

NEW MATERIALS CANDIDATES FOR NEW-GENERATION NUCLEAR POWER PLANT APPLICATIONS

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Abstract

The global population, currently around 7 billion, is projected to reach 10 billion by 2050. Meeting the growing energy demand is essential for improving quality of life and socio-economic development requires accessible and sustainable energy sources. Traditional fossil fuels such as oil, coal, and natural gas pose serious environmental challenges and are not viable long-term solutions. A transition to cleaner and more efficient energy sources is critical to mitigate the impacts of global climate change. Nuclear energy, which emits no greenhouse gases, offers a promising alternative while other renewable technologies continue to evolve. In this context, MAX phases—a class of thermodynamically stable nano-laminated ceramics—are being explored for their potential use in nuclear reactors, particularly in fuel cladding applications due to their resilience in extreme environments.

Keywords: MAX phase, X-Ray Diffraction Residual stress, Irradiation

1. Introduction

With the global population projected to reach 10 billion by 2050, the demand for sustainable and affordable energy to improve quality of life and support economic development is increasing. Nuclear energy, due to its lack of greenhouse gas emissions, presents a viable alternative to fossil fuels, and MAX phase materials are emerging as a promising option for use in nuclear reactors.

The MAX phases, first identified by Nowotny in the 1960s as Hägg phases with M_2AX structure (Nowotny, 1971), remained largely unexplored until 1996, when Barsoum et al. synthesized bulk Ti_3SiC_2 , revealing their unique properties (Barsoum & Hounag, 1993). With over half of the 342 known MAX phases discovered after 2018, and 28 M, 28 A, and 6 X elements incorporated so far, research in this field has rapidly advanced (Dahlqvist et al., 2024).

MAX phases ($M_{n+1}AX_n$) are layered, hexagonal ternary carbides and/or nitrides where M is an early transition metal, A is a group 13–16 element, X is C and/or N, and $n = 1, 2$, or 3. Their structure consists of strong $M-X$ bonds forming M_6X octahedra interleaved with weaker $M-A$ layers. The value of n

determines the number of M layers between A layers, giving rise to their unique mix of metallic and ceramic properties (Gonzalez-Julian, 2021; Griseri et al., 2019; Patel et al., 2014; Tallman, 2015).

MAX phases exhibit a unique combination of metallic and ceramic properties—they are machinable, thermally and electrically conductive, lightweight, and resistant to oxidation, thermal shock, and creep. With moderate hardness (2–8 GPa), high fracture toughness, and excellent high-temperature performance, they are ideal for demanding applications such as nuclear reactors (Barsoum, 2013; Barsoum & Radovic, 2011; Patel et al., 2014; Radovic & Barsoum, 2013; Tallman et al., 2015). Their deformation occurs via ripplation and kink-band formation under compression (Barsoum et al., 2019; Gruber et al., 2020; Sudhakar et al., 2023).

Ceramic materials are usually not manufactured by plastic deformation, mechanical machining or casting but their production involves powders shaping and successive consolidation at high temperatures (Gruber et al., 2020). The MAX phases were typically fabricated using methods such as hot press sintering (HPS) (Tallman et al., 2016), spark plasma sintering (SPS) (Haemers et al., 2020), mechanical alloying (MA) (Tallman et al., 2016), self-propagating high-temperature synthesis (SHS) (Barsoum, 2013), physical vapor deposition (PVD) (Hoffman et al., 2012) and chemical vapor deposition (Clark et al., 2016).

- **Ti₃AlC₂:** The Ti₃AlC₂ samples were fabricated by ball milling stoichiometric mixtures of pre reacted Ti₂AlC and TiC powders for 24 h. The mixture was press at 1200 °C for 4 h under a load corresponding to a stress of ~30 MPa and a vacuum of 10⁻¹ Pa (Zhao et al., 2021).
- **Ti₃SiC₂:** Samples of Ti₃SiC₂ were prepared by ball milling stoichiometric mixtures of Ti (99.9%, –325 mesh) Si (99.5%, –325 mesh), and C powders (99.9%) for 24 h, which were then press at 1450 °C for 6 h under a load corresponding to a stress of ~30 MPa and a vacuum of 10⁻¹ Pa (Hull, 1975).

