystallization)

[†]Wide variation in grain width—average value given

Table 2. Orientation dependence of tensile properties of Al-Li 8090 alloys.

Alloy designation	Property	Specimen orientation with respect to the rolling direction, (deg.)				
		0(L)	30	45	60	90(LT)
PLATES						
Lital A-T8E51 (12.5 mm thick)	0.2% Y.S., MPa	485	447	393	382	467
	UTS., MPa	555	509	478	513	534
	Elongation (25 mm GL)	0.054	0.061	0.115	0.096	0.07
	Uniform elongation	0.054	0.058	0.075	0.07	0.06
	n	0.053	0.063	0.077	0.11	0.05
Lital A-T8E51 (8 mm thick)	0.2% Y.S., MPa	471	—	370	—	457
	UTS., MPa	561	—	472	—	547
	Elongation (10 mm GL)	0.05	—	0.057	—	0.075
	Uniform elongation	0.05	—	0.057	—	0.07
	n	0.065	—	0.09	—	0.068
SHEETS						
Lital A-T6 (3 mm thick)	0.2% Y.S., MPa	456	424	352	350	432
	UTS., MPa	528	487	476	464	538
	Elongation (25 mm GL)	0.023*	0.044	0.14	0.141	0.062
	Uniform elongation	0.023	0.044	0.102	0.096	0.062
	n	0.068	0.058	0.111	0.101	0.086
Lital C-T81 (1.6 mm thick)	0.2% Y.S., MPa	355	326	324	315	310
	UTS., MPa	452	461	448	450	449
	Elongation (25 mm GL)	0.062	0.104	0.116	0.10	0.10
	Uniform elongation	0.06	0.08	0.097	0.088	0.092
	n	0.108	0.114	0.107	0.123	0.131
Lital C-T3 (1.6 mm thick)	0.2% Y.S., MPa	211	204	206	198	203
	UTS., MPa	328	328	330	315	331
	Elongation (25 mm GL)	0.17	0.174	0.178	0.19	0.162
	Uniform elongation	0.145	0.14	0.16	0.15	0.14
	n	0.16	0.161	0.155	0.162	0.163

*Average of 3 tests. All the specimens failed in curvature

test specimen orientation. Serrated flow is more pronounced in Lital C in T3 condition as compared to Lital A in T6 condition and for a given material in specimens oriented at 45–60° to the rolling direction (figure 2). The serrated plastic flow can be understood in terms of the interaction of solute elements with the dislocations resulting in Portevin-Le-Chatelier effect (Behnood and Evans 1989) and to progressive shearing of δ' precipitates (Gregson *et al* 1988).

The directionality in tensile properties of Al-Li alloys has been attributed to (i) crystallographic texture (Doorbar *et al* 1986; Gregson and Flower 1985; Peel *et al* 1986, 1987; Smith 1987; Welpmann *et al* 1984), (ii) mechanical fibring which includes grain anisotropy, degree of recrystallization, banding of small grains and directionality of coarse as well as fine inclusions (Wilson 1975; Forsyth and Stubbington 1975; Doorbar *et al* 1986; Peel *et al* 1986, 1987) and (iii) nature and distribution of strengthening precipitates (Dinsdale *et al* 1988; Lee and Kim 1989). In order to bring out the effect of various metallurgical factors on the tensile anisotropy in 8090 Al-Li alloys, the anisotropy data in table 2 are regrouped and presented in table 3. An

Figure 2. Load-elongation curves showing varied degrees of serrated plastic deformation in (a) Lital C-T3 and (b) Lital A-T6 sheets.

analysis of the data in table 3 reveals that the crystallographic texture and the degree of recrystallization have a pronounced effect on the anisotropy of these alloys.

The $\{111\}$ and $\{220\}$ pole figures obtained indicate that the commonly observed $\{110\} \langle 112 \rangle$ hot-rolling texture is predominant in the cross-rolled 12.5 mm thick plate. On the other hand, the other Al-Li alloy products now studied with straight-rolled texture possess an additional cold-rolled component of $\{112\} \langle 111 \rangle$, which was seen to be more pronounced at thinner gauges. The textural observations are consistent

Table 3. Metallurgical factors and the degree of anisotropy in tensile properties of Lital A and Lital C alloy products.