Irradiation, particularly by high-energy particles like neutrons, can displace atoms from their lattice sites, initiating knock-on events that cause structural damage in materials (Anderson, 1964; Barsoum & El-Raghy, 2001). The extent of irradiation damage in MAX phases depends mainly on temperature and fluence, with higher doses increasing defect density (Barsoum & El-Raghy, 2001). At low temperatures, defects form black spot clusters, while at higher temperatures, increased mobility of interstitials and vacancies promotes recombination and reduces residual damage (Was et al., 2002). X-ray diffraction (XRD) is a technique used to analyze material structure, phase composition, and microstrains by probing large sample volumes (Cullity & Smoluchowski, 1957). X-rays, first discovered by Röntgen in 1895, are electromagnetic waves with wavelengths between 0.5–2.5 Å, and characteristic X-rays—first identified by Barkla in 1909—are emitted when inner-shell vacancies are filled by outer electrons, producing element-specific radiation (Cullity & Smoluchowski, 1957; Klug & Alexander, 1974).

Many recent studies have explored the damage tolerance and amorphization resistance of several MAX phases irradiated with heavy ions (Bugnet et al., 2010; Nappé et al., 2009; Prabhu et al., 2020; Whittle et al., 2010). In 2009, Nappé et al. explored the effect of 90 MeV Xe ions on Ti₃SiC₂, showing surface layer amorphization of the Ti₃SiC₂ and impurity phases, as well as selective grain boundary sputtering due to nuclear interactions (Ward et al., 2018). Liu et al. characterized Al-doped Ti₃SiC₂, irradiated to a maximum dose of 3.25 dpa with high energy Kr and Xe ions, via X-ray diffraction, XRD, and nanoindentation showing an expansion of the c-lattice parameter, c-LP, and irradiation induced hardness increase that could be annealed out at 800 °C (Liu et al., 2010). They also reported the possible existence of a β-Ti₃SiC₂ phase to explain the presence of new peaks found in the post irradiated XRD patterns (Liu et al., 2010).

Soon after, Le Flem et al. reported on selected area electron diffraction, SAED, patterns using transmission electron microscopy, TEM, of $\text{Ti}_3(\text{Si,Al})\text{C}_2$ irradiated with 92 MeV Xe ions up to 6.67 dpa, which further the main conclusion is that Ti_3SiC_2 does not appear to be very sensitive to amorphization, probably because of the high metallic character of the bonding (Le Flem et al., 2010). With increasing dose, diffraction peaks were found to disappear due to lattice disturbances (Le Flem et al., 2010). In 2010, Bugnet et al. explored the effect of 100 keV Ar^{2+} ion irradiation on multilayered $(\text{Ti,Al})_N/\text{Ti}_2\text{Al}_{N_x}$ nitrogen-deficient thin films it was observed that the multilayered structure preserved up to a dose of 12 dpa, probably because nitrogen vacancies served as traps for interstitial point defects (Wang et al., 2013). Similarly, Whittle et al. reported that Ti_3AlC_2 and Ti_3SiC_2 possess very high resistance to amorphization, even up to 25 dpa, irradiated with 1 MeV Kr^{2+} and 1 MeV Xe^{2+} (Wang et al., 2013). Nappé et al. reported that nuclear collisions with 4 MeV Au ions, up to a dose of 4.3 dpa, resulted in a 2.2% swelling in Ti_3SiC_2 (Nappé, Maurice, et al., 2011). Also in 2011, Nappé et al. investigated the structural changes of Ti_3SiC_2 under a variety of ions and energies, concluding that Ti_3SiC_2 is not sensitive to electrical interactions, and confirmed that nuclear collisions lead to an increase in c-LP and a decrease in the a lattice parameter, a-LP, and a concomitant increase in lattice microstrains (Nappé, Monnet, et al., 2011).