Description	Lital A–T8E51 (12.5 mm thick)	Lital A–T8E51 (8 mm thick)	Lital A–T6 (3 mm thick)	Lital C–T81 (1.6 mm thick)	Lital C–T3 (1.6 mm thick)
Degree of aging	Peak-aged	Peak-aged	Peak-aged	Near peak-aged	Under-aged
Degree of recrystallization	PR	PR	PR	R	R
Crystallographic texture	Strong, cross-rolled	Strong, straight-rolled	Strong, straight-rolled	—	—
Maximum drop in 0.2% YS (%)	21	21	23	13	6
Maximum drop in UTS (%)	14	16	12	1	–1
Maximum increase in elongation (%)	113	30	—	87	12

with those reported for Al–Li 8090 alloy products (Peters *et al* 1985, 1986a,b; Fox *et al* 1986; Palmer *et al* 1986). The calculation of Schmid factors as a function of orientation with respect to the rolling direction predicts that in these Al–Li alloys the minimum in tensile strength and maximum in tensile ductility occur at orientations 50–60° to the rolling direction (Peters *et al* 1985; Fox *et al* 1986). In a recent study conducted on 8090 and 8091 Al–Li alloys, Tempus *et al* (1991) predicted similar variation in tensile anisotropy through calculations based on the Taylor and Sachs models. The observed anisotropy in the present study can be understood on similar lines.

The degree of aging is seen to have a moderate effect on the anisotropy in strength, though the same influences the ductility significantly. In the case of Lital C sheets, the alloy in T81 condition shows a higher degree of anisotropy in yield strength as compared to that in T3 condition (table 3). Engler and Lucke (1991) observed that the precipitation stage that emerges with artificial aging changes the crystallographic texture of 8090 Al–Li alloy. The observed effect of degree of aging on the anisotropy may once again be attributed to the associated crystallographic texture.

The Lital C sheets in either of the heat-treated conditions show relatively lower degree of anisotropy in tensile properties when compared to the other Al–Li alloy products considered here. This could be attributed primarily to the lower degree of mechanical fibring. The Lital C sheets possess near recrystallized grain structure coupled with low inclusion constant, no appreciable precipitate free zones (PFZs) and the absence of PFZ-associated coarse equilibrium precipitates (see table 1). However, the strength levels of recrystallized alloys, in general, are lower as compared to the partially or fully unrecrystallized alloys. Recrystallization, therefore, imparts isotropy in tensile properties at the cost of strength.

3.3 Low cycle fatigue

The variation of fatigue life with plastic strain amplitude at half-life, generally referred to as Coffin–Manson (C–M) plot, is shown in figure 3 for the 12.5 mm thick plate in

Figure 3. Coffin-Manson (C-M) plots for longitudinal and long-transverse test directions.

L and LT directions. The alloy exhibits superior fatigue resistance in the LT direction as compared to the L direction. The fatigue life in LT direction at any particular strain amplitude is 40–100% higher than that in the L direction. The variation in resistance against strain-controlled fatigue bears parallelism with tensile ductility, as expected. The alloy in LT direction possesses higher ductility (7%) as compared to that in L direction (5.4%).

Studies on the low cycle fatigue of 8090 Al–Li alloys are limited to microstructural effects and test environment (Coyne *et al* 1981; Srivatsan and Coyne 1986a,b; Srivatsan 1988). In a different study conducted on extruded bars of 8090 Al–Li alloy, Khireddine *et al* (1989) reported considerable anisotropy in LCF resistance. The alloy exhibited superior LCF resistance at core section in longitudinal direction as compared to that at the surface sections in L direction and to that at the core section of LT direction. The observed anisotropy has been attributed to the variation in grain size as well as crystallographic texture.

The 12.5 mm thick Lital-A alloy in both the test orientations exhibited a transition in C-M plots at a strain $\Delta\epsilon_p/2 \approx 2 \times 10^{-3}$. A detailed analysis revealed that the transition is associated with a gradual change in deformation and fracture modes (Eswara Prasad *et al* 1989). With increase in strain amplitude, the fracture mode changes from higher energy transgranular shear to lower energy ductile intergranular fracture. Highly localized plastic deformation due to shearing of Al_3Li (δ') precipitates at higher strain amplitudes results in intergranular fracture. These features were found to be common for specimens tested in both L and LT test directions.