Diffraction Origins and Theory

In 1912, von Laue discovered that X-rays diffract from periodic atomic structures in crystals, and in 1913, the Braggs formulated Bragg's Law, which relates the X-ray wavelength (λ), interplanar spacing (d_{hkl}), and incident angle (θ) as (Cullity & Smoluchowski, 1957; Klug & Alexander, 1974):

$$n\lambda = 2d_{hkl} \sin\theta \quad (1)$$

This principle enables X-ray diffraction (XRD), a powerful technique for determining atomic structures in crystalline materials. XRD uses monochromatic X-rays (e.g., Cu $K\alpha$, $\lambda = 1.54 \text{ \AA}$) that scatter elastically from electron clouds, producing interference patterns. Constructive interference occurs at specific angles, producing diffraction peaks used to calculate d-spacing and analyze crystal structure (Bol, 1967; Prabhu et al., 2020).

2. Experimental

XRD patterns of irradiated SiC samples were obtained using a Bruker D8 Discover diffractometer in Bragg-Brentano configuration, for pristine and irradiated conditions. Scans were made with Cu $K\alpha$ radiation ($\lambda = 1.54 \text{ nm}$, 45 Kv and 40 mA). Figure 1 shows the XRD pattern for both types of SiC samples, Indicating the planes with the highest diffraction intensities.

Figure 2 and Figure 3 show that the XRD patterns for samples within the $35.2\text{--}36^\circ$ and $119.6\text{--}120.4^\circ$ ranges. It was observed that the XRD graphs of samples (16BB01) shifted a little bit to the right side. However, the lattice parameters were determined and yielded to be identical for the two specimens, being: $3.08 \text{ \AA}$ for “a” of both specimens, and $15.119 \text{ \AA}$ and $15.120 \text{ \AA}$ for “c” of specimen 16BB01 and 40BB01, respectively. Thus, the observed peak shift is only an artificial effect and not a change in the crystal structure. The depth of irradiation is too small to detect any changes, so it needs to be use Small Angular X-ray Scattering (SAXS) is to detect lattice distortions caused by irradiation.

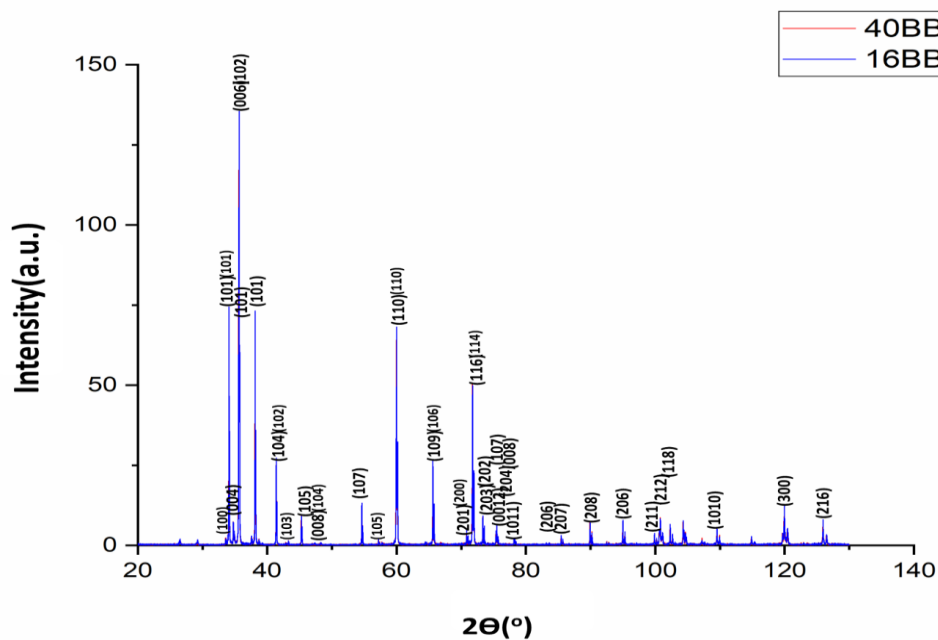


Figure 1. XRD pattern of SiC

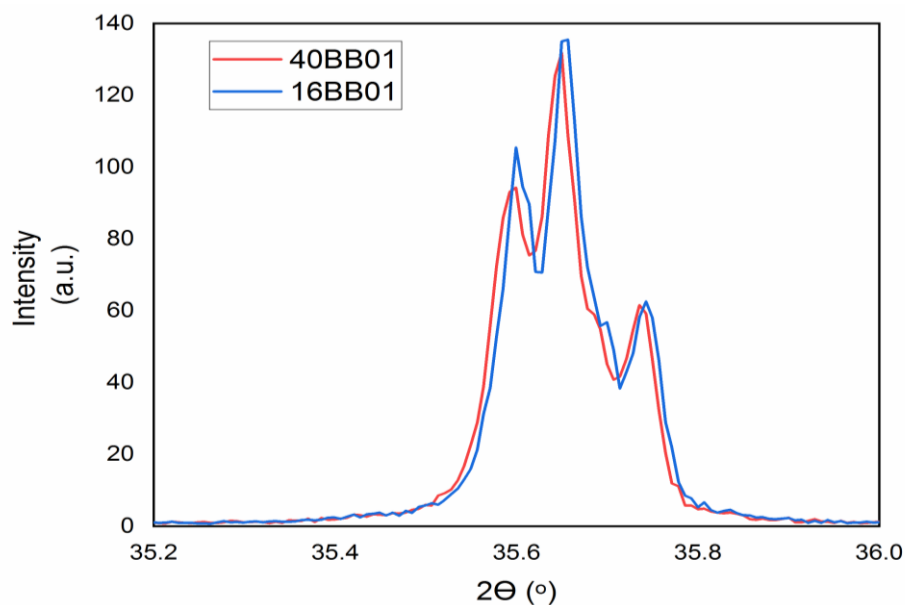


Figure 2. XRD pattern of SiC zoom in between 35.2°–36°

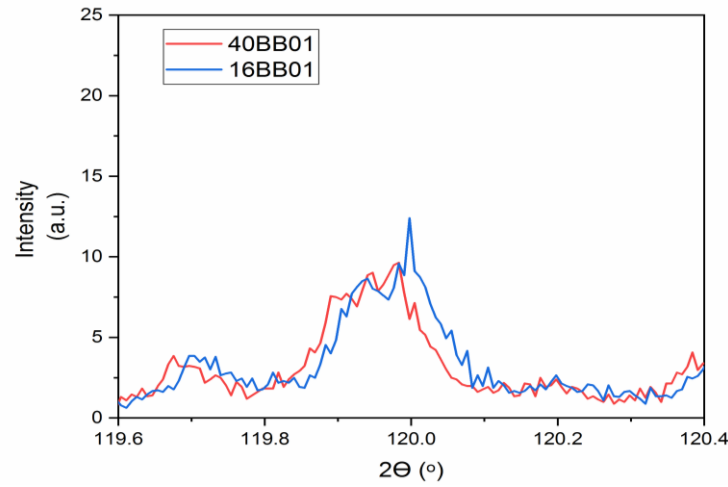


Figure 3. XRD pattern of SiC zoom in between 119.6°–120.4°

3. Preparations for residual stress measurements

The Table 1 shows six types of X-ray tubes available for residual stress measurements along with their corresponding wavelengths. According to the Excel tables provided, the distances between planes (i.e., the d values), θ angles, 2θ angles, and intensities have been calculated based on Bragg's law. For further analysis, all six tubes were used to calculate the 2θ angles.

Table 1. illustrates the types of tubes and wavelength ($\text{\AA}$)

Tube	Wavelength
Cr	2.291
Mn	2.1018
Ti	2.7497
Cu	1.542
Co	1.789
Fe	1.93736

To investigate residual stress, measurements at larger Bragg angles are required. Therefore, it is necessary to prepare an Excel chart capable of calculating the interplanar spacing (d-value), Bragg angle (θ), (2θ), and corresponding intensity.

3.1 SiC–Ti tube

First, reflections were calculated for Ti tube to determine Bragg angles and intensities. Table 2 and 3 show the data in first and second order respectively, when using this tube. It was observed that the SiC{109} diffraction peak occurs at a large Bragg angle ($\sim 150^\circ$) and exhibited high intensity in first order.