3.4 Strength differential under monotonic and cyclic conditions

Cyclic stress response behaviour of 8090 alloy plate of 12.5 mm thickness in T8E51 condition analysed in terms of the variation of only tensile (σ_T) or only compressive (σ_c) stress amplitude with number of cycles revealed the occurrence of strength differential (S–D) (Eswara Prasad *et al* 1992). The variation of S–D with plastic strain for the three test directions (L, L + 45° and LT) is shown in figure 4. For cyclic loading

Figure 4. Variation of strength differential with plastic strain under monotonic and cyclic loading conditions in L, L + 45° and LT test directions.

conditions, the strength differential ($|\sigma_c| - \sigma_T$) at half-life is plotted against the corresponding plastic strain amplitude ($\Delta\epsilon_p/2$). On the other hand, the monotonic strength differential data are derived from pure tension and compression tests using specimens of low cycle fatigue specimen geometry. The data revealed an orientation dependent S-D effect. The stress amplitude in compression (σ_c) is higher than that in tension (σ_T), depicting a normal S-D effect at all strain levels in the L + 45° and long-transverse directions. A similar behaviour was observed at strain amplitudes $\Delta\epsilon_p/2 > 4.5 \times 10^{-3}$ in the longitudinal direction, for which, at lower strains, the S-D is more than compensated by the Bauschinger effect. The 2–3% tensile prestretch given to the material as part of thermomechanical treatment results in Bauschinger effect, which obscures S-D effect at lower strains in L direction. This study explains the behaviour generally reported but not clearly explained in literature (Lee *et al* 1990) regarding lower compressive yield strength in L direction of Al-Li alloys with a prior history of stretch during processing.

The trends in S-D are found to be similar under cyclic and monotonic loading conditions. However, a comparison between the cyclic and monotonic S-D data

presented in figure 4, brings out two aspects. First, the cross-over of the cyclic S-D from a negative to positive value in the case of L direction occurs at a lower strain level ($\Delta\epsilon_p/2 = 4.5 \times 10^{-3}$) as compared to monotonic loading ($\epsilon_p = 6 \times 10^{-3}$). Secondly, the magnitude of cyclic S-D at any particular strain level is nearly 20–40% lower than that under monotonic conditions. Among the three test directions considered, the S-D is of largest magnitude in the L + 45° direction.

Several explanations for the S-D effect based on residual stresses, microcracks, interactions between dislocations and interstitial solute atoms and a stress concentration at the particle-matrix interface have been proposed in the literature (Leslie and Sober 1967; Hirth and Cohen 1969, 1970; Kalish and Cohen 1969; Olsen and Ansell 1969; Chait 1972). Of these, two mechanisms, namely (i) based on a stress intensification at the matrix-particle interface suggested by Olsen and Ansell (1969) and (ii) based on dislocation-solute atom interactions proposed by Hirth and Cohen (1969, 1970) appear relevant here. To investigate the applicability of these two explanations, the S-D data were obtained for 12.5 mm thick plate in four different heat-treated conditions, viz. the solution-treated, the peak-aged (T6), the overaged (T7) and the as-received stretch plus peak-aged (T8E51) conditions. Detailed analysis of the data revealed (Eswara Prasad *et al* 1992) that the S-D observed in the present alloy cannot be explained in terms of either the solute-dislocation interactions or the stress intensification at the particle-matrix interface.

3.5 Anisotropy in fracture resistance

The plane-strain fracture toughness (K_{Ic}) values obtained for Al-Li alloys with different notch orientations (see figure 5) are listed in table 4. The 12.5 mm thick plate possesses nearly 75% higher K_{Ic} in the L-T orientation as compared to that in the T-L orientation. The fractured surfaces also exhibit considerable differences in the two test orientations in terms of the degree of delamination. The height of delaminated ridges is found to be significantly higher (1.5–3 mm) in the L-T orientation as compared to that in the T-L orientation (0.2–0.5 mm). Considerably higher degree of anisotropy in fracture resistance of Al-Li alloys is again evident when the K_{Ic} data obtained in S-L or S-T orientations are compared with that in L-T orientation (Venkateswara Rao and Ritchie 1988). The fracture toughness (K_{Ic}) values in S-T

Figure 5. Crack plane orientation code for plate and sheet products.

Table 4. Fracture toughness data of Al-Li-Cu-Mg alloys.