Table 2. the Bragg angles and intensities for SiC with Ti tube in first order

SiC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
2.7497	7.5587	15.1174	0.182908	10.47985	20.9597	1	0	0	2
2.7497	3.77935	7.5587	0.372322	21.33249	42.66498	1	0	0	4
2.7497	2.62732	5.25464	0.550707	31.55319	63.10638	412	1	0	1
2.7497	2.51581	5.03162	0.57816	33.12614	66.25227	999	0	0	6
2.7497	2.51581	5.03162	0.57816	33.12614	66.25227	999	1	0	2
2.7497	2.35785	4.7157	0.622533	35.66851	71.33702	436	1	0	3
2.7497	2.17957	4.35914	0.68257	39.1084	78.2168	174	1	0	4
2.7497	2.00046	4.00092	0.75772	43.41415	86.8283	65	1	0	5
2.7497	1.88967	3.77934	0.81476	46.6823	93.36461	1	0	0	8
2.7497	1.8318	3.6636	0.848888	48.63769	97.27538	1	1	0	6
2.7497	1.6786	3.3572	0.959746	54.98937	109.9787	80	1	0	7
2.7497	1.54032	3.08064	1.103022	63.19853	126.3971	441	1	1	0
2.7497	1.42145	2.8429	1.314031	75.28845	150.5769	168	1	0	9

Table 3. the Bragg angles and intensities for SiC with Ti tube in second order

Second order									
SiC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
2.7497	7.5587	15.1174	0.372322	21.33249	42.66498	1	0	0	2
2.7497	3.77935	7.5587	0.814757	46.68214	93.36428	1	0	0	4

3.2 SiC–Mn tube

Second, the calculation was performed for Mn tube. Table 4 and 5 show the peak data in first and second order respectively. It was clear that in first order, Bragg angle was large ($\sim 149^\circ$) for the SiC{208} reflection (to be more precise, the second order of the SiC{104}), whereas its intensity is moderate. However, the second order diffraction of the SiC{208} (precisely, the fourth order of the SiC{104}) comes at the same large Bragg angle with relatively large intensity.

Table 4. the Bragg angles and intensities for SiC with Mn tube in first order

SiC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
2.1018	7.5587	15.1174	0.139484	7.991827	15.98365	1	0	0	2
2.1018	3.77935	7.5587	0.281778	16.14467	32.28935	1	0	0	4
2.1018	2.62732	5.25464	0.411505	23.57751	47.15502	412	1	0	1
2.1018	2.51581	5.03162	0.430933	24.69062	49.38124	999	1	0	2
2.1018	2.35785	4.7157	0.461959	26.4683	52.93661	436	1	0	3
2.1018	2.17957	4.35914	0.503118	28.82653	57.65305	174	1	0	4
2.1018	2.00046	4.00092	0.553102	31.69041	63.38081	65	1	0	5
2.1018	1.88967	3.77934	0.589721	33.7885	67.577	1	0	0	8
2.1018	1.8318	3.6636	0.611014	35.0085	70.01701	1	1	0	6

SiC									
2.1018	1.54032	3.08064	0.750851	43.02057	86.04113	441	1	1	0
2.1018	1.42145	2.8429	0.832053	47.67314	95.34628	168	1	0	9
2.1018	1.3288	2.6576	0.912219	52.26632	104.5326	33	2	0	1
2.1018	1.31419	2.62838	0.926722	53.09727	106.1945	343	2	0	2
2.1018	1.28954	2.57908	0.952629	54.5816	109.1632	56	2	0	3
2.1018	1.25978	2.51956	0.986668	56.53193	113.0639	23	0	0	12
2.1018	1.22174	2.44348	1.035596	59.3353	118.6706	14	1	0	11
2.1018	1.19393	2.38786	1.076288	61.66678	123.3336	1	1	1	8
2.1018	1.17892	2.35784	1.100445	63.05085	126.1017	1	2	0	6
2.1018	1.13491	2.26982	1.183614	67.81607	135.6321	17	2	0	7
2.1018	1.08978	2.17956	1.302874	74.6492	149.2984	51	2	0	8

Table 5. the Bragg angles and intensities for SiC with Mn tube in second order

Second order									
SiC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
2.1018	7.5587	15.1174	0.281778	16.14467	32.28935	1	0	0	2
2.1018	3.77935	7.5587	0.589719	33.7884	67.5768	1	0	0	4
2.1018	2.62732	5.25464	0.92726	53.12807	106.2561	412	1	0	1
2.1018	2.51581	5.03162	0.988927	56.66134	113.3227	999	0	0	6
2.1018	2.51581	5.03162	0.988927	56.66134	113.3227	999	1	0	2
2.1018	2.35785	4.7157	1.100437	63.05037	126.1007	436	1	0	3
2.1018	2.17957	4.35914	1.302858	74.64824	149.2965	174	1	0	4

3.3 SiC–Cr tube

Third, reflections for the Cr tube were calculated and Table 6 and 7 show the data in first and second order respectively. It was observed that in first order the Bragg angle of the SiC{205} and SiC{118} reflections were moderate ($\sim 139^\circ$) and ($\sim 147^\circ$) with small intensities.

Table 6. the Bragg angles and intensities for SiC with Cr tube in first order

SiC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
2.291	7.5587	15.1174	0.152133	8.716601	17.4332	1	0	0	2
2.291	3.77935	7.5587	0.307938	17.64356	35.28712	1	0	0	4
2.291	2.62732	5.25464	0.451144	25.84866	51.69733	412	1	0	1
2.291	2.51581	5.03162	0.472732	27.08556	54.17112	999	0	0	6
2.291	2.51581	5.03162	0.472732	27.08556	54.17112	999	1	0	2
2.291	2.35785	4.7157	0.507306	29.06647	58.13294	436	1	0	3
2.291	2.17957	4.35914	0.553376	31.70611	63.41222	174	1	0	4
2.291	2.00046	4.00092	0.609696	34.93301	69.86602	65	1	0	5
2.291	1.88967	3.77934	0.651262	37.31456	74.62912	1	0	0	8

SiC									
2.291	1.8318	3.6636	0.675569	38.70723	77.41447	1	1	0	6
2.291	1.6786	3.3572	0.75106	43.03256	86.06513	80	1	0	7
2.291	1.54032	3.08064	0.838553	48.04556	96.09112	441	1	1	0
2.291	1.42145	2.8429	0.937139	53.69409	107.3882	168	1	0	9
2.291	1.3288	2.6576	1.039312	59.54822	119.0964	33	2	0	1
2.291	1.31419	2.62838	1.058537	60.64973	121.2995	343	1	1	6
2.291	1.31419	2.62838	1.058537	60.64973	121.2995	343	2	0	2
2.291	1.28954	2.57908	1.093633	62.66055	125.3211	56	2	0	3
2.291	1.25978	2.51956	1.141565	65.40683	130.8137	23	0	0	12
2.291	1.2579	2.5158	1.144842	65.5946	131.1892	33	2	0	4
2.291	1.22174	2.44348	1.215655	69.65187	139.3037	14	1	0	11
2.291	1.22045	2.4409	1.218516	69.8158	139.6316	14	2	0	5
2.291	1.19393	2.38786	1.284997	73.62488	147.2498	1	1	1	8
2.291	1.17892	2.35784	1.332121	76.32492	152.6498	1	2	0	6

Table 7. the Bragg angles and intensities for SiC with Cr tube in second order

Second order									
SiC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
2.291	7.5587	15.1174	0.307938	17.64356	35.28712	1	0	0	2
2.291	3.77935	7.5587	0.65126	37.31444	74.62888	1	0	0	4
2.291	2.62732	5.25464	1.059255	60.69086	121.3817	412	1	0	1
2.291	2.51581	5.03162	1.144833	65.5941	131.1882	999	0	0	6
2.291	2.51581	5.03162	1.144833	65.5941	131.1882	999	1	0	2
2.291	2.35785	4.7157	1.332104	76.32392	152.6478	436	1	0	3

3.4 SiC–Cu tube

With using a copper tube, according to the data, the Bragg angles are too small for residual stress measurements in first order (Table 8).