Alloy designation	Crack plane orientation	0.2% YS (MPa)	K_{Ic} or K_c (MPa $\sqrt{m}$)	Thickness required for valid K_{Ic} (mm)	Nature of load-COD curve
Alcan 8090 plate (12.5 mm thick)	L-T	485	28	8.4	Gradual load drop
	T-L	465	16	3.2	Sudden load drop
	S-L*	—	13	1.8	Sudden load drop
Alcan 2090* -T8E41 plate	L-T	552	36	10.6	—
	T-L	548	24	4.8	—
	S-L	372	17	5.2	—
	S-T	372	16	4.6	—
Alcan 8090-Lital A sheet (3 mm thick)	L-T	456	51 [†]	—	Rising R-curve
	T-L	426	50 [†]	—	Flat R-curve

*Data from Venkateswara Rao and Ritchie (1988)

[†]Plane-stress fracture toughness (K_c)

and S-L orientations (see figure 5) are more than 50% as compared to the K_{Ic} value in L-T orientation (table 4). Such an anisotropy in fracture resistance of Al-Li alloys has been attributed to microstructural features. The anisotropic, nearly unrecrystallized grain structure and marked deformation texture lead to easy delamination in the thickness direction (Venkateswara Rao and Ritchie 1988; Roven *et al* 1990). In a recent study on 8090 Al-Li alloy plate Lynch (1991) has shown that double aging treatment involving reaging, after initial artificial aging, at high temperatures of 200–210°C (rapid heating to the reaging temperature) for shorter durations of up to 300s, significantly increases toughness in the thickness (S-L) direction with only a marginal loss in strength and thereby reduces the anisotropy in fracture resistance. The beneficial effect of double aging treatment has been attributed to the decrease in the extent of lithium segregation at the grain boundaries.

The plane-stress fracture toughness (K_c) values obtained in the L-T and T-L orientations for Lital A sheet of 3 mm thickness are listed in table 4. The alloy in both the test orientations possesses nearly the same K_c value. However, the nature of the $K_R - \Delta a$ plots (known as *R*-curves) is different for the two test directions. The alloy exhibits progressively rising *R*-curve with deflected crack path in the L-T orientation as against the flat *R*-curve behaviour in the T-L orientation with straight crack path (figure 6). The increased resistance to crack extension in L-T orientation as compared to the T-L orientation can be explained using a mixed-mode analysis (Kamat and Eswara Prasad 1990). The deflected cracks in L-T orientation are assumed to be initially inclined cracks with the inclination angle being the same as deflection angle, θ . The analysis shows that

$$(K_R)_{\text{Deflected crack}} = (K_R)_{\text{Undeflected crack}} \times \sec^{\frac{1}{2}} \theta \quad (1)$$

or

$$(K_R)_{\text{L-T orientation}} = (K_R)_{\text{T-L orientation}} \times \sec^{\frac{1}{2}} \theta \quad (2)$$

As the angle of deflection (θ) increases, the fracture resistance in L-T orientation increases. The theoretically predicted *R*-curve in L-T orientation (figure 6a), using the $K_R - \Delta a$ data obtained for undeflected crack in T-L orientation and the deflection

Figure 6. (a) $K_R - \Delta a$ plots of Lital A sheet (2 mm thick) and the corresponding crack path morphologies in (b) T-L and (c) L-T test directions.

angle measured for L-T orientation, is seen to match well with the experimentally obtained R-curve (Kamat and Eswara Prasad 1990). The progressively increasing fracture resistance in the L-T orientation can therefore be attributed to the crack

path deflection. Metallographic observations reveal that the cracks follow the shear bands in L–T orientation and get deflected constantly. The shear bands are formed along {111} planes which are grossly oriented at 45–55° to the rolling direction. On the other hand, the cracks in T–L orientation follow the high angle grain boundaries and result in relatively straight crack paths (Eswara Prasad *et al* 1991).

4. Conclusions

Marked anisotropy in mechanical properties, namely the tensile, fatigue and fracture, has been observed in quaternary Al–Li–Cu–Mg alloys in both plate and sheet product forms. The extent of anisotropy was seen to be as high as 110% in tensile ductility and more than 50% in fracture resistance measured in terms of K_{Ic} . Such an anisotropy in mechanical properties has primarily been attributed to the nearly unrecrystallized pancake microstructure and strong crystallographic texture.

Acknowledgements

The authors are grateful to Dr P Rama Rao, for constant encouragement and keen interest in this work. Thanks are due to Prof. V Kutumba Rao for valuable comments on the manuscript and Dr K N Raju and Dr R Sunder for providing material for conducting these studies. The help received from Mr K S Prasad in conducting the microstructural investigations and Dr R K Basu in texture analysis is gratefully acknowledged.

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