Table 8. the Bragg angles and intensities for SiC with Cu tube in first order

SiC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
1.542	7.5587	15.1174	0.102179	5.854447	11.70889	1	0	0	2
1.542	3.77935	7.5587	0.205446	11.77116	23.54232	1	0	0	4
1.542	2.62732	5.25464	0.297839	17.06491	34.12982	412	1	0	1
1.542	2.51581	5.03162	0.311474	17.84614	35.69228	999	0	0	6
1.542	2.51581	5.03162	0.311474	17.84614	35.69228	999	1	0	2
1.542	2.35785	4.7157	0.33312	19.08635	38.17271	436	1	0	3
1.542	2.17957	4.35914	0.361566	20.71621	41.43242	174	1	0	4
1.542	2.00046	4.00092	0.395654	22.66928	45.33856	65	1	0	5

1.542	1.88967	3.77934	0.420271	24.07975	48.15949	1	0	0	8
1.542	1.8318	3.6636	0.434434	24.89126	49.78252	1	1	0	6
1.542	1.6786	3.3572	0.47722	27.34268	54.68536	80	1	0	7
1.542	1.54032	3.08064	0.524229	30.03609	60.07217	441	1	1	0
1.542	1.42145	2.8429	0.573296	32.84743	65.69486	168	1	0	9
1.542	1.3288	2.6576	0.619002	35.46621	70.93242	33	2	0	1
1.542	1.31419	2.62838	0.626945	35.92128	71.84256	343	1	1	6
1.542	1.31419	2.62838	0.626945	35.92128	71.84256	343	2	0	2
1.542	1.28954	2.57908	0.640863	36.71876	73.43752	56	2	0	3
1.542	1.25978	2.51956	0.658602	37.7351	75.4702	23	0	0	12
1.542	1.2579	2.5158	0.659759	37.8014	75.60279	33	2	0	4
1.542	1.22174	2.44348	0.682928	39.1289	78.2578	14	1	0	11
1.542	1.22045	2.4409	0.683788	39.17818	78.35637	14	2	0	5
1.542	1.19393	2.38786	0.702027	40.22317	80.44634	1	1	1	8
1.542	1.17892	2.35784	0.712845	40.84299	81.68598	1	2	0	6
1.542	1.13491	2.26982	0.746875	42.79279	85.58559	17	2	0	7
1.542	1.08978	2.17956	0.785929	45.03043	90.06086	51	2	0	8

3.5 Ti_3AlC_2 -Cu tube

Table 9 and 10 show the Bragg angles and intensities in the case of Cu tube. It can be seen that the $Ti_3AlC_2\{219\}$ reflection has Bragg angle at $\sim 136^\circ$, which is large enough, but has a very small intensity. Thus, the small intensity must be compensated with extended collection time.

Table 9. the Bragg angles and intensities for Ti_3AlC_2 with Cu tube in first order

TiAlC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
1.542	6.8	13.6	0.113627	6.51033	13.02066	39	0	0	2
1.542	2.633	5.266	0.297177	17.02697	34.05395	19	1	0	0
1.542	2.585	5.17	0.302868	17.35308	34.70615	6	1	0	1
1.542	2.277	4.554	0.345432	19.79181	39.58362	100	1	0	3
1.542	2.267	4.534	0.34702	19.88279	39.76557	18	0	0	6
1.542	2.082	4.164	0.37935	21.73517	43.47034	3	1	0	4
1.542	1.7177	3.4354	0.465485	26.67031	53.34062	14	1	0	6
1.542	1.52	3.04	0.531976	30.47995	60.9599	14	1	1	0
1.542	1.3164	2.6328	0.625729	35.85163	71.70325	2	2	0	0
1.542	1.3106	2.6212	0.62893	36.03506	72.07012	9	1	0	9
1.542	1.2624	2.5248	0.656997	37.64313	75.28627	11	1	1	6
1.542	1.1383	2.2766	0.744122	42.63503	85.27005	2	2	0	6
1.542	0.9926	1.9852	0.889486	50.96378	101.9276	2	2	0	9
1.542	0.9719	1.9438	0.916196	52.49418	104.9884	3	2	1	3
1.542	0.9086	1.8172	1.013254	58.0552	116.1104	2	1	1	12
1.542	0.8311	1.6622	1.188167	68.07696	136.1539	2	2	1	9

Table 10. the Bragg angles and intensities for Ti_3AlC_2 with Cu tube in second order

Second order									
TiAlC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
1.542	6.8	13.6	0.228755	13.10667	26.21334	39	0	0	2
1.542	2.633	5.266	0.625674	35.84848	71.69696	19	1	0	0
1.542	2.585	5.17	0.639156	36.62095	73.2419	6	1	0	1
1.542	2.277	4.554	0.74396	42.62576	85.25152	100	1	0	3
1.542	2.267	4.534	0.748027	42.85881	85.71762	18	0	0	6
1.542	2.082	4.164	0.834013	47.78545	95.5709	3	1	0	4
1.542	1.7177	3.4354	1.114549	63.85894	127.7179	14	1	0	6

3.6 Ti_3AlC_2 -Ti tube

For residual stress measurements in the Ti_3AlC_2 MAX phase, the use of Cr and Mn X-ray tubes is not feasible due to the occurrence of fluorescence excitation. In the periodic table, Cr and Mn are the second and third neighboring elements to Ti, respectively. As a result, their characteristic X-ray energies can excite Ti atoms, leading to increased background noise and preventing the clear detection of XRD peaks. Using the Ti tube, the Bragg angle is too small ($\sim 129^\circ$ for the $Ti_3AlC_2\{110\}$), and the intensity is also small. Table 11 and 12 show the Bragg angle in first order and second order respectively, when using the Ti tube.

Table 11. the Bragg angles and intensities for Ti_3AlC_2 with Ti tube in first order

TiAlC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
2.7497	6.8	13.6	0.203587	11.66469	23.32938	39	0	0	2
2.7497	2.633	5.266	0.549383	31.47732	62.95464	19	1	0	0
2.7497	2.585	5.17	0.560792	32.131	64.262	6	1	0	1
2.7497	2.277	4.554	0.648258	37.14246	74.28492	100	1	0	3
2.7497	2.267	4.534	0.651604	37.33414	74.66828	18	0	0	6
2.7497	2.082	4.164	0.721286	41.32662	82.65324	3	1	0	4
2.7497	1.7177	3.4354	0.927965	53.16848	106.337	14	1	0	6
2.7497	1.52	3.04	1.130221	64.75691	129.5138	14	1	1	0

Table 12. the Bragg angles and intensities for Ti_3AlC_2 with Ti tube in second order

Second order									
TiAlC									
Lambda	d	2d	RAD	Theta	2Theta	Intensity	h	k	l
2.7497	6.8	13.6	0.416287	23.85151	47.70301	39	0	0	2

4. Conclusions

1. The Bragg–Brentano technique, while effective in many cases, has limitations in detecting lattice distortions between irradiated and non-irradiated areas of MAX phases due to the low depth of irradiation affected zones.
2. Due to the small depth of irradiation, the use of Small Angular X-ray Scattering (SAXS) is crucial in detecting lattice distortions caused by irradiation.
3. For residual stress measurements in the TiAl_3C_2 MAX phase, it's important to note that the use of Cr and Mn X-ray tubes is not feasible due to the occurrence of fluorescence excitation. This highlights the challenges we face in our research. The best tube which can be used for measuring is Cu tube.

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Thermodynamic Stability

MAX phases are thermodynamically stable over a wide range of temperatures because of:

Strong M–X bonds (ionic/covalent character)

Favorable enthalpy of formation compared to competing phases

Their layered structure allows strain accommodation during thermal cycling

Applications

- **High-temperature coatings**
- **Electrical contacts and heating elements**
- **Nuclear materials** (radiation damage resistance)
- **Lightweight structural components**
- **Precursors to MXenes** (2D materials obtained by etching out the A-layer